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Fusion Welding of AerMet[®] 100 Alloy

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Fusion Welding of AerMet® 100 Alloy

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Abstract

A database of mechanical properties for weldment fusion and heat-affected zones was established for AerMet® 100 alloy, and a study of the welding metallurgy of the alloy was conducted. The properties database was developed for a matrix of weld processes (electron beam and gas-tungsten arc) welding parameters (heat inputs) and post-weld heat treatment (PWHT) conditions. In order to insure commercial utility and acceptance, the matrix was commensurate with commercial welding technology and practice. Second, the mechanical properties were correlated with fundamental understanding of microstructure and microstructural evolution in this alloy. Finally, assessments of optimal weld process/PWHT combinations for confident application of the alloy in probable service conditions were made. The database of weldment mechanical properties demonstrated that a wide range of properties can be obtained in welds in this alloy. In addition, it was demonstrated that acceptable welds, some with near base metal properties, could

be produced from several different initial heat treatments. This capability provides a means for defining process parameters and PWHT's to achieve appropriate properties for different applications, and provides useful flexibility in design and manufacturing. The database also indicated that an important region in welds is the softened region which develops in the heat-affected zone (HAZ) and analysis within the welding metallurgy studies indicated that the development of this region is governed by a complex interaction of precipitate overaging and austenite formation. Models and experimental data were therefore developed to describe overaging and austenite formation during thermal cycling. These models and experimental data can be applied to essentially any thermal cycle, and provide a basis for predicting the evolution of microstructure and properties during thermal processing.

Executive Summary

AerMet® 100 Alloy is an alloy recently developed by Carpenter Technology Corporation and is a very high strength, high fracture toughness alloy designed for use in structural components in aerospace applications. A variety of possible commercial and military applications for the alloy have been identified and include armor plate, landing gear, gas turbine main shafts for military and commercial aircraft, mechanics hand tools, bolts and fasteners, and bicycle frames. Many of these applications require weldments joining AerMet® 100 to itself, or to another alloy steel or high temperature alloy. At the outset of this project, however, little work had been done to assess the welding behavior of the alloy, and thus applications within the Department of Energy (DOE) and the commercial sector were limited to those not requiring welding. Development of suitable welding technology allows for greater flexibility in DOE designs for a variety of hardware, and increases the market potential for the alloy in general.

As a result of this need, a cooperative research and development agreement (CRADA) was established between Carpenter Technology Corporation and Sandia National Laboratories to evaluate the welding response of this alloy. Carpenter is the developer and sole producer of the alloy and therefore has a thorough understanding of the basic characteristics of the alloy. Sandia has expertise in welding science and technology and microstructural analysis. Carpenter supplied all the materials required for this program and also fabricated all of the test samples and conducted parallel heat treatment and x-ray analysis studies during the course of the project. Sandia performed the welding, mechanical testing, and microstructural and metallurgical analyses pertaining to welding response.

The objectives of the project can be summarized as follows. First, a database of mechanical properties for weldment fusion and heat-affected zones (FZ and HAZ) was established. This database was established within a matrix of weld processes (electron beam and gas-tungsten arc) welding parameters (heat inputs) and post-weld heat treatment (PWHT) conditions. In order to insure commercial utility and acceptance, the matrix was commensurate with commercial welding technology and practice. Second, the mechanical properties were correlated with fundamental understanding of microstructure and microstructural evolution in this alloy. Finally, determinations of optimal weld process/PWHT combinations for confident application of the alloy in probable service conditions were made.

The database of weldment mechanical properties demonstrated that high strength levels can be achieved in either solution treated or aged base plate with either the cold wire feed GTA or EB welding process. In general, for direct aged (at temperatures near 482°C) welds in either base plate, the overaged region of the HAZ is the strength limiting factor, while the fusion zone controls the impact toughness. The data also indicate that within the range of GTA welding parameters and PWHT conditions evaluated, suitable procedures can be defined which provide useful

combinations of strength and toughness. Compared to the GTA welds in the direct aged condition, the EB welds generally have somewhat superior strength, but not necessarily superior toughness. This observation seems to imply that the finer solidification features which are usually associated with EB welding do not play a dominant role in overall mechanical properties. It is believed that the underlying reasons for this are likely associated with the martensitic transformation and the structure that results from it. Thus, the characteristics of the solid state transformation would appear to dominate over the details of the solidification microstructure. In addition, it was demonstrated that acceptable welds, some with near base metal properties, could be produced from several different initial heat treatments, and this provides useful flexibility in design and manufacturing. The database also provides a means for defining process parameters and PWHT's to achieve appropriate properties for different applications.

The mechanical properties database indicated that an important region in welds is the softened region which develops in the heat-affected zone (HAZ). Weld thermal cycle simulations, dilatometry, electron microscopy, and austenite fraction measurements indicated that the development of this region is governed by a complex interaction of precipitate overaging and austenite formation. This softened region is bounded at lower peak temperatures by the rate at which precipitate coarsening occurs, and it is bounded at higher temperatures by the formation of transformable austenite (*i.e.* austenite which transforms to martensite on cooling). Therefore, within the welding metallurgy studies, experimental data and models were developed to describe overaging and austenite formation kinetics during thermal cycling. The aging model is based on precipitate strengthening theory and classical coarsening kinetics, and, when combined with the additivity principle, accurately reproduces the softening which occurs during weld HAZ thermal cycles. Austenite formation kinetics were examined by development of continuous heating transformation (CHT) diagrams for heating rates between 0.1 and 2500°C/sec, and indicates that the formation of austenite at high heating rates may occur by a shear transformation. The models and experimental data can be applied to essentially any thermal cycle, and provide a basis for predicting the evolution of microstructure and properties during welding, heat treatment, and other processing. In addition, the models can be used to assess the effects of in-service thermal exposures, such as fires or other thermal excursions.

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I. Introduction

AerMet® 100 alloy is a 0.23wt % C, 13.4% Co, 3.1% Cr, 11.1% Ni 1.2% Mo proprietary steel alloy [Refs 1, 2] designed by Carpenter Technology Corporation for aerospace structural applications. The alloy is designed to exhibit a combination of high strength, high fracture toughness, and exceptional resistance to stress corrosion cracking. The nominal properties for the alloy are 285 ksi (1965 MPa) ultimate tensile strength, 115 ksi√in (126.4 MPa√m) K_{Ic} , and 35 ksi√in (38.5 MPa√m) K_{Isc} in a 3.5% NaCl solution, and make it a candidate for applications such as navel aircraft landing gear. The alloy is the most recent development in the family of secondary hardening Ni-Co-Cr steels which includes the HY180 and AF1410 alloys. Following normal heat treatment (solution treating, quenching, and aging), these alloys are low carbon, Fe-Ni lath martensitic alloys that attain their properties through the precipitation of M_2C (where M is Mo and Cr) carbides during aging. The AerMet® 100 alloy has a high hardenability, and was designed to be air cooled from the solution treatment temperature in section sizes up to 2 in (51 mm). The normal heat treatment for this alloy consists of solution treatment at 885°C for one hour, air cooling, refrigeration at -73°C for one hour, air warming, and aging at 482°C for 5 hours.

In comparison to other types of high strength steels, such as the quenched and tempered steels, the general weldability of the Ni-Co-Cr steels is usually considered to be quite good [Ref 3]. This improvement results primarily from the fact that the martensite which forms in these alloys is relatively soft and ductile. Thus, in the as-quenched condition, which typifies much of the weld fusion and heat-affected zones, the alloys are resistant to the problems (particularly cold cracking) which are often associated with welding of high strength quenched and tempered steels.

Since AerMet® 100 is a relatively recent development, there have been few studies [Refs 4, 5] of its welding behavior. Work by Novotny and Englehart [Ref 5] on gas-tungsten arc welds in AerMet® 100 indicated that reasonable joint strength and toughness could be achieved in the alloy by direct aging after welding, and that near base metal properties could be achieved if the welds were solution treated and aged after welding. The study by Kautz *et. al.* [Ref 4] demonstrated that the alloy is amenable to processing by both electron and laser beam welding, although no properties data was reported. Given the general similarity between AerMet® 100 and AF1410, it is likely that both the fabrication weldability (ability to produce defect free welds) and the service weldability (ability of welds to meet service requirements) would qualitatively be similar. AF1410 was developed specifically for welded applications, and is known to have a relatively low susceptibility to welding defects. In addition, high joint strength and toughness, relative to the base metal, can be achieved in AF1410 though appropriate combinations of welding process procedures and post-weld heat treatment [Refs 6, 7].

The objectives of the project can therefore be summarized as follows. First, a database of mechanical properties for weldments was established. This database was established within a

matrix of weld processes (electron beam and gas-tungsten arc) welding parameters (heat inputs) and post-weld heat treatment (PWHT) conditions. In order to insure commercial utility and acceptance, the matrix was commensurate with commercial welding technology and practice. A second objective was to develop an understanding of the welding metallurgy of this alloy, particularly as it pertains to microstructure and properties evolution in the weld heat affected zone. Toward this end, two major features of the response of the alloy to rapid thermal cycling were examined. These features are the kinetics of aging and kinetics of austenite formation during heating. They are important in that they are the primary factors which control the severity and extent of the softened region of the weld heat-affected zone. Description of these processes, in relation to welding process parameters and in conjunction with the baseline mechanical properties data, provide a basis for predicting the evolution of microstructure and properties during welding, heat treatment, and other processing. In addition, the models can be used to assess the effects of in-service thermal exposures.

II. Experimental Procedures

2.1 Welding Matrix and Procedures

In order to facilitate comparisons between the various processing conditions and the welding metallurgy study, a single heat of 15.9 mm thick AerMet® 100 Alloy plate and a single heat of matching 1.6 mm diameter weld filler wire were used for the entire project. The compositions of these materials are shown in Table 1. Both alloy compositions meet the requirements for Aerospace Materials Specification AMS 6532 and the wire and base plate compositions are closely matched.

Table 1. Alloy Compositions

Element	Base Plate (Heat 90608)	Weld Filler Wire (Heat 90548)
Ni	11.11	11.03
Co	13.43	13.43
Cr	2.96	2.98
Mo	1.19	1.18
C	0.24	0.24
Mn	<0.01	0.01
Si	<0.01	<0.01
Al	0.006	0.005
Ti	0.017	0.010
P	0.003	0.003
S	0.0010	0.0007
N	<0.0010	<0.0010
O	0.0027	0.0027

For the baseline weldment properties study, a matrix consisting of base plate heat treatment condition, welding procedures, and post-weld heat treatment conditions was established. In order to assure commercial utility, the matrix was designed to be commensurate with typical current practice for similar alloys. Within this matrix, two base plate conditions, solution annealed and fully heat treated (solution treated plus aged at 482°C for 5 hr), were investigated. Because of the known need to maintain a high degree of weldment cleanliness in high strength-high toughness alloys, the welding processes selected were electron beam (EB) and cold wire feed gas-tungsten arc (GTA) welding. For the EB welding, procedures were developed to produce a full penetration weld in 15.9 mm thick plate in a single weld pass. For the GTAW process, procedures were developed to represent both low and high heat input processing. Three basic post-weld heat

treatment (PWHT) conditions were examined. These conditions were as-welded, direct aged (*i.e.* without an re-solution heat treatment), and a low temperature direct age. Direct PWHT aging of welds in solution treated plate was conducted using the standard refrigeration treatment and age for AerMet® 100 Alloy (-73°C/1 hr + 482°C/5 hr). In order to minimize base plate overaging in the welds in aged base plate, these welds were heat treated using a slightly lower aging temperature (-73°C /1 hr + 468°C/5 hr). The low temperature PWHT (direct aging at 204°C for 1hr) was intended to represent a paint curing cycle, and to determine to the effects of cementite precipitation on weldment properties. In addition, one EB weld and one GTA weld received a full solution treatment followed by the standard aging heat treatment. All welds were made parallel to the rolling direction and the full matrix of weldment conditions is summarized in Table 2.

Table 2. Weldment Matrix

Weld Designation	Base Plate Condition	Welding Process	Heat Input	Post-Weld Heat Treatment
GSL-AW	Solution Anneal	GTAW	Low	As-Welded
GSL-NA	Solution Anneal	GTAW	Low	Direct Age (-73°C/1 hr+482°C/5 hr)
GSL-LA	Solution Anneal	GTAW	Low	Direct Age (204°C/1hr)
GSH-AW	Solution Anneal	GTAW	High	As-Welded
GSH-NA	Solution Anneal	GTAW	High	Direct Age (-73°C /1 hr+482°C/5 hr)
GSH-LA	Solution Anneal	GTAW	High	Direct Age (204°C /1hr)
GAL-AW	482°C/5 hr Age	GTAW	Low	As-Welded
GAL-NA	482°C /5 hr Age	GTAW	Low	Direct Age (-73°C /1 hr+468°C/5 hr)
GAL-LA	482°C /5 hr Age	GTAW	Low	Direct Age (204°C /1hr)
GAH-AW	482°C /5 hr Age	GTAW	High	As-Welded
GAH-NA	482°C /5 hr Age	GTAW	High	Direct Age (-73°C /1 hr+468°C/5 hr)
GAH-LA	482°C /5 hr Age	GTAW	High	Direct Age (204°C /1hr)
ESL-AW	Solution Anneal	EBW	1 Pass	As-Welded
ESL-NA	Solution Anneal	EBW	1 Pass	Direct Age (-73°C /1 hr+482°C/5 hr)
ESL-LA	Solution Anneal	EBW	1 Pass	Direct Age (204°C /1hr)
EAL-AW	482°C /5 hr Age	EBW	1 Pass	As-Welded
EAL-NA	482°C /5 hr Age	EBW	1 Pass	Direct Age (-73°C /1 hr+468°C/5 hr)
EAL-LA	482°C /5 hr Age	EBW	1 Pass	Direct Age (204°C /1hr)
GSL-SA	Solution Anneal	GTAW	Low	Solution Anneal + Normal Age
ESL-SA	Solution Anneal	EB	1 Pass	Solution Anneal + Normal Age

Prior to welding, all material was cleaned using a caustic cleaner (Alconox) and abrasive pads, followed by water and ethanol rinses. Gas-tungsten arc welds were produced in the 15.9 mm thick plates by using the parameters shown in Table 3. A U-groove joint design was used for weld metal testing and a J-groove design was used to facilitate testing of heat-affected zone (HAZ) properties. The U- and J-groove designs were selected to have identical cross-sectional areas, so that the number of fill passes was the same for either joint type. For the parameters shown in Table 3, 18 and 10 stringer bead passes were required for the low and high heat input schedules, respectively.

Table 3. GTA Welding Parameters

Parameter	Low Heat Input Schedule	High Heat Input Schedule
Arc Current (A)	250	250
Arc Voltage (V)	12	12
Travel Speed (mm/sec)	5.92	2.54
Wire Diameter (mm)	1.6	1.6
Wire Feed Rate (mm/sec)	23.7	18.6
Shielding, 80He/20Ar (l/min)	16.5	16.5
Trailing Shield, Ar (l/min)	11.8	11.8
Heat Input (J/mm)*	402	945
Interpass Temperature (°C)	150	150

* Assumes an arc efficiency of 0.80

In addition to the trailing shield, aluminum foil dams (approximately 50 mm high and running the length of the weld) spaced 25 mm on either side of the GTA torch were used in an attempt to minimize oxide formation during welding. Although oxide formation was minimal, a stainless steel brush was used to clean each stringer bead prior to deposition of subsequent passes.

Electron beam welding procedures were selected to produce a single pass, full penetration weld in the 15.9 mm thick plate, while maintaining a weld width of approximately 2.5 mm at the midsection of the plate. The EB welding parameters are shown in Table 4.

Table 4. EB Welding Parameters

Accelerating Voltage (kV)	100
Beam Current (mA)	75
Travel Speed (mm/sec)	10.2
Heat Input (J/mm)	740

In order to assure maximum cleanliness for the EB welds, the vacuum chamber was backfilled with argon, and re-evacuated prior to welding.

2.2 Analytical Procedures

A typical weld plate and the layout for mechanical test samples is shown in Figure 1. Test samples included longitudinal base metal tensile samples, transverse and longitudinal tensile samples, transverse base metal Charpy impact samples, and fusion zone and fusion line Charpy impact samples. Fusion line Charpy samples were produced by blanking out oversized samples which were then macroetched, marked, and finish machined. Tensile tests were conducted at

(a)

(b)

Figure 1. (a) Photograph of test weld plate. (b) Layout of weld plates and mechanical test samples.

room temperature on 6.4 mm diameter by 25.4 mm gage length samples at a strain rate of 10^{-3} sec^{-1} using a servo-hydraulic load frame in accordance with ASTM Specification E8. Charpy impact tests were conducted using a calibrated Satech impact testing machine at room temperature in accordance with ASTM Specification E23.

Weld thermal cycle simulations were used to assess the HAZ characteristics of AerMet® 100. These simulations were conducted on a Gleeble 1500 thermomechanical simulator and a variety of thermal cycles, representing various regions of the HAZ, were used. The specific HAZ thermal cycles were determined from either the Gleeble software or analytical descriptions of weld thermal cycles and are described in detail in the pertinent results sections. The weld thermal cycle simulator was also used for short time isothermal aging experiments, as well as dilatometric measurements of critical temperatures as a function of heating rate. All thermal cycle testing was conducted on 6.35 mm diameter rods in either the solution heat treated or fully aged condition.

Optical metallography was used to characterize weld metal and HAZ structures. Sample preparation for the microstructural examinations included mounting and polishing through 0.05 mm colloidal silica using standard metallographic procedures. Analytical electron microscopy (AEM) was used for further characterization of selected samples. For these analyses, thin foil specimens were prepared by electrolytic thinning in a 10% perchloric acid in methanol solution at -40°C and 15 VDC. A Phillips EM30 analytical electron microscope, operating at an accelerating potential of 300 kV and equipped with an Oxford Instruments energy dispersive spectrometer (EDS), was used for analysis of the foils. Scanning electron microscopy (SEM) was also used on selected samples. Samples were generally examined in the secondary electron imaging mode in an Amray 1645 SEM operating at 20 kV. Identification of microstructural constituents was conducted by AEM.

Differential thermal analysis experiments were performed using a Netsch STA 429 Thermal Analyzer. Tungsten (>99.99% purity) was used as the reference material. Both the reference and the samples were held in high purity alumina crucibles during testing. All tests were run in a flowing helium atmosphere at heating and cooling rates of 0.33°C/sec. The peak temperature during testing was approximately 1500°C. Previous experience [Ref 8] with this equipment and procedures indicated that a reproducibility of approximately 2°C in measured temperatures could be expected. Interpretation of the DTA curves was conducted using the convention established by MacIssac *et al.* [Ref 9].

III. Results and Discussion

3.1 General Welding Behavior

Cross-sections of the low heat input GTA welds are shown in Figure 2. As noted in Section 2, the welds were intended to be representative of typical current weld practice, and this was essentially achieved. Both welds of Figure 2 have the same cross-sectional area and the same

(a)

(b)

Figure 2. Cross-sections of low heat input GTA welds. (a) U-groove. (b) J-groove.

number of fill passes. The J-groove joint design provides a reasonably flat interface for testing of fusion line properties. In general, GTA weld processing of this alloy was found to be quite acceptable and no solidification or cold cracking was observed. Some difficulties were encountered in producing a full penetration root pass on the J-groove joint, although this is primarily a consequence of the joint design and is not thought to be related to any materials characteristic. Since the tensile and Charpy sample dimensions are much smaller than the plate thickness, no difficulties were encountered in removing samples well away from the root area of the J-groove welds. As determined by X-ray radiography, some welds contained occasional lack of fusion defects at the junctions between stringer passes which was composed of aligned pores. These defects were rare, but did affect some of the tensile test results and are discussed below.

The macrostructure of the EB welds is shown in Figure 3. The weld width at the midsection of the plate is approximately 2 mm, and this is considered reasonable for normal joint fit-up in material of this thickness. As with the GTA welds, no cracking was encountered during fabrication of the EB welds, and, as might be expected, there were no processing differences

Figure 3. Cross-section of EB weld.

between the solution treated and fully aged base plates. These observations are in agreement with earlier work on AerMet® 100 EB welding [Ref 4], where it was found that EB processing of this alloy is generally favorable.

3.2 Mechanical Properties Database

3.2.1 GTA Weld Tensile Behavior

The results of tensile tests of the weldment matrix are discussed in this section (raw data for all tensile tests, including the location of failure, are given in Appendix A). The average tensile properties of AerMet® 100 GTA welds in solution treated and fully aged base plates are shown in Table 5 and Table 6, respectively. From these data, several comparisons and observations can be made and, in order to simplify comparisons, this section is divided into several subsections which are based on sample orientation. In these comparisons, it is important to recognize that longitudinal tests represents all-weld metal tests while transverse tests sample the fusion zone, HAZ, and base metal simultaneously.

Table 5. GTA Weld Tensile Results (Solution Base Plate)

Heat Input	PWHT	Orientation	Y.S. MPa (ksi)	T.S. MPa (ksi)	Elongation (%)	R.A. (%)
Low	As-Welded	Trans	1489 (216)	1834 (266)	13.0	58.6
		Long	1565 (227)	1923 (279)	14.0	48.1
Low	-73/1+482/5	Trans	1765 (256)	2013 (292)	11.2	45.5
		Long	1765 (256)	1971 (286)	11.0	37.6
Low	204/1	Trans	1496 (217)	1820 (264)	12.8	56.6
		Long	1523 (221)	1937 (281)	11.7	34.2
High	As-Welded	Trans	1399 (203)	1847 (268)	6.5	42.3
		Long	1427 (207)	1916 (278)	14.2	50.4
High	-73/1+482/5	Trans	1744 (253)	1882 (273)	5.1	12.1
		Long	1813 (263)	1909 (277)	11.4	43.8
High	204/1	Trans	1482 (215)	1834 (266)	6.4	58.3
		Long	1510 (219)	1903 (276)	13.4	50.0
Base Metal	Solution Treated	Long	1399 (203)	1861 (270)	15.7	56.3
	-73/1+482/5	"	1799 (261)	2013 (292)	13.6	56.2
	204/1	"	1413 (205)	1827 (265)	14.0	52.0

Notes: Results are average of three tests.
 Transverse tests sample BM,FZ, and HAZ.
 Longitudinal tests are all-weld metal.

Table 6. GTA Weld Tensile Results (Aged Base Plate)

Heat Input	PWHT	Orientation	Y.S. MPa (ksi)	T.S. MPa (ksi)	Elongation (%)	R.A. (%)
Low	As-Welded	Trans	1551 (225)	1923 (279)	6.6	21.5
		Long	1565 (227)	1923 (279)	14.0	48.0
Low	-73/1+468/5	Trans	1730 (251)	1951 (283)	7.2	25.0
		Long	1785 (259)	2034 (295)	8.3	28.9
Low	204/1	Trans	1599 (232)	1930 (280)	8.0	27.1
		Long	1544 (224)	1971 (286)	14.4	50.0
High	As-Welded	Trans	1523 (221)	1868 (271)	5.9	7.4
		Long	1427 (207)	1916 (278)	14.2	50.4
High	-73/1+468/5	Trans	1737 (252)	1903 (276)	5.5	18.8
		Long	1820 (264)	1965 (285)	12.9	51.6
High	204/1	Trans	1530 (222)	1861 (270)	5.7	11.7
		Long	1551 (225)	1916 (278)	14.8	51.2
Base Metal	Full Age	Long	1758 (255)	2040 (296)	13.7	54.2
	-73/1+468/5	"	1785 (259)	1958 (284)	14.0	57.6
	204/1	"	1765 (256)	2034 (295)	14.0	55.6

Notes: Results are average of three tests.

Transverse tests sample BM,FZ, and HAZ.

Longitudinal tests are all-weld metal.

3.2.1.1 Longitudinal Tests

Comparing first the longitudinal (all-weld metal) tests in the as-welded condition, it appears that heat input generally has only a small effect on weld metal properties. The lower heat input, however, has a slightly higher yield strength than the high heat input, which is likely due to finer solidification feature sizes as well as additional aging during the higher number of passes in the low heat input welds. These values also compare favorably with the base metal in the solution treated condition. Further, the as-welded properties of the low heat input welds approach the AMS 6532 transverse minimums in the fully aged condition (YS = 1620 MPa, UTS = 1930 MPa, Elongation = 8%, RA = 45%) and this observation implies that as-welded properties may be acceptable for many applications. Direct aging of all-weld metal (longitudinal) samples at 482°C for 5 hr results in tensile properties which are comparable with those of fully aged base metal. A similar observation can be made for samples direct aged at 462°C, although the low heat input ductility was apparently impaired by weld porosity (discontinuous lack of fusion defects)

(a)

(b)

Figure 4. Examples of typical fractures in GTA welds showing (a) normal cup-and-cone fracture, and (b) ductility impaired by porosity.

similar to that shown in Figure 4. Careful inspection of the data for individual samples in Appendix A indicates that all low ductility tests for low heat input samples are associated with this type of weld defect. Thus, it is believed that weld defects, rather than an inherent weld metal characteristic, are responsible for the low ductility observed in some of the testing. Finally, the low temperature (204°C) aging cycle does not appear to have a significant effect on strength levels relative to the as-welded condition.

3.2.1.2 Transverse Tests

Considering now the transverse tests for the solution treated base plate, the as-welded strength levels are essentially the same as that for the all-weld metal tests. However, in the case of high heat input welds, the as-welded ductility is comparatively low. Examination of these samples (Appendix A) indicated that failure had occurred by either localized cup-and-cone fracture near the edge of the gage length, or in some cases by shear failure. An example of the shear failure is shown in Figure 5, and was found to be even more prevalent in the high heat input direct aged ($482^{\circ}\text{C}/5\text{ hr}$) transverse tests. This mode of failure is somewhat surprising, but is thought to be a consequence of the J-groove joint design used in the transverse tests and the relatively extensive softened region in the HAZ of the high heat input welds. As will be discussed in detail in Section 3.3, the softened (overaged) region occurs in both the low and high heat input welds, but its width is substantially greater in the high heat input welds. This softening, coupled with the near 45° inclination of the fusion boundary relative to the tensile loading axis (see Figure 2) in the current tests, results in a general tendency for shear failure. Further, upon aging, the fusion zone strength levels increase while those in the overaged region of the HAZ remain comparatively low. Thus, the tendency for shear failure is more prevalent in post-weld aged samples. As with reduced ductility due to porosity, the shear failure is primarily a consequence of the specific weld joint design and procedures, rather than a limiting materials characteristic. The shear failure does, however, illustrate the importance of the HAZ overaged region in controlling weldment properties, and for this reason it is considered extensively in Section 3.3.

Figure 5. Example of shear failure observed in high heat input GTA weld transverse tests.

In spite of these difficulties, a number of conclusions can be drawn from the transverse test data. For example, low heat input weldments in the solution treated base plate exceed the AMS 6532 transverse minimums and approach typical base metal values following direct aging at 482°C for 5 hr. Similarly, for the aged base plate, the low heat input welds approach AMS 6532 transverse minimums in the as-welded condition, and exceed these minimums in the direct aged (462°C) condition. It therefore seems apparent that high joint strength efficiencies can be achieved through suitable selection of weld joint design and GTA welding parameters. The low temperature direct age has little or no effect on strength levels for the transverse tests.

3.2.2 EB Weld Tensile Behavior

The tensile test results for EB welds in AerMet® 100 plate are summarized in Table 7 and Table 8, and the raw data for all tests is given in Appendix A. All EB weld tensile tests were conducted in the transverse orientation. Of the conditions shown in Table 7 and Table 8, the most likely to be applied in practice are the direct aged welds in solution treated base plate, as-welded in the aged base plate, direct aged welds in the aged plate, and possibly the low temperature PWHT in the aged plate.

For the direct aged (482°C) welds in the solution treated base plate, the tensile properties exceed the AMS 6532 longitudinal minimums, and are essentially equivalent to base plate typical values. Elongation and reduction of area are quite high, indicating that little or no strain localization occurs in these transverse tests and that HAZ softening is not a dominant factor for these welding parameters. As with the GTA welds, the low temperature age appears to have little effect on the weldment strength.

Table 7. EB Weld Tensile Results (Solution Base Plate)

Condition	Y.S. MPa (ksi)	T.S. MPa (ksi)	Elongation (%)	R.A. (%)
As-Welded	1351 (196)	1820 (264)	14.3	60.4
PWHT -73/1+482/5	1785 (259)	1992 (289)	12.3	55.2
PWHT 204/1	1406 (204)	1813 (263)	16.4	60.0
<u>Base plate</u> Solution Treated	1385 (201)	1896 (275)	15.9	55.0
-73/1+482/5	1806 (262)	2013 (292)	13.5	55.2
204/1	1406 (204)	1813 (263)	14.5	54.0

Notes: Results are average of three tests.
Transverse tests sample BM,FZ, and HAZ.

Table 8. EB Weld Tensile Results (Aged Base Plate)

Condition		Y.S. MPa (ksi)	T.S. MPa (ksi)	Elongation (%)	R.A. (%)
As-Welded		1606 (233)	1992 (289)	12.4	53.3
PWHT -73/1+468/5		1751 (254)	1992 (289)	14.8	60.7
PWHT 204/1		1537 (223)	1861 (270)	9.3	49.5
<u>Base plate</u>	Full Age	1765 (256)	2034 (295)	13.9	55.2
	-73/1+468/5	1765 (256)	1951 (283)	14.8	60.0
	204/1	1765 (256)	2040 (296)	13.7	54.3

Notes: Results are average of three tests.
Transverse tests sample BM,FZ, and HAZ.

In the as-welded condition, welds in the aged base plate approach AMS 6532 transverse minimums (YS = 1620 MPa, UTS = 1930 MPa, Elongation = 8%, RA = 45%), so that it seems apparent that as-welded properties may be suitable for many applications. Clearly, this observation is useful for situations where post-weld heat treatment is not feasible. It is also interesting to note that the elongation values are quite high, indicating that strain localization, although likely in the fusion zone, is not a strong factor in these tests. This observation is somewhat surprising given the fact that the fusion zone in these welds is a relatively small part of the gage length, and is expected to be in essentially the solution treated condition.

The tensile properties of direct aged (468°C) welds in aged base plate exceed AMS 6532 longitudinal minimums (YS = 1620 MPa, UTS = 1930 MPa, Elongation = 10%, RA = 55%) and are very similar to the base plate values. Thus, joint efficiencies for this condition are very close to 100%. As observed for all the EB weldments, the softened region of the HAZ does not appear to play a strong role in determining tensile behavior for these welds, although samples in this condition fail outside of the weld fusion zone. In addition, the 462°C/5 hr PWHT does not have a significant effect on the base metal properties.

Interestingly, low temperature PWHT of EB welds in aged base plate appears to result in a slight degradation of ductility in these tests. This degradation was not as apparent in the GTA welds in the same condition, in part because of the porosity and shear failure problems discussed earlier. The loss of ductility in EB welds heat treated at 204°C is thought to result from precipitation of cementite (Section 3.3) and strain localization in the weld fusion zone during testing.

3.2.3 GTA Weld Impact Behavior

The results of Charpy impact tests of the weldment matrix are discussed in this section (raw data for all impact tests are presented in Appendix B). Table 9 and Table 10 display the average centerline (weld metal) and fusion line impact energies for GTA welds in solution treated and aged base metal, respectively. Also shown are the base metal values for each condition. In general, the toughness of the weld fusion line is always superior to the weld fusion zone for all conditions, so that the fusion zone region of welds is likely to control toughness for the entire weldment. As might be expected from the difficulty associated with locating the notch on tests of this type and the fact that the fusion line is not completely planar, it is also important to note that fusion line tests are, in general, often scattered.

The as-welded centerline values (in either the solution treated or aged base plate) represent the basic toughness of the as-deposited weld metal. For this condition, the weld metal has roughly 60-70% of the impact toughness of fully aged base plate, while the fusion line toughness for welds in either solution treated or aged base plate are approximately 85 and 120%, respectively, that of normally aged base metal. As was noted in Section 3.5.1, the strength level for as-deposited low heat input GTA welds approaches that for the AMS 6532 transverse minimums. Thus, the as-welded condition in aged base plate has reasonable tensile and impact properties relative to

Table 9. GTA Weld Charpy Impact Results (Solution Base Plate)

Heat Input	PWHT	Base Metal Impact Energy, J (ft-lb)	Notch Location	Impact Energy J (ft-lb)
Low	As-Welded	61.2 (45.1)	Centerline Fusion Line	29.3 (21.6) 41.4 (30.5)
Low	-73/1+482/5	49.2 (36.3)	Centerline Fusion Line	25.1 (18.5) 47.9 (35.3)
Low	204/1	71.9 (53.0)	Centerline Fusion Line	22.2 (16.4) 38.6 (28.5)
High	As-Welded	68.5 (50.5)	Centerline Fusion Line	26.6 (19.6) 35.3 (26.0)
High	-73/1+482/5	50.0 (36.9)	Centerline Fusion Line	22.8 (16.8) 52.6 (38.8)
High	204/1	69.6 (51.3)	Centerline Fusion Line	22.5 (16.6) 36.3 (26.8)

Note: All values are averages of three tests per condition.

Table 10. GTA Weld Charpy Impact Results (Aged Base Plate)

Heat Input	PWHT	Base Metal Impact Energy, J (ft-lb)	Notch Location	Impact Energy J (ft-lb)
Low	As-Welded	45.0 (33.2)	Centerline Fusion Line	29.3 (21.6) 59.0 (43.5)
Low	-73/1+468/5	48.3 (35.6)	Centerline Fusion Line	22.2 (16.4) 46.5 (34.3)
Low	204/1	46.0 (33.9)	Centerline Fusion Line	26.2 (19.3) 46.4 (34.2)
High	As-Welded	42.7 (31.5)	Centerline Fusion Line	26.6 (19.6) 54.6 (40.3)
High	-73/1+468/5	48.3 (35.6)	Centerline Fusion Line	21.4 (15.8) 50.4 (37.2)
High	204/1	41.5 (30.6)	Centerline Fusion Line	28.5 (21.0) 60.3 (44.5)

Note: All values are averages of three tests per condition.

normally aged wrought material, and is likely to be a useful condition for many applications. In terms of toughness, the GTA weld heat input has only a small effect on as-welded properties, although the high heat input welds are generally inferior to the low heat input welds. Note that only one set of tests was conducted for base plate in the aged condition, and that these values are reported in both Table 9 and Table 10.

For direct aged (482°C) welds in solution treated plate, the centerline impact toughness is approximately half that of the base plate following PWHT. Here again, the lower heat input welds have slightly superior impact toughness in comparison to the high heat input. The fusion line toughness for this condition is similar to the base metal, so that in terms of toughness the weld metal is the lowest toughness region of the weldment. As might be expected from the aging data given by Novotny [Ref 10], direct aging (468°C) of welds in aged base plate results in slightly lower fusion zone toughness (and higher strength). Therefore there is some latitude in the selection of PWHT, depending on weld requirements for strength and toughness.

The low temperature PWHT appears to have differing effects on the base metal and weld metal, depending on base metal condition and welding parameters, but, except for its effect on solution treated base metal, the effect is relatively slight.

3.2.4 EB Weld Impact Behavior

Average Charpy impact test results for the EB weldments are shown in Table 11 and Table 12 for the solution treated and aged base plates, respectively (raw data for all tests included in Appendix B). Base metal impact values for the corresponding conditions are given in Table 9 and Table 10.

For centerline tests of the solution treated base plate, the as-welded and the direct aged (482°C) values are similar to those observed in the GTA welds for the same conditions. Thus, there is no apparent toughness advantage for the EB welding under the present conditions. For the reasons noted in Section 3.4.3, fusion line tests are generally somewhat scattered, but invariably exceed those measured at the weld centerline.

Direct aged (468°C) impact energies in the aged base plate are comparatively low (note that a retest of this condition resulted in similar values). It is possible that these low values result from two primary sources. First, the PWHT temperature (which was originally selected to minimize overaging of the base plate) generally results in lower impact toughness and higher strength. The aging data given by Novotny [Ref 10] for wrought material shows that impact toughness is a very strong function of the 5 hr aging temperature for temperatures in the range of 450-510°C. In fact, for wrought material, the toughness of material aged at 468°C is nominally near 36 J, while that for material aged at 482°C is usually near 55 J. Second, the low toughness region is confined in the EB weld impact samples to a very narrow (approximately 2 mm) region along the crack path. It is believed that the first of these effects is more predominant, and to test this possibility, samples from EB welds in aged plate were direct aged at 482°C for 5 hr, and the results are included in

Table 11. EB Weld Charpy Impact Results (Solution Base Plate)

Condition	Notch Location	Impact Energy J (ft-lb)
As-Welded	Centerline	30.0 (22.1)
	Fusion Line	48.8 (36.0)
PWHT -73/1+482/5	Centerline	24.0 (17.7)
	Fusion Line	39.6 (29.2)
PWHT 204/1	Centerline	42.3 (31.2)
	Fusion Line	61.3 (45.2)

Note: All values are averages of three tests per condition.

Table 12. EB Weld Charpy Impact Results (Aged Base Plate)

Condition	Notch Location	Impact Energy J (ft-lb)
As-Welded	Centerline	26.7 (19.7)
	Fusion Line	37.0 (27.3)
PWHT -73/1+468/5	Centerline	15.7 (11.6)
	Retest	13.6 (10.0)
	Fusion Line	26.4 (19.5)
PWHT -73/1+482/5	Centerline	25.9 (19.1)
PWHT 204/1	Centerline	39.5 (29.1)
	Fusion Line	58.2 (42.9)

Note: All values are averages of three tests per condition.

Table 12. As shown, substantially higher impact values, which are similar to those for the GTA welds in the same conditions, were obtained. Thus, it seems apparent that the weld metal displays a similar sensitivity to aging temperature as that observed in wrought material [Ref 10]. Moreover, it is clear that there is PWHT temperatures must be carefully selected and based on a trade-off between specific requirements for weld toughness and strength. Of course, the additional aging of the base plate during PWHT must also be considered.

It is interesting to note that the low temperature (204°C) PWHT has a different effect on EB weld metal toughness than it does on GTA weld metal toughness. For the GTA welds the 204°C PWHT resulted in slight reductions in toughness, while the same heat treatment results in an improvement in centerline toughness for the EB welds (for both base plate conditions). Although the underlying reasons for this difference are not clear, it is likely that it is a result of the multiple reheating cycles, which probably exceed 204°C, in the GTA welds. Thus, for EB welding situations in which a full post-weld aging treatment is not feasible or needed from a strength perspective, some toughness gain can be realized through a low temperature PWHT.

3.2.5 Fully Re-Heat Treated Weldments

The average tensile properties for fully re-heat treated welds are shown in Table 13 and raw data for these welds is given in Appendix A. Based on prior work [Ref 5], the GTA welds were expected to have properties similar to that of the base metal. Unfortunately, however, the GTA weld tensile samples contained a significant amount of porosity resulting from lack of fusion (see notes in Appendix A), and thus displayed poor ductility. Electron beam weldments display, as

Table 13. Tensile Properties for Fully Heat Treated Welds

Welding Process	Heat Input	PWHT	Specimen Orientation	Yield MPa (ksi)	Tensile MPa (ksi)	Elong. (%)	R.A. (%)
GTAW	Low	Full	Base Metal	1794 (260.2)	1969 (285.5)	14.7	59.1
			Transverse	1802 (261.3)	1932 (280.2)	2.1	4.2
			Longitudinal	1685 (244.4)	1924 (279.1)	6.9	24.2
EB	1 Pass	Full	Base Metal	1794 (260.2)	1969 (285.5)	14.7	59.1
			Transverse	1838 (266.5)	1991 (288.8)	14.4	60.8

Notes: Full PWHT includes 885°C/1hr, air cool, -73°C/1 hr + 482°C/5 hr.
Values are averages of three tests per condition.

Table 14. Charpy Impact Properties for Fully Heat Treated Welds

Welding Process	Heat Input	PWHT	Notch Location	Impact Energy J (ft-lb)
GTAW	Low	Full	Base Metal	54.1 (39.9)
			Centerline	21.6 (15.9)
			Fusion Line	23.5 (17.3)
EB	1 Pass	Full	Base Metal	54.1 (39.9)
			Centerline	33.9 (25.0)
			Fusion Line	48.1 (35.5)

Notes: Full PWHT includes 885°C/1hr, air cool, -73°C/1 hr + 482°C/5 hr.
Values are averages of three tests per condition.

expected, strength and ductility similar to that of the base plate. Based on the EB results, the data from direct aged weldments, and from previous work [Ref 5], it is believed that fully reheat treated GTA welds without defects would also display tensile behavior similar to that for wrought material.

The results of Charpy impact tests for the fully heat treated welds are shown in Table 14 and the raw data for all tests is given in Appendix B. Centerline values for the GTA weld are somewhat lower than those obtained for direct aged welds, although the strength levels are similar.

Thus, it is likely that impact tests were affected by the same porosity encountered in the tensile tests. The fully treated EB weld shows impact energies which are approximately 63% and 90%, for centerline and fusion line tests, respectively, that for the base plate and are appreciably higher than for direct aged (482°C) welds. Therefore, the fully heat treated condition, for cases where it can be applied, provides some toughness advantage relative to directly aged welds. Of course, full heat treatment of welds is not always applicable and the potential distortion which is likely to be encountered during the austenitization step must be considered.

3.2.6 Mechanical Testing Summary

Because of the large number of sample conditions and test configurations, it is appropriate to summarize some of the basic conclusions which can be drawn from the baseline weld properties study.

High strength levels can be achieved in either solution treated or aged base plate with either the cold wire feed GTA or EB welding process. In general, for direct aged (at temperatures near 482°C) welds in either base plate, the overaged region of the HAZ is the strength limiting factor, while the fusion zone controls the impact toughness. The data also indicate that within the range of GTA welding parameters and PWHT conditions evaluated, suitable procedures can be defined which can provide useful combinations of strength and toughness. Compared to the GTA welds in the direct aged condition, the EB welds generally have somewhat superior strength, but not necessarily superior toughness. This observation seems to imply that the finer solidification features which are usually associated with EB welding do not play a dominant role in overall mechanical properties. It is believed that the underlying reasons for this are likely associated with the martensitic transformation and the structure that results from it. Thus, the characteristics of the solid state transformation would appear to dominate over the details of the solidification microstructure.

For the GTA welding conditions chosen, the welding parameters for the low heat input welds are probably marginal for reproducibly fabricating defect free welds. Nevertheless, for those GTA welds which are free from welding defects, the lower heat input welds generally display superior strength and toughness for the same PWHT conditions. Although the differences are not large, the data implies that low heat input conditions are preferable. This conclusion is most apparent in the shear failures of the high heat input welds. Clearly, this type of failure could be avoided by a more suitable choice of the joint geometry in the high heat input welds, but it is important to note that shear failure did not occur in the low heat input welds, even though the joint configuration was the same. Thus, the narrower (and likely higher strength) softened region in the HAZ of low heat input welds behaves more favorably even in joints with poor geometry.

3.3 Welding Metallurgy

3.3.1 Melting and Solidification

Initial characterization of the welding metallurgy of AerMet® 100 Alloy included the use of differential thermal analysis (DTA) to determine the temperature ranges associated with solid state, melting, and solidification reactions. The result of this analysis for solution treated starting material is shown in Figure 6. For this heating rate, austenite begins to form in the temperature

Figure 6. DTA scan for solution treated AerMet® 100 Alloy for heating and cooling rates of 0.33°C/sec. Temperatures shown are averages of two tests.

range of 450-500°C (the lower and upper critical temperatures are discussed in more detail in a later section), and the formation of austenite is complete at approximately 807°C. Melting of austenite initiates at 1435°C, and continues until the liquidus is reached at 1498°C. On-cooling, and after some undercooling, solid begins to form at 1466°C, and the on-cooling solidus is reached at 1445°C. The apparent M_s temperature for the as-solidified alloy is near 307°C.

From these analyses, several features of the alloy are apparent. First, as might be expected, the transformation to austenite occurs over a relatively broad temperature range. As will

be discussed in detail below, however, the transformation to austenite in this alloy is complex and both the mode and critical temperatures associated with the transformation are strongly dependent on the heating rate. With respect to welding, the temperatures of primary interest are the on-heating liquidus and the on-cooling solidus, as these define the temperature range over which weld solidification occurs and hot cracking can develop. For AerMet® 100 Alloy, this temperature range is approximately 53°C, which is significantly smaller than the 150-200°C solidification temperature range normally encountered in hot cracking sensitive alloys. Thus, the narrow solidification temperature range of this alloy accounts in part, for its apparent resistance to hot cracking. Conversely, the DTA results show that the alloy solidifies as austenite, and in general alloys which solidify as austenite have a lower resistance to hot cracking than alloys which solidify as ferrite and subsequently transform to austenite. This difference is related to the lower solubility of impurity species (particularly P and S) in austenite, and the segregation of these impurities to the liquid during solidification. As a result of the toughness requirements for this alloy, however, it is normally produced from high purity starting materials and with the use of very clean processing so that the levels of impurity species are quite low (Table 1). Thus, even though the alloy solidifies as austenite, the narrow solidification temperature range and low impurity contents can be expected to result in a low susceptibility to solidification hot cracking, as was observed in the fabrication of the EB and GTA test welds.

The apparent M_s temperature of 307°C is significantly higher than the 225°C normally reported for wrought AerMet® 100 Alloy. This difference, although it was not investigated in detail, is likely due to composition gradients developed during solidification. For this reason, the M_s temperature for weld fusion zones are likely to be similar to the value measured by DTA.

3.3.2 HAZ Microstructure Evolution

The mechanical response of the GTA welds implies that the evolution of microstructure and properties in the HAZ of AerMet® 100 Alloy welds is an important feature of the overall service weldability of the alloy. This implication is further illustrated in Figure 7 which shows the structure of, and a microhardness traverse across, a single bead-on-plate GTA weld in aged plate in the as welded condition. Several important regions are apparent in the hardness traverse. In the fusion zone the hardness is relatively low, and the microstructure consists essentially of fresh martensite. Thus, the hardness is similar to that for solution treated base metal and the fusion zone can be expected to respond to post-weld aging treatments (as was observed in the tensile test results). The near fusion line region of the HAZ also consists of fresh martensite and also responds to post-weld aging. In locations further out in the HAZ the hardness steadily increases toward the unaffected aged base plate, but there is a trough in the hardness (in the range of

(a)

(b)

Figure 7. (a) Macrostructure of low heat input bead-on-plate GTA weld in aged plate. (b) Hardness traverse across weld zones.

approximately 3.5-4.5 mm). The region adjacent to, and within, the trough is particularly important since it does not respond to post-weld aging. As a result, the primary emphasis of the welding metallurgy study was placed on characterization of the microstructure and properties evolution in this region of the HAZ. This characterization included determination of the temperatures and microstructures associated with the trough, as well as development of kinetic descriptions of the softening and reaustenitization processes.

3.3.2.1 Gleeble Testing

Gleeble simulations were used to determine the temperatures associated with the hardness trough in the HAZ. For these simulations, thermal cycles similar to those shown in Figure 8 were used. The thermal cycles correspond to those expected for the high heat input GTA weld, and peak temperature was varied between 460 and 910°C. Hardness and microstructure were evaluated for each peak temperature.

Figure 8. Typical thermal cycle used for Gleeble HAZ simulations. The thermal cycle corresponds to the high heat input GTA welds and a 675°C peak temperature.

The results of the thermal cycle simulations on both aged and solution treated base plate are shown in Figure 9. For both the aged and solution treated base plates, the minimum in hardness occurs for the 710°C peak temperature thermal cycle. In addition, for the aged base plate, post-simulation aging does not affect the level or temperatures associated with the hardness trough. This implies that the softening is associated with overaging, although it is also conceivable that the formation of reverted austenite (discussed in more detail below) also contributes to the softening. The increase in hardness at peak temperatures of 760°C and above is associated with the formation of new austenite (and fresh martensite on cooling). It is also interesting to note that some hardening apparently occurs in the solution treated material at the peak temperatures of 560 and 610°C. This implies that in the temperature range examined, the aging reactions in AerMet® 100 can be quite rapid. Moreover, the steep recovery in hardness at temperatures of 760°C and above implies that the austenite formation kinetics are also quite rapid in this temperature range for the heating rates associated with fusion welding.

(a)

(b)

Figure 9. Hardness of weld simulation samples as a function of peak temperature.
(a) Aged base plate. (b) Solution treated base plate.

3.3.2.2 Metallographic Analysis

Representative transmission electron micrographs of the HAZ simulations are shown in Figure 10. In general, the TEM evaluations shown in Figure 10 and additional extraction replica analyses revealed several important features. First, relatively large elongated (rod like) alloy

Figure 10. TEM micrographs of HAZ simulation samples. (a) 660°C. (b) 710°C. (c) 760°C.

carbides were evident in samples cycled to peak temperatures near 660 and 710°C. For samples treated above this temperature, precipitation was less evident, and those precipitates that remained were generally spherical in shape. The overall microstructures of the 660 and 710°C samples were similar, although there was evidence that recovery had proceeded further in the 710°C samples. As determined by EDS analysis, little reverted austenite formed in the 660°C samples, while austenite could be detected along martensite lath boundaries in the 710°C simulations.

Attempts to quantitatively determine the room temperature austenite fraction in the Gleeble simulations were unsuccessful. This difficulty appeared to be the result of the relatively small sample area available for analysis (although mounts with multiple samples were tried), as well peak overlap and the very fine scale of the reverted austenite which is present as thin films along the martensite lath boundaries. In an effort to overcome these difficulties, small (approximately 13 mm cubes) were rapidly heat treated in a salt pot at Carpenter Technology Corporation. For these experiments, the salt pot was maintained at temperatures replicating the peak temperatures in the Gleeble experiments. The samples (both solution treated and fully aged) were instrumented with a thermocouple located at the center of the cube, and were treated by immersion in the salt pot until the sample temperature reached the bath temperature. This was followed immediately by a water quench. The heat treatment required only a few seconds, but the heating time was significantly longer than that for the Gleeble tests. Hardness and fraction austenite were determined for each peak temperature and the results of these tests are shown in Figure 11. The results indicate that the

Figure 11. Hardness and fraction austenite for salt pot heat treated samples compared with Gleeble HAZ simulations.

salt pot heat treatments generally reproduce the softened region observed in the Gleeble simulations, although the heat treatments at 760°C resulted in lower hardness than those for the Gleeble simulations. It is believed that the differences at the 760°C peak temperature result from differences in the mechanism of austenite formation at different heating rates and these differences are discussed in greater detail in a subsequent section. In any case, an increasing fraction of reverted austenite is observed with increasing peak temperature in the salt pot treated samples. At the higher peak temperatures (above 760°C) there is likely much more austenite present at temperature, but the composition of this austenite is such that it transforms to martensite on cooling to room temperature [Ref. 10, 11]. The amount of austenite in samples rapidly heated to peak temperatures above 760-810°C is similar to that observed in material solution treated under normal conditions.

Thus, the basic microstructural evolution in the softened region of the HAZ consists principally of precipitate overaging (with the formation of 10-15% reverted austenite and possibility of some recovery). This softened region is bounded at lower peak temperatures by the rate at which precipitate coarsening occurs, and it is bounded at higher temperatures by the formation of transformable austenite. For the HAZ simulations it appears that precipitate dissolution can lag behind the formation of new transformable austenite, so that at high heating rates and peak temperatures near 760°C, fresh martensite, which effectively contains preexisting precipitates, can form. Such a mechanism accounts for the high hardness (relative to the fully solution treated material) for peak temperatures just above those associated with the hardness trough (*i.e.* at a position of about 3.5 mm in Figure 7). Since the softened region is bounded by two distinct processes, it is appropriate to develop kinetic descriptions of these processes so that the effects of welding process changes can be estimated numerically.

3.3.2.3 Overaging Kinetics

Description of the overaging process in the HAZ of AerMet® 100 Alloy requires a numerical model which can be applied to the rapid thermal cycle associated with the welding process. In this section a semi-empirical model is therefore developed and applied to the thermal cycles used in the HAZ simulations. The model is based on classical coarsening models, and the additivity rule is used, Section 3.2.3.4 to apply this model to weld thermal cycles. Aging data for secondary hardening steels is usually portrayed by means of a time-temperature parameter such as the Larson-Miller parameter (*e.g.* by the construction of a master tempering plot). However, such data cannot easily be adapted to weld thermal cycles because the form of these hardness or strength versus aging parameter curves are usually complex. In the current work, therefore, an attempt is made to define a parameter for which the relationship between the parameter and hardness is simple. Since most available data is in the form of hardness as a function of time and temperature

for a variety of aging conditions, the parameter is first developed for AerMet® 100 hardness data [Ref 10], and then shown to be applicable to strength data.

It is first assumed that hardness is proportional to shear strength and that analysis for the strengthening effect of incoherent precipitates [Ref 12] applies,

$$H \propto \tau = \frac{\alpha \mu b}{\lambda} \quad [1]$$

where H is the hardness, τ is the shear strength, α is a geometric factor, μ is the shear modulus, b is the burgers vector, and λ is the interparticle spacing. For overaging, the volume fraction of precipitates is assumed constant, so that $\lambda \propto r$, where r is the precipitate radius and the precipitates are assumed to be spherical. Thus, in the overaging regime,

$$H \propto \frac{1}{r} \quad [2]$$

During aging, the precipitates are assumed to follow classical coarsening kinetics [Ref 12] (*i.e.* Lifshitz and V. V. Slysov [Ref 13], and Wagner [Ref 14], or LSW kinetics), so that

$$\bar{r}_t^3 = \bar{r}_0^3 + Kt, \quad [3]$$

where $\bar{r}_0$ is the mean initial particle radius and $\bar{r}_t$ is the particle radius at time t . Through direct measurements of carbide size during isothermal heat treatment, Yoo et. al. [Ref 15] and Lee et. al. [Ref 16] have shown that the carbide coarsening kinetics in AerMet® 100 Alloy are well approximated by the cubic dependence of Equation 3. The coefficient K in Equation 3 is generally given by an equation of the form

$$K = \frac{8}{9} \frac{D \sigma V_m C_\alpha(\infty)}{RT} f(\phi), \quad [4]$$

where D is the diffusivity, σ is the interfacial energy, V_m is the molar volume of precipitate, $C_\alpha(\infty)$ is the solubility of the solute, R is the ideal gas constant, T is the temperature, and $f(\phi)$ is a function which depends on the particular coarsening theory ($f(\phi) = 1$ for the LSW theory). If it is assumed that all factors other than the diffusivity are not functions of temperature and the temperature dependence of D can be represented by

$$D = D_0 \exp\left(\frac{-Q}{RT}\right), \quad [5]$$

then Equation 4 can be written as

$$K = K_1 \frac{\exp\left(\frac{-Q}{RT}\right)}{T}, \quad [6]$$

where K_1 is a constant. Combination of Equations 3 and 6 yields an expression for the coarsening rate as a function of time and temperature:

$$\bar{r}_t^3 = \bar{r}_0^3 + K_1 \frac{\exp\left(\frac{-Q}{RT}\right)}{T} t. \quad [7]$$

Substitution of Equation 7 into Equation 2 (with $\bar{r}_t = r$) and rearranging leads to the result

$$H \propto K_2, \quad [8]$$

where K_2 has the form

$$K_2 = \frac{1}{\left(\bar{r}_0^3 + \frac{K_1 t \exp\left(\frac{-Q}{RT}\right)}{T} \right)^{1/3}} \quad [9]$$

Thus, a plot of H versus K_2 should be linear. The most direct means to apply this methodology is to search for values of $\bar{r}_0$, K_1 , and Q which maximize the linearity of the plot. This searching can easily be accomplished in a spreadsheet application. Alternatively, if values for any of the three parameters are known, such as an initial precipitate size or the activation energy for the diffusion of a known precipitating species, these can be used to reduce the number of search parameters.

The procedure described above was used to fit the isothermal aging data for AerMet® 100 given by Novotny [Ref 10], and the results are shown in Figure 12. The optimized fit parameters

Figure 12. Plot of hardness versus optimized K_2 for isothermally aged [Ref 10] AerMet® 100. Dashed lines are 90% prediction limits.

for this data set are $\bar{r}_0 = 2.47 \times 10^{-8}$, $K_1 = 0.02718$, and $Q/R = 38990$, and the hardness can be represented by a line of the form.

$$\text{HRC} = 38.18 + 4.690 \times 10^{-7} (K_2) \quad r^2 = 0.9559 \quad [10]$$

These parameters linearize the data over a wide range of aging conditions, including those conditions near peak hardening and overaged. The apparent activation energy for the softening, 77,470 cal/mole, is higher than that which might be expected for coarsening of alloy carbides in AerMet® 100. For example, the activation energy for the diffusion of Mo in ferrite is given [Ref 17] as 57,700-60,000 cal/mole (the diffusion of Cr would be expected to have a similar activation energy). However, given the assumptions inherent in the model, such as spherical precipitates, *etc.*, this discrepancy is not thought to be a major drawback with respect to application of the model. Moreover, the fitting is not exceptionally sensitive to the activation energy within a specific range. For example, if Q is fixed at 60,000 cal/mole and the fit is optimized by varying only $\bar{r}_0$ and K_1 , the r^2 for the H vs. K_2 correlation is 0.9488 ($\bar{r}_0 = 1.65 \times 10^{-7}$, $K_1 = 1.01 \times 10^{-4}$) which is similar to the $r^2 = 0.9559$ obtained with the three parameter fit.

As a test, estimates from the model can be compared with additional aging data given by Novotny [Ref 10]. This data was developed for 5 hr aging over the temperature range from 400 to

Figure 13. Comparison of overaging model predictions (solid hardness line) with 5 hr aging data from Novotny [Ref 10]. Dashed lines are 90% prediction limits.

800°C, and was conducted to evaluate the formation of reverted austenite during aging. The range of temperatures is substantially larger than those used to determine the fit coefficients. The results of this comparison are shown in Figure 13. Also shown in Figure 13 is the room temperature austenite fraction as determined by X-ray diffraction. As shown, the model provides accurate predictions in the temperature range (450-550°C) where little reverted austenite forms during aging, and provides reasonable predictions in the temperature range (550-700°C) where appreciable fractions of stable reverted austenite form. As might be expected, for temperatures above approximately 700°C, where unstable austenite (*i.e.* fresh solution treated martensite at room temperature) is formed, the model significantly underestimates the hardness. The overaging model therefore appears to be generally applicable at temperatures up to approximately 700°C, even for relatively long aging times. For shorter aging times, such as weld thermal cycles (where less reverted austenite might be expected to form), the model should provide even better estimates over this temperature range.

The data of Figure 13 implies that the formation of reverted austenite in the temperature range between 500 and 700°C does not have a large effect on the hardness of AerMet® 100, and it is interesting to consider the possible reasons for this observation. Based on qualitative EDS measurements of martensite laths and austenite in HAZ simulation samples heat treated to peak temperatures in this range, the compositions of the austenite and martensite are not dramatically different, with only small differences in the levels of Co, Ni, and Cr observed. Since these

elements are substitutional in both constituents, it is not unreasonable to expect that the hardness of both phases would be similar. Moreover, the matrix hardness in this alloy following heating to this temperature range is likely to be dominated by the lath structure, which is effectively little changed by the formation of thin austenite films along the lath boundaries. Thus, although the overaging model does not explicitly incorporate the formation of reverted austenite (or recovery), this does not appear to have serious implications with respect to its application. The model does not incorporate the possibility that the carbide precipitate composition is a function of temperature. Ayer and Machmeier [Ref 18] have shown that the carbide stability in AerMet®100 is dependent on the aging temperature. However, in the overaged regime for which the current model is primarily applicable, the precipitates are generally very large and incoherent. Thus, their role in strengthening is not likely to be strongly dependent on their composition, and this factor would not be expected have serious implications with respect to application of the model.

3.3.2.4 HAZ Thermal Cycles

Application of the overaging model to weld HAZs requires a means to estimate the cumulative effect of thermal cycle on the kinetics. A commonly used approach for estimating this effect is the additivity principle [Ref 19, 20]. To apply the additivity principle, a thermal cycle is modeled as a series of small isothermal steps. At each step, the time elapsed is a fraction of the time required, at that temperature, to achieve a given amount of transformation or hardness. When these fractions sum to unity, the given hardness has been achieved. Formally, this can be written as

$$\int_0^t \frac{dt}{t_a(T)} = 1, \quad [11]$$

where t is the time and $t_a(T)$ is the time to reach the given hardness isothermally. For discrete time steps, Equation 11 can be written in the summation form, or

$$\sum_0^t \frac{\Delta t}{t_a(T)} = 1. \quad [12]$$

From Equation 10, $H = H_0 + mK_2$, where H_0 and m are the intercept and slope of the fit to the isothermal aging data, respectively, it is apparent that

$$H = H_0 + m \frac{1}{\left(\bar{I}_0^3 + \frac{K_1 t \exp\left(\frac{-Q}{RT}\right)}{T} \right)^{1/3}} \quad [13]$$

or, upon rearranging

$$t = t_a(T) = \frac{T \left[\left(\frac{m}{H - H_0} \right)^3 - \bar{I}_0^3 \right]}{K_1 \exp\left(\frac{-Q}{RT}\right)}. \quad [14]$$

For an isothermal hold, Equation 14 represents the time required to reach a given hardness, H , at a given temperature, T . Using a series of small isothermal steps to represent a thermal cycle, the summation of Equation 12 can be solved by using a goal seeking routine to find the hardness which satisfies the summation for that thermal cycle. Alternatively, the summation can be solved sequentially by calculating the softening during a given time step, adjusting the temperature to the temperature associated with the next time step and adjusting the start time to that which it would have been had the temperature always been at the new temperature, and iterating this procedure over the entire thermal cycle. In either case, identical results should be achieved.

Using Equations 12 and 14, together with the coefficients derived from the isothermal aging data of Novotny [Ref 10] and the actual thermal cycle from the Gleeble HAZ simulations, the hardness for the HAZ simulations can be estimated. The results of these estimates are shown in Figure 14, and the agreement between predicted and measured hardness is quite good. Thus, the overaging model, developed from isothermal aging data and applied to weld thermal cycles through the use of the additivity principle, appears to provide an appropriate means for estimating the HAZ softening in AerMet® 100. However, it should also be noted that the accuracy of predictions based on this model is dependent on the accuracy to which the thermal cycle at any HAZ location is known. For the Gleeble simulations in Figure 14, the thermal cycle is directly measured and accurately known. For general application to welds, the thermal cycle at a given HAZ location must be estimated from thermal models of the weld process and parameters, and in these cases the accuracy is therefore partially dependent on the accuracy of the thermal model. In addition, real welds, and in particular multipass welds, can have multiple thermal cycles and complex residual stress distribution within the HAZ. Although the model should be capable of treating the multiple

Figure 14. Comparison of model predictions and measured hardness for HAZ simulations in aged base plate. Dashed lines are 90% prediction limits.

thermal cycles, the potential effect of residual stresses on the softening behavior is not incorporated in the current model.

It is also interesting to consider the applicability of the aging model for estimating properties other than hardness. As shown by Novotny [Ref 10], the ultimate tensile strength of AerMet® 100 is essentially linearly related to hardness over a wide range. Thus, it might be expected that tensile strength should also be described by the current model. Figure 15 shows

Figure 15. Ultimate tensile strength versus K_2 for AerMet® 100. The plots were developed using (a) coefficients from hardness correlation, and (b) coefficients optimized on UTS data. Dashed line are 90% prediction limits.

Figure 16. Yield strength versus K_2 for AerMet® 100. The plots were developed using (a) coefficients from hardness correlation, and (b) coefficients optimized on YS data. Dashed line are 90% prediction limits.

UTS data from Novotny [Ref 10] versus K_2 using the coefficients developed from the hardness data, Figure 15(a), as well as with coefficients optimized on the UTS data directly, Figure 15(b). For the correlation using the UTS data, the fit coefficients were $\bar{\Gamma}_0 = 9.02 \times 10^{-9}$, $K_1 = 0.000390$, and $Q/R = 37750$, although the correlation coefficient for the direct UTS fits is only marginally superior to the correlation using the hardness coefficients ($r^2 = 0.9680$ vs. $r^2 = 0.9646$). In either case, however, it appears that the model should be generally applicable for estimating the tensile strength of AerMet® 100 following thermal cycling.

Correlation with the yield strength is not as promising as that for the tensile strength. Figure 16 shows YS data from Novotny [Ref 10] versus K_2 using the coefficients developed from the hardness data, Figure 16(a), as well as with coefficients optimized on the YS data directly, Figure 16(b). Although the coefficients linearize the YS data for relative short aging times and/or lower aging temperatures (high K_2), the error associated with significant overaging (low K_2) is appreciable ($r^2 = 0.8379$). In order to overcome this difficulty, direct fits of the YS data required substantial weighting of the data in the overaged regime. For the direct YS fits ($\bar{\Gamma}_0 = 1.95 \times 10^{-8}$, $K_1 = 0.04095$, and $Q/R = 41130$) of Figure 16 the data becomes even more of a bimodal distribution between low and high K_2 values. Thus, although a representation of the YS data was achieved, it is not fully satisfactory and would required additional validation. Nevertheless, it appears that the current approach is generally applicable to estimating hardness, tensile strength, and potentially, yield strength.

3.3.2.5 Austenitization Kinetics

As discussed in Section 3.3.2.1, the high temperature edge of the HAZ softened region is bounded by the formation of new transformable austenite. For this reason, it is important to

develop a description of the austenite formation kinetics at rapid heating rates, and this is the subject of this section. In this portion of the work, the austenite formation kinetics were examined by Gleeble dilatometry at heating rates in the range of 0.1 to 2500°C/sec. Figure 17 shows the typical dilatation response of solution treated AerMet® 100 for a heating rate of 100°C/sec. In principal the lower and upper critical temperatures can be estimated from these curves by determining the initial departure from, and final approach to, the linear portions of the curve associated with the low and high temperature constituents, respectively. In practice, these temperatures were determined as the temperatures at which the difference in dilatation between the curve and the extrapolated linear fit of the curve preceding the initial contraction was either 2 or 98% of the dilatation difference between the low and high temperature linear extrapolations (*i.e.* a lever rule approach). A similar approach can be used to estimate the apparent fraction transformed over the entire transformation range. However, for these determinations it should be noted that the conversion of dilatation data to fraction transformed is very approximate since the austenite which forms is likely to have a composition which depends on temperature (as does the decomposing martensite/ferrite) and therefore a lever rule estimation of the fraction transformed is not strictly correct. In addition, heating rates of 200°C/sec and higher were found to exhibit unusual features which further complicate interpretation of the dilatation data. Figure 18 shows the apparent austenite fraction (determined by lever rule conversion of the dilatation data) for several representative heating rates. From these curves, several trends are apparent. First, the lower critical temperature initially increases with heating rate, and then becomes essentially constant at the higher heating rates. The behavior of the upper critical temperature with heating rate is not, however, as clearly defined. Finally, for heating rates of 200°C/sec and above, a relatively sharp and reproducible break (denoted by an x on the curve for 300°C/sec in Figure 18) appears in the

Figure 17. Typical dilatation curves for solution treated AerMet® 100. (a) Entire heating and cooling curve. (b) Detail of austenite formation region on-heating.

Figure 18. Apparent austenite fraction as determined by lever rule conversion of dilatation data for several heating rates. x denotes break in 300°C/sec curve.

curves following an initially rapid transient. It should also be noted that at temperatures above the break in the 300°C/sec curve shown, the curve becomes extremely jagged. However, this is not thought to be real, but rather, appears to be an artifact of the data collection equipment and is due to aliasing of the Gleeble power control and data collection electronics (its appearance can be altered by varying the data collection rates). As will be discussed in more detail below, it is believed that the break which occurs for heating rates of 200°C/sec above is the termination of the formation of austenite, and that the approach to the final expansion line is homogenization of the austenite by diffusional adjustments. For this reason, in subsequent continuous heating transformation (CHT) diagrams, the upper critical temperature for heating rates above 200°C/sec is depicted as the temperature at which the break in the curve appears. The temperature at which the dilatation reaches the high temperature expansion line is depicted as a separate point on the diagrams.

Figure 19 shows the CHT diagram for solution treated AerMet® 100 for heating rates in the range of 0.1 to 2500°C/sec. The diagram for this alloy is extremely complex, and appears to consist of two distinct heating rate regions. The first of these is the lower heating rate range from 0.1 to 25°C/sec, and the other is the range including 50°C/sec and above. These regions are believed to constitute different mechanisms of austenite formation, and are discussed separately.

Considering the low heating rate regime, the details of which are shown in Figure 20, several important features are apparent. Both the 98% and apparent 50% austenite boundaries are approximately isothermal, which implies that for these heating rates the austenite formation (after

Figure 19. Continuous heating transformation diagram for solution treated AerMet® 100. Error bars for 100°C/sec are $\pm 2\sigma$ for four tests.

Figure 20. Continuous heating transformation diagram for solution treated AerMet® 100 for heating rates from 0.1 to 25°C/sec.

nucleation) essentially follows the equilibrium phase boundaries. The lower critical temperature increases with increasing heating rate, which is typical of the increasing delay in nucleation with increasing heating rate. Given the rapidity with which aging can occur in this alloy at temperatures in the range of the lower critical temperature (Section 3.3.2.3), however, it is also conceivable that

the martensite matrix composition can change during the heating below the critical temperature. Thus, some variation in the lower critical temperature resulting from changes in the martensite matrix composition cannot be discounted. This variation would be expected to be different, or nonexistent, for tests on aged or fully aged material, although heating tests for these starting microstructures were not conducted in the present work.

The CHT diagram for AerMet® 100 at heating rates of 50°C/sec and above is considerably more complex and is shown in Figure 21. For the solution treated material, the lower critical temperature is essentially constant (at approximately 700°C) for heating rates in the range of 50 to 300°C/sec, and decreases monotonically at heating rates above 300°C/sec. Conversely, the temperature tentatively associated with homogenization increases with heating rate as would be expected.

The decrease in lower critical temperature with increasing heating rate is interesting and unusual. For a diffusion controlled transformation, it would normally be expected that the critical temperature would continue to increase with increasing heating rate, as was observed for the lower heating rate tests. Moreover, the plateau and decrease in the lower critical temperature are associated with the region where the break in the dilatation response appears (see Figure 18). Thus, it seems likely that the mechanism of transformation is different in the high heating rate regime, and that the redistribution of solute (precipitation of alloy carbides) prior to transformation

Figure 21. Continuous heating transformation diagram for solution treated AerMet® 100 for heating rates from 50 to 2500°C/sec. Error bars for 100°C/sec are $\pm 2\sigma$. Open points are from multiple heating rate tests.

plays an important role in determining the critical temperatures at the high heating rates. As a test of this possibility, multiple heating rate tests were conducted in which the sample was heated at 50°C/sec to 675°C (*i.e.* to just below the lower critical temperature for that heating rate), and then rapidly heated at either 500, 1000, or 2500°C/sec. The results of these tests are shown as open points on Figure 21. For these tests, the lower and upper critical temperatures are effectively constant, at 700-710 and 735-740°C, respectively, and the dilatation curve did not show the break which was characteristic of the single heating rate tests. The formation of austenite in the high heating rate regime is therefore consistent with a shear (martensitic) type transformation in that it exhibits constant start and finish temperatures.

It is interesting to consider the possibility that austenite formation at high heating rates occurs by a massive transformation mechanism. Bibby and Parr [Ref 21] have shown that for pure iron, the decomposition of austenite can occur by a massive transformation mechanism for cooling rates intermediate between those resulting in normal nucleation and growth and those resulting in a martensitic transformation. For the massive transformation, the reaction start temperature was independent of cooling rate in a manner similar to that for martensitic transformations (though at a higher temperature). No data was given for the transformation finish temperature, and no definitive studies of the cooling or heating rate dependence of the finish temperature for massive transformations could be found. Since the massive transformation involves a diffusional growth mechanism (albeit very short range), however, it seems unlikely that the finish temperature would be independent of cooling or heating rate. Moreover, the Fe-Ni system is known to transform martensitically on heating and cooling, so it is perhaps not surprising that the Fe-Ni-Co system could as well. Thus, for the present, it is reasonable to conclude that the on heating transformation to austenite in AerMet® 100 is martensitic-like. Clearly, additional work is required to unequivocally identify the mechanism of the transformation.

As shown above, the high heating rate transformation start and finish temperatures are heating rate dependent for the solution treated material. Conversely, it is expected that the start and finish temperatures for aged or overaged annealed material would be independent of heating rate and similar to that measured in the multiple heating rate tests (Aged and overaged starting materials would be expected to have similar matrix compositions since the normal aging procedure is past the peak aging conditions for this alloy [Ref 10]. Since appreciable aging can occur prior to the transformation at low heating rates (*i.e.* less than 50°C/sec) for any initial microstructure, the low heating rate transformation characteristics of aged or overaged material would be expected to be similar to that shown in Figure 20.

In any case, the CHT diagram provides a means for estimating the temperature (location) of the high temperature extreme of the softened region of weld HAZs. For most welding processes likely to be used for AerMet® 100 (*e.g.* GTAW, EB), weld HAZ heating rates are above the

200°C/sec heating rate where the transformation mechanism becomes martensitic-like. Thus, because the transformation is athermal, the dilatometry data or CHT diagram can be applied directly to estimate the extent of the softened region. This estimation can be accomplished simply by superimposing the dilatation curve or critical temperatures from the CHT diagram onto the peak temperature versus position curve for a weld HAZ. A plot of this type for the HAZ simulations discussed above (which used a heating rate of approximately 300°C/sec) is shown in Figure 22.

As shown, the data from the CHT diagram provides a reasonable estimate of the high temperature limit of the softened region. However, it should be noted that the dashed line shown on Figure 22 is not intended to imply an estimation of hardness due to the formation of the austenite, but rather that the softening due to overaging is superceded by the formation of austenite. As with the model for overaging, the accuracy of estimates for austenite formation based on the CHT diagram are dependent to some extent on the accuracy to which the thermal cycle is known. In this case, however, the relative insensitivity of the critical temperatures at high heating rates implies that this should not be a major limitation for estimates of weldment behavior.

Figure 22. Combined plot of hardness, overaging model predictions, and austenite formation data for HAZ simulations in aged base plate.

IV. Summary and Conclusions

The database of weldment mechanical properties demonstrated that high strength levels can be achieved in either solution treated or aged base plate with either the cold wire feed GTA or EB welding process. In general, for direct aged (at temperatures near 482°C) welds in either base plate, the overaged region of the HAZ is the strength limiting factor, while the fusion zone controls the impact toughness. The data also indicate that within the range of GTA welding parameters and PWHT conditions evaluated, suitable procedures can be defined which provide useful combinations of strength and toughness. Compared to the GTA welds in the direct aged condition, the EB welds generally have somewhat superior strength, but not necessarily superior toughness. This observation seems to imply that the finer solidification features which are usually associated with EB welding do not play a dominant role in overall mechanical properties. It is believed that the underlying reasons for this are likely associated with the martensitic transformation and the structure that results from it. Thus, the characteristics of the solid state transformation would appear to dominate over the details of the solidification microstructure. It was demonstrated that acceptable welds, some with near base metal properties, could be produced from several different initial heat treatments, and this provides useful flexibility in design and manufacturing. The database also provides a means for defining process parameters and PWHT's to achieve appropriate properties for different applications.

The mechanical properties database indicated that an important region in welds is the softened region which develops in the heat-affected zone (HAZ). Weld thermal cycle simulations, dilatometry, electron microscopy, and austenite fraction measurements indicated that the development of this region is governed by interaction of precipitate overaging and austenite formation. This softened region is bounded at lower peak temperatures by the rate at which precipitate coarsening occurs, and it is bounded at higher temperatures by the formation of transformable austenite (*i.e.* austenite which transforms to martensite on cooling). Therefore, within the welding metallurgy studies, experimental data and models were developed to describe overaging and austenite formation kinetics during thermal cycling. The aging model is based on precipitate strengthening theory and classical coarsening kinetics, and, when combined with the additivity principle, accurately reproduces the softening which occurs during weld HAZ thermal cycles. Austenite formation kinetics were examined by development of continuous heating transformation (CHT) diagrams for heating rates between 0.1 and 2500°C/sec, and indicates that the formation of austenite at high heating rates may occur by a shear transformation. The models and experimental data can be applied to essentially any thermal cycle, and provide a basis for predicting the evolution of microstructure and properties during welding, heat treatment, and other processing.

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Appendix A. Weldment Tensile Data

Table A1. GTA Weldment Tensile Data (Solution Base Plate)

Heat Input	Post-Weld Heat Treatment	Specimen Orientation	Yield (ksi)	Tensile (ksi)	Elongation (%)	Reduction (%)	Ave Yield (ksi)	Ave Tensile (ksi)	Ave Elong (%)	Ave Reduct (%)	Comments
Low	As-Welded	Base Metal	198.1	276	16.1	56.4	201.3	274.5	15.9	55.0	CC, COG
			190.5	276	15.7	55.4					"
			215.4	271.4	15.9	53.2					"
		Transverse	217.7	265.4	13.2	59.2	216.1	265.7	13.0	58.6	CC, EOG
			214.1	266.4	12.8	57.7					"
			216.4	265.3	13.0	59.0					"
		Longitudinal	224.1	279.7	14.7	48.7	227.3	279.4	14.0	48.1	CC, COG
			226.6	279.7	13.3	46.6					"
			231.1	278.7	13.9	48.9					"
Low	-100F/1h+900F/5h	Base Metal	266.9	285.2	12.3	51.0	261.8	291.9	13.5	55.2	CC, COG
			260.8	295.7	14.0	57.7					"
			257.7	294.9	14.3	57.0					"
		Transverse	260.9	290.2	10.3	44.1	256.0	291.9	11.2	45.5	CC, COG
			259.8	297.8	11.8	45.9					"
			247.2	287.7	11.4	46.4					"
		Longitudinal	248.4	286.2	13.7	53.0	255.6	286.0	11.0	37.6	CC, COG
			257.7	287.5	10.6	35.2					asymet CC, pore initiation, COG
			260.6	284.2	8.6	24.5					asymet CC, pore initiation, =COG
Low	400F/1h	Base Metal	203.7	262.8	14.6	55.4	204.2	263.0	14.5	54.0	CC, COG
			204.7	263.3	14.0	52.7					"
			204.2	262.8	14.9	53.8					"
		Transverse	215.4	263.8	12.9	58.0	216.9	263.5	12.8	56.6	CC, EOG
			221.5	263.3	12.1	53.2					"
			213.9	263.3	13.4	58.5					"
		Longitudinal	217.2	280.3	13.5	45.4	221.2	280.7	11.7	34.2	CC, COG
			225.1	281.1	9.9	21.1					asymet CC, pore initiation, COG
				280.8		36.1					asymet CC, pore initiation, COG
High	As-Welded	Base Metal	204.3	257.1	14.7	58.2	203.9	266.8	15.5	57.5	CC, COG
			201.1	271.4	15.6	56.4					"
			206.3	272.0	16.1	58.0					"
		Transverse	212.5	267.0	8.2	59.2	202.6	267.8	6.5	42.3	CC, EOG (mm from FL)
			197.7	268.5	5.8	57.7					"
			197.7	267.9	5.5	10.1					SHEAR, EOG (para to FL?)
		Longitudinal	198.8	284.2	13.8	48.9	206.8	278.0	14.2	50.4	CC, COG
			212.9	270.5	14.4	51.6					"
			208.8	279.2	14.3	50.6					"
High	-100F/1h+900F/5h	Base Metal	260.8	289.8	12.4	53.4	260.3	291.8	13.6	57.1	CC, COG
			258.7	292.3	14.5	59.5					"
			261.3	293.3	14.0	58.5					"
		Transverse	251.0	273.8	4.5	9.4	253.2	273.1	5.1	12.1	SHEAR, EOG (para to FL?)
			253.6	272.2	6.6	17.7					"
			255.1	273.2	4.3	9.3					"
		Longitudinal	262.8	278.1	10.6	42.2	262.8	277.0	11.4	43.8	CC, COG
			262.8	276.0	11.7	48.2					"
			262.7	276.9	12.0	40.9					"
High	400F/1h	Base Metal	209.5	264.9	15.3	57.7	206.5	262.8	15.5	58.0	CC, COG
			208.4	260.3	14.7	57.7					"
			201.7	263.3	16.4	58.5					"
		Transverse	215.9	268.4	5.9	57.5	215.4	266.3	6.4	58.3	CC, EOG
			211.9	266.4	4.7	58.0					"
			218.3	264.2	8.5	59.3					"
		Longitudinal	212.0	278.3	13.2	49.4	218.8	276.0	13.4	50.0	CC, COG
			219.2	276.7	13.3	50.6					"
			225.1	273.0	13.8	49.9					"

CC = cup and cone fracture, COG = failure near center of gage, EOG = edge of gage, SHEAR = shear type failure mode.

Table A2. GTA Weldment Tensile Data (Aged Base Plate)

Heat Input	Post-Weld Heat Treatment	Specimen Orientation	Yield (ksi)	Tensile (ksi)	Elongation (%)	Reduction (%)	Ave Yield (ksi)	Ave Tensile (ksi)	Ave Elong (%)	Ave Reduct (%)	Comments
Low	As-Welded	Base Metal	258.7	294.7	13.7	54.1	256.2	295.0	13.9	55.2	CC, COG
			257.7	294.2	14.5	58.7					"
			252.1	296.2	13.5	52.8					"
		Transverse	229.9	276.2	3.1	8.6	225.3	279.2	6.6	21.5	Asymet CC, pore initiation, COG
			225.8	281.7	5.7	15.5					Asymet CC, COG
			220.3	279.8	11.0	40.5					"
		Longitudinal	224.1	279.7	14.7	48.7	227.3	279.4	14.0	48.1	CC, COG
			226.6	279.7	13.3	46.6					"
			231.1	278.7	13.9	48.9					"
Low	-100F/1h+875F/5h	Base Metal	261.8	284.4	14.9	59.7	256.2	283.6	14.8	60.0	CC, COG
			251.6	282.9	14.3	58.7					"
			255.1	283.4	15.3	61.7					"
		Transverse	248.7	282.0	4.4	7.8	251.4	282.9	7.2	25.4	Asymet CC, COG
			251.6	283.7	2.9	5.5					Pore cluster, CC, COG
			254.0	283.1	14.4	63.0					CC, C-EOG
		Longitudinal	264.9	298.0	4.1	10.2	259.1	295.2	8.3	28.9	Asymet CC, pore initiation, COG
			253.2	292.4	12.4	47.6					CC, COG
Low	400F/5h	Base Metal	254.1	296.2	13.8	55.0	256.0	295.5	13.7	54.3	CC, COG
			250.5	293.7	13.9	55.5					"
			263.4	296.7	13.4	52.3					"
		Transverse	232.5	281.1	11.2	41.2	231.9	279.8	8.0	27.1	Asymet CC, pore initiation, COG
			237.3	279.7	7.6	26.1					CC, COG
			226.0	278.5	5.3	14.0					Pore cluster, CC, COG
		Longitudinal	223.7	286.6	14.2	48.3	224.3	286.2	14.4	50.0	CC, COG
			224.9	286.0	14.4	50.0					"
			224.3	286.0	14.7	51.7					"
High	As-Welded	Base Metal	258.7	296.7	13.4	53.4	254.6	296.2	13.4	53.2	CC, COG
			251.6	295.2	13.6	53.9					"
			253.6	296.7	13.1	52.3					"
		Transverse	219.2	272.2	6.0	7.1	220.5	271.1	5.9	7.4	SHEAR, =COG
			222.1	271.0	6.0	7.1					"
			220.1	270.2	5.8	7.9					"
		Longitudinal	198.8	284.2	13.8	48.9	206.8	278.0	14.2	50.4	CC, COG
			212.9	270.5	14.4	51.6					"
			208.8	279.2	14.3	50.6					"
High	-100F/1h+875F/5h	Base Metal	268.5	285.4	12.8	55.5	262.0	284.9	13.3	55.1	CC, COG
			263.9	284.4	12.8	51.7					"
			253.6	284.9	14.3	58.2					"
		Transverse	252.8	275.2	6.2	21.2	252.2	275.9	5.5	18.8	SHEAR, =COG
			251.4	276.5	5.7	20.5					"
			252.5	276.1	4.6	14.6					"
		Longitudinal	264.4	287.6	12.9	52.3	263.6	284.6	12.9	51.6	CC, COG
			266.0	286.9	12.5	49.5					"
			260.4	279.4	13.3	53.0					"
High	400F/5h	Base Metal	259.3	294.7	14.4	57.7	259.6	294.7	14.2	56.9	CC, COG
			257.2	295.7	13.2	52.8					"
			262.3	293.7	15.1	60.2					"
		Transverse	222.4	273.1	5.1	9.4	222.2	270.1	5.7	11.7	SHEAR, COG
			223.2	267.9	6.2	10.9					"
			220.9	269.3	5.9	14.7					"
		Longitudinal	224.4	278.1	15.1	51.2	225.0	278.5	14.8	51.2	CC, COG
			227.2	278.7	14.6	51.2					"
			223.4	278.6	14.7	51.2					"

CC = cup and cone fracture, COG = failure near center of gage, EOG = edge of gage,
SHEAR = shear type failure mode.

Table A3. EB Weldment Tensile Data

Base Plate Condition	Heat Input	Post-Weld Heat Treatment	Specimen Orientation	Yield (ksi)	Tensile (ksi)	Elongation (%)	Reduction (%)	Ave Yield (ksi)	Ave Tensile (ksi)	Ave Elong (%)	Ave Reduct (%)	Comments
Solution	1-Pass	As-Welded	Base Metal	198.1	276	16.1	56.4	201.3	274.5	15.9	55.0	CC, COG
				190.5	276	15.7	55.4					"
				215.4	271.4	15.9	53.2					"
			Transverse	202.5	263.8	14.3	60.6	195.8	264.3	14.3	60.4	CC, EOG
				186.7	264.8	14.4	60.6					"
				198.1	264.3	14.1	60.1					"
Solution	1-Pass	-100F/1h+900F/5h	Base Metal	266.9	285.2	12.3	51.0	261.8	291.9	13.5	55.2	CC, COG
				260.8	295.7	14.0	57.7					"
				257.7	294.9	14.3	57.0					"
			Transverse	256.2	287.5	12.6	55.5	258.7	288.5	12.3	55.2	Asymet CC, COG
				262.8	287.5	12.1	55.0					"
				257.2	290.6	12.3	55.0					"
Solution	1-Pass	400F/1h	Base Metal	203.7	262.8	14.6	55.4	204.2	263.0	14.5	54.0	CC, COG
				204.7	263.3	14.0	52.7					"
				204.2	262.8	14.9	53.8					"
			Transverse	208.8	262.3	16.0	60.6	204.2	262.7	16.4	60.0	CC, EOG (Double neck)
				203.8	263.4	16.4	58.2					"
				200.1	262.3	16.7	61.2					"
Full Age	1-Pass	As-Welded	Base Metal	258.7	294.7	13.7	54.1	256.2	295.0	13.9	55.2	CC, COG
				257.7	294.2	14.5	58.7					"
				252.1	296.2	13.5	52.8					"
			Transverse	233.7	285.9	12.5	53.6	232.7	289.5	12.4	53.3	CC, =COG (Prob HAZ)
				230.3	291.4	12.7	53.2					"
				234.0	291.1	12.1	53.2					"
Full Age	1-Pass	-100F/1h+875F/5h	Base Metal	261.8	284.4	14.9	59.7	256.2	283.6	14.8	60.0	CC, GOG
				251.6	282.9	14.3	58.7					"
				255.1	283.4	15.3	61.7					"
			Transverse	254.6	290.6	14.6	59.1	254.3	289.3	14.8	60.7	CC, = COG (Prob HAZ)
				252.0	288.0	15.0	61.9					"
				256.4	289.3	14.7	61.1					"
Full Age	1-Pass	400F/1h	Base Metal	254.1	296.2	13.8	55.0	256.0	295.5	13.7	54.3	CC, COG
				250.5	293.7	13.9	55.5					"
				263.4	296.7	13.4	52.3					"
			Transverse	220.4	271.9	9.7	51.7	222.6	270.0	9.3	49.5	CC, COG
				224.1	267.4	9.0	48.1					"
				223.3	270.6	9.2	48.8					"

CC = cup and cone fracture, COG = failure near center of gage, EOG = edge of gage,
SHEAR = shear type failure mode.

Table A4. Tensile Properties for Re-Heat Treated Weldments

Base Plate Condition	Welding Process	Heat Input	Post-Weld Heat Treatment	Specimen Orientation	Yield (ksi)	Tensile (ksi)	Elongation (%)	Reduction (%)	Ave Yield (ksi)	Ave Tensile (ksi)	Ave Elong (%)	Ave Reduct (%)	Comments
Solution	GTAW	Low	100F/1h+900F/5h	Base Metal	262.8	286.2	14.0	58.5	260.2	285.5	14.7	59.1	CC, COG
					259.2	284.4	15.0	59.9					"
					258.7	286.0	15.1	58.9					"
				Transverse	262.8	280.3	2.3	2.4	261.3	280.2	2.1	4.2	Lack of Fusion
					259.8	277.7	2.3	6.3					"
						282.7	1.7	4.0					"
				Longitudinal	250.4	282.5	8.9	29.9	244.4	279.1	6.9	24.2	CC, Pore Initiation, COG
					238.4	279.9	5.9						"
					244.3	274.8	6.0	18.4					CC, Pore Cluster, COG
Solution	EB	1-Pass	100F/1h+900F/5h	Base Metal	262.8	286.2	14.0	58.5	260.2	285.5	14.7	59.1	CC, COG
					259.2	284.4	15.0	59.9					"
					258.7	286.0	15.1	58.9					"
				Transverse		288.5	13.8	58.2	266.5	288.8	14.4	60.8	CC, COG
					263.9	289.5	15.6	65.1					"
					269.0	288.5	13.7	59.2					"

CC = cup and cone fracture, COG = failure near center of gage, EOG = edge of gage,
SHEAR = shear type failure mode.

Appendix B. Weldment Charpy Impact Data

Table B1. GTA Weldment Charpy Impact Data (Solution Base Plate)

Heat Input	Post-Weld Heat Treatment	Notch Location	Impact Energy (ft-lb)	Average Energy (ft-lb)	Comments
Low	As-Welded	Base Metal	42.8	45.1	OK
			46.0		"
			46.6		"
		Centerline	20.0	21.6	OK
			23.9		"
			21.0		"
		Fusion Line	29.8	30.5	OK
			28.9		"
			32.9		"
Low	-100F/1h+900F/5h	Base Metal	36.9	36.3	OK
			33.8		"
			38.2		"
		Centerline	17.5	18.5	OK
			19.1		"
			19.0		"
		Fusion Line	35.8	35.3	OK
			35.2		"
			35.0		"
Low	400F/1h	Base Metal	54.3	53.0	OK
			52.9		"
			51.9		"
		Centerline	16.5	16.4	OK
			15.8		"
			17.0		"
		Fusion Line	28.9	28.5	OK
			27.1		"
			29.4		"
High	As-Welded	Base Metal	50.1	50.5	OK
			48.5		"
			53.0		"
		Centerline	22.0	19.6	OK
			20.9		"
			16.0		"
		Fusion Line	25.1	26.0	OK
			25.5		"
			27.4		"
High	-100F/1h+900F/5h	Base Metal	37.0	36.9	OK
			35.8		"
			37.9		"
		Centerline	15.1	16.8	OK
			14.7		"
			20.7		"
		Fusion Line	37.6	38.8	OK
			38.9		"
			40.0		"
High	400F/1h	Base Metal	55.0	51.3	OK
			50.2		"
			48.8		"
		Centerline	16.3	16.6	OK
			16.5		"
			17.0		"
		Fusion Line	28.0	26.8	OK
			25.3		"
			27.1		"

Table B2. GTA Weldment Charpy Impact Data (Aged Base Plate)

Heat Input	Post-Weld Heat Treatment	Notch Location	Impact Energy (ft-lb)	Average Energy (ft-lb)	Comments
Low	As-Welded	Base Metal	32.5	33.2	OK
			34.2		"
			32.8		"
		Centerline	20.0	21.6	OK
			23.9		"
			21.0		"
		Fusion Line	43.0	43.5	OK
			44.8		"
			42.8		"
Low	100°F/1h+875°F/5h	Base Metal	38.8	35.6	OK
			34.0		"
			34.0		"
		Centerline	16.0	16.4	OK
			16.2		"
			17.0		"
		Fusion Line	34.5	34.3	OK
			33.0		"
			35.5		"
Low	400°F/1h	Base Metal	34.2	33.9	OK
			32.0		"
			35.5		"
		Centerline	20.0	19.3	OK
			18.0		"
			19.9		"
		Fusion Line	35.2	34.2	OK
			34.5		"
			33.0		"
High	As-Welded	Base Metal	31.0	31.5	OK
			30.5		"
			33.0		"
		Centerline	22.0	19.6	OK
			20.9		"
			16.0		"
		Fusion Line	39.4	40.3	OK
			39.6		"
			42.0		"
High	-100°F/1h+875°F/5h	Base Metal	38.8	35.6	OK
			34.0		"
			34.0		"
		Centerline	15.5	15.8	OK
			15.9		"
			16.0		"
		Fusion Line	38.7	37.2	OK
			37.5		"
			35.3		"
High	400°F/1h	Base Metal	28.2	30.6	Small Delam
			32.0		OK
			31.5		Small Delam
		Centerline	19.9	21.0	OK
			21.5		"
			21.7		"
		Fusion Line	47.9	44.5	OK
			45.5		"
			40.0		"

Table B3. EB Weldment Charpy Impact Data

Base Plate Condition	Heat Input	Post-Weld Heat Treatment	Notch Location	Impact Energy (ft-lb)	Average Energy (ft-lb)	Comments
Solution	1-Pass	As-Welded	Base Metal	42.8	45.1	OK
				46.0		"
				46.6		"
			Centerline	24.5	22.1	OK
				19.5		OK, some glossy, small lip
				22.2		"
			Fusion Line	38.8	36.0	OK
				35.0		"
				34.2		"
Solution	1-Pass	-100F/1h+900F/5h	Base Metal	36.9	36.3	OK
				33.8		"
				38.2		"
			Centerline	19.0	17.7	OK, v. small shear lip
				17.2		"
				17.0		OK
			Fusion Line	29.0	29.2	"
				29.0		"
				29.5		"
Solution	1-Pass	400°F/1h	Base Metal	54.3	53.0	OK
				52.9		"
				51.9		"
			Centerline	29.0	31.2	OK
				30.5		"
				34.2		"
			Fusion Line	46.8	45.2	OK
				42.0		"
				46.8		"
Full Age	1-Pass	As-Welded	Base Metal	32.5	33.2	OK
				34.2		"
				32.8		"
			Centerline	20.4	19.7	OK, small shear lip
				18.8		"
				20.0		"
			Fusion Line	33.4	27.3	OK
				29.3		"
				19.1		"
Full Age	1-Pass	-100F/1h+875F/5h	Base Metal	38.8	35.6	OK
				34.0		"
				34.0		"
			Centerline	9.9	11.6	glossy cntline sep, no shear
				12.0		"
				13.0		v. small shear lip
			Fusion Line	26.2	19.5	OK, small shear lip
				16.3		"
				16.0		"
		-100F/1h+875F/5h Retest	Centerline	5.1	10.0	glossy cntline sep, no shear
				14.9		"
		-100F/1h+900F/5h	Centerline	19.0	19.1	OK
				19.1		OK, small amt glossy
Full Age	1-Pass	400°F/1h	Base Metal	34.2	33.9	OK
				32.0		"
				35.5		"
			Centerline	28.0	29.1	OK
				31.3		"
				28.0		"
			Fusion Line	40.9	42.9	OK
				44.2		"
				43.5		"

Table B4. Charpy Impact Data for Re-Heat Treated Weldments

Weld Designation	Base Plate Condition	Welding Process	Heat Input	Post-Weld Heat Treatment	Notch Location	Impact Energy (ft-lb)	Average Energy (ft-lb)	Comments
GSL-SA	Soution	GTAW	Low	Solution Treat Plus -100F/1h+900F/5h	Base Metal	39.9	39.9	OK
						40.0		"
						39.7		"
					Centerline	16.5	15.9	Ok
						14.9		"
						16.2		"
					Fusion Line	25.0	17.3	Lack of fusion
						13.0		"
						14.0		"
ESL-SA	Soution	EB	1-Pass	Solution Treat Plus -100F/1h+900F/5h	Base Metal	39.9	39.9	OK
						40.0		"
						39.7		"
					Centerline	25.0	25.0	OK
						26.0		"
						24.0		"
					Fusion Line	35.5	35.5	OK
						35.5		"

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