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NATIONAL METAL CASTING RESEARCH INSTITUTE

Final Report

VOLUME 2: DIE CASTING RESEARCH

**By
D. Jensen**

August 1995

Work Performed Under Contract No. DE-FC07-92ID13164

**For
U.S. Department of Energy
Office of Industrial Technologies
Washington, D.C.**

**By
University of Northern Iowa
Cedar Falls, Iowa**

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For
U.S. Department of Energy
Sponsored by the Office of the Assistant Secretary
for Energy Efficiency and Renewable Energy
Office of Industrial Technologies
Washington, D.C.

By
Metal Casting Center
Department of Industrial Technology
University of Northern Iowa
Cedar Falls, Iowa

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Overview

The contents of this document constitute the final report for Thrust Area Two: Die Casting Research Consortium of the U.N.I. Metal Casting Center. This thrust area consisted of a coordinated approach to investigation of topics of interest the U.S. die casting industry. With the aid and financial support of the North American Die Casting Association and the U.S. Department of Energy, five inter-related research projects were undertaken at three universities, with the University of Northern Iowa bearing the responsibility for program coordination, reporting, and technology transfer. The primary investigators/coordinators and their associated institutions were:

Carroll Mobley, Rajiv Shivpuri	- Ohio State University;
Dan Quick	- University of Northern Iowa;
John Conrad, Fred Bradley	- University of Wisconsin - Madison;
Diran Apelian	- Worcester Polytechnic Institute;
William Walkington, Joseph Martori	- North American Die Casting Association
Robert Trimberger, Manohar Sohal	- U.S. Department of Energy

In addition to the research and project management activities, two well-respected metal casting professionals served consulting roles; professors John F. Wallace of Case Western Reserve University, and Carl Loper, Jr. of the University of Wisconsin - Madison. In total, over fifteen people from the five universities cooperated, along with several more from NADCA to perform the research and disseminate the information to the die casting industry.

Initial funding for these activities was supplied by the U.S. Department of Energy - Office of Industrial Technology, with matching funding provided, in part, by NADCA. These monies were earmarked under the Department of Energy Metal Casting Competitiveness Research Act of 1990, Public Law 101-425; and became available in April, 1992. All components of the project were completed as of July 1, 1994. The specific research sub-projects completed were:

Development and Evaluation of Die Coatings - J. Conrad primary investigator;

Accelerated Die Life Characterization of Die Materials - R. Shivpuri, primary investigator;

Evaluation of Fluid Flow and Solidification Modelling Programs as Applied to the High Pressure Die Casting Industry - F. Bradley, primary investigator;

Selection and Characterization of Aluminum-based Die Casting Alloys - D. Apelian, primary investigator;

Influence of Die Materials and Coatings on Die Casting Quality - C. Mobley, primary investigator.

The activities and individual projects of the consortium are detailed in the following section, with final reports for each of the research projects comprising the remaining sections

Program Tasks

The five projects, mentioned in the previous section, required coordinated interactions between the universities. In the first two projects, the coatings to be applied and tested were selected by the project personnel based on work performed by Wallace, who acted as a consultant. The coatings were applied using new plasma ion-implantation processes and equipment under development by Conrad at UW. Following the application of the coatings, the materials were subjected to wear testing in equipment and tooling provided and staffed under the supervision of Shivpuri at OSU. Due to the unique nature and availability of the equipment and expertise necessary, these projects could not have been carried out without the mutual cooperation of these institutions.

Similarly, inter-project cooperation was also necessary in the completion of the computer solidification project. Three computer solidification packages were to be tested for adaptability to high pressure die casting use; however, an agreement with one of the package suppliers was not possible for Bradley's UW project. This testing was completed, in full, thanks to the inter-project cooperation with Mobley at OSU, where an agreement with the package supplier was in effect.

Finally, the technology transfer and project coordination activities required of the UNI-MCC would have been much less effected were it not for the cooperation of the OSU researchers, Apelian's team at WPI, and the staff at NADCA. Close interactions with these people enabled the consortium to locate activities in Illinois, Ohio, and Massachussets -- resulting in more input from parties in an inter-regional way.

Specific Tasks

In addition to the above mentioned research sub-projects, the scope of work included the following responsibilities pertinent to the UNI-MCC (Department of Energy Grant Proposal, p. 128):

- 1.) Providing fiscal accountability of the income and disbursements for the total program.
- 2.) Consolidation of research progress and final reports from each of the other participating universities to provide the necessary report documents to the Metal Casting Industrial Advisory Board and the DOE.
- 3.) Organize and conduct a two-day symposium/workshop directed to defining the current body of knowledge and state-of-the-art technology relating to die materials for the high pressure die casting of aluminum-based alloys.
- 4.) Organize and conduct a two-day workshop to define the critical research needs of the U.S. die casting industry and to assist in the preparation of research proposals to address those needs.

In order to facilitate the accomplishment of those tasks, the scope of work for each of the five sub-projects included the following (Department of Energy Grant Proposal, p. 127):

- a.) Providing the UNI-MCC with written progress and final reports describing the die casting research activities and projects conducted as part of this program at their sites.
- b.) Attending and actively participating in both of the die materials symposium/workshop and the research definition workshop conducted by the UNI-MCC.

These tasks were accomplished in the manner laid out in the following section.

Chronology of Activities

Prior to the actual receipt of federal funding, it was necessary to iron-out differences with the Metal Casting Industrial Advisory Board. A meeting was arranged with the primary investigators from most of the research projects, both consortium consultants, the UNI-MCC, and Mr. D. Cocks of the MCIAB and Mr. R. Trimberger for the DOE. This meeting was held on February 6, 1992 at the Ohio State University, leading to mutual understandings for the project objectives and clearing the way for funding.

After the official project kickoff in April, 1992, sub-contracts were arranged for each of the five research projects at their respective university, and the research project activities were enabled. In coordination with NADCA, the first workshop was held in Rosemont, IL on July 20-21, 1992. In order to re-synchronize with the federal funding time frames, the subject of workshop focused on necessary research objectives, hence the title Research Needs of the U.S. Die Casting Industry Workshop. Presentations from all of the research projects were obtained, as well as reports from the following NADCA research and development sub-committees:

Cast Materials Sub-Committee, by H. Brucher of Doehler-Jarvis;

Process Technology Sub-Committee, by D. Albright of Hydro Magnesium and G. Pribyl of Heick Die Casting;

Die Materials Sub-Committee, by W. Smith of DCD Technologies and H. Brucher;

Computer Modelling Sub-Committee, by W. Walkington of Walkington Engineering and A. Miller of Ohio State University.

This workshop was hosted by D. Quick of the UNI-MCC, and featured a summary of previous research by J. Wallace of CWRU. Wallace served as a consultant to the Die Coatings and Wear Testing projects, as per his professional services agreement with UNI. The proceedings from this workshop were transcribed, edited, then published and sent to the 40 registered participants in June of 1993.

To meet the reporting requirements of the DOE, monthly progress reports were obtained via telephone from each of the research sub-projects, consolidated with those at UNI, and (fax) forwarded to M. Sohal for the DOE. Quarterly progress reports for these projects, preferably written, were likewise obtained, consolidated, and (mail) forwarded to Sohal for the DOE.

The second workshop, entitled Defining the State of Die Cast Technology Workshop, was conducted on March 25-26, 1993 at the Worcester Polytechnic Institute in Worcester, MA. Timing was selected to allow for reports of preliminary results from each of the research projects. Presenters at this workshop featured evaluations of the current state of die casting from various perspectives, as well as foreseen areas for future development of the industry. The meeting was hosted by D. Apelian and moderated by D. Quick. In addition, an open panel discussion was held on the first evening, moderated by C. Loper of UW, which was very well received. The activities of this workshop were transcribed, then forwarded to Loper, who was retained to serve as a consultant in expediting the editing and publication of the second workshop proceedings. Problems with this arrangement arose, and it became necessary to re-acquire the services of D. Jensen to complete those proceedings and forward the copies to the 45 registered participants at the start of July, 1994.

The individual research projects were completed by the early summer of 1993; and final reports on these projects obtained in the time period from late May to August of that year. Results of these some of these projects were presented at the NADCA 17th International Die Casting Congress and Exposition, held in Cleveland, OH in October of 1993. Additionally, three summary presentations were given by the UNI-MCC for the local NADCA chapter, and the NADCA organization has committed resources for future technology transfer activities in the dissemination of these research results. The detailed findings of the research projects are presented in the following sections, and the financial summary for this thrust area is affixed in the appendix.

Conclusion

Upon the acceptance of this final report by the Department of Energy (also forwarded to NADCA), the specific activities of this thrust area are concluded. It should be noted that enough of the consortium aspects and activities that took place under this funding were deemed worthwhile to NADCA and the die casting industry, that it appears a more permanent consortium may be established at the Ohio State University.

Establishing a consortium with more permanence should be a primary goal for the future. The inevitability of funding delays, particularly at the federal level, played havoc with the ability of the consortium institutions to acquire or retain the technological personnel and equipment time necessary to carry out the physical research activities. It appears unlikely that the federal processes can be adjusted to facilitate coordination with these kinds of research activities. Therefore, in this person's view, establishing a more permanent consortium base, if imbued with the ability to smooth the transient funding discontinuities, should result in a much more efficient research effort for the die casting industry.

Sub-Project 1

Development and Evaluation of Die Coatings

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R.P. Fetherston

A. Chen

INTRODUCTION

The Department of Energy Grant (No. DE-FG07-92ID13164) provided through a subcontract by the Metal Casting Center at the University of Northern Iowa (grant No. 22243-08-C) has been used to apply the Plasma Source Ion Implantation (PSII) technology, a surface modification technique developed at the University of Wisconsin-Madison, to improve the service life of dies used in the die casting industry.

Die castings are made by rapidly forcing molten metal into a die cavity cut into a steel block. The most commonly used steel for dies is the AISI H-13 steel. High pressure moves the metal through troughs into the die cavity in a fraction of a second. Total cycle time is measured in seconds. The casting may or may not need additional processing since it is accurately cast into intricate shapes. The finished shape is the prime advantage of the die casting process. However, in order to accomplish this, expensive tooling is required. Depending upon the application, cost of the die can be approximately 10,000 times the selling price of the casting. Dies deteriorate in use due to a variety of reasons, including, heat checking, soldering and erosion, so that castings begin to exhibit decreased surface quality and general appearance. Eventually dies need to be replaced at high cost to the user. The die casting industry and consumers could benefit significantly if the lifetime of the dies could be prolonged.

The PSII technology has been used to *deposit thin coatings of refractory metals, onto H-13 die steel test pins* which were subsequently evaluated at the Ohio State University at Columbus. In addition, the Department of Energy/University of Northern Iowa grant has *spawned collaborations involving PSII research with a variety of industries and professional agencies involved in die casting in the United States.*

PLASMA SOURCE ION IMPLANTATION CONCEPT AND PROCESS

Conventional ion implantation has been shown to be effective in improving the wear, corrosion, fatigue and friction properties of materials. For example, ion implantation has been shown to improve substantially the life

of cutting tools such as punches, taps, dies and cutting inserts. However, conventional ion implantation has not received widespread industrial acceptance, due to its relatively high cost. The high cost is due largely to the fact that most ion beam materials processing facilities employ ion implanters which were optimized for semiconductor wafers fabrication, not for materials applications. In particular, conventional ion implantation is a line of sight process, and thus extensive target manipulation is required for the implantation of three dimensional objects.

Plasma Source Ion Implantation (PSII) is a non-line-of-sight technique for surface modification of materials which is optimized for ion implantation of non-planar targets [1-15]. In Fig. 1 we compare PSII with conventional ion implantation. Conventional ion implantation is a line-of-sight process in which ions are extracted from an ion source, accelerated as a directed beam to high energy, and then raster scanned across the target.

If the target is non-planar, target manipulation is required to implant all sides of the target. The necessary target manipulation adds complexity and reduces the size of the target which can be implanted. This target manipulation problem in conventional ion implantation is exacerbated by the need to provide adequate heat sinks at the target to limit temperature rise during the implantation. Depending on the target size and geometry, target manipulation (translation and rotation) may be required to implant all sides of the target. The PSII approach to implantation of complex-shaped targets is much simpler. In PSII, the target is placed directly in the plasma source and pulsed to high negative potential relative to the chamber walls. Ions are accelerated normal to the target surface, across the plasma sheath thus eliminating not only the line-of-sight problems of conventional ion implantation, but also the retained dose problem.

Even with sophisticated target manipulation systems, the performance of beamline implanters is still limited by the retained dose problem, i.e., the maximum dose retained by the target is governed by the angle of incidence of the beam and additionally at high fluences, by sputtering. The angular dependence of the retained dose imposes restrictions on the degree of dose uniformity achievable with line-of-sight, beamline ion implantation of three dimensional targets. In order to achieve reasonable dose uniformity on targets with curved surfaces, target masking may be employed to restrict the ion beam angle of incidence to values less than 20° to 30° . However, for

targets with complicated geometry, it may be either difficult, impractical, or virtually impossible to devise masks which satisfy this requirement.

Conventional Beamline Implantation

- Line of sight process
- Beam rastering and target manipulation are required to achieve uniform implantation

Plasma Source Ion Implantation

- Plasma sheath surrounds target
- Ions bombard all surfaces of target without beam rastering or target manipulation

Fig. 1. Comparison of conventional beamline ion implantation and Plasma Source Ion Implantation (PSII).

Even if the target has sufficient symmetry to accommodate masking, the masking degrades the system performance, since the fraction of the beam which scrapes off on the mask constitutes a loss in production throughput. Furthermore, unless the mask material is made of the same material as the target, sputtering of the mask may contaminate the target. Elimination of the ion accelerator stage, raster scan apparatus and target manipulator hardware makes PSII a much simpler and more cost-effective technology than conventional beamline ion implantation.

Compared to accelerator technology, PSII provides the opportunity of a superior implant (due to the amelioration of the retained dose problem) at a lower cost (due to the decreased complexity and potentially greater processing throughput of the PSII process). Our experiments on PSII (see references) have demonstrated (1) efficient implantation of ions to concentrations and depths required for surface modification, (2) dramatic improvement in the life of manufacturing tools in industrial applications, (3) acceptable dose uniformity on non-planar targets without target manipulation, and (4) that such uniformity can be achieved in a batch processing mode.

INVOLVEMENT OF THE PSII RESEARCH PROGRAM WITH THE DIE CASTING INDUSTRY

- In early 1988 the PSII research program was involved in a research program with Stroh Die Casting, Milwaukee. This research was aimed at determining the effectiveness of PSII nitrogen implantation in improving the life of A-2 tool steel shears used for trimming the flashes off die castings.
- In June 1991, the PSII research program in collaboration with Harley Davidson Inc was awarded a research grant from the State of Wisconsin Industrial and Economic Development Research Fund for research entitled "Improved Lifetime of Die Casting Molds by PSII". This collaboration also involved EST Die casting and Ram Tools a die machining company, both in the Milwaukee area. The primary goal of this research was to reduce (by PSII) the propensity of heat checking of the dies which manifested in the form of a poor casting surface finish. We have deposited palladium and diamond-like carbon (DLC) coatings on test pins which were inserted in the gating area of the casting. The casting surface facing these pins was periodically monitored for surface finish. The tests were inconclusive but it was noted in many instances that the casting surface facing the palladium coated pins did show a decrease in total crack length and the casting surface finish was better for the DLC coated pins. A drawback of these tests was the long times that were involved. Based on our recent discussions with personnel at EST and Harley Davidson, we will be treating a full dies. These dies have been selected, based upon their size, cost and their rapid wear rate. Based on

the results detailed in the later part of this report, tungsten coatings will be deposited on these dies using the PSII-IBED process.

- In early 1992 we submitted a proposal to the International Lead Zinc Research Organization (ILZRO) for research on producing zero-draft core pins by depositing DLC coatings. This work was originally intended to be done in collaboration with Madison Kipp Cast, a company involved in zinc and magnesium die castings. This interaction could not be pursued due to non-availability of funds.
- Our program has received a Department of Energy grant through a subcontract from the soon to be established Center of Die Casting at the Ohio State University. The research will involve PSII coatings on actual dies to be tested at aluminum, zinc and magnesium die casting companies in the Ohio and Wisconsin areas

PERSONNEL VISITS

- Visit by K. Sridharan and M. Shamim of the PSII research program to the U.S. Die Casting Industry Research Needs Workshop, Chicago, IL, July, 1992. A presentation entitled "Potential for the use of Plasma Surface Modification in the Die Casting Industry" was given by K. Sridharan
- Visit by K. Sridharan, M. Shamim and A. Chen of the PSII research program to the NADCA workshop, Milwaukee, WI
- Visit by Prof. J.R. Conrad of the PSII research program to Die Casting Research Consortium, Columbus, OH, February, 1992
- Atleast two visits by Mr. William Walkington and Prof. Carol Mobley to the PSII facilities in Madison, WI
- Two-way visits between PSII program and Madison Kipp Cast personnel

PSII-IBED COATING TREATMENT AND TESTING OF PINS

Coating Selection

The selection of coatings was based on the work in the early 1980s by Mirtich and Wallace [16] in which it was demonstrated that coatings of certain refractory metals and their compounds could reduce the propensity for cracking by thermal fatigue. We initiated work on developing a data base of the relevant properties for the die casting application for a wide range of

prospective metal coatings. Our opinion is that the solubility of the coating material in molten aluminum and its melting point are of paramount importance. *Platinum* (Pt), *molybdenum* (Mo) and *tungsten* (W) were selected, and the relevant properties of these metals are shown in Table I.

Table I: Relevant properties of Mo, Pt and W for application as die coating materials.

IBED Metal	Mo	Pt	W
Yield Strength (x1000 psi)	90	—	—
Tensile Strength (@20°Cx1000psi)	75	18	264-590
Young's Modulus (x10 ⁶ psi)	48	21.3	50
Ductility (%)	—	35-40	1-3
Coeff. of linear Exp. (x10 ⁻⁶ psi)	5	9	4.5
Thermal Conductivity (W/cm°C)	1.4	0.73	1.78
Melting Point (°C)	2610	1769	3410
Solubility in Al (Vol%) @ 725°C	0	2	—
@ 925°C	2	7	—
ΔG ^o _f Oxidation (kcal/mol)	MoO ₄ -199.9	PtO ₂ 40.1	WO ₄ -182.6

Radio Frequency (RF) Plasma Source for Producing Coatings

The plasma in the PSII chamber is generated using tungsten filaments. However, the tungsten filaments used for the production of the primary electrons have short lives and are somewhat unsuitable for long runs that are required for producing thick IBED (Ion Beam Enhanced Deposition) films. We have developed a RF source can be coupled inductively or capacitively through a matching network to the vacuum chamber to produce the plasma. A matching network for RF power supply for parallel plate sputtering and plasma generation is being fabricated. We are also designing an inductively-coupled system to produce a cleaner plasma. An illustration of the design of

the RF source for PSII-IBED processing and the relevant PSII parameters used for depositing these coatings are shown in Fig. 2 and Table II, respectively.

Fig. 2. A schematic illustration of the RF source designed for producing PSII-IBED coatings.

Table II: Relevant PSII -IBED parameters used for depositing coatings.

Argon Sputter Cleaning Energy/Dose	5 kV, 5×10^{16} atoms/cm ²
Implantation Voltage	15 kV
Operating Pressure	1.9×10^{-3} Torr
Repetition Rate	55 Hz
Discharge Current	1.5 A
Discharge Voltage	125 V
Cathode Voltage	1 kV
IBED Time	120 minutes
Coating Thickness	2 μ m
Target/Cathode Separation	4 inches

Testing Procedure at Ohio State University (from Ref. 17)

Monitoring the wear performance of a coating or a die material for die can be a long and expensive procedure. An accelerated test has been devised at the OSU where a pyramidal test pin design is employed (see Fig. 3). Here the sharp corners of the pin act as stress concentrators and are expected to readily erode, solder and exhibit heat checking. The pins were tested under the following conditions:

Average gate velocity: 50 m/s

Average furnace temperature: 1320 deg. F

Lubricant spray: 2 seconds with Chem-Trend, Safety Lube 4000

Number of shots: 1000

The wear of the test pins were measured by a precision balance with a weight accuracy of 0.0001 g. The results of the study are shown in Fig. 4.

Fig. 3 . The pyramidal pin design used for accelerating the deterioration of coatings.

Fig. 4. Performance of various coatings with respect to the untreated H-13 die.

The study shows that W coatings have the potential to improve the wear life of the dies. It is our opinion that the Pt and Mo coatings may have delaminated in certain areas of the pins thereby contributing to the weight loss measurements. Further optimization of PSII process parameters for improved adhesion may be required.

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Sub-Project 2

Accelerated Die Life Characterization of Die Materials

R. Shivpuri

The Ohio State University

1.0 INTRODUCTION

1.1 Introduction

The die casting process uses high injection pressure to push high temperature molten metal into a die cavity representing the part geometry. Figure 1 shows a typical horizontal cold chamber die casting machine, including the die plates, the toggle mechanism for locking the plates, the shot sleeve, shot accumulators and intensifier, etc.[1].

A typical die casting process can be divided into the following stages: molten metal pouring, injection and cavity filling, solidification, and part ejection. These four stages of the die casting process are shown in Figure 2. [1]

During the injection process, the molten metal is pushed into the die cavity with high gate velocity. Usually the gate velocity is in the 100 fps to 200 fps range, and the molten metal temperature is in the 1100 °F to 1300 °F range. This high velocity and high temperature molten metal injection can cause erosive wear of the die surface exposed to the molten stream. Most die casting dies have complicated geometrical features such as cores, pins, ribs, and corners which are especially prone to erosive wear.

Major defects associated with die wear are erosion, heat checking, soldering and cracking. In this study sponsored by Department of Energy, the erosive wear resistance of metallic coatings was evaluated as follow:

(1) W, Mo, and Pt coatings were applied on H-13 pins by ion implantation techniques at the University of Wisconsin, Madison,

(2) Multi-pin erosive wear tests were conducted at the Ohio State University, and

(3) Results of tests were analyzed at the Ohio State University.

1.2 Erosion in Die Casting

The driving force for erosive wear or washout of die cavity surfaces during the filling stage may be derived from the following sources: solid particle impingement, liquid droplet impingement, and cavitation-erosion. In addition, during the part solidification stage, the molten aluminum alloy chemically attacks the die steel surface possibly leading to corrosive wear, dissolution, and sometimes soldering.

The mechanism of erosive wear is shown schematically in Figure 3. The impingement damage may result directly from impact pressure and velocity, or from shearing effects at the surface caused by high velocity radial flow of the spreading droplet, or both. In die casting case, as the impingement particles hit the die surface, the die surface response may be either brittle fracture or plastic deformation [2].

For evaluating the wear behavior of die materials and different surface treatments and coatings for die casting application, it was decided to use the wear of pins as a surrogate measure of die wear. The choice of pins as test coupons presented several advantages including ease of coating, assembly and disassembly, and

destructive evaluation of coatings and substrate properties and quality. It also provided an inexpensive technique for quantifying surface wear phenomena of the die.

2.0 TEST DIE AND TEST PROCEDURE

2.1 Test Die at The Ohio State University

Industrial experience with die wear indicates that core pins or die inserts exposed to the liquid metal attack in front of the gates exhibit the highest level of erosive wear (wash out) and soldering. Consequently, the wear of core pin surfaces was chosen to represent the accelerated wear of die surfaces exposed to high flow velocities. To further accelerate the wear process, a pyramidal test pin design with sharp corners was selected for our tests. It was expected that these sharp pin corners will readily erode, solder and exhibit heat checking along the edges.

In order to evaluate different materials or surface coatings, "a multi-pin flat plate die" with six test pins was designed and fabricated for these experiments (Figure 4). One of the reasons for choosing the multi-pin design was the ease of assembly, disassembly and measurement. In addition, the multi-pin design allows the testing of several pins simultaneously, thus providing multiple test sites for comparative evaluations. The flat cavity (plate) design allows for the study of the effect of filling and solidification on die wear because of its geometrical simplicity. In addition, the test pins can be rotated or their shape changed,

permitting the study of the angle of attack and the surface geometry on wear loss. Each test pin batch contains one control pin to check repeatability of each campaign.

2.2 Accelerated Test Procedure

Wear is a gradual loss of material. Wear test could be very long term (> 100,000 shots) and costly. In order to accelerate the wear rate, an operating procedure was developed to provide an extreme environment on the test pin so that quantifiable wear loss can occur in a reasonable number of shots. The operating conditions chosen to accelerate the erosive wear rate are listed below:

(1) Casting alloy A390 with a high silicon content

The A390 alloy is identified as hypereutectic aluminum-silicon casting alloy. The advantages of A390 are superior wear resistance, high fluidity and high thermal conductivity. The nominal chemical composition of A390 is 16-17% Si, 4-5% Cu and 0.6-1.1% Fe. The melting point of A390 is around 1220°F. Relatively speaking, the primary silicon particles in partially solidified melt are bigger and have irregular shapes and sharp corners. Furthermore the silicon being very hard, A390 is more aggressive than other aluminum die casting alloys which results in higher abrasive wear rate for the die surface.

(2) High gate velocity

The average magnitude of gate velocity in the die casting application is around 130 fps. In the early tests on our single pin test die, we had used the gate velocity of over than 200 fps.

However, with that high a gate velocity, the gate area seizes prematurely because of enhanced soldering problem. Due to gate seize-up the machine had to be shut down and the dies cleaned, this interrupted the test operation. The details of this problem can be found in the report ERC/NSM-C-91-34 [3]. To avoid this problem, in this multiple pin die test, the gate velocity was chosen to be lower around 160-180 fps.

3.0 PILOT TEST VERIFICATION

(Sponsored by Engineering Research Center for Net shape Manufacturing)

3.1 Experimental Design Strategy: Pilot Stage

Before using the multi-pin test die and test pins to evaluate the coatings for erosive wear, it was important to verify the validity of die design and test procedure for conducting the erosive wear test. The experimental design program selected for this was conducted as the "pilot stage". The objectives of this pilot stage tests were:

- (1) To test the validity of the die design
- (2) To check the assumptions on the test pin locations
- (3) To estimate the minimum shots for quantifiable wear

3.2 Machine Operation Conditions

The operation condition of pilot run was shown as follows:

- (a) Furnace temperature: 1300-1350°F
- (b) Gate velocity: 180 fps
- (c) Accumulator pressure: 1000 psi
- (d) Spray time: 3 sec.
- (e) Pin material: H-13 pin, hardness HRc 46

3.3 Performance Measurements

The wear of test pin was measured by a precision balance with the weight accuracy of 0.0001g.

3.4 Results of the Pilot Test

The weight loss of the pins was the primary measure for erosion in this experiment. The pilot test results are plotted in Figure 5. Based on these results, the following conclusions were made:

(1) The front-row pins show an approximately equal weight loss rate and similar surface wear profiles. The weight loss in pin location 2 is slightly larger than the other front-row pins. This variance may come from a few sources such as the variabilities in weight loss measurement or the machining of the die (maybe it is slightly closer to the gate). Pin location 2 was used as "control location" in the following tests.

(2) The rear-row pins do not have as large a weight loss and surface wear as the front-row pins. This is because the velocity impingement on the surface of the pin in the rear row is far less than the front row.

(3) The number of shots needed for the coated pin experiments is expected to be larger than 600. From statistical analysis it was determined that after 600 shots the weight changes were significant.

4.0 ION IMPLANTATION TEST PINS STUDY

(Sponsored by the Department of Energy)

4.1 Introduction

Ion implantation and related surface treatments using energetic ion beams are being increasingly employed for wear resistant applications. In this project, three different ion implantation surface treatments (W, Mo, Pt) were applied on the H-13 pin surface at the University of Wisconsin, Madison (UW-M). The detailed specification and process parameters for preparing these coatings will be provide by UW-M.

The selection of the W, Mo, and Pt for coatings was based on the results of an earlier heat checking study by Prof. Wallace at Case Western Reserve University. During his heat checking dunk test, Prof. Wallace found the three sputter coatings of W, Mo and Pt to have the best resistance to heat checking.

4.2 Test Result and Test Condition

The results of ion implantation coating tests are shown in Figure 6. The test conditions were:

- gate velocity average:50 m/s
- furnace temperature average:1320°F
- lubricant spray: 2 secs with Chem-Trend, Safety-Lube
4000

The result shows that the W coating has the best resistance to erosive wear. Other coatings like Pt and Mo did not show major improvement over H-13 wear behavior.

4.3 Analysis and Discussion

Mechanical properties such as tensile strength, yield strength, ductility, toughness and hardness are frequently of most importance when selecting metals for dies. However, for a given application a compromise between these and other properties is usually required. Generally speaking, good erosive wear resistant die coatings should have the following properties:

- good strength
- good toughness
- good hardness
- high thermal conductivity
- compatible thermal expansion with substrate

Strength and toughness analysis

The yield strength vs Young's modulus information can be found in Figure 7 [4] for various metallic elements. The higher the yield strength of the material, the stronger the material will be. From

this figure we can see that the yield strength of W is almost twice that of Mo, and ten times that of Pt. This plot was taken for pure metal in a fully work-hardened condition. The elastic modulus is an important property of a material, because together with the yield stress, it governs the total amount of volume elastic energy which the material can store. Additionally, the toughness of the material is used to describe both the strength and ductile properties. From the above, it can be concluded that the W is better than both Mo and Pt.

Modulus of elasticity of selected metals in tension ($\times 10^6$ psi) [5]

Metal	W	Mo	Pt
Modulus of Elasticity	50	47	21.3

Hardness analysis

The erosive wear resistance can be correlated with the hardness of the surface. From this erosive wear (Pt, Mo, W) test, we found the W coating to be the best. Figure 8 shows a plot of the hardness vs resistance to abrasive wear for pure annealed metals and steel. Thus under identical conditions, W has better abrasive wear resistance than Mo due to its higher hardness [6]. Recently researchers have linked the surface energy consideration with the friction and wear problem. Since both the surface energy and the mechanical strength of a material depend on the strength of its bonds, that means the high surface energies are associated with high strength parameters. The Figure 9 shows the γ/p vs p plot for

several materials [4]. The γ is surface energy and p the hardness. It can be seen that soft materials have higher γ/p ratios than do harder materials. Also, that non metals have lower ratios than do metals of comparable hardness. Furthermore, the low γ/p ratios are associated with better surface interaction behavior, low friction, smaller wear particles, smooth surfaces, and less adhesion. From this figure it can be observed that W is better than both Mo and Pt for abrasive resistance. Therefore, from the friction or sliding applications sometimes non-metals are better than metals, and hard materials are better than soft materials [4]. From different tests conducted at OSU, it was found that the VC is one of the best coatings to resist erosive wear in die casting operation due to its extreme hardness [7]. This result is in agreement with Figure 9.

Oxide Layer Hardness Analysis

In die casting environment, pure Mo and W coatings will very easily oxidize and form an oxide layer; Pt is more stable. In die casting operation environment, the molten aluminum is also easily oxidized. The hardness of oxides of W and Al are of the same order, but the oxide of Mo is more than 10 times less hard. When the hard aluminum oxide particles impinge on the Mo coating surface, higher wash out effect can be expected [8]. The hardness of various oxides are shown in Figure 10 with provides ratio of metal hardness to oxide hardness.

Others

Some other properties of the three coatings are listed as below. These properties effect the heat checking resistance rather than erosive wear resistance of the coatings. Since these three coatings were selected based on Prof. Wallace's heat checking test, detailed discussions can be found in Ref. [9].

Thermal Conductivity of selected metals (cal/cm²/cm/s/°C)

Metal	W	Mo	Pt
Thermal Conductivity	0.397	0.34	0.165 (17°C)

Coefficient of linear thermal expansion of selected metals (in/in.
°C)

Metal	W	Mo	Pt
Coefficient of linear thermal expansion	4.6	4.9	8.9

4.4. Conclusions

The primary conclusions from this erosive test are:

1. W coating showed improvement in the erosive wear resistance of H-13 steel substrate while Pt and Mo coatings did not provide any significant protection.

2. A possible reason for W to provide increased erosive wear resistance is its good hardness and high yield strength. The hardness of W increases even further in oxidizing environment.

ACKNOWLEDGMENTS

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Figure 1 General Layout for a Cold Chamber Die Casting Machine [1]

Figure 2 The Four Sequence of Operation of Cold Chamber Die Casting Machine [1] : pouring (a), injection (b), die opening (c) and ejection (d).

Impinging liquid drop produces compression at impact site on solid which may cause yielding and formation of depression *a*; outward radial flow creates shear forces which may form erosion waves in ductile solid *b*, or fractures in brittle solid *c*

Figure 3 A Schematic Shows Liquid Drop Impingement [2]

Figure 4 A Schematic of Test Die and Test Pin

Figure 5 Weight Loss of Pilot Run Test Pins [7]

Figure 6 Weight Loss of W, Mo, Pt and H-13 Pins

Figure 7 Young's Modulus vs Yield Strength Plot for Metallic Elements [4]

Figure 8 The Effect of Hardness on the Abrasive Wear Resistance for Pure Annealed Metals and Steel [6]

Figure 9 Plot of the γ/p ratio vs Hardness for Materials [4]

Figure 10 Correlation between Metal Hardness and Its Oxide Hardness [9]

Sub-Project 3

Evaluation of Fluid Flow and Solidification Modelling Programs

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CASTHERM

INTRODUCTION

The CSIRO Division of Manufacturing Technology, Australia, has developed this software package specifically to assist the die casting die designer to calculate the temperature distribution in the die. The temperature calculation is carried out using a mathematical technique known as the Boundary Element Method.

CASTHERM is the result of research being conducted since 1972. The temperature distribution information helps the user to interactively position heating/cooling systems in the die.

CASTHERM HIGHLIGHTS

- PC-based computer aided software package.
- Two-dimensional heat flow analysis using boundary element method to speed up calculations and simplify die geometry input.
- IGES transfer format for geometric information to and from other CAD/CAM systems.
- To assist the design of the cooling and heating system of die casting dies.

CASTHERM DESIGN VARIABLES

Using CASTHERM, the die designer can study the effect of design variables on the thermal performance of a die casting die. Some of the variables are:

- The production rate.

- Casting size and shape.
- The construction of the die.
- The material of the die.
- The cooling channel position.
- The coolant flow rate and material.
- The application of die spray.
- The size and position of electric heaters.
- The use of thermal insulation between the die and machine frame.

HARDWARE REQUIREMENTS

The following hardware and software are required:

- An IBM AT or compatible personal computer of type 80286, 80386 or 80486 and math coprocessor.
- Minimum 1 megabyte of RAM for 80286-based machine (2 Mbyte for 80386).
- 40 Megabyte hard disk.
- 1 high density floppy disk drive (1.2 Mbyte 5.25 inch or 1.44 Mbyte 3.5 inch).
- DOS 3.3 or later version.
- High resolution color graphics capability, EGA, VGA or better.
- A Microsoft or compatible mouse.
- An MS-Windows supported plotter.
- An MS-Windows supported printer.

CASTHERM PROGRAM OVERVIEW

CASTHERM provides an integrated environment for the design of cooling and/or heating systems in pressure die casting dies. All the program modules access a single data file.

Current implementation of CASTHERM consists of

File Module: provides facilities to store, retrieve and print out files of calculated results and design geometry.

Project Information Module: provides the facilities for defining the die design project information, properties of casting alloys and die materials.

Thermal Zones Module: provides the means of specifying quantitatively the heat load distribution into the die.

Die Section Module: provides the facilities for creating and processing the geometry of die sections and associated cooling and/or heating lines. It also provides the facilities to manipulate the display and to produce hard copy plots of the die temperature.

Analysis Module: provides comprehensive facilities for defining the thermal conditions at the boundary of the die section and cooling channels and for calculating the die temperature from which the design of the thermal features of the die can be optimized.

IGES Module: provides the facilities for transferring data to and from other CAD systems.

DIE DESCRIPTIONS

The die casting simulations performed using CASTHERM were conducted with two die casting die drawings supplied by Ohio State University. These dies are briefly described below and CAD drawings have been included.

2mm Flat Plate Die

The part produced by this die is a simple 2mm thick plate. Four thermocouples were located in the die to record the time-temperature variation with die cycles. The die has water channels in both ejector and the moving half of the die. The cooling effect of the water channels is asymmetric.

Doehler-Jarvis Bowl Die

This die was from Doehler-Jarvis. The bowl casting is 165mm outside diameter with a floor thickness of 5mm. Thermocouples were located in the die at and near the inside surface of the bowl. The bowl wall is flat and 7mm thick at the contact location with the thermocouple probes.

Learning CASTHERM

CASTHERM provides training sessions for new customers. I was given a four hour hands on training with this software. Prior to this training, my knowledge of Die Casting was limited to book knowledge and site visits. In spite of this I had a good feel for the software after the demo.

The program manual is well detailed. It has clear instructions to load the software into the computer. The program comes with a demo design file. The manual gives step by step instructions to view this demo design file. By going through this procedure the user can get a comfortable feeling the software and is ready to start using the software. Going through the entire manual and experimenting with my own die designs took me about a day.

Running Project Information Module

This module helps to specify the information related to the project, the casting alloy(s) and the die material(s) used in a particular design.

The following information needs to be entered for Project Info:

1. Die Name
2. Section Name
3. Units used (SI units or English units)

The following information needs to be entered for a Casting alloy:

1. Alloy name
2. Density
3. Specific Heat
4. Latent Heat
5. Liquidus temperature
6. Solidus temperature

The following information needs to be entered for a Die Material:

1. Material name
2. Conductivity
3. Specific Heat
4. Density

The data specified for casting and die materials is stored in separate data files and can be recalled into future die designs, this saves time and allows for consistent data usage.

Please refer to tables 1 and 2 for the casting and material data used for this report.

Running Thermal Zones Module

The casting, runner system and overflows are divided into one or more thermal zones. A typical casting has varying section thicknesses. Due to this section thickness variation the heat input into the die is not distributed uniformly. Thermal zones are formed by defining regions whose average wall thickness can be considered to be a constant. For example, if a casting has uniform thickness, then the thermal zones could be casting, runner and biscuit.

After dividing the casting and runner system into thermal zones, for each of the thermal zone the following information needs to be provided:

1. Production rate
2. Volume of molten metal in this thermal zone which is in contact with the die
3. Surface of molten metal in this thermal zone which is in contact with the die
4. Pouring temperature of the casting alloy
5. Ejection temperature of the casting
6. The expected average die temperature at the zone

Based on the above information, calculations are made to compute Heat Load and Heat flux which is further used in the analysis module discussed later.

Please refer to tables 3 and 4 for information entered for this report.

Running Die Section Module

As per the CASTHERM manual, in standard die drawing practice, there should always be a

***** DMT-CASTHERM ALLOY INFORMATION *****

Die Name: plate
Section Name: Ejector half
Units : SI Units

ALLOY Name:	AL
Density:	2500.00 kg/cub.m
Specific Heat:	1088.40 J/kg. deg C
Latent Heat:	393000.00 J/kg
Liquidus:	590.00 deg C
Solidus:	580.00 deg C

Table 1. Alloy Information.

***** DMT-CASTHERM MATERIAL INFORMATION *****

Die Name: plate
 Section Name: Ejector half
 Units : SI Units

Material Name: H13
 Conductivity: 28.00 W/m deg C
 Specific Heat: 460.00 J/kg. deg C
 Density: 7850.00 kg/cub.m

Material Name: Steel
 Conductivity: 35.00 W/m deg C
 Specific Heat: 460.00 J/kg. deg C
 Density: 7800.00 kg/cub.m

Table 2. Material Information.

***** DMT-CASTHERM THERMAL ZONE INFORMATION *****

Die Name: plate
Section Name: Ejector half
Units : SI Units

Casting Name: plate
Production Rate: 200.00 shots/hour

Zone Name: runner
Volume: 20.13 cub.cm
Surface Area: 42.97 sq.cm
Pour Temp: 675.00 deg C
Ejection Temp: 325.00 deg C
Expected Die Temp: 250.00 deg C
Heat Load: 2164.02 Watts
Heat Flux: 503640.47 W/sq.m

Zone Name: casting
Volume: 24.65 cub.cm
Surface Area: 121.29 sq.cm
Pour Temp: 675.00 deg C
Ejection Temp: 325.00 deg C
Expected Die Temp: 250.00 deg C
Heat Load: 2649.24 Watts
Heat Flux: 218421.93 W/sq.m

Zone Name: flowoff
Volume: 1.77 cub.cm
Surface Area: 6.24 sq.cm
Pour Temp: 675.00 deg C
Ejection Temp: 325.00 deg C
Expected Die Temp: 250.00 deg C
Heat Load: 190.24 Watts
Heat Flux: 304991.99 W/sq.m

Table 3. Thermal Zone Information - Plate.

***** DMT-CATHERM THERMAL ZONE INFORMATION *****

Die Name: horse shoe
 Section Name: Ejector half
 Units: SI Units

Casting Name: horse shoe
 Production Rate: 200.00 shots/hour.

Zone Name: biscuit
 Volume: 157.00 cub.cm
 Surface Area: 314.00 sq.cm
 Pour Temp: 650.00 deg C
 Ejection Temp: 325.00 deg C
 Expected Die Temp: 250.00 deg C
 Heat Load: 9769.72 Watts
 Heat Flux: 311137.50 W/Sq.m

Zone Name: runner
 Volume: 29.25 cub.cm
 Surface Area: 48.75 sq.cm
 Pour Temp: 650.00 deg C
 Ejection Temp: 325.00 deg C
 Expected Die Temp: 250.00 deg C
 Heat Load: 1820.15 Watts
 Heat Flux: 373365.00 W/Sq.m

Zone Name: section1
 Volume: 6.95 cub.cm
 Surface Area: 17.38 sq.cm
 Pour Temp: 650.00 deg C
 Ejection Temp: 325.00 deg C
 Expected Die Temp: 250.00 deg C
 Heat Load: 432.48 Watts
 Heat Flux: 248910.00 W/Sq.m

Zone Name: section2
 Volume: 6.95 cub.cm
 Surface Area: 17.38 sq.cm
 Pour Temp: 650.00 deg C
 Ejection Temp: 325.00 deg C
 Expected Die Temp: 250.00 deg C
 Heat Load: 432.48 Watts
 Heat Flux: 248910.00 W/Sq.m

Table 4. Thermal Zone Information - Horseshoe.

die section drawn through the shot hole and the casting. This normally can be used as the first section for thermal analysis. Other sectional planes can be selected after having reviewed the result of the analysis of the first section. Usually three to five sectional planes are sufficient to design a cooling system for a die. For this report, I have chosen a vertical plane section through the center of the casting and the shot biscuit, please refer to figures .1 and 2

Building Models

First step in building models is a rough sketch of the die section showing the die insert(s) and the holder block. The various dimensions are converted into corner coordinates. The model can be built using other CAD systems and an IGES file can be read into CASTHERM. The Die Section Module menu provides drawing primitives like line, arc and circle. Using these drawing primitives the model is drawn.

Each closed loop of lines and arcs is defined as a "component" in CASTHERM. The component menu in Die Design Module helps in selecting loop which define a component. Each component is specified with a different material. All the components form the die section that is being modeled. The heating and cooling lines are also defined as components. In the Analysis module they are defined as heating or cooling lines.

Time to build models

Step	Plate Casting	Horse shoe casting
1. Entering thermal zone information.	20 minutes	20 minutes
2. Reading the die drawing and making a rough sketch with	30 minutes	30 minutes

Figure 1A. Plate Die.

Figure 1B. Plate Die.

Figure 2. Horseshoe Die.

coordinates.

3. Time required to input the model 25 minutes 35 minutes
and define the components.

I feel that the time required to do the above tasks will be considerably lower after gaining some experience. This should be fairly fast as the software is very simple to use and menus and options are intuitive.

Running Analysis Module

This module helps the user to do the following:

1. Assign the thermal conditions on the boundary edges of each die section.
2. Add, delete and edit cooling/heating channels.
3. Calculate the temperature distribution along the die boundary.

If the calculated temperature distribution is not acceptable, the user can:

1. Move, add, delete or edit the cooling/heating channel and/or
2. Change the thermal conditions on the boundary edges.

Specifying boundary conditions

There are six boundary types that can be specified in CASTHERM and they are classified into two major categories, namely:

1. **Heat flux type:** Constant heat flux applied at the boundary surfaces of the die. Two types of boundary conditions in this category.
 - Die/Cast : at the die to casting interface and
 - Die/Heat : at the die and electric cartridge heater interface.

2. Convective type: The thermal transfer into the die is by thermal convection. Four types of boundary conditions in this category.

- Die/Air : at the die surface which are exposed to the ambient air.
- Die/Mc : at the die surfaces which are in contact directly with the machine platen.
- Die/Cool : at the cooling/heating channel surfaces which are in contact with the water or oil coolant. If high temperature oil is used, the same channel acts as die heater.
- Die/Die : at the die surfaces which are in contact with the surfaces of the adjoining die components.

The following data needs to be provided for each of the above six boundary conditions:

For Die/Cast boundary type:

If die spray is applied, then specify

1. Spray duration.
2. Volume flow rate of the die spray.

For Die/Cool boundary type:

1. Coolant type oil or water.
2. Channel type fountain or flow thru.
3. Channel dimensions.
4. Coolant temperature.
5. Coolant flowrate.

For Die/Heater boundary type:

1. Heater dimensions.

2. Heater wattage.

For Die/Air boundary type:

1. Air temperature.
2. Heat transfer coefficient.

For Die/Die boundary type:

1. Contact type rear or side of the die component.
2. If air gap specify gap thickness.

For Die/Machine boundary type:

1. Machine frame temperature.
2. Machine locking force.
3. Die contact area.

Please refer to tables 5 and 6 for boundary condition data used for this report.

Calculating the Temperatures

The die temperatures can be calculated after specifying all the boundary conditions.

CASTHERM uses a mathematical technique known as the Boundary Element Method to calculate the temperature. The user needs to specify the boundary element size for the calculations. Using this number the program creates a boundary element mesh. Finer the mesh, longer it takes to calculate the temperatures.

The temperatures computed are average temperatures at the boundary element nodes.

Referring to figure 3 we can see that the temperature fluctuations in the die are periodic in

Table 5. Boundary Conditions - Plate.

***** DMT-CASTHERM BOUNDARY CONDITONS *****

Die Name: plate
Section Name: Ejector half
Units : SI Units

1. Boundary type Die/Cast
Thermal Zone flowoff
Sprayed for 1.00 secs at 0.30 l/min
Percentage heat removed 46.70
Heat Flux 162556.37 W/sq.m
2. Boundary type Die/Cast
Thermal Zone casting
Sprayed for 2.00 secs at 0.60 l/min
Percentage heat removed 37.87
Heat Flux 135713.49 W/sq.m
3. Boundary type Die/Cast
Thermal Zone runner
Sprayed for 4.00 secs at 0.60 l/min
Percentage heat removed 49.62
Heat Flux 253729.19 W/sq.m
4. Boundary type Die/Die
Side Contact
No air Gap
Hcoef 501.25 W/sq.m deg C
5. Boundary type Die/Mc
Machine Frame Temperature 50.00 deg C
Machine Locking Force 25.00 tonne
Die area in contact with frame 1600.00 sq.cm
Hcoef 286.89 W/sq.m deg C

Table 6. Boundary Conditions - Plate

***** DMT-CASTHERM BOUNDARY CONDITONS *****

Die Name: horse shoe
section Name: ejector half
Units : SI Units

Boundary type Die/Cast
Thermal Zone section1
Sprayed for 2.00 secs at 0.60 l/min
Percentage heat removed 53.35
Heat Flux 116116.66 W/sq.m
Boundary type Die/Cast
Thermal Zone section2
Sprayed for 2.00 secs at 0.60 l/min
Percentage heat removed 53.35
Heat Flux 116116.66 W/sq.m
Boundary type Die/Cast
Thermal Zone runner
Sprayed for 2.00 secs at 0.60 l/min
Percentage heat removed 18.83
Heat Flux 303047.71 W/sq.m
Boundary type Die/Cast
Thermal Zone biscuit
Sprayed for 2.00 secs at 0.60 l/min
Percentage heat removed 13.21
Heat Flux 270046.95 W/sq.m
Boundary type Die/Die
Side Contact
No air Gap
Hcoef 501.25 W/sq.m deg C
Boundary type Die/Die
Side Contact
No air Gap
Hcoef 501.25 W/sq.m deg C
Boundary type Die/Mc
Machine Frame Temperature 50.00 deg C
Machine Locking Force 600.00 tonne
Die area in contact with frame 3600.00 sq.cm
Hcoef 1368.41 W/sq.m deg C
Boundary type Die/Air
Air Temperature 25 deg C, Hcoef 36.48 W/sq.m deg C
Boundary type Die/Air
Air Temperature 25 deg C, Hcoef 36.48 W/sq.m deg C
Boundary type Die/Cool
Type: Water drilled through
Channel Diameter 9.50 mm
Coolant Temperature 50.00 deg C

Table 6 (continued). Boundary Conditions - Plate.

Coolant Flow Rate 3.00 l/min
Hcoef 4704.34 W/sq.m deg C
· Boundary type Die/Cool
Type: Water drilled through
Channel Diameter 9.50 mm
Coolant Temperature 50.00 deg C
Coolant Flow Rate 3.00 l/min
Hcoef 4704.34 W/sq.m deg C
· Boundary type Die/Cool
Type: Water drilled through
Channel Diameter 9.50 mm
Coolant Temperature 50.00 deg C
Coolant Flow Rate 3.00 l/min
Hcoef 4704.34 W/sq.m deg C
· Boundary type Die/Cool
Type: Water drilled through
Channel Diameter 9.50 mm
Coolant Temperature 50.00 deg C
Coolant Flow Rate 3.00 l/min
Hcoef 4704.34 W/sq.m deg C

Figure 3. Temperature Fluctuations About Average Die Temperature.

nature. During the metal injection stage, the temperature of die surfaces which are in contact with the molten metal rise to almost the pouring temperature of the metal. This temperature will then drop until the ejection temperature. If die spray is applied, the surface temperature drops rapidly which will rise again due to the heat transfer from the die. Two other observations can also be made from the same figure.

1. After certain number of die cycles, the surface temperatures of the die fluctuates around a constant average temperature, this is the temperature computed by the program.
2. The average temperature during the initial few die cycles is not constant and slowly increases to a constant value.

According to the CASTHERM manual, the average die surface temperature affects the casting quality. It should be the strategy of the die designer to ensure that there is a sufficient heating and/or cooling capacity such that this average temperature can be maintained at the appropriate level during production.

Simulation Times

The time to run the simulations for both the casting was about 2 minute on an IBM compatible 386 25MHz personal computer. It only takes about 15 to 20 seconds on a 486 machine. Cooling/heating lines can be moved or added or deleted in about few minutes.

Simulation Results

Please refer to figures 4, 5, 6 and 7 for the simulated temperature results for both plate and horse shoe castings. As can be seen from these figures the die surface temperatures are

Figure 4. Simulated Temperature Results - Plate

Figure 5. Simulated Temperature Results - Plate.

Figure 6. Simulated Temperature Results - Horseshoe.

Figure 7. Simulated Temperature Results - Horseshoe.

considerably lower in the cases with water cooling channels. Also please note that the temperatures are in the reasonable working range. If the temperatures are not in a reasonable range then we can either move the water channels, or change the dimensions of the water channels, or add/delete water channels or change the water flowrate through the water channels.

CASTHERM Review

Positives

1. The program is extremely easy to use.
2. The manual is well written and easy to read and use.
3. The program runs on IBM personal computer (386's and 486's).
4. Any person who can read die casting drawings, has some familiarity with computers and the die casting process can learn this software and be productive.
5. This software is extremely useful in locating cooling and heating lines in a short time. A typical casting might only take about 30 to 45 minutes. Thus this software can also be used improve die quotations.
6. Provides IGES file transfer format to and from other CAD systems. Thus models can be built in other CAD systems and imported into CASTHERM and vice versa.

Negatives

1. The program can only solve 2 dimensional problems. One needs to usually solve multiple sections to accurately locate cooling/heating lines.
2. The program calculates only average temperatures. This is good for locating cooling/heating lines, however we have no idea about the maximum temperatures, minimum temperatures and cooling rates which also affect the quality of castings and die life.

ProCAST

INTRODUCTION

ProCAST is a software package developed to meet the simulation requirements of any casting industry. It can do both 2d and 3d simulations using a finite element mesh. The full geometric complexity of industrial castings including the gating, risering, vents, flow-off, mold etc. can be modeled. Any casting and mold material can be modeled.

ProCAST Features:

Fluid Flow:

- 3d fluid flow, using Navier-Stokes and energy equations.
- Transient fluid flow, flow during the filling stage.
- Fluid turbulence modeling.
- Trapped gas model accounts for vents and sand permeability.

Heat Transfer:

- Transient nonlinear heat conduction in three dimensions.
- Conduction heat transfer in solid with fluid flow.

Radiation:

- Diffuse, gray body radiation model.
- Automatic view factor calculations with moving relative surfaces.
- Radial and mirror symmetry options

Solidification:

- Micro modeling, solidification kinetics.
- Enthalpy formulation for handling phase change.

Input data:

- Temperature dependent density, conductivity, specific heat, viscosity, and emissivity.
- Time and/or temperature dependent interface heat transfer coefficients.

Boundary Conditions:

- Time dependent temperature, pressure, and velocity boundary conditions.
- Time and/or temperature dependent film coefficients for flux boundary conditions.
- Pressure dependent velocity boundary conditions

Finite element mesh:

- Interfaces with PATRAN, IDEAS, IFEM, GFEM, ProEngineer and ANVIL,
- Brick, tetrahedron, wedge, quad and triangular elements.
- Coincident node technique for modeling the casting-mold interface.
- Automatic generation of coincident node faces.
- Multi-point constraints.

Simulation output:

- Automatic calculation of isochrons, temperature gradients, solidification rates, and cooling rates.
- Automatic calculation of metallurgical indicators for porosity, grain size and morphology, etc.
- Time-temperature curve plots.
- Sophisticated contour and vector plotting program for viewing all results.

ProCAST Software System:

A complete ProCAST analysis requires the use of five or six packages. The table below orders the flow of information through these software packages. The ProCAST software comes with PreCAST, DataCAST, ProCAST, ViewCAST and PostCAST software packages.

Software Package	Purpose
1. PATRAN, IDEAS, ANVIL	Built FEM model of casting/Mold
2. PreCAST	Apply boundary conditions, define materials
3. DataCAST	Check the model
4. ProCAST	Do simulations
5. PostCAST, ViewCAST, PATRAN and IDEAS	View simulation results

PATRAN, IDEAS or ANVIL can be used to build a finite element model for the analysis.

This model should contain at a minimum the following information:

1. Nodal x, y and z coordinates.
2. Element connectivity.
3. Material number assigned to each element.

PreCAST

PreCAST uses interactive menus to add more specific information to the finite element model.

With PreCAST, you can

1. Generate a coincident node interface.
2. Assign interface heat transfer coefficient data.
3. Locate slave-master nodes and their associated weights in a multi-point constraint.
4. Specify material properties.
5. Specify all required information to define the thermal and fluid flow boundary conditions.
6. Specify emissivities, temperatures and velocities to enclosure or solid elements.
7. Set run time options which will be used in ProCAST.
8. Build Two-dimensional FEM models.

DataCAST

DataCAST reviews the total model, performs extensive error checking, and converts all the units into a standard system (cgs). It also prepares a summary file which describes the complete analysis model. This summary file provides one form of model documentation.

ProCAST

ProCAST performs the simulation analysis. The results of the simulation are reported in both hard copy and post processing files. The post processing files can be generated using PostCAST. ProCAST can run in the background, this frees up the workstation.

PostCAST

PostCAST provides the following information:

1. Temperature results at each time step.
2. Pressure results at each time step.

3. Velocity results at each time step.
4. Heat flux results at each time step.
5. Time to reach a given temperature (Isochrons).
6. Temperature versus time results for various nodes.
7. Solidification rate, cooling rate, and temperature gradient results which can be combined into a factor indicative of microscopic features.
8. Row sum errors from the radiation model.

With the exception of item six above, all PostCAST results can be displayed graphically using PATRAN, IDEAS or ViewCAST

SOFTWARE & HARDWARE REQUIREMENTS

Software Requirements

UNIX operating system, a mesh building software like PATRAN or IDEAS or ANVIL or ProEngineer.

Currently porting ProCAST to Open VMS based operating software running on DEC systems.

Hardware Requirements

Workstations from: Sun, Silicon graphics, DEC, CRAY, Convex etc.

DIE DESCRIPTIONS

The die casting simulations performed using ProCAST were conducted with two die casting die drawings supplied by Ohio State University. These dies are briefly described below and CAD drawings have been included.

2mm Flat Plate Die

The part produced by this die is a simple 2mm thick plate. Four thermocouples were located in the die to record the time-temperature variation with die cycles. The die has water channels in both ejector and the moving half of the die. The cooling effect of the water channels is asymmetric.

Doehler-Jarvis Bowl Die

This die was from Doehler-Jarvis. The bowl casting is 165mm outside diameter with a floor thickness of 5mm. Thermocouples were located in the die at and near the inside surface of the bowl. The bowl wall is flat and 7mm thick at the contact location with the thermocouple probes.

Building models

PATRAN was used to build the models for the above two castings. Finite element computation time depends on the number of elements in the model. There are two ways to reduce the number of elements in a model. First, recognize symmetry in the casting/die. For this reason, only half the plate casting was modeled. Further simulation time reductions can be achieved by eliminating parts of the casting/die which are far away from the places which are of interest to us, as the effect would be insignificant. For this reason only quarter of the horse shoe casting was modeled. Second, by having the coarsest mesh we can get by. The mesh coarseness/fineness (i.e. less or more elements) depends on the accuracy sort in the computations. In general finer meshes provide more accurate/detailed information, however require longer simulation time. So it is better to have finer mesh where more detailed information is desired and coarser mesh elsewhere. This reduces the number of elements in the model. Knowing where we need a finer mesh and where we need coarser mesh is developed through experience.

After choosing the part of the casting/die to be modeled, make a rough sketch. Divide up this model into many small volumes. Each volume is such that it can be modeled using PATRAN and when meshed will have surface nodes matching with the surface nodes of the neighboring volumes. This way we get a model where the mesh is continuous. This is absolutely necessary for finite element simulation program to do computations.

Once the model is divided up into small volumes as explained above, we have to use the primitives available in PATRAN to construct and mesh each volume separately. The procedure starts off by inputting the corner coordinates of a volume. The corner coordinates are joined by lines, straight or curved based on the options available. Surfaces of each volume are formed by specifying the lines which form the boundaries of each surface. The volume itself is formed by specifying the end surfaces. Each volume needs to be meshed. The user has to ensure continuity of mesh between each volume. At the time of meshing, the material id of the volume needs to be specified. The user has to give a different material id for each part of the die (ejector half, moving half, water channels etc.). Mesh continuity has to be ensured across different material id's.

The Multi-point feature in ProCAST allows some tolerance with respect to mesh continuity across different material id's. In some cases, this could drastically reduce the number of elements required in the model and thus the simulation time. However, the results are only valid if pure thermal simulation is performed. In case of fluid flow simulations, pressure gradients are not well behaved across a multi-point interface.

After building the entire model, PATRAN features to compact and eliminate the unused the element and node id's are evoked. A neutral format version of the model is saved from the PATRAN. The format of this neutral file is such that it can be read into PreCAST for applying boundary conditions.

Learning PATRAN

The best way to learn PATRAN is attending their year-round seminars/training sessions. PATRAN cannot be learned completely in one training session. PATRAN learning is process, it's usage becomes easier with experience. This is similar to choosing the part of the casting/die to be modeled and choosing the mesh fineness in each part to the model (as explained above). A person who has the ability to read drawings and mentally visualize them will be a faster learner. In any case the time required to build the models (of similar complexity) reduces drastically with some experience. Please refer to the next section for examples of time reduction.

I have learned PATRAN through reading the manuals. I did not attend any training sessions or had any prior knowledge/experience. However, I had coursework in finite element methods. The PATRAN manuals are hard to read and understand to develop models. The best way I found was by executing the examples given in the manual. However, all the examples in the manual are for building components. Whereas for simulation it is necessary to build component (in our case casting and/or gating) and the die around it and ensure mesh continuity. This type of model building is an order of magnitude more complex than building just components. I had learned to build die casting models by examining some of the sample files I had obtained from ProCAST. I recommend to learn PATRAN by attending their training sessions and specifically asking them to help you with respect to die models.

Model building times

Learning PATRAN: 3 weeks

Plate Casting:

Time to make the rough sketch and divide up into small volumes (as discussed above) = 2 hours.

Time to build the finite element model using PATRAN = 10 hours.

Total time = 12 hours.

A second model of the plate casting die was made after gaining some simulation experience. It was found that the time required for simulation could be cut down by reducing the number of elements. Also the effect of round cross-section and square cross-section water channel was negligible. Thus, the second model had square cross-section water channels and less number of elements. The total time to build this model was much lower, about 6 hours.

Horse shoe casting:

Time to make the rough sketch = 3 hours

Time to build the finite element model = 8 hours

Total time = 11 hours

This casting was more complex than the simple plate casting. I had to further learn PATRAN commands to build the model. Time lost in trying out various methods to build the model is not included in the 11 hours mentioned above. The experience gained clearly indicated that an experience modeler could build the same model in about 4 to 5 hours.

Note on including water channels:

Both the above models were built with water channels in them. The same models without water channels would take considerably less time. For example the horse shoe casting would only take 8 hours compared to 11 hours. If a priori we know where to place the water channels, then we can build a model with water channels. In ProCAST, we can simulate with

and without the water channel by indicating the water channel material id as being present or empty at the start of simulation. Doing this has helped me to compare the simulation results, as the mesh remains the same and thus I had to plot the time-temperature curves at the same nodes in the model.

Learning ProCAST

ProCAST provides periodic two-day training courses. They also do on-site training.

I have learned ProCAST by watching the demos given at the university and also experimenting with the software on my own. The people at ProCAST were very helpful at answering my questions and doubts. It took me about 2 weeks to learn. However, I feel taking a training course from them would help to considerably reduce the training time required and learn the details of interest to you. You should keep in mind that ProCAST can do a lot more than you would ever use.

Applying boundary conditions using PreCAST

The neutral file generated by PATRAN is read using PreCAST. The pull down menus in PreCAST are arranged in the order in which they need to be used. The sequence is as follows:

1. Interface information.
2. Material information.
3. Boundary conditions.
4. Initial conditions.
5. Simulation parameter.

Interface information:

PreCAST automatically determines the pairs of material id's which are adjacent. If two material id's do not form a single part of the mold (like ejector half or moving half or water channel etc.) then an interface needs to be created. An interface heat transfer coefficient needs to be specified for each of the interfaces. This coefficient can be a constant or a function of time or a function of temperature. A database of interface coefficients is maintained and can be used. New entries can also be made.

The following information was used in the models simulated for this report:

Ejector half - moving half interface : Constant function $0.31 \text{ cal/cm}^2/\text{C}/\text{sec}$.

Casting - die interface : Time function $1.0 \text{ cal/cm}^2/\text{C}/\text{sec}$ till casting ejection and zero till end of die cycle.

Ejection time used for plate casting was 7 seconds.

Ejection time used for horse shoe casting was 10 seconds.

Cycle time used for plate casting was 46 seconds.

Cycle time used for horse casting was 10 seconds.

Material information:

ProCAST database provides data for some aluminum, magnesium, copper, nickel, steel and iron based alloys.

Examples of aluminum alloys: A390, Al 95.5% Cu 4.5%, Al 85% Cu 15%, Al 92% Mg 8%, Al 86% Mg 14%.

Examples of Magnesium alloys: AZ31B, AZ91B, HM11XA, K1A, ZK51A.

Example of Zinc: ZA-5.

The following new material information was entered into the database and used for this report:

Die material : H13

specific heat 0.11 cal/gm/c.c.

conductivity 0.0587 cal/cm/C/sec.

density 7.8 gm/c.c.

Casting material: Aluminum alloy

specific heat 0.23 cal/gm/C

conductivity 0.23 cal/cm/C/sec

fraction solid: 1.0 @ 538 °C and 0.0 @ 593 °C

density 2.74 gm/c.c.

solidus temperature 538 °C

liquidus temperature 593 °C

latent heat 93 cal/gm

viscosity was left blank as only pure thermal simulation was done for this report.

Boundary conditions:

Four types of boundary conditions were necessary for the simulations. The first boundary condition was due to air cooling of the die surface. This was modeled as simple cooling in air by specifying the surface film coefficient (a heat transfer coefficient) and the surrounding ambient temperature. The second boundary condition was at the planes of symmetry. At the planes of symmetry the heat flux was set to zero to simulate symmetric heat transfer conditions. The third boundary condition was for the casting-die interface to indicate air cooling of the die after the casting was ejected and also spraying effects. The fourth boundary

condition was for modeling water channels. In the simulation done for this report, the water channels were modeled as an heat sink. They can also be modeled by specifying a heat transfer coefficient for cooling in air, just like the external surfaces of the die.

The following information was used for the simulations:

External die surface cooling film coefficient = $0.00023 \text{ cal/cm}^2/\text{C}/\text{sec}$

Ambient temperature = 50°C

Casting-die interface boundary condition was a function of time as given below:

for plate casting

Time (seconds)	heat transfer coefficient ($\text{cal/cm}^2/\text{C}/\text{sec}$)	comment
0	0.0	molten metal injected
7	0.00023	casting ejected, die is air cooling
36	0.1	spray
38	0.00023	end of spray, die is air cooling
46	0.0	end of die cycle

for horse shoe casting

Time (seconds)	heat transfer coefficient ($\text{cal/cm}^2/\text{C}/\text{sec}$)	comment
0	0.0	molten metal injected
10	0.00023	casting ejected, die is air cooling
15	0.4	spray
17	0.00023	end of spray, die is air cooling
35	0.0	end of die cycle

Water channels were modeled as heat sinks with capacity of -13.3 cal/cc/sec.

Initial conditions:

The temperature of the die and casting at beginning of the simulations needs to be specified.

The simulation was set as die casting simulation. This makes sure that the die temperatures at the end of the previous die cycle are used. At the beginning of every die cycle the casting temperature is same as that specified in initial condition.

The following data was used in the simulations:

Initial temperature of the die = 100°C

Initial temperature of the casting = 650°C

Simulation parameters:

The following are some of the simulation parameters of interest:

Number of cycles = 5 (simulate for 5 die casting cycles).

Starting time step = 0.001 sec (time step for finite element calculations).

Maximum time step allowed = 0.25 sec (this upper limit to time step was necessary to assure significant number of calculations during the lubricant spraying and thus obtained a smoother time-temperature curve).

Simulation requirements

The following table shows the CPU time and hard disk requirement for the four simulations.

The simulations were done on a Sun Sparc Station 2.

Figure 1. Simulated Time & Temperature Curve - Plate Without Cooling

Figure 2. Simulated Time & Temperature Curve - Plate Without Cooling.

Figure 3. Simulated Time & Temperature Curve - Plate Without Cooling.

Figure 4. Simulated Time & Temperature Curve - Plate Without Cooling.

Figure 5. Simulated Time & Temperature Curve - Plate.

PROCAST TEMPERATURE VS TIME RESULTS

Figure 6. Simulated Time & Temperature Curve - Plate.

Figure 7. Simulated Time & Temperature Curve - Plate.

Figure 8. Simulated Time & Temperature Curve - Plate.

PROCAST TEMPERATURE VS. TIME RESULTS

Figure 9A. Plate Die.

Figure 9B. Plate Die.

should be. It is also possible that the casting does not need any water cooling.

Horse shoe casting

Please refer to figures 10 to 12 for time-temperature curves without water cooling. Please refer to Figure 13 for location of the time-temperature curves. The distance from the casting surface are 0.0, 0.01 and 0.5 inch.

From the time-temperature curves we can see that the die temperatures are slowly rising with each die cycle and also the temperature fluctuations are decreasing with distance from the casting surface. The time-temperature curves with water cooling at the same location are identical. This fact was verified by viewing the effect of water cooling channel at the temperature probe location. ViewCAST temperature plots with each time step clearly showed that the effect of water channel was not felt at the temperature probe location in the first five die cycles. This implies that the more die cycles need to be simulated to see the effect of water cooling.

ProCAST Review

Positives

1. ProCAST can do simultaneous simulation of fluid flow, micro modeling of solidification and heat transfer. In thin castings, part of the casting could solidify as the mold cavity is being filled.
2. It can also do pure fluid flow or heat transfer simulations.
3. It can model die spraying, heating elements, cooling channels, insulating materials etc.
4. Simulation results can be viewed using PostCAST, ViewCAST or any other post processing software like PATRAN.
5. The following are some of the types of simulation results that can be viewed:

Figure 10. Simulated Time & Temperature Curve - Horseshoe.

Figure 11. Simulated Time & Temperature Curve - Horseshoe.

Figure 12. Simulated Time & Temperature Curve - Horseshoe.

Figure 13. Horseshoe Die.

- Time - Temperature curves at any location in the casting or die.
 - Isochrons, heat fluxes, cooling rates, temperature gradients
 - Pressure, fluid velocities, turbulent energy, turbulent dissipation
 - Micro structure
 - A movie of temperature, fluid velocity etc. changes with each simulation step at any cross-section in the model.
 - Calculate and display criteria like : Niyama criteria for solidification porosity.
6. Results from each simulation give many insights about the die casting process. Some of the insights could be used for other castings. Thus using ProCAST has long term benefits in term of overall die casting process understanding.
 7. It can run in the background, i.e a user could start the program and the computer could be used for other purposes.
 8. The user could use PostCAST or ViewCAST and view the results of an simulation running in the background and decide whether to let the background simulation to continue or not.
 9. ProCAST is adding the code for residual stresses. When this new version is ready, we can predict the residual stresses in the castings, any deformations in the casting and the die.
 10. Die casting simulation is just one of the casting processes ProCAST can model. It can help in modeling new processes, or a combination of various processes.

Negatives

1. Changing the location of cooling/heating lines involves rebuilding the entire mesh and restarting the entire simulation. This is time consuming.
2. Building a model of the entire die and simulating it takes considerable time. Usually only part of the die (as explained in the section "Building Models") is modeled.
3. The user of the program needs to be familiar with reading die casting drawings, finite

element mesh generation and have the necessary background in fluid flow, solidification and heat transfer to be able to interpret the results and take corrective actions.

4. Simulating 20 to 30 die cycles to attain a constant temperature fluctuation takes considerable time. However, this simulation time can be reduced by guessing the approximate average die temperature and starting the simulation with this temperature as the initial die temperature. With this trick the number of die cycles required for attaining constant temperature fluctuations can be reduced to around five die cycles.

Sub-Project 4

Selection and Characterization of Aluminum-based Die Casting Alloys

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**Final Report
for the Aluminum Die Casting Alloy Research Project**

**A STUDY OF ALLOY-MICROSTRUCTURE-
PERFORMANCE INTERACTION**

Phase I

Submitted to
DOE, NADCA, UNI and Task Group
by
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June 1993

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Final Report
for the Aluminum Die Casting Alloy Research Project
Phase I

1 Introduction

The aluminum die casting alloy research project is part of the die casting research program, which is sponsored by DOE and NADCA and managed by UNI. The aluminum die casting alloy research has been conducted at the Aluminum Casting Research Laboratory (ACRL) at WPI under subcontract 22243-08-B between WPI and UNI. The project coordinator was Prof. D. Apelian and the work was conducted by Dr. L. Wang and ACRL staff and students. This research is supervised by a task group headed by Mr. Herb Brucher (Doehler-Jarvis).

The research project was divided into two phases: Phase I included a critical literature review and a low resolution study of 380 alloy. Phase II will be a detailed study of alloy - microstructure - performance interaction. Phase I has been completed and took 18 months to complete. Officially, the project started in April 1992, when funding was approved. The literature review actually began in Feb. 1992 and was completed by April 30, 1993. In July 1992 a preliminary literature review was submitted to DOE, UNI and the chairman of the task group. The study of the mechanical properties of 380 alloy on a low resolution basis started in Nov. 1992. The microstructure analysis was divided into two parts. Part I: the effects of cooling conditions on the microstructure of 380 alloy was conducted during August through October 1992. Part II: the microstructure analysis of the test specimens was carried out along with the mechanical property tests. Early in the project, a Gantt Chart (the detailed time schedule and tasks) was formulated. This schedule proposed the end of May 1993 as the due time for the completion of Phase I. The schedule was later revised twice, and the time for the completion of the study was moved to June 15, 1993. The work in Phase I has been completed in accordance with the revised Gantt Chart. During the process, reports on the progress of the project were periodically sent to UNI and to the chairman of the task group, Mr. Herb Brucher. In addition, the progress and the status of the project were presented at two workshops held in July 1992 and April 1993.

The following is the final report on Phase I of the project. The report includes a general statement and two appendixes. The general statement describes the overall conditions and the time frame for the fulfillments of the tasks proposed in the original plan. This section will also highlight the research process, and briefly discuss the results in the two research areas: the critical literature review and the experiments. The two appendixes are (A) a critical literature review of aluminum die casting alloys and (B) a detailed technical report for the low resolution study of 380 alloy.

2 Critical Literature Review

Objectives

Aluminum die casting alloys are the most widely used alloys in the die casting industry. They constitute the largest part of the die casting metals measured by both weight and dollar value. The total die casting industry sales for North America amounted to \$7.7 billion in 1989. The share of aluminum alloys was about \$4.7 billion, i.e., about 61 % of the total die casting sales. In the expanding use of aluminum alloys in die casting, research and development work have played a central role. A large effort has been made in this area, and many relevant articles have been published in a number of journals and publications. However, there has not been a thorough review paper dedicated exclusively to aluminum die casting alloys. A thorough review of the current status of research and development in aluminum die casting alloys is essential to the future study and development of alloys. In order to explore the potential properties, expand the applications of aluminum die casting alloys, and compete with other materials and with foreign competitors in the global market, the American die casting industry recognizes the necessity for a intensive research. Accordingly, DOE and NADCA initiated and funded this project: "a study of alloy - microstructure - performance interaction". A thorough critical literature review of aluminum die casting alloys was proposed as the first step in the research project . The objectives of the review were:

- To summarize the current knowledge and collect composition and property data of aluminum die casting alloys.
- To gather information about the recent research and development work in aluminum die casting alloys.

- Based on the current status of aluminum die casting alloy research, help set detailed research goals and plans for phase II of the project.

Since this research aims to serve the die casting industry, the review was focused on the practical aspects of aluminum die casting alloys and related research. It did not cover pure theoretical investigation. Also, it did not include research areas of aluminum casting alloys, such as modification of eutectic Si, grain refinement, heat treatment and porosity. These have been studied extensively and frequently reviewed.

Literature Search and Information Collection

The literature search was conducted mainly at WPI and MIT libraries and from the ACRL publication collections. The titles of recent published papers were first searched from World Aluminum Abstracts, and Metal Abstracts. Electronic retrieval was conducted under the key words "Casting Alloy" from the World Aluminum Abstracts, starting from 1968. The Metal Abstracts was combed under subtitle "Aluminum Based Alloys" for the last ten years. From the abstracts of the papers, whose titles are related to aluminum die casting alloys, papers of interest were selected. These papers were searched and key papers were reviewed. Based on the information from relevant papers, books, data from reference books and information and knowledge accumulated in ACRL's research, the critical literature review was written and compiled. It contains the following:

- Introduction
- Classification of Die Casting Alloys
- Specifications of Aluminum Die Casting Alloys
- Effects of Alloying Elements and Impurities
- Solidification and Microstructure
- Properties
- Typical Die Casting Alloys
- Discussions
- References

Highlights of the Critical Literature Review

• The literature indicates that a large effort has gone into the research and development of aluminum alloys. The work has resulted in an ever increasing number of alloys, alloy standardization, a better understanding of the fundamentals of aluminum alloy, a great improvement in properties, and broadening the applications of aluminum alloys. Research on aluminum and aluminum alloys has been accelerated in the last two decades, this is reflected in the increasing number of papers published each year.

• As an important segment of aluminum alloys, aluminum die casting alloys have also seen remarkable development since the cold-chamber die casting machine was invented. During the last half century, numerous Al die casting alloys were created. In the Handbook of International Alloy Compositions and Designations, Vol. III, Aluminum, 1980, more than 300 aluminum alloys were recorded worldwide for die castings (not including ingot alloys). In American Aluminum Association (AA) standardized aluminum alloys, 23 are for die casting. These cover almost all the existing Al die casting alloys commercially used worldwide in terms of compositions.

• These 23 AA issued aluminum die casting alloys can be divided into several groups. A representative alloy from each group is listed below.

380.0 is an Al-Si-Cu system alloy. It is generally considered the best "all-purpose" die casting alloy and is the most commonly used. This alloy has very good castability, good mechanical properties, and generous limitations on impurities, however, it has only fair corrosion resistant properties.

360.0 is an Al-Si-Mg system alloy. It is also a general purpose alloy. This alloy has very good castability, good corrosion resistance, but fair machinability, less generous impurity limitations, and slightly lower tensile strength and elongation than 380.0.

390.0 is a hypereutectic Al-Si-Cu-Mg system alloy. It has outstanding wear resistance, low thermal expansion coefficient, high thermal conductivity, elevated temperature strength and hardness, very good castability, but fair

corrosion resistance, low ductility and poor machinability. It is mainly used to replace grey cast iron in engine parts and as wear resistant components.

413.0 is basically an eutectic Al-Si binary alloy. It has excellent castability, very good corrosion resistance, but poor machinability and poor surface finish ability. It suits castings with large, complex shape, thin wall and good corrosion resistance.

C443.0 is basically a hypoeutectic Al-Si binary alloy. It has high ductility, very good corrosion resistance, good machinability, but fair castability, low strength. This alloy is used for the castings, where above-average ductility and excellent corrosion resistance are needed.

518.0 is an Al-Mg system alloy. It has excellent corrosion resistance, high strength, high ductility, excellent machinability, good polishing characteristics, good fatigue property, low density, but poor castability. This alloy suits the castings requiring the highest corrosion resistance with no complicated or intricate shape, also ornamental parts and hardware.

In all the Al die casting alloys the main alloying elements are Si, Cu, Mg, and Fe, and the common impurities are Mn, Ni, Cr, Zn, Sn and Pb. The Al die casting alloys may also contain melt treatment elements, such as Ti, B, Na, Sr, Sb and P. Their major effects on the aluminum die casting alloys are as follows:

Si imparts good castability, increases strength, reduces ductility, but does not affect corrosion resistance.

Cu increases hardness and strength, but reduces corrosion resistance.

Mg makes Al-Mg alloys have bright surface, high ductility, and excellent corrosion resistance, but poor castability. In Al-Si alloys Mg associated with Si increases strength through heat treatment, but may form oxides and intermetallic inclusions.

Fe is generally the major impurity in Al casting alloys. It reduces ductility and may form intermetallics with other elements and cause hard spots. However,

in Al die casting alloys Fe is added as an alloying element to reduce die soldering.

Mn, Co, Ni, Mo, and Be can act as Fe correctors to reduce the Fe embrittlement effect.

Mn is a commonly used Fe corrector, but it may form primary intermetallics with Fe and Si as hard spots.

Ni enhances elevated temperature properties.

Cr slightly increases strength, but may form sludge.

Pb and Sn improve machinability of Al-Si alloys.

Zn has no significant effect.

Be reduces oxidation and the formation of inclusions in Mg-bearing alloys.

The literature search shows that though a great deal of research and development work has been conducted on Al alloys, relatively little work has been done on die casting alloys, especially in the U. S. For example, searching Metal Abstracts for the past ten years more than 30,000 papers were collected under subtitle "Aluminum Base Alloys." However, only about 1000 papers were directly associated with aluminum casting alloys. Of the 1000 papers, about 300 were directly related to die casting, from which about 200 were technical and almost half of them were written in foreign languages, mostly German, Japanese, Russian, or French.

The research and development work on Al die casting alloys conducted thus far is scattered and does not provide sufficient information to systematically evaluate and optimize the alloys. The information provided also does not fully reflect the progress of the die casting techniques, the advances in quality control of alloys and castings, or reflect the accumulated knowledge about the alloys. For example, for the major Al die casting alloys, such as 380.0, 360.0, and 518.0 (former name 218), the compositions and the property data found in current references are almost the same as those published in the 1950's in Metal Digest. During the four decades, progress in die

casting techniques, alloy treatment and quality control have resulted in considerable improvement in the casting quality and properties of these alloys, but these improvements are not reflected in the current literature. Casting designers and producers have for long had questions about the scientific basis for setting the contents of alloying and impurity elements in aluminum die casting alloys. Though the effects of these elements, individually and in combinations, on microstructure and properties have been repeatedly investigated and the results have been published, questions still remain. Take 380.0 as an example, the upper limit of the total alloying and impurity element contents (20.45%) differs from the lower limit (total 11.8%, assuming the lower limit for Fe is 1.3%) by about 8.65%. A variation of alloying and impurity elements in an aluminum alloy by 8.65% will make a considerable difference in the microstructure and properties of the alloy. However, only one value for each property is given in the current standards. Is it the lowest, mean or average value? If it is the lowest, which composition does the value correspond to? If it is mean or average, in what range do the properties vary within the composition limits? More importantly, what is the optimum composition to meet certain requirements (i. e. certain properties)? Which element(s) is critical and which can be allowed to vary in a wider range for certain requirements? All these questions need to be answered.

Some of the data and information currently available can not be trusted by die casting designers and producers. For example:

On many occasions the data from different sources, or even in the same reference, may be conflicting. For example, the tensile strength of 380.0 alloy is shown as 48 ksi in the ASM Metal Handbook, as 46 ksi in ASTM B85, and as 43 ksi in Metal Digest (July 1982, A1-239). In Metal Digest (April 1983, A1-245), the room temperature properties of 360.0 alloy are tensile strength 44 ksi, elongation 2.5%, but in the same reference the data listed in typical short-time elevated temperature tensile properties show that at 75 °F (24 °C) for 30 min the tensile strength is 47 ksi and the elongation is 3%. Is this alloy so temperature sensitive? The testing standard does not set strict temperature requirement for the room temperature property test. What is the room temperature? Can the properties be used for other temperatures, for example, 16 °C?

Accordingly, designers generally use about 65-75% of the test specimen properties in their calculations. This results in only very few aluminum alloy castings, including die castings, are currently used in sensitive applications, where high reliability is required.

- The 65 - 75 % factor used by designer is a reflection of their mis-trust in the published property data. This mistrust is brought about, not only by the contradicting published values explained above, but also by the fact that the designer believe, and rightfully so, that the properties of a casting will vary with the mold design, the production conditions, the operating parameters, and variation in the alloy chemistry within the prescribed limits. Test specimens, which are the source of the published property data, are usually produced under controlled conditions, and so do not reflect the effect of these parameters. Several attempts were made to relate test specimen properties to actual casting properties; however, these few studies did not result in generally acceptable relationships.

- A systematic study of composition-microstructure-property relationship of Al die casting alloys is therefore necessary for the Al die casting industry and would have great technical and economical significance. Designers need reliable property data to evaluate casting performance. Alloy and casting producers need to understand how composition and casting conditions affect the properties and how to optimize the composition and production conditions for certain requirements. The systematic study will provide this information and data and will help build the foundation for the production of more reliable and lighter aluminum die casting components while reducing production costs, and opening new markets for aluminum alloy die castings.

3 Low Resolution Study of 380 alloy

As part of Phase I of the Aluminum Die Casting Alloy Research Project a low resolution study was performed on alloy 380. The objective of this study is two folds:

- To investigate the alloy-microstructure-performance interaction, and
- To determine whether or not a more comprehensive study is needed.

The low resolution study was performed on 380 alloy since this is the most commonly used aluminum die casting alloy. Two compositions representing the two extreme

limits of the alloy chemistry were chosen. The effects of the alloy chemistry on its microstructures and its mechanical properties were evaluated and documented.

3-1 Mechanical Property test

Specimen Preparation

Wabash alloys, Gibbs Die Casting Corp, and Doehler Jarvis produced and inspected the alloy and specimens. All expenses for specimen preparation are their donation to this project.

Wabash Alloys produced the two compositions of the alloys. The compositions of the alloys produced and the alloy specification are shown in the following Tables. 1709 lbs of the L alloy (composition at the lower limits of the alloy specification) and 1721 lbs of the H alloy (composition at the upper limits of the alloy specification) were produced. The total cost of the alloys was \$6860.00.

Compositions of the L and H alloys used in the tensile property tests

	Cu	Si	Fe	Mn	Mg	Zn	Ni	Pb	Sn	Ti	Cr		Al
L	2.90	7.37	0.67	0.24	0.02	0.85	0.02	0.02	0.01	0.03	0.02		Remainder
H	3.91	9.51	1.27	0.28	0.27	2.89	0.53	0.06	0.03	0.04	0.08		Remainder

Specification of the A380.0 alloy composition

	Cu	Si	Fe	Mn	Mg	Zn	Ni	Pb	Sn	Ti	Cr	Other Total	Al
A380	3.0-4.0	7.5-9.5	1.3	0.50	0.10	3.0	0.50		0.35			0.50	Remainder

Gibbs Die casting Corp die cast one hundred test bars, fifty for each composition. The test bars were a modification of the standard tensile test specimen – ASTM B557 - 84. The modification is in the grip sections of the specimen. These sections were made flat instead of round. The flat surface was used for hardness tests.

Doehler Jarvis x-ray inspected all the test bars. Eight specimens of alloy L and four specimens of alloy H were found to have defects within their reduced section. Eighty test bars, forty for each composition, were sent to WPI for testing and analysis.

Testing

Mechanical property tests were conducted at WPI. An Instron testing machine, Model 8502 was used. The testing procedure followed the ASTM standard B 557 - 84, "Standard Methods of Tension Testing Wrought and Cast Aluminum- and Magnesium-Alloy Products". Tensile strength, yield strength, and elongation were measured. A chart recorder was used to record and display load-displacement diagram. The data were also stored in a computer for analysis. Forty specimens for each composition were tested. Due to apparent defects some specimens showed very low properties and some curves did not show an initial straight line (elastic deformation range) to allow property calculation. Therefore, the data from only 35 specimens of each composition were used in the property calculations. Hardness tests were carried out on a Rockwell Hardness Tester, Model 3JR, Number 10661. 20 specimens for each composition were tested. Ten measurements were taken on each specimen.

Results

The tensile property and hardness values obtained in this study are shown in following Table. The Standard Brinell Hardness given in the Table was converted from the Rockwell Hardness Scale B measurements. Typical properties, as listed in different references, are also shown in the Table for comparison. Standard deviation for each property value is given.

**The measured tensile property and hardness values of A380.0 alloy at two composition levels
and the Specifications of the Properties from Different Sources**

		Tensile Strength ksi (MPa)	Yield Strength ksi (MPa)	Elongation %	Hardness Rb (HB)
L Alloy	Average	44.94±1.68 (309.86±11.58)	20.42±1.91 (140.80±13.17)	6.46±0.85	33.4±4.5 (69.4±9.4)
	Min/Max	41.38 / 48.22	16.97 / 24.41	4.3 / 8.2	21.2 / 47.0
H Alloy	Average	50.17±1.04 (345.92±7.17)	32.17±1.48 (221.81±10.20)	2.55±0.28	62.3±3.5 (98.3±5.5)
	Min/Max	48.22 / 52.87	29.08 / 35.00	2.00 / 3.05	49.6 / 71.0

Specifications of A380 Tensile and Hardness Properties in Different References

1*	47 (324)	23 (159)	4.0	(75)
2*	47 (325)	23 (160)	4.0**	
3*	47 (320)	23 (160)	3.5**	(75)
4*	47 (325)	23 (160)	3.5**	
5*	47 (325)	23 (158)	3.5**	

- * 1 Rooy, E. L., "Aluminum and Aluminum Alloys", ASM Handbook, 9th Edition, Vol. 15, p743-770.
- 2 ASM Metal Handbook, 10th Edition, Vol. 2, 1992.
- 3 1992 Annual Book of ASTM Standard, Vol 15, 1990.
- 4 ASM Metal Handbook, 9th Edition, Vol. 15, 1990.
- 5 "Aluminum Alloy A380 (Aluminum Die Casting Alloy)", Metal Digest, Al-68, June 1986.
- ** Tests were conducted on 2 inch diameter specimens.

3-2 Microstructure of 380 alloy

A microstructural analysis of 380 alloy was conducted. The objective of the study was to characterize the microstructure of the test specimens and to investigate the effects of the casting conditions on the microstructure.

Microstructure and Phase Identification

Several broken specimens of each composition were sectioned at different locations. The samples were mounted, polished, and examined by optical and scanning electronic microscopy (SEM). An energy dispersive x-ray spectrometer (EDX) was used for phase identification and composition analysis. An image analysis system was used to evaluate the size and distribution of the principal phases. The samples were also etched for dendrite arm spacing (DAS) and grain size measurements. The following features were observed in the specimens.

- Generally, the microstructure of both alloys consist of alpha aluminum dendrites, a mixture of fine acicular and near modified eutectic Si particles, primary Si particles, Fe- and Cu-bearing phases and some complex intermetallics in the interdendritic areas. The microstructures of both alloys were very fine, DAS ranging between 5 and 10 μm . However, the morphology, size and distribution of the phases present in the L alloy differed significantly from those in the H alloy and from those in sand and permanent mold castings .
- When the polished surfaces of the samples were observed visually, the cross sectional area of L alloy specimens showed a ring-shape segregation consisting of three distinct regions, a core area, a ring surrounding the core, and an outer layer. The outer layer was about 1000 μm in width and the ring was about 400 μm in width. The DAS was slightly coarser at the center area than in the outer layer and there were microstructural differences among these three regions.
- In the ring region, larger amounts of Cu-bearing phase and Si eutectic structure were detected than in the outer layer and the core area. The majority of Cu-bearing particles were highly branched with sizes of 5 to 10 μm . In the outer layer and the core area, dominant Cu-bearing phases were small individual particles elongated along the interdendritic areas. The eutectic Si particles were

coarser and primary Si particles were only observed in the ring region for the L alloy specimens; with sizes of 2 to 4 μm .

- The Fe-bearing phase were present as polygonal particles, no Chinese script or needle phases were observed, which are the dominant Fe-bearing phases in sand and permanent mold castings of similar alloy compositions. The sizes of these polygonal particles were distributed in two distinct ranges. There were a few larger particles with diameters ranging from 5 to 10 μm . The remaining particles were small with sizes 1 to 2 μm , but in large amounts. The concentration of the Fe-bearing particles was apparently denser in the ring and the outer layer regions than in the core area.
- In the H alloy specimens, the ring structure like that in the L alloy specimens was not observed visually. However, slightly larger amounts of fine eutectic Si were observed at locations very close to the sample surface. Toward the center the eutectic Si particles became less, but coarser, with the coarsening of the dendritic structure. No obvious Cu- and Fe-bearing phases segregated.
- Due to the larger Si contents, the H alloy specimens contained more Si phases, both eutectic and primary. In the thin outer layer, about 500 μm in width, the dominant Si phase was fine eutectic. Away from this layer towards the center, considerable amounts of primary Si particles precipitated. In the L alloy, however, there were only small amounts of the primary Si found in the ring area. The sizes of the primary Si in the H alloy were also slightly greater than those in the L alloy (4 to 6 μm vs 2 to 4 μm).
- The Fe- and Cu-bearing phases were also in larger amounts in the H alloy than in the L alloy specimens. The Fe-bearing phase consisted of large amounts of the larger particles, unlike in the L alloy where the dominant Fe-bearing particles were the smaller ones.
- The non-uniformity in the microstructure was observed not only across the cross sectional area, but also along the length of the specimen. Generally, the structure was coarse at the location close to the gate and was refined with the distance of melt flow. However, little difference was found between the middle of the reduced section and the end of the specimen.

Microstructures of 380 alloy solidified under different cooling rates

In this part of the study the effects of the cooling condition (cooling rates) on the microstructure of 380 alloy was investigated. The initial objectives were to estimate the cooling rates in aluminum alloy die castings, and to provide information for relating the properties measured from test specimens to the properties of die castings. A 380 alloy, from Doehler Jarvis production line, was investigated. Furnace cooling, sand mold, graphite mold, steel mold, water cooled copper mold, and wedged copper mold were used to provide different cooling rates. Three thermocouples were placed in each mold to record the thermal history, from which the cooling rates were calculated. The samples were then sectioned and analyzed. The solidification data obtained from these experiments are shown in the Table on the next page. By comparing these microstructure with those of die castings of the same alloy it is found that the cooling rates die castings are in the range of 100 to 150 °C/s.

3-3 Fracture surface analysis

Fractographic analysis was conducted on several specimens of each composition. The fracture surfaces of H alloy specimens show small facets dominant on the surfaces, which are the cleaved Si particles, and a little fibrous structure. The overall fracture is brittle in nature. The fracture surfaces of L alloy specimens also show small facet, but more fibrous structure, indicating that the specimens underwent significant deformation. On the surfaces of specimens of both alloys macro- and micro-porosity and inclusions are detected.

3-4 Discussions and suggested future work

The mechanical property tests show that the variation of composition of 380 alloy within its specification has significant effects on the mechanical properties. For example, the tensile strength, yield strength and the hardness (Rb) of H alloy specimens are 11.6%, 57.5% and 86.5% respectively higher than those of L alloy specimens. Elongation of L alloy specimens is 153% higher than that of H alloy specimens.

Summary of Solidification Data of 380 Alloy Under Different Cooling Rates

Casting Condition		Cooling Rate (C/s)	DAS (μm)	Microstructural Characteristics
Furnace Cooled		0.015	100.5	Very coarse Si flakes, huge Fe rich needles, big equiaxed chinese script, eutectic-like and clusters of large Cu rich lumpy particles
Sand Mold		0.14	59.7	No Fe rich needle, chinese script elongated, and disintegrated, no primary silicon particles
Graphite Mold		1.9	24.8	No Fe rich needle, small amount of chinese script confined in interdendritic areas, irregular polygonal Fe rich phase and primary silicon particles appear.
Steel Mold		2.2	17.7	Similar to but finer than in graphite mold
Water Colled Copper Mold		3.4	15.8	Similar to but even finer than in graphite mold
Wedged Copper Mold	Top	91.7	13.5	Similar to but much finer than in graphite mold
	Middle	~120	6	Finer than in top portion, no Chinese scrip phase found
	Bottom	>500	3.1	All Fe and Cu rich phases confined in interdendritic areas as lumpy particles, very fine Si particles

- This property variation within the specified composition range was not reflected in existing specifications and no work in the published literature shows these data.

- The effects of many elements on mechanical properties of aluminum die casting alloys have been studied. For example, increasing Si, Cu, and Fe contents increases strength and reduces ductility, this was confirmed in this study. The H alloy, which has higher Si, Cu and Fe contents, possesses higher tensile strength, yield strength and hardness, but lower ductility than the L alloy. However, several questions remain to be answered. For example, to what extent these elements affect the properties, individually and in combination. What element makes the most contribution to the property change, and to which property? What is the optimum composition for certain property requirements?

- The mechanical properties of the as-cast die castings are affected not only by the alloy composition, but also by other factors, such as casting condition, gating design, defects, and outside environmental effects. Some effects of these factors were reflected in this study. For example, though specimens of the same alloy in this study had the same composition, and were produced under controlled conditions following the same procedure, the properties still varied. Like among the L alloy specimens, the tensile strength and elongation varied from 41.38 to 48.22 ksi and 4.3 to 8.2% .

- The microstructure at the reduced section of the tensile specimens was similar to the microstructure between the upper and middle portions of the wedged copper mold casting. By comparing the DAS it is estimated that the cooling rate at this portion of the tensile specimen was about 100 to 150 °C/s.

- In the die casting, the high cooling rate, fast filling of the die cavity and solidifying under high pressure result in fine, unique microstructure in A380 alloy. This microstructure is very different from that in sand and permanent mold castings. In addition, the variation in composition affects the microstructure significantly. The difference in microstructure presumably causes the variation in mechanical properties.

Due to time constraints, the effects of impurity on microstructure were not investigated in this phase of the study. The variation in the microstructure between the H and L alloys may not be due to the effect of the alloying elements only. In 380 die casting alloy the total impurity contents could be 4.95%, which should affect the microstructure, and consequently, affect the properties significantly. Their effects should not be ignored.

Based on the experiments and the literature review, it is evident that a detailed systematic study of the alloy-microstructure-performance relationship is definitely necessary for aluminum die casting alloy development and for the competitiveness of the U. S. die casting industry.

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Review:

ALUMINUM DIE CASTING ALLOYS

Submitted to:

DOE, NADCA, UNI and Task Group

as part of the Phase I

of Aluminum Die Casting Alloy Research Project

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I Introduction

When aluminum was first discovered 176 years ago [1, 2], it was very precious because it was difficult to produce. For a long time it was only used for decoration and utensils. It was the invention of the electrolytic reduction method (Hall-Heroult Method) that made possible the industrial production of aluminum and its alloys. Because of their many superior properties, aluminum alloys have had ever expanding use since they were mass produced. In 1991, the world consumption of aluminum alloys reached 19841.5×10^3 tons, and the American consumption was 5933.8×10^3 tons[3].

Since the commercial use of aluminum alloys, casting has been one of the most common methods of manufacturing. Generally, Al alloy components can be produced by almost all casting methods. The most commonly used casting processes are sand, permanent mold, and die casting. Among all the processes die casting produces the largest portion of all Al alloy castings. In the 1980's about 68% of the total Al alloy castings were die cast[4]. Die casting of Al alloys was made possible by the development of the cold chamber die casting machine and the improvement of die materials. The demands for large amounts of identical Al alloy parts, in the transportation, construction and appliance industries, have resulted in a broad market for Al alloy die castings. This has been especially true since the "energy crisis" of the 1970's when the automobile industry began to make cars more fuel efficient by reducing weight. More and more, Al alloy parts have been introduced in cars and other vehicles, replacing ferrous alloys. Estimates are that by the end of this decade Al usage could increase almost 30% to 210 lbs per vehicle[5]. The Al alloy die castings have gained dramatically in the last two decades. In the die casting industry Al remains the dominant material. In 1992, a total of 405,000 tons of Al die castings were produced in the U. S. while the Mg and Zn die castings were totalled 18,500 and 117,000 tons, respectively[6]. The total die casting industry sales for North American amounted to \$7.7 billion in 1989[7]. The share for Al was about \$4.7 billion, consisting of about 61% of the total die casting sales (see Fig. 1-1). After a two-year decline caused by the recession, 1992 seemed to be the beginning of a slow climb. The forecast for 1992 yielded a projection of \$6.9 billion for the total sales. While there was still a drop in the total sales figure from 1989, the sales for Al die castings were

estimated to be about \$4.7 billion, the same as that in 1989. The trend for 1993 continues upward. The forecast is \$7.3 billion for the total sales and \$5 billion for Al castings. Fig. 1-2 shows total the U. S. Al die casting shipments during 1989-93. The estimated shipment total for 1993 is over 1.5 billion pounds.

In addition to the demands of the market and progress made in technology and equipment, the research and development of Al alloys have also played a very important role in their dramatic growth. For example, the age hardening capacity of the system Al-Cu-Mg found by Alfred Wilm in 1907^[2], and the modifying technique of hypoeutectic Al-Si alloys patented by A. Pacz in 1921^[8] have both greatly contributed to the expansion of utilizing this most rewarding nonferrous commodity. Because of the expanding application of Al alloys and the competition with other materials, the research and development of Al alloys have been intensified in recent years. Just browsing through some information journals, such as Metal Abstracts, one will clearly see that the number of papers related to Al alloys, increases every year. For example, Table 1-1 lists the number of collected papers under the subtitle "Aluminum Base Alloys" in Metal Abstracts. It shows that in the last ten years the number of published papers related to research and development of Al alloys has doubled.

TABLE 1-1 Number of Papers under Subtitle "Aluminum Base Alloys"
in Metal Abstracts

YEAR	1970	1980	1985	1990	1991	1992
Number of Papers	1100	2300	2700	4000	4950	>6000

This research has resulted in the development and applications of new alloys, such as, Al alloy composites, Al-Li system alloys, hypereutectic Al-Si alloys. In addition, increased knowledge and the improvement in research techniques have led to a better understanding of the phenomena occurring in Al alloys, such as, during solidification, melt treatment, heat treatment. The results have helped to optimize the production processes and enhanced the properties of the existing alloys. It has been reported ^[9] that now the normal level of mechanical properties is generally well above that obtainable as recently as fifteen years ago. The research and development work on Al alloys is certain to continue in the future to meet new demands and competition

Fig. 1-1 North American die casting shipments by alloy type

Fig. 1-2 Total U. S. aluminum die casting shipment

However, despite the large number of papers published each year on aluminum based alloys, only a very small portion deal with Al die casting alloys. By electronic retrieval from World Aluminum Abstracts starting from 1968 and searching Metal Abstracts for the last ten years, only about 200 papers were found, which are related directly to some aspects of Al die casting alloys. Of this number most of the work was conducted outside of the U. S. and about half of the technical papers were written in Japanese, German, Russian or French. This percentage does not correspond to the portion of Al die castings produced in the U. S. If this situation continues it would certainly not help Al casting alloys to compete with other materials, nor would it help the U. S. Al die casting industry to compete in the world market.

The present work will offer a critical review of all the available research and development work on Al die casting alloys. It will summarize what has been done in recent years, recount the achievements, and analyze what still needs to be done in the field of Al die casting alloy research, which, in turn, would benefit the die casting industry.

This review will be presented in two sections. The first section will give a general description of Al die casting alloys and include the following: alloy classification, specifications, effect of alloying and impurity elements, microstructure, and properties. The second section will concentrate on some of the most commonly used Al die casting alloys, such as 360, 380, 390, 413, C443, and 518.

II Classification of Aluminum Casting Alloys

Almost all elements can be added to Al, either by solidifying in Al as a solid solution or forming intermetallics. By adding different elements and their combinations, numerous alloys have been developed. In order to simplify alloy selection and arranging production, major industrial countries have specified their alloys. Based on existing alloys, the American Standard of Test Materials (ASTM) has issued a Cast Aluminum and Aluminum Alloy Designation System (ANSI/ASC H35.1 - 1990)^[10] as a national standard, classifying all the casting Al alloys.

In this standard, a system of four digit numerical designations is used to identify Al and Al alloys in the form of castings and foundry ingots. The first digit indicates the alloy group as follows:

Al, 99.00% and greater	1xx.x
Al alloys grouped by major alloying elements	
Copper	2xx.x
Silicon, with added Cu and/or Mg	
	3xx.x
Silicon	4xx.x
Magnesium	5xx.x
Zinc	7xx.x
Tin	8xx.x
Other Element	9xx.x
Unused Series	6xx.x

In the 1xx.x group, the second two of the four digits indicate the minimum Al percentage. These digits are the same as the two digits to the right of the decimal point in the minimum Al percentage when it is expressed to the nearest 0.01%.

In the 2xx.x through 9xx.x alloy groups, the second two of the four digits have no special significance but serve only to identify the different alloys.

The last digit, which is to the right of the decimal point indicates the product form. xxx.0 indicates casting. xxx.1 and xxx.2 indicate ingot.

For xxx.1 ingot, limits for alloying elements and impurities are the same as for the alloys in the form of castings except for the followings:

For Die Casting	For Ingot
Maximum Fe%	
Up thru 1.3	0.3 less than casting
Over 1.3	1.1 maximum
Maximum Mg%	
Less than 0.50	0.05 more than casting
0.5 and greater	0.1 more than casting
Maximum Zn%	
Over 0.25 thru 0.6	0.10 less than casting
Over 0.6	0.1 less than casting

xxx.2 indicates an ingot that has chemical composition limits that differ but fall within the limits of an xxx.1 ingot.

A modification of the original alloy or impurity limits is indicated by a serial letter such as A, B, C, etc. before the numerical designation, like A380.0. X is assigned for experimental alloys.

III Specifications of Aluminum Die Casting Alloys

In the U. S. the Aluminum Association has registered 239 designations of Al casting alloys and ingot alloys^[11]. Of this total 23 are for die castings. Other societies, government organizations, and some individual users and producers of Al alloys have also issued their own designations of Al casting alloys. Some of these alloy specifications were made public and some are still proprietary. Since 1953 the Aluminum Digest has issued 31 alloys for Al die castings and ingots, 14 out of which have different designations and compositions from those of the standardized ones. The Handbook of International Alloy Compositions and Designations, Vol. III, 1980, listed 51 Al die casting alloys in the U. S., among which six are distinct from those standardized ones. Internationally, though the ISO has published standards of Al casting alloys for international use, major industrial countries still remain their own designations and specifications. In Europe with the advent of the Single European Market, uniform standards for Al ingots and castings will apply throughout the market^[12], but currently existing designations and specifications issued by individual countries are still in use. Therefore, today numerous designations of Al casting alloys exist in the world. The Handbook of International Alloy Compositions and Designations, Vol. III, 1980, listed incompletely 327 Al die casting alloys worldwide (see Table 3-1)^[12]. In addition, new alloys have been patented, while some alloys are obsolete or have been replaced by other alloys. It is difficult to tell exactly how many Al casting alloys and Al die casting alloys exist and are used in the world or even in the U. S.

However, despite the fact that some alloys have different designations given by different institutions or countries, they actually have the same or similar specifications (composition). Moreover, the most commonly used alloys in the production are actually limited. These alloys are generally listed in most of the references. Tables 3-2, through 3-6 list some specifications of Al die casting alloys collected from different sources. Table 3-2 contains all the Al die casting alloys registered by the American Aluminum Association. Table 3-3 lists the alloys issued in Metal Digest, which are distinct from those AA registered ones. Table 3-4 shows the U. S. alloys, which are listed in The Handbook of International Alloy Compositions and Designations, Vol. III,

1980, but are not identical to the standardized ones. Tables 3-5 (a) to (f) list some alloys used in other countries as retrieved from The Handbook of International Alloy Compositions and Designations, Vol. III , 1980. Table 3-6 contains the Al die casting alloys, which are found in ASM Engineering Handbook^[13] and Smithells Metals Reference Book^[14], but not included in above tables. By comparing the compositions of all these Al die casting alloys, one can find that the alloys registered by the Aluminum Association cover almost all other alloys. For a given alloy with a designation from another system, one can almost always find an equivalent (or similar) one in the American Aluminum Association register of alloy. In this review, when an alloy is mentioned, it generally refers to the American Aluminum Association registered alloy.

Characteristics of Aluminum Die Casting alloys

Requirements for Aluminum die casting alloys

Developing or selecting an alloy depends mainly on the alloy's performing characteristics (mostly mechanical properties), process properties, economics, and effect on the environment. In addition to these general requirements, there is a special requirement for Al die casting alloys and there are some requirements of process properties that Al die casting alloys should follow more strictly than other Al casting alloys. These include the following:

Good resistance to die soldering. Because the die materials currently used are all ferrous alloys, the diffusion at pouring and solidification temperatures may lead to sticking of the casting to the die.

Good fluidity. In die casting the die cavity is filled under pressure, that should ease filling. However, in the die casting the metallic die has high cooling capacity, the die castings generally have complex shape and thin cross section, and prefer low pouring temperature to alleviate the hot attack to the expensive die. These factors make the filling problems in die casting more serious than in other casting processes.

Resistance to hot tearing. The complex configuration, thin wall and high stiffness of the die make the hot tearing a much more critical problem in die casting than in other casting processes.

Unique features of Aluminum die casting alloys

High Fe content. By looking at all the Al die casting alloys listed in Tables 3-2 through 3-5, one can notice that all these alloys contain 1 to 2% Fe, which is much higher than in other casting alloys. For example, in sand casting alloys Fe contents are generally less than 0.5%. The higher Fe contents in Al die casting alloys are to alleviate the soldering effect. However, Fe is generally considered to be detrimental to the mechanical properties. In permanent mold casting, ferrous dies are also generally used, therefore soldering may also be a problem. But because the metal is filled into the die under gravity but not under high pressure, resulting in much less friction and impingement to the die surfaces than in the die casting, the die is allowed to have a thicker and therefore stabler coating. The coating can significantly reduce the soldering tendency. Also the dies used in permanent mold casting are generally less expensive than those used in die casting. So, it is not necessary to specify high Fe contents in permanent mold Al casting alloys.

No Al-Cu system and heat treatable die casting alloys. It can also be observed that none of the Al die casting alloys belong to the 2xx.x group and some of the alloys containing Mg in group 3xx.x, such as 356 and 357, are not used as die casting alloys. Alloys in the 2xx.x group are Al-Cu alloys. These alloys and the Al-Si-Mg alloys, 356-T6 and 357-T6, have the best mechanical properties of all the Al casting alloys. However, these superior properties are obtained after their heat treatment. So far, die castings are generally considered as not heat treatable because of blister problems caused by pressured air entrained in the casting. In addition, the 2xx.x alloys usually have poor hot tearing resistance. Mg is considered to be deleterious in Al die casting alloys when Si and Fe contents are high, and is usually limited at a low level ($<0.1\%$ in the U. S.).

No 1xx.x Al die casting alloys. In the U. S. no alloy in the 1xx.x group is specified as a die casting alloy. Actually, alloys in this group have very good conductivity and elongation and have been used, especially in the electrical industry. For example, a slot bar and a ring in a cage-type rotor of an induction motor are usually die cast with pure Al [15]. However, because of their high melting temperature, low tensile strength, hardness, poor castability and poor machinability, these alloys are not widely used in die casting. Some alloys belonging to this group are specified as die casting alloys in

other countries (see Table 3-6). It seems that standardization of some alloys in this group as die casting alloys is useful and necessary.

Groups of Aluminum die casting alloys

All the Al die casting alloys registered by the American Aluminum Association are in groups of 3xx.x, 4xx.x and 5xx.x. According to their compositions, the alloys in a group can also be divided into sub-groups. In the same sub-group the alloys have similar compositions and properties, differing from those in other groups and sub-groups. 360.0, 361.0, 364.0, and 369.0 can be considered a sub-group, having 360.0 as a representative. They are Al-Si-Mg alloys. In these alloys Si content ranges from 7.5 to 12.0%, which is below the binary Al-Si eutectic content. Mg content is between 0.2-0.6%. 380.0, 383.0, 384.0 and 385.0 are Al-Si-Cu alloys and can be grouped together as a 380.0 sub-group. In these alloys Si and Cu contents range from 7.5 to 13% and 2 to 4.5%, respectively. 385.5 has a eutectic Si content and the rest of them have a hypoeutectic Si content. 390.0, 392.0 and 393.0 are also Al-Si-Cu alloys, but they have Si contents in the hypereutectic Al-Si range. Si and Cu contents in this sub-group are 16 to 23% and 0.4 to 5.0%, respectively. 390.0 is generally referred to as the representative of this sub-group. It seems that in each sub-group the Si content increases with an increase in the series.

In the Al-Si alloys, group 4xx.x, 413.0 is a eutectic and C443.0 is a hypereutectic alloy. 343.0 has a composition similar to C443.0 in terms of Si content, but it is placed in the 3xx.x group. This is probably because it contains relatively large amounts of Zn and Sn. Alloys of group 5xx.x include 515.0, 516.0 and 518.0. They are Al-Mg alloys, containing Mg 2.5-4.0%, 2.5-4.5% and 7.5-8.5%, respectively.

Table 3-1 Number of Al Die Casting Alloys Worldwide

(Listed in the Handbook of International Alloy Compositions and Designations, Vol. III, 1980)

Country	W Germany	U. S. A.	Italy	Russia	Belgium	Bulgaria	Spain	Norway	Poland
Number	58 (38)*	51(23+6)**	34	24	16	16	15	14	13

Country	S Africa	Austria	Denmark	France	Czech-	E Germany	Japan	Yugoslavia	Sweden
Number	12	10	9	9	8	8	7	6	5

Country	Korea	India	Netherlands	Switzerland	G Britain	Canada	Australia	Total
Number	4	3	3	2	***	***	***	327

Only Al die casting alloys are counted. Total number listed is 327. Ingot alloys are not included.

* Total listed is 58, of which 38 are distinct.

** Total listed is 51, of which 29 (23+6) are distinct and 23 are in the "Registration Record of Aluminum Association Alloy Designation.

*** G. Britain, Canada and Australia do not distinguish die casting alloys.

Table 3-2 Compositions of Registered Aluminum Die Casting Alloys Used in Cast Shapes
(Listed in the ASM Metal Handbook, Vol. 15, 9th Edition, 1988)

Alloy	Composition %												
	Si	Fe	Cu	Mn	Mg	Cr	Ni	Zn	Sn	Ti	Others Except Al		Al
											Each	Total	
343.0	6.7-7.7	1.2	0.50-0.9	0.5	0.1	0.1	...	1.2-2.0	0.5	...	0.1	0.35	Remainder
360.0	9.0-10.0	2.0	0.6	0.35	0.40-0.6	...	0.50	0.50	0.15	...	...	0.25	Remainder
A360.0	9.0-10.0	1.3	0.6	0.35	0.40-0.6	...	0.50	0.50	0.15	...	...	0.25	Remainder
361.0	9.5-10.5	1.1	0.50	0.25	0.40-0.6	0.20-0.30	0.20-0.30	0.50	0.1	0.2	0.05	0.15	Remainder
364.0	7.5-7.9	1.5	0.20	0.1	0.20-0.40	0.25-0.50	0.15	0.15	0.15	...	0.05*	0.15	Remainder
369.0	11.0-12.0	1.3	0.5	0.35	0.25-0.45	0.30-0.40	0.05	1.0	0.1	...	0.05	0.15	Remainder
380.0	7.5-9.5	2.0	3.0-4.0	0.50	0.10	...	0.50	3.0	0.35	...	...	0.50	Remainder
A380.0	7.5-9.5	1.3	3.0-4.0	0.50	0.10	...	0.50	3.0	0.35	...	...	0.50	Remainder
B380.0	7.5-9.5	1.3	3.0-4.0	0.50	0.10	...	0.50	1.0	0.35	...	...	0.50	Remainder
383.0	9.5-11.5	1.3	2.0-3.0	0.50	0.10	...	0.30	3.0	0.15	...	...	0.50	Remainder
384.0	10.5-12.0	1.3	3.0-4.5	0.50	0.10	...	0.50	3.0	0.35	...	...	0.50	Remainder
A384.0	10.5-12.0	1.3	3.0-4.5	0.50	0.10	...	0.50	1.0	0.35	...	...	0.50	Remainder
385.0	11.0-13.0	2.0	2.0-4.0	0.50	0.30	...	0.50	3.0	0.3	...	...	0.50	Remainder
390.0	16.0-18.0	1.3	4.0-5.0	0.10	0.45-0.65	...	...	0.10	...	0.20	0.10	0.20	Remainder
B390.0	16.0-18.0	1.3	4.0-5.0	0.50	0.45-0.65	...	0.10	1.5	...	0.20	0.10	0.20	Remainder
392.0	18.0-20.0	1.5	0.40-0.80	0.20-0.60	0.80-1.20	...	0.50	0.50	0.30	0.20	0.15	0.50	Remainder
393.0	21.0-23.0	1.3	0.7-1.1	0.1	0.7-1.3	...	2.0-2.5	0.10	...	0.10-0.20	0.05**	0.15	Remainder
413.0	11.0-13.0	2.0	1.0	0.35	0.1	...	0.50	0.50	0.15	...	...	0.25	Remainder
A413.0	11.0-13.0	1.3	1.0	0.35	0.1	...	0.50	0.50	0.15	...	...	0.25	Remainder
C443.0	4.5-6.0	2.0	0.6	0.35	0.1	...	0.50	0.50	0.15	...	...	0.25	Remainder
515.0	0.50-1.0	1.3	0.2	0.40-0.6	2.5-4.0	...	...	0.10		...	0.05	0.15	Remainder
516.0	0.30-1.5	0.35-1.0	0.3	0.15-0.40	2.5-4.5	...	0.25-0.04	0.20	0.1	0.10-0.20	0.05***	...	Remainder
518.0	0.35	1.8	0.25	0.35	7.5-8.5	...	0.15	0.15	0.15	...	...	0.25	Remainder

Note: The alloys written in shadow are those, that are also listed in ASTM B85.

* 0.02-0.04 Be. ** 0.08-0.15 V. *** 0.10 max Pb.

Table 3-3 Specifications of Aluminum Die Casting Alloys in the U. S.
(Published in Metal Digest, and distinct from the AA registered alloys)

Designation	Time	AA Reference Number	Chemical Composition (%)														Others	
			Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Ni	Pb	Sn	Others	Each	Total		
Al 218	10, 1953	518	0.3	1.8	0.20	0.30	7.5-8.5											
Apex 418	8, 1957		0.35	1.3	0.20	0.35	7.5-8.5		0.15		0.15				0.10			
Alsiloy No.1	9, 1957	A413.0	11.0-13.0	1.3	0.60	0.30			0.5		0.50							
Apex 36	61,958	A360.0	9.0-10.0	1.0	0.60	0.30	0.4-0.6		0.5		0.50				0.20			
Apex Z39	10, 1958		8.5-9.5	1.0	3.0-4.0	0.50	0.10		2.0-3.0		0.50				0.30			
Al 13	11, 1958	413	11.0-13.0	2.0	0.60	0.35	0.10		0.5		0.50		0.15			0.25		
Apex 39	1, 1959		8.5-9.5	1.0	3.0-4.0	0.50	0.10		1.0		0.50				0.30			
Birmal P.83	12, 1960	B380.0	7.5-9.5	1.3	3.0-4.0	0.50	0.10		1.0		0.50		0.30			0.50		
Apex 39-11	2, 1961		10.5-11.5	1.0	3.0-4.5	0.50	0.10		1.0		0.50				0.30			
Apex Z39-11	2, 1963		10.5-11.5	1.0	3.0-4.0	0.50	0.10		2.0-3.0		0.50				0.30			
Hi-Flo	12, 1963		9.0-10.5		2.0-3.0				2.0-3.0									
Apex 28	5, 1964		9.5-10.5	1.0	1.7-2.3	0.50	0.10		1.0		0.50				0.20			
Alcan 164	8, 1964		11.5-13.0	0.6-1.2	3.0-4.0	0.10									0.05	0.15		
Alcan 143	11, 1965		7.5-9.5	0.6	3.0-4.0	0.1	0.05		0.1						0.1	0.15		
Al A13	8, 1967	A413.0	11.0-13.0	1.3	0.6	0.3	0.1		0.5		0.5		0.1			0.2		

Table 3-4 Specifications of Aluminum Die Casting Alloys in the U. S.

(Listed in Handbook of International Alloy Compositions and Designations, Vol. III, Aluminum, 1980,
and distinct from the AA registered alloys)

Designation	Type	AA Reference Number	Chemical Composition (%)													
			Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Ni	Pb	Sn	Others	Others	
															Each	Total
	AlSi5.2		4.5-6.0	2.0	0.6	0.35	0.10		0.50		0.50		0.15			0.25
	AlSi5Cu		4.5-5.5	1.3	3.0-4.0	0.50	0.10		1.0		0.50		0.30			0.50
	AlCu4Si	208.0	2.5-3.5	1.2	3.5-4.5	0.50	0.10		1.0	0.25	0.35					0.50
	AlSi12Cu	384	11.0-13.0	1.3	3.0-4.0	0.50	0.10		1.0		0.50		0.35			0.50
B384.0	AlSi12Cu		11.0-13.0	2.0	2.0-4.0	0.50	0.30		3.0		0.50		0.30			0.50
L514.0	AlMg3.2 SiMn		0.50-1.0	1.3	0.20	0.40-0.60	2.5-4.0		0.10						0.05	0.15

Table 3-5 (a) Specifications of Aluminum Die Casting Alloys in Germany -- I

(Listed in Handbook of International Alloy Compositions and Designations, Vol. III, Aluminum, 1980)

Designation	Type	AA Reference Number	Chemical Composition (%)										
			Si	Fe	Cu	Mn	Mg	Zn	Ti	Ni	Pb	Sn	Others
			Others	Each	Total								
22/D	AlSi6.2CuMnMg		5.0-7.5	1.3	3.0-5.0	0.3-0.6	0.1-0.3	2.0	0.15	0.3	0.3	1	
226/2	AlSi8.5CuMn		7.5-7.9	0.8	2.0-3.5	0.2-0.5	0 - 0.3	2.0	0.15	0.3	0.2	0.1	
226/3	AlSi8.5CuMn		7.5-9.5	0.8	2.0-3.5	0.2-0.5	0 - 0.3	3.0	0.15	0.3	0.2	0.1	
226/311	AlSi8.5CuMn		7.5-9.5	1.3	2.0-3.5	0.2-0.5	0 - 0.3	0.7	0.15	0.3	0.2	0.1	
226/D	AlSi8.5CuMn		7.5-9.5	1.3	2.0-3.5	0.2-0.5	0 - 0.3	1.2	0.15	0.3	0.2	0.1	
228	AlSi10CuMn		9.011.0	0.8	3.0-4.0	0.3-0.5	0.3-0.5	1.2	.10-0.2	0.3	0.2	0.1	
230/D	AlSi12.2	413.0	11.0-13.5	1.0	0.10	0 - 0.4	0.05	0.1	0.15				0.05 0.15
231/331	AlSi12.2Mn		11.0-13.5	1.3	0.4	0.2-0.5	0.3	0.5	0.15	0.2	0.2	0.1	
231/D	AlSi12.2Mn		11.0-13.5	1.3	1.0	0.2-0.5	0.3	0.5	0.15	0.2	0.2	0.1	
232	AlSi9		7.0-11.0	1.	1.6	0 - 0.5	0.3	1.5	0.15	0.4	0.2	0.2	0.05 0.15
233/D	AlSi10MgMn		9.0-11.0	1.3	0.3	0.2-0.5	0.20-0.50	0.3	0.15	0.1			0.05 0.15
239/D	AlSi10Mg		9.0-11.0	1.0	0.10	0 - 0.4	0.20-0.50	0.10	0.15				0.05 0.15
311	AlSi8.5CuMn	380.0	7.5-9.5	1.3	2.0-3.5	0.2-0.5	0 - 0.3	1.2	0.15	0.3	0.2	0.2	0.05 0.15
332	AlSi10		8.0-12.0	0.8	0.4	0 - 0.5	0 - 0.5	0.40	0.20	0.3	0.2	0.1	0.05 0.15
333	AlSi8.2Cu		7.0-9.0	1.0	1.0-2.5	0.5	0.3	1.6-2.5	0.2	0.5	0.3	0.1	0.05 0.15
341	AlMg7.7Mn		0 - 1.3	1.2	0.2	0.2-0.5	6.0-9.5	0.40	0.20				0.05 0.15
341/B	AlSi7.7Mn		0 - 1.3	0.8	0.2	0.2-0.5	6.0-9.5	0.40	0.20				0.05 0.15
349	AlMg8.5Mn		0 - 2.5	1.0	0.05	0.2-0.5	7.0-10.0	0.1	0.15				0.05 0.15
349/B	AlMg8.5Mn		0 - 2.5	1.0	0.03	0.2-0.5	7.0-10.0	0.1	0.15				0.05 0.15
380/A	AlSi8.7Cu		7.5-10.0	1.3	2.5-4.0	0.5	0.3	1.2	0.15	0.5	0.2	0.2	0.05 0.15
380/B	AlSi8.7Cu		7.5-10.0	1.3	2.5-4.0	0.5	0.3	3.0	0.15	0.5	0.2	0.2	0.05 0.15

Table 3-5 (a) Specifications of Aluminum Die Casting Alloys in Germany -- II

(Listed in Handbook of International Alloy Compositions and Designations, Vol. III, Aluminum, 1980)

Desig- nation	Type	AA Re- ference Number	Chemical Composition (%)													
			Si	Fe	Cu	Mn	Mg	Zn	Ti	Ni	Pb	Sn	Others	Others		
														Each	Total	
3.1263	AlCu4.5Si		2.0-4.0	1.1	3.0-6.0	0.5	0.5	2.5	0.5	0.5				0.05	0.15	
3.2152	AlSi6.2CuMnMg		5.0-7.5	1.3	3.0-5.0	0.3-0.6	0.1-0.3	2	0.15	0.3	0.3	0.1		0.05	0.15	
3.2162	AlSi8.5CuMn		7.5-9.5	1.3	2.0-3.5	0.2-0.5	0 - 0.3	1.2	0.15	0.3	0.2	0.1		0.05	0.15	
3.2183	AlSi12.2Mn		11.0-13.5	1.3	2	0.2-0.5	0 - 0.3	0.5	0.15	0.2	0.1	0.1		0.05	0.15	
3.2382	AlSi10Mg		9.0-11.0	1.0	0.1	0 - 0.4	0.20-0.50	0.10	0.15					0.05	0.15	
3.2463	AlSi8ZnCu		7.0-9.0	1.0	1.0-2.5	0.5	0.3	1.6-2.5	0.2	0.5	0.3	0.1		0.05	0.15	
	AlSi10		8.0-12.0	1.2	0.4	0 - 0.5	0 - 0.5	0.4	0.2	0.3	0.2	0.1		0.05	0.15	
	AlSi12Mn		11.0-13.0	1.2	0.4	0.2-0.7	0.5	0.4	0.05	0.2				0.05	0.15	
3.2582	AlSi12.2		11.0-13.5	1.0	0.10	0 - 0.4	0.05	0.10	0.15					0.05	0.15	
3.2584	AlSi12.2		11.0-13.5	0.6	0.05	0.5	0.05	0.1	0.15					0.05	0.15	
3.2982	AlSi12.2Mn		11.0-13.5	1.3	1.0	0.2-0.5	0.3	0.5	0.15	0.2	0.2	0.1		0.05	0.15	
3.2983	AlSi10MnMg		9.0-11.0	1.3	0.3	0.2-0.5	0.20-0.50	0.3	0.15	0.1				0.05	0.15	
	AlMg7.7Mn		0 - 1.3	1.2	0.2	0.2-0.5	6.0-9.6	0.40	0.20					0.05	0.15	
	AlMg8.5Mn		0 - 2.5	1.0	0.05	0.2-0.5	7.0-10.0	0.1	0.15					0.05	0.15	
	AlSi6.5CuMn		5.0-8.0	1.5	2.0-4.0	0.2-0.6	0.5	0.70	0.20	0.5	0.3	0.1		0.05	0.15	
GD-Al Si6Cu3	AlSi6.5CuMn		5.0-8.0	1.5	2.0-4.0	0.2-0.6	0.5	0.70	0.20	0.5				0.05	0.15	
GD-Al Si12(Cu)	AlSi12Mn		11.0-13.0	1.2	0.4	0.2-0.7	0.5	0.4	0.05	0.2				0.05	0.15	

Table 3-5 (b) Specifications of Aluminum Die Casting Alloys in Russia

(Listed in Handbook of International Alloy Compositions and Designations, Vol. III, Aluminum, 1980)

Designation	Type	AA Reference Number	Chemical Composition (%)												
			Si	Fe	Cu	Mn	Mg	Zn	Ti	Ni	Pb	Sn	Others	Others	
														Each	Total
AL 2	AlSi11.5		10.0-13.0	1.5	0.6	0.5	0.1	0.3							2.8
AL 3	AlSi5CuMnMg		4.5-5.5	1.5	1.5-3.0	0.6-0.9	0.35-0.6	0.3		0.05	0.01	+ Ce 0.5	Ce+Zr 0.5		2.0
AL 3W	AlSi5CuMnMg		4.0-6.0	1.3	1.5-3.5	0.2-0.8	0.2-0.8								
AL 4	AlSi9.2MnMg		8.0-10.0	0.9	0.3	0.2-0.5	0.17-0.3	0.3	0.15		0.05	0.01			1.4
AL 4W	AlSi9.5MnMg		8.0-11.0	1.3	1.0	0.2-0.5	0.2-0.4	0.5		0.3					3.0
AL 5	AlSi5CuMg		4.5-5.5	1.5	1.0-1.5	0.5	0.35-0.6		0.15			0.01	Be 0.1		
AL 6	AlSi5.2Cu		4.5-6.0	1.5	2.0-3.0	0.3	0.1	0.3							
AL 7-4	AlCu4Si		0.75-2.25	1.0	3.8-4.2	0.50	0.30	0.50							
AL 9	AlSi7Mg		6.0-8.0	1.5	0.2	0.5	0.2-0.4	0.3	0.15	0.3	0.05	0.01	Be 0.1		2.1
AL 9W	AlSi7Mg		6.0-8.0	1.3	1.5	0.6	0.2-0.5	0.5		0.3					4.1
AL10W	AlCu7SiMg		4.5-6.5	1.3	6.0-8.0	0.5	0.2-0.5	0.6		0.5					
AL 11	AlZn9.5SiMg		6.0-8.0	1.5	0.6	0.5	0.1-0.3	7.0-12.0							
AL 13	AlMg5SiMn		0.8-1.3	1.5	0.1	0.1-0.4	4.5-5.5	0.2					Zr 0.15		1.9
AL 22	AlMg11.7SiTiBe		0.8-1.2	1.2			10.5-13.0	0.1	0.05-0.15				Zr 0.2, B2 0.03-0.07		1.4
AL 23	AlMg6.5ZrTiBe		0.2	0.2	0.15	0.1	6.0-7.0	0.1	0.05-0.15				Zr0.05-0.20, Be0.02-0.10		0.6
Al 23-1	AlMg6.5ZrTiBe		0.05	0.05	0.05	0.1	6.0-7.	0.05	0.05-0.15				Zr0.05-0.20, Be0.02-0.10		0.2
AL 27	AlMg10.5ZrTiBe		0.2	0.2	0.15	0.1	9.5-11.5	0.1	0.05-0.15				Zr0.05-0.20, Be0.05-0.15		0.6
AL 28	AlMg5.5MnTi		0.3	0.5	0.3	0.4-1.0	4.8-6.3		0.05-0.15						0.8
AL 29	AlMg7SiMn		0.5-1.0	0.9	0.1	0.25-0.60	6.0-8.0	0.2					Zr 0.01, Be0.01		
Al 32	AlSi8CuMnMgTi		7.5-8.5	0.9	1.0-1.5	0.3-0.5	0.3-0.5	0.3	0.1-0.3						1.2
AMg7L	AlMg7SiMn		0.5-1.0	0.9		0.25-0.6	6.0-8.0						Be 0.01		
AMg9	AlMg9MnBe		0.4	0.4	0.05	0.2-0.5	8.5-9.5						Cr 0.1, Zr 0.1, Be 0.0001-0.005		
AMg10	AlMg9.5						10.0-11.5								
AMg11	AlMg10.7						10.0-11.5								

Table 3-5 (c) Specifications of Aluminum Die Casting Alloys in Japan

(Listed in Handbook of International Alloy Compositions and Designations, Vol. III, Aluminum, 1980)

Designation	Type	AA reference Number	Chemical Composition (%)													
			Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Ni	Pb	Sn	Others	Others	
															Each	Total
ADC 1	AlSi12		11.0-13.0	1.3	0.6	0.3	0.3		0.5		0.5		0.1			
ADC 3	AlSi9.5Mg		9.0-10.0	1.3	0.6	0.3	0.4-0.6		0.5		0.5		0.1			
ADC 5	AlMg7.5		0.3	1.2	0.2	0.3	4.0-11.0		0.1		0.1		0.1			
ADC 6	AlMg3.2Mn		1.0	0.8	0.12	0.4-0.5	2.5-4.0		0.4		0.1		0.1			
ADC 7	AlSi5.2		4.5-6.0	1.3	0.6	0.3	0.3		0.5		0.5		0.1			
ADC 10	AlSi8.5Cu		7.5-9.5	1.3	2.0-4.0	0.5	0.3		1.0		0.5		0.3			
ADC 12	AlSi11.2Cu		10.5-12.0	1.3	1.5-3.5	0.5	0.3		1.0		0.5		0.3			

Table 3-5 (d) Specifications of Aluminum Die Casting Alloys in France

(Listed in Handbook of International Alloy Compositions and Designations, Vol. III, Aluminum, 1980)

Designation	Type	AA Reference Number	Chemical Composition (%)													
			Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Ni	Pb	Sn	Others	Others	
															Each	Total
A 5	Al99.50		+Cu+Fe 0.55	+Cu+Fe 0.55	0.05 +Fe+Si 0.55	0.05	0.05		0.1	0.1	0.1		0.05			
A-G6	AlMg6.7		1.0	1.3	0.2	0.6	5.0-8.5		0.4	0.2	0.1		0.1	(Be 0.05)		
A-G10	AlMg9.7		1.0	1.3	0.2	0.6	8.5-11.0		0.4	0.2	0.1		0.1	Be 0.05		
A-S5Fe	AlSi5.5Fe		4.0-7.0	1.3-2.0	0.6	1.0	0.6	0.2	0.5		0.5		0.1			
A-S5U4Fe	AlSi5.2Cu		4.5-6.0	1.3	3.0-4.0	0.7	0.6				0.3		0.3			
A-S9G	AlSi10Mg		9.0-11.0	1.3	0.3	0.5	0.15-0.9		0.5	0.2	0.5		0.1			
A-S9U3A	AlSi8.7Cu		7.5-10.0	1.3	2.5-4.0	0.5	0.3		1.2	0.2	0.5	0.2	0.2			
A-S9U3B	AlSi8.7Cu		7.5-10.0	1.3	2.5-4.0	0.5	0.3		3.0	0.2	0.5	0.2	0.2			
A-S10U4	AlSi10Cu		9.0-11.0	1.3	3.0-4.0	0.5	0.5		0.8	0.2	0.5		0.1			
A-S12	AlSi12.2		11.0-13.50	1.3	0.6	0.3	0.2		0.5	0.2	0.5		0.1			

Table 3-5 (e) Specifications of Aluminum Die Casting Alloys in Australia

(Listed in Handbook of International Alloy Compositions and Designations, Vol. III, Aluminum, 1980)

Designation	Type	AA Reference Number	Chemical Composition (%)													
			Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Ni	Pb	Sn	Others	Others	
															Each	Total
AP 313	AlSi8.5CuFe	380.2	7.5-9.5	0.7-1.1	3.0-4.0	0.10	0.10		0.10		0.10		0.20			0.20
AP 315	AlSi11.2-CuFe	384.2	10.5-12.0	0.6-1.0	3.0-4.5	0.10	0.10		0.10		0.10		0.10			0.20
AP 401	AlSi12Fe	413.2	11.0-13.0	0.7-1.1	0.10	0.10	0.07		0.10		0.10		0.10			0.20
AP 403	AlSi5.2	443.2	4.5-6.0	0.6	0.10	0.10	0.05		0.10	0.20					0.05	0.15
AP 503	AlMg8	518.2	0.30	0.7	0.10	0.10	7.5-8.5		0.10	0.20	0.05		0.05		0.05	0.15
AP 605	AlSi9.5-FeMg	360.2	9.0-10.0	0.7-1.1	0.10	0.45-0.60			0.10		0.10		0.10			0.20
AS 401	AlSi12	A413.1	11.0-13.0	1.0	0.6	0.50	0.15		0.40	0.20	0.50	0.10	0.15			0.25
BP 313	AlSi8.5Cu3.5	A380.2	7.5-9.5	0.6	3.0-4.0	0.10	0.10		0.10		0.10				0.05	0.15
BP 605	AlSi9.5Mg	A360.2	9.0-10.0	0.6	0.10	0.05	0.45-0.60		0.05							0.15
CA 313		380.0	7.5-9.5	1.3	3.0-4.0	0.50	0.30	0.10	3.0	0.20	0.50	0.35	0.25		0.05	0.20
AA 315		384.1	10.5-12.0	1.3	3.0-4.5	0.50	0.10	0.10	1.0	0.20	0.50	0.25	0.35		0.05	0.20
DA 401		413.1	10.0-13.0	1.0	0.60	0.50	0.25	0.10	0.40	0.20	0.50	0.15	0.05		0.05	0.20
EA 401		413.2	11.0-13.0	0.6	0.15	0.50	0.10	0.10	0.15	0.20	0.10	0.10	0.05		0.05	0.20

Table 3-5 (f) Specifications of Aluminum Die Casting Alloys in Czechoslovakia

(Listed in Handbook of International Alloy Compositions and Designations, Vol. III, Aluminum, 1980)

Designation	Type	AA Reference Number	Chemical Composition (%)													
			Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Ni	Pb	Sn	Others	Others	
															Each	Total
424331	AlSi9.7-MgMn		9.00-10.50	1.50	0.20	0.10-0.40	0.20-0.50		0.30	0.15						1.80
424337	AlSi10-CuMnMg		9.00-11.00	1.50	0.80-1.50	0.35-0.50	0.05-0.40		0.30		0.20	0.05	0.05			
424338	AlSi8-CuMn		7.50-8.50	0.90	3.00-4.00	0.30-0.45	0.10		0.90		0.50	0.20	0.20			0.50
424381	AlSi6Cu		4.5-7.0	1.5	2.0-3.0	0.3	0.1		0.3							
424382	AlSi6-CuMn		5.00-7.00	1.50	2.00-3.00	0.20-0.60	0.60		0.80		0.50	0.20	0.10			
424384	AlSi9.5Mn		8.00-11.00	1.50	1.80	0.30-0.60	0.50		1.00		0.40	0.20	0.20			4.50
425519	AlMg9.7-SiCa		0.40-0.80	1.20	0.05	0.05	9.00-10.50		0.25	0.05				Ca 0.10 -0.15		
424568	AlMg8.5-MnBe		1.20	0.80	0.05 + Zn0.30	0.20-0.60	7.00-10.00		*+Cu 0.30					Be0.002 -0.01		

Table 3-6 Specifications of Aluminum Die Casting Alloys Used Internationally and Outside of the U. S. -- I
(Listed in ASM Engineering Handbook, 1980, but not Included in above Tables)

Specification Number	Designation	Country *	Chemical Composition (%)												
			Si	Cu	Mn	Mg	Ni	Zn	Ti	Sn	Fe	Pb	Each	Total	Al
SABS 992	Al-99.7A	S. Afr.	0.20	0.02				0.06			0.25		0.03		99.7
SABS 989	Al-99.7A	S. Afr.	0.20	0.02				0.06			0.25		0.03		99.5
SABS 992	Al-99.5A	S. Afr.	0.30	0.03				0.07			0.4		0.03		99.5
SABS 989	Al-99.5A	S. Afr.	0.30	0.03				0.07			0.4		0.03		99.5
NFA 57 703	A5-Y4	Fra.		0.05	0.05	0.05	0.05	0.1	0.1	0.05	#				99.45
UNI 5076	GD-AlSi12Cu2Fe	Ita.	11.0-12.5	1.75-2.5	0.5	0.3	0.3	0.8	0.15	0.1	0.7-1.0	0.15			bal.
UNI 5075	GD-AlSi8.5Cu3.5Fe	Ita.	8.0-9.5	3.0-4.0	0.3	0.3	0.3	0.9	0.15	0.1	0.7-1.0	0.15			bal.
NF A 57 703	A-S9 U3-Y4	Fra.	7.5-10.0	2.5-4.0	0.5	0.3	0.5	1.2	0.2	0.2	1.3	0.2			bal.
SABS 992	Al-Si10Cu2FeA	S. Afr.	9.0-11.5	0.7-2.5	0.5	0.3	1	1.20	0.2	0.2	1	0.3		0.15	bal.
SABS 989	Al-Si10Cu2FeA	S. Afr.	9.0-11.5	0.7-2.5	0.5	0.3	1	1.2	0.2	0.2	0.5-0.9	0.3		0.15	bal.
SABS 992	Al-Si8Cu4FeA	S. Afr.	7.5-9.5	3.0-4.0	0.5	0.1	0.5	3	0.2	0.2	1.3	0.3		0.15	bal.
SABS 989	Al-Si8Cu4FeA	S. Afr.	7.5-9.5	3.0-4.0	0.5	0.1	0.5	3	0.2	0.2	0.5-1.2	0.3		0.15	bal.
NS 17 530	NS 17 530-05	Nor.	7.5-10.0	2.0-4.0	0.5	0.3	0.3	1.2	0.2	0.2	1.3	0.3			bal.
NS 17 532	NS 17 532-05	Nor.	7.5-10.0	2.0-4.0	0.5	0.3	0.3	1.2	0.2	0.2	1.3	0.3			bal.
SIS 14 42 52	SIS Al 42 52-10	Swe.	7.5-10.0	2.0-4.0	0.5	0.3	0.3	1.2	0.2	0.2	1.3	0.3			bal.
ISO R 164	Al-Si8Cu3Fe	ISO	7.0-9.5	2.5-4.5	0.6	0.15	0.3	1.2	0.2	0.2	1.3	0.3			bal.
DIN 1725	GD-AlSi8Cu3/3.2162.05	Ger.	7.5-9.5	2.0-3.5	0.2-0.5	0.3	0.3	1.2	0.15	0.1	0.8	0.2	0.05	0.15	bal.
DIN 1725	GD-AlSi10MgCu/3.2983.05	Ger.	9.0-11.0	0.3	0.2-0.5	0.2-0.5	0.1	0.3	0.15		0.6		0.05	0.15	bal.
UNI 5074	GD-AlSi9MgFe	Ita.	9.0-10.0	0.5	0.35	0.4-0.6	0.5	0.4	0.15	0.15	0.7-1.2	0.15			bal.
NS 17 520	NS 17 520-XX	Nor.	9.0-11.0	0.2	0.5	0.2-0.4	0.1	0.3	0.2	0.05	0.5	0.05			bal.

: Cu+Fe+Si=0.55

*: Ita. - Italy, Ger. - Germany, Fra. - France, S. Afr. - South Africa, Nor. - Norway, Swe. - Sweden

Table 3-6 Specifications of Aluminum Die Casting Alloys Used Internationally and Outside of the U. S. -- II
(Listed in ASM Engineering Handbook, 1980, and Smithells Metals Reference Book, 1983,
but not Included in above Tables)

Specification Number	Designation	Country *	Chemical Composition												
			Si	Cu	Mn	Mg	Ni	Zn	Ti	Sn	Fe	Pb	Each	Total	Al
ISO R164	Al-Si5Cu3Fe	ISO	4.0-6.5	2.0-4.5	0.2-0.7	0.15	0.3	0.5	0.2	0.2	1.3	0.3			bal.
ISO R164	Al-Si5MgFe	ISO	3.5-6.0	0.1	0.6	0.4-0.9	0.1	0.1	0.2	0.05	1.3	0.1			bal.
ISO R164	Al-Si12Fe	ISO	11.0-13.5	0.1	0.5	0.1	0.1	0.10	0.15	0.05	1.3	0.1			bal.
ISO R164	Al-Si12CuFe	ISO	11.0-13.5	1.2	0.5	0.3	0.2	0.5	0.2	0.1	1.3	0.1			bal.
UNI 5079	GD-ALSi13Fe	Ita.	11.0-13.0	0.8	0.3	0.3	0.2	0.5	0.15	0.1	0.7-1.0	0.15			bal.
NS 17 512	NS 17 512-05	Nor.	11.0-13.5	0.6	0.5	0.3	0.2	0.5	0.2	0.1	1.3	0.1			bal.
SABS 992	Al-Si12FeA	S. Afr.	10.0-13.0	0.6	0.5	0.1	0.5	0.5	0.15	0.15	1.2				bal.
SIS 14 42 52	SIS Al 42 54-10	Swe.	7.5-10.0	2.0-4.0	0.5	0.3	0.3	***	0.2	0.2	1.3	0.3			bal.
SIS 14 42 47	SIS Al 42 47-10	Swe.	8.0-10.0	0.2	0.5	0.5	0.1	0.3	0.2	0.05	0.8-1.3	0.1			bal.
UNI 5077	GD-ALSi5Fe	Ita.	4.5-6.0	0.4	0.3	0.2	0.3	0.5	0.15	0.1	0.7-1.0	0.1			bal.
ISO R164	Al-Si5Fe	ISO	4.0-6.0	0.1	0.5	0.1	0.1	0.1	0.2	0.1	1.3	0.1			bal.
SABS 992	Al-Mg8FeA	S. Afr.	0.35	0.25	0.35	7.5-8.5	0.15	0.15		0.15	1.8			0.25	bal.
UNI 5080	GD-ALMg7.5Fe	Ita.	0.3	0.05	0.4	7.0-8.0	0.05	0.1	0.2	0.05	0.7-1.0	0.05	**		bal.
LM6		Eng.	12												bal.
LM18		Eng.	4.5-5.5												bal.
LM24		Eng.	8	3.5											bal.
LM30		Eng.	16.0-18.0	4.5		0.6									bal.
LM16		Eng.	4.5-5.5	1		0.5									bal.

*: Ita. - Italy, Ger. - Germany, Fra. - France, S. Afr. - South Africa, Nor. - Norway, Swe. - Sweden, Eng. - England

** : Be = 0.005, *** : Zn = 1.2-3.0

IV Effects of Alloying and Impurity Elements on the Microstructure and Properties of Al Die Casting Alloys

In an alloy an element is considered an alloying element or an impurity depending on the amount added and its function. If an element is intentionally added into an alloy and if it, or with other element(s), offers the alloy desired properties and/or microstructure, it is considered an alloying element. If an element is not intentionally added and it is not expected to affect the alloy's properties and microstructure, it is considered an impurity. For a given alloy the alloying element is always controlled within a certain range. Generally, the impurity is the inevitable element brought into the alloy from charging materials, flux, refractories, tools or environment. Because of the large number of impurities usually existing in commercial alloys, it is almost impossible to eliminate all of them economically and practically. Therefore, the impurity in an alloy is usually specified under an upper limit to minimize its effect. Actually, in most commercial alloys, despite a low content of each impurity, because of their large number they do affect the properties to some extent. Moreover, for some elements it is hard to tell whether they are alloying elements or impurities from their contents or functions, because many elements may have either beneficial or detrimental effects on different properties in different circumstances.

In Al die casting alloys, the main alloying elements are Si, Cu, Mg, and Fe. The common impurities are Mn, Ni, Cr, Zn, Pb, Sn, and Ti. The solid solubilities of these and some other common elements vary considerably (Table 4-1)^[9]. These elements exist in the Al die casting alloys in the form of solid solution, separate phases and/or intermetallic compounds, affecting the alloy properties. Specifying an alloy often results from a balance of the effects of all added and impurity elements and from a trade-off among the best combination of elements for desired properties and other considerations, such as economics. Because all these elements may affect an alloy's microstructure or properties to some extent, their effects have been the focus of research in alloy development since the very beginning of alloy research. However, despite the fact that this study has been conducted for so long, because the effects of an element, especially in combination with other elements, are complicated, there are still a number of unanswered questions. The remainder of this chapter will review the work on the effects of all these elements on the microstructure and properties of Al die

casting alloys. This review will be carried out from a practical point of view, intending to be helpful to the production, but will not be involved in the detailed mechanism.

Table 4-1 Solid Solubility of Elements in Aluminum

Element	Temperature °C (°F)		Maximum solid solubility (wt %)
Cadmium	649	(1200)	0.4
Cobalt	657	(1215)	<0.02
Copper	548	(1018)	5.65
Chromium	661	(1222)	0.77
Germanium	424	(795)	7.2
Iron	655	(1211)	0.05
Lithium	600	(1112)	4.2
Magnesium	450	(842)	17.4
Manganese	658	(1216)	1.82
Nickel	640	(1184)	0.04
Silicon	577	(1071)	1.65
Silver	566	(1051)	55.6
Tin	228	(442)	0.06
Titanium	665	(1229)	1.3
Vanadium	661	(1222)	0.4
Zinc	443	(829)	70.0
Zirconium	660.5	(1221)	0.28

4-1 Effects of Alloying Elements

Silicon

The effect of the addition of Si to Al was investigated by St. Claire-Deville as early as 1856, but Al-Si alloys were not widely used until 1920, when Pocz discovered the possibility of improving their mechanical properties by Na modification^[16]. Since then, Al-Si alloys and their variations have been the most important commercial casting alloys. The main reasons for the wide application of Al casting alloys with Si as the major alloying element are their superior casting characteristics and economics. Si is relatively cheap. It imparts some properties improvements to the alloys. Si is also one of the few elements that may be added to Al without a loss of its weight advantage.

Binary Al-Si alloys are a simple eutectic system and its freezing range is narrow, compared with the Al-Cu system in the commonly used composition range. By increasing the Si content ($> 1.6\%$), the freezing range of the alloys decreases until the Si content reaches the eutectic composition, about 12.5% . In tandem, by increasing the Si content ($> 1.6\%$), viscosity at constant temperature decreases, even past the eutectic composition, in spite of the factor that the liquidus temperatures rises and freezing range widens. This is because of the high latent heat of Si which is 4.5 times greater than that of an equivalent volume of Al^[17]. These two factors contribute to the excellent fluidity of Al-Si alloys, especially with a Si range of 7 to 18% . The addition of Si in Al-Mg alloys can also improve their fluidity.

Al-Si alloys have low shrinkage because there is an expansion during the Si phase transformation from liquid to solid. The change in volume of Al-Si alloys during solidification decreases linearly with increasing Si content, and reaches zero value at $25\% \text{Si}$. The low shrinkage and the narrow freezing range of Al-Si alloys result in their good hot tear resistance, soundness, and good weldability.

The electrolytic potential of Si is -0.26V and the alloy with $1\% \text{Si}$ in solution has a potential of -0.81V , as compared with -0.85V for Al. The high difference in potential between the two phases should lead to very rapid corrosion. However, Si is inert to most corroding environments and the corrosion resistance of the alloys is the same as or better than that of Al in most solutions.

The mechanical properties of binary Al-Si alloys depend less on composition than on the distribution and shape of the Si particles. Small, rounded, evenly distributed particles, either eutectic or primary, result in high ductility with relatively high strength; faceted, acicular Si crystals may produce slightly higher strength, but much lower ductility, impact, and fatigue resistance. Fig. 4-1 shows the change of properties as a function of the Si content. Because the amount and morphology of Si in Al-Si alloys are the major factors determining mechanical properties, Al-Si alloys were considered to be among the least sensitive to casting variables, such as gas content, design of castings, rate of cooling, and feeding. The strength of Al-Si alloys declines very rapidly with increasing temperature. Fatigue resistance is low, especially when primary crystals of Si are present or when Si is in the unmodified form. Artificial aging increases tensile strength but reduces fatigue resistance. Si in Al-Mg alloys reduces ductility and notch toughness without a compensating increase in strength.

Fig. 4-1 Mechanical properties of Al-Si alloys as a function of Si content

Thermal expansion is reduced with increasing Si content appreciably. The decrease is more than linear and the coefficient for the range 300-700K, which is 25×10^{-6} m/m/K for pure Al, becomes 20×10^{-6} at 10%Si and 12×10^{-6} at 40%Si. This is beneficial to some applications. However, because of the difference in expansion coefficients between Si particles and matrix, thermal stresses build up in the two phase alloys, which may reach values well above the yield point and lead to thermal fatigue. Because the Si particle is hard, Al-Si alloys are somewhat difficult to machine.

All the above information was taken from work done a long time ago. Recent work about the effects of Si is devoted mostly to the effects of modification and heat treatment on Si morphology and alloy properties. The morphology of Si crystals and Si-bearing phases will be discussed in section V "Solidification and Microstructure of Aluminum Die Casting Alloys," and section VII "Typical Aluminum Die Casting Alloys." The modification of Si phases will be mentioned in the pertinent sections. For more information about the effect of Si on the mechanical and physical properties of Al-Si alloys, such as radiation, thermodynamics, thermal, and electrical properties, see references [18, 19, 20].

Copper

Cu is added mainly to increase the strength of Al alloys. This is done at low temperature mostly by heat treatment, and at high temperature through the formation of compounds with Fe, Mg, Ni, etc. Cu exists in Al-Cu alloys either by solving in the Al matrix or forming intermetallics, like CuAl_2 and, when there are other elements in the alloy, some other compounds. As the Cu content increases, hardness of the alloys increases continuously, but strength and especially ductility depend on whether the Cu is in solid solution, or as spheroidized and evenly distributed particles or as a continuous network at the grain boundaries. Dissolved Cu produces the highest increase in strength while still retaining substantial ductility; with the network not only does the ductility disappear, but even the measurable strength is reduced because of the extreme brittleness. Fig. 4-2 shows the properties as functions of Cu content. When CuAl_2 is spheroidized and evenly distributed, strength is a function of the mean free path between particles. The modules of elasticity increase proportionally to Cu content. At a temperature near the melting point the modules are approximately one-half of that at room temperature. Damping capacity is substantially reduced with

Fig. 4-2 Mechanical properties of Al-Cu alloys as function of Cu content.
H = work hardened; O = annealed; T₄ = quenched and naturally aged; T₆ = quenched and artificially aged

increasing Cu content. Cu additions increase the strength at high temperature and decrease the creep rate of Al by a factor of 2 to 10. Fatigue strength is raised by Cu in solid solution, but in equilibrium conditions ($<1\% \text{Cu}$) the effect is limited. Fatigue strength of casting alloys with intermixed coarse and fine dendrites is low.

Cu decreases the electrolytic potential, and also the corrosion resistance. Al-Cu alloys, which have the least negative potential, should have the maximum resistance to contact corrosion but, whereas in pure Al and most corrosion-resistant alloys the oxide film produces passivity, in Al-Cu alloys the Cu dispersed in the oxide prevents complete passivation, and severe deterioration can be produced not only when the alloys are in contact with other materials, but also when alloys with different Cu contents or differently cold worked or heat treated are in contact. Cu-bearing alloys tend to pit severely in the annealed condition and when age hardened may be susceptible to intergranular or stress corrosion. The increase of corrosion rate ranges from a doubling or trebling of the weight loss with 3% Cu in Al 99% to a 2000-fold increase by the addition of 0.1% Cu to 99.998% Al. The higher the purity the more pronounced the effect of the first addition. The structure of alloys has a decided influence; Cu in solid solution has less effect than Cu in CuAl_2 , provided that the aging does not progress to the point where the alloys become susceptible to intergranular or stress corrosion. The corrosion resistance of casting alloys with cored dendrites and CuAl_2 particles in the interdendritic areas is particularly poor. The density of the alloys increases and the thermal expansion increases with increasing Cu contents. Electric conductivity is very sensitive to copper in solution, but it is affected little by the Cu-bearing compounds. In an alloy with 5% Cu in solution the conductivity is approximately half that of pure Al. Because of the relatively wide freezing ranges of Al-Cu alloys, fluidity, feedability and hot tear resistance of the alloys with 4-6% Cu are very poor. Castability becomes better when Cu content is increased. However, the specific gravity increases when Cu content is increased. Weldability is relatively poor for most Al-Cu alloys.

Because that Al-Cu alloys have poor corrosion resistance, poor castability and higher specific gravity, there has been a trend for Al-Cu system casting alloys to be replaced by other systems of Al based casting alloys. And because the best improvement of strength by addition of Cu is obtained after heat treatment and die castings are generally not heat treatable, there is no Al-Cu binary alloy being used as die casting alloy. In Al die casting alloys Cu is widely added in other Al alloy systems, mostly in Al-Si alloys,

to improve their hardness, strength, fatigue, creep resistance, and machinability, at the expense of corrosion resistance. It may also increase their high temperature strength slightly. In these alloys, Cu, if not in solution, is present as CuAl_2 in alloys in which Mg is not present or present in amounts too small to tie all the Cu as $\text{Cu}_2\text{Mg}_8\text{Si}_6\text{Al}_5$. The amount of Cu in solution is reduced by Ni or Mn additions in favor of forming the Cu-Ni and Cu-Mn compounds. In Al-Mg alloys, Cu may reduce pitting corrosion by enhancing general corrosion, and, within the content limit, has little or no effect on mechanical properties.

There has been a few recent technical papers devoted to the effect of Cu addition on Al die casting and other casting alloys, though they were not systematic studies. Franek, Alexander [21] studied solidification of the P modified Al-Si-Cu alloy, Al-20Si4Cu. The alloy was modified with 0.1% P and cooled at a rate range of 1-100 °C/min to determine the structure and concentration distribution of Cu in primary Si, alpha-phase solid solution and ternary Al-CuAl₂-Si eutectic mixture. Hardness measurements showed that Cu was contained mainly in a supersaturated solution which broke up at over 200 °C. Cu increased hardness at room and elevated temperatures but not at greater than 200 °C. Gasparyan, L. A. et al[22] investigated the effect of Si and Cu contents on the mechanical and casting properties of some Al-Si-Cu alloys. It was found that approximately the same changes in mechanical properties were produced by the introduction of Si $\leq 4\%$ or Cu $\leq 4\%$, although the casting properties of Al-Cu were inferior to those of Al-Si alloys. On the other hand, the Al-Cu alloys were more easily machined by cutting. The combined introduction of Cu and Si significantly increased tensile strength and hardness but sharply reduced the elongation. Ohuchi, H.[23] studied the effect of Cu and Mg on the thermal shock property of Al-20% Si alloy. The work indicated that the alloy containing such heat treatable alloying elements as Cu and Mg had low resistance to thermal shock. Karpachev, V. M. et al[24] studied the effect of composition on heat resistant (high temperature) strength of hypoeutectic Al-Si-Cu alloys. The experiments indicated that Cu had the stronger positive effect on the high temperature strength of the alloys. Jorstad, J. L.'s work[25] showed that Cu and Mg improved the machinability of Al die casting alloys. The addition of Cu and Mg improved the machined surface finish and decreased the tendency of an alloy to build up on a cutting tool edge.

Magnesium

Alloys 515, 516, and 518, are Al-Mg alloys, in which Mg is added as an alloying elements. Other than in these alloys Mg is considered to be an impurity element for most of Al die casting alloys.

Al-Mg binary alloys have high strength with good ductility, excellent corrosion resistance, weldability, and surface finish. Fig. 4-3 shows hardness and strength as a function of Mg content in Al-Mg alloys. Hardness, strength, and fatigue resistance increase, and ductility decreases with increasing Mg content. At the Al end of the Al-Mg binary phase diagram there is a eutectic, Liq. $\rightarrow$ Al + Mg_5Al_8 at 35% Mg, 723 K. The solid solubility of Mg in Al is 17.4 w% and 1.9 w% at temperatures of 723K and 300 K, respectively. This results in a good heat treatable response of the alloys with Mg content above 5-6%. The data shown in Fig. 4-3 are all after heat treated or cold worked. Mg is one of the few elements that lower the elasticity modules of Al. The decrease in strength with increasing temperature is less rapid than in other Al alloys. Mg tends to increase slightly the thermal expansion of Al. Electric resistance increases almost linearly with the Mg in solution. Mg additions lower the potential of Al. The specific volume of the oxide film is larger than that of the metal from which it forms, and this produces a very impervious film, especially when it absorbs water and transforms to hydroxide. The alloys have a better corrosion resistance to salt water and mild alkali than pure Al.

The castability of the alloys is poor at low Mg content (2-4%) and somewhat better at higher Mg (7-12%). The amount of eutectic present is relatively low even in the high-Mg alloys, and there is appreciable hot tear tendency, especially at low Mg contents. The strong tendency of the Mg to react with refractories and to oxides may introduce into the melt more or less finely dispersed inclusions that reduce the fluidity and have an adverse effect on the properties. The mechanical properties of castings are sensitive to grain size. In spite of a relatively wide freezing range in these alloys, segregation of Mg is not too pronounced. There is appreciable coring, so that the eutectic phase, Mg_5Al_8 , may appear in alloys with as little as 3-4% Mg.

Al-Mg die casting alloys, like 515, 516, 518, are mainly used for the castings, which require good surface appearance, corrosion resistance, and moderate strength with good ductility. These alloys have good as-cast surface finish and high reflectivity when

Fig. 4-3 Mechanical properties of Al-Mg alloys as function of Mg content.
 H_V = Vicker hardness; UTS = ultimate tensile strength; YS =
 yield strength; H = cold worked; O = annealed

electropolished and anodized. Besides the addition of Mg as the main alloying element, these alloys also contain Fe and small amounts of Si, Cu, Mn, Zn, Ni and Sn. Fe contents are 1.3%, 0.35-1.0% and 1.8% for 515, 516, and 518, respectively. In these alloys containing Mg and Fe as main elements, there is a ternary eutectic, liq. $\rightarrow$ Al + FeAl₃ + Mg₅Al₈, most probably at 0.15%Fe, 33%Mg, 723K. The eutectic is completely divorced. The solid solubility of Fe in Al is probably reduced by Mg. The solid solubility of Mg in Al is also reduced by the Fe addition. In nonequilibrium conditions, the tendency for the divorcing of the eutectic is strongly increased, so that large, massive crystals of FeAl₃ can form, even at low Fe concentration. In addition, the Mg₅Al₈ compound tends to appear, even at Mg contents of 2-3%. Fe additions have some grain refining effect on casting Al-Mg alloys. The addition of Fe to Al-Mg alloys tends to reduce corrosion resistance, but has little or no effect on stress corrosion or exfoliation.

When both Mg and Si are added to Al, Mg and Si form Mg₂Si, which give the alloy very strong heat treatment response. When heat treated the tensile strength, yield strength, and elongation are all increased significantly. In sand casting Al-Si-Mg alloys, the Si contents are always over 5% and Mg contents range from 0.5 to 1.5%, like 356, 357. In these alloys, the Si contents are much more than the proper ratio for forming Mg₂Si (Mg/Si = 1.73). The addition of Si imparts to the alloys good castability, while Mg greatly increases the response of Al-Si alloys to heat treatment by forming the Mg₂Si constituents. The addition of Mg to Al sand casting alloys with 5-11% Si increases strength and hardness and decreases ductility, as shown in Fig. 4-4. In addition to the response to heat treatment, the presence of Mg may increase strength and creep resistance of Al-Si alloys.

In Al die casting alloys containing Mg, a high level (>0.4%) Fe leads to the formation of Al₅FeSi and Al₈FeMg₃Si₆, which can not dissolve during T4^[9]. In Al-Si alloys, Mg, if not in solution, is present as Mg₂Si unless when Cu is also present and Cu₂Mg₈Si₆Al₅ is formed. In casting conditions, Mg₂Al can be found together with CuAl₂ and Si. In the absence of Cu, Co, Cr, Mn, Mo, or Ni, high Fe and Mg result in the appearance of FeMg₃Si₆Al₈. In some important Al die casting alloys, like 380, 413, C443 etc, Mg is considered an impurity and specified at a low level, less than 0.1%. The reasons for limiting Mg content could be as follows. Mg can easily oxidize to form MgO microsize particles within the melt. At high holding temperatures

Fig. 4-4 Effect of Mg on tensile properties of Al-Si alloys in the solution heat treated condition; determined on standard test bars cast in green sand (W. E. Sich, "Physical Metallurgy of Aluminum Alloys," American Society for Metals, 1949, p143)

(> 745 °C), spinel, which is a complex Al-Mg oxide, usually forms and grows rapidly forming inclusions in the melt^[26]. As mentioned above, the strong tendency of the Mg to react with refractories and to oxides may also introduce into the melt more or less finely dispersed inclusions. When proper amounts of Mg are added to heat treatable Al-Si alloys, the finely dispersed Mg₂Al precipitates or G - P zone, after solution heat treated and aged, strongly strengthen the alloys. However, coarse as-cast MgAl compounds have little strengthening effect. At higher levels, (e. g. >0.3%) of Mg, Mg₂Al may form^[9]. In commercial Al die casting alloys, Mg may also form complex intermetallics with other elements, like Al₈FeMg₃Si₆, Cu₂Mg₈Si₆Al₅. The formed inclusions and intermetallics reduce the fluidity and have an adverse effect on the properties.

Recently the low limit of Mg content (max 0.1%) specified in these Al die casting alloys has been questioned. First, no sound technical basis for specifying this limitation has been found. Though the effects of Mg on Al alloys have been studied extensively, the investigations have been conducted mostly on relatively simple systems, binary or ternary alloys. Most commercial Al die casting alloys are complex alloy systems containing several alloying and impurity elements. Probably because the large number of elements encountered in these alloys, their low, varying concentrations, and the possibility of interactions between them, the systematic study of the effects of elements, such as Mg, on properties of commercial alloys can be very complicated and difficult. This kind of study has not been found during this literature review. Second, the limitations set on similar alloys in other countries are higher (0.3% or more) than in the U. S. For example, the British alloy LM26 (0.5-1.5% Mg), German alloy 226/D (0.3% Mg), French alloy A-S9U3A (0.3% Mg), Italian alloy 5075 (0.3% Mg), and Japanese alloys ADC10, ADC12 (0.3% Mg) (as shown in Table 7-x), are all equal or equivalent to the U. S. alloy 380, but have higher Mg contents. Since allowing higher Mg content, for example in the 380 alloy, can substantially reduce the cost of the "de-mag" operation, diecasters want to confirm whether it is feasible. Thus, a systematic investigation of the effects of Mg, and the same to other elements, on properties of Al die casting alloys needs to be conducted and may result in economic significance.

Though this study has not been systematically conducted for Al die casting alloys, some relevant work has been carried out for die casting and other Al casting alloys. Jorstad,

J. L.^[27] studied the effect of minor elements and impurities on the machinability of the 380 alloy. It was found that a small addition of Mg (0.3%) to a primary 380 die casting alloy improved machinability. Mg hardened the alloy matrix and thereby reduced the build up on a tool edge, resulted in shorter and tighter chips, reduced the friction coefficient between tool and workpiece, and provided a better surface finish. This addition gave the primary 380 alloy desirable machining characteristics similar to those of a secondary alloy. However, tool wear rate for the Mg-modified primary alloy was significantly lower than those for the secondary alloy. The effect of Mg on machinability of the 380 alloy was stronger than that of other minor/impurity elements. Kumagai, S. et al^[28] studied the effects of impurities on the mechanical properties and machinability of alloy ADC12, and found that the addition of Mg up to 0.5% resulted in an improvement in mechanical properties. The addition of Mg, Mn, Sn, Pb, and Bi improved machinability. DasGupta, R. et al^[29, 30] studied the effect of Mg additions on the microstructure and properties of unmodified (for Mg ranging from 0.07 to 0.59%) and Sr-modified (for Mg ranging from 0.08 to 0.63%) 319 alloy. 319 is a sand and permanent mold casting alloy. In this alloy the same question also exists. The 319 alloy has a composition similar to 380 except for Si content (5.5-6.5% Si for 319 and 7.5-9.5% Si for 380) and also has a Mg content limitation of 0.1%. They found that increasing Mg did not contribute to a significant change in microstructure of either the as-cast or the heat treated (T5) structure. In unmodified samples Si was angular plate like, while in Sr-modified condition it was fibrous. In T6 condition the Sr-modified samples had a globular eutectic structure regardless of Mg content. The mechanical properties of both unmodified and Sr-modified specimens under either the as-cast or the heat treated conditions were not significantly influenced by an addition in Mg content. Joenoes, A. T. et al^[31], working on the effect of Mg addition on the microstructure of both unmodified and modified Al-Si alloys, showed that Mg decreased the homogeneity of the resulting microstructure. At 1% it slightly refined the Si phase, but it had a negative effect on Sr modification. Karpachev, V. M. et al^[24] studied the effect of Mg, Mn and Cr on heat resistance properties of a Russian Al-Si-Cu alloy AK5M4 (5.0%Si, 4.0%Cu, 0.8%Fe, 0.5%Ni, 1.0%Zn and Al). The variation of Mg, Mn, and Cr tested were 0.2-2.0%, 0.2-1.6% and 0-0.6%, respectively. Creep rupture time and rate were measured at 300 °C under stress of 45 MPa. The result indicated a complex relationship between heat resistance and composition, which was related with the change in microstructure. The highest levels of creep rupture (>40 h) were achieved with AK5M4 containing 1.13-1.74%Mg, 0.34-0.86%Mn and 0-0.3%Cr. As mentioned above, experiments^[23] showed that Mg

and Cu lowered thermal shock resistance of Al-20%Si alloy. It was also reported^[32] that the addition of Mg allowed a reduction in the stabilization annealing temperature.

It seems that the results of the studies on the effect of Mg on the microstructure and properties of Al die casting alloys and similar alloys are contradictory. This is probably because the experiments conducted in different research facilities were under different conditions and the alloy compositions were also different. The claims about the effects of Mg on the microstructure and properties of Al die casting alloys, which resulted in the low content limitation of Mg for some alloys, do not have sound supporting evidence in the published literature. If there is any such evidence, the experiments may have been conducted under conditions and alloy compositions different from those in die castings, and therefore they would have to be considered only tentative. The understanding of the effects of Mg on Al die casting alloys are still not well established.

Iron

Fe is always present in Al. It is often added unintentionally through the use of steel tools for melting and casting and through the mixing of Fe or rust to remelting material. For most Al alloys, Fe is the main impurity and efforts are made to keep it as low as economically possible. However, in Al die casting alloys Fe is added intentionally mainly to minimize die soldering.

In a binary Al-Fe system, the equilibrium solid solubility is of the order of 0.03 to 0.05% Fe at eutectic temperature (928 K) to a value of the order of 0.00x% at 700K. The phase in equilibrium with Al is usually designated as FeAl₃ (40.7%Fe). With increasing Fe content the shrinkage in solidification decreases linearly to approximately 3% for a 5%Fe alloy. The viscosity of molten alloys increases, but there is no appreciable change in surface tension. The thermal conductivity is reduced. The resistivity increases to approximately 2.75×10^{-8} ohm. m at 0.05%Fe and 2.9×10^{-8} ohm. m at 1%Fe. The electrolytic potential difference between the FeAl₃ (-0.39 to -0.58V) and the matrix Al (-0.80V) reduces corrosion resistance.

In an Al-Fe-Si system, there are two ternary phases that can be in equilibrium with Al: Fe₂SiAl₈ (alpha) and FeSiAl₅ (beta). Another phase, FeSi₂Al₄ (delta), is often present in high-Si alloys, and a fourth phase, FeSiAl₃ (Gamma), forms in high-Fe and high-Si

alloys. More ternary phases form at higher Fe and Si contents. The invariant reactions in the Al corner are shown in Table 4-2.

In non-equilibrium conditions with small amounts of Mn, Cr, and Cu present, another phase with a composition of 27-35% Fe and 6-8% Si may form. This phase is an extension of the MnSiAl or CrSiAl phases in which most of the Mn or Cr is replaced by Fe. For most commercial casting alloys, because of non-equilibrium solidification, it is not uncommon to find alloys in which FeAl_6 , FeAl_3 , Fe_2SiAl_8 , FeSiAl_5 , and FeSi_2Al_4 may coexist with one another and with Si. Basically, it can be expected that FeAl_3 , Fe_2SiAl_8 , and FeSiAl_5 can appear in all the alloys in which they are primary. In Al-Si alloys when Cu is present, the Fe is usually in the Al- FeSiAl_5 -Si eutectic as thin platelets interspread with the Si needles or rods. If there is more than 0.8% Fe, primary FeSiAl_5 crystals appear. Fast cooling tends to shift the eutectic toward higher Fe contents and to disperse the FeSiAl_5 crystals. Occasionally, especially in alloys with low Si content or containing Mg, Cu, or Zn, FeSiAl_5 is formed as Chinese script; when Cu is present, Fe may be associated with it, probably as Cu_2FeAl_7 . In the presence of Mg, the compound $\text{FeMg}_3\text{Si}_6\text{Al}_8$ is found, also with a Chinese script appearance when eutectic and as globular crystals when primary. Fe forms $(\text{FeMn})_3\text{Si}_2\text{Al}_{15}$ with Mn, often in the shape of Chinese script, thus removing the embrittling effect of FeSiAl_5 . The constituents formed by Ni, together with Cu and Fe, are not known with certainty.

The FeSiAl_5 may be found in the Chinese script shape characteristic of Fe_2SiAl_8 or in the platelet form characteristic of FeSi_2Al_4 . Thus, identification of the phases from the shape alone may be misleading. The morphology and size of Fe-bearing phases in Al casting alloys depend on the alloy composition and the casting condition. The most commonly found morphologies are Chinese script and needle (platelet). Isolated block phase and petal-like phase may also exist. The size of the phases is generally reduced with increasing cooling rate. In recent publications the Chinese script phase is generally referred to as $\text{Al}_{15}(\text{Fe,Mn})_3\text{Si}_2$, and the needle phase as Al_5FeSi ^[33]. The Chinese script phase may form at low cooling rate, for example $< 0.2\text{ }^\circ\text{C/s}$, and at high cooling rate, $> 6\text{ }^\circ\text{C/s}$ for 380 alloy^[33]. At low cooling rate the Chinese script phase forms first from liquid before the alpha-Al dendrites form, and it results in big particles. At high cooling rate it forms along with the formation of eutectic Si particles in the interdendritic region. When there is Mn in the alloy, it can alter the

Table 4-2 Invariant Reactions at the Aluminum End of the Al-Fe-Si System

Reaction	A	B	C	Temp. (°K)	Composition							
					Liquid		A		B		C	
					%Fe	%Si	%Fe	%Si	%Fe	%Si	%Fe	%Si
liq. $\rightarrow$ Al + FeAl ₃				928	1.9	—	0.05	—	36			
liq. $\rightarrow$ Al + Si				850	—	12.5	—	1.65	—	99.9		
liq. $\rightarrow$ Al + Si + FeSiAl ₃				849	0.7	12.0	0.01	1.60	0.0x	99.8	~25	~15
liq. + FeAl ₃ $\rightarrow$ Fe ₂ SiAl ₈ + FeSiAl ₃				983	7.5	12.5	36	0.2	33.0	7.0	34	17.0
liq. + FeSiAl ₃ $\rightarrow$ Fe ₂ SiAl ₈ + FeSiAl ₃				948	6.0	13.0	34	17	33.0	7.0	25.0	15.0
liq. + FeSiAl ₃ + FeSiAl ₄ $\rightarrow$ FeSiAl ₃				973	7	14	34	17	25.5	25.5	25.0	13.0
liq. + FeAl ₃ $\rightarrow$ Al + Fe ₂ SiAl ₈				903	2.5	4.0	36	0.1	0.05	0.6	33	7.0
liq. + Fe ₂ SiAl ₈ $\rightarrow$ Al + FeSiAl ₃				885	1.7	6.5	33.0	7.0	0.04	1.1	25	13
liq. + FeSi ₂ Al ₄ $\rightarrow$ FeSiAl ₃ + Si				869	1.5	14	25.5	25.5	25.0	13	< 0.1	99.8+

$\text{Al}_{15}(\text{Fe},\text{Mn})_3\text{Si}_2$ phase forming field, e. g., the composition range, at which the $\text{Al}_{15}(\text{Fe},\text{Mn})_3\text{Si}_2$ phase forms, as shown in Fig. 5-3[33]. Because the needle phase is brittle and its shape is detrimental to tensile properties, it is preferable to have less needle phase or to change the needle phase to Chinese script if they can not be eliminated. For practical purposes, Mn is used as a Fe corrector. However, if the total Mn plus Fe content is above 0.8%, the $(\text{FeMn})_3\text{Si}_2\text{Al}_{15}$ crystals are primary and they appear as hexagonal globules (although the compound is cubic). These globules do not embrittle the alloy, but reduce appreciably the machinability; therefore, the Mn content is often controlled for only partial correction of Fe to prevent the formation of primary crystals. Cr and Ni are also used as Fe correctors, but the $(\text{CrFe})_4\text{Si}_4\text{Al}_{13}$ or $(\text{CrFe})_5\text{Si}_8\text{Al}_2$ compounds and FeNiAl_9 are more elongated and produce some brittleness. The best Fe corrector is Co, because it does not combine with Si and thus the amount of extraneous crystals formed is more limited. Mo has the same effect as Co.

In an Al-Fe-Mg ternary system there is a ternary eutectic, $\text{Liq.} \rightarrow \text{Al} + \text{FeAl}_3 + \text{Mg}_5\text{Al}_8$ most probably at 0.11%Fe, 33%Mg, 723K. The eutectic is completely divorced. The solid solubility of Fe in Al is reduced by Mg addition. In nonequilibrium conditions, massive crystals of FeAl_3 can form, even at low Fe concentration. In addition, the Mg_5Al_8 compound tends to appear, even at Mg content of 2-3%.

Fe in Al-Mg alloys has some grain refining effect. In Al-Mg alloys Fe is generally present as coarse crystals (when $\text{Mg} > 2\%$), which reduce ductility, creep, and fatigue resistance. Fe addition in Al-Mg alloys also tends to reduce corrosion resistance, but has little or no effect on stress corrosion or exfoliation.

Fe improves hot tear resistance and decreases the tendency for die sticking or soldering in die casting. The effect of Fe on mechanical properties depends on the Fe content and morphology of Fe-bearing compounds. Generally, the Fe is present as large primary or pseudo-primary crystals or as intermetallic compounds with other elements, like AlFeSi phases, which increase the hardness but decrease the ductility. These essentially insoluble phases are responsible for an improvement in strength, especially at elevated temperature. As the fraction of the insoluble phase increases with increased Fe content, castability, such as flowability and feeding characteristics, are adversely

affected. The presence of FeAl_3 produces pitting, thus decreasing corrosion resistance. However, Fe, together with Ni in Al-Si alloys, produces particularly good resistance to high temperature water or steam. Fe participates in the formation of sludging phases with Mn, Cr, and other elements.

Some recent work investigated the formation and characteristics of Fe-bearing compounds in Al casting alloys and the effect of the addition of Fe on the properties of alloys. Richard, M.^[34] studied morphologies of phases containing Fe and Mn and their formation in Al-Si9Cu3 die casting alloys. It was found that the introduction of Fe and Mn into the alloys led to the formation of several constituents containing Fe and Mn. These intermetallic hard spots consisted of a particularly coarse and dense form which results from decantation at too low a temperature during holding. The work gave the decantation temperature and the minimum holding temperature in terms of their segregation factor, which is expressed by the sum $(\text{Fe} + 2\text{Mn} + 3\text{Cr})$. The melting temperature must allow for this segregation factor and also the possible presence of coarse constituents in the charges. Flores, A. et al^[35] studied kinetics of the formation, growth, and sedimentation of the $\text{Al}_{15}(\text{Fe}, \text{Mn})_3\text{Si}_2$ phase in Al-Si-Cu alloys. It was found that when the composition in an Al-Si-Cu alloy exceeded 0.60% Fe, 0.50% Mn and 8% Si, it would form $\text{Al}(\text{Fe}, \text{Mn})\text{Si}$ type phase, commonly called sludge, at temperatures in the range of 610-660 °C. This phase led to the formation and sedimentation of very hard inclusions which produced a detrimental effect on mechanical properties of the alloy. The initial chemical composition exerted a strong influence on the kinetics of the process. Awano, Y. et al^[36] studied non-equilibrium crystallization of AlFeSi compounds in melt superheated Al-Si alloy castings. The work indicated that the morphology of an AlFeSi compound changed from needle-like to a Chinese script by heating the melt to high temperature (i.e., by melt superheating without any Fe corrector such as Mn in Al-Si alloy casting). They investigated the factors affecting this phenomenon and the effect of melt superheating on the strength of the squeeze castings of JIS AC4C alloy (normally 6.5-7.5% Si, 0.5% Fe, 0.2% Cu, 0.3% Mn, 0.20-0.40% Mg, 0.3% Zn and 0.2% Ti, equivalent to 356.0) with high Fe content. It was confirmed that when the melt was superheated, the Fe compound in Chinese script crystallized in the wide range of Fe and Si compositions, but otherwise, Fe must form the compound in the needle-like form. The AlFeSi compound in a Chinese script form crystallized through non-equilibrium solidification at a later period than the crystallization of the AlFeSi compound in the needle-like form, and it was insusceptible to the thermal history of the melt after superheating. In Al-Si-Fe alloy

castings three types of AlFeSi compounds in needle-like form, indefinite form, and a Chinese script form were observed together with small AlMgFe compounds in various shapes. The addition of Mg inhibited the crystallization of the AlFeSi compound in a Chinese script form. When solidified in a short time, the compound crystallized at a lower superheating temperature. The toughness of T6 treated squeeze casting increased when superheated at an elevated temperature. This was caused not only by the crystallization of the AlFeSi in a Chinese script form but also by the reduction in the particle size of the compound in a needle-like form.

Karpachev, V. M. et al's work^[24] indicated that Cu had the strongest positive effect on the high temperature strength of the Al-Si-Cu alloys, and alloying by Fe and Ni made it possible to reduce the content of Cu without reducing the high temperature strength of the alloys. It was reported^[32] that Fe addition reduced the tendency of hot tearing and segregation of the BS LM24 alloy. Holecek, S.^[37] studied the negative effect of Fe, less than 2%, in Al-Si alloys on mechanical properties. The alloys investigated were Na- and P-modified Al alloys containing 12-13% Si and 1.5-2% Cu. The work examined the compensation of the Fe effect by addition of Mn, Cr, Mo, Co and/or V and recommended the compensation for alloys containing high Fe contents which had unacceptable mechanical properties. Kobayashi, T. et al and Egashira, H. et al^[38, 39] conducted research on fatigue crack propagation characteristics in Al-Si system casting alloys. They used high purity Al-Si alloys (Si between hypo to hyper range) and investigated the effect of Fe, Mg, and Mn. It was found that the crack propagation rate (da/dN) in a high stress intensity range was retarded by the addition of Fe and Mn in an Al-Si-Mg alloy. Another work^[40] carried out a statistical investigation of the microstructural feature causing fatigue fracture in an Al casting alloy. It was found that Fe content had a pronounced influence. One Toyota patent^[41] claimed that contents of Fe, Mn, and Ni were strongly related to creak initiation (hot tearing) of Al die casting alloys. The creak can be prevented by controlling the content of Fe, Mn and Ni, even if recovered or scrap Al or chips were used.

Klein, F. et al^[42, 43] studied the die soldering tendency of the Al-alloy GD-AlSi9Cu3. In die casting, the Al alloy GD-AlSi9Cu3, as well as the alloy GD-AlSi7MgSb, tends to adhere to dies made of hot work tool steel. The sticking occurred for various reasons: shrinkage processes in cooling the castings in the tool; poor strength of castings due to the high temperature prior to charging the tool; high material accumulation, as well as some operating and casting design factors, such as fast

opening the closing unit; wedging of castings at opening the tool; and use of lateral cores and materials. The formation of erosion was only observed in casting the GD-AlSi7MgSb alloy and not in casting the GD-AlSi9Cu3 alloy. The steady Al sticking on certain tool parts was caused by Fe diffusion from the core into the boundary zone of the Al layer and the formation of the AlFeSi phase. The strong sticking effect was caused by the relatively homogeneous phase. Remedies could include raising the Fe content of the Al alloys to $>0.5\%$, redesign of the die and its ejection mechanism and changes in the operating procedures. At high melt flowing speed in the casting slot or at short charging time, the sticking effect was reduced because of finer structure and higher strength of castings. Also, the melt modification by Na led to reduction of the sticking effect.

4-2 Effects of Impurities

Commercial Al casting alloys almost always contain many impurity elements. Some of them are brought in inevitably by charge materials, flux, and/or reaction of the alloy with refractories and operating tools. In most of the casting alloys Fe is the main impurity of this kind and has deleterious effects on some properties, as mentioned previously. However, in die casting alloys Fe is purposely added to alleviate die sticking and is considered as an alloying element. Some of the other elements, however, do not significantly affect properties by themselves; they may be intentionally added or controlled in certain amounts to compensate for the negative effects of some other elements. For example, Mn and Cr are used as Fe correctors. Generally, most impurity elements, because of their small amounts, are present in solid solutions and have little or no effects on alloy properties, but under certain conditions and/or with other elements, they may form intermetallic compounds in the interdendritic area during the eutectic reaction or form primary phases. The formation of these compounds may affect some properties. Whether the effects are beneficial or detrimental depends on alloy composition, casting condition, and what property is concerned. In this section the impurity elements most commonly found in Al die casting alloys will be discussed. The concentrations of the impurities concerned in this work will only be those found in commercial alloys or within alloy specifications, and only the effect on commercial die casting alloys will be discussed.

Manganese

In Al die casting alloys Mn is usually controlled in the amount less than half of Fe content acting as a corrector for Fe. Fe in Al casting alloys generally reduces ductility of the alloys drastically and decreases their corrosion resistance. As mentioned previously, the addition of Mn may compensate for these negative effects of Fe. In the presence of Mn in commercial alloys the compound $(\text{Fe, Mn})_3\text{Si}_2\text{Al}_{15}$ may form often in the shape of Chinese script, thus removing the embrittling effect of the possible needle phase, FeSiAl_{15} , which is formed when Mn is absent. Decreasing corrosion resistance by Fe takes place because the FeAl_3 formed produces pitting. The addition of Mn may absorb Fe in the Mn compound, as in MnAl_3 to form the compound $(\text{Fe, Mn})\text{Al}_6$, which has a potential approximately the same as that of Al, removing the pitting effects of FeAl_3 . However, if the total Mn plus Fe content is above 0.8%, the $(\text{Fe, Mn})_3\text{Si}_2\text{Al}_{15}$ crystals may be primary and they appear as hexagonal globules (although the compound is cubic). These globules do not embrittle the alloy, but as hard spots they appreciably reduce the machinability. Increasing Mn concentration also correspondingly detracts from castability and casting soundness. Therefore, the Mn content is often controlled for only partial correction of Fe to prevent the formation of primary crystals. As was also mentioned previously, Flores, A.^[35] studied the kinetics of the formation, growth, and sedimentation of the $(\text{Fe, Mn})_3\text{Si}_2\text{Al}_{15}$ phase in Al-Si-Cu alloys. It was found that the $(\text{Fe, Mn})_3\text{Si}_2\text{Al}_{15}$ phase (sludge) would form when the composition in the alloy exceeded 0.60% Fe, 0.5% Mn and 8% Si at temperatures in the range of 610-660 °C. This phase led to the formation and sedimentation of very hard inclusions which produce a detrimental effect on the mechanical properties of the alloy. The initial chemical composition exerts a strong influence on the kinetics of the process.

In an Al-Mn binary system, the possible values of the solid solubility of Mn in Al are 1.8% at 930K, 1.0% At 900k and 0.2% At 700K. In low Mn content the phase in equilibrium with Al is MnAl_6 . Mn is used extensively in wrought alloys, mainly as a hardener to enhance the strength of the alloys through work hardening and as a corrector for Fe. In the absence of work hardening, Mn offers no significant strengthening benefits in Al casting alloys. However, secondary Al casting alloys always contain high Mn levels brought in by scraps of wrought materials. In commercial alloys the solubility of Mn in Al is reduced by the presence of Fe and Si, and Mn may form compounds with Fe, Al, and Si, such as $(\text{Fe, Mn})_3\text{Si}_2\text{Al}_{15}$,

(Fe,Mn)Al₆. In Al-Si alloys Mn additions may improve high-temperature properties slightly, enhance fatigue resistance, and reduce shrinkage^[32]. Mn and Zn in secondary alloys, where many elements are present in small amounts, compensate for Cu and Ni. Corrosion resistance was reported as very close to that of primary alloys. Mn in Al-Mg (>2%) alloys tends to form coarse particles and slightly increases the low-temperature properties and creep resistance. Some evidence^[26] exists that a high volume fraction of MnAl₆ may beneficially influence internal casting soundness. It was reported^[41] that Mn with Fe and Ni was strongly related to crack initiation (hot tearing). Mn can also be employed to alter the response in chemical finishing and anodizing.

Nickel

In Al-Si alloys Ni slightly increases the strength at room and elevated temperatures and may slightly increase ductility if it acts as a Fe corrector; otherwise it reduces ductility. Ni reduces the coefficient of thermal expansion. Ni and Fe in Al-Si alloys provide the alloys with particular good resistance to high-temperature water or steam.

Chromium

Cr, like Mn, is usually used as Fe corrector. In Al-Si alloys Cr slightly increases the strength at room and elevated temperatures and slightly reduces ductility. Cr may form sludge with Mn and Fe. Like Mn, Cr in Al-Mg (>2%) alloys tends to form coarse particles and increases the low-temperature properties and creep resistance.

Zinc

Zn in amounts up to 1% in Al alloys is in solid solution and does not form any visible phase. Higher Zn amounts together with Mg may lead to the formation MgZn₂. Generally, Zn does not have a significant effect on Al casting alloys, and so, for many of the alloys Zn is specified at higher levels than other impurities, for example, up to 3% in some Al die casting alloys. However, it does have an influence on some properties of Al casting alloys. In Al-Si alloys Zn decreases high-temperature strength, tends to increase hot tearing tendency, but improves machinability. Zn up to 2-3% has no effect on corrosion properties. In secondary alloys, where many elements are present in small amounts, Zn and Mn compensate for Cu and Ni, and corrosion was

reported as very close to that of primary alloys. In Al-Mg alloys Zn enhances castability and strength.

Klein F.^[44] studied the effects of Zn content on the mechanical properties of the Al die casting alloy GD-AlSi8Cu3. The experiments were carried out on die cast flat test bars with Zn contents of 1.2% and 2%. The results indicated that the alloy with 2% Zn had higher tensile strength than that with 1.2% Zn, even after one year of aging. For high rates of pouring at the ingate the influence of the higher Zn content was low for thin-walled castings in comparison with castings of greater wall thickness. But this advantage was cancelled by an increased tendency to hot tearing.

Tin and Lead

Sn and Pb, if present together with Mg in Al-Si alloys, tend to enter a Mg_2Si phase. In Al-Si alloys Sn and Pb decrease the high-temperature strength, and improve machinability, but Sn has a deleterious effect on the corrosion resistance of the alloys. Pb in Al-Mg alloys is used to improve machinability, supposedly without a loss of strength and corrosion resistance.

Other elements

Be was reported to correct for Fe. It slightly increases the strength at room and elevated temperatures of Al alloys and may also increase ductility, if acting as a Fe corrector; otherwise, it decreases it. Be additions of as low as a few ppm may be effective in reducing oxidation at high temperatures, especially at the molten state, and reducing associated inclusions in Mg-containing compositions. At higher levels ($>0.04\%$) Be affects the form and composition of Fe-bearing intermetallics, markedly improving strength and ductility.

Co and Mo are Fe correctors and Co is considered to be the best Fe corrector. In Al-Si alloys both of them may slightly increase the strength at room and elevated temperatures and may also slightly increase ductility, if acting as Fe correctors; otherwise they reduce it. Co also may increase fatigue resistance.

In Al-Si alloys Bi improves wear resistance and machinability. Ag additions were reported to increase elongation. V at 0.1-0.2% was reported to refine the FeMn

compounds. Zr slightly increases strength, reduces ductility, and has a grain refining effect. Cd in concentrations exceeding 0.1% improves machinability. Rare earth elements increase creep resistance and reduce gas porosity. Ca has a deleterious effect on corrosion resistance, increases H solubility, and is responsible for casting porosity at trace concentration levels.

Generally, in Al-Si alloys most elements decrease fluidity of the molten alloy and enhance hot tearing resistance, except for Zn. Elements with high melting temperature tend to be beneficial to high-temperature resistance.

4-3 Melt Treating Elements

Several elements may have a grain refining effect, but the most commonly used grain refiner in Al casting alloys is Ti-B master alloy. Ti and B are usually added in amounts well within their solid solution and do not form any separate phase. Fe reduces their solubility so that in the presence of Fe, less Ti and B are needed for grain refinement. However, Ti is often employed at levels greater than those required for grain refinement to reduce cracking tendencies in hot tearing compositions. Na, Sr, and Sb are effective modifiers of hypoeutectic Al-Si alloys. Besides modifying eutectic Si particles, these elements may have some other effects. For example, additions of Na and Sr increase porosity, Sb improves machinability, but by combining with Mg, Sb reduces the response to heat treatment. Na is embrittling in Al-Mg alloys. Grain refinement and modification have been research focuses of Al casting alloys for many decades and still attract many researchers. Numerous papers and books have discussed these areas. Because these topics are concerned more with general Al casting alloys and have been well documented, this literature review will skip them. However, though grain refinement renders little or no improvement in mechanical properties to Al die casting alloys, it improves fluidity, consequently resulting in castings with uniform microstructure and less defects. Its beneficial effect on refining and evenly distributing porosity may also have some benefits to die castings. So, grain refinement is beneficial to Al die casting alloys and is recommended for certain applications.

P and some other elements have an effect on refining primary Si particles of hypereutectic Al-Si alloys. The most effective and widely used commercial refiner is P. However, in hypoeutectic Al-Si alloys, P may react with Na and Sr, diminishing their modifying effect. The application of P as primary Si refiner has also been

investigated extensively and is well documented. Considering that the review of the refinement of hypereutectic Al-Si alloys has been carried out frequently, the present work will not repeat the general information about it, but because hypereutectic Al-Si alloys are widely used in die castings, this review will cover some recent development which is related to die casting. This will be covered when discussing the 390 alloy in Section VII.

V Solidification and Microstructure of Aluminum Die Casting Alloys

Microstructure of an alloy generally refers to kinds of phases presented in the alloy, size, morphology and distribution of these phases, dendrite arm spacing (DAS), grain structure, and sometimes also to porosity, inclusions and their morphology, size, and distribution. This review will focus on the various aspects of phases and leave the porosity and inclusions to the project of "effect of casting quality on properties of Al die casting alloys." The microstructure of Al die casting alloys generally consists of primary phases if there are any, alpha Al dendrites, Al-Si eutectic and some interdendritic phases, depending on the alloy composition, processing and processing history. The primary phases are mostly primary Si crystals, as in hypereutectic Al-Si alloys, or some intermetallic compounds, such as AlFeSi phases. In all commercial Al die casting alloys there are alpha Al dendrites, which solidify first or following or together with the formation of the primary phases. In the later stage of the formation of the dendrite network in Al-Si alloys the Al-Si eutectic structure forms in the interdendritic areas. Along with and/or following the formation of Al-Si eutectic, various intermetallic compounds precipitate. Later in this section the solidification and microstructures of typical hypoeutectic, eutectic, and hypereutectic Al-Si die casting alloys and an Al-Mg die casting alloy will be discussed.

5-1 Solidification and microstructure of hypoeutectic Al-Si-Cu die casting alloys

Backerud, L. et al^[33] studied the solidification characteristics and microstructure of various Al casting alloys. They used a unique thermal analysis method in combination with metallography and modern microanalytical techniques. The thermal analysis method consists of two thermocouples placed in a small cylinder mold, one close to the inner wall and the other at the center of the mold. When the molten Al alloy solidifies in the mold, the thermocouples record two cooling curves of the solidification process. They analyzed the cooling curve of the center of the specimen, first derivation of this cooling curve, and the temperature differences between the center and wall versus time. The microstructures and the compositions of all the phases in the specimen were also analyzed through metallography, energy dispersive x-ray analysis (EDX) and x-ray powder diffraction technique (XRD). These analyses reveal several important solidification characteristics of the alloys in a quantitative manner. The information

generated includes reactions during solidification (temperature interval, phases and fraction of solid (f_s) precipitated in each reaction) and the point of dendrite coherency (temperature and fraction of solid at this point). In reality, the point of dendrite coherency is the point at which the solid phase network is established throughout the entire casting, and from this moment phenomena such as macrosegregation, shrinkage, porosities, and hot tearing start to develop.

The 380 die casting alloys were intensively investigated in their experiments. In order to elucidate the influence of certain alloying elements on the structure development, various compositions in the 380 alloy family were studied. The compositions of the 380 alloys investigated in their study are shown in Table 5-1.

Table 5-1 Compositions of the 380 alloys used in the experiments

Alloy	Composition (%)	Element							
		Si	Fe	Cu	Mn	Mg	Zn	Ti	Fe/Mn
A380.1	AA-standard	7.5-9.5	1.0	3.0-4.0	0.5	0.10	2.9	0.35	2.0
B380.1	AA-standard	7.5-9.5	1.0	3.0-4.0	0.5	0.10	0.9	0.35	2.0
Eu380	Actual sample	9.1	1.0	3.2	0.38	0.34	1.1	--	2.6
A380.1	Actual sample	8.9	0.88	3.26	0.24	0.06	2.05	0.03	3.8
B380.1	Actual sample	9.3	0.89	3.42	0.28	0.02	0.76	0.04	3.2
(1)	Actual sample	7.2	0.52	1.2	0.18	0.10	0.45	0.013	2.9
(2)	Actual sample	11.0	0.51	1.5	0.11	0.04	0.48	0.01	4.6
(6)	Actual sample	9.3	0.52	2.6	0.11	0.59	0.51	0.013	4.7
(7)	Actual sample	8.8	1.32	2.4	0.12	0.12	0.50	0.01	11.0
(8)	Actual sample	8.7	1.35	2.5	0.48	0.12	0.53	0.012	2.8
(12)	Actual sample	9.0	0.63	2.5	0.40	0.04	0.47	0.014	1.6

Among these alloys, European-type 380 (Eu380), A380.1, and B380.1 represent a rather narrow concentration range of interesting elements, i.e., Si 8.9-9.3, Fe 0.88-1.0%, Cu 3.2-3.4% and Mn 0.24-0.38%. The 380 (1) was prepared to illustrate the effect of low contents of Si, Cu, Fe, and Mn on the structure development. The 380 (2) is characterized by a very high Si content, while the Cu level is kept low. The 380

(6) has high Mg content. The 380 (7) contains a very high amount of Fe, while the Mn content is low. The 380 (8) has a high Fe content, and at the same time that the Mn content is rather high, the Fe/Mn ratio = 2.8. The 380 (12) is used to demonstrate the influence of low Fe and high Mn contents on the structure development.

A summary of the pre-eutectic reactions observed in the experiments is given in Table 5-2. All reactions observed in the 380 type alloys are described in Table 5-3, and the phases observed in the solidified samples are given in Table 5-4. Note that the reaction number (4) is only observed in samples with high Mg content.

The sequence of solidification, especially the pre-eutectic reactions can be explained by looking at the phase diagrams. The compositions listed in Table 5-1 are demonstrated in Fig. 5-1 a), b) and c), where the Si and Fe contents of each sample are plotted in three diagrams representing a) low ($\sim 0.1\%$), b) medium ($\sim 0.25\%$), and c) high ($\sim 0.4\%$) Mn contents, respectively. The solid dots (rectangle, circular, and triangle) in the phase diagrams represent the initial composition of each alloy. From the equilibrium point of view, one should expect the primary reactions to be as follows:

1. For sample 380(1), 380(6), 380(2), A380 and 380(12): α -aluminum
2. For sample B380 and Eu380: $\text{Al}_{15}(\text{MnFe})_3\text{Si}_2$ -phase
3. For sample 380(7) and 380(8): Al_5FeSi -phase

These processes may occur, but only at very low cooling rates. At cooling rates of approximately 1°C/s or higher, all samples start to solidify by forming a dendritic network. Irrespective of the actual composition, the coherency point is reached at a temperature only $1-2^\circ\text{C}$ below the starting growth temperature and after the formation of $\sim 16\% f_s$. This means that at this cooling rate (and higher) the kinetics of phase formation favors dendritic growth of α -aluminum before any intermetallic precipitates.

The situation is demonstrated in Fig. 5-2 a), b), and c), which shows the actual composition (hollow dot) after the development of the dendritic network, calculated according to the Scheil equation. It is from these new compositions that we have to consider which should be the second phase to precipitate, the Al_{15} - or Al_5 phases, or possibly the two together. Following the precipitation of the second phase, the main eutectic occurs. During and after the main eutectic reaction the formation of some

(a) Low Manganese content

(b) Medium Manganese content

(c) High Manganese content

Fig. 5-1 Location of actual composition of samples of the 380-type alloys in AlFeSi-phase diagrams at various Mn-levels

Fig. 5-2 Calculated segregation effects after formation of a dendritic network

intermetallic phases takes place. The kinds and the fractions of the second phases (pre-eutectic) and the phases formed together with (co-eutectic precipitates) and after Al-Si eutectic depend on cooling rates as well as the compositions. The following are microstructures of various 380 alloys, showing the effects of compositions and cooling rates.

Table 5-2 Quantification of pre-eutectic reactions in the various 380 type alloys

Sample	Start of solidification °C	Coherency point		Precipitation AlFeSi phase f_s	Start of eutectic	
		f_s	°C		f_s	°C
Eu380	577	18	576	23-27	33	561
A380	572	16	571	12-21	24	561
B380	575	16	574	18-21	26	564
380(1)	600	18	598	30-41	(50)*	567
380(2)	571	18	571	--	(18)*	571
380(6)	582	18	582	26-33	38	560
380(7)	581	17	581	20-27	35	564
380(8)	583	16	582	23-29	42	566
380(12)	578	14	577	18-24	28	565
Mean value	580	16±2	579	22-28	33	565
		DC-1°C**				

* Extreme values not included in mean value

** DC--Dropping Temperature between start of solidification and coherency point

Table 5-3 Reactions during solidification of the various 380 type alloys

Reaction No	Reaction
1	Development of dendritic network
2	Precipitation of AlMnFe-containing phases
3	Main eutectic reaction involving precipitation of Si and Mn-Fe-containing phases
4	Precipitation of Mg_2Si
5	Precipitation of Al_2Cu
6	Formation of complex eutectics, containing Al_2Cu and $Al_5Mg_8Si_2Cu_2$

Table 5-4 Phases observed by microscopy / SEM / EDX

No.	1	2	3	4	5	6	7
Phase	Alpha-Al*	Al ₁₅ Mn ₃ Si ₂ *	Al ₅ FeSi	Si*	Al ₂ Cu*	Al ₅ Cu ₂ Mg ₈ Si ₆	Mg ₂ Si*
Charac- teristics	Dendrites	Brown script	Needles	Grey	Pink particles		Black

* Confirmed by XRD.

Eu380 (9.1%Si, 3.2%Cu, 1.0%Fe, 0.38%Mn)

The principal reactions in the Eu380 alloy are shown in Table 5-5. Formation of a dendritic network is followed by precipitation of the AlSiFe phase, which at low cooling rates (e.g., 0.2 °C/s) consists of Al₁₅- particles (see Fig. 5-3), but at higher cooling rates (e.g., 0.6, 5 °C/s) consists of the Al₅-phase (Figs 5-4 and 5-5).

The co-eutectic precipitation of the Fe-bearing phase occurs as the Al₅-phase at the intermediate cooling rate (0.6 °C/s, Fig. 5-2), but as the Al₁₅-phase at higher cooling rates, from 5 °C/s (Fig. 5-5) and up to 200 °C/s. At higher cooling rates the growth kinetics obviously favor the formation of the cubic Al₁₅-phase above that of the monoclinic Al₅-phase at this composition of the sample.

At the end of the solidification process, the precipitation of Al₂Cu (5) and the more complex eutectic reaction (6) take place. Fig. 5-6 shows compact Al₁₅-particles sedimented in a sample that was cooled at a very low rate. These particles form what is called "sludge."

Table 5-5 Solidification characteristics of the Eu380 alloy under various cooling rates

Cooling rate (°C/s)	0.2	0.6	5.0
Dendrite arm spacing (μm)	74	57	28
Solidification temp. range (°C)	61	5	19
Solidification time (s)	93	103	111
End temperature (°C)	479	474	465
Coherency point (f _s %/°C)		18/576

Fig. 5-3 Microstructure from a sample of alloy Eu380 cooled at a rate of 0.2 °C/s, showing large Al₁₅(Fe, Mn)₃Si₂ (F, brown), Si (S, gray), Al₂Cu (C, pink); X560

Fig. 5-4 Microstructure from a sample of alloy Eu380 cooled at a rate of 0.6 °C/s, showing large Al₅FeSi (F, brown), Si (S, gray), Al₂Cu (C, pink); X560

Fig. 5-5 Microstructure from a sample of alloy Eu380 cooled at a rate of 5 °C/s, showing large Al_5FeSi (F, brown), Si (S, gray), Al_2Cu (C, pink); X560

Fig. 5-6 Microstructure of alloy Eu380 from the bottom of an extremely slowly cooled sample showing sedimental $\text{Al}_{15}(\text{Fe}, \text{Mn})_3\text{Si}_2$ particles. Such particles form what is called "sludge"; X220

Table 5-6 Summary of reactions in the Eu380 alloy

Cooling rate (°C/s)	Reaction No.	T (°C)	t (s)	f _s (%)	Acc. f _s (%)
0.2	1	578-577	60	14	14
	(1-2)	577-572	28	5	19
	2	572-562	61	8	27
	(2-3)	564-561	23	4	31
	3	561-559	156	29	60
	(3-5)	559-491	424	35	95
	5	491-478	35	3	98
	(5-6)	487-485	18	1	99
	6	485-479	22	1	100
0.6	1	577-576	29	18	18
	(1-2)	576-567	16	5	23
	2	567-566	1	4	27
	(2-3)	566-561	19	6	33
	3	561-557	53	28	61
	(3-5)	557-489	147	32	93
	5	489-485	14	4	97
	(5-6)	485-480	13	2	99
	6	480-474	14	1	100
5.0	1	576-575	7	27	27
	(1-2)	575-564	2	4	31
	2	564	2	4	35
	(2-3)	564-558	3	7	42
	3	558-552	11	32	74
	(3-5)	552-491	16	18	92
	5	491-483	4	5	97
	(5-6)	483-479	5	1	98
	6	479-465	5	2	100

In the *A380 alloy* (8.9%Al, 3.26%Cu, 0.88%Fe, 0.24%Mn), at low cooling rates (e.g., 0.3 °C/s) the Al₁₅-phase grows simultaneously with the alpha-Al, but at a rate of ~1 °C/s the separation of the two phases is obvious. This separation suggests that the Al₁₅-phase forms during the main eutectic reaction. In the main eutectic reaction at ~1 °C/s, Si and the Al₅-phase were found to precipitate together and this is also the case at a cooling rate of 6 °C/s. The final reactions are shown in Fig. 5-7 where the brown particles between the pink Al₂Cu crystals constitute the Al₅Mg₈Cu₂Si₆.

In the *B380 alloy* (9.3%Si, 3.42%Cu, 0.89%Fe, 0.28%Mn), the composition and hence the structure development in the sample are similar to the phase of A380. At a low cooling rate the Al₁₅-phase precipitates together with the dendrites, while the Al₅-phase dominates at the higher cooling rates. In the co-eutectic precipitation there is a mix of the two Fe-bearing phases.

The380(1) (7.2%Si, 1.2%Cu, 0.52%Fe, 0.18%Mn) alloy has low contents of Si, Cu, Fe, and Mn. As expected, the pre-eutectic part of the sample is very large (~50% f_g) and the dendrites occupy a large part of the total volume. In the eutectic regions the Al₁₅-phase as well as the Al₅-phase are found. This mix of phases is also seen at cooling rates up to 50 °C/s, while the Al₁₅-phase, due to kinetic reasons, dominates at still higher cooling rates. It is clearly shown that the amount of final eutectics is reduced in this alloy, due to the low Cu content.

The380(2) (11.0%Si, 1.5%Cu, 0.51%Fe, 0.11%Mn) alloy is characterized by a very high Si content, while Cu and Mn levels are low. Though the composition is very close to the eutectic, the solidification process starts with the formation of a dendritic network. This leads to a supersaturation of Si in the remaining liquid and some primary Si particles will form before the start of the main eutectic reaction. Even at very high cooling rates, the Al₅-phase is the only Fe-bearing phase precipitating in areas, which would be expected at the low Mn concentration of this alloy.

The 380(6) (9.3%Si, 2.6%Cu, 0.52%Fe, 0.11%Mn, 0.59%Mg) alloy contains a high Mg content, but low Mn. The high Mg content leads to the formation of the Mg₂Si phase (Fig. 5-8) in addition to Si, AlFeSi, Al₂Cu and complex eutectics. At the low

Fig. 5-7 Microstructure from a sample of alloy A380(1) cooled at a rate of 0.3 °C/s, showing small dark-brown particles between the Al₂Cu crystals (C, pink), representing the final precipitates of the phase Al₅Mg₈Si₆Cu₂ (M); X560

Fig. 5-8 Microstructure from a sample of alloy 380(6) cooled at a rate of 0.6 °C/s, showing black Mg₂Si phase, in addition to Al₅FeSi (F, brown), Si (S, gray), Al₂Cu (C, pink) and complex eutectics; X560

Mn content, the Al_5 -phase is the only Fe-bearing phase observed at all cooling rates (0.6-100 °C/s).

The 380(7) (8.8%Si, 2.4%Cu, 1.32%Fe, 0.12%Mn) alloy contains very high Fe content, while the Mn content is low. The Fe/Mn ratio is 11. Following the formation of dendrites, the second phase consists of the formation of very large Al_5 -particles, and such large particles are found even up to high cooling rates (e.g., 100-150 °C/s).

The 380(8) (8.7%Si, 2.5%Cu, 1.35%Fe, 0.48%Mn) alloy has a high Fe content, and at the same time Mn content is rather high, the Fe/Mn ratio =2.8. As is expected, a strong reaction precipitating a mix of Al_{15} - and Al_5 - pre-eutectic particles takes place. Also, in eutectic areas the mix of Fe-bearing phases is shown up to high cooling rates (e.g., 40-50 °C/s).

The 380(12) (9.0%Si, 2.5%Cu, 0.63%Fe, 0.40%Mn) alloy contains low Fe and high Mn contents with the Fe/Mn ratio=1.6. As can be expected, the Al_{15} -phase dominates as Fe-bearing compound. No trace of the Al_5 -phase is found in a wide range of cooling rates (0.7-100 °C/s).

5-2 Solidification and microstructure of eutectic Al-Si die casting alloy 413

In Backerud, L. et al's work the A413.0 alloy was chosen as the eutectic Al-Si die casting alloy for study. The standard composition and the composition of the alloy studied are shown in Table 5-7. The actual sample is slightly hypoeutectic. The solidification process starts with the development of a dendritic network, which, due to growth kinetics, occupies a larger volume of the sample than expected from the equilibrium diagram. While the dendrites grow, Si is enriched in the interdendritic regions, leading to the frequent formation of primary Si particles in the structure. During the coupled, eutectic reaction, Al_5FeSi dominates as the Fe-bearing phase at low cooling rates (e.g., 0.3 °C/s), while $Al_{15}(Mn, Fe)_3Si_2$ takes over at high cooling rates (e.g., 5 °C/s). Tables 5-8, 5-9, 5-10, and 5-11 show the solidification reactions, phases, and solidification characteristics of the A413.1 alloy tested.

Table 5-7 Composition of the 413 alloy

Alloy	Composition (%)	Element							
		Si	Fe	Cu	Mn	Mg	Zn	Ti	Ni
A413.1	AA-standard	11.0-13.0	1.0	1.0	0.35	0.10	0.50	--	0.50
413.1	Actual sample	11.4	0.46	0.09	0.18	0.03	1.1*	--	

* The sample tested is outside specification

Table 5-8 Reactions during solidification of the 413.0 alloy

Reaction No	Reaction	Suggested temperature (°C)
1	Development of dendritic network	574
2a	Liq. => Al + Al ₅ FeSi	572
2b	Liq. => Al + Al ₁₅ (Mn, Fe) ₃ Si ₂	572
2c	Liq. => primary Si	572
3a	Liq. => Al + Si + Al ₅ FeSi	575
3b	Liq. => Al + Si + Al ₁₅ (Mn, Fe) ₃ Si ₂	573

Table 5-9 Phases observed by microscopy / SEM / EDX

No.	1	2	3	4	5
Phase	Alpha-Al*	Primary Si	Si	Al ₅ FeSi	Al ₁₅ Mn ₃ Si ₂ *
Characteristics	Dendrites	Big gray particles	Gray	Needle	Brown script

Table 5-10 Solidification characteristics of the 413.0 alloy under various cooling rates

Cooling rate (°C/s)	0.3	0.7	5.0
Dendrite arm spacing (μm)	88	63	34
Solidification temp. range (°C)	17	17	28
Solidification time (s)	540	230	42
End temperature (°C)	557	557	546
Coherency point (f_s %/°C)			

Table 5-11 Summary of reactions in the 413.0 alloy

Cooling rate (°C/s)	Reaction No.	T (°C)	t (s)	f_s (%)	Acc. f_s (%)
0.3	1	574-573	39	8	8
	(1-2)	573-572	45	9	17
	2	572	35	8	25
	3	572-570	280	55	80
	(3-end)	570-557	141	20	100
0.7	1	574	16	8	8
	(1-2)	574-572	31	14	22
	2	572	26	13	35
	3	572-569	119	53	88
	(3-end)	569-557	38	12	100
5.0	1	574	5	10	10
	(1-2)	574	5	17	27
	2	574-573	4	11	38
	3	573-567	24	59	97
	(3-end)	567-546	4	3	100

5-3 Solidification and microstructure of hypereutectic Al-Si-Cu die casting alloys

Backerud, L. et al studied solidification and microstructure of two compositions of the 390 alloys. The compositions, standard and investigated, are shown in Table 5-12. The actual sample for alloy A390.1 was made up in the laboratory and the composition was lower in Si, higher in Cu than the AA standard, and had no P. The sample of alloy B390.1 had Si and Cu within the specification, and contained some P and higher Mn than that in the standard composition.

Table 5-12 Composition of alloys 390.0 and B390.1

Alloy	Composition (%)	Element							
		Si	Fe	Cu	Mn	Mg	Zn	Ti	Other
A390.1	AA-standard	16.0-18.0	0.4	4.0-5.0	0.1	0.50-0.65	0.10	0.20	
B390.1	AA-standard	13.0-16.0	1.0	4.0-5.0	0.50	0.50-0.65	1.4	0.20	
390.0	Actual sample	15.0*	0.3	5.5*	0.01	0.65	0.065	0.01	
B390.1	Actual sample	17.45	0.74	4.81	0.27	0.56	1.00	0.06	

* The sample tested is outside specification

The solidification reactions, data, and phases of these two alloys are listed together in Tables 5-13, 5-14, 5-15 and 5-16. The information about alloy B390.1 is given in parentheses.

In alloy 390.0, because it has no P, the primary Si crystals are very large. The eutectic areas show eutectic Si, Al_5FeSi needles, large amounts of Al_2Cu , and some $\text{Al}_5\text{Mg}_8\text{Cu}_2\text{Si}_6$ particles. Most of the Mg_2Si seems to have been consumed during the reaction (5).

In alloy B390.1, due to the higher Si content and more effective nucleation (contains some P), the density of primary Si particles is higher. However, increasing the cooling rate from 0.5 to 4 °C/s does not seem to influence the primary precipitation to any significant extent. There is a small "overshoot" in the primary reaction, resulting in the formation of some dendrites in areas around agglomerates of primary Si particles. During the eutectic reaction $\text{Al}_{15}(\text{Mn}, \text{Fe})_3\text{Si}_2$ as well as Al_5FeSi are found together with the same phases as in the 390.0 alloy.

Table 5-13 Reactions during solidification in the 390.0 and B390.1 alloys

Reaction No	Reaction	Suggested temperature (°C)
1	Primary Si	(636)* [670]**
2	Development of dendritic network	(561) [557]
3	Liq. => Al + Si + Al ₅ FeSi	575
[3a]	[Liq. => Al + Si + Al ₁₅ (Mn, Fe) ₃ Si ₂]	[573]
[3b]	[Liq. => Al + Si + Al ₅ FeSi]	[575]
4	Liq. => Al + Si + Mg ₂ Si	555
5	Liq. + Mg ₂ Si => Al + Si + Al ₂ Cu + Al ₅ Mg ₈ Cu ₂ Si ₆	512
6	Liq. => Al + Al ₂ Cu + Al ₅ Mn ₈ Cu ₂ Si ₆ + Si	507

* The information in () is for 390.0 only

** The information in [] is for B390.1 only

Table 5-14 Phases observed by microscopy / SEM / EDX in alloys 390.0 and B390.1

No.	1	2	3	4	5	6	7
Phase	Alpha-Al	Primary Si	Si [Al ₁₅ -(Mn, Fe) ₃ Si ₂]	Al ₅ FeSi	Al ₂ Cu	Al ₅ Mg ₈ Cu ₂ Si ₆	
Characteristics	Dendrites	(Big gray) [Gray, large bulk]	Gray	Brown scrip	Needle	Pink	Gray brown bulk

Table 5-15 Solidification characteristics of the 390.0 and B390.0 alloys under various cooling rates

Cooling rate (°C/s)	(0.4)	[0.3]	(0.6)	[0.5]	(5.0)	[4.0]
Dendrite arm spacing (μm)		[43]		[31]		[19]
Solidification temp. range (°C)	(141)	[186]	(144)	[186]	(139)	[217]
Solidification time (s)	(957)	[1054]	(417)	[470]	(59)	[72]
End temperature (°C)	(493)	[484]	(492)	[482]	(483)	[448]
Coherency point (f _s %/°C)						

Table 5-16 Summary of reactions in the 390.0 and B390.1 alloys

Cooling rate (°C/s)	Reaction No.	T (°C)	t (s)	f _s (%)	Acc. f _s (%)
(0.4) [0.3]	1	(634-617)[670-663]	(49) [21]	1	1
	1-2	(617-561)[663-557]	(242)[384]	(5) [11]	(6) [12]
	2	(561) [557-560]	(22) [40]	(4) [10]	(10) [22]
	(2-3)	(561-560)	(13)	(3)	(13)
	3	(560-558)[560-557]	(108)[112]	(21) [25]	(34) [47]
	(3-4)	(558-540)	(498)	(41)	(75)
	(4)	(540-503)	(14)	(14)	(89)
	[3-5]	[557-496]	[408]	[45]	[92]
	[5]	[496-497]	[15]	[3]	[95]
	5-6	(503-493)[497-495]	(11) [18]	(11) [2]	(100) [97]
	[6]	[495-484]	[56]	[3]	[100]
(0.6) [0.5]	1	(636-602)[668-652]	(60) [25]	(5) [3]	(5) [3]
	1-2	(602-561)[652-556]	(61) [143]	(4) [10]	(9) [13]
	2	(561) [556-560]	(16) [11]	(9) [11]	(18) [24]
	(2-3)	(561-557)	(8)	(5)	(23)
	3	(557-556)[560-555]	(80) [91]	(31) [36]	(54) [60]
	(3-4)	(556-536)	(78)	(26)	(80)
	(4)	(536-534)	(17)	(4)	(84)
	[3-5]	[555-496]	[155]	[31]	[91]
	(4-5)	(534-502)	(56)	(7)	(91)
	[5]	[496-497]	[21]	[6]	[97]
	(5+6)	(502-492)	(41)	(9)	(100)
(5.0) [4.0]	[6]	[497-482]	[24]	[3]	[100]
	1	(622-562)[665-616]	(15) [13]	(10) [10]	(10) [10]
	1-2	(562-557)[616-556]	(2) [15]	(3) [11]	(13) [21]
	2	(557) [556-557]	(3) [9]	(11) [36]	(24) [57]
	(2-3)	(557-554)	(2)	(7)	(31)
	3	(554-547)[557-549]	(14) [5]	(55) [16]	(86) [73]
	(3-4)	(547-517)	(14)	(5)	(91)
	(4)	(517-503)	(3)	(1)	(92)
	[3-5]	[549-496]	[14]	[9]	[82]
	5-6	(503-483)[494-485]	(6) [3]	(8) [5]	(100) [93]
	[6]	[485-448]	[11]	[7]	[100]

5-4 Solidification and microstructure of Al-Mg die casting alloy 518

Backerud, L. et al also studied solidification and microstructure of the 518 alloy. The composition investigated and the standard composition are shown in Table 5-17. The solidification reactions, phases, and solidification characteristics of the alloy are listed in Tables 5-18, 5-19, 5-20 and 5-21.

Table 5-17 Composition of alloy 518

Alloy	Composition (%)	Element							
		Si	Fe	Cu	Mn	Mg	Zn	Ti	Other
518.2	AA-standard	0.25	0.7	0.10	0.10	7.6-8.5	0.05		
518	Actual sample	0.10	0.16	0.01	0.01	7.6	0.02		

Table 5-18 Reactions during solidification in the 518.2 alloy

Reaction No	Reaction	Suggested temperature (°C)
1	Development of dendritic network	612
3	Liq. => Al + Al ₃ Fe	612
3	Liq. => Al + Al ₃ Fe + Mg ₂ Si	587
4	Liq. => Al + Si + Mg ₂ Si	555
5	Liq. => Al + Al ₃ Fe + Mg ₂ Si + Al ₈ Mg ₅	448

Table 5-19 Phases observed by microscopy / SEM / EDX in alloy 518.2

No.	1	2	3	4	5
Phase	Alpha-Al	Al ₃ Fe	Mg ₂ Si	Al ₈ Mg ₅	
Characteristics	Dendrites	Gray needles	Light brown particles	Black particles	

Table 5-20 Solidification characteristics of the 518.2 alloy under various cooling rates

Cooling rate (°C/s)	0.2	0.4	6.0
Dendrite arm spacing (μm)	93	58	32
Solidification temp. range (°C)	177	174	183
Solidification time (s)	995	454	52
End temperature (°C)	435	437	428
Coherency point ($f_s\%$ /°C)		29/609	

Table 5-21 Summary of reactions in the 518.2 alloy

Cooling rate (°C/s)	Reaction No.	T (°C)	t (s)	f_s (%)	Acc. f_s (%)
0.2	1+2	612-611	77	21	21
	(2-3)	611-532	490	72	93
	3	532-518	65	3	96
	(3-4)	518-437	343	3	99
	4-end	437-435	20	1	100
0.4	1+2	611-609	51	29	29
	(2-3)	609-524	230	62	91
	3	524-512	24	2	93
	(3-4)	512-438	140	5	98
	4-end	438-437	9	2	100
6.0	1+2	611-609	10	35	35
	(2-3)	609-548	21	54	89
	3	548-511	7	7	96
	(3-4)	511-439	11	2	98
	4-end	439-428	3	2	100

The solidification sequence starts with the development of a dendritic network around 610 °C. In an Al-Mg binary alloy system, one should expect a final eutectic reaction involving the precipitation of the Al_8Mg_5 phase to occur at 450 °C. However, the Fe concentration (initially 0.16%) will increase in interdendritic areas and reach a level above 1.8% somewhere below 550 °C (it can be confirmed by use of the Scheil equation). Under this circumstance Al_3Fe will precipitate. The Al_3Fe is easily recognized by its "needle" form. When at the low level of Si (Si = 0.1%), some Mg_2Si may also form together with Al_3Fe during a third reaction.

Finally, a complex eutectic reaction completes the solidification:

The equilibrium temperature of this reaction is given as 449 °C. Due to nucleation and the kinetics of growth, this reaction has been measured to occur around 440-430 °C. (The Mg_2Si phase appears black on the photomicrographs.) Due to segregation, the relative rate and amount of the final eutectic reaction increases with the cooling rate.

Quantitative description of the microstructure of Al foundry alloys

Jaquet, J. C. et al^[45] also developed a program to quantitatively describe the microstructure of Al foundry alloys. In their program, the microconstituents considered to be relevant were the Al and Al-Si eutectic phase, the Si and intermetallic particles as well as microporosity. The corresponding microstructural parameters chosen were the area fraction of the eutectic phase, the mean chord of the Al and Al-Si eutectic phase, the area fraction, length, breadth, shape factor, and distribution of the Si + intermetallic particles, as well as the volume fraction, length, and shape factor of the micropores. The program showed that there existed a clear relationship between the mean chord and the secondary dendrite arm spacing of the Al phase.

The microstructures of Al alloys, wrought and casting alloys including die casting alloys, have been intensively investigated and well documented since Al was discovered. Except a very few, like Backerud et al's work mentioned above, however, most of the studies are scattered and qualitative. Because the commercial alloys mostly contain many impurity elements, these alloys are actually very complicated systems and

may have a large number of phases in their structure. This can be clearly seen in Tables 5-22, 5-23, 5-24, and 5-25, which show the phases formed in the Al-Si alloys, the eutectics present in Al-Si alloys, the phases formed in Al-Mg₂Si alloys, and the possible low-melting eutectics in Al-Mg₂Si alloys, respectively^[18]. The microstructure depends highly on casting conditions as well as on alloy composition. It is not uncommon that the results from different studies, which were claimed to be conducted on the same alloy, were not consistent. A possible reason for this is that the studies conducted separately were with different alloy compositions or under different casting conditions. Therefore, for practical purposes, it is desirable that the study of Al die casting alloys be conducted under production die casting conditions. Unfortunately, not much information of this kind has been retrieved during this literature review. Even the research that is directly related to die casting alloys is very limited. The relevant work conducted recently is presented as follows. For a better understanding of all Al base die casting alloys, much more work needs to be done.

5-5 Microstructure formed under die casting conditions

The microstructure may not be uniform in different portions of a casting, due to thickness variation and different distance from gates. This nonuniformity even occurs across a cross section. Suzuki, M. et al^[46] studied this phenomenon in the ADC12 Al alloy (11.82%Si and 2.06%Cu). It was found that the dendrite cell size of the die casting was less than 1/10 the size of that found in sand castings. The size was linearly smaller (larger?) from the surface toward the center of the casting. Eutectic Si was in spherical form (1 μ m dia.) in the layer of 0.05-0.1 mm from the surface, becoming acicular and larger toward the center. The axis-ratio was varied in the range of 3.4-13.8. The ratio of the crystallized alpha-Al area was rapidly increased with a depth of up to 0.5 mm from the surface, and it was kept nearly constant toward the center. In Al die castings, the surface area is called "chill zone." The surface layer having a microhardness (Vickers) > 105, corresponding to an alpha-phase content < 60%, was defined as the "chill zone."^[47] The chill zone was thinner in heavy-section castings. Casting parameters, such as metal temperature, injection pressure, and gate velocity, have little effect on the chill zone formation, but the die temperature had a significant effect.

Table 5-22 Phase Formed in Al-Si Alloys

Si	Si < 12% eutectic	Si > 12% eutectic and primary	Mg > 0.2% Mg ₂ Si	Fe > 0.05% FeSiAl ₃	Mn > 0.1% (FeMn) ₃ Si ₇ Al ₁₁	Cr > 0.1% (CrFe) ₄ Si ₄ Al ₁₁	Fe > Mg FeMg ₃ Si ₄ Al ₄	Mg > 2 Cu Cu ₂ Mg ₈ Si ₄ Al ₃
Na	Na < 0.01% dispersed	Na > 0.01% (NaAl)Si ₂						
P	Si < 12% dispersed	Si > 12%, P > 0.01% AlP, often inside Si crystals						
Fe	Fe < 0.7% FeSiAl ₃ eutectic	Fe > 0.7% FeSiAl ₃ primary	Co > 0.1% (CoFe) ₇ Al ₄	Cr > 0.1% (CrFe) ₄ Si ₄ Al ₁₁ or (CrFe) ₃ Si ₄ Al ₂	Mn > 0.2% (FeMn) ₃ Si ₇ Al ₁₁	Mo > 0.1% AlFeMo(?)	Ni > 0.1% FeNiAl ₄	Mg < Fe FeMg ₃ Si ₄ Al ₄
Mg	Mg < 0.2% in solid soln.	Mg > 0.2%, Cu < 0.2% Fe < Mg Mg ₂ Si	Cu < 1/2 Mg	Fe > Mg				
	Cu < 1%	Cu > 1%, Mg < 0.2% Fe < 0.3%	Cu > 1%, Mg > 2 Cu					
Cu	in solid soln.	CuAl ₂	CuMg ₈ Si ₄ Al ₃					
Co	(CoFe) ₇ Al ₄							
Cr	(CrFe) ₄ Si ₄ Al ₁₁ or (CrFe) ₃ Si ₄ Al ₂							
Mn	(FeMn) ₃ Si ₇ Al ₁₁							
Ni	FeNiAl ₉	NiAl ₃	(CuNi) ₂ FeAl ₇	(CuFeNi)Al ₆	(CuFeNi) ₂ Al ₃	Cu ₄ NiAl ₇	(FeNi)Al ₉	and FeMnNi compounds
Zn	in solid soln.							

Table 5-23 Eutectic Present in Al-Si Alloys. If there are eutectics containing Al-Si-Mg₂Si and Mn, Cr, Co, and Ni phases, their melting points should be in the range of 820-850 K. Zn in solution may lower the melting points of the eutectics by 1-2 K for every 1% Zn

Phases in eutectic	% Si	Composition of liquid			Temperature	
		% Fe	% Cu	% Mg	(°K)	(°C)
Al-Si	12.5				850	577
Al-Si-FeSiAl ₃	12	0.7			848	575
Al-Mg ₂ Si-Si	13			5	828	555
Al-CuAl ₂ -Si	5.2		26.7		798	525
Al-CuAl ₂ -Si-Cu ₂ Mg ₈ Si ₆ Al ₃	6		28	2.2	780	507
Al-FeMg ₃ Si ₆ Al ₈ -Mg ₂ Si-Si	12.9	0.15 (Fe + Mn)		4.9	828	555
Al-Si-(FeMn) ₃ Si ₂ Al ₁₃	11.7	0.75			846	573
Al-Mg ₂ Si-Si-(FeMn) ₃ Si ₂ Al ₁₃	12	< 1 (Co + Fe)		< 5	828	555
Al-Si-(CoFe) ₂ Al ₉	13	0.7 (Ni + Fe)			848	575
Al-Si-FeNiAl ₉	~12	~0.7 (Cr + Fe)			~848	~575
Al-Si-(FeCr) ₄ Si ₄ Al ₁₃	11	0.5			848	575

Table 5-24 Phase Formed in Al-Mg₂Si Alloys

Mg	Mg < 0.3% in solid soln.	Mg > 0.3% Mg ₂ Si	Cu > 0.1%, Si > Mg Cu ₇ Mg ₈ Si ₄ Al ₃	Fe > 0.5%, Si > Mg FeMg ₃ Si ₄ Al ₄			
Si	Si < 0.2% in solid soln.	Si > 0.2% Mg ₂ Si	Cu > 0.1% Si > Mg Cu ₇ Mg ₈ Si ₄ Al ₃	Fe > 0.5% Si > Mg FeMg ₃ Si ₄ Al ₄	Mg < 2 Si Fe > 1/2 Si Fe ₂ SiAl ₄	Mg < 2 Si Fe < Si FeSiAl ₃	Mn or Cr > 0.1% (FeMn) ₃ Si ₂ Al ₁₁ , or (CrFe) ₄ Si ₄ Al ₁₃
Fe	Mg > 2 Si FeAl ₃ or FeAl ₆ *	Mg ≤ 2 Si Fe ₂ SiAl ₄	Si > 2 Fe Si > Mg FeSiAl ₃	Si < Fe Si > Mg FeMg ₃ Si ₄ Al ₄	Mn or Cr > 0.1% Si < 0.6 Mg (FeMn)Al ₆ or (CrFe)Al ₇	Mn or Cr > 0.1% Si > Mg (FeMn) ₃ Si ₂ Al ₁₁ , or (CrFe) ₄ Si ₄ Al ₁₃	
Cu	Cu < 0.3% in solid soln.	Mg = 1.2-2 Si, Cu > 0.3% CuMgAl ₂		Si = Mg, Cu > 0.3% Cu ₇ Mg ₈ Si ₄ Al ₃			
Mn	Mg > 2 Si (FeMn)Al ₆	Si ≥ Mg (FeMn) ₃ Si ₂ Al ₁₁					
Cr	Mg > 2 Si (CrFe)Al ₇	Si > Mg (CrFe) ₄ Si ₄ Al ₁₃					
Zn	Zn < 0.5% in solid soln.						
Pb	PbMg ₇						
Bi	Bi ₇ Mg ₃						

* With rapid cooling FeAl₆ may form; homogenisation and subsequent work transforms it into equilibrium compounds.

Table 5-25 Possible Low-Melting Eutectics in Al-Mg₂Si Alloys

	Temp. (°K)	%Cr	%Cu	Composition			
				%Fe	%Mg	%Mn	%Si
Al-Mg ₂ Si	868				8.15		4.75
Al-Si	850						12.5
Al-Mg ₂ Si-Si	823				5		13
Al-Mg ₂ Si-(FeMn) ₃ Si ₂ Al ₁₃ -Si	< 823			< 0.5	< 5	< 0.5	~ 12
Al-Mg ₂ Si-FeMg ₃ Si ₆ Al ₈ -Si	822			0.15	4.9		12.9
Al-Mg ₂ Si-(CrFe) ₄ Si ₄ Al ₁₃ -Si	< 823	< 0.5		< 0.5	< 5		~ 12
Al-CuAl ₂ -Mg ₂ Si	787		31.5		3.9		2.3
Al-Cu ₂ Mg ₈ Si ₆ Al ₅ -CuAl ₂ -Si	780		28		2.2		6.0
*Al-Mg ₂ Si-Mg ₃ Al ₈	722				32		0.4
*Al-Mg ₂ Si-Mg ₃ Al ₈ -FeAl ₃	~ 722			< 0.1	~ 32		~ 0.4
*Al-Mg ₂ Si-(FeMn)Al ₆ -Mg ₃ Al ₈	~ 722			< 0.1	~ 32	< 0.1	~ 0.4
*Al-Mg ₂ Si-(FeCr)Al ₇ -Mg ₃ Al ₈	~ 722	< 0.1		< 0.1	~ 32		~ 0.4

* Under commercial conditions Mg₃Al₈ does not appear unless the excess Mg is at least 2%.

Metal flowline patterns may be present in Al die castings^[48]. The flowline patterns were the result of differences in the solidification structure. Their thicknesses ranged from 5 to 50 μm . Bright parts were rich in primary alpha crystals and dark parts in eutectic Si. The orientation was random.

Lin, W. et al's^[49] study showed that there were metastable phases and residual stress in Al die castings (ADC12 alloy) due to quick cooling. The metastable phases and residual stress caused instability of the die castings. They studied the mechanism of the formation of the instability and developed a two-stage or high temperature aging technique. It was claimed that the new technique could eliminate the dimension instabilization and was time and energy efficient.

Yamamoto, Y. et al^[50] studied metal flow and solidification behavior of Al die castings. It was found that when molten metal (ADC10 or ADC12) was poured into the shot sleeve of the die casting machine at 640 °C, its temperature lowered rapidly to its solidifying temperature and the metal was injected into the die cavity in a solid-liquid mixture state.

Nishi, N, et al^[51] investigated the solidification phenomena within a shot sleeve and formation of microstructures of the JIS-ADC10 alloy die castings with samples prepared using a 90 metric tone locking force cold chamber die casting machine. The microstructures of JIS-ADC10 alloy after die casting consisted of coarse alpha-Al phase with dendrite arm spacing $> 10 \mu\text{m}$, fine alpha-Al phase with dendrite arm spacing 5-10 μm and fine eutectic structures. The amount of coarse alpha-Al phase increased as the time interval, between when the molten metal was poured into the shot sleeve and the injection, increased and as the pouring temperature decreased. The coarse alpha-Al phase formed in the shot sleeve before injecting metal into the die cavity. Two other types of abnormal structures which solidified in the shot sleeve existed in the die castings microstructure in addition to the coarse alpha-Al phase. The first was a scattered structure which exhibited a planar interface with the matrix. The second was a massive structure which agglomerated coarse alpha-Al phase and coarse eutectic Si.

Iwahori, H^[52] studied the scattered structure and indicated that commercial Al die castings of the Japanese casting alloy ADC12 often included a number of scattered structures caused by destruction of the solidified layers in a metal sleeve at a shot. The

quantity of scattered structures was governed by the solidification behavior of molten metal in a metal sleeve at a shot and could be estimated from the heat capacity of molten metal in the sleeve/unit area calculated by using die-casting conditions.

Nishi, N. et al^[53] also studied the effects of the pressure during metal filling on the microstructure. It was found that with increasing injection velocity and/or temperature, the solidified microstructure became finer, and the measured mechanical properties (tensile and Charpy impact test) became superior. Extensive shrinkage microporosity and a coarse dendrite structure were present in die castings cast under low pressure. Kainov, V. M. et al^[54] studied the effect of the cooling rate on the microstructure and properties of Al-Si die casting alloy A12.

Awano, Y. et al^[36] found that the morphology of an AlFeSi compound changed from a needle-like to a Chinese script form by heating the melt to high temperature (i.e., by melt superheating without any Fe corrector, such as Mn in an Al-Si alloy). They investigated the factors affecting this phenomenon through the squeeze castings of the JIS AC4C (equivalent to 356) alloy with high Fe content. It was confirmed that, when the melt was superheated, the Fe compound in Chinese script form crystallized in the wide range of Fe and Si compositions, but otherwise, Fe must form the compound in the needle-like form. The AlFeSi compound in a Chinese script form crystallized through non-equilibrium solidification at a later period than the crystallization of the compound in the needle-like form and was insusceptible to the thermal history of the melt after superheating. In Al-Si-Mg alloy castings three types of AlFeSi compounds, in the needle-like form, indefinite, and Chinese script forms, were observed together with a small AlMgFe compound in various shapes. The addition of Mg inhibited the crystallization of the AlFeSi compound in a Chinese script form.

5-6 Primary Si particles and Al-Si eutectic structure

The morphology, size, and distribution of primary and eutectic Si can change markedly during solidification depending on conditions such as cooling rate, the presence of impurities, and the ease of nucleation. These changes usually alter the properties of the alloys. Because solidification in foundry practice is complex, so is the microstructure of the castings. Often more than one basic morphological type is present for each phase. There is disagreement over whether certain morphologies are primary or eutectic. These difficulties arise because the Si morphologies are highly sensitive to

solidification variables, such as growth rate, temperature gradient, and composition, and, consequently, they readily merge into one another. Moreover, there are differences in morphology identification and nomenclature. Nevertheless, there are nine basic Si morphologies that can be observed in Al-Si alloy castings. These are described briefly below:

- 1 *Star-like primary Si* is typical of untreated alloys and slow cooling rates. It appears under three subtypes, namely, *solid star*, *radial star* and *lamellar star*.
- 2 *Polyhedral primary Si*.
- 3 *Plate-like Si* has not been conclusively classified either as primary or eutectic Si.
- 4 *Dendritic primary Si* is found in Na-treated and Sr-treated alloys, which freeze at a comparatively high undercooling.
- 5 *Spheroidal primary Si* is typical of Na-treated alloys and high cooling rates.
- 6 *Feathery Si* is sometimes referred to as *skeleton*, *web*, *fishspine*, *lamellar*, *feathery dendritic Si* etc. It was suggested that this phase be regarded as eutectic Si, but the matter is still open. It is typical of high undercoolings and Si depletion.
- 7 *Y-like Si* is sometimes referred to as *angular Si*. It is classified as eutectic Si.
- 8 *Flake-like eutectic Si* is typical in unmodified and P-treated alloys. Sections appear wavy or as straight acicular. Very often the flake-like eutectic Si is observed in a wheatsheaf configuration.
- 9 *Fibrous eutectic Si* is typically found in modified alloys. It appears as round particles on a polished surface.

A variety of transitional and hybrid morphologies can be observed in the structure of castings, but these are all derivatives of the above nine basic types.

Primary Si particles in hypereutectic Al-Si alloys

The present work will not review the refinement of primary Si in hypereutectic Al-Si alloys. For information on that subject readers are referred to references[9, 55, 56, 57, 58, 59, 60, 61, 62, 63]. The present work intends to highlight only the recent development about primary Si study.

Ohmi, T. et al^[64] studied the effects of undercooling and cooling rate on the primary Si size of hypereutectic Al-Si alloys, Al-32%Si and Al-22%Si. Three cooling methods were used. One was a melt-filled graphite crucible with a small diameter, which was air cooled or immersed in a molten metal bath for slow cooling. The second was a wedge shaped Cu mold. In the third method the melt was directly injected into a Pb bath. The relationship between the cooling rate (R) and the degree of undercooling (T_m) was estimated on the basis of the classical nucleation theory. The results showed that the primary Si size (D) was reduced with an increase of T_m and an exponential relation between D and T_m was obtained. The mean grain size of primary Si of Al-22%Si was 0.7 times as large as that of Al-32%Si for the same cooling rate. The predicted cooling rates were ~ 500 K/s for Al-22%Si and ~ 590 K/s for Al-32%Si to obtain the mean grain size of primary Si of $10\text{ }\mu\text{m}$ that is the proper size for practical use. Another study^[65] indicated that by adding Al-5%P refiner and solidifying at cooling rates $5\text{--}100\text{ }^\circ\text{C/s}$, the mean grain size of primary Si can be reduced to $16\text{--}37\text{ }\mu\text{m}$. Kurdyumov, A. V. et al^[66] studied the Si segregation in hypereutectic Al-Si alloy pistons and found the endurance of the piston depended on the even distribution of the Si crystals through the casting. Khamani, M. S. et al^[67] studied Si distribution in Al-17%Si alloy and found that there was a significant change in density and Si distribution through the sample cross-section even for samples solidified with a high cooling rate. Ikegai, Y. et al^[68, 69] studied the macroscopic distribution of P in hypereutectic Al-Si alloys and the effects of various conditions on the P distribution. The macroscopic behavior was investigated by autoradiography, in that 200 or 80 ppi ^{32}P was added at $860\text{ }^\circ\text{C}$. The area with high P concentration can be detected easily by exposing the alloy to an X-ray photographic film. The pattern of the P distribution and the defects of the castings were found to be strongly related. It was also observed that the distribution of P depended on the form of P added in the alloy (Cu-8.4%P or hexametasodium phosphate flux). P density was high where Si precipitation density was high. For Cu-8.4%P, the P density was high at a cold shut. The distribution pattern was not influenced by the P concentration but by the cooling rate. Ohmi, T et al^[70, 71] developed a duplex casting process to refine primary Si particles in hypereutectic Al-Si alloys. In this process, two kinds of molten alloys with different compositions were cast in mold in sequence at a given interval. When the liquidus temperature of the second molten alloy was higher than the first one, the primary Si crystals were drastically refined. Furthermore, they were finer when the temperature of each molten metal at the mixing was lower. Under the most effective condition for refinement, the mean size of the primary Si crystals was refined to $\sim 40\text{ }\mu\text{m}$, while it

was 70-90 μm in the ingots produced by the conventional casting. It was also observed that the microsegregation of the primary Si was alleviated while the primary Si crystals were refined by the duplex casting process. This process was used to examine the modification of Al-22%Si and Al-22%Si-2.5%Mg alloys. The combination of Na-treated Al-12%Si or Na-treated Al-12%Si-5%Mg alloy as the first melt and P-treated Al-32%Si alloy as the second melt was used. In the conventional casting, the simultaneous addition of Na and P tended to coarsen primary Si, while refined primary Si and modified eutectic Si crystals were observed in the ingot obtained by the combined use of the duplex casting process and additions of Na and P. In the vicinity of the side surface of the ingots markedly refined primary Si crystals were observed, which showed the synergistic effect of the duplex casting process and P addition to the second melt.

Eutectic Al-Si structure

Abbott, T. B. et al^[72] studied the structure of fully eutectic Al-Si alloy castings. In their experiments, Sr-modified Al alloys containing 14-15% Si were cast with fully eutectic structures by using heated molds and high purity materials. In alloys containing the additional elements, Mg, Cu, or Ni, a distinct eutectic colony structure was evident, outlined by intermetallic compounds. At the edge of the castings, the eutectic colony structures and the Al grains (revealed by anodizing) were of a similar scale. Further from the mold walls, the eutectic colonies contained several Al grains. This was believed to result from the blocking of the growth of the Al phase by the Si phase. Savas, M. A. et al^[73] studied a commercial eutectic Al-Si alloy, ETIAL 140. It was found that the slowly cooled and unmodified castings contained coarse Si flakes. Rapid cooling and also the addition of Sr and Li master alloys reduced the eutectic interface spacing and refined the Si phase. It was noted, however, that the size of the intermetallic phase increased abruptly when the Sr content was $>0.04\%$. Unlike the Sr effect, the Li addition did not influence the intermetallic size significantly.

VI Properties of Al Die Casting Alloys

6-1 Mechanical properties

Tables 6-1, 6-2, 6-3 (a), and 6-3 (b) show the mechanical properties of various Al die casting alloys collected from different sources.

Effect of composition

It is obvious that compositions have a dominant effect on properties of Al die casting alloys. However, very little research work has been conducted on this aspect recently. From the literature searched in this review, only the following work is found to be relevant. Gasparyan, L. A. et al's^[22] study showed that approximately the same change in mechanical properties was produced by the introduction of Si $\leq 4\%$ or Cu $\leq 4\%$. The combined introduction of Cu and Si significantly increased the ultimate tensile strength and Brinell hardness number but sharply reduced the elongation. An alloy containing Cu 4% and Si 14% had a maximum ultimate tensile strength of 24 kg/mm² and a Brinell hardness number of 89 kg/mm² with an elongation of 2%. Hemon, Y.^[74] compared the mechanical properties of Al die casting alloys. In the experiments, cut-off samples of several die casting alloys were investigated to show the possibility of obtaining elongation better than those observed with A-S9U3 (equivalent to 380). In the conditions of test the maximum elongation was obtained with A-S5GT (AlSi4.6MgTi). This improvement in elongation was accompanied by a reduction in tensile strength of about 30%. Holecek, S.^[37] studied the effect of Fe on mechanical properties of Al-Si alloys. The alloys investigated were Na- and P-modified Al alloys containing 12-13%Si and 1.5-2%Cu. The result indicated that Fe ($\leq 2\%$) had a negative effect on the mechanical properties. This negative effect of Fe could be compensated by the addition of Mn, Cr, Mo, Co and/or V. The effect of the variation in composition of alloy LM24 (equivalent to 380) on mechanical properties was summarized as follows^[32]: Si (7.5-9.5%) increases high temperature strength and machinability whereas Cu (3-4%) increases hardness; Zn content $> 3\%$ gives rise to shrinkage deformation; Mg (0.3% maximum) allows a reduction in the stabilization annealing temperature; Fe (1.3% maximum) reduces the tendency of hot cracking and segregation; Mn (0.5% maximum) can reduce shrinkage and increase high temperature

Table 6-1 Typical Mechanical Properties of Aluminum Die Casting Alloys
(Listed in the ASM Metal Handbook Volume 15, 9th Edition, 1988)

Alloy			Temper	Tensile Strength, ksi (MPa)		Yield Strength (0.2% Offset), ksi (MPa)		Elongation in 2 in. or 50 mm, %		Hardness HB*	Shear Strength ksi (MPa)	Fatigue Strength (R. R. Moore Specimen), 500000000 Cycles, ksi (MPa)
ANSI	ASTM	UNS										
360.0	SG100B	A03600	F**	44 (300)	47 (324)	25 (170)	25 (172)	2.5	3.0	75	28 (190)	20 (140)
A360.0	SG100A	A13600	F	46 (320)	46 (317)	24 (170)	24 (165)	3.5	5.0	75	26 (180)	18 (120)
364.0			F		43 (296)		23 (159)	7.5	...			
380.0	SC84B	A03800	F	46 (320)	48 (331)	23 (160)	24 (165)	2.50	3.0	80.00	28 (190)	20 (140)
A380.0	SC84A	A13800	F	47 (320)	47 (324)	23 (160)	23 (159)	3.50	4.0	75.00	27 (190)	20 (140)
383.0	SC102A	A03830		45 (310)		22 (150)		3.50			...	...
384.0	SC114A	A03840	F	48 (330)	47 (324)	24 (170)	25 (172)	2.50	1.0		29 (200)	20 (140)
390.0	SC174A	A03900	F	40.5 (280)	40.5 (279)	35.0 (240)	35 (241)	<1	1.0	120.00	...	...
			T5		43 (296)		38.5 (265)	1.0	1.0			
B390.0	SC174B	A23900		46.0 (320)		36.0 (250)		<1			...	...
392.0	S19	A03920	F	42.0 (290)	42 (290)	39.0 (270)	38 (262)	<1	<0.5		...	...
413.0	S12B	A04130	F	43 (300)	43 (296)	21 (140)	21 (145)	2.5	2.5	80	25 (170)	19 (130)
A413.0	S12A	A14130	F	42 (290)	35 (241)	19 (130)	16 (110)	3.5	3.5	80	25 (170)	19 (130)
C433.0	S5C	A34430		33 (230)		14 (100)		9		50	19 (130)	17 (120)
443.0			F		33 (228)		16 (110)		9.0	50		
513.0			F		40 (276)	22 (152)	22 (152)		10.0			
515.0			F		41 (283)		...		10.0			
518.0	G8A	A05180	F	45 (310)	45 (310)	28 (190)	27 (186)	5	8.0	80	29 (200)	20 (140)

Note: The data written in shadow are those which are listed in ASTM B85.

* 500 Kg (1100 lb) load on 10 mm (0.4 in) ball; ** As-cast

Table 6-2 Mechanical Properties of Aluminum Die Casting Alloys
(Listed in the Metal Digest, and distinct from the AA registered alloys)

Alloy Designation	Published Time	Tensile Strength, ksi (MPa)	Yield Strength (0.2% Offset), ksi (MPa)	Elongation in 2 in. or 50 mm, %	Hardness HB*	Shear Strength ksi (MPa)	Fatigue Strength (R. R. Moore Specimen), 500000000 Cycles, ksi (MPa)
Al 218	10, 1953	45 (310)	27 (186)	8		27 (186)	23 (159)
Apex 418	8, 1957	45 (310)	27 (186)	8	80	28 (193)	23 (159)
Alsiloy No.1	11, 1957	39 (269)	21 (145)	2		25 (172)	19 (131)
Apex 36	6, 1958	45 (310)	24 (165)	3	75	27 (186)	19 (131)
Apex Z39	10, 1958	47 (324)	24 (165)	3.5	80	28 (193)	20 (138)
Al 13	11, 1958	43 (296)	21 (145)	2.5		28 (193)	19 (131)
Apex 39	1, 1959	47 (324)	24 (165)	3.5	80		
Birmal P.83	12, 1960	40.4 (279)	15.7 (108)	3	85	26 (179)	20 (138)
Apex 39-11	2, 1961	48 (331)	28 (193)	2.5	85	29 (200)	20 (138)
Apex Z39-11	2, 1963	47 (324)	24 (165)	2.5	85	28 (193)	19 (131)
Hi-Flo	12, 1963	48-51 (331-352)	21-23 (145-159)	5 - 6	74		18 (124)
Apex 28	5, 1964	45 (310)	24 (165)	4	80	26 (179)	18 (124)
Alcan 164	8, 1964	47 (324)	25 (172)	1		30 (207)	21 (145)
Alcan 143	11, 1965	47 (324)	23 (159)	4		30 (207)	20 (138)
Al A13	8, 1967	35-39 (241-269)	16-21 (110-145)	3.5-2.0		25-28 (172-193)	17-19 (117-131)

* 500 Kg (1100 lb) load on 10 mm (0.4 in) ball.

Table 6-3 (a) Typical Mechanical Properties of Aluminum Die Casting Alloys Used in Other Countries
(Listed in ASM Engineering Handbook, 1980)

Specification Number	Designation	Country *	Status	Tensile strength (MPa)	Yield Strength (MPa)	Elongation (%) in 2 in. (50mm)	Hardness HB
UNI 5076	GD-ALSi12Cu2Fe	Ita.	As cast	265	115	1	85-100
DIN 1725	GD-ALSi12Cu/3.2982.05	Ger.	As cast	220	140	1 **	60-80
UNI 5075	GD-ALSi8.5Cu3.5Fe	Ita.	As cast	215	145	1	85-105
NF A 57 703	A-S9 U3-Y4	Fra.	As cast	200		1	
SABS 992	Al-Si10Cu2FeA	S. Afr.	As cast	215	140	2	65-90
SABS 989	Al-Si10Cu2FeA	S. Afr.	Chill cast	145			
SABS 992	Al-Si8Cu4FeA	S. Afr.	As cast	230	110	1	85
SABS 989	Al-Si8Cu4FeA	S. Afr.	Chill cast	180		1.5	
NS 17 530	NS 17 530-05	Nor.	As cast	245	176	1	
NS 17 532	NS 17 532-05	Nor.	As cast	245	176	1	
SIS 14 42 54	SIS Al 42 54-10	Swe.	As cast	250	180	1	70-110
SIS 14 42 52	SIS Al 42 52-10	Swe.	As cast	250	180	1	70-110
ISO R 164	Al-Si8Cu3Fe	ISO					
DIN 1725	GD-ALSi8Cu3/3.2162.05	Ger.	As cast	240	160	0.5 **	80-110
DIN 1725	GD-ALSi10Mg/3.2983.05	Ger.	As cast	220	140	1 **	70-90
UNI 5074	GD-ALSi9MgFe	Ita.	As cast	195	145	2	70-90
NS 17 520	NS 17 520-XX	Nor.					
DIN 1725	GD-ALSi10Mg/3.2382.05	Ger.	As cast	220	140	1 **	70-90
DIN 1725	GD-ALSi6Cu4/3.2152.05	Ger.	As cast	220	150	0.5 **	70-100

*: Ita. - Italy, Ger. - Germany, Fra. - France, S. Afr. - South Africa, Nor. - Norway, Swe. - Sweden

** : in 4 in. (100mm)

Table 6-3 (b) Typical Mechanical Properties of Aluminum Die Casting Alloys Used in Other Countries
(Listed in ASM Engineering Handbook, 1980 and Smithells Metals Reference Book, 1983)

Specification Number	Designation	Country*	Status	Tensile Strength (MPa)	Yield Strength (MPa)	Elongation (%) in 2 in. (50mm)	Hardness HB
LM30		Eng.	As cast	275		1	120
ML16		Eng.	As cast	295		1	
NFA 57 703	A-S12-Y4	Fra.	As cast	170		2	
DIN 1725	GD-Si12/3.2582.05	Ger.	As cast	220	140	1 **	60-80
LM18		Eng.	As cast	205	110		
LM6		Eng.	As cast	270	145		
UNI 5079	GD-ALSi13Fe	Ita.	As cast	225	130	1.5	75-95
NS 17 512	NS 17 512-05	Nor.	As cast	196	147	2	
SABS 992	Al-Si12FeA	S. Afr.	As cast	205	115	1	70-90
NFA 57 703	A-59 G-Y4	Fra.	As cast	180		1	
SIS 14 42 47	SIS Al 42 47-10	Swe.					
UNI 5077	GD-ALSi5Fe	Ita.	As cast	175	100	2.5	50-65
LM24		Eng.	As cast	320	150	2	85
NF A 57 703	A-G10-Y4	Fra.					
DIN 1725	GD-ALMg9/3.3292.05	Ger.	As cast	200	140	1 **	70-100
SABS 992	Al-Mg8FeA	S. Afr.	As cast	195	115	1	60-80
UNI 5080	GD-ALMg7.5Fe	Ita.	As cast	195	120	1.5	60-80
NFA 57 703	A-G6-Y4	Fra.					

* : Ita. - Italy, Ger. - Germany, Fra. - France, S. Afr. - South Africa, Nor. - Norway, Swe. - Sweden, Eng. - England

** : in 4 in. (100mm)

strength, and when in excess, it gives rise to an Fe-Mn phase which is a drawback for machining; Ni (0.5% maximum), Sn (0.2% maximum) Pb (0.3% maximum) and Ti (0.2% maximum) have minor effects. Maglia, P. et al^[75] studied the effects of composition variation and die casting conditions on the mechanical properties of PN 424352 alloy (a Czech alloy, but the specification of this alloy was not found in references). The tested specimens indicated high values of Rm (room temperature strength), Rp0.2 (yield strength), and HB with very good reproducibility. However, ductility of die castings could not be guaranteed. The castings with higher contents of additions showed that yield increased by 8% and Rm by 2%. Koritta, J. et al^[76] (written in Czech) tested 42 experimental Al-Si-Cu casting alloys with variation in contents of Si, Cu, Fe, Mn, Mg, and Zn. Based on the mechanical properties of the tested alloys, they developed a regression mathematical model and determined relations for the optimum strength of alloys by a mathematical analysis in terms of higher impurity contents of Fe and Zn.

Besides these standard alloys, some patented alloys claimed to have special or improved properties. For example, Stroganov, G. B.^[77] patented a die casting alloy, containing 6-8%Si, 4.5-5.5%Cu, 0.3-0.8%Mg, 0.3-1.2%Ni, and 1.2%Fe. The alloy preferably contained up to 0.5%Cd for improving corrosion resistance. After low-temperature aging the mechanical properties were as follows: tensile strength 34-36 kg/mm², yield (0.2% proof) strength 32-34 kg/mm², elongation 2-4%, short duration strength 24-27 kg/mm², and long duration (prolonged) strength 7 kg/mm². Nishi, H. et al^[78] developed two dispersion strengthened Al die casting alloys, Al-5Ni-5Mg and Al-2Mn-4.5Mg. These alloys exhibited improved ductility, compared with the conventional Al-Si based die casting alloys, such as Al-8Si-3Cu (380). They were 11, 13, and 3%, respectively. The tensile strengths and the Charpy impact strengths of these alloys, compared with the 380 alloy, were 340, 310, and 300 MPa; 3, 6, and 1 x 10⁵ J/m², respectively. The reasons for the strengthening were claimed to be the fine uniform dispersion of needle-shaped Al₃Ni or Al₆Mn particles and the strengthened matrix by the Mg and/or Mn in solid solution.

Factors other than composition affecting mechanical properties

In addition to the composition, some other factors, such as casting design, and the operating parameters (melt temperature, mold temperature, pressure cycle time, and

feed channel characteristics), may also affect microstructure and, therefore, affect mechanical properties.

Effect of locations and dimensions

Pushmashev, P. I.^[79] studied the effects of dimensional parameters on the mechanical properties of Al die castings. The alloy tested contained 9.2%Si, 1.4%Cu, 0.65%Mg, 0.5%Mn, 0.2%Cr, 0.2%Ti, and 0.4%Fe. Casting temperature was 600 to 750 °C and die temperature was 50 to 250 °C. The properties were evaluated as functions of the parameter spr, which is the ratio of the area of the cross-section to its perimeter. The result indicated that tensile properties were inversely related to spr. However, as spr increased, the scattering of data decreased noticeably.

Susuki, M. et al^[80] studied the "mass effect" on various properties of Al-12%Si-2.5%Cu (JIS ADC12) die castings as a result of changes in wall thickness from 2 to 10 mm. It was found that the specific gravity, coefficient of thermal expansion, modules of elasticity, hardness, UTS, 0.2%YS, elongation, bending-strength, flexibility, compressive strength, compressive YS, compressive strain, shearing strength, impact strength, and fatigue strength were all dependent on the wall thickness. The wall thickness (t) was related to the material strength (S) by the general equation:

$$S = a \cdot t^{-n}$$

Values of a and n were given in their paper for each of the properties.

Susuki, M. et al's study^[81] also found that the segregation of Si, Cu, and Fe caused variations in specific gravity, coefficient of thermal expansion, and lattice constant between the chill zone and the inner parts of the die castings. The mechanical properties of the chill zone were 30-65% higher, due to the finer structure. The hardness and wear resistance were reduced by the formation of primary alpha-crystal on the chilled surface. The corrosion resistance of the chilled surface was superior to that of the inner parts in alkaline solutions, while the opposite behavior was observed for acidic solutions.

Hotta, S. et al^[82] studied causes for the scattering of strength of Al die casting alloys ADC10 and ADC12. The scattering of the tensile strength values was investigated on

specimens that were cut from different locations of Al die castings. It was found that the specimens cut from places where metal flow from the gate was $<150\text{-}200\text{ mm}$ did not contain defects. The strength of non-defect specimens was $200\text{-}350\text{ MPa}$ and was almost constant. But this value varied with the kind of castings or their thicknesses, corresponding to the DAS. The strengths at locations beyond the above-mentioned distance were lower, and the apparent specific gravities were also lower. These specimens contained defects and had strength of $<200\text{ MPa}$. These defects could be classified into lump, thin, and scattered structure defects. While the lump defects reduced both apparent specific gravity and tensile strength the thin and scattered structure defects reduced the tensile strength, but hardly affected the apparent specific gravity. The tensile strength varied almost linearly with the fraction of defect area on the fracture surface.

Effect of operating parameters

Czikel, J.^[83] studied the preparation of test bars for obtaining reliable data for die casting alloys. The influence of the position of the test bar in the mold, the bar density, and the casting time and temperature on the hardness, strength, elongation, and toughness of GD AlSi9Cu3 alloy was investigated. The study showed that data obtained by test bars prepared in various ways differed significantly. Czikel, J. et al.^[84] also tested die casting properties using a modified die for preparing specimens. In a common die, impact bending test and tensile test specimens were produced. The test bars were applied for static and dynamical examination in one single operation. Tests were conducted on the AlSi9Cu3 (equivalent to 380) alloy and included density, impact bending strength, mechanical properties and hardness, the proof stress relation, the tensile strength-hardness quotient, and the R value. It was found that pouring temperature and pouring rate affected properties. Density-impact bending strength and density-R value were found to be interrelated.

Nishi, N. et al.'s^[53] study showed that the pressure during metal filling affected microstructure and mechanical properties. With an increase in injection velocity and/or a rise in temperature, the solidified microstructure became finer, and the measured mechanical properties (tensile and Charpy impact test) became superior. Maglia, P. et al.'s^[75] experiments also showed that all the mechanical properties of the PN 424352 alloy were affected by a change of speed of filling piston and the temperature of the melt. Bratukhin, V. A. et al.^[85] studied the effects of additional pressure applied

during the solidification of die casting high-temperature Al alloys ATsR-IV and VAL8 on structure, porosity, and mechanical characteristics. It was found that porosity of the alloys decreased when they solidified under pressure and ductility of the alloys was improved.

Iwahori, H.^[52] experiments demonstrated that the scattered structures caused by the destruction of the solidified layers in the metal sleeve at shot seriously affected the mechanical properties. The scattered structure/normal structure interface had a tensile strength of only 9 km/mm². The scattered structure lowered the strength of commercial die castings and caused pressure leakage defects near the gate.

Effect of heat treatment

Jacob, S.^[86] studied the effect of heat treatment on elongation of Al die castings. In his experiments, cylindrical samples of 6.18 mm diameter were used and cast in a machine having an injection pressure of 110 MPa and closing force of 1200 KN. The alloy composition was 9.95%Si, 3.6%Cu, 0.5%Fe, 0.5%Zn, and 0.02%Ti. Samples were divided into three groups. Group I was untreated. Groups II and III were heat-treated for 4 hrs at 300 °C. Group III was cooled slowly in the oven while group II was removed after 4 hrs. The results showed a 40% improvement in elongation combined with a 20% decrease in rupture strength and Brinell hardness in both heat-treated groups of samples.

Fatigue properties The die casting process is known to be capable of producing very high fatigue properties. The fatigue properties are influenced by the die casting conditions. Oedegaard, J. A.^[87] reviewed static and dynamic mechanical properties in Al casting alloys and focused on the fatigue mechanisms. It was reported that three major microstructural parameters, porosity and shrinkage defects, DAS, and fraction, size and crystallographic orientation of intermetallic compounds, had significant effects on the fatigue properties for a wide range of Al casting alloys. For heat treatable alloys of low porosity, the heat treating conditions and thus the precipitate structure seemed to influence the fatigue behavior. The fatigue crack path was reported to run through the eutectic. Klein, F. et al^[88] studied the effects of the filling velocity, the final pressure during solidification, and specimen cross-sectional area in four-point bending fatigue on fatigue properties for the alloy GD- AlSi9Cu3 . The results showed a marked dependence of fatigue properties on die casting conditions. The dependence was not

only on fatigue, but also on the scattering of load cycles at a given load. Chijiwa, K. et al^[89] studied the fatigue fracture of Al die casting alloys. The fatigue crack growth was measured using CT specimens of the ADC12 alloy. The specimens were varied in thickness from 2.9 to 25.4 mm. The same test was carried out with heat treated die castings and casting specimens of alloy AC4C (equivalent to 356) specimens for comparison. The results indicated that in ADC12 both fatigue fracture toughness K_{fc} and the threshold stress intensity factor range K_{th} were independent of the thickness of CT specimens. Heat treated ADC12 caused an increase of K_{fc} , but little change in K_{th} . K_{fc} in ADC12 was the same as in the AC4C casting using a sand mold, but smaller than in the AC4C casting using a metal mold. The fatigue fracture of GK-AlSi12CuNiMg alloy was studied^[40] and it was found that the Fe content of the alloy had a pronounced influence. Egashira, H. et al^[39] investigated the effects of Si particle size, strength of the matrix, and addition of Mn and Fe on the fatigue crack propagation for high purity Al-Si casting alloys. The fine distribution of Si particles resulted in the retardation of the crack propagation rate (in the da/dN -K curve) in an Al-12%Si alloy. In Al-7%Si and Al-6.3%Si-0.32%Mg alloys the retardation of the crack propagation rate was pronounced in a high stress intensity range when the strength of the matrix was high. At the threshold region, however, the crack propagation rate was retarded in the matrix of lower strength due to the crack closure effect. The fatigue crack propagation rate (da/dN) in a high stress intensity range was retarded by the addition of Mn and Fe in an Al-Si-Mg alloy.

6-2 Other properties

Tables 6-4, 6-5 (a) through 6-5 (c) show the physical and process properties of some Al die casting alloys listed in different sources.

Electrical resistance Shikata, T. et al^[15] studied and developed an Al die casting alloy with high electrical resistance. This alloy was to be used in the cast rotor of a high starting torque motor used for hoist and machine tools. The effects of Cr, Mn, Ti, Fe, and Zr on electrical and mechanical properties of the Al-8.5%Si-3%Cu alloy were studied. From this study, an Al-8.5%Si-3%Cu-1.0%Cr-0.5%Mn-0.3%Ti alloy was obtained as a base alloy of high electrical resistance. The alloy had electrical conductivity of 22-24 % IACS, tensile strength of 18-20 kg/mm², and elongation of 2-3%, and showed good castability.

Table 6-4 Typical Physical Properties of Aluminum Die Cast Alloys
(Listed in ASM Metal Handbook, Volume 15, 9th Edition, 1988)

Alloy	Temper	Specific Gravity*	Density Kg/m3 (lb/in3)	Approximate Melting Range C (F)	Electrical Conductivity % IACS	Thermal Conductivity at 25 C (77 F) W/m . K	Coefficient of Thermal Expansion per C x 10E-06 (per F x 10E-06)	
							20-100 C (68-212 F)	20-300 C (68-570 F)
360.0	F**	2.68	2685 (0.097)	570-590 (1060-1090)	37	0.35	20.9 (11.6)	22.9 (12.7)
A360.0	F	2.68	2685 (0.097)	570-590 (1060-1090)	37	0.35	21.1 (11.7)	22.9 (12.7)
364.0	F	2.63	2630 (0.095)	560-600 (1040-1110)	30	0.29	20.9 (11.6)	22.9 (12.7)
380.0	F	2.76	2740 (0.099)	520-590 (970-1090)	27	0.26	21.2 (11.8)	22.5 (12.5)
A380.0	F	2.76	2740 (0.099)	520-590 (970-1090)	27	0.26	21.1 (11.7)	22.7 (12.6)
384.0	F	2.70	2713 (0.098)	480-580 (900-1080)	23	0.23	20.3 (11.3)	22.1 (12.3)
390.0	F	2.73	2740 (0.099)	510-650 (950-1200)	25	0.32	18.5 (10.3)	
	T5	2.73	2740 (0.099)	510-650 (950-1200)	24	0.32	18.0 (10.0)	
413.0	F	2.66	2657 (0.096)	575-585 (1070-1090)	39	0.37	20.5 (11.4)	22.5 (12.5)
A413.0	F	2.66	2657 (0.096)	575-585 (1070-1090)	39	0.37		
443.0	F	2.69	2685 (0.097)	575-630 (1070-1170)	37	0.34		
518.0	F	2.53	2519 (0.091)	540-620 (1000-1150)	24	0.24	24.1 (13.4)	26.1 (14.5)
A535.0	F	2.54	2547 (0.092)	550-620 (1020-1150)	23	0.24	24.1 (13.4)	26.1 (14.5)

* The specific gravity and weight data in this table assume solid (void-free) metal. ** AS-cast.

Table 6-5 (a) Die Casting and Other Characteristics (ASTM B85, 1991)
(Listed in 1992 Annual Book of ASTM Standards, Vol. 02.02)

Alloy			Approximate Melting Temperature Range, F	Die Casting Characteristics				Other Characteristics						Strength of Elevated Temperatures
ANSI	ASTM	UNS		Resistance to Hot Cracking	Pressure Tightness	Die Filling Capacity	Anti- Soldering To the Die	Resistance to Corrosion	Machin- ing	Polish- ing	Electro- plating	Anodizing (Appear- ance)	Chemical Oxide Coating	
360.0	SG100B	A03600	1035-1105	1	2	3	2	2	3	3	2	3	3	1
A360.0	SG100A	A13600	1035-1105	1	2	3	2	2	3	3	2	3	3	1
380.0	SC84B	A03800	1000-1100	2	2	2	1	4	3	3	1	3	4	3
A380.0	SC84A	A13800	1000-1100	2	2	2	1	4	3	3	1	3	4	3
383.0	SC102A	A03830	960-1080	1	2	1	2	3	2	3	1	3	4	2
384.0	SC114A	A03840	960-1080	2	2	1	2	5	3	3	2	4	5	2
390.0	SC174A	A03900	945-1200	4	4	1	2	3	5	5	3	5	5	3
B390.0	SC174B	A23900	950-1200	4	4	1	2	3	5	5	3	5	5	3
392.0	S19	A03920	1025-1245	4	3	1	2	2	5	5	3	5	5	3
413.0	S12B	A04130	1065-1080	1	1	1	1	2	4	5	3	5	3	3
A413.0	S12A	A14130	1065-1080	1	1	1	1	2	4	5	3	5	3	3
C443.0	S5C	A34430	1065-1170	3	3	4	4	2	5	4	2	2	2	5
518.0	G8A	A05180	995-1150	5	5	5	5	1	1	1	5	1	1	4

Note: Rating System - The various alloys are rated 1 to 5 according to the negative qualities in the listed categories.
A rating 1 gives the best performance, 5 the poorest performance.

Table 6-5 (b) Characteristics of Aluminum Die Casting Alloys
(Listed in ASM Metal Handbook, Vol. 15, 9th Edition, 1988)

Alloy	Resistance To Hot Tearing	Pressure Tightness	Fluidity	Shrinkage Tendency	Corrosion Resistance	Machinability	Weldability
360	1	1	2	2	3	4	
A360.0	1	1	2	2	3	4	
364	2	2	1	3	4	3	
380	2	1	2	5	3	4	
A380.0	2	1	2	5	3	4	
484	2	2	1	3	3	4	
390	2	2	2	2	4	2	
413	1	2	1	2	4	4	
C443.0	2	3	3	2	5	4	
515	4	5	5	1	2	4	
518	5	5	5	1	1	4	

Table 6-5 (c) Characteristics of Aluminum-Silicon Casting Alloys
(From J. E. Gruzleski's book "The Treatment of Liquid Aluminum-Silicon Alloys")

Alloy	Casting Method	Resistance To Tearing	Pressure Tightness	Fluidity	Shrinkage Tendency	Corrosion Resistance	Machinability	Weldability
319	S, P	2	2	2	2	3	3	2
332	P	1	2	1	2	3	4	2
355	S, P	1	1	1	1	3	3	2
A356.0	S, P	1	1	1	1	2	3	2
A357.0	S, P	1	1	1	1	2	3	2
380	D	2	1	2		5	3	4
390	D	2	2	2		2	4	2
413	D	1	2	1		2	4	4
443	P	1	1	2	1	2	5	1

Shrinkage Garber, L. W. et al^[90] investigated the effects of pouring temperature and solidification and cooling rate on density and shrinkage of the A380 alloy. The results indicated that pouring temperature did not affect density or shrinkage. Internal and external shrinkage volumes were found to decrease and increase, respectively, when the solidification and cooling rate were increased. The total shrinkage volume was equal to the sum of the internal and external shrinkage volumes and was found to be constant and independent of the solidification and cooling rate for a given weight of metal.

Die soldering Klein, K. et al^[42, 43] studied the die soldering tendency of alloy GD-AlSi9Cu3 in die casting. The reasons for the die soldering were summarized as follows: the shrinkage process in cooling the casting in the tool, the wedging of castings when opening the tool, poor strength of castings due to the high temperature prior to charging the tool, high material accumulation, the rapid opening of the closing unit, and the use of lateral cores and materials. The formation of erosion was only observed in casting the GD-AlSi7MgSb alloy but not in casting the GD-AlSi9Cu3 alloy. Due to the erosion caused by tool troubles, parts cast later remained attached to the die. By correctly arranging the ejectors the sticking could be reduced. At too high a melt flowing speed in the casting slot or at short charging times, the sticking effect was reduced because of the finer structure and higher strength of the castings. Also the melt modification by Na led to a reduction of the sticking effect. The steady Al sticking on certain tool parts was caused by Fe diffusion from the core into the boundary zone of the Al layer and the formation of the AlFeSi phase. The strong sticking effect was caused by the relatively homogeneous phase.

Wear resistance Savas, M. A. et al^[73] studied the control of mechanical properties and wear resistance of a commercial eutectic Al-Si alloy, ETIAL 140. It was found that the slowly cooled and unmodified castings which contained coarse Si flakes showed highest wear resistance and lowest UTS, elongation and hardness. Rapid cooling and also the addition of Sr and Li master alloys reduced the eutectic interface spacing and refined the Si phase. This generally corresponded to a significant increase in all properties except the wear resistance. It was noted, however, that the size of the intermetallic phase increased abruptly when Sr content was $>0.04\%$ which resulted in a sharp reduction in all mechanical properties. Unlike the Sr effect, the Li addition did not influence the intermetallic size significantly and, therefore, the mechanical properties were not impaired. In addition, the resistance also remained relatively

unaffected because the Li solid solution hardened the alpha Al dendrites. More about the property of wear resistance will be presented in the following.

6-3 Properties of hypereutectic Al-Si alloys

The 390 alloy is the most dimensionally stable Al die casting alloy at elevated temperatures^[91]. Its yield and tensile strengths are redundant up to 204 °C, which is approximately 42% of the final solidification temperature. At this temperature the alloy retains approximately 72% of the room temperature strength. Thermal expansion of the alloy is the lowest of the Al-Si die casting alloys and its thermal conductivity is relatively high. The alloy requires surface protection in corrosive environments. The high Si content detracts from its machinability.

Hanaoka, H. et al^[92] investigated the influence of surface condition on the wear resistance of hypereutectic alloys. Four specimens of the A390 (16%Si) alloy with Si precipitations extruded by 0, 1, 2, and 5 μm were prepared by etching the matrix phase. The wear was measured by a pin-disk method at a wear speed of 1.9 m/s and a pressure of 0.95 kg/mm². When the wear surface was not lubricated, the seizure of the matrix phase was a problem for the 0 μm extrusion specimen. For 2 and 5 μm extrusion specimens, Si particles fell off from the surface and the falling off of the particles increased with the wear rate. The study indicated that the 1 μm extrusion specimen showed the best resistance against wear and seizure. Ayel, M. J.^[93] conducted seizing tests, simulating the friction of the piston on the cylinder lining for various materials and surface treatments. The resistance to seizing and to wear, with and without lubrication, were studied. It was found that the Al-Si hypereutectic alloys behaved better than cast iron. Clegg, A. J. et al^[94] studied the influence of structural modification on the wear resistance of an Al-Si hypereutectic alloy. The wear tests were carried out on LM30 using a specially designed pin-on-hardened-steel ring machine. The melt was either without, or with Na, S, or Na+S addition, providing, respectively, eutectic Si modification, primary Si refinement, and a combination. The additions were 0.8% NaF or/and 0.5%S. The structural modification did not produce any significant difference in wear resistance. It appeared that the refinement of primary Si particles might well improve machinability but not significantly alter the wear resistance.

Yilmaz, F.^[95] studied the effects of growth velocity R , Si volume fraction V_f , and interparticle spacing on the hardness and tensile properties of directional solidified Al-Si alloys (14-17%Si). The relationship between the hardness and growth velocity in various Si morphologies was defined and shown to be of the same form as that for yield strength. It was found that, as a result of a high Si volume fraction, the hardness value in complex regular crystals was almost 30-35% higher than that in eutectic crystals for any given interparticle spacing. Krukovskaya, G. N.^[96] studied the morphology of primary Si in hypereutectic Al-Si alloys and correlated the mean dimension of primary Si particles, the type of linking of the polyhedra, and the mechanical strength of the alloys.

Yamamoto, Y. et al^[97] studied the differences in transverse strength at different positions in 390 alloy die castings by measuring the metal filling behavior in die cavities. A part in a die casting was found to have low transverse strength. This casting was die cast at slow injection velocity with a solid-liquid mixture which had a high volume fraction of solids and scattered structures. In this part primary Si easily segregated. It was also found that alloy 390 was die cast without fusion penetration difficulty.

Ohuchi, H.^[23] studied the effect of Cu and Mg on the thermal shock of the Al-20%Si alloy. Contents of Cu and Mg tested were up to 3% and 5%, respectively. The experiments were conducted on castings of 23 mm in diameter and 23 mm in height which were subjected to 500 °C for 20 min and then quenched into ice water. When the castings containing Mg were subjected to thermal shock, the occurrences of a number of microcracks resulted in porosity in the castings. The castings containing Cu < 0.5% deformed and cracks on their surfaces caused a slight decrease in apparent specific gravity. The alloy containing Cu 1% or more behaved similarly as the Mg-containing alloys. In general, the alloys containing such heat treatable alloying elements as Cu and Mg had low resistance to thermal shock.

VII Typical Aluminum Die Casting Alloys

7-1 360.0 alloy

360 type alloys are not used as commonly as 380 alloys in die casting. This is probably because 360 alloys have a more strict limitation on impurity contents, which increases the production costs, and their mechanical properties are not superior to 380 alloys. Probably, because they are not used very commonly, not much research work has been done on them recently. However, 360 alloys have good hot tearing and corrosion resistances and good elevated temperature strength, and this makes the alloys find applications in outboard motor parts, instrument cases, cover plates, and marine and aircraft castings. These alloys also have the potential to be heat treated for improving mechanical properties, which may expand the market for these alloys.

Composition

The 360.0 is an Al-Si-Mg alloy and its composition limits are shown in Table 7-1. The name of this alloy, 360.0, is used in Aluminum Association (AA) and American National Standard Institute (ANSI) Alloy designations and it is also the generally used trade name in the U.S. Its UNS (SAE-ASTM United Numbering System for Metals and Alloys) Number is A03600. It used to be designated as alloy SG100B in ASTM B85. This alloy also has other designations in the U. S., which include the following: Government QQ-A-591 class 12, SAE 309, and AMS 4290. Among these alloys the SAE 309 has the same specification as A360.0 with Fe content of 1.3%. The QQ-A-591 class 12 is specified to have less contents of Mn (0.30%), Sn (0.1%) and the total other elements (0.20%) than the 360.0. The alloys A360, 361, 364, and 368 can all be considered as members of the 360 family, due to similar compositions and properties.

Table 7-1 Composition limits of the 360.0 alloy

	Composition (%)									
	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ni	Sn	Other Total
360.0	9.0-10.0	2.0	0.6	0.35	0.40-0.60	--	0.50	0.50	0.15	0.25

Properties and operating parameters

Mechanical properties at room temperature

	<u>360.0</u> [98]	<u>360.0</u> [26]	<u>360.0</u> [10, 99]
Tensile strength (psi)	47,000	47,000	44,000
Yield strength (psi)	25,000	25,000	25,000
Elongation (%)	3.0	3.0	2.5
	(1/4" diam.)	(2" diam.)	(2" diam)
Hardness (HB)		75	
Shear strength (psi)	30,000		28,000
Fatigue limit (psi)	19,000		20,000
Modules of elasticity (psi)	10,300,000		
Modules of rigidity (psi)	3,850,000		
Poisson's ratio	0.33		

Tensile properties at elevated temperature

Probably because of fewer impurities in this alloy, which leads to a higher solidus temperature, this alloy was rated^[10] to have excellent elevated temperature strength. Typical tensile properties for separately cast test bars of 360.0 at elevated temperature are given in Table 7-2^[99].

Table 7-2 Typical tensile properties for separately cast test bars of 360.0 at elevated temperature^[99]

Temperature		Tensile strength		Yield strength		Elongation
°C	°F	MPa	ksi	MPa	ksi	%
24	75	325	47	170	25	3
100	212	305	44	170	25	2
150	300	240	35	165	24	4
205	400	150	22	95	14	8
250	500	85	12	50	7.5	20
315	600	50	7	30	4.5	35
370	700	30	4.5	25	3	40

Physical properties^[98]

Density at 68 °F (20 °C)	2.64 g/cm ³ (0.095 lb/in ³)
Liquidus temperature	1159 °F (596 °C)
Solidus temperature	1035 °F (557 °C)
Thermal expansion, $\mu\text{m/m K}$	

68 to 212 °F (20 to 100 °C) ----- 21.0

68 to 392 °F (20 to 200 °C) ----- 22.0

68 to 572 °F (20 to 300 °C) ----- 23.0

Specific heat at 212 °F (100 °C)	0.23 cal/g °C
Latent heat of fusion	93 cal/g
Thermal conductivity at 77 °F (25 °C)	0.35 cal/cm ² cm °C s
Electrical volume conductivity	37 % IACS
Electrical resistivity at 68 °F (20 °C)	6.16 microhm-cm
Radiography	

Type AA or similar film with no intensifying screen, cardboard exposure holder, 36-in focal distance, X-ray unit of 140-kv capacity, 2 H&D film density

Operating parameters

Melting temperature range	1200-1400 °F (649-760 °C)
Die casting temperature range	1175-1300 °F (635-704 °C)
Type of flux for remelting for casting	

Gaseous type, nitrogen or chlorine; or
solid type, aluminum chloride base

Joining

Rivet compositions; 6053-T4, 6053-T6, 6053-T61. Soft solder after copper plating, then use methods applicable to copper-base alloys. Resistance welding; flash method

Castability

Because it has a high Si content and small amounts of other elements (total maximum 4.95% excluding Si) comparing with other Al die casting alloys (e.g., total maximum 10.25% for 380.0 and 7.65% for 390.0 excluding Si), it has a fairly narrow freezing range (39 °C). That renders the alloy very good for hot tearing resistance and for pressure tightness. Fluidity of this alloy was rated good, as good as that of the 380 alloy, by Rooy, E. L.^[26]. This may be based on the consideration that the alloy has higher Si content, is closer to eutectic than the 380 alloy, and has a narrow freezing temperature range. However, in another reference^[10] it was rated not as good as the 380 alloy. This may be because that, although it has a narrow freezing range, its solidus (1035 °F, (557 °C)) is higher than other Al die casting alloys (e.g., 1000 °F (538 °C) for the 380.0 alloy), and this results in only a fairly good filling capacity. The spiral length for a 10%Si alloy is ~15 in for 1700 °F (704 °C) casting temperature and 22 in for 1400 °F (760 °C)^[100]. This alloy has good antisoldering properties because of the high Fe content.

Machinability

Fair due to high Si and low Cu, comparing with the 380 alloy.

Other properties

Low Cu content gives this alloy good corrosion resistance. The presence of a relatively large amount of Mg makes the alloy easily oxidized, forming oxide inclusions.

Property control

Within the composition limit, lowering Fe content increases ductility. For example, A360.0 (1.3%Fe) has higher elongation than 360.0 (2.0%Fe). Lowering Cu content may reduce strength slightly but improves corrosion resistance. Because of a relatively large amount of Mg, this alloy has a potential to enhance its tensile properties by heat

treatment. Klein, F.[101] studied this possibility. The heat treatment has been made possible due to a recent improvement in total venting of the die cavity by using a vacuum, or by changing the die casting process.

7-2 380.0 alloy

The 380 alloys are used as general purpose Al die casting alloys and are most commonly used. This is because these alloys have good castability and mechanical properties and have a generous limitation on impurities. Corrosion behavior of these alloys is not attractive due to relatively large amounts of Cu. Their large amounts of impurities result in low solidus temperature and may affect elevated temperature strength. Because they are most commonly used, they have received the most attention in Al die casting alloy research.

Composition

The 380.0 is an Al-Si-Cu alloy and its composition limits are shown in Table 7-3. The name of this alloy, 380.0, is used in AA and ANSI Alloy designations and is also the generally used trade name in the U.S. Its UNS number is A03800. This alloy has other designations in the U. S., which include the following: Government QQ-A-591 class 11, SAE 308, and AMS 4291. It used to be designated as alloy SC84B in ASTM B85. The alloys A380, B380, 383, 384, A384 and 385 can all be considered as members of the 380 alloy family, due to similar compositions and properties. In foreign countries there are a lot of Al die casting alloys (Table 7-4), that can be considered as variations of the 380 alloy.

Table 7-3 Composition limits of the 380.0 alloy

	Composition (%)									
	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ni	Sn	Other Total
380.0	7.5-9.5	2.0	3.0-4.0	0.50	0.10	--	3.0	0.50	0.35	0.50

Table 7-4 Different International Designations for the 380.0 Alloy Type
(From J. E. Gruzleski's book "The Treatment of Liquid Aluminum-Silicon Alloys" and
the Handbook of International Alloy Composition and Designation, Vol. III, 1980)

Country	Specification	Designation	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Ni	Pb	Sn	Zr	Others	
															Each	Total
International	SO DIS 352	AlSi8Cu3Fe	7.5-9.5	1.3	2.5-4.0	0.6			1.2							
Belgium	NBN 436	DG AlSi8Cu3Fe	7.0-9.5	1.3	2.5-4.5	0.6			1							
Canada	HA3	SC 84N	7.5-9.5	0.6	3.0-4.0	0.1	0.05		0.1						0.05	0.15
		SC 84R	7.5-9.5	0.70-1.2	3.0-4.0	0.5	0.45-0.75		0.70-1.2	0.05					0.05	0.15
Denmark	DS 3002	4254	7.5-10.0	1.3	2.0-4.0	0.5	0.3		3	0.2	0.3	0.3	0.2			
Finland	SFS 568	GAISI9Cu3Fe	7.5-10.0	1.25	2.0-4.0	0.5	0.3		1.2	0.2	0.3	0.3	0.2		0.05	0.15
France	NF A 57-703	A-S9U3A-Y4	7.5-10.0	1.3	2.5-4.0	0.5	0.3		1.2	0.2	0.5	0.2	0.2			
W. Germany	DIN 1725/2	G-AlSi8Cu3(226)	7.5-9.5	0.8	2.0-3.5	0.2-0.5	0-0.3		1.2	0.15	0.3	0.2	0.1		0.05	0.15
		GD-AlSi8Cu3(226D)	7.5-9.5	1.3	2.0-3.5	0.2-0.5	0-0.3		1.2	0.15	0.3	0.2	0.1		0.05	0.15
Italy	UNI 3601	G-AlSi8.5Cu3.5	7.5-9.5	0.8	3.0-4.0	0.5	0.05		0.1	0.1	0.1					
	UNI 5075	GD-AlSi8.5Cu3.5Fe	8.0-9.5	1.1	3.0-4.0	0.4	0.3		1	0.15	0.3	0.15	0.1		0.1	1.6
Japan	JIS H 5302	ADC 10	7.5-9.5	1.3	2.0-4.0	0.5	0.3		1		0.5		0.3			
	JIS H 5302	ADC 12	10.5-12.0	1.3	1.5-3.5	0.5	0.3		1		0.5		0.3			
Netherlands	NEN 6022	AlSi8Cu3	7.0-9.5	0.7	2.5-4.5	0.6	0.15		1	0.2	0.3	0.3	0.2			
Norway	NS 17530	AlSi9Cu3	7.5-10.0	1.3	2.0-4.0	0.5	0.3		1.2	0.2	0.3	0.3	0.2			
Spain	NE 38-203-	L2630	7.5-10.0	1	2.5-4.0	0.5	0.3		3	0.2	0.5	0.3	0.2			
Sweden	SIS 144251	4251	6.0-8.0	0.7	2.0-3.0	0.5	0.3		2	0.2	0.3	0.2	0.1			
	SIS 144252	4252	7.5-10.0	1.3	2.0-4.0	0.5	0.3		1.2	0.2	0.3	0.3	0.2			
Switzerland	VSM 10895	G-AlSi8Cu3	7.5-9.5	1.3	2.0-3.5	0.2-0.5	0.3		2	0.15	0.3	0.2	0.2		0.1	0.3
UK	BS1490	LM24	7.5-9.5	1.3	3.0-4.0	0.5	0.1		3	0.2	0.5	0.3	0.2			
USA	STM B179-8	A380.0	7.5-9.5	1.3	3.0-4.0	0.5	0.1		3		0.5		0.4			0.5

Solidification and microstructure

The solidification and microstructure of the 380 type alloys have been discussed in Section V. The DAS of the 380 alloy under die casting conditions is generally 5 to 10 μm [51]. The phases observed in 380 type alloys can be $\alpha\text{-Al}$ dendrites, $\text{Al}_{15}(\text{Fe}, \text{Mn})_3\text{Si}_2$ (brown Chinese script), Al_5FeSi (needles), Si (grey), Al_2Cu (pink particles), $\text{Al}_5\text{Cu}_2\text{Mg}_8\text{Si}_6$, and Mg_2Si (black). Under die casting conditions, the $\alpha\text{-Al}$ dendritic network always forms first. At low cooling rates, the second phase precipitates before the main eutectic reaction. The second phase can be $\text{Al}_{15}\text{-}$ or $\text{Al}_5\text{-}$, or possibly the two together. In the alloys with Si contents very close to eutectic composition, primary Si particles may precipitate as the second phase. However, under die casting conditions, such second phases generally do not form. Following the precipitation of the second phase or the formation of the dendritic network, the main eutectic occurs. In tandem and afterwards, the formation of various phases takes place in the interdendritic areas. The kinds and the fractions of the second phases and the phases formed together with and after Al-Si eutectic depend on the cooling rate as well as the composition.

When lowering the Si, Fe, and/or Mn contents, the pre-eutectic part of the alloy becomes larger and dendrites occupy a larger part of the total volume. When the Si content is high, for example, very close to eutectic, the dendrites are still the first phase to form under die casting conditions. With high Fe : Mn ratio (high Fe and/or low Mn), large $\text{Al}_5\text{-}$ particles form and become a dominant Fe-bearing phase. When the ratio is low, the $\text{Al}_{15}\text{-}$ will dominate as a Fe-bearing phase. When both Fe and Mn contents are high, large amounts of mix of $\text{Al}_{15}\text{-}$ and $\text{Al}_5\text{-}$ will precipitate as pre- and co-eutectic phases. If the Mg content is high in the 380 alloys, the Mg_2Si phase may form in addition to the formation of Si, AlFeSi , Al_2Cu and complex eutectics.

A study[46, 47] showed that the dendrite cell size of 380 (ADC12) die castings was more than $1/10 \times$ smaller than that of sand castings. The size was linearly smaller (larger?) from the surface toward the center of the casting. Eutectic Si was in spherical form (1 μm diam.) in a layer of 0.05-0.1 mm from the surface, becoming acicular and larger toward the center. The axis ratio was varied in the range of 3.4-13.8. The ratio

of the crystallized alpha-phase area rapidly increased with a depth of up to 0.5 mm from the surface, and then it was kept nearly constant toward the center. The surface layer, having a microhardness (Vicker) >105 and corresponding to an alpha-phase content $<60\%$ was defined as the "chill zone." The die temperature had a significant effect on the chill zone, and the chill zone was thinner in heavy-section castings. The nonuniformity of the microstructure occurred not only across a cross section but also among different sections of the castings due to the different thickness and distance from the gates. Generally, the structure is coarser at heavier sections and/or near the gates.

The casting parameters also affect the microstructure significantly. When the casting temperature is low, the molten metal starts to solidify in the shot sleeve before being injected into the die cavity, forming a coarse alpha-Al phase^[51]. When this metal is injected into the die cavity, it results in the castings, consisting of coarse alpha-Al phase with DAS $>10\mu\text{m}$, a fine alpha-Al phase with DAS $5\text{--}10\mu\text{m}$, and fine eutectic structures. The amount of the coarse alpha-Al phase increases as the time from pouring molten metal into the shot sleeve to injection increases and as the pouring temperature decreases. Two other types of abnormal structures which solidify in the shot sleeve may exist in the die casting microstructure in addition to the coarse alpha-Al phase. The first is a scattered structure which exhibits a planar interface with the matrix. The second is a massive structure which agglomerates the coarse alpha-Al phase and coarse eutectic Si. The scattered structure is caused by the destruction of the solidified layers in the shot sleeve during injection^[52]. Increasing the injection velocity and/or the pouring temperature results in a finer microstructure. Extensive shrinkage microporosity and coarse dendrite structure are present in die castings cast under low pressure. Metal flowline patterns may be present in 380 Al die castings^[48]. The flowline patterns are the result of differences in the solidification structure. Their thickness ranges from 5 to $50\mu\text{m}$. Bright parts are rich in primary alpha-crystals and dark parts in eutectic Si. The orientation is random. The melting history of an Al-Si casting alloy was found to affect the microstructure^[36]. Heating the melt to a high temperature can change the morphology of the AlFeSi compound from a needle-like to a Chinese script form.

Properties and operating parameters

Mechanical properties at room temperature

	380.0 ^[26]	380.0 ^[98]	380.0 ^[10, 99]
Tensile strength (psi)	48,000	48,000	46,000
Yield strength (psi)	24,000	24,000	23,000
Elongation (%)	3.0	3.0 (1/4" diam.)	2.5 ^[10] , 3.5 ^[99] (2" diam)
Hardness (HB)	80		
Shear strength (psi)		31,000	28,000
Fatigue limit (psi)		21,000	20,000
Modules of elasticity (ksi) x 10 ³		10.3	10.3 ^[99]
Modules of rigidity (ksi) x 10 ³		3.85	3.85 ^[99]
Poisson's ratio		0.33	

Tensile properties at elevated temperature

The elevated temperature strength was rated as 3, inferior to the 360.0 (rated as 1)^[10]. This is probably because of the large amounts of impurities in this alloy, which lead to a lower solidus temperature. The elevated temperature tensile properties for the 380.0 alloy are given in Table 7- 5^[102].

Table 7-5 Typical tensile properties for separately cast test bars of 360.0 at elevated temperature^[102]

Temperature		Tensile strength		Yield strength		Elongation
°C	°F	MPa	ksi	MPa	ksi	%
24	75	324	47	165	24	3.5
100	212	310	45	165	24	4
150	300	234	34	152	22	5
205	400	165	24	110	16	8
250	500	90	13	55	8	20

Physical properties:[98]

Density at 68 °F (20 °C)	2.71 g/cm ³ (0.095 lb/in ³) 2.74 g/cm ³ (0.098 lb/in ³)[99]
Liquidus temperature	1100 °F (593 °C) 1072-1056 °F (578-569 °C)[33]
Solidus temperature	1000 °F (538 °C) 894-865 °F (479-463 °C)[33]
Specific heat at 212 °F (100 °C)	0.23 cal/g °C
Latent heat of fusion	93 cal/g
Thermal conductivity at 77 °F (25 °C)	0.23 cal/cm ² cm °C s
Electrical volume conductivity	27 % IACS
Electrical resistivity at 68 °F (20 °C)	7.50 microhm-cm

Operating parameters

Melting temperature range	1200-1400 °F (650-760 °C)
Die casting temperature range	1175-1300 °F (635-704 °C)
Type of flux for remelting for casting	

Gaseous type, nitrogen or chlorine; or
solid type, aluminum chloride base

Joining

Rivet compositions; 6053-T4, 6053-T6, 6053-T61. Soft solder after copper plating, then use methods applicable to copper-base alloys. Resistance welding; flash method. Generally, it has poor brazeability and poor weldability.

Castability	380 type of alloys all have good castability
-------------	--

Other properties

Heat treatment

Conventional die castings are not heat treatable. However, they may be annealed or stress relieved. Annealing: for more ductility, heat to 500-700 °F, hold at the temperature for 4-6 hrs and cool in the furnace or in still air. Stress relieving: heat to 350-500 °F, hold at the temperature for 4-6 hrs and cool in air. Castings of the B380 alloy produced under pore-free die casting procedures (roughly analogous to vacuum technology) may be solution heat-treated for maximum strength and ductility without blistering.

Corrosion

Because of the presence of relatively large amounts of Cu content, the 380 type alloys generally have poor corrosion resistance.

7-3 390.0 alloy

In the last two decades, there has been a growing interest in the hypereutectic Al-Si alloys with 14 to 25% Si, such as the 390 alloy. These alloys are natural composites containing hard primary particles of non-metallic Si embedded in a eutectic matrix. Due to this unique microstructure, these hypereutectic Al-Si alloys possess outstanding wear resistance, a thermal expansion coefficient lower than that of other commonly-used Al alloys and not much higher than that of iron, good elevated temperature strength and hardness (Yield and tensile strengths are redundant up to 400 °F. At this temperature, which is about 42% of the final solidification temperature, the alloy retains about 72% of its room temperature strength), very good casting characteristics (fluidity, in particular), high thermal conductivity, (which is about three times that of grey cast iron), fairly good corrosion resistance, and an excellent strength to weight ratio. These properties make them suitable for use mainly in the automotive industry for the production of engine blocks, pistons, cylinders, pump components, pulleys and other parts. Other uses are in aircraft engines and consumer items, such as in outboard motors and chain-saw engines.

Composition

The 390.0 is the major hypereutectic Al-Si alloy, also containing Cu and Mg as alloying elements. Its composition limits are shown in Table 7-6. The name of this alloy, 390.0, is used in AA and ANSI Alloy designations and is also the generally used trade name in the U.S. Its UNS number is A03900. This alloy used to have the name of SC174A in ASTM B85. The alloys B390, 392, and 393 can all be considered as members of the 390 die casting alloy family, due to their similar compositions and properties.

Table 7-6 Composition limits of the 390.0 alloy

Composition (%)											
										Other	
	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ni	Ti	Each	Total
380.0	16.0-18.0	1.3	4.0-5.0	0.10	0.45-0.65	--	0.10	--	0.20	0.10	0.20

Solidification and microstructure

Solidification and microstructure of the 390 type alloys have been discussed in Section V. The microstructure of commercial hypereutectic Al-Si alloys has two major features: the primary phase and the eutectic. The primary phase consists of virtually pure Si crystals. The eutectic consists of an Al solid solution (of mainly Si, Cu, and Mg) and Si, which is referred to as eutectic Si. The alloys (both unmodified and modified) form eutectic cells and the eutectic Si is interconnected within the cells. In the eutectic cells, there may exist other phases, such as Al_5FeSi needles, $\text{Al}_{15}(\text{Mn}, \text{Fe})_3\text{Si}_2$, Al_2Cu , and $\text{Al}_5\text{Mg}_8\text{Cu}_2\text{Si}_6$. Alpha-Al haloes around both primary and eutectic Si crystals are frequently observed.

Properties and operating parameters

Mechanical properties at room temperature

	<u>390</u> [103]*	<u>390.0</u> [99, 104]**	
		<u>As-diecast</u>	<u>T5</u>
Tensile strength ksi (MPa)	40.5 (290)	40.5 (280)	43 (295)
Yield strength ksi (MPa)	35 (241)	35 (240)	38 (260)
Elongation (%)	< 1.0	1	--
Hardness (HB)***	120	120	
Shear strength ksi (MPa)			
Fatigue limit ksi (MPa)****	20 (138)	20 (140)	--
Modules of elasticity ksi (MPa)	11,900 (81,000)	11,800 (81,200)	

* Tensile properties are average values obtained from ASTM standard round die casting specimens, 1/4" diameter, produced on a cold chamber die casting machine

** Tensile properties and hardness are determined from standard cast-to-size tensile specimens 6.4 mm (1/4") diameter for die castings tested without machining the surface

*** 500 kg load, 10 mm diameter ball

**** RR Moore, 5×10^8 cycles

Tensile properties at elevated temperature

The elevated temperature strength was rated as 3, the same as alloy 380.0 but inferior to 360.0 (rated as 1)[10]. This is probably because of large amounts of impurities in this alloy, which lead to a lower solidus temperature. The elevated temperature tensile properties for this alloy are given in Table 7-7[103, 104].

Table 7-7 Elevated temperature tensile strength of alloy 390.0[103, 104]

Temperature		Tensile strength*		Yield strength		Elongation
°C	°F	MPa	ksi	MPa	ksi	%
24	75	276	40	241	35	<1.0
38	100	(195)	(28)			
95	200	(285)	(41)			
100	212	282	41	282	41	<1.0
150	300	255 (265)	37 (38)	255	37	<1.0
205	400	200 (210)	29 (30)	200	29	<1.0
250	500	131 (125)	19 (13)	110	16	2.0

* The data in () are taken from reference^[104], and rest of the data are taken from reference^[103].

Wear resistance

The resistance against wear and seizure of hypereutectic alloys was found to be better than that of cast iron. The surface of hypereutectic Al-Si alloys can be treated to gain better wear resistance. The treatment is to remove a small amount of Al matrix either by mechanical, chemical, or electro-chemical means to cause the primary Si crystals to stand out from the matrix. Experiments^[92] showed that without lubrication, 1 μm extrusion specimens had best resistance against wear and seizure. The wear resistance is also affected by the primary Si size and distribution. The mean size of primary Si of 10 μm was considered to be proper for practical use^[65]. Even distribution of the primary Si crystals renders the alloy the best wear resistance. However, other experiments^[94] indicated that the refinement of primary Si particles might not significantly alter the wear resistance.

Physical properties^[103]

Density at 68 °F (20 °C)	2.74 Mg/m ³ (0.099 lb/in ³) ^[103] 2.73 Mg/m ³ (0.099 lb/in ³) ^[104]
Liquidus temperature	1200 °F (549 °C)*
Solidus temperature	945 °F (507 °C)**
Thermal expansion, $\mu\text{m/m K}$	

68 to 212 °F (20 to 100 °C) ----- 18.5
18.0^[104]

Thermal conductivity at 77 °F (25 °C)	0.32 cal/cm ² cm °C s
Electrical volume conductivity	25 % IACS
Electrical resistivity at 68 °F (20 °C)	6.16 microhm-cm

* and ** For an alloy with 17.45%Si (B390.1^[33]), the primary Si was found to start precipitating at 1238-1225 °F (670-663 °C) and the last liquid solidified at 905-838 °F (485-448 °C).

Operating parameters

Castability

The 390.0 alloy has excellent castability. Fluidity in the binary Al-Si system is maximum at about 17%Si, which is the midpoint of the specified Si range of alloy 390.0^[103]. This is attributed to the high latent heat of Si, which is 4.5 times greater than that of an equivalent volume of Al^[17]. Solidification shrinkage and hot tearing are reduced at this Si level and casting pressure tightness is enhanced. Application of slow shot speed technology will significantly reduce air/gas porosity^[103]. The weak point of this alloy in the castability is its relatively poor shrinkage feeding characteristics^[105]. This may not be a problem for die casting because die castings usually do not use feeding. However, in another reference^[10] it is rated to have excellent fluidity, and its hot tear resistance and pressure tightness are rated poor.

Machinability

Since the introduction of hypereutectic Al-Si alloys to industry, machinability has been a great concern. Si and other hard phases act as abrasives and tend to reduce cutting tool life. Cu and Mg increase the alloy matrix hardness, improve the machined surface finish, and decrease the tendency of an alloy to build up on the cutting tool edge. Heat treatment acts in the same manner. Refinement of primary Si and modification of eutectic Si substantially improve tool life. It was reported^[105] that, compared to the unrefined condition, refinement of primary Si in the C89 alloy (20%Si) reduced tool wear on the order of 36% in die casting specimens. The machining problems, in terms of tool life, have been greatly reduced with the development of cutting tool materials and the design of new polycrystalline diamond tools and coated tools in particular. Proper cutting parameters and cutting fluids are also essential for prolonging tool life when machining the 390 alloy.

Other properties

Heat treatment

The 390 alloy is most often used in the as-cast condition. Due largely to its relative high Cu content, the alloy will respond to elevated temperature aging after casting without requiring a solution treatment (T5) resulting in an increase of approximately 20 - 3 ksi in tensile and yield strength and about HB 25. Moreover, the alloy 390, when used at elevated temperatures, tends to undergo a permanent dimensional change (growth) caused by the precipitation of Si from a solid solution. This growth can be stabilized by a T5 treatment.

Corrosion resistance

One of the advantages that Al alloys offer as a replacement for cast iron in engines is their vastly superior corrosion resistance. However, among Al alloys the relatively large amount of Cu in the 390 alloy detracts from corrosion resistance. Instances where corrosion may be an issue call for the use of protective organic or inorganic coatings.

7-4 413.0 alloy

The 413.0 alloy is a eutectic type Al-Si die casting alloy with 2%Fe. Due to its Si and Fe contents, this alloy has excellent castability, anti-soldering property, and good solidification shrinkage tendency. The low Cu content (<1%) gives the alloy very good corrosion resistance. But it has poor machinability and poor surface finish ability. It is used for castings where excellent castability and corrosion resistance are needed, such as large and complex-shaped castings and also thin-walled castings for pressure tightness and pressure-proof. Typical applications are meter cases and automobile fittings^[106].

Composition, Solidification and microstructure

The composition of the 413.0 alloy is given in Table 7-8. The name of this alloy, 413.0, is used in AA and ANSI Alloy designations and is also the generally used trade name in the U.S. Its UNS number is A04130. It used to have the name of S12B in ASTM B85. A413.0 also belongs to this type (eutectic) of Al-Si alloys. The only difference between the two alloys in the composition is Fe content, 2.0%Fe for 413.0 and 1.3%Fe for A413.0.

Table 7-8 Composition limits of the 413.0 alloy

	Composition (%)										
	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ni	Sn	Other	Total
413.0	11.0-13.0	2.0	1.0	0.35	0.10	--	0.50	0.50	0.15	0.25	

Solidification and microstructure of the 413 type alloys have been discussed in Section V. The solidification starts with the development of an alpha-Al dendrite network due to growth kinetics regardless of the actual composition, even with the Si content greater than eutectic. Then, primary Si crystals appear frequently in the structure at relatively low cooling rates, but under die casting conditions they do not form. During the eutectic reaction, the Chinese script Fe-bearing phase $Al_{15}(Mn, Fe)_3Si_2$ dominates over the needle phase Al_5FeSi at high cooling rates. The morphology and size of eutectic Si depend on the cooling rate and melt treatment.

Properties and operating parameters

Mechanical properties at room temperature

	413.0 ^{[26]*}	413.0 ^{[104]*}
Tensile strength ksi (MPa)	43 (296)	43 (300)
Yield strength ksi (MPa)	21 (145)	21 (145)
Elongation (%)	2.5	2.5
Hardness (HB)**	80	
Shear strength ksi (MPa)		25 (170)
Fatigue limit ksi (MPa)***		19 (130)

* Separately cast test bars at as-cast condition

** 500 kg load, 10 mm diameter ball

*** RR Moore, 5×10^8 cycles

Tensile properties at elevated temperature

The elevated temperature strength was rated as 3, the same as alloy 380.0 but inferior to 360.0 (rated as 1)^[10]. The data are given in Table 7-9^[33].

Table 7-9 Typical tensile properties for separately cast test bars of as die-cast 413.0 alloy at elevated temperature^[33]

Temperature		Tensile strength*		Yield strength		Elongation
°C	°F	MPa	ksi	MPa	ksi	%
-195	-320	360	52	160	23	1.5
-80	-112	310	45	145	21	2
-28	-18	303	44	145	21	2
24	75	295	43	145	21	2.5
100	212	255	37	140	20	5
150	300	220	32	130	19	8
205	400	165	24	105	15	15
250	500	90	13	60	9	30
315	600	50	7	30	4.5	35
370	700	30	4.5	15	2.5	40

Physical properties^[104]

Density at 68 °F (20 °C)	2.657 Mg/m ³ (0.096 lb/in ³)
Liquidus temperature	1065 °F (574 °C) ^{[33]*}
Solidus temperature	1035-1015 °F (557-546 °C) ^{[33]*}
Thermal expansion, μ -in/in °C	

	μ m/m K	μ -in/in °F
68 - 212 °F (20 - 100 °C)	20.4	11.3
68 - 392 °F (20 - 200 °C)	21.4	11.8
68 - 572 °F (20 - 300 °C)	22.4	12.4

Specific heat	936 J/kg K
Latent heat of fusion	389 kJ/kg
Thermal conductivity at 77 °F (25 °C)	121 W/m K
Electrical volume conductivity at 68 °F (20 °C)	31 % IACS
Electrical resistivity at 68 °F (20 °C)	55.6 n-ohm-m

* For an alloy with 11.4%Si, 0.4%Fe, 0.09%Cu, 0.18%Mn, 0.03%Mg, and 1.1%Zn, cooled at 0.3 to 5.0 °C/s

Operating parameters

Melting temperature	650 to 760 °C (1200 to 1400 °F)
Die casting temperature	635 to 705 °C (1175 to 1300 °F)

Castability

This alloy has excellent castability. Its fluidity, hot tearing resistance, pressure tightness, and anti-soldering are all rated 1 (the highest) among all Al die casting alloys^[10]. It is a most readily castable alloy. It is, however, subject to abrupt solidification when approaching the setting point and is less suitable for a cold-chamber die casting machine than the 360 alloy with a lower Si content.

Machinability

The 413 alloy is rather difficult to machine because of rapid tool wear caused by the high Si content and its tendency to drag.

Joining

Rivet compositions: 6053-T4, 6053-T6, 6053-T61

Soft solder: After Cu plating, then use methods applicable to Cu base alloys.

Welding: Highly suitable for flash welding. Inert-gas welding or oxyacetylene torch welding produce good results. Metal-arc welding is done with flux-coated electrodes using direct current and reverse polarity (work negative). No flux is required in the inert-gas shielded arc A. C. welding process.

Other properties

Heat treatment

The 413 alloy cannot be hardened or strengthened by any heat treatment. Ductility can be improved by heat treatment consisting of heating the die castings to a temperature between 500 to 700 °F (260 to 371 °C) for 4 to 6 hours and cooled in the furnace or still air.

Corrosion resistance

The 413 alloy exhibits high corrosion resistance under both ordinary atmospheric and marine conditions. Its resistance may be increased by anodic treatment. Exposure to strong alkaline solutions must be avoided. Strongly alkaline substances tend to dissolve the natural oxide coating of the casting and attack the metal. Mildly alkaline mixtures at room temperature present a less severe condition, but exposures of this type should be approached with caution.

7-5 C443.0 alloy

The C443.0 alloy is basically a hypoeutectic type Al-Si binary die casting alloy with relatively low Si and low Cu contents. Due to its relatively low Si content, this alloy has high ductility among Al-Si base die casting alloys and good machinability. The low Cu content (<0.6%) gives the alloy very good corrosion resistance. Though the low Si content results in a fairly broad solidification range (~105 °F or 58 °C), and thus less fluidity and a somewhat increased tendency to hot tearing, this alloy is readily die castable. It is used for castings where above-average ductility and excellent

corrosion resistance are needed. Typical applications are pipe fittings, marine fittings, and carburetor bodies.

Composition

The composition of C443.0 alloy is given in Table 7-10. The name of this alloy, C443.0, is used in AA and ANSI Alloy designations and is also the generally used trade name in the U.S.. Its UNS number is A34430. It also has designations as SAE 304, and Government QQ-A-591. It used to have the name of S5C in ASTM B85. This alloy is the only specified hypoeutectic type Al-Si binary Al die casting alloy.

Table 7-10 Composition limits of C443.0 alloy

	Composition (%)									
	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ni	Sn	Other Total
C443.0	4.5-6.0	2.0	0.6	0.35	0.10	--	0.50	0.50	0.15	0.25

Properties and operating parameters

Mechanical properties at room temperature

	C443.0 ^{[107]*}	C443.0 ^{[104]*}
Tensile strength ksi (MPa)	33 (228)	33 (230)
Yield strength ksi (MPa)	14 (97)	16 (110)
Elongation (%)	9.0	9.0
Hardness (HB)**	50	65
Shear strength ksi (MPa)	19 (131)	21 (145)
Fatigue limit ksi (MPa)***	17 (117)	17 (115)
Poisson's ratio		0.33
Elastic modules: ksi x 10 ³ (GPa)		
Tension		10.3 (71.0)
Shear		3.85 (26.5)

* Fe content is specified as 1.0% max.

** 500 kg load, 10 mm diameter ball

*** RR Moore, 5 x 10⁸ cycles

Tensile properties at elevated temperature

The elevated temperature strength is poor, rated as 5, the worst among all Al die casting alloys^[10]. The data are given in Table 7- 11^[104].

Table 7-11 Typical tensile properties for separately cast test bars of as die-cast C443.0 alloy at elevated temperature^[104]

Temperature		Tensile strength		Yield strength		Elongation
°C	°F	MPa	ksi	MPa	ksi	%
24	75	230	33	110	16	9
100	212	195	28	110	16	9
150	300	150	22	105	15	10
205	400	110	16	85	12	25
250	500	60	9	40	6	30
315	600	35	5	25	3.5	35
370	700	25	3.5	15	2.5	35

Physical properties^[104]

Density at 68 °F (20 °C) 2.69 Mg/m³ (0.097 lb/in³)

Liquidus temperature 1170 °F (630 °C)

Solidus temperature 1065 °F (575 °C)

Thermal expansion

	μm/m K	μ-in/in °F
68 - 212 °F (20 - 100 °C)	22	12.2
68 - 392 °F (20 - 200 °C)	23	12.8
68 - 572 °F (20 - 300 °C)	24	13.3

Specific heat (at 100 °C) 963 J/kg K

Latent heat of fusion 389 kJ/kg

Thermal conductivity at 77 °F (25 °C) 142 W/m K (as-cast)

163 W/m K (annealed)

Electrical volume conductivity at 68 °F (20 °C) 37 % IACS (as-cast)

Electrical resistivity at 68 °F (20 °C) 46.6 n ohm-m (as-cast)

41.0 n ohm-m (annealed)

Operating parameters

Melting temperature	650 to 760 °C (1200 to 1400 °F)
Die casting temperature	635 to 705 °C (1175 to 1300 °F)

Castability

Castability of this alloy is relatively poor. Its hot tearing resistance and pressure tightness are rated 3 and its fluidity and anti-soldering are rated 4^[10]. However, this alloy is readily die castable.

Machinability

Machinability of this alloy is favored by its relatively low Si content^[107]. However, in some references it is rated 4 and 5^[10], the worst among Al die casting alloys.

Joining

Rivet compositions: 6053-T4, 6053-T6, 6053-T61

Soft solder: After Cu plating, then use methods applicable to Cu base alloys.

Resistance weld: flash method

Other properties

Heat treatment

The C443.0 alloy cannot be hardened or strengthened by any heat treatment, due to low Cu and Mg contents. It is most often put into service in the as-cast (F) condition.

Corrosion resistance

Because of low Cu content the C443.0 alloy has very good corrosion resistance.

7-6 518.0 alloy

The 518.0 alloy is an Al-Mg binary die casting alloy with 1.8%Fe and low impurity contents. It has an excellent combination of high strength and corrosion resistance. It also has high ductility, excellent machinability, good polishing characteristics (can be brightly polished and anodized), good fatigue property (comparable to wrought Al alloys), low density, but poor castability. Its uses include marine fittings, ornamental hardware, outboard motor brackets, ornamental automotive parts, and other applications requiring the highest corrosion resistance but less complicated or intricate shape.

Composition

The composition of the 518.0 alloy is given in Table 7-12. The name of this alloy, 518.0, is used in AA and ANSI Alloy designations and is also the generally used trade name in the U.S. Its UNS number is A05180. Its Government designation is QQ-A-591. It used to have the name of G8A in ASTM B85. The other Al-Mg type die casting alloys include 515.0 and 516.0.

Table 7-12 Composition limits of the 518.0 alloy

	Composition (%)									
	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ni	Sn	Other Total
518.0	0.35	1.80	0.25	0.35	7.5-8.5	--	0.15	0.15	0.15	0.25

Solidification and microstructure

The solidification sequence starts with the development of an alpha-Al dendritic network. During cooling, in the interdendritic area solute is enriched, leading to an Al_3Fe phase and then to form Mg_2Si . Finally, a complex eutectic reaction completes the solidification.

The microstructure of the solidified alloy consists of alpha-Al dendrites with Al₃Fe needles, Mg₂Si black particles, and a light brown blocky Al₈Mg₅ phase in the interdendritic areas.

Properties and operating parameters

Mechanical properties at room temperature

	<u>518.0</u> ^[108]	<u>518.0</u> ^[104]
Tensile strength ksi (MPa)	45 (310)	45 (310)
Yield strength ksi (MPa)	28 (193)	28 (190)
Elongation (%)	5-8	5-8
Hardness (HB)*		80
Impact strength. Charpy V-notch J (ft lbf)		7 (9)
Shear strength ksi (MPa)	30 (207)	
Fatigue limit ksi (MPa)**	23 (159)	23 (160)
Poisson's ratio	0.33	
Elastic modules: ksi x10 ³ (GPa)		
Tension	10.3 (71.0)	
Shear	3.8 (26)	

* 500 kg load, 10 mm diameter ball

** RR Moore, 5 x 10⁸ cycles

Tensile properties at elevated temperature

The elevated temperature strength is relatively poor, rated as 5^[10]. This is probably related to its low solidus temperature. The data are given in Table 7-13^[108].

Table 7-13 Typical short time elevated temperature tensile properties of the 518.0 alloy^{[108]*}

Temperature		Tensile strength		Yield strength		Elongation
°C	°F	MPa	ksi	MPa	ksi	%
24	75	310	45	186	27	8
100	212	276	40	172	25	12
150	300	234	34	159	23	15
205	400	172	25	131	19	18
260	500	110	16	83	12	22

* After holding 1/2 hour at test temperature.

Physical properties^[108]

Density at 68 °F (20 °C)	2.57 Mg/m ³ (0.093 lb/in ³)	
Liquidus temperature	1150 °F (620 °C)	
	1134 °F (612 °C)*	
Solidus temperature	995 °F (535 °C)	
	819-803 °F (437-428 °C)*	
Thermal expansion		
	μm/m K	μ-in/in °F
68 - 212 °F (20 - 100 °C)	24.1	13.4
68 - 392 °F (20 - 200 °C)		13.9
68 - 572 °F (20 - 300 °C)		14.3
Thermal conductivity	96.2 W/m K	
Electrical volume conductivity at 68 °F (20 °C)	25 % IACS	

* For the 518 alloy with 0.10%Si, 0.16%Fe, 0.01%Cu, 0.01%Mn, 7.6%Mg, and 0.02%Zn, cooled at 0.2 to 6.0 °C/s^[33]

Operating parameters

Castability

Castability of this alloy is poor. Its hot tearing resistance, pressure tightness, fluidity, and anti-soldering are all rated 5, the worst among all Al die casting alloys^[10]. However, this alloy can be die cast readily by the cool-chamber high pressure method. It is difficult to cast by the air-machine or low-pressure method. It is impractical to use this alloy for making casting shapes of great intricacy.

Machinability and surface finish

Die castings of the 518.0 alloy require little machining because their surfaces are smooth and accurate. Light cuts and simple set ups are the rule when machining is required. This alloy has excellent machinability; in fact, it is free-cutting and is best machined with very sharp, smooth tools. The cut should be deep enough to prevent burnishing or rubbing the cast surface. When polishing is applied, high polish is needed in order to obtain a uniform appearance after anodizing.

Joining

It has a poor capacity for brazing or welding.

Other properties

Heat treatment

The castings of this alloy are normally used in the as-die-cast condition. This alloy cannot be hardened nor strengthened by any heat treatment. When machining- to extremely close tolerance, the castings should be stress-relief annealed by heating to 350-500 °F for 4-6 hours, then cooled in still air.

Corrosion resistance

The 518 alloy has excellent resistance to corrosion and discoloration. Its corrosion resistance is rated 1, the best among all Al die casting alloys. For added surface protection, the finish-machined castings can be treated by one of a number of specialized surface treatments for Al alloys. However, it is difficult to obtain a uniform appearance after anodizing.

VIII Conclusions and Discussions

- 1 A large effort has gone into the research and development of aluminum alloys. The work has resulted in an ever increasing number of alloys, alloy standardization, a better understanding of the fundamentals of aluminum alloy, a great improvement in properties, and broadening the applications of aluminum alloys. Research on aluminum and aluminum alloys has been accelerated in the last two decades, this is reflected in the increasing number of papers published each year.
- 2 As an important segment of aluminum alloys, aluminum die casting alloys have also seen remarkable development since the cold-chamber die casting machine was invented. During the last half century, numerous Al die casting alloys were created. In the Handbook of International Alloy Compositions and Designations, Vol. III, Aluminum, 1980, more than 300 aluminum alloys were recorded worldwide for die castings (not including ingot alloys). In American Aluminum Association (AA) standardized aluminum alloys, 23 are for die casting. These cover almost all the existing Al die casting alloys commercially used worldwide in terms of compositions.
- 3 These 23 AA issued aluminum die casting alloys can be divided into several groups. A representative alloy from each group is listed below.

360.0 is an Al-Si-Mg system alloy. It is also a general purpose alloy. This alloy has very good castability, good corrosion resistance, but fair machinability, less generous impurity limitations, and slightly lower tensile strength and elongation than 380.0.

380.0 is an Al-Si-Cu system alloy. It is generally considered the best "all-purpose" die casting alloy and is the most commonly used. This alloy has very good castability, good mechanical properties, and generous limitations on impurities, however, it has only fair corrosion resistant properties.

390.0 is a hypereutectic Al-Si-Cu-Mg system alloy. It has outstanding wear resistance, low thermal expansion coefficient, high thermal conductivity, elevated temperature strength and hardness, very good castability, but fair corrosion resistance, low ductility and poor machinability. It is mainly used to replace grey cast iron in engine parts and as wear resistant components.

413.0 is basically an eutectic Al-Si binary alloy. It has excellent castability, very good corrosion resistance, but poor machinability and poor surface finish ability. It suits castings with large, complex shape, thin wall and good corrosion resistance.

C443.0 is basically a hypoeutectic Al-Si binary alloy. It has high ductility, very good corrosion resistance, good machinability, but fair castability, low strength. This alloy is used for the castings, where above-average ductility and excellent corrosion resistance are needed.

518.0 is an Al-Mg system alloy. It has excellent corrosion resistance, high strength, high ductility, excellent machinability, good polishing characteristics, good fatigue property, low density, but poor castability. This alloy suits the castings requiring the highest corrosion resistance with no complicated or intricate shape, also ornamental parts and hardware.

- 4 In all the Al die casting alloys the main alloying elements are Si, Cu, Mg, and Fe, and the common impurities are Mn, Ni, Cr, Zn, Sn and Pb. The Al die casting alloys may also contain melt treatment elements, such as Ti, B, Na, Sr, Sb and P. Their major effects on the aluminum die casting alloys are as follows:

Si imparts good castability, increases strength, reduces ductility, but does not affect corrosion resistance.

Cu increases hardness and strength, but reduces corrosion resistance.

Mg makes Al-Mg alloys have bright surface, high ductility, and excellent corrosion resistance, but poor castability. In Al-Si alloys Mg associated

with Si increases strength through heat treatment, but may form oxides and intermetallic inclusions.

Fe is generally the major impurity in Al casting alloys. It reduces ductility and may form intermetallics with other elements and cause hard spots. However, in Al die casting alloys Fe is added as an alloying element to reduce die soldering.

Mn, Co, Ni, Mo, and Be can act as Fe correctors to reduce the Fe embrittlement effect.

Mn is a commonly used Fe corrector, but it may form primary intermetallics with Fe and Si as hard spots.

Ni enhances elevated temperature properties.

Cr slightly increases strength, but may form sludge.

Pb and Sn improve machinability of Al-Si alloys.

Zn has no significant effect.

Be reduces oxidation and the formation of inclusions in Mg-bearing alloys.

5 Though a great deal of research and development work has been conducted on Al alloys, relatively little work has been done on die casting alloys, especially in the U. S. For example, searching Metal Abstracts for the past ten years more than 30,000 papers were collected under subtitle "Aluminum Base Alloys." However, only about 1000 papers were directly associated with aluminum casting alloys. Of the 1000 papers, about 300 were directly related to die casting, from which about 200 were technical and almost half of them were written in foreign languages, mostly German, Japanese, Russian, or French.

6 The research and development work on Al die casting alloys conducted thus far is scattered and does not provide sufficient information to systematically evaluate and optimize the alloys. The information provided also does not fully reflect the

progress of the die casting techniques, the advances in quality control of alloys and castings, or reflect the accumulated knowledge about the alloys. For example, for the major Al die casting alloys, such as 380.0, 360.0, and 518.0 (former name 218), the compositions and the property data found in current references are almost the same as those published in the 1950's in Metal Digest. During the four decades, progress in die casting techniques, alloy treatment and quality control have resulted in considerable improvement in the casting quality and properties of these alloys, but these improvements are not reflected in the current literature. Casting designers and producers have for long had questions about the scientific basis for setting the contents of alloying and impurity elements in aluminum die casting alloys. Though the effects of these elements, individually and in combinations, on microstructure and properties have been repeatedly investigated and the results have been published, questions still remain. Take 380.0 as an example, the upper limit of the total alloying and impurity element contents (20.45%) differs from the lower limit (total 11.8%, assuming the lower limit for Fe is 1.3%) by about 8.65%. A variation of alloying and impurity elements in an aluminum alloy by 8.65% will make a considerable difference in the microstructure and properties of the alloy. However, only one value for each property is given in the current standards. Is it the lowest, mean or average value? If it is the lowest, which composition does the value correspond to? If it is mean or average, in what range do the properties vary within the composition limits? More importantly, what is the optimum composition to meet certain requirements (i. e. certain properties)? Which element(s) is critical and which can be allowed to vary in a wider range for certain requirements? All these questions need to be answered.

- 7 Some of the data and information currently available can not be trusted by die casting designers and producers. For example:

On many occasions the data from different sources, or even in the same reference, may be conflicting. For example, the tensile strength of 380.0 alloy is shown as 48 ksi in the ASM Metal Handbook, as 46 ksi in ASTM B85, and as 43 ksi in Metal Digest (July 1982, Al-239). In Metal Digest (April 1983, Al-245), the room temperature properties of 360.0 alloy are tensile strength 44 ksi, elongation 2.5%, but in the same reference the data listed in typical short-time elevated temperature tensile properties show that at 75 °F (24 °C) for 30 min the

tensile strength is 47 ksi and the elongation is 3%. Is this alloy so temperature sensitive? The testing standard does not set strict temperature requirement for the room temperature property test. What is the room temperature? Can the properties be used for other temperatures, for example, 16 °C?

Accordingly, designers generally use about 65-75% of the test specimen properties in their calculations. This results in only very few aluminum alloy castings, including die castings, are currently used in sensitive applications, where high reliability is required.

- 8 The 65 - 75% factor used by designer is a reflection of their mis-trust in the published property data. This mistrust is brought about, not only by the contradicting published values explained above, but also by the fact that the designer believe, and rightfully so, that the properties of a casting will vary with the mold design, the production conditions, the operating parameters, and variation in the alloy chemistry within the prescribed limits. Test specimens, which are the source of the published property data, are usually produced under controlled conditions, and so do not reflect the effect of these parameters. Several attempts were made to relate test specimen properties to actual casting properties; however, these few studies did not result in generally acceptable relationships.
- 9 A systematic study of composition-microstructure-property relationship of Al die casting alloys is therefore necessary for the Al die casting industry and would have great technical and economical significance. Designers need reliable property data to evaluate casting performance. Alloy and casting producers need to understand how composition and casting conditions affect the properties and how to optimize the composition and production conditions for certain requirements. The systematic study will provide this information and data and will help build the foundation for the production of more reliable and lighter aluminum die casting components while reducing production costs, and opening new markets for aluminum alloy die castings.

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**Technical Report
for
The Aluminum Die Casting Alloy Research Project**

**A STUDY OF
ALLOY-MICROSTRUCTURE-PERFORMANCE
INTERACTION**

**PHASE I
A LOW RESOLUTION STUDY OF 380 ALLOY**

Submitted to
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1 Introduction

As part of Phase I of the Aluminum Die Casting Alloy Research Project a low resolution study was performed on alloy 380. The objective of this study is two folds:

- To investigate the alloy-microstructure-performance interaction, and
- To determine whether or not a more comprehensive study is needed.

The low resolution study was performed on 380 alloy since this is the most commonly used aluminum die casting alloy. Two compositions representing the two extreme limits of the alloy chemistry were chosen. The effects of the alloy chemistry on its microstructures and its mechanical properties were evaluated and documented.

2 Mechanical Property test

2-1 Specimen Preparation

Wabash alloys, Gibbs Die Casting Corp, and Doehler Jarvis produced and inspected the alloy and specimens. All expenses for specimen preparation are their donation to this project.

Wabash Alloys produced the two compositions of the alloys. The compositions of the alloys produced and the alloy specification are shown in Tables I and II. 1709 lbs of the L alloy (composition at the lower limits of the alloy specification) and 1721 lbs of the H alloy (composition at the upper limits of the alloy specification) were produced. The total cost of the alloys was \$6860.00.

Gibbs Die casting Corp die cast one hundred test bars, fifty for each composition. The test bars, shown in Fig. 1 were a modification of the standard tensile test specimen -- ASTM B557 - 84. The modification is in the grip sections of the specimen. These sections were made flat instead of round. The flat surface was used for hardness tests.

Doehler Jarvis x-ray inspected all the test bars. Eight specimens of alloy L and four specimens of alloy H were found to have defects within their reduced section. Eighty test bars, forty for each composition, were sent to WPI for testing and analysis.

2-2 Testing

Testing Procedure

Mechanical property tests were conducted at WPI. An Instron testing machine, Model 8502 was used. The testing procedure followed the ASTM standard B 557 - 84, "Standard Methods of Tension Testing Wrought and Cast Aluminum- and Magnesium-Alloy Products". Tensile strength, yield strength, and elongation were measured. A chart recorder was used to record and display load-displacement diagram. Two typical curves, one for each composition, as displayed on the chart record are shown in Fig 2. The X and Y axes are displacement and load, respectively. The data of load vs displacement were also stored in a computer for analysis. Forty specimens for each composition were tested. Due to apparent defects some specimens showed very low properties and some curves did not show an initial straight line (elastic deformation range) to allow property calculation. Therefore, the data from only 35 specimens of each composition were used in the property calculations.

The tensile strength (TS) was calculated by dividing the maximum load by the original cross-sectional area of the reduced section of the specimen. The load value at fracture is the maximum load for the specimen. In the testing machine this maximum value is automatically stored in its computer operating system and displayed. The maximum load can also be calculated from the curve of load vs displacement displayed on the chart or stored in the recording computer. The maximum load stored in the machine's computer operating system was used in the TS calculation in this study, which will be reported in the following section: Results. The as-die cast specimens used were not perfectly round. The dimensions of the cross sectional area also slightly varied from specimen to specimen. The maximum and minimum diameters at the center of the reduced section were measured for each specimen. The average of the maximum and minimum diameters was used as the diameter for determining the original cross sectional area of the specimen.

The elongation is the increase in length of the gage length, expressed as a percentage of the original gage length. The original gage length of 2.000" was carefully measured and marked. The increase in length of the gage length was determined by carefully fitting the ends of the fractured specimen together and measuring the distance between the gage marks. The elongation can also be calculated based on the curve of load vs displacement. In this method the increase in length (plastic extension) is estimated by subtracting the elastic extension from the total extension at fracture. This requires that the curve shows a clear initial straight line, which represents the elastic deformation of the specimen. However, many curves in these experiments did not show perfect initial straight lines or the curves had two connected straight lines with perturbation at the beginning (see Fig. 2), probably caused by the realignment of the machine. Thus, determining the plastic extension in this way would involve a degree of subjectivity. In addition, the plastic extension, determined this way, occurs presumably on the whole reduced section, and not in the gage length only. The reduced section of the specimen is designed as 2 1/4", but is difficult to measure. Therefore, the calculation based on the curves was only used as a check on the results from the direct measurements, and showed good agreement.

Yield strength was determined by the "offset method" at an offset of 0.2%. In this method a straight line is drawn on the stress-strain diagram parallel to the initial straight line on the curve of stress vs strain. This line is placed at a distance of 0.2% of the length of the reduced section from the initial straight line in the direction of the strain axis. The stress at the point, where the straight line drawn and the stress-strain curve intersect, is the yield strength. In these experiments the load vs displacement curves showed two straight lines at the beginning of loading, and the first line was shorter than the second. Initially, the calculations were carried out on both lines to estimate the yield strengths. It was found out that the values were very low based on the first lines. They were less than 10 ksi for both compositions, which is much lower than the specification, 23 ksi. Moreover, the lengths of these first lines varied significantly, comparing with their total lengths. The occurrence of the two lines may be caused by the machine realignment and the first short line was usually ignored. In these experiments the yield strengths were calculated based on the second line, which showed reasonable agreement with the specification and had a relatively narrow variation.

Hardness tests were carried out on a Rockwell Hardness Tester, Model 3JR, Number 10661. Rockwell Hardness Scale B (load 100 kg, 1/16" ball) was used. 20 specimens for each composition were tested. Ten measurements were taken on each specimen, five on each end. The measurements were conducted on the centerline of the flat areas in the modified portions of the specimen. The first and the last measurements were about 3/8" from the tips of the specimen, and the remainder of the measurements were about 3/16" apart.

2-3 Results

Calculated Property Data

The tensile property and hardness values obtained in this study are shown in Table III. The Standard Brinell Hardness given in the Table was converted from the Rockwell Hardness Scale B measurements. Typical properties, as listed in different references, are also shown in Table III for comparison. Standard deviation for each property value is given.

Error Estimation and Analysis

The error in the diameter measurements of the reduced section of the specimens was ± 0.001 in. This measuring error translates to an error of $\pm 0.8\%$ in the cross sectional area of the specimen. The load error in the machine reading is estimated as maximum ± 10 lbs, which may introduce a maximum error of $\pm 0.5\%$ (minimum load in this experiment was ~ 2000 lbs). As a result of the measurement errors (cross sectional area and load errors) the tensile strength may have a maximum error of $\pm 1.3\%$. The standard deviations of the measured tensile strengths are $\pm 2.1\%$ ($= \pm 1.04/50.17$) and $\pm 3.3\%$ ($= \pm 1.68/44.94$) for H and L alloys, respectively.

The maximum measuring error in the gage lengths, original and after fracture was ± 0.002 ". This value may introduce maximum errors of 0.1% and 5.0% for the original gage length (2.000") and the increase of the gage length after fracture (minimum of 0.040), respectively. This may lead to a maximum error of 5.1% for the elongation. The standard deviations of the measured elongations are $\pm 11.0\%$ ($\pm 0.28/2.55$) and $\pm 13.2\%$ ($\pm 0.85/6.46$), for H and L alloys, respectively.

The error in yield strengths is estimated to be $\pm 3\%$, based on the load and drawing errors. The standard deviations of measured yield strengths are $\pm 4.6\%$ ($\pm 1.48/32.17$), and $\pm 9.4\%$ ($\pm 1.91/20.42$) for H and L alloys, respectively.

The hardness was checked with standard specimen for ten measurements and the error was $\pm 2\%$. The standard deviations of the hardnesses in the results are $\pm 5.6\%$ ($= \pm 3.5/62.3$) and $\pm 13.5\%$ ($= \pm 4.5/33.4$) for H and L alloys, respectively.

Comparing the errors caused by the measurements to the standard deviations, it can be seen that the standard deviations are greater for all the properties. Thus, the results reported with the standard deviation are representative for all the properties of the two compositions of A380.0 Alloy.

The variations in the properties may be caused by the measuring errors, such as in length, diameter of the specimen, and in load, and by the non-uniformity of the specimens, such as segregation, porosity, inclusions, and the deformation of specimens before tests. The standard deviations in all the properties of L alloys show greater values than those of H alloy in these experiments. The possible reasons are that the non-uniformity in the specimens of the L alloy is more severe than in the specimens of the H alloy. For example for the L alloy, many specimens were apparently bent after die cast, but very few of H alloy specimens were bent. Moreover, more L alloy specimens were detected with defects by X-ray inspection, also more severe segregation (will be discussed in the next section) was found in L alloy specimens.

Attempts to determine Young's modulus based on the recorded load vs displacement curves did not succeed. The values calculated from the second straight lines were only 1182 and 1021 ksi for the H and L alloys, respectively, which are far below the value in the specification (10300 ksi). When calculated from first lines the values were only 1500 to 2000 ksi, which is still too low. Fig. 3 shows a typical load vs displacement curve (left), based on the specified properties of A380.0 alloy (elongation is taken as 4%). For comparison the typical curves of H (middle) and L (right) alloys recorded are also shown (the same as in Fig. 2). It is clearly shown that the slopes, which determine the Young's modulus, of the curves for both H and L alloys are much smaller than that of the standard alloy. The value obtained by drawing method could only be approximate because the measuring error is very large. The questions why the slopes of the load vs displacement curves of the specimens in these experiments are

much smaller than the typical value and why there are two straight lines in the load vs displacement curves still remain to be answered.

Original data, Calculations and Data Distribution

The original data and the calculations are presented in Tables 4 through 7. Figs 4 through 7 show the distributions of tensile strength, elongation, yield strength, and hardness values of the tested specimens, respectively.

3 Microstructure of 380 alloy

A microstructural analysis of 380 alloy was conducted. The objective of the study was to characterize the microstructure of the test specimens and to investigate the effects of the casting conditions on the microstructure.

3-1 Microstructure and Phase Identification

Several broken specimens of each composition were sectioned at different locations. The samples were mounted, polished, and examined by optical and scanning electronic microscopy (SEM). An energy dispersive x-ray spectrometer (EDX) was used for phase identification and composition analysis. An image analysis system was used to evaluate the size and distribution of the principal phases. The samples were also etched for dendrite arm spacing (DAS) and grain size measurements. The following features were observed in the specimens.

- The microstructures of the L and H alloys are shown in Figs 8 and 9, respectively. Generally, the microstructure of both alloys consist of alpha aluminum dendrites, a mixture of fine acicular and near modified eutectic Si particles, primary Si particles, Fe- and Cu-bearing phases and some complex intermetallics in the interdendritic areas. The microstructures of both alloys were very fine, DAS ranging between 5 and 10 μm . However, the morphology, size and distribution of the phases present in the L alloy differed significantly from those in the H alloy and from those in sand and permanent mold castings .
- When the polished surfaces of the samples were observed visually, the cross sectional area of L alloy specimens showed three distinct regions, a core area, a

ring surrounding the core, and an outer layer. This structure, shown in Fig. 10, was taken using an optical microscope at 25X magnification. The outer layer was about 1000 μm in width and the ring was about 400 μm in width. The microstructures of these three regions of the L alloy specimens are shown in Figs 8 and 11. The DAS was slightly coarser at the center area than in the outer layer and there were microstructural differences among these three regions.

- In the ring region, larger amounts of eutectic structure were detected, and the eutectic Si particles were also coarser than those in the core area and outer layer (see Fig. 8). The outer layer contained more eutectic Si particles than the core area. Primary Si particles were only observed in the ring region for the L alloy specimens; with sizes of 2 to 4 μm .
- In the ring region the amounts of the Cu-bearing phases were apparently larger than in the other regions (see Fig. 11). The majority of them were highly branched phases, their sizes ranging from 5 to 10 μm . A small amount of the Cu-bearing phases exhibited as small individual particles elongated along the interdendritic areas. In the outer layer and the core area, dominant Cu-bearing phases were small elongated particles. It appears that there is slightly more Cu-bearing phases in the outer layer than in the core area. Considerable amounts of the Cu-bearing particles were associated with Fe-bearing phases. This can be observed in the SEM photos (Fig. 12 and Fig. 15 (a)), where the Cu-bearing particles are the brightest. Even on Cu-bearing phases themselves these brighter portions can be observed, for example on the branched Cu-bearing phase (Fig. 13). Fig. 14 shows the EDX spectra of the brighter part and the remainder of the Cu-bearing phase. It indicates that the brighter part contains more copper than the less bright part.
- The Fe-bearing phase were present as polygonal particles, no Chinese script or needle phases were observed, which are the dominant Fe-bearing phases in sand and permanent mold castings of similar alloy compositions. The sizes of these polygonal particles were distributed in two distinct ranges. There were a few larger particles with diameters ranging from 5 to 10 μm (Fig. 15 (a)). The remaining particles were small with sizes 1 to 2 μm (Fig. 15 (b)), but in large amounts. The concentration of the Fe-bearing particles was apparently denser in

the ring and the outer layer regions than in the core area. The EDX spectrum (Fig. 16) shows that this phase also contains considerable amounts of Mn and Cr.

- In the H alloy specimens, the ring structure like that in the L alloy specimens was not observed visually. However, slightly larger amounts of fine eutectic Si were observed at locations very close to the sample surface. Toward the center the eutectic Si particles became less, but coarser, with the coarsening of the dendritic structure (see Fig. 9). No obvious Cu- and Fe-bearing phases segregated.
- Due to the larger Si contents, the H alloy specimens contained more Si phases, both eutectic and primary. In the thin outer layer, about 500 μm in width, the dominant Si phase was fine eutectic. Away from this layer towards the center, considerable amounts of primary Si particles precipitated (Fig. 9). In the L alloy, however, there were only small amounts of the primary Si found in the ring area. The sizes of the primary Si in the H alloy were also slightly greater than those in the L alloy (4 to 6 μm vs 2 to 4 μm).
- The Fe- and Cu-bearing phases were also in larger amounts in the H alloy than in the L alloy specimens (Fig. 17). The Fe-bearing phase consisted of large amounts of the larger particles, unlike in the L alloy where the dominant Fe-bearing particles were the smaller ones.
- The non-uniformity in the microstructure was observed not only across the cross sectional area, but also along the length of the specimen. Generally, the structure was coarse at the location close to the gate and was refined with the distance of melt flow. However, little difference was found between the middle of the reduced section and the end of the specimen.

3-2 Microstructures of 380 alloy solidified under different cooling rates

In this part of the study the effects of the cooling condition (cooling rates) on the microstructure of 380 alloy was investigated. The initial objectives were to estimate the cooling rates in aluminum alloy die castings, and to provide information for relating the properties measured from test specimens to the properties of die castings.

Experimental Procedure

Alloy: 380 alloy from Doehler-Jarvis production lines.

Mold and Design: To obtain different cooling rates; sand, graphite, permanent (steel) and water cooled and wedged copper molds were used. A furnace cooling sample was also taken in each run. The design and dimensions of these molds and the thermocouple arrangement are shown in Fig. 18.

Cooling Curve Measurement: Type "K" thermocouples were used for measuring the cooling curves at three locations of each molds. The thermocouples had a 304 SS sheath of diameter 0.062", and were firmly held at predetermined locations in the molds. The thermocouples were all connected to a data acquisition system for recording and displaying the temperature histories.

Melting and Pouring: In each run about 5 lbs of 380 ingots was melted in a 5 lb clay-reinforced graphite crucible in a 4.8 KW resistance electric furnace. The melt was degassed with a lance using pure nitrogen at temperature range of about 680 to 720 °C for 30 min. The pouring temperature was 750 ± 5 °C.

Analysis

Sampling: For each heat, hydrogen and chemical analysis specimens were taken. The castings were sectioned for analysis. Three samples were taken from the wedged copper mold casting at areas surrounding the thermocouple beads. One sample was taken from each of the other castings at a location surrounding the middle thermocouple bead.

Sample Preparation and Analysis: All the samples were mounted and polished for metallographic analysis, which included optical microscopy, SEM, EDX analysis and DAS and grain size measurements. The cooling curves were analyzed for investigating cooling rate and solidification characteristics.

Results

Summary of Microstructural Characteristics: Table VIII is a summary of this experiment. It shows the casting condition, measured cooling rates, DAS, and the microstructures observed.

Cooling Curves: The cooling curves at different locations in sand, permanent (steel), water cooled copper and wedged copper molds castings are shown in Fig. 19.

Microstructure: The cooling rate varies significantly with different mold materials and casting thicknesses. For example, the furnace cooling rate is 1.5×10^{-2} °C/s, the cooling rate in the sand mold is 1.4×10^{-1} °C/s, while it is greater than 5×10^2 °C/s in the thin portion (2 mm thick) of the wedged copper mold. The microstructure of 380 alloy is affected by the cooling rate. The typical microstructures of the cast specimens are shown in Figs 20 to 24. It can be seen that under various cooling conditions the microstructures of 380 alloy has the following characteristics.

With increasing cooling rate the DAS and grain size decrease. For example, the DAS is about 100 μm in the furnace cooled sample, 60 μm in sand mold casting, while it is only about 3 μm in the thin portion of the wedged copper mold casting. Accordingly, almost all the phases are refined. In addition to the size of the phases, the morphology and the composition of the phases are altered with the variation in cooling rate.

At very slow cooling, e. g. in the furnace cooled sample, a greater amount of large Fe rich needles (longer than 1000 μm) are formed (Fig. 20 (a)). The Fe rich phase also forms large (about 200 μm diameter) well developed, nearly equiaxed Chinese script (Fig 20 (b) and Fig 21 (b)). There is almost no Fe rich irregular polygonal particles. The Cu rich phases consist of clusters of large lumpy particles (up to about 80 μm in diameter, Fig. 21(a)), which have a smooth polished surface, and individual and clusters of branched (eutectic-like) particles (Fig. 21(b)), which exhibit granular surfaces. Silicon particles are coarse and acicular but few are primary silicon.

- With increasing cooling rate, the Fe rich needles disappear, the Chinese script becomes smaller, elongated, fragmented, and difficult to observe (Figs 22, 23 and 24). Consequently, irregular polygonal particles appear and are subsequently refined. At very high cooling rates, e. g. in the thin portion of the wedged copper mold, the Fe rich phase forms fine particles in the interdendritic areas.
- At cooling rates greater than that in the sand mold ($> 1.4\text{ }^{\circ}\text{C/s}$), the Fe rich needles are not detected (Figs 22 and 23). This is very different from what was observed in 319 alloy, where the Fe rich needles readily formed in sand and permanent mold castings. Since 380 and 319 are similar in that both are Al-Si-Cu alloys, containing Fe and Mn, this phenomenon needs further investigation.
- In the sand mold castings, the Fe rich Chinese script is large and well developed, however, it is elongated (Fig. 22(a)). At higher cooling rates, greater than that in the graphite mold ($> 1.9\text{ }^{\circ}\text{C/s}$), the Chinese script becomes very small ($< 15\text{ }\mu\text{m}$ in diameter) and forms in the interdendritic areas (Figs 23 and 24 (a)). Its boundaries are not clearly delineated from the matrix. The Chinese script can hardly be seen at high cooling rates ($> 100\text{ }^{\circ}\text{C/s}$), e.g. in the middle and thin portions of the wedged copper mold casting (Fig. 24 (a) and (b)).
- The irregular polygonal Fe rich particles form at cooling rates greater than that in the graphite mold ($> 1.9\text{ }^{\circ}\text{C/s}$, Fig. 22 (b)). At this cooling rate the particles are relatively big, about 10 to 15 μm in diameter. With the increase in cooling rate the irregular polygonal particles become refined (Figs 23 and 24). In the thin portion of the wedged copper mold (cooling rate $\sim 500\text{ }^{\circ}\text{C/s}$) these particles cannot be observed (Fig. 24 (c)).
- Though copper rich phases become significantly refined with increasing the cooling rate, their morphologies do not show significant change. However, at very high cooling rates, e.g. in the thin portion of the wedged copper mold (cooling rate $\sim 500\text{ }^{\circ}\text{C/s}$) only tiny, lumpy particles form in the interdendritic regions.
- The 380 alloy used for testing was not modified. Silicon forms eutectic flakes, which are refined with increasing cooling rate. However, considerable amount

of primary silicon particles are observed throughout the samples at a cooling rate of about 2 °C/s (graphite mold) and greater (Figs 22 (b), 23 and 24 (a)). At lower cooling rates, e.g. in the sand mold casting, no primary silicon particles can be seen (Fig. 22(a)). In a furnace cooled sample, several large primary silicon particles are found (Fig. 20 (b)), but they are all concentrated in a small area. At very high cooling rates, the primary silicon cannot be found (Fig. 24 (b) and (c)).

4 Fracture surface analysis

Fractographic analysis was conducted on several specimens of each composition. Fig. 25 (a) shows the fracture surfaces of H alloy specimens. It shows small facets dominant on the surfaces, which are the cleaved Si particles, and a little fibrous structure. The overall fracture is brittle in nature. The fracture surface of a L alloy specimen is shown in Fig. 25 (b), which also shows small facets, but more fibrous structure, indicating that the specimen underwent significant deformation. On the surfaces of specimens of both alloys macro- and micro-porosity (Fig. 26) and inclusions (Fig. 27) are detected.

5 Discussions:

5-1 Properties

- The mechanical property tests show that the variation of composition of 380 alloy within its specification has significant effects on the mechanical properties. For example, the tensile strength, yield strength and the hardness (Rb) of H alloy specimens are 11.6%, 57.5% and 86.5% respectively higher than those of L alloy specimens. Elongation of L alloy specimens is 153% higher than that of H alloy specimens.
- This property variation within the specified composition range was not reflected in existing specifications and no work in the published literature shows these data.

- All the published property values fell between those of the H and L alloys. This indicates that the compositions of the alloys tested for establishing the specifications were probably between the H and L alloy compositions.
- The effects of many elements on mechanical properties of aluminum die casting alloys have been studied. For example, increasing Si, Cu, and Fe contents increases strength and reduces ductility, this was confirmed in this study. The H alloy, which has higher Si, Cu and Fe contents, possesses higher tensile strength, yield strength and hardness, but lower ductility than the L alloy. However, several questions remain to be answered. For example, to what extent these elements affect the properties, individually and in combination. What element makes the most contribution to the property change, and to which property? What is the optimum composition for certain property requirements?
- The mechanical properties of the as-cast die castings are affected not only by the alloy composition, but also by other factors, such as casting condition, gating design, defects, and outside environmental effects. Some effects of these factors were reflected in this study. For example, though specimens of the same alloy in this study had the same composition, and were produced under controlled conditions following the same procedure, the properties still varied. Like among the L alloy specimens, the tensile strength and elongation varied from 41.38 to 48.22 ksi and 4.3 to 8.2%.

5-2 Microstructure

- The microstructure at the reduced section of the tensile specimens was similar to the microstructure between the upper and middle portions of the wedged copper mold casting. By comparing the DAS it is estimated that the cooling rate at this portion of the tensile specimen was about 100 to 150 °C/s.
- In the die casting, the high cooling rate, fast filling of the die cavity and solidifying under high pressure result in fine, unique microstructure in A380 alloy. This microstructure is very different from that in sand and permanent mold castings. In addition, the variation in composition affects the microstructure significantly. The difference in microstructure presumably causes the variation in mechanical properties.

Due to time constraints, the effects of impurity on microstructure were not investigated in this phase of the study. The variation in the microstructure between the H and L alloys may not be due to the effect of the alloying elements only. In 380 die casting alloy the total impurity contents could be 4.95 %, which should affect the microstructure, and consequently, affect the properties significantly. Their effects should not be ignored.

Based on the experiments and the literature review, it is evident that a detailed systematic study of the alloy-microstructure-performance relationship is definitely necessary for aluminum die casting alloy development and for the competitiveness of the U. S. die casting industry.

Table I Compositions of the L and H alloys used in the tensile property tests

	Cu	Si	Fe	Mn	Mg	Zn	Ni	Pb	Sn	Ti	Cr	Al
L	2.90	7.37	0.67	0.24	0.02	0.85	0.02	0.02	0.01	0.03	0.02	Remainder
H	3.91	9.51	1.27	0.28	0.27	2.89	0.53	0.06	0.03	0.04	0.08	Remainder

Table II Specification of A380.0 alloy

	Cu	Si	Fe	Mn	Mg	Zn	Ni	Pb	Sn	Ti	Cr	Other Total	Al
A380	3.0-4.0	7.5-9.5	1.3	0.50	0.10	3.0	0.50		0.35			0.50	Remainder

Table III The measured tensile property and hardness values of A380.0 alloy at two composition levels and the specifications of the properties from different sources

		Tensile Strength ksi (MPa)	Yield Strength ksi (MPa)	Elongation %	Hardness Rb (HB)
L Alloy	Average	44.94±1.68 (309.86±11.58)	20.42±1.91 (140.80±13.17)	6.46±0.85	33.4±4.5 (69.4±9.4)
	Min/Max	41.38 / 48.22	16.97 / 24.41	4.3 / 8.2	21.2 / 47.0
H Alloy	Average	50.17±1.04 (345.92±7.17)	32.17±1.48 (221.81±10.20)	2.55±0.28	62.3±3.5 (98.3±5.5)
	Min/Max	48.22 / 52.87	29.08 / 35.00	2.00 / 3.05	49.6 / 71.0

Specifications of A380 Tensile and Hardness Properties in Different References

1*	47 (324)	23 (159)	4.0	(75)
2*	47 (325)	23 (160)	4.0**	
3*	47 (320)	23 (160)	3.5**	(75)
4*	47 (325)	23 (160)	3.5**	
5*	47 (325)	23 (158)	3.5**	

* 1 Rooy, E. L., "Aluminum and Aluminum Alloys", ASM Handbook, 9th Edition, Vol. 15, p743-770.

2 ASM Metal Handbook, 10th Edition, Vol. 2, 1992.

3 1992 Annual Book of ASTM Standard, Vol 15, 1990.

4 ASM Metal Handbook, 9th Edition, Vol. 15, 1990.

5 "Aluminum Alloy A380 (Aluminum Die Casting Alloy)", Metal Digest, Al-68, June 1986.

** Tests were conducted on 2 inch diameter specimens.

Table IV Original data in tensile property tests of H alloy and calculations -- I

Sample : Modified standard tensile specimen for die casting, according to ASTM B557-84

The sample is round with diameter of 0.250 in and gage length of 2.000 in

Test machine : Instron 8502, Waveform setting was the ramprate 0.06 in/min

Chart record setting : X - Position, Speed = 0.020 inch sample displacement per inch distance on X axis

Y - Load, Speed = 250 lbs per inch distance on Y axis

Sample Series	Diameter (in)			Area	Load (lb)		Displace-	UTS	Y	E	Date	Note
	Max	Min	Ave	sq in	Max	0.20%	ment (in)	ksi	ksi	%		
H 11-1	0.2590	0.2463	0.2527	0.0501	2454	1486	0.040	48.95	29.64	2.00	5/24/93	out 1/2
H 11-2	0.2545	0.2422	0.2484	0.0484	2471	1629	0.050	51.01	33.62	2.50	6/8/93	out 1/2
H 12-1	0.2575	0.2470	0.2523	0.0500	2456	1659	0.048	49.14	33.20	2.40	6/8/93	out 1/2
H 12-2	0.2550	0.2426	0.2488	0.0486	2264	--	0.043	46.57	--	2.15	6/8/93	
H 13-1	0.2570	0.2485	0.2528	0.0502	1860	1744	0.024	37.07	34.77	1.20	6/8/93	D, SB, Out 1/2
H 13-2	0.2540	0.2430	0.2485	0.0485	2466	1488	0.046	50.85	30.68	2.30	5/24/93	out
H 14-1	0.2580	0.2472	0.2526	0.0501	2444	1659	0.041	48.77	33.10	2.05	6/10/93	Out 1/2
H 14-2	0.2540	0.2430	0.2485	0.0485	2463	1582	0.051	50.78	32.61	2.55	6/10/93	
H 15-1	0.2545	0.2455	0.2500	0.0491	2496	1558	0.057	50.85	31.74	2.85	6/8/93	Out 1/2
H 15-2	0.2540	0.2440	0.2490	0.0487	2392	1529	0.044	49.12	31.39	2.20	6/10/93	D
H 20-1	0.2570	0.2458	0.2514	0.0496	2315	1390	0.044	46.64	27.99	2.20	6/10/93	D, Out 1/2
H 20-2	0.2540	0.2430	0.2485	0.0485	2441	1537	0.053	50.33	31.69	2.65	6/10/93	Out 1/2
H 21-1	0.2570	0.2468	0.2519	0.0498	2469	1467	0.057	49.54	29.43	2.85	6/10/93	Out 1/2
H 21-2	0.2550	0.2430	0.2490	0.0487	2491	1646	0.061	51.15	33.79	3.05	6/10/93	
H 22-1	0.2572	0.2470	0.2521	0.0499	2435	1522	0.046	48.78	30.49	2.30	6/10/93	Out 1/2
H 22-2	0.2540	0.2430	0.2485	0.0485	2400	1559	0.048	49.48	32.14	2.40	6/10/93	Out 1/2
H 23-1	0.2575	0.2465	0.2520	0.0499	2405	1584	0.044	48.22	31.77	2.20	6/10/93	Out 1/2
H 23-2	0.2530	0.2420	0.2475	0.0481	2320	1542	0.045	48.22	32.05	2.25	6/10/93	
H 24-1	0.2555	0.2458	0.2507	0.0493	2472	1626	0.055	50.10	32.96	2.75	6/10/93	Out 1/2
H 24-2	0.2540	0.2425	0.2483	0.0484	2475	1694	0.052	51.13	35.00	2.60	6/10/93	Out 1/2

Table IV Original data in tensile property tests of H alloy and calculations -- II

Sample Series	Diameter (in)			Area	Load (lb)		Displace-	UTS	Y	E	Date	Note
	Max	Min	Ave	sq in	Max	0.2 %	ment (in)	ksi	ksi	%		
H 25-1	0.2565	0.2465	0.2515	0.0497	2460	1551	0.048	49.52	31.22	2.40	6/10/93	Out 1/2
H 25-2	0.2540	0.2425	0.2483	0.0484	2462	1687	0.051	50.86	34.86	2.55	6/10/93	
H 26-1	0.2565	0.2470	0.2518	0.0498	2482	1589	0.060	49.86	31.92	3.00	6/10/93	
H 26-2	0.2520	0.2410	0.2465	0.0477	2519	1571	0.059	52.78	32.91	2.95	6/10/93	SB
H 33-1	0.2560	0.2465	0.2513	0.0496	2439	1638	0.053	49.19	33.04	2.65	6/10/93	
H 33-2	0.2530	0.2420	0.2475	0.0481	2296	--	0.044	47.72	--	2.20	6/10/93	
H 34-1	0.2570	0.2475	0.2523	0.0500	2532	1558	0.048	50.67	31.18	2.40	5/24/93	out 1/2
H 34-2	0.2530	0.2425	0.2478	0.0482	2449	1548	0.050	50.80	32.11	2.50	5/24/93	out 1/2
H 35-1	0.2600	0.2475	0.2538	0.0506	2168	--	0.048	42.87	--	2.40	6/8/93	
H 35-2	0.2540	0.2428	0.2484	0.0485	2498	1695	0.058	51.55	34.98	2.90	6/8/93	D
H 60-1	0.2560	0.2462	0.2511	0.0495	2529	1531	0.047	51.07	30.93	2.35	6/8/93	O
H 60-2	0.2532	0.2422	0.2477	0.0482	2471	1559	0.053	51.28	32.36	2.65	6/8/93	
H 70-1	0.2590	0.2445	0.2518	0.0498	2458	1636	0.047	49.38	32.86	2.35	6/8/93	
H 70-2	0.2540	0.2425	0.2483	0.0484	2426	1648	0.051	50.12	34.05	2.55	6/8/93	
H 71-1	0.2560	0.2468	0.2514	0.0496	2456	1443	0.043	49.48	29.08	2.15	6/8/93	Out 1/2
H 71-2	0.2540	0.2430	0.2485	0.0485	2417	1499	0.054	49.83	30.91	2.70	6/8/93	
H 72-1	0.2575	0.2462	0.2519	0.0498	2505	1595	0.057	50.28	32.02	2.85	5/24/93	out 1/2
H 72-2	0.2540	0.2425	0.2483	0.0484	2468	1504	0.053	50.99	31.08	2.65	6/8/93	O
H 73-1	0.2565	0.2468	0.2517	0.0497	2513	1565	0.060	50.53	31.46	3.00	6/8/93	
H 74-2	0.2535	0.2422	0.2479	0.0482	2473	1627	0.052	51.26	33.72	2.60	6/8/93	Out 1/2
							Average	50.17	32.17	2.55		
							Stdev	±1.04	±1.48	±0.28		

* B = The specimen was bent visually; SB = The specimen was slightly bent visually;

D = There was defect detected by X-ray inspection; On = The fracture was right at the gage mark;

O = The fracture was outside the gage length; Out 1/2 = The fracture was outside the middle half of the gage length

** -- = No reliable date could be taken

The data in the shadow were omitted in the property calculations due to apparent defects or no reliable data

Table V Original data in tensile property tests of L alloy and calculations -- I

Sample : Modified standard tensile specimen for die casting, according to ASTM B557-84

The sample is round with diameter of 0.250 in and gage length of 2.000 in

Test machine : Instron 8502, Waveform setting was the ramprate 0.06 in/min

Chart record setting : X - Position, Speed = 0.020 inch sample displacement per inch distance on X axis

Y - Load, Speed = 250 lbs per inch distance on Y axis

Sample Series	Diameter (in)			Area	Load (lb)		Displace-	UTS	Y	E	Date	Note
	Max	Min	Ave	sq in	Max	0.20%	ment (in)	ksi	ksi	%		
L 10-1	0.2525	0.2435	0.2480	0.0483	2189	934	0.111	45.32	19.33	5.55	6/10/93	Out 1/2
L 10-2	0.2510	0.2405	0.2458	0.0474	2190	883	0.133	46.17	18.62	6.65	6/10/93	Out 1/2
L 11-1	0.2525	0.2430	0.2478	0.0482	2276	942	0.140	47.21	19.54	7.00	6/10/93	
L 11-2	0.2500	0.2400	0.2450	0.0471	2148	1043	0.143	45.56	22.12	7.15	6/10/93	Out 1/2
L 12-1	0.2515	0.2440	0.2478	0.0482	2086	857	0.126	43.27	17.79	6.30	6/10/93	D, Out 1/2
L 12-2	0.2500	0.2412	0.2456	0.0474	2107	966	0.142	44.48	20.40	7.10	6/10/93	Out 1/2
L 13-1	0.2520	0.2440	0.2480	0.0483	2016	853	0.112	41.73	17.66	5.60	6/10/93	Out 1/2
L 13-2	0.2495	0.2405	0.2450	0.0471	2040	1000	0.086	43.27	21.22	4.30	5/24/93	On
L 14-1	0.2565	0.2445	0.2505	0.0493	2292	892	0.141	46.51	18.10	7.05	6/10/93	Out 1/2
L 14-2	0.2495	0.2405	0.2450	0.0471	2262	993	0.145	47.98	21.06	7.25	6/10/93	D, Out 1/2
L 15-1	0.2530	0.2440	0.2485	0.0485	2152	866	0.124	44.37	17.85	6.20	6/10/93	Out 1/2
L 15-2	0.2565	0.2428	0.2497	0.0490	2245	1176	0.120	45.86	24.03	6.00	6/10/93	SB
L 34-1	0.2530	0.2440	0.2485	0.0485	2007	823	0.097	41.38	16.97	4.85	6/8/93	D, O
L 34-2	0.2574	0.2420	0.2497	0.0490	2269	1195	0.124	46.33	24.41	6.20	6/10/93	
L 35-1	0.2550	0.2435	0.2493	0.0488	2287	1060	0.144	46.87	21.72	7.20	6/10/93	O
L 35-2	0.2570	0.2430	0.2500	0.0491	2254	1161	0.115	45.92	23.65	5.75	6/10/93	D, B, Out 1/2
L 36-1	0.2520	0.2442	0.2481	0.0483	1951	--	0.073	40.36	--	3.65	5/24/93	On
L 36-2	0.2505	0.2400	0.2453	0.0472	2205	1050	0.118	46.68	22.24	5.90	5/24/93	On
L 37-1	0.2525	0.2440	0.2483	0.0484	1911	844	0.080	39.48	17.43	4.00	6/10/93	D, O
L 37-2	0.2495	0.2412	0.2454	0.0473	2116	976	0.145	44.76	20.64	7.25	6/10/93	Out 1/2

Table V Original data in tensile property tests of L alloy and calculations -- II

Sample Series	Diameter (in)			Area	Load (lb)		Displace-	UTS	Y	E	Date	Note
	Max	Min	Ave	sq in	Max	0.20%	ment (in)	ksi	ksi	%		
L 38-1	0.2530	0.2450	0.2490	0.0487	2144	891	0.138	44.03	18.31	6.90	6/10/93	Out 1/2
L 38-2	0.2500	0.2410	0.2455	0.0473	2067	936	0.144	43.67	19.78	7.20	6/10/93	
L 39-1	0.2520	0.2442	0.2481	0.0483	2086	1113	0.126	43.15	23.03	6.30	6/10/93	D, B
L 39-2	0.2500	0.2410	0.2455	0.0473	2079	866	0.146	43.92	18.29	7.30	6/10/93	Out 1/2
L 40-1	0.2530	0.2435	0.2483	0.0484	2058	898	0.076	42.52	18.55	3.80	6/10/93	D, O
L 40-2	0.2500	0.2410	0.2455	0.0473	2074	966	0.134	43.81	20.41	6.70	6/8/93	O
L 41-1	0.2525	0.2445	0.2485	0.0485	2088	--	0.139	43.05	--	6.95	6/8/93	SB, Out 1/2
L 41-2	0.2495	0.2410	0.2453	0.0472	2093	975	0.135	44.31	20.63	6.75	6/8/93	O
L 42-1	0.2525	0.2440	0.2483	0.0484	2028	--	0.114	41.90	--	5.70	5/24/93	Out 1/2
L 42-2	0.2495	0.2410	0.2453	0.0472	2045	966	0.124	43.29	20.45	6.20	6/8/93	D, SB, Out 1/2
L 70-1	0.2575	0.2450	0.2513	0.0496	2209	1125	0.108	44.55	22.69	5.40	6/8/93	Out 1/2
L 70-2	0.2510	0.2400	0.2455	0.0473	2259	948	0.148	47.72	20.03	7.40	6/8/93	Out 1/2
L 71-1	0.2570	0.2460	0.2515	0.0497	2324	1097	0.164	46.78	22.09	8.20	6/8/93	
L 71-2	0.2490	0.2410	0.2450	0.0471	2103	918	0.143	44.61	19.47	7.15	6/8/93	Out 1/2
L 72-1	0.2520	0.2442	0.2481	0.0483	2121	949	0.117	43.87	19.63	5.85	5/24/93	SB, On
L 72-2	0.2510	0.2410	0.2460	0.0475	2128	894	0.138	44.77	18.81	6.90	6/8/93	Out 1/2
L 73-1	0.2600	0.2455	0.2528	0.0502	2205	1115	0.100	43.95	22.23	5.00	6/8/93	
L 73-2	0.2505	0.2405	0.2455	0.0473	2126	983	0.137	44.91	20.77	6.85	6/8/93	SB
L 74-1	0.2530	0.2440	0.2485	0.0485	2110	985	0.114	43.50	20.30	5.70	6/8/93	Out 1/2
L 74-2	0.2505	0.2410	0.2458	0.0474	2287	970	0.142	48.22	20.46	7.10	6/8/93	Out 1/2
							Average	44.94	20.42	6.46		
							Stdev	±1.68	±1.91	±0.85		

* B = The specimen was bent visually; SB = The specimen was slightly bent visually;

D = There was defect detected by X-ray inspection; On = The fracture was right at the gage mark;

O = The fracture was outside the gage length; Out 1/2 = The fracture was outside the middle half of the gage length

** -- = No reliable date could be taken

The data in the shadow were omitted in the property calculations due to apparent defects or no reliable data

Table VI Hardness data of H alloy specimens

Specimen : Modified standard tensile specimen for die casting, according to ASTM B557-84

Testing location : On the flat surfaces at the two ends of the specimen

Ten measurements are conducted on each specimen, five on each end

The first measurement starts at 3/8" from the end with the ingate print

The measurements are about 3/16" apart from each other

Testing machine: Rockwell Hardness Tester, Model 3JR, Tester Number 10661

Rockwell Hardness Scale B; Load - 100 Kg, 1/16" Ball

Sample Series	Measured Value										Average		
	1	2	3	4	5	6	7	8	9	10	Ave.	Stdev	
H 11-1	52.2	57.9	64.0	63.7	62.0	67.0	61.5	61.3	64.2	64.4	61.8	4.2	
H 11-2	65.9	67.7	66.8	64.9	68.6	69.0	69.7	67.2	66.7	62.9	66.9	2.0	
H 12-1	63.5	62.5	65.1	61.2	64.0	63.7	58.4	59.7	56.1	59.8	61.4	2.9	
H 13-1	54.8	63.2	63.5	63.1	57.9	63.2	56.0	58.2	57.8	59.5	59.7	3.3	
H 13-2	62.5	66.8	69.2	69.7	70.6	69.3	68.6	68.3	65.0	62.7	67.3	2.9	
H 14-1	55.0	54.7	49.6	55.8	54.9	64.0	60.4	51.8	60.3	54.5	56.1	4.3	
H 15-1	53.0	54.0	52.8	57.0	53.2	64.0	55.4	57.5	55.9	53.0	55.6	3.4	
H 23-1	66.4	66.8	62.4	64.5	65.3	68.8	67.7	56.7	59.0	57.0	63.5	4.5	
H 23-2	62.5	61.0	58.0	58.9	59.3	65.7	52.1	54.8	56.0	58.8	58.7	3.9	
H 24-1	63.8	66.5	65.7	67.2	65.0	66.6	61.9	55.0	59.0	63.6	63.4	3.9	
H 24-2	66.5	65.5	59.6	57.6	65.3	61.5	66.1	59.4	58.0	65.8	62.5	3.7	
H 25-1	54.4	63.8	62.4	66.3	68.4	66.0	67.0	65.8	66.7	62.6	64.3	4.0	
H 34-1	60.0	64.1	62.1	62.1	62.8	67.8	63.9	61.6	55.9	59.1	61.9	3.2	
H 34-2	66.4	69.5	67.5	67.5	70.0	65.4	60.5	63.7	62.4	64.7	65.8	3.0	
H 71-1	52.1	64.2	65.4	67.8	67.8	67.5	69.3	64.9	63.9	55.1	63.8	5.7	
H 71-2	66.0	68.8	65.9	62.9	63.0	68.7	67.1	66.0	64.9	66.1	65.9	2.0	
H 72-1	61.8	62.2	61.0	62.7	61.6	66.7	63.0	55.1	66.1	63.1	62.3	3.2	
H 72-2	66.7	61.6	64.3	61.2	65.1	71.0	65.4	66.8	66.0	63.0	65.1	2.9	
H 73-1	57.5	64.0	56.9	59.0	57.8	59.2	66.7	59.6	59.9	56.3	59.7	3.3	
H 74-2	66.2	64.9	61.1	59.0	59.2	57.2	53.4	57.2	61.2	53.8	59.3	4.2	
										Ave	62.3	3.5	
										Std	3.3		

Table VII Hardness data of L alloy specimens

Specimen : Modified standard tensile specimen for die casting, according to ASTM B557-84

Testing location : On the flat surfaces at the two ends of the specimen

Ten measurements are conducted on each specimen, five on each end

The first measurement starts at 3/8" from the end with the ingate print

The measurements are about 3/16" apart from each other

Testing machine: Rockwell Hardness Tester, Model 3JR, Tester Number 10661

Rockwell Hardness Scale B; Load - 100 Kg, 1/16" Ball

Sample Series	Measured Value										Average		
	1	2	3	4	5	6	7	8	9	10	Ave.	Stdev	
L 10-1	37.6	46.1	38.9	44.9	47.0	42.3	38.8	39.0	41.4	39.1	41.5	3.4	
L 11-1	34.2	37.9	37.5	41.9	32.7	39.3	39.0	39.1	35.3	34.1	37.1	2.9	
L 11-2	25.3	32.9	28.3	28.1	36.7	34.1	33.0	34.0	34.0	30.6	31.7	3.5	
L 12-1	31.4	37.2	28.5	27.7	30.8	33.0	30.9	34.5	28.9	30.0	31.3	2.9	
L 13-1	26.3	23.5	23.0	30.0	24.8	22.0	25.2	23.2	26.9	23.0	24.8	2.4	
L 13-2	35.3	39.0	37.8	43.3	43.4	28.3	33.7	30.1	29.3	32.1	35.2	5.5	
L 14-1	25.8	27.8	32.2	29.0	33.9	24.8	35.9	28.8	29.5	36.7	30.4	4.1	
L 36-1	26.2	32.9	35.5	32.5	33.0	33.1	31.5	26.2	27.0	25.3	30.3	3.7	
L 36-2	39.7	35.7	38.4	37.8	36.0	35.2	33.2	36.0	32.9	34.7	36.0	2.2	
L 37-1	25.0	29.0	31.0	28.8	29.4	33.4	36.3	30.0	25.7	25.0	29.4	3.6	
L 37-2	35.6	36.3	31.6	35.7	36.2	21.2	24.9	28.0	25.3	25.8	30.1	5.7	
L 38-1	26.3	30.5	27.5	28.5	33.1	32.7	33.2	22.9	26.8	26.0	28.8	3.5	
L 38-2	29.1	32.8	26.8	24.6	28.0	35.0	29.7	34.1	24.0	24.4	28.9	4.1	
L 42-1	27.3	36.0	35.7	35.0	37.9	32.7	30.3	29.1	28.0	26.4	31.8	4.1	
L 72-1	25.5	34.0	32.1	35.9	35.7	39.8	38.1	35.0	34.0	33.0	34.3	3.9	
L 72-2	37.0	39.5	38.1	36.1	31.3	33.0	34.8	38.0	29.9	32.8	35.1	3.2	
L 73-1	42.5	36.6	44.9	43.8	44.0	45.3	44.7	42.8	40.0	34.0	41.9	3.8	
L 73-2	38.0	41.7	39.6	34.4	39.0	40.5	35.0	37.8	37.9	34.9	37.9	2.5	
L 74-1	30.4	32.0	35.0	31.4	31.0	34.2	38.5	34.4	31.0	32.0	33.0	2.5	
L 74-2	44.4	42.1	37.5	43.1	37.9	38.7	35.0	38.5	36.9	34.0	38.8	3.4	
										Ave	33.4	3.5	
										Std	4.5		

Table VIII Summary of solidification data of 380 alloy under different cooling rates

Casting Condition		Cooling Rate (C/s)	DAS (μm)	Microstructural Characteristics
Furnace Cooled		0.015	100.5	Very coarse Si flakes, huge Fe rich needles, big equiaxed chinese script, eutectic-like and clusters of large Cu rich lumpy particles
Sand Mold		0.14	59.7	No Fe rich needle, chinese script elongated, and disintegrated, no primary silicon particles
Graphite Mold		1.9	24.8	No Fe rich needle, small amount of chinese script confined in interdendritic areas, irregular polygonal Fe rich phase and primary silicon particles appear.
Steel Mold		2.2	17.7	Similar to but finer than in graphite mold
Water Colled Copper Mold		3.4	15.8	Similar to but even finer than in graphite mold
Wedged Copper Mold	Top	91.7	13.5	Similar to but much finer than in graphite mold
	Middle	~120	6	Finer than in top portion, no Chinese scrip phase found
	Bottom	>500	3.1	All Fe and Cu rich phases confined in interdendritic areas as lumpy particles, very fine Si particles

Standard dimensions, in.	
G -- Gage length	2.000 ± 0.005
D -- Diameter (see Note)	0.250 ± 0.005
R -- Radius of fillet, min	3
A -- Length of reduced section, min	2.25
L -- Over-all length, min	9
B -- Distance between grips, min	4.5
C -- Diameter of end section, approximate	3/8

Note - The reduced section may have a gradual taper from the ends toward the center, with the ends not more than 0.005 in. larger in diameter than the center.

Fig. 1 Modified standard test specimen for die castings

Fig. 2 Typical load vs displacement curves of H alloy (left) and L alloy (right) specimens displayed on the chart record

Fig. 3 The load vs displacement curve (left) drawn according to the specified mechanical property data (elongation was taken as 4%) compared with the actually recorded curves for the H (middle) and L alloy (right) specimens

Fig. 4 The distribution of tensile strength of the L and H alloys vs number of specimens tested (total 35 valid specimens)

Fig. 5 The distribution of yield strength of the L and H alloys vs number of specimens tested (total 35 valid specimens)

Fig. 6 The distribution of elongation of the L and H alloys vs number of specimens tested (total 35 valid specimens)

Fig. 7 The distribution of hardness (Rb) of the L and H alloys vs number of measurements (total 20 specimens and 200 measurements)

(a)

(b)

(c)

Fig. 8 Typical microstructures of L alloy specimen at (a) outer layer, (b) ring region, and (c) core area of the cross section

(a)

(b)

(c)

Fig. 9 Typical microstructures of H alloy specimen at (a) outer, (b) middle, and (c) center area of the cross section

Fig. 10 Ring structure in the cross section of the L alloy specimen

(a)

(b)

(c)

Fig. 11 Typical microstructures of L alloy specimen at (a) outer layer, (b) ring region, and (c) core area; SEM photographs

Fig. 12 Copper-bearing phases associated with Fe-bearing phase

Fig. 13 Branched Cu-bearing phase; the brighter part contains more Cu than the remainder of the phase

(a)

(b)

Fig. 14 EDX spectra of branched Cu-bearing phase (see Fig. 13); (a) the less bright part, and (b) the bright part

(a)

(b)

Fig. 15 Iron-bearing polygonal particles; (a) large particles with diameter of 5 to 10 μm , and (b) small particles with diameter of 1 to 2 μm

Fig. 16 EDX spectra of the Fe-bearing phase

Fig. 17 Morphologies of typical Fe- and Cu-bearing phases in H alloy specimen; the polygonal particles are the Fe-bearing phase and the small, elongated, bright particles are Cu-bearing phase

Fig. 18 Schematic diagram of molds; (a) wedged copper mold, and (b) sand, graphite, and permanent (steel) mold

Fig. 19 Cooling curves of 380 alloy cast in (a) sand mold, (b) permanent (steel) mold, (c) water cooled copper mold, and (d) wedged copper mold

(a)

(b)

Fig. 20 Microstructures of furnace cooled 380 alloy sample showing (a) large Fe-bearing needles (arrow), and (b) large, well-developed, equiaxed Fe-bearing Chinese script (A) and primary Si particles (arrow)

(a)

(b)

Fig. 21 Microstructures of furnace cooled 380 alloy sample showing (a) cluster of large, lumpy Cu-bearing particles (arrow), and (b) branched Cu-bearing phase (arrow) and the Fe-bearing Chinese script (A)

(a)

(b)

Fig. 22 Microstructures of 380 alloy samples cast in (a) sand mold, and (b) graphite mold

(a)

(b)

Fig. 23 Microstructures of 380 alloy samples cast in (a) permanent (steel) mold, and (b) water cooled copper mold

(a)

(b)

(c)

Fig. 24 Microstructures of 380 alloy samples cast in wedged copper mold; at (a) top, wider, (b) middle, and (c) bottom, narrower section

Fig. 25 Fracture surfaces of (a) H alloy specimen, and (b) L alloy specimen

(a)

(b)

Fig. 26 Pores observed under SEM in fracture surfaces of (a) L alloy specimen, and (b) H alloy specimen

(a)

(b)

Fig. 27 Inclusions observed in fracture surfaces of A380.0 die casting alloy

Sub-Project 5

Influence of Die Materials and Coatings on Die Casting Quality

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FINAL REPORT

QUALITY MEASURES FOR DIE CASTINGS

by

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prepared for

**Center for Applied Research in Metal Casting
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QUALITY MEASURES FOR DIE CASTINGS

To be acceptable as near net shape manufactured components, die castings must meet the specified service requirements set forth by the customer. Consistent with Munsinger's (1) description of quality, the quality of a component or system is herein taken to indicate that the particular component or system meets or conforms to the designated requirements. The term "quality measures", as used herein, refers to the collection of designated requirements and/or properties which must be met to insure that the die cast component is acceptable for the intended service or application. Obviously, the successful completion of any business transaction involving manufactured products requires communication and understanding of the significance of all of the quality measures by the customer and manufacturers. As an aid in establishing the communication of suggested and typical quality measures for die castings, the American Die Casting Institute (ADCI) formulated and published a collection of property and performance specifications for die castings (2). These ADCI documents consist of four series, which address the Engineering (E Series), the Metallurgical (M Series), the Commercial (C Series), and the Quality (Q Series) Standards for die castings. A general reference which addresses metallic materials specifications is the Metallic Materials Specification Handbook (3).

The requirements typically associated with die castings are listed in Table 1. Also included in Table 1 are the properties and/or features of the die casting which are used to judge that the part meets the given requirements.

Table 1: Typical Die Casting Quality Measures

<u>Requirement</u>	<u>Associated Property/Features</u>
Dimensional	Dimensions and Tolerances
Surface Finish	Surface Finish
Pressure Tightness	No thru-thickness, interconnected porosity
Load Bearing Ability	Mechanical Properties
Heat Treatable	Low Gas Content
Weldable	Low Gas Content
Machinable, (without exposing porosity)	Adequate "Skin" Thickness

The dimensional and surface finish requirements are inherent to essentially all manufactured components which are to be assembled in larger systems. The capabilities of the die casting industry to produce die castings of specific alloys with given dimensions, tolerances, and surface finishes is reported in several references (2,4). The requirement for pressure tightness is attributed to those die castings which are to serve as liquid or gas containers. Most die castings are required to be load bearing, and as such, most have mechanical properties, such as tensile yield strengths, ultimate tensile strengths, failure elongations, and impact toughness, which meet or exceed the specified minimum values appropriate for the given application. In order to improve the mechanical properties of the die castings, in many cases it is desirable that they could be heat treated. However, heat treatment of die castings, without their blistering or altering surface finish, requires that the contained gas content of the die casting be below a threshold value. In some cases, it is desirable or required that the die castings be joined via the use of a welding process. Similar to the heat treating of die castings, the successful welding of die castings generally requires that the contained gas content of the die casting be less than a threshold value. A microstructural feature common to most die castings is the so-called "skin", which is a layer of low porosity material, typically 0.012 to 0.024 (0.3 to 0.6 mm) thick (5), which forms where the liquid alloy contacts the die surfaces. Any machining or surface removal method which removes the entire "skin" thickness generally eliminates the pressure tightness capability of the die casting and significantly reduces the mechanical properties of the casting.

Methods for detecting flaws, or lack of acceptability, for die casting alloy are given in Reference (3). Flaw detection tests include (1) crack tests, using penetrants or chalk tests, (2) X-ray, (3) ultrasonics, (4) pressure tests, and (5) chemical etching (3).

It has long been recognized and reported that porosity may occur in solidified products. Biringuccio (6), in 1544, presented a written description of porosity as a potential manufacturing defect in castings. Die castings have a reputation of containing significant levels of porosity. Several recent studies and reports (7-9) have addressed the causes and measured the levels of porosity in die castings. Porosity in die castings is significant because it is directed related to or influences all of the quality measures listed in Table 1. Equivalently, the quality measures for die castings are all functions of the porosity level associated with the casting. Therefore, an extremely important quality assessment for die castings is their total or global porosity and information concerning the porosity distribution within the part. Reference 7 provides a comparison of several methods for characterizing porosity in die castings. The methods used in Reference 7 included (1) density measurement using Archimedes' method, (2) radiographic methods, including plate and tomography methods, (3) ultrasonic frequency attenuation method, and (4) metallographic examination for characterizing the location, amount, and type of porosity, and (5) the vacuum fusion method for determining the quantity of gases contained in die cast samples.

The specific relations between the quality measures listed in Table 1 and the varying porosity and gas contents of die castings are more qualitative than quantitative. It is certainly anticipated that the ability to provide pressure tightness is a function of the porosity level associated with the component. However, the lack of pressure tightness requires that the porosity be interconnected through the thickness of the component. While zero porosity would guarantee pressure tightness, the total or global porosity level could be relatively small and still allow for a interconnected path through the thickness of the component, thereby failing the pressure tightness requirements.

Several reports and papers (10-18) are available which document the effects of porosity on the mechanical properties of non-ferrous castings made by die casting and related casting processes. Essentially all of these references indicate that increasing amounts of porosity decrease the tensile properties, such as yield strength, ultimate tensile strength, and failure elongation. However, no single relationship between a given mechanical property and porosity has evolved or been developed for a reasonably broad class of alloys. Most of the researchers studying the effects of porosity on mechanical properties of castings report large scatter bands in their data, which are frequently attributed to difficulties in characterizing and quantifying the porosity level or distribution within a tested samples.

The ability to heat treat and/or weld die castings is primarily a function of the amount of gas contained within the sample. As discussed in Reference 9, the total porosity in die castings results from both shrinkage porosity and gas. To produce heat treatable and the highest strength component for the given alloy, the die caster must eliminate both shrinkage porosity, via directional solidification and liquid feeding procedures, and gas, via gate and venting design and elimination of gas sources. Die cavity design, including gate and vent locations, sizes and shapes, are fundamental to producing low porosity die castings. Methodologies for die design and the associated selection of appropriate operating parameters and procedures for producing acceptable quality die castings are discussed in References 5 and 19.

Modeling of the die casting processes allows the die casting engineer to predict and correlate the fluid flow and thermal fields in the die casting process with conditions which allow the production of acceptable quality parts. A number of numerical computer programs are available for modeling the thermal fields and fluid flows in 2 and 3 dimensions for the die casting process. Included in these computer programs are PROCAST, available from Universal Energy Systems; MAGMASOFT, available from MAGMA Foundry Technology; AFSOLID, available from American Foundrymen's Society, DUCTcast, available from DUCT CAD/CAM Systems.

Several of the above commercial computer programs for modeling the die casting processes were used as part of the University of Northern Iowa/Department of Energy die casting study (20). The current computer programs should allow the die casting engineer to predict the gate freezing time and, in turn, the shrinkage porosity distribution in die castings. The current computer programs do not yet provide predictions of the gas related porosity in die castings.

Die casting engineers and customers would also like to be able to predict and control the microstructural features in die castings. For example, the ability to control "skin" thickness over the entire die casting is extremely important in the production of pressure tight components. While the use of microstructural features, such as the secondary dendrite arm spacings in hypoeutectic aluminum alloys (21), is gaining acceptance and usage for components made by casting processes other than die casting, the die casting industry and customers thereof have not yet made microstructural features a significant part of their quality assurance procedures. This is an methodology and area which should be pursued by the die casting industry.

As die casting is an established manufacturing industry, a part quality definition and assurance methodology exists. However, new markets and a greater number of satisfied customers would result from the development of methods to reduce and better control the amount of porosity and contained gas in die castings. The reduction of porosity levels and contained gas contents would significantly and simultaneously improve all of the current quality measures for die castings.

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**MODELING THE THERMAL FIELDS IN THE
COLD CHAMBER DIE CASTING PROCESS USING MAGMASOFT**

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FOREWORD

This document has been prepared for the Engineering Research Center for Net Shape Manufacturing (ERC/NSM). This Center was established on May 1, 1986 and is funded by the National Science Foundation and the member companies. The focus of the Center is net shape manufacturing with emphasis on cost-effective manufacturing of discrete parts. The research concentrates on manufacturing from engineering materials to finish or near-finish dimensions via processes that use dies and molds. In addition to conducting industrially relevant engineering research, the Center has the objectives to *a*) establish close cooperation between industry and the university, *b*) train students, and *c*) transfer the research results to interested companies.

This report describes the OSU ERC/NSM experiences during the initial usage of the MAGMASOFT casting simulation software. The report includes a description of the capabilities of the MAGMASOFT program. The results of simulations of two casting and die systems, (1) the ERC's 2 mm thick flat plate die and casting system and (2) the ERC bowl die and associated castings are included. The report describes the requirements and procedures for geometry entry, mesh generation, simulations, and results analysis.

This report is published by the ERC/NSM at Ohio State, and has been made available for inclusion in the University of Northern Iowa(UNI)/North American Die Casting Association (NADCA)/Department Of Energy (DOE) casting simulation report. With the exception of the supportive information provided by MAGMASOFT presented in Appendix D, the opinions and views set forth in this report are those of the authors. The simulations conducted as part of this study were all performed using MAGMASOFT version 2.03, which is currently installed on an IBM Risc/6000 320E Workstation, with only 32 Megabytes memory, while MAGMA recommends at least 64 MB for simulation of the die casting systems.

Information about the ERC for Net Shape Manufacturing can be obtained from the office of the Director, Taylan Altan, located at the Baker Systems Engineering Building, 1971 Neil Avenue, Columbus, Ohio 43210-1271, phone 614/292-9267 or Fax number (614)-292- 7219.

EXECUTIVE SUMMARY

MAGMASOFT is a commercially available software package developed for simulating the filling, cooling, and solidification of castings. The MAGMASOFT program was made available to the OSU ERC/NSM in the Summer of 1992 as a tool for modeling the thermal fields and associated phenomena in die casting. The MAGMASOFT program, Version 2.03, was installed on an OSU IBM Risc/6000 320E Workstation equipped with 32 MB memory. The MAGMASOFT program has been used to model the thermal fields in the dies and castings of several ERC/NSM systems, including the dies and castings identified as (1) the 2 mm thick plate casting and (2) the bowl casting.

This report describes the procedures used in applying the MAGMASOFT program and system to model the thermal fields in the die casting systems (i.e., dies and casting) associated with the OSU ERC/NSM 2 mm thick plate and the bowl die casting systems. The capabilities and limitations of the MAGMASOFT program used in this study are described. The calculated temperatures associated with three thermocouples located in the 2 mm thick plate were within 2° C of the experimental values throughout the casting cycle. Less agreement between the simulated and experimental temperatures was obtained for the bowl casting, where the calculated temperatures were on average 30° C different than the observed at eight thermocouple locations. The lesser agreement between the simulated and experimental temperature histories for the bowl casting was attributed to the use of large dimension mesh in the vicinity of the thermocouple locations and the resulting uncertainty in the location of the simulated thermal response relative to the die casting interface and experimental thermocouple location.

While it was concluded that the MAGMASOFT program is a powerful tool to be used in support of the design and analysis of the die casting system, successful and effective use of the program requires (a) prior knowledge of the appropriate boundary conditions (i.e., heat transfer coefficients as functions of time and/or temperature at all interface locations) and (b) additional Workstation memory to allow for finer enmeshment and effectively record and recall the model results. These two requirements would be applicable for all modeling programs.

1.0 INTRODUCTION

MAGMASOFT is a commercially available software package developed for simulating the filling, cooling, and solidification of castings. The software has been used extensively in Europe, and is finding increasing usage in the U.S. The U.S. automotive industry has been a major user of the MAGMASOFT program. The software has been used to model sand casting processes, the Cosworth process, investment casting, and low and high pressure die casting processes. Descriptions of the MAGMASOFT program and its applications are given in References 1 through 8.

The MAGMASOFT program simulations of the filling, cooling, and solidification of the casting enable the user to better understand the filling and solidification patterns, the die and casting thermal conditions, the feeding ability of risers, the cycle timing, shrinkage, cooling rates, thermal gradients, and solidification times. The software can be used to aid in designing molds and provides insights into the occurrence and location of potential defects, such as shrinkage porosity and hot tears, in the castings.

The MAGMASOFT program offers many features which are necessary or helpful for the analysis of permanent mold and die casting processes. For example the program can simulate a single casting cycle, or it can simulate any number of sequential cycles. The analysis of sequential cycles allows the user to examine how quickly the die thermal conditions approach the quasi-steady state condition. The program can also be used to quantitatively define the effects of the die preheating and cooling line placement and extent of cooling on the thermal response of the dies and castings.

2.0 MAGMASOFT PROGRAM INFORMATION AND REQUIREMENTS

The following sections provide a general overview of the MAGMASOFT 2.03 version used in this study and the OSU ERC/NSM hardware platform used for its operation.

2.1 History of MAGMASOFT at the OSU ERC/NSM

In July of 1992, three graduate students from the OSU ERC/NSM went to MAGMA's office in Chicago and participated in a three day training course. By the end of July, MAGMASOFT Version 2.02 was installed at the ERC/NSM.

Upon acquiring the MAGMASOFT Version 2.02 at the ERC/NSM, the graduate students used the program to simulate castings and associated dies and molds for several of the ERC/NSM member companies. The Version 2.02 was relatively easy to use; however, long times were required for entry of the part geometry since the user had to enter all points with the mouse interactively at the computer.

In December, 1992 the upgraded MAGMASOFT Release 2.03 was installed at the ERC/NSM. The 2.03 version contained an updated postprocessor and a new preprocessor. The preprocessor was a complete change in the CAD system. A one day training was given at OSU on the new preprocessor in late December.

The new preprocessor redefined geometry entry for the system. The user could still enter the geometry interactively with a mouse, however, the use of the scripted commands and command files made part geometry entry much quicker and easier than with Version 2.02. These commands are described in more detail later in the report.

Since the Graduate student had no previous experience using the new geometry entry system, it was decided to document the learning and usage history for this version. This report provides an indication of the time and effort required to learn and use the Version 2.03 MAGMASOFT program.

2.2 MAGMASOFT Simulation Program Release 2.03

MAGMASOFT Version 2.03 is a stand alone simulation package, which allows for geometry entry, simulation, and viewing of results. It does not require a separate software package for geometry entry, mesh generation, and viewing of the results. The program consists of 6 main modules identified as:

- Project Management Module
- Preprocessor Module
- Enmeshment Module
- Simulation Module
- Postprocessor Module
- Data Management Module

Typically the user will create a project in the Project Management Module, enter the geometry in the Preprocessor Module, mesh the geometry in the Enmeshment Module, perform simulations in the Simulation Module, and then examine the results in the Postprocessor Module. The user can update or add thermophysical data at any time by using the Data Management Module.

2.2.1 Project Management Module

The project management module indicates to each module which version of the project is currently being worked on. The versions are numbered starting with v00. The project management module allows multiple versions of a project to study various design alterations to consider things such as different gating, and cooling. The module also tracks relevant information, such as casting weight, volume, and die temperatures.

The use of different versions allows the user to simulate the current casting, then use a newer version to study potential part modifications. If a newer version is not created, the results of the prior simulation can be overwritten if the simulation is run again.

2.2.2 Preprocessor Module

The preprocessor is a mini-CAD system, which allows the user to enter the casting geometry using solid models built from circles, polygons, and boxes. The

current system will also read command files to create the geometry. The command files are easy to create and consist primarily of part dimensions from the prints. The system can also perform solid rotations and sweeps. The user can also specify locations of simulated thermocouples for data collection.

2.2.3 Enmeshment Module

The enmeshment module requires few inputs from the user, and is easy to use. The user enters the thickness requirements of the elements along each of the coordinate axes, the accuracy level, and a few other optimization parameters, which are discussed later.

2.2.4 Simulation Module

The simulation module takes the mesh generated from the geometry and simulates fluid and heat flow phenomena. The simulation module also has the capability to simulate cyclic batch processes, such as die casting and permanent mold casting. The following sections give some details on the simulation aspects.

2.2.4.1 Mold filling and fluid flow module - MAGMAfill

The simulation allows the user to examine the filling sequence and calculate the temperature distribution of the melt as the melt cools while filling. Currently the solvers are limited to single phase laminar flow. In die casting the flow is frequently multiphase (entrained gas) and turbulent. However, the solver provides a reasonable approximation to the initial temperature distribution, which is often better than assuming a constant temperature in the casting cavity at the beginning of solidification. The module has four levels of accuracy, with 1 being the simplest:

1. Fast solver for bottom pouring conditions.
2. Navier-Stokes' equation neglecting temperature dependent viscosity.
3. Navier-Stokes' equation including temperature dependent viscosity.
4. Navier-Stokes' equation based on SOLA-VOF technology.
Temperature dependent viscosity and special boundary conditions to determine free surfaces are taken into account.

The solvers using the Navier-Stoke's equations require more computational time. Typical times on the OSU equipment have been approximately 5 hours for a mesh size of 800,000 control volumes with any of the Navier-Stoke's solvers.

2.2.4.2 Solidification module - MAGMASolid

The solidification module uses the thermophysical properties from the programs' database, and the initial temperature profile from the filling module if desired to calculate the thermal fields in the die and casting. The module also calculates feeding flows.

2.2.4.3 Batch module - MAGMASolid

The batch module manages the filling and solidification modules. The user can simulate a cyclic process such as die casting for as many cycles as desired. The user can indicate which cycles to perform: filling, solidification, or both. The user also specifies the die close and open times. The program will also transfer the die thermal profile from cycle to cycle.

2.2.5 Postprocessor Module

The postprocessor allows the user to review the results of the filling and solidification simulation. The user can view from different viewpoints (i.e. rotate the axis system), slice through the part, view different materials, enlarge specific areas, and view velocity vectors. The following output is available for display.

1. Filling Flow Pattern - view the filling at different time steps with velocity vectors.
2. Temperature Distributions - view the temperatures in the casting and die, at different times, and the solidification sequence. The solidification sequence is viewed with a color scale where red represents liquid, shades of green represents mushy metal, and blue represents solidified metal.
3. Criteria Functions - thermal gradients, cooling rates, solidification times, feeding, and shrinkage (Niyama) are available.
4. Cooling Curves - data collected from the thermocouple locations. This can also be used for modeling filling times, by specifying thermocouple locations in the casting cavity.

2.2.6 Data Management Module

The data management module contains the thermophysical data for mold, casting, cooling, heating, and surrounding materials. The thermophysical data includes the density, conductivity, specific heat, and viscosity as a function of temperature. The data file also contains the liquidus, solidus, and initial

temperature. The database also contains heat transfer coefficients for various boundaries or interfaces. The user has the freedom to use data in the database or enter data considered more appropriate or representative of their material. The MAGMASOFT thermophysical property database includes density, conductivity, specific heat, and viscosity as a function of temperature for 39 alloys and 28 die materials.

2.3 Hardware Requirements

The software is designed to work on most UNIX based engineering workstations. The workstation used at OSU consists of:

System Hardware

- IBM Risc/6000 320E Workstation
- 32 MB of Memory
- 2.9 GB fixed disk
- 19" High resolution color monitor
- 1/4" and 8 mm tape drives

System Software

- AIX Version 3.2 Operating System (UNIX)
- X Windows System Version 11 Release 4 (X11R4)

MAGMA recommends that systems using MAGMASOFT for die casting should have a minimum of 64 MB of memory. It should be noted that of the IBM Risc stations, the 320E model is considered one of the slowest models. A typical workstation configured with the appropriate memory and disk space will cost between \$25,000 - 35,000.

MAGMA typically uses Silicon Graphics machines, but has also installed the software on IBM, Sun, Hewlett Packard, DEC Stations, and Apollo machines.

3.0 MAGMASOFT SYSTEM INPUTS AND OUTPUTS

The following sections describe the information required for the MAGMASOFT program to perform the casting simulation. Both required and optional inputs, and defaults available for input are described. The input for each of the program modules is described.

3.1 Project Management Module

The project module mainly indicates to the other modules which project is currently being examined. If the project is a new project, the user only needs to enter a project name, and an optional comment if desired. For projects that have been created, the user can select the desired project from the list. The project manager will also allow the user to create new versions of the project. This allows the user to examine changes in the design without losing the original geometry or simulation data. Projects can also be deleted from the project manager.

3.2 Preprocessor Module

The preprocessor is the program's geometry or CAD system. Entry of the geometry is relatively easy; however, it is one of the more time consuming modules of software for the user. The following section describes the highlights of the system. Some further detail is given later with the description of the entry of the OSU die castings.

The user has two options for entering part geometry. The interactive screen input can be used, or a scripted test file describing the geometry can be used. The scripted method was used to enter the OSU dies. The scripting allows the user to take dimensional information directly from the part drawings. To enter the same on the screen would require locating the mouse cross hairs at the precise dimensions. The script files are convenient, because if there is a mistake in the geometry, the script file can be easily edited. To edit specific points on the screen is not available.

When the preprocess is started the screen appears with numerous options. The options have various functions. When the screen first appears, there are four

views showing, the y-x, y-z, z-x, and a 3-D view. Some of the functions are briefly described in the following sections.

Circle, Box, Polygon

These are the basic building tools. The user can make cylinders, cones, rectangular boxes, boxes of varying sizes, and any shaped polygons. The user also has the ability to use commands to enter spheres and cubes.

Hidden Line Mode

This button allows the user to change the view to the hidden line mode. This function, in conjunction with specifying which materials to show, can aid in obtaining a 3-D view of the geometry.

Zoom

This feature indicates to the system to zoom in or out. The user then can use the mouse to specify the zoom location.

View

The view command allows the user to switch to viewing only one of any of the four views. This is helpful if geometry is being entered interactively.

Units

This simply lets the user select the basic units. The units available are millimeter, centimeter, inch, foot, and meter.

Grid

The grid is useful if the user is interactively entering the geometry. The grid can be turned off or on, and set to the desired increment. For example, a grid of one would cause the mouse to move in units of one across the screen.

Accuracy and Angle

This section allows the user to set the accuracy and angle. The accuracy defines how many segments curved surfaces are broken into. The angle determines the amount of revolution.

Materials

This is important to understand for future modules. In the preprocessor the user only classifies the materials by function, such as "Cast_Alloy" or "Cooling." Typical casting materials are prespecified and additional "User_Defined" materials are available. Later in the simulation module, these generic categories are assigned specific material, such as "Cast_Alloy" might be assigned A380.

Control Points

This section lets the user set the locations where temperature data is to be taken. The user can specify up to 24 points.

Revolution

This command allows the user to specify an arbitrary axis of rotation, the 2-D object to rotate, and the amount of rotation. Any of the three primitive shapes can be rotated.

Sweep

The sweep command is similar to the rotate, except that the user specifies a 3-D trajectory over which the selected object is moved by entering the x, y, and z coordinates, by either a scripted command on two different views in the CAD system. An example would be a cylinder in the shape of an S.

The system also has a number of functions for filing, selecting objects, screen color, deleting selected objects, and redisplaying. Other functions such as to rotate, to scale, and to move are available through the command input section at the bottom of the screen.

3.3 Enmeshment Module

The enmeshment module creates the FDM mesh based on the geometry entered in the preprocessor. The program divides the geometry into non-overlapping elements called control volumes (CV's). The CV's are essentially small bricks with edges parallel to the given Cartesian coordinates.

The enmeshment module requires six parameters. These 6 parameters allow the user significant influence over the mesh. The parameters are:

Minimum Wallthickness	minimum wall thickness
Minimum Subdivision Elementsize	minimum element size into which the object is subdivided
Mesh Accuracy	accuracy of the mesh
Mesh Design Options	
Angle	angle changes
Smooth	smoothing factor
Ratio	length-to-width ratio

Table 3.3-1 Enmeshment Parameters

The first step of the meshing algorithm is to check the angles between any three adjacent points. If the angle that is made from these three points with respect to a straight line is less than the specified *Angle*, the middle point is eliminated.

The program then puts a "mark" on the x, y, and z axis from all points used in the geometry. The program then checks the distance between these marks. If the distance between any of the marks is less than the *Minimum Wallthickness*, one of the markers will be eliminated. Special rules are applied near the boundary of the casting.

The distance between the markers is then subdivided. The *Mesh Accuracy* parameter controls the number of elements that will be placed equidistant between the markers.

The *Minimum Subdivision Elementsize* is then used to check the distance between each marker. If the distance between any markers is smaller than the specified size they are eliminated.

The last two parameters, the *Smooth and Ratio*, are important in forcing the geometry to be closer in shape to the actual geometry. This can influence the

accuracy of the filling. The *Smooth* controls the rate of change of the shape of the elements. It compares the ratio of the length of an element to its neighbor. If this ratio is too large, the larger element is subdivided. The *Ratio* compares the ratio of the x, y, and z thicknesses for a given element. If this ratio is too large, the element is subdivided. For example, a *Ratio* value of 1 would produce cubic elements.

These parameters are easy to use. The user really only needs to have an idea about the minimum thickness of the features of the geometry. For example, in die casting, a thin section of the casting is typically the gate. If the user had a 1 mm gate, and put a *Minimum Wallthickness* of 2 mm in that direction, the gate would vanish. Similar difficulties can arise in thin curved walls. The elements can become disconnected if the parameters are not properly specified. This will cause part of the casting not to fill properly in the simulation.

All of the parameters are required and all offer defaults. The defaults are listed below:

Minimum Wallthickness	5 mm
Minimum Subdivision Elementsize	5 mm
Mesh Accuracy	5
Mesh Design Options	
Angle	60
Smooth	60
Ratio	60

Table 3.3-2 Enmeshment Parameters Default Values

These parameters are typical of values used for simulation of sand casting; however, they usually need to be changed for die castings. For example, the *Minimum Wallthickness* and *Minimum Subdivision Elementsize* values are too large for many of the die casting features.

The module also has a calculate feature. The module will quickly (less than a few seconds) tell the user the number of elements in each coordinate direction, and the total number of elements. The user can then adjust the parameters until the desired mesh size is obtained. One objective in changing the mesh size is to impact the simulation time. Larger dimension meshes, or equivalently smaller number of elements, run faster.

3.4 Simulation Module

The simulation module has two main functions, the filling simulation and the solidification simulation. The first input by the user is the production method. MAGMASOFT currently offers a *Shape* module and a *Batch* module. The *Shape* module is for single castings. The *Batch* module is for batch, or cyclical casting, such as permanent mold or die casting.

If the *Shape* is selected, the user can select to perform a filling simulation, solidification simulation, or both. If the *Batch* is selected, the user must select the number of cycles to simulate, which cycles to perform filling simulations, which to perform solidification simulations, and the die open and close times for the casting cycle. The following sections describe the information for the material specifications, filling module, and solidification modules.

3.4.1 Material Data

The material data entry section consists of several screens. The first screen asks for assignment of materials to materials group specified in the preprocessor and their initial temperatures, and the second asks for the heat transfer information. These screens interface with the data management module, described later.

The first screen lists all of the generic material groups specified by the user in the preprocessor. The user must select each material group and assign a material to it from the material database. For example, when a material group is selected, such as "Cast_Alloy," a subscreen appears with 39 potential materials from the database (there are 28 different mold materials in the database). The user makes a selection from this list, and specifies the initial temperature. The database contains a default initial temperature. However, some materials can be both mold materials and casting alloys and the default initial temperature is typically above the liquidus temperature.

The user must then specify the heat transfer coefficients. The screen shows all of the interfaces that were created in the preprocessor. A heat transfer coefficient must be specified for each interface. When the user selects a specific interface, a subscreen appears listing the heat transfer coefficient files in the database. These files are often functions of temperature, but the database also contains a set of constant values with respect to temperature.

3.4.2 Filling Data

The filling data entry requires seven inputs, which can all be defaulted. The user must select which solver level the software is to use. Solver 1 is the default solver. The four basic solvers are:

- Solver 1 Fast solver for bottom pouring conditions
- Solver 2 Navier-Stokes' equation neglecting temperature dependent viscosity
- Solver 3 Navier-Stokes' equation including temperature dependent viscosity
- Solver 4 Navier-Stokes' equation based on SOLA-VOF technology.
Temperature dependent viscosity and special boundary conditions to determine free surfaces are taken into account

Next the filling dependency is entered as either time based or flux based. If time based is selected, the user must enter a filling time (for example 30 milliseconds for a die casting). There are also three flux parameters. The value of the flux in the x, y, and z direction (m^3/s) is entered if the filling is based on flux. If the filling is based on time, these inputs indicate the direction of filling, and a value of unity, 1, is entered.

The last input to this module determines which results will be saved. There is an automatic default, which saves data files at every 10 mass percent filled. The user can also select to save by percent filled, or at a given time increment, or a combination of both percent filled and time.

3.4.3 Solidification Data

The solidification requires nine inputs; however, most are defaulted and do not need to be changed. The first input indicates whether data should be taken from the filling simulation, or assumed to be at a constant temperature. This is simply a "yes" or "no" option with the later being the default.

If the casting design includes a feeder, the program can calculate the feeding flow. There are two feeding inputs required. The first input is simply whether to calculate feeding flow or not. Again, this is simply a "yes" or "no" option with the later being the default.

If the user specifies "yes" to calculating feeding flow, the amount of feeding flow needs to be specified. The value is entered as a fraction solid (in %) of the

solidification interval, for which macroscopic feeding is no longer possible. The default value is 0.0%. This value is alloy dependent and is between 25 and 35% for aluminum alloys.

The program has the flexibility to control when the solidification simulation stops. There are three options: automatic, time, and temperature. The default is automatic. This stops the solidification when all sections of the casting are below the solidus temperature. The time and temperature based will stop the simulation based on an entered time or temperature respectively.

Much like the filling solver, the solidification solver has four levels. These levels differ primarily by the method and size of the time stepping increments. Fast is the default solver. The following describes the four categories:

Fast	the fastest solver
Efficient	slightly more accurate than "Fast"
Advanced	faster than "Accurate" and is the second most accurate mode
Accurate	the highest possible accuracy

There are two input fields for criterion temperatures. The first is defaulted to the solidus temperature plus 10 percent of the solidification range. This value is used for calculation of the Niyama criteria (Niyama, 1981, 1982). The second is defaulted to the liquidus temperature plus 2 °C. This value is used to calculate the cooling rate and the temperature gradient. Either of these default values can be changed.

As in the filling module, the final input determines when results will be saved. There is an automatic default, which saves at 10, 50, and 90 percent solidified. The user can also select to save by percent solidified, a given time increment, or a combination of both percent solidified and time.

3.5 Postprocessor Module

The postprocessor provides a way to examine the mesh, and the results of the filling and solidification simulations. The postprocessor allows the user to view 3-D images of any material(s) (i.e. casting, mold, cooling media) during the filling and solidification. The postprocessor also shows the results from the various criteria functions listed below:

1. Thermal gradient
2. Cooling rate
3. Total solidification time
4. Solidification time – Liquid to Solidus
5. Feeding criteria
6. Niyama criteria

The postprocessor also allows the viewing of cooling curves from the locations specified in the preprocessor. These curves are useful in die casting and permanent mold casting to understand when the dies reach the quasi-steady state thermal condition.

When entering the postprocessor module, the user must specify to examine the filling and solidification results, the mesh, or the cooling curves. For the mesh the user can then select which materials to view. The default is to view the casting only.

If the user selected to view the results files, the first menu allows the user to select which result file is to be examined. This can either be a filling, a solidification, or a criterion file. There is no default result file, one must be selected. In the same menu, the user can select which materials are to be displayed, the default for this again is the casting only. A third selection on the menu is the color scale. The default is a continuous color scale. The user can also use a scale where liquid metal is red, mushy metal is a scale of green, and solid metal is blue. This type of scale is useful for examining the solidification sequence. At this point, the result is displayed.

When either the mesh or one of the result files are displayed, a graphics menu also appears. This menu allows the user to perform rotations, slice through the part, clip and enlarge the object, and show velocity vectors for filling. The user can also choose to view the previous or next sequential result file.

Another type of output is the cooling curves. When this is selected, a menu appears and the user can select one or more of the cooling curves from the locations specified in the preprocessor. The user then selects to view either results during filling, solidification, combination (filling and solidification), or batch. The batch is for multiple cycles, such as in die casting. The user also has the option to change the x and y scales; however, default values are selected based on the range of the data in the files.

3.6 Data Management Module

The software requires thermophysical data for all materials involved in the process. MAGMA provides an extensive database of material properties as a function of temperature with the software, which can be used immediately without modification. MAGMA continues to add material values to the database. MAGMA obtains their data by experimental studies. The user can get data on specific alloys from various sources, such as the Thermophysical Properties of Matter (9) or the Purdue University Center for Information and Numerical Data Analysis and Synthesis (CINDAS) (10), and add those data to the computer database.

The Data Management Module provides an easy means of viewing the current data, both in tabular and graphic forms, editing the data, and entering company proprietary data for in-house materials. The module contains two main sections, one for the thermophysical data for the materials, and one for the heat transfer coefficients between the materials.

The thermophysical section is for entering or editing the thermal conductivity, density, viscosity, and the product of density and specific heat as a function of temperature. The latent heat of fusion is included in the product of density and specific heat data. The data files also contain the liquidus, solidus, and initial default temperature, which can be changed as input in the simulation module. The material group can also be defined in this section. The material groups are assigned in the preprocessor. For example, the cast object is called "Cast_Alloy" in the preprocessor, and then assigned a specific material in the simulation module. This allows the user to easily switch materials from any of the wide selection in the database.

The heat transfer coefficient section is for entering the information concerning the heat transfer between two materials. The heat transfer coefficient can be entered as a constant value, function of time, or a function of temperature.

Another feature of the database is the memo file attached to each material. This is a space provided identifying the source of the data. Also recorded in the memo space is the name of the last user to update the file and on what date.

4.0 DIE AND CASTING INPUT DESCRIPTIONS

The die casting simulations performed using MAGMASOFT were conducted with two die castings currently used at the ERC/NSM. These dies are briefly described below and CAD drawings have been included in Appendix A.

4.1 2 mm thick Plate Die and Casting

This part is a relatively simple plate die casting. It consists of a 2 mm thick plate with a fan gate and three overflows. The die has also been equipped to hold four thermocouple probes. The die has been used in ERC/NSM studies of the effects of die spray and cooling on part quality. The plate die and resultant castings have been used in previous ERC/NSM studies, as reported in Reference 11.

4.1.1 Geometry Entry

The geometry entry of the die and casting for the 2mm thick plate was straight forward. The part and gating are easily represented by geometrical shapes. The following briefly describes the steps taken to enter the geometry using MAGMASOFT's internal CAD system. The geometry has been entered using a scripted file. The commands have also been included with the description. The complete description has been included in Appendix B.

The plate was the first geometry that was entered into the MAGMASOFT CAD system. The times indicated are the combined times to study the drawing, to read the coordinates from the drawing, and enter them into the system. The times were somewhat long since this was a new system. The times were logged in a lab notebook along with project notes, as the geometry was being entered. The table below summarizes the time required to enter the die geometry.

Using the calculate feature of the meshing program, the parameters were adjusted until an appropriate mesh was obtained. The adjustment of these parameters took less than 5 minutes, and the generation of the mesh less than a minute.

Minimum Wallthickness	
x	0.5 mm
y	1 mm
z	1 mm
Minimum Subdivision Elementsize	
x	0.5 mm
y	1 mm
z	1 mm
Mesh Accuracy	
x	3
y	3
z	3
Mesh Design Options	
Angle	5
Smooth	3
Ratio	10

Table 4.1-2 Enmeshment Parameters

Figure 4.1-2 shows the meshing of the casting.

4.1.3 Process Parameters

The program asks for a number of process parameters in the main simulation module. The following describes the parameters used in this simulation. The thermophysical properties of the casting, die, and cooling systems were accepted as the default values from the MAGMASOFT program. The heat transfer coefficients characterizing the casting-die interface, the die-air interface, and the die-cooling lines were obtained from C. Mobley. The input heat transfer coefficients were based on experimental values previously determined at the ERC/NSM and were the same values as provided to all parties participating in the modeling activities. The process parameters, such as total cycle time, fill time, die-casting contact time, spray duration and initiation times, and initial temperatures, were provided by C. Mobley as input to the MAGMASOFT program.

MAGMApost

Return

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MAGMA

Information

☐ Cast Alloy ☐ Insulation ☒ Cooling Ch ☐ Inlet
☐ Core Band ☐ Chills ☒ User Def1 ☐ Feederneck
Hold Band Perm Hold User Def2 Feeder

Project: H140
Version: v00
Number of CV's: 70380
Thu Apr 0 13:51:13 1993

CAD Material Classification	Assigned Material	Initial Temperature
Casting Alloy (1 1)	A390	650°C
Permanent Mold (6 1)	H-13	100°C
User Defined 1 (8 1)	H-13	100°C
Cooling Media	Water	27°C

Table 4.1-3 Material Classification

Interface	Heat Transfer Coefficients (W/m ² °K)
A390 - H13 (Die-Casting)	8,000
H13 - Water (Die-Coolant)	1,500
H-13 - Air (Die-Air)	0.9
H13 - H13 (Die-Die)	2,000

Table 4.1-4 Heat Transfer Coefficients

Process Parameter	Value
Fill Time	28 ms
Cooling Lines	
Temperature	27 °C
Flow Rate	2.25 gpm
Melt Temperature	650 °C
Die Preheat	100 °C
Biscuit Thickness	16 mm
Cycle Time	
Closed	5 sec
Open	35 sec

Table 4.1-5 Process Parameters

While the user can input fill times of their choice, the subsequent reporting of the fill time on the computer screen will be to the nearest 10 milliseconds. Thus, the above input value of 28 milliseconds is reported thereafter as 30 milliseconds fill time.

The table below shows the simulation cycle time, percentage solidified, and the actual time from the start of the simulation. Note that the percent solidified represents both the casting and biscuit.

Simulation Cycle Time (sec)	Percent Solidified
0.00	0
0.12	42
0.40	60
1.60	82
3.48	95
5.00	99
6.08	100
40.00	End of Cycle

Table 4.1-6 Solidification Simulation Times

The times reported for percent solidified are based on a 16 mm length biscuit. The biscuit region of the casting is the last section to complete solidification. Use of different length biscuits would alter the computed time for complete solidification.

The total time for computing the solidification for one cycle was 113 minutes. Most of this time was calculating the die cooling during the open part of the cycle. This time would probably decrease significantly with increased memory. The computer time included CPU time and time for paging of the virtual memory. With the low system memory, the CPU utilization is typically less than 30 percent of the time.

Figures 4.1-3 to 4.1-5 show the solidification sequence of 42, 60, and 82 percent respectively. The red indicates liquid, blue solid, and green mushy metal. The solidification sequence indicates that there should be some shrinkage porosity near the gate and overflows. Fig. 4.1-6 shows the solidification times. Figure 4.1-7 shows the thermal distribution in the die at the end of the cycle.

Similarly for the filling calculations, the table below shows the calculation times based on file write times.

- MAGMApost
- Return
- Settings
- User Scale
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- Clipping
- Previous
- Same
- Next

Temperature [C]

- 665
- 652
- 640
- 628
- 615
- 602
- 590
- 578
- 565
- 552
- 540
- 528
- 515
- 502
- 490

Project: F1401P
 Version: 900
 Cycle: 5
 Number of CV's: 105040
 Thu Apr 6 13:37:22 1995

Information

T.Dine =
 Percent Solidified = 41.96 (%)

MAGMA

MAGMApost

Return

Settings

User Scale

Viewpoint

Clipping

Previous

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Temperature [C]

665

652

640

628

615

602

590

578

565

552

540

528

515

502

490

2

1

0

MAGMA

Project: 111111

Version: 1.00

Cycle: 5

Number of (V.S.): 10000

Time: Apr 01 11:00:23 1993

Information

T ime

Percent solidified = 60.07 [%]

MAGMA

MAGMapost

Return

Settings

User Scale

Viewpoint

Clipping

Previous

Same

Next

Uld. Time
Uxc]

3.50

3.26

3.01

2.77

2.53

2.29

2.04

1.80

1.56

1.31

1.07

0.83

0.59

0.34

0.10

Project Title

Version: v0.0

Cycle: 5

Number of CV's: 7058.40

Time: Apr 8 14:02:40 1993

Information

MAGMA

MAGMA

Simulation Cycle Time (sec)	Percent Filled
0.000	0
0.012	40
0.015	50
0.017	60
0.021	70
0.023	80
0.025	90
0.028	95
0.030	100

Table 4.1-7 Filling Simulation Times

The filling simulation is based on the filling of the casting cavity (i.e., considers only the casting cavity from the gate to the overflows) and does not include considerations of the flow in the shot sleeve, biscuit, and runner portions of the system.

The total computer time for filling for one cycle was 617 minutes, or 10.28 hours. The filling simulation is more complex and requires more memory for calculations. Figures 4.1-8 to 4.1-11 show the filling sequence of 40, 60, 80, and 100 percent fill.

Figure 4.1-12 shows the resulting temperatures versus time for five cycles for the thermocouple (cooling curve) locations specified in the preprocessor. They were located 0.5, 1.5, and 12.5 mm from the surface of the casting as indicated the numbers in Fig. 4.1-1.

4.1.4 Results

The following section provides a brief overview of the results of the simulation. These results can be compared with the experimental data provided.

- **Solidification Time**

The program allows the user to examine the solidification sequence, identify the last regions to solidify, and solidification times. The following reports

MAGMA

MAGMA

MAGMA

MAGM/Post

Return

Settings

User Scale

Viewpoint

Clipping

Previous

Save

Quit

Temperature [C]
 622.5
 620.0
 617.5
 615.0
 612.5
 609.9
 607.4
 604.9
 602.4
 599.9
 597.4
 594.9
 592.4
 589.9
 587.4

Stop

Information

Time
 Percent filled 100.00 [-]

Output: 133.dat
 Version: VII
 Number of CVs: 612000
 Run Mar 21 11:36:40 1993

MAGMApost

Quit

Help

3D-Results

3D-Mesh

C-Curves

MAGMA

Thu Apr 8 14:05:46 1993
 Reading cooling curve data ...
 ended
 Reading cooling curve data ...
 ended

Information

1: ■	4: ■	7: ■	10: ■	13: ■	16: ■	19: ■	22: ■
2: ■	5: ■	8: ■	11: ■	14: ■	17: ■	20: ■	23: ■
3: ■	6: ■	9: ■	12: ■	15: ■	18: ■	21: ■	24: ■

the solidification time at the thermocouple location, the hot spot in the die (located slightly above the gate area), the biscuit, and at the gate:

<u>Location in Casting</u>	<u>Solidification Time (sec)</u>
Gate	0.27
Near Thermocouple	0.34
Last to Freeze/Hot spot	0.41
Biscuit	6.08

The analysis of the solidification sequence for the 2 mm thick plate indicates that the gate (70 mm wide and 1.5 mm thick) completes solidification in 0.27 seconds while the last region of the plate to freeze requires 0.41 seconds. These calculated data indicate that a portion of the plate casting will not solidify while connected to the pressurized liquid source associated with the biscuit. This indicates that some solidification shrinkage is expected to occur in the plate casting (12). The percent solidified versus cycle time reported in Table 4.1-6 is based on the complete casting system, i.e., the casting cavity, runner, and biscuit.

Data relating the fraction solidified in the casting cavity up to and including the time of gate freezing would allow the calculation of the total percentage shrinkage porosity expected in this casting. While the determination of the percent shrinkage porosity in a given casting cavity would enhance the program's value, it was not easily obtained using this version of the MAGMASOFT program. The computer program also does not consider the formation of porosity via gas entrapment or generation.

- **Thermal Characteristics**

The program tracks the thermal conditions in the die. The user can view the die and examine graphically the thermal fields, or through the use of the cooling curves the program will track the temperatures at a specific location.

The cooling curve shown in Fig. 4.1-12 shows the thermal profiles at the thermocouple location specified in the preprocessor. The maximum temperatures indicated at a depth of 0.5 mm from the casting surface are near 226° C for the fifth cycle. The die is essentially isothermal at a temperature of 115° C at the onset on a casting cycle after it has reached the quasi-steady state operating condition. The temperature-time records

indicate that the die has nearly reached the quasi-steady state thermal condition after 5 cycles. The thermal data associated with the fifth cycle are not expected to differ significantly (more than 10° C) from the quasi-steady state value. The initial die temperature was set at 100° C in the simulation program, and not room temperature, in order that the system reached the quasi-steady state condition in fewer cycles.

The fifth cycle was the only one to include filling. It appears that at this particular location, the effect on filling only increased the 0.5, and 1.5 mm locations by approximately 10° C. The filling typically only effects the thermal response of the die in the vicinity of the die-casting interface.

The calculated cooling curve data agree quantitatively with the experimental cooling curve data. Experimental data reached a maximum temperature of 228° C at a depth of 0.5 mm below the surface. The coolest temperature at this location was 115° C. The table below summarizes the maximum and minimum simulated and experimental temperatures:

Location (mm)	Experimental Data		Simulation Data	
	min °C	max °C	min °C	max °C
0.5	110	228	115	226
1.5	115	200	125	202
12.5	115	135	120	136

Table 4.1-8 Cooling Curve Comparison

The simulation and experimental values for the minimum and maximum temperatures at the three thermocouple locations agree closely. In most of the cases, the agreement is within the assumed experimental temperature error of ± 2 C.

- **Filling Characteristics**

As expected there is some cooling of the liquid alloy as it fills the casting cavity. In the filling analysis, the temperature of the liquid alloy at the gate was taken to be 625 C. Based on that gate temperature, and using the level 4 filling solver, the liquid near the top of the casting cooled 25° C (i.e., to a temperature of 600 C) during filling. Most of the liquid in the casting cavity

cooled 10° C during filling. The filling analysis indicates that there should be no filling related defects, such as cold shuts for the specified system.

Figure 4.1-12 shows the thermal history in the die at the three locations results with filling conducted only on the fifth cycle. There is a slight difference in the temperature profile. The point nearest to the surface shows that there is some extra heat being removed from that region as the metal flows past.

4.2 Bowl Die

This die has been used in an experimental program to determine heat transfer coefficients as a function of alloy superheat, intensification pressure, and die surface finish (13). The die has also been used to study the effect of venting and vacuum assisted venting on part quality. The die was modified for the vacuum assisted study. The die and part configuration used for this simulation study was that associated with the vacuum assisted study.

4.2.1 Geometry Entry

The geometry entry of the die and casting for the bowl die was straight forward. The part and gating are easily represented by geometrical shapes. The following briefly describes the steps taken to enter the geometry using MAGMASOFT's internal CAD system. The geometry was entered using a script file. The commands have also been included with the description. The complete description has been included in Appendix C.

The times indicated are the combined times to read the coordinates from the drawing and enter them into the system. The table below summarizes the time required to enter the die geometry.

Die Feature	Time (minutes)
Die	8
Bowl Casting	32
Gate and Runner	29
Biscuit and Inlet	3
Cooling Lines	20
Cooling Curves	10
Total Time	102

Table 4.2-1 Geometry Entry Times

The total time required to review the drawings and enter the information into the CAD system was approximately 102 minutes. This was the second part that was entered by the user into the CAD system. Although the time is slightly slower than the 2mm thick plate, the complexity of the part is greater.

Figure 4.2-1 shows the three dimensional view of the casting as it is represented in the MAGMA preprocessor. The red is the casting, the blue is the cooling lines, the brown and orange represent the two die halves, and the numbers represent the cooling curve locations.

4.2.2 Mesh Generation

The parameters listed in the table below resulted in a mesh of 778,128 elements, and 23,800 filling elements. Using the calculate feature of the meshing program, the parameters were adjusted until an appropriate mesh was obtained. The adjustment of these parameters took less than 5 minutes, and the generation of the mesh less than a minute.

Minimum Wallthickness	
x	1 mm
y	1 mm
z	1.2 mm
Minimum Subdivision Elementsize	
x	1 mm
y	1 mm
z	1.2 mm
Mesh Accuracy	
x	1
y	2
z	1
Mesh Design Options	
Angle	5
Smooth	3
Ratio	20

Table 4.2-2 Enmeshment Parameters

Figure 4.2-2 shows the meshing of the casting.

4.2.3 Process Parameters

The program asks for a number of process parameters in the main simulation module. The following describes the parameters used in this simulation.

CAD Material Classification	Assigned Material	Initial Temperature
Casting Alloy (1 1)	A380	625°C
Permanent Mold (6 1)	H-13	200°C
User Defined 1 (8 1)	H-13	200°C
Cooling Media	Water	27°C

Table 4.2-3 Material Classification

Heat Transfer ($\text{W/m}^2 \text{ } ^\circ\text{K}$)	
A380 - H13	42,000
H13 - Water	800
H13 - Air	0.9
H13 - H13	2,000

Table 4.2-4 Heat Transfer Coefficients

MAGMapost

Return

Settings

User Scale

Viewpoint

Clipping

Previous

Same

Next

1
2
3

Project: Offroad

Version: V10

Rendered CV: 8/1/8

In May 11 08:33:03 (PM)

Information

Cast Alloy

Core Sand

Mold Sand

Insulation

Chills

Perm Mold

Cooling Ch

Exhaust Doff

Upper Doff

Inlet

Peelbinock

Pocket

MAGMA

372

Process Parameter	Value
Fill Time	40 ms
Cooling Lines	
Temperature	27 °C
Melt Temperature	625 °C
Die Preheat	200 °C
Biscuit Length	16 mm
Cycle Time	
Closed	10 sec
Open	20 sec

Table 4.2-5 Process Parameters

The following numbers represent approximate times to calculate the solidification results. These figures are based on the file write times and not actual CPU time, which for the current IBM machine is typically much less because of memory limitations. The table below shows files which were saved for the percentage solidified and the time from the start of the simulation.

Simulation Time (sec)	Percent Solidified
0.00	0
0.18	21
0.43	41
1.00	60
3.45	80
7.07	90
10.00	95
28.67	99
30.00	End of Cycle (99+)

Table 4.2-6 Solidification Simulation Times

The total computer time for modeling solidification for one cycle was 487 minutes, or 8.1 hours. Most of this time was calculating the die cooling during the open part of the cycle.

Figures 4.2-3 to 4.2-5 show the thermal fields in the casting when the casting is 41, 60, and 80 percent solidified, respectively. The red indicates liquid, blue solid, and green mushy (i.e., two phase) material. The pictured solidification sequence indicates that shrinkage porosity is expected to occur in the six post areas and in

MAGMAPost

Return

Settings

User Scale

Viewpoint

Clipping

Previous

Same

Next

Temperature [C]

580.0

575.7

571.4

567.1

562.9

558.6

554.3

550.0

545.7

541.4

537.1

532.9

528.6

524.3

520.0

Stop

MAGMA

Information

Time

Percent solidified = 40.33 [%]

Project: 011001

Version: 0.01

Cycle: 5

Runout of CV: 8.7/10/28

1st May 11 08:33:03 (UTC)

MAGMapost

Return

Settings

User Scale

Viewpoint

Clipping

Previous

Save

Next

Temperature [C]
 580.0
 575.7
 571.4
 567.1
 562.9
 558.6
 554.3
 550.0
 545.7
 541.4
 537.1
 532.9
 528.6
 524.3
 520.0

Stop

Project: Offroad
 Version: v1.0
 Cycle: 5
 Number of Cycles: 1/10/20
 In May 11/10/2016 1991

Information

Time

Percent solidified =

60.11 [s]

MAGMA

MAGI/Adjust

Return

Settings

User Scale

Viewpoint

Clipping

Previous

Same

Next

Temperature [C]

580.0

575.7

571.4

567.1

562.9

558.6

554.3

550.0

545.7

541.4

537.1

532.9

528.6

524.3

520.0

Two

MAGMA

Information

T.100

Percent not lifted = 80.21 [-]

Project: 03100

Version: 000

Cycle:

Rendered: CVS 7/6/94

For May 11th 15 00, 1994

the box area (i.e., the flat portion of the wall where the thermocouple probes were located) of the casting. Fig. 4.2-6 shows the solidification times. Figure 4.2-7 shows the thermal distribution in the die at the end of the cycle.

Table 4.2-8 lists the percent fill of the cavity as a function of the filling time. The percent fill in the cavity is directly proportional to the filling time, with complete filling in the specified 40 milliseconds.

Percent Filled	Simulation Time (sec)
0	0.000
20	0.008
30	0.012
40	0.016
50	0.020
60	0.024
70	0.027
80	0.032
90	0.036
95	0.038
100	0.040

Table 4.2-7 Filling Simulation Times

The total computer time for modeling the filling during one cycle was 708 minutes, or 11.8 hours. Figures 4.2-8 to 4.2-11 show the liquid front location and the associated temperatures within the liquid for the specified filling times, or equivalently, the percent fill for the 40, 60, 80, and 100 percent fill cases, respectively.

Figure 4.2-12 shows the resulting thermal histories within the die at several specified locations and depths for five cycles. The experimental and simulated thermocouples constituting the thermocouple probes at the bowl wall were located 0.5, 1.5, and 12.5 mm from the surface of the casting on both sides of the box area of the casting, as indicated by the numbers in Fig. 4.2-1. The seventh and eighth locations were by the gate and floor of the casting, respectively.

MAGIApost

Return

Settings

User Scale

Viewpoint

Clipping

Previous

Same

Next

Use |

5.00

4.65

4.30

3.95

3.60

3.25

2.90

2.55

2.20

1.85

1.50

1.15

0.80

0.45

0.10

Close

Project: 010001
 Version: 010
 Cycle: 5
 Number of CVS: 7/01/28
 1st May 11 10:29 (b.1111)

Information

MAGMA

5 km/s

2
1
y

- MAGMapost
- Return
- Settings
- User Scale
- Viewpoint
- Clipping
- Previous
- Same
- Next

Temperature
[C]

- 629.2
- 619.4
- 616.5
- 613.7
- 610.9
- 608.1
- 605.2
- 602.4
- 599.6
- 596.8
- 593.9
- 591.1
- 588.3
- 585.5
- 582.6

Stop

MAGMA

Information

Time
Percent Filled 39.99 [-]

Project: C:\magma
Version: v8.1
Cycle: 1
Number of CV's: 75 (R20)
1st May 11 08:14 (1993)

MAGMapost

Return

Settings

User Scale

Viewpoint

Clipping

Previous

Save

Next

Temperature

171

310

299

289

278

268

257

246

236

225

214

204

193

183

172

100

MAGMA

Information

Project: 1000000

Version: 0.001

Cycle: 5

Number of (X,Y): 701/20

In May 14 1991 15 1991

Percent: 80.11111111

99.14 1.1

5 km/s

z
y
x

MAGMApost

Return

Settings

User Scale

Viewpoint

Clipping

Previous

Same

Next

Temperature [C]

622.1

619.2

616.4

613.5

610.6

607.7

604.8

602.0

599.1

596.2

593.3

590.4

587.6

584.7

581.8

Information

Project: Ollind

Version: v11

Cycle: 1

Number of CV's: 75

File: 11025337

Percent filled: 79.36

Unit: [s]

MAGMA

5 km/s

Z
Y
X

MAGMApost

Return

Settings

User Scale

Viewpoint

Clipping

Previous

Same

Next

Temperature [C]

622.2

619.3

616.4

613.6

610.8

608.0

605.1

602.3

599.4

596.6

593.8

590.9

588.1

585.2

582.4

Show

MAGMA

Information

Time
Percent filled 59.99 [%]

Project: Ottawa
Version: v11
Cycle: 1
Number of CVS: 13450
In May 11th 2407 1991

MAGMApost

Quit

Help

3D-Results

3D-Mesh

C-Curves

MAGMA

Information

1: ☐	4: ☐	7: ☐	10: ☐	13: ☐	16: ☐	19: ☐	22: ☐
2: ☐	5: ☐	8: ☐	11: ☐	14: ☐	17: ☐	20: ☐	23: ☐
3: ☐	6: ☐	9: ☐	12: ☐	15: ☐	18: ☐	21: ☐	24: ☐

5 km/s

2
y
x

MAGMapost

Return

Settings

User Scale

Viewpoint

Clipping

Previous

Same

Next

Temperature [C]

622.0

619.1

616.1

613.2

610.2

607.3

604.3

601.4

598.4

595.5

592.5

589.6

586.6

583.7

580.7

Step

MAGMA

Information

7 bin

Percent Filled 100.00 [%]

Report Created
Version 9.01
Cycle 1
Run Date of CV 5: 25.4.20
17 May 11 00:26:55 (PST)

4.2.4 Results

The following section presents the results of the simulations for the bowl casting and die and makes a comparison of the computed results with the experimental data.

- **Solidification Time**

The times to complete solidification of the casting at the thermocouple location (i.e., in the vicinity of the plate section of the wall), biscuit, and at the gate for the fifth casting cycle are given below:

<u>Location</u>	<u>Solidification Time (sec)</u>
Thermocouple (Box area)	3.8
Gate Region near Post	1.9
Biscuit	Small amount still liquid at 30 sec

The biscuit length used in this simulation was 16 mm. As with the previously described plate casting, information relating the gate freezing time and the volume of liquid that remains in the casting cavity at that time, can be used to predict the amount of shrinkage porosity in the resultant casting. In analyzing the thermal fields in the casting, it was difficult to define the volume of liquid that existed in the casting cavity at the gate freezing time of 2.07 seconds. By identifying the casting cavity as a subregion of the casting system, the prediction of the liquid fraction in the casting cavity at the gate freezing time, and, in turn, the amount of shrinkage porosity in the casting, should be possible. Based on the thermal analysis of the casting, shrinkage in each of the size posts, and in the box area of the casting is expected.

While not used in this analysis, it should be noted that the MAGMASOFT program will allow for the thermal analysis of the solidified (or partially solidified) casting after it is ejected from the die cavity. Such computations would be useful in the prediction of the thermal stresses and distortion of the cast component following its removal from the die casting machine.

- **Thermal Characteristics**

The cooling curve in Fig. 4.2-12 shows the thermal profiles at the thermocouple location specified in the preprocessor. The maximum temperatures indicated at a depth of 0.5 mm from the casting surface are near 600° C. At the surface, the die cools to 315° C at the end of the cycle. The program also tracks the average temperature in the die. The peak average temperature in the die reaches 522° C, and the minimum average temperature is 202° C. It is important to note that the initial die temperature was set at 200° C in the simulation program, and not room temperature.

The experimental cooling curve data indicate a similar profile. Experimental data reached a maximum temperature of 518° C at a depth of 0.5 mm below the surface. The coolest temperature at this location was 357° C. The table below summarizes the maximum and minimum temperatures from the simulation and experimental trail:

Experimental Data			Simulation Data		
Location (mm)	min °C	max °C	min °C	max °C	Location (mm)
0.25 Inside	357	518	324	476	0.5 Inside
1.3 Inside	357	501	324	476	1.5 Inside
12.7 Inside	347	379	279	349	12.5 Inside
0.2 Outside	289	503	286	465	0.5 Outside
1.5 Outside	288	467	286	465	1.5 Outside
12.7 Outside	272	322	265	279	12.5 Outside
Gate	163	284	247	381	Gate
Floor	240	300	252	333	Floor

Table 4.2-8 Cooling Curve Comparison

The average difference between the experimental and simulated temperatures at the specified locations was about 30° C. The largest difference was associated with the gate location, where the difference was 97° C. This difference and those associated with the temperature indicators near the die-casting interface are attributed to three factors: (1) the mesh size and (2) the different locations for the experimental and simulated temperature determinations, and (3) the experimental data being for a

water lubricant sprayed condition. Because of available memory capacity of the IBM Risc 6000 system, the mesh size about the die-casting interface regions was relatively coarse (i.e., 1 mm for the thickness dimension) and the corresponding accuracy for the temperature indicating location was ± 0.5 mm. Thus, the simulated temperature location set nominally at 0.5 mm from the die-casting interface could have been at or 1 mm from the die-casting interface. The combination of large mesh size and differences between the experimental and simulated thermal response locations limits the ability to make any quantitative comparison of these data. Increased memory capability would allow for a finer mesh, and, in turn, smaller mesh dimensions, which would significantly reduce the discrepancy between experimental and simulated thermal values.

The difference between the experimental and simulated temperature values for the bowl casting could also originate from the non inclusion of the lubricant spray effect in the MAGMASOFT analysis. The bowl die castings for which the thermal data are presented were sprayed with a water-based lubricant with spraying times of about 2 seconds. The MAGMASOFT program used to calculate the thermal fields in the casting did not allow for or include the thermal response associated with the spraying. While the inclusion of the thermal response of the lubricant application would be expected to alter the simulated thermal values, it should be noted that most of the simulated temperatures near the casting-die interface or less than the experimental values, and therefore the inclusion of the spray cooling effect is not thought to be the factor which account for the observed differences between the experimental and simulated temperature values.

- **Filling Characteristics**

As expected, the simulation indicates some cooling of the liquid alloy as it fills the casting cavity. In the filling analysis, the temperature of the liquid alloy at the gate was taken to be 625° C. Based on that gate temperature, and using the level 4 filling solver, the liquid near the top of the casting cooled 45° C (i.e., to a temperature of 580° C) during filling. Most of the liquid in the casting cavity cooled 20° C during filling. The filling analysis indicates that there should be no filling related defects, such as cold shuts for the specified system.

5.0 CURRENT AND FUTURE CAPABILITIES

The following sections summarize the current capabilities of MAGMASOFT with respect to die casting, and some potential areas of improvement that would benefit die casting simulation.

5.1 Current Capabilities

The graduate student working on this project has found the system to be user friendly and integrated to allow entry of the geometry. It has been well thought out, and designed specifically with the foundry in mind. It is totally integrated with a preprocessor and a postprocessor and requires no other third party software. It has the capability to model both fluid flow in filling and the cooling and solidification of the casting. For batch type casting, the program also has the ability to cycle the die according to a specified cycle time. This is very important in die casting. This allows the user to examine steady state profiles.

The software also has many other capabilities. One very notable capability is the ability to model feeding. This allows the user to specify a feeder, and the associated feederneck. The program will calculate feeding flow based in part on percent solids, until feeding is no longer available. The output will display the casting as a percent density. The program can also display such features as cooling rate, thermal gradient, and Niyama criteria.

5.2 Future Capabilities

Listed below are some items that could improve the current software with respect to the ability to model die casting. Many of these improvements are expected to be released soon, and others are under development. Further detail has been provided by MAGMA in the next section about the future development plans.

- *Incorporation of Die Lubricant and Spray Effect*
The low pressure die cast module currently has the ability to include the effects of die lubricant and spray. The current version installed at the ERC does not have this capability. Future releases are expected to include this.

- *Atomized Flow at the Gates*
With some gate configurations and metal velocities, the metal tends to atomize at the gate. The current flow solver is limited in its prediction of atomization. A new release of a high pressure die casting flow solver by the end of the year should have a significantly improved capability to model the flow phenomena in high pressure die casting. The advanced flow solver should also aid in the analysis of air entrapped porosity. The new solver will also account for back pressures in the die.
- *Filling based on Shot Profile*
Filling of die castings is a complex problem. There is interest in accurately modeling the 3-D wave formation in the shot sleeve. The speed profile of filling can often have as many as 7 levels of speed transitions. If air is entrapped during this phase, there can be a significant impact on flow at the gate and resulting porosity.
- *Cooling Curve Locations in Mesh Postprocessor*
The ability to determine the location of the cooling curve locations after the mesh has been generated, through the postprocessor could assist in interpreting the quality of the mesh and cooling curve results.
- *Input from other CAD systems*
This feature has started to be implemented at many of the domestic user sites. Current import capabilities include CATIA, SDRC, PATRAN, and IGES file formats.
- *Stress Analysis*
A stress analysis module has been developed, and should be released soon domestically. Description provided in Reference 8.

- *Die Preheat Analysis*

The program currently handles modeling of heating lines, such as oil. A module to examine the preheating practice of surface techniques, such as torches, could aid in improving the preheat practices. Currently, the user typically inputs a bulk temperature of the die for preheat. This would be very similar to a die spray module, in that the interface conditions on the surface could be changed during the preheat process.

- *Fraction Liquid when Gate Freezes*

The program tracks the fraction liquid for the entire casting cavity. It would be useful if it tracked separately the fraction liquid past the gate. From the amount of liquid remaining past the gate, calculations could be made on the amount of porosity attributable to shrinkage after the gate has solidified.

6.0 REFERENCES

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APPENDIX A - CAD GEOMETRY DRAWINGS

Figure 3.1 Ejector Die Half of the Pac Man Die

Figure 3.2 Cover Die Half of the Pac Man Die

APPENDIX B - 2mm THICK PLATE GEOMETRY ENTRY

The following describes in detail and lists the commands in the text file used to create the geometry.

- *Die (5 minutes)*

The die was entered as an ejector half and a fixed half. The command *set mat* changes the material representation in the CAD system. The material 6 1 is called permanent mold and the material 8 1 is a user defined material. Both of these are eventually set to H-13 steel in the simulation module. The separation allows for easy viewing of the die surface in the postprocessor. The *set cube* command defines a cube of the specified material based on the x, y, and z coordinates of two end points on the diagonal.

```
#  
#   Form Box  
#  
set mat 6 1  
set cube 100 150 200 0 -150 -300      ejector die  
set mat 8 1  
set cube -100 150 200 0 -150 -300     fixed die
```

- *Plate (20 minutes)*

The plate was represented by three cubes and four cylinders. Together the seven components makeup the main plate. The *set cyl* command specifies a cylinder with the bottom center at the first set of x, y, and z coordinates and the second center being at the second set of x, y, z coordinates. The last two numbers represent the radius at the top and bottom.

```
#  
#   Casting  
#  
set mat 1 1  
set cube 0 -50.8 -59.7 2 50.8 59.7  
set cube 0 -50.8 -49.5 2 -60.2 49.5  
set cube 0 50.8 -49.5 2 60.2 49.5  
set cyl 2 -50.8 -49.5 0 -50.8 -49.5 10.2 10.2  
set cyl 2 -50.8 49.5 0 -50.8 49.5 10.2 10.2  
set cyl 2 50.8 -49.5 0 50.8 -49.5 10.2 10.2  
set cyl 2 50.8 49.5 0 50.8 49.5 10.2 10.2
```

- *Overflows (16 minutes)*

The overflows have a draft on the sides. These can be constructed by specifying two boxes, and the height of each box. The system is then told to close the solid. The gate to the overflow was then specified as a thin cube. These two objects were combined and called Overflow 2. This was then copied and moved to the first and third location to make all three of the overflows.

```
#
#   Overflows
#
begin macro
begin box
set inpv 1
set point 0 -6 60.7
set point 0 6 70.7          set first box
begin box
set point 0 -6 62.7
set point 0 6 67.7          set second box
set inpv 0
set point 0 0 0
set point 10 0 0            set height of boxes
close solid
set cube 0 -4.5 59.7 1 4.5 60.7  inlet to overflow
end macro
name sel Overflow 2
copy sel
name sel Overflow 1
move rel 0 35.56 0
copy sel
name sel Overflow 3
move rel 0 -71.12 0
```

- *Gate (20 minutes)*

The gate was one of the more difficult sections to enter. The gate land area was entered using the *set cube* command. The fan area had to be entered as two polygons to obtain the taper along the edges. This left the fan at a constant thickness. The material had to be changed to permanent mold and a triangular cut made on the top half of the fan.

```
#
#   Gate
#
set cube 0 35 -61.7 1 -35 -59.7  gate land
```

name sel Gate	
# Fan Area	
begin macro	
begin line	
set inpv 1	
set point 0 10 -98.7	
set point 0 17 -80.7	
set point 0 35 -61.7	
set point 0 -35 -61.7	
set point 0 -17 -80.7	
set point 0 -10 -98.7	set on polygon
close line	
set point 0 9 -98.7	
set point 0 16 -78.7	
set point 0 35 -61.7	
set point 0 -35 -61.7	
set point 0 -16 -78.7	
set point 0 -9 -98.7	set second polygon
close line	
set inpv 0	
set point 0 0 0	
set point 9 0 0	set height of polygons
close solid	drew object
set mat 6 1	
begin line	
set inpv 2	
set point 9 0 -78.7	
set point 1 0 -61.7	
set point 9 0 -61.7	triangle points
close line	
set inpv 1	
set point 0 35 0	
set point 0 -35 0	triangle height
close solid	
end macro	
name sel Fan	

- *Runner (6 minutes)*

The runner was constructed with two boxes to obtain the taper along the sides. The end of the runner entering the biscuit was constructed with a tapered cylinder.

```
#
# Runner
#
```

```

set mat 1 1
begin macro
begin box
set inpv 1
set point 0 10 -172.7
set point 0 -10 -98.7          bottom of runner
begin box
set point 0 9 -172.7
set point 0 -9 -98.7          top of runner
set inpv 0
set point 0 0 0
set point 9 0 0              height of runner
close solid
set cyl 0 0 -172.7 9 0 -172.7 10 8    end of runner
end macro
name sel Runner

```

- *Biscuit and Inlet (5 minutes)*

The biscuit is only a cylinder. At the plunger side of the cylinder, a thin layer of 10 1 material is added. This is the inlet. The inlet indicates to the program where the part is filled from for the filling simulation.

```

#
#   Biscuit and Inlet
#
set cyl 0 0 -174.7 -15 0 -174.7 25 25
set mat 10 1
set cyl -15 0 -174.7 -16 0 -1174.7 25 25

```

- *Cooling Lines (22 minutes)*

The cooling lines consisted only of cylinders. For the simulation only the top line on the ejector side was used. The experimental data only used this cooling line, or no cooling lines.

```

#
#   Cooling Lines
#
set mat 7 1
#   Ejector Side
set cyl 28.7 150 0 28.7 -150 0 4.76 4.76
set cyl 30.7 150 67.3 30.7 -150 67.3 4.76 4.76
#   Fixed Side
set cyl -22.86 -32.36 -39.88 -22.86 32.36 -39.88 4.76 4.76
set cyl -22.86 27.6 -44.88 -22.86 27.6 200 4.76 4.76

```

set cyl -22.86 -27.6 -44.88 -22.86 -27.6 200 4.76 4.76

- *Cooling Curve Locations (6 minutes)*

The cooling curves were specified at specific points in the casting corresponding to the locations in the experimental tests. The experimental tests used one thermocouple at the surface. Since the Finite Difference Method tracks temperatures at the center of the element, tracking the surface temperatures is difficult; however the thermocouples were located at 0.25, 1.5, and 12.5 mm from the surface. To ensure proper meshing around the thermocouple points, small cubes of die material were placed at each location.

```
#
#   Cooling Curve Locations
#
#   Fine cubes to ensure proper meshing
set mat 6 1
set cube 2 12 -31.5 2.5 13 -32
set cube 3 12 -31.5 4 13 -32
set cube 14 12 -31.5 15 13 -32
begin cc
set point 2.25 12.5 -31.75      Curve at 0.25 mm away
set point 3.5 12.5 -31.75      Curve at 1.5 mm away
set point 14.5 12.5 -31.75     Curve at 12.5 mm away
end cc
```

APPENDIX C - BOWL DIE

The following describes in detail and lists the commands in the text file used to create the geometry. Most of the specific commands were described in the prior appendix on the 2mm thick plate and will not be repeated in this section.

- *Die (8 minutes)*

The die was entered as an ejector half and a fixed half.

```
#  
#   Form Box  
#  
  
set mat 6 1  
set cube 0 227 229 -98.5 -227 -368  
set mat 8 1  
set cube 0 227 229 149 -227 -368
```

- *Bowl (32 minutes)*

The bowl was represented by multiple cylinders. The casting required a main cylinder, then the core area was removed. The

```
#  
#   Casting  
#  
set mat 1 1  
set cyl -3 0 0 53 0 0 82 82      main casting  
set mat 8 1  
set cyl 0 0 0 54 0 0 78 79      larger core  
set cyl -3 0 0 0 0 0 35.6 35.6  smaller core  
begin line  
set inpv 1  
set point 0 0 -35.6              polygon core to cut box shape  
set point 0 0 35.6  
set point 0 -62.5 35.6  
set point 0 -67.5 39.6  
set point 0 -84.5 39.6  
set point 0 -84.5 -39.6  
set point 0 -67.5 -39.6  
set point 0 -62.5 -35.6  
close line  
set inpv 0  
set point -3 0 0
```

```

set point 53 0 0
close solid
#      6 Posts
set inpv 1
set mat 1 1
set cyl -3 0 -75.2 53 0 -75.2 6.8 5.4
copy sel
rot sel 60
copy sel
rot sel 60
copy sel
rot sel 60
copy sel
rot sel 60
copy sel
rot sel 60
#      End Box
set cube -3 72 -27 53 83 27

```

- *Gate and Runner (29 minutes)*

The gate was one of the more difficult sections to enter. The gate land area was entered using a polygon. The curved runner was also entered using a polygon. The inner polygon was entered using points that would provide the proper taper to the runner. The points for the lines were approximated from the part drawings.

```

#
#      Gate and Runner
#
#      Gate Land Area
begin line
set inpv 1
set point 0 0 -86
set point 0 0 -79
set point 0 22 -76
set point 0 39 -68
set point 0 49 -62
set point 0 40 -76
set point 0 20 -83
close line
set inpv 0
set point 0 0 0
set point 2 0 0
close solid
name sel Gate

```

Curved Runner Area

```
begin line
set inpv 1
set point 0 0 -82
set point 0 -9 -102
set point 0 -13 -117
set point 0 -14 -132
set point 0 -15 -172
set point 0 -19 -218
set point 0 22 -218
set point 0 19 -172
set point 0 18 -152
set point 0 19 -132
set point 0 24 -109
set point 0 32 -92
set point 0 42 -73
set point 0 29 -79
set point 0 10 -83
close line
set point 0 1 -93
set point 0 -4 -104
set point 0 -8 -118
set point 0 -9 -132
set point 0 -10 -172
set point 0 -13 -218
set point 0 17 -218
set point 0 14 -172
set point 0 13 -152
set point 0 14 -132
set point 0 20 -108
set point 0 27 -91
set point 0 30 -85
set point 0 22 -89
set point 0 7 -93
close line
set inpv 0
set point 0 0 0
set point 7 0 0
close solid
set cyl 0 0 -220 7 0 -220 27 19
```

- *Biscuit and Inlet (3 minutes)*

The biscuit is only a cylinder. At the plunger side of the cylinder, a thin layer of 10 1 material is added. This is the inlet. The inlet indicates to the program where the part is filled from for the filling simulation.


```

#
#      Biscuit and Inlet
#
set cyl 0 0 -215 -15 0 -215 45 45
set mat 10 1
set cyl -15 0 -215 -16 0 -215 45 45

```

- *Cooling Lines (20 minutes)*

The cooling lines consisted only of cylinders. Some of the cooling lines that were loops or did not go straight through the die were created using the sweep command. This is done by setting a contour, typically a circle the size of the cooling line, setting the trajectory of the cooling line, and creating the sweep.

```

#
#      Cooling Lines
#
#      set mat to 13 1 (Boundary) if line not used
set mat 7 1
#      Ejector Side
set cyl 33.3 227 -112.7 33.3 -227 -112.7 4.8 4.8
#      Inner Core Lines
set con
begin circle
set inpv 1
set point 149 28.6 39.7
set point 149 33.4 39.7
set tra
set point 149 28.6 39.7
set point 22.2 28.6 39.7
set point 22.2 -28.6 39.7
set point 149 -28.6 39.7
close line
mark ct
make sweep
copy sel
move rel 0 0 -79.4
#      Biscuit Cooling
set con
begin circle
set inpv 0
set point 28.6 28.6 -368
set point 28.6 33.4 -368
set tra

```

```

set point 28.6 28.6 -368
set point 28.6 28.6 -203.2
set point 28.6 -28.6 -203.2
set point 28.6 -28.6 -368
close line
mark ct
make sweep
#      Fixed Side
set cyl -31.8 227 0 -31.8 -227 0 4.8 4.8
set cyl -31.8 227 -57.2 -31.8 -227 -57.2 4.8 4.8
set cyl -23.8 227 -112.7 -23.8 -227 -112.7 4.8 4.8

```

- *Cooling Curve Locations (6 minutes)*

The cooling curves were specified at specific points in the casting corresponding to the locations in the experimental tests. Thermocouples were located at 0.25, 1.5, and 12.5 mm from the surface. To ensure proper meshing around the thermocouple points, small cubes of die material were placed at each location.

```

#
#      Cooling Curves
#      Fine cubes to ensure proper meshing
#
#      Center of End Box Locations (Interior and Exterior)
set mat 8 1
set cube 23 83 -1 24 83.5 1
set cube 23 84 -1 24 85 1
set cube 23 95 -1 24 96 1
set cube 23 71.5 -1 24 72 1
set cube 23 70 -1 24 71 1
set cube 23 59.5 -1 24 60.5 1
set mat 6 1
#      Gate Location
set cube -1.5 -84.5 14.5 -2.5 -85.5 15.5
#      Floor of Casting Location
set cube -5.5 -.5 57.5 -6.5 .5 58.5

begin cc
#      Center of End Box Locations (Interior and Exterior)
set point 25 83.25 0
set point 25 84.5 0
set point 25 95.5 0
set point 25 71.75 0
set point 25 70.5 0
set point 25 60 0

```

Gate Location
set point -2 15 -85
Floor of Casting Location
set point -6 0 58
end cc