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**HEAVY-SECTION STEEL TECHNOLOGY PROGRAM
SEMI-ANNUAL PROGRESS REPORT
FOR PERIOD ENDING FEBRUARY 28, 1971**

OAK RIDGE NATIONAL LABORATORY
operated by
UNION CARBIDE CORPORATION
for the
U.S. ATOMIC ENERGY COMMISSION

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National Technical Information Service
U.S. Department of Commerce
3285 Port Royal Road, Springfield, Virginia 22151
Price: Printed Copy \$3.68; Microfilm \$0.96

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Contract No. W-7405-eng-26

**HEAVY-SECTION STEEL TECHNOLOGY PROGRAM
SEMIANNUAL PROGRESS REPORT
For Period Ending February 28, 1971**

F. J. Witt, Program Director

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DECEMBER 1971

**OAK RIDGE NATIONAL LABORATORY
Oak Ridge, Tennessee 37830
operated by
UNION CARBIDE CORPORATION
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Foreword

The Heavy-Section Steel Technology (HSST) Program is a USAEC-sponsored effort for investigating the effects of flaws, variations of properties, stress raisers, and residual stress on the structural reliability of present and contemplated water-cooled reactor pressure vessels. The cognizant engineer for the USAEC is J. R. Hunter. At ORNL, the program is under the Pressure

Vessel Technology Program, of which G. D. Whitman is Director. The HSST program is being carried out in very close cooperation with the nuclear power industry. Prior reports in this series are ORNL-4176, ORNL-4315, ORNL-4377, ORNL-4463, ORNL-4512, ORNL-4590, and ORNL-4653.

Summary

The Heavy-Section Steel Technology (HSST) program is one of the major USAEC safety-oriented research efforts concerned with the in-service structural integrity of the massive pressure vessels typical of those used in large boiling-water and pressurized-water reactor power plants. Emphasis is placed on the effects of flaws, material inhomogeneities, and discontinuities on the behavior of the vessels under startup, operating, cool-down, and accident conditions.

Research and development under the HSST program is of necessity broad based because of the extensive interrelating fields of study involved: metallurgy, chemistry, material properties, inspection, analytical and experimental stress and strain analyses, environmental effects, fracture mechanics, and the general area of fracture behavior. The material being investigated is mainly the ASTM A 533, grade B, class 1 steel and the submerged-arc weld. Less emphasis is placed on the forging-grade reactor vessel steel, ASTM A 508, class 2, and electroslog and shielded-metal-arc welds.

The HSST program is coordinated with efforts by other government agencies and the manufacturing and utility sectors of the nuclear power industry both in the United States and abroad. These total efforts should culminate in the quantification of safety assessments that are needed by the USAEC regulatory bodies, the professional code-writing bodies, and the nuclear power industry.

There are 12 tasks of the HSST program. In this report, the activities under these tasks are grouped under administration and procurement, unirradiated materials investigations, irradiated materials studies, and pressure vessel investigations. In addition to the investigations performed at Oak Ridge National Laboratory, work is being performed under nine subcontracts and three cooperative efforts. One progress report, two technical reports, and two technical manuscripts were distributed during the reporting period.

Extensive investigations of tensile and impact properties throughout the length, width, and thickness of the 55-ton HSST plates continue. Surface material was seen to vary by 19,000 psi in yield strength; however, at quarter thickness much less variation was noted. Thirty ft-lb Charpy impact temperature varies as much as 30°F at quarter thickness for the longitudinal direction.

Under an informal cooperative effort, Siemens AG has developed extensive fracture toughness data on 22 Ni Mo Cr 37 forging steel, which is very similar to ASTM A 508, class 2 forging steel. Tensile, Charpy V-notch impact, drop weight, and fracture toughness tests were made. Also, a crack growth fatigue study was made. Three different heats were used in the investigation. The forging steel, in comparable thickness with HSST 12-in.-thick plates, exhibited properties as good as those of the plate steel. Two of the three forging steels were shown to have a lower transition temperature in fracture toughness than the plate steel. Almost identical crack growth rates were observed between the forging and plate steels.

The main emphasis of the investigation of gross strain crack tolerance of steels has been a study of the effects of crack size on the gross strain crack tolerance as a function of temperature. Three crack sizes were studied. A significant increase to strain above 0.2% occurred at increasing temperatures with increasing flaw size. This was true for gross section strain, net section strain, and crack opening displacement.

The investigation of crack arrest toughness has been essentially completed, and it was found that the arrest toughness is lower than dynamic initiation toughness, which is lower than static initiation toughness. In particular, the arrest toughness has not exhibited an abrupt transition to higher toughness levels even at 120°F, which to date is the highest temperature at which tests have been made.

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Three-dimensional elastic-plastic analyses using a finite-element computer program have been performed on a tensile specimen containing a part-through semi-elliptic crack. The first case analyzed was that of an elastic, perfectly plastic material. The case currently being run is that of a strain-hardening material. The results of this analysis can be compared with the flawed tensile test results.

Compact tension specimens were tested from -50 to 550°F . One-inch (1 T) specimens have now been tested with both longitudinal and transverse orientations. From these results the fracture toughness parameter K_{Icd} was determined. This work was performed as a preliminary step in testing specimens 6 and 9 in. thick.

The investigation of crack propagation in water reactor environment, excluding irradiation, has been completed. It was concluded that environment had no effect on fatigue crack growth of A 533, grade B, class 1 steel and weldment. The data compared very well with those obtained on other heats of the steel at 75°F .

The investigations of irradiation effects at Naval Research Laboratory, Westinghouse Electric Corporation, and Oak Ridge National Laboratory are essentially complete. A major effort at Hanford Engineering Development Laboratory (HEDL) and a cooperative effort with Verein Deutscher Eisenhüttenleute continue.

At HEDL a 4-in.-thick compact tension dummy specimen gamma-heat experiment was designed to determine specimen temperature profiles in a reflector position facility of the Experimental Test Reactor. Some 50 thermocouples will be used in the experiment.

Compact tension specimens irradiated to 8×10^{19} neutrons/cm² ($E > 1$ MeV) at 540°F were tested. A fracture toughness K_{Ic} value of 46,200 psi $\sqrt{\text{in.}}$ was found at 200°F , and a value of 63,300 psi $\sqrt{\text{in.}}$ was found at 245°F . The low values imply that irradiation has a progressive embrittling effect on steel which does not saturate at a given fluence level. A series of high-fluence tensile specimens were also tested. These tests indicated that changes in yield and ultimate strength are directly proportional to fluence and further verified that saturation does not take place.

The work sponsored by Verein Deutscher Eisenhüttenleute features Charpy impact and tensile tests. The materials being investigated are three low-alloy steels, including A 533 steel furnished by the HSST program. Submerged-arc welds and heat-affected zones are also being investigated. Irradiations are being performed in

the DIDO reactor. Fluences are 1, 5, and 10×10^{19} ($E > 1$ MeV) at 572 and 752°F . The irradiations have been completed, and results should be available in late 1971.

A technique for fatigue sharpening large cracks by local pressurization was applied to six large 6-in.-thick tensile specimens. The fatiguing of machined flaws $1/4$ to $1/2$ in. deep was successfully accomplished on all the specimens.

A technique for generating small sharp flaws using a method of hydrogen charging in a high residual stress region was developed. The high residual stress region was made by use of an electron beam welder to produce an autogeneous weld on the surface of a specimen. A 10% solution of hydrogen sulfate serves as the electrolyte in charging the hydrogen into the steel. Flaws in a series of small pressure vessels were generated in this manner.

Ten large flawed tensile specimens (cross section 18 in. wide by 6 in. thick) have now been tested. Seven were tested in the longitudinal direction and three in the transverse direction. Extensive experimental data have been obtained from all the tests. Fairly low gross section strains (around 1%) were obtained up to 100°F with accompanying flat fractures. Mixed mode fractures with gross section strains around 3% were found at 130°F , thus indicating a very rapid toughness transition behavior. The longitudinal specimens indicated a transition to higher toughness levels than the transverse specimens. All the specimens were instrumented with acoustic emission devices. Acoustic emissions were studied as a function of load or stress intensity factor. The cumulative emissions were shown to be an exponential function of the stress intensity. The exponent appeared to vary with test temperature.

The fabrication of the intermediate test vessels continues, with delivery of the first vessel scheduled for mid-1971. High strength values were found to exist for the cylindrical test courses of the vessels. Following an intensive metallurgical and testing study, techniques for reducing the strength to levels comparable with A 533, grade B, class 1 plate were developed. These procedures are being applied to the test courses.

A study of the relationship between fracture toughness K_{Ic} and Charpy impact data was completed. Although a correlation of K_{Ic} with Charpy energy appeared to exist for several situations, no general conclusion could be drawn that confidence existed in such correlation throughout the full toughness range.

1. Program Administration and Procurement

With the emergence of nuclear reactors as one of the major heat resources, the United States electrical power industry has been quick to recognize the potential of nuclear reactor systems for power generation. The peak period for nuclear plant orders was in 1967, when two-thirds of the announced power generating plants were nuclear. It is now conservatively estimated that by 1980 about 25% of the total electrical output in the United States will be from nuclear plants. In the United States there are currently over 100 civilian nuclear power stations in operation (approximately 20), under construction (approximately 55), and on order (approximately 35). The total electrical capacity in service exceeds 8300 MW(e). Some of the larger plants now under construction have capacities exceeding 1100 MW(e). The two major systems that have evolved to commercial status in the United States are the boiling-water reactor (BWR) and the pressurized-water reactor (PWR) systems.

Concurrent with the growth of the nuclear power industry has been the increase in size of the plants themselves. This increase has resulted mainly from scaling up small systems with some modification in pressures and operating temperatures; consequently the pressure vessels of the systems have increased in diameter and thickness. The plate for fabricating some larger pressurized-water vessels approaches 12 in. in thickness, with diameters of the vessels from 14 to 16 ft. The standard operating temperature is around 550°F. Pressures range upward to around 2500 psi for the PWR systems; for BWR systems the vessels are designed for about half that pressure. Equally significant to the increased sizes is the economic incentive to place these large plants near densely populated and industrial areas.

The main concern of the Heavy-Section Steel Technology (HSST) Program is the massive pressure vessels,

with emphasis being placed on the effects of flaws, material inhomogeneities, and discontinuities on the behavior of the vessels under startup, operating, cool-down, and accident conditions. This interest stems from the fact that, inherently, flaws (discontinuities), though possibly quite small, exist in vessels and that better in-service inspection techniques must be developed to detect these for the service life of the plants, some being designed to operate 40 years. The objective of the HSST program is to develop a technology for reliably estimating a conservative margin of safety against fracture of nuclear pressure vessels during the service life of the plants, particularly if such fractures might endanger the public.

The material used to fabricate most American water reactor vessels is ASTM A 533, grade B, class 1 plate steel or the equivalent forging-grade steel, ASTM A 508, class 2.¹ The welds most commonly used are the submerged-arc or the shielded-metal-arc weld, but electroslag welds have been used in some vessels. All these steels and weldments are being investigated, but the major emphases are on the plate steel and the submerged-arc welds.

The HSST program has a professional staff charged with administering, evaluating, and interpreting the research programs and results. There are 2 advisory committees staffed by 28 scientists and engineers with expertise in the areas of HSST research. Only 8 of these experts are under subcontract. Nine technical subcontracts (Table 1.1) were in force during this reporting period. The work under these subcontracts is summarized elsewhere in this report. Also during this period

1. 1971 *Annual Book of ASTM Standards*, Part 4, pp. 594-604, 646-51, American Society for Testing and Materials, April 1971.

Table 1.1. Research subcontracts sponsored by the HSST program

UCCND Subcontract No.	Title	Company	Initiation date	Termination date
3078	Testing of Heavy-Section Compact K_{Ic} Specimens	Westinghouse Electric Corporation	Jan. 1, 1968	June 30, 1970
3134	Gross Strain Measurement of Fracture Toughness of Steel	TRW, Inc.	Sept. 1, 1968	Apr. 30, 1972
3152	Strain Rate and Crack Arrest Studies	Materials Research Laboratory	Sept. 1, 1968	Sept. 30, 1971
3153	Three-Dimensional Elastic- Plastic Stress and Strain Analysis for Fracture Mechanics	Brown University	Oct. 1, 1968	Mar. 31, 1972
3196	Elastic-Plastic Fracture Criterion for Flawed Structures	Westinghouse Electric Corporation	Jan. 1, 1969	Dec. 31, 1971
3202	Large Tensile Tests	Southwest Research Institute	Jan. 1, 1969	Oct. 31, 1971
3209	Irradiation Effects on Pressure Vessel Materials	Westinghouse Electric Corporation	Dec. 23, 1968	Sept. 30, 1970
3290	Crack Growth Rate in an Environment of High- Temperature Reactor Water	Westinghouse Electric Corporation	June 1, 1969	Apr. 30, 1971
3398	Correlation of Impact and Toughness Data	University of Illinois	Feb. 1, 1970	Nov. 15, 1970

the sixth progress report,² two technical reports,^{3,4} and two technical manuscripts^{5,6} were published. Twelve papers on HSST research were reviewed for a special issue of *Nuclear Engineering and Design*.

The major procurement items are the intermediate test vessels. These are discussed in Chapter 4.

2. F. J. Witt, *HSST Program Semiannual Progr. Rep. Aug. 31, 1967*, ORNL-4167; *Feb. 29, 1968*, ORNL-4315; *Aug. 31, 1968*, ORNL-4377; *Feb. 28, 1969*, ORNL-4463; *Aug. 31, 1969*, ORNL-4512; *Feb. 28, 1970*, ORNL-4590.

3. T. R. Mager, *Fracture Toughness Characterization Study of A 533 Grade B Class 1 Steel*, WCAP-7538 (September 1970).

4. T. R. Mager, *Notch Preparation in Compact Tension Specimens*, WCAP-7579 (September 1970).

5. D. A. Canonico, *Transition Temperature Considerations for Thick-Wall Nuclear Pressure Vessels*, ORNL-TM-3114 (October 1970).

6. G. D. Whitman and F. J. Witt, *HSST Program*, ORNL-TM-3055 (October 1970).

2. Investigation of Unirradiated Materials

These investigations are divided into studies of unirradiated materials and studies of irradiation effects. The studies of unirradiated materials include inspection, characterization, variability determinations, transition-temperature investigations, fracture mechanics studies, and fatigue-crack-propagation tests. Current work in these areas is discussed here. Additional results for unirradiated material are given in Chapter 3, 'Investigations of Irradiated Materials.

MATERIALS INSPECTION

C. E. Childress

The inspection activities of the HSST program are normally reported in this section; however, the status of the fabrication of the intermediate test vessels is discussed in Chapter 4.

MATERIALS CHARACTERIZATION AND VARIABILITY STUDIES

R. G. Berggren W. J. Stelzman
T. N. Jones

Tensile test results have been obtained from section 01C1 of HSST plate 01 (ASTM A 533, grade B, class 1 steel). The plate section locations are shown in Fig. 2.1, and the results are reported in Table 2.1. Table 2.2 compares the yield strengths of the 01C1 region with the portions of plate 01 previously investigated. The comparison results were obtained using the standard 0.505- and the 0.176-in.-gage-diam tensile specimens; however, test data indicate that either specimen produces comparable yield results.

The 01C1 region yield-strength data show the same insensitivity to specimen orientation in the longitudinal (R) or transverse (W) directions¹ as in previous plate locations except for the back surface (1 T) level, which shows a significant difference. After reexamining the

data for the 1 T level, it is believed that additional data from R-oriented specimens will reduce this difference. The 01C1 region exhibits lower yield strengths than the nearby 01E1 region or the central 01K regions at both the front (0 T) and back (1 T) surfaces and the one-fourth and three-fourths thickness levels but equal yield strengths at midlevel. In contrast, the 01C1 yield strengths are equivalent to the 01MU region yield

Fig. 2.1. Test locations in HSST plate 01.

1. R. G. Berggren, *HSST Program Semiannu. Progr. Rep. Feb. 28, 1969*, ORNL-4463, pp. 20-21.

Table 2.1. Tensile properties of 12-in.-thick HSST plate section 01C1,^a
ASTM A 533, grade B, class 1 steel

Depth in plate (d/12)	Orientation ^b	Strength property (psi)			Elongation (%)		Reduction in area (%)
		Upper yield	Lower yield	Ultimate ^c tensile	Uniform	Total ^d	
		$\times 10^3$	$\times 10^3$	$\times 10^3$			
0	R		73.0	90.3	10.6	18.5	75.0
	W	76.9	74.6	91.5	13.1	26.4	71.8
$\frac{1}{4}$	R	66.3	64.1	85.7	10.7	20.1	72.4
	W	66.9	65.1	86.5	10.8	19.4	70.2
$\frac{1}{2}$	R	69.5	66.4	88.4	10.6	19.7	70.7
	W	67.6	67.0	88.8	10.3	18.7	69.9
$\frac{3}{4}$	R	64.3	62.4	83.9	11.8	20.7	72.2
	W	64.1	63.4	85.0	11.6	20.2	70.7
1	R	71.6	67.4	87.0	13.3	23.6	75.0
	W		71.0	90.1	11.5	20.3	73.8

^aAverage of two or more 0.178-in.-gage-diam specimens tested at 75°F and a strain rate of 0.016 in. in.⁻¹ min.⁻¹.

^bSpecimen axial dimension in R, major rolling direction, or W, transverse to major rolling direction, of plate.

^cDefined as the maximum load after the yield point divided by the original area of cross section.

^dLength-to-diameter ratio: 7.

Table 2.3. Variation of Charpy V-notch impact properties
of 12-in.-thick HSST plate 01, ASTM A-533,
grade B, class 1 steel

Table 2.2. Variation of yield stress^a in 12-in.-thick HSST
plate 01, ASTM A 533, grade B, class 1 steel

Plate section ^b	Orientation ^c	Depth in plate (psi)				
		0T	1T	$\frac{1}{4}T$	$\frac{3}{4}T$	$\frac{1}{2}T$
		$\times 10^3$	$\times 10^3$	$\times 10^3$	$\times 10^3$	$\times 10^3$
C1	R	73.0	67.4	64.1	62.4	66.4
C1	W	74.6	71.0	65.1	63.4	67.0
E1	R	80.7 ^d		63.9 ^d		
K1-1	R	82.5				
K1-1	W	79.1 ^d		68.5 ^d		67.1 ^d
K1-1	W					67.5
K1-3	R	79.2 ^d	75.0 ^d	68.8 ^d	66.3 ^d	66.7 ^d
K1-3	T				66.4	66.7
K3	R	79.0	77.7	68.5		
K3	W	79.4	78.1	69.0		
K3	W	77.4 ^d	76.0 ^d	69.3 ^d		
K3	T			68.1		
K4	R	81.2		68.7		68.1
MU1	R	73.7 ^d		61.7 ^d		58.4 ^d
MU1	W	73.1 ^d		61.6 ^d		58.0 ^d
MU2	R	72.5 ^d		63.2 ^d		
SH	R	56.3	59.3 ^d	64.6 ^d		

^aTest temperature, 75°F.

^bRefers to plate section and location in Fig. 2.1.

^cSpecimen axial dimension in direction of (R) or transverse (W) to major rolling direction of plate.

^dResults from standard 0.505-in.-gage-diam specimens; remainder of results from 0.178-in.-gage-diam miniature specimens.

Plate section ^a	Orientation ^b	Depth in plate				
		0T	1T	$\frac{1}{4}T$	$\frac{3}{4}T$	$\frac{1}{2}T$
Transition Temperature at 30 ft-lb (°F)						
C1	RW	-130	-125	5	5	25
C1	WR	-105	-100	15	25	40
E1	RW	-140		0		
K1-1	WR			20		50
K3	RW	-140	-135	-5		
K3	WR	-105	-120	20		
K4	RW	-140		-5		15
MU1	RW	-120		25		30
MU1	WR	-85		25		40
MU2	RW	-100		25		
SH	RW		-115	10		
Shelf Energy (ft-lb)						
C1	RW	122	125	124	125	108
C1	WR	100	103	110	100	92
E1	RW	116		125		
K1-1	WR			96		85
K3	RW	114	110	120		
K3	WR	86	94	96		
K4	RW			120		116
MU1	RW	125		136		129
MU1	WR	93		110		100
MU2	RW	125		125		
SH	RW		125	124		

^aRefers to plate section and location in Fig. 2.1.

^bFirst letter indicates orientation of axial center line of specimen and second letter indicates fracture propagation direction. R and W refer to principal rolling direction and width direction of plate respectively.

strengths at both surfaces and at the one-fourth and three-fourths thickness levels but higher at the mid-thickness. The OISH region yield strengths still remain well below the values obtained elsewhere in the plate.

Charpy V-notch impact data were also obtained from section OIC1; the results are presented in Table 2.3. At a correlation energy of 30 ft-lb the OIC1 impact data indicate a lower transition temperature and higher shelf energy from longitudinal (RW) than from transverse (WR) oriented test specimens.¹ This appears to be consistent with results from other portions of the plate. The OIC1 impact results are in good agreement with results for the other regions of the plate.

Fracture Toughness Tests² on Forgings from 22 Ni Mo Cr 37

E. Klausnitzer
Siemens Aktiengesellschaft
West Germany

Introduction

Since 1967 Siemens AG has been working on a research program entitled, "Fracture Toughness Testing on Pressure Vessel Steels and Application of the Results on Reactor Pressure Vessels." This program is funded by the government of the Federal Republic of Germany. We are investigating the theoretical and experimental basic theories of fracture mechanics and their application to pressurized-water reactors (PWR); therefore the measurement of the fracture toughness of the steel, 22 Ni Mo Cr 37, is very important. This steel is used by Siemens AG in the form of forgings with wall thicknesses from 4 to 40 in., especially for reactor pressure vessels and other primary components of nuclear reactors. The tests have been done on several forgings, some of which were embrittled, with various fabrication histories.

Some of the results on unirradiated specimens are given in this discussion; no data on irradiated specimens are available. The irradiated specimens were taken out of the KWO (nuclear power plant Obrigheim) in August 1970 and will be tested in 1971.

Test Program

According to the German testing standards for pressure vessels it has to be shown for each forging that the

specified values for chemical composition and mechanical properties, especially the tensile and impact values, are reached. Since it is important to guard against brittle fracture in nuclear reactor pressure vessels, Siemens AG also specified, for its first PWR (MZFR, built 1961-65), the determination of the NDT temperature. Pilot testing of the fracture toughness was done first, and systematical investigations on fracture mechanics have been under way since 1966.

To satisfy the applicable testing standards and to evaluate the safety against brittle fracture by use of fracture mechanics, the test material was subjected to the following tests:

1. tensile tests, to obtain the yield strength, ultimate tensile strength, total elongation, and reduction in area;
2. Charpy V-notch tests for the notch impact strength and the percentage of brittle fracture;
3. drop-weight tests for the NDT temperature;
4. measurements of the crack growth under fatigue load;
5. measurements of the fracture toughness.

Test Materials

For these tests the following forgings from 22 Ni Mo Cr 37 were used.

Material A. A 110-in.-diam by 36-in.-thick forging plate originally intended for use as the tube plate of a steam generator and subsequently rejected after cracks appeared after the annealing was available. Test material from this forging plate was obtained from the supplier.

Material B. For further investigations there was available a 28-in.-diam by 28-in.-thick forging bar fabricated especially for brittle fracture testing.

Material C. Enough testing material for this program was available following acceptance testing of a 139-in.-OD, 9-in. wall thickness, 119-in. length cylindrical forged ring for a nuclear reactor pressure vessel. Specimens from this forging ring will also be tested in the irradiated condition.

The chemical composition of the forgings is generally within the specification of the 22 Ni Mo Cr 37, as shown in Table 2.4. The slightly excessive carbon content of material C is considered insignificant. Apart from slight deviations the chemical composition required by the 22 Ni Mo Cr 37 specification corresponds also with that required by ASTM A 508, class 2.

2. This work was sponsored by Siemens Aktiengesellschaft (Siemens AG) of the Federal Republic of Germany and is reported as a cooperative exchange with the HSST program.

Table 2.4. Chemical composition and heat treatment of 22 Ni Mo Cr 37 forgings

Identification	Chemical composition (wt %)									Heat treatment		
	C	Mn	P	S	Si	Al	Ni	Cr	Mo	Austenitized	Tempered	Stress relieved
22 Ni Mo Cr 37 specification	0.17 0.25	0.50 1.00	Max 0.015	Max 0.015	0.05 0.35	Max 0.05	0.50 1.00	0.30 0.50	0.60 0.80			
22 Ni Mo Cr 37 forging plate A (110 in. diam, 36 in. thick)	0.24	0.91	0.013	0.013	0.23	0.03	0.78	0.33	0.67	1740°F for 16 hr, water quench	1250°F for 20 hr, furnace cool	
22 Ni Mo Cr 37 forging bar B (28 in. diam, 28 in. length)	0.25	0.66	0.013	0.011	0.26	0.02	0.96	0.44	0.59	1650°F for 6 hr, water quench	1290°F for 10 hr, air cool	1150°F for 30 hr, air cool
22 Ni Mo Cr 37 forging ring C (139 in. diam, 9 in. thick, 119 in. length)	0.27	0.76	0.015	0.011	0.30	0.045	0.83	0.35	0.64	1600°F, for 6 hr, water quench	1270°F for 15 hr, furnace cool	1110°F for 5 hr, furnace cool during fabrication of the vessel 6 times at 1000–1110°F for 6–21 hr, furnace cool

All testing materials were melted in an electric furnace and vacuum degassed. The hot working by forging was done at 2100 and 1560°F.

The heat treatment of the test material is also given in Table 2.4. As already mentioned, material A is a rejected piece that was not further stress relieved. The several stress-relief procedures for material C result from conformance to the manufacturing procedure required for the reactor pressure vessel. Typical grain structures of each of the test materials are shown in Fig. 2.2. The grain size of material A is somewhat greater than that of materials B and C. All materials have normal annealing grain structure.

Performance of the tests and specimen orientation

The following specimens were used for the tests:

1. tensile specimens, B 6 X 30 DIN 50 125 (i.e., 6 mm diam, 30 mm testing length);
2. Charpy V-notch test specimens according to ASTM E 23-60, type A;
3. P2 drop-weight specimens according to ASTM E 208-69;
4. WOL-1X, WOL-2X, WOL-4X specimens and 1 T-CTS, 2 T-CTS, and 3 T-CTS specimens according to ASTM E 399-70 T.

All specimens were tested according to the relevant German (DIN) or American (ASTM) regulations. The test temperature was 75°F, and the frequency was maintained between 10 and 100 Hz. The K_{Ic} measurements were done in a cooling bath (nitrogen or methanol).

Material A. The inset in Fig. 2.3 shows a part of the broken forging from which five rods were taken out for investigations by trepanning. Rod 5 was available for the tests at Siemens AG. This rod was divided equally into six approximately 6-in.-thick disks. The initial tests indicated that the Charpy V values were the lowest in the midthickness of the forging. This region was selected for the fracture toughness testing to measure the lower limit of the fracture toughness. The specimens were oriented to reduce the possibility of a gradient in the properties in the fracture area.

Material B. Specimen layout from the segments available from material B is shown in Fig. 2.4. The specimens were oriented so that the fracture planes are radially positioned. The forging direction could not be taken into account. In principle, the distribution of the specimens is the same for all three segments; a detailed view is given in Fig. 2.5.

Material C. All materials testing on forged rings for reactor pressure vessels is done on a ring about 8 in. high which is cut off after annealing. Figure 2.6 shows a

Fig. 2.2. Metallographic examination of 22 Ni Mo Cr 37 forgings. Top to bottom: material A, 36 in. thick; material B, 28 in. thick, material C, 9 in. thick. 200X

forged ring for the KWO reactor pressure vessel, a 40-in.-long segment of which was available for the examinations described here. According to the test standards, the specimens were mainly taken out of the quarter thickness. The fracture planes of the specimens are perpendicular to the maximum loading direction of the reactor pressure vessel.

Test results

A close examination of the distribution of the test specimens shown in Figs. 2.3 to 2.6 reveals that the tests cannot give a complete evaluation of the variability of the properties of the forgings. For this, a much greater volume of material would be necessary. Nevertheless, the results are very valuable because they are the first detailed measurements of the fracture toughness of forgings of German fabrication.

The test results are shown in Figs. 2.7 to 2.13. The graphs contain the single values and the best-fit curves for the tests from materials A, B, and C.

Material A. All test results were obtained from midthickness specimens and represent therefore the worst mechanical properties of the examined forging. Figure 2.7 shows the results from the Charpy V-notch tests at temperatures from -58 up to $+212^{\circ}\text{F}$ and also the results of the drop-weight tests. In addition, the percentage of brittle fracture was determined from the Charpy V-notch tests. The results from the tensile tests and the fracture toughness tests at temperatures between -321 to -13°F are given in Fig. 2.8, which also shows the yield strength obtained from the tensile tests from materials B and C for calibration of the K_{Ic} values.

Material B. As previously reported at the Fourth Annual Information Meeting of the Heavy-Section Steel Technology Program,³ test results show significant variations in the properties of heavy forgings made of 22 Ni Mo Cr 37. It is apparent that the Charpy V-notch values vary considerably over the wall thickness. Similar results have been found for material B, as shown in Fig. 2.9. There is a significant gradient in the axial direction (segment A-B-C) and also in the radial direction (surface and center).

The same tendency is observed for the percentage of brittle fracture. The available test material was insufficient to determine the NDT temperature as a function

3. H. Cerjak and W. Debray, "Properties of Nuclear Grade Forging Steel," paper 7, presented at the Fourth Annual Information Meeting of the Heavy Section Steel Technology Program, Mar. 31 and Apr. 1, 1970.

Fig. 2.3. Specimen orientation for a 36-in.-thick forging plate.

Fig. 2.4. Specimen orientation for a 28-in.-thick forging bar.

Fig. 2.5. Specimen orientation for a 28-in.-thick forging bar.

Fig. 2.6. Specimen orientation for a 9-in.-thick forging ring.

of the specimen orientation. The value given in Fig. 2.9 applies to the surface of segment B.

Yield strength and fracture toughness values for the temperature range -321 to -31°F are shown in Fig. 2.10. The yield strength has a scatter less than $\pm 10\%$ to the best-fit curve for specimens taken throughout the whole forging. The dependence of the fracture toughness on wall thickness cannot yet be determined systematically due to lack of material, but the tendency is clearly evident from the results of the WOL-4X specimen tested at -31°F . The two lower values apply to segment B and the two upper values to segment C. The general trend corresponds qualitatively with that from the Charpy values.

Material C. Material C comes from a relative thin walled forging. It is to be expected, therefore, that the

properties will not change considerably over the wall thickness of the tested region. Extensive tensile and Charpy V-notch tests, which are not herein detailed, indicate no difference between quarter-thickness and midthickness locations. The results from Charpy V-notch tests shown in Fig. 2.11 for the temperature range between -130 and $+194^{\circ}\text{F}$ are typical, therefore, for the main region of the wall thickness.

Figure 2.12 demonstrates the temperature dependence of the fracture toughness for the range -321 to -58°F . Tests with WOL-4X specimens at -31°F were invalid because the following condition was not met:

$$a, B \geq 2.5 \left(\frac{K_{Ic}}{\sigma_{0.2}} \right)^2$$

Fig. 2.7. Charpy V-notch test results, brittle fracture, and NDT temperature for a 36-in.-thick (material A) forging plate.

Fig. 2.8. Temperature dependence of K_{IC} fracture toughness and yield strength for a 36-in.-thick (material A) forging plate.

Fig. 2.10. Temperature dependence of K_{Ic} fracture toughness and yield strength for a 28-in.-thick (material B) forging bar.

Fig. 2.9. Charpy V-notch test results and NDT temperature for a 28-in.-thick (material B) forging bar.

Fig. 2.11. Charpy V-notch test results, brittle fracture, and NDT temperature for a 9-in.-thick (material C) forging ring.

Fig. 2.12. Temperature dependence of K_{Ic} fracture toughness and yield strength for a 9-in.-thick (material C) forging ring.

Measurements of the crack growth during cycling have been made by use of ultrasonic testing for part of the WOL-2X and WOL-4X specimens shown in Fig. 2.12. The results are shown in Fig. 2.13. Note that the resulting scatter band corresponds to that found by Clark and Wessel.⁴ Since the specimens were subsequently used also for the measurement of the fracture toughness, only low ΔK values during cycling were investigated according to ASTM E 399.

Discussion of results

As already mentioned the reported values are the first results from extensive tests on fracture toughness of German forgings conforming to 22 Ni Mo Cr 37, which is used for nuclear reactor pressure vessels.

4. W. G. Clark, Jr., and E. T. Wessel, *Application of Fracture Mechanics Technology to Medium-Strength Steels*, Scientific Paper 69-1D7-BFPWR-P2, Westinghouse Research Laboratories.

It is reasonable to compare the measured values with those from the American plate steel ASTM A 533, grade B, class 1 because this steel has similar properties and is also used in nuclear reactor pressure vessels. For this comparison the extensive results from HSST plate 02 from the HSST program,⁴⁻⁶ material D, are used.

Figure 2.14 is a montage of the results of the Charpy V-notch tests from these materials. It can be seen that there is very great scatter in the Charpy V-notch values.

5. E. T. Wessel, *Linear Elastic Fracture Mechanics for Thick-Walled, Welded Steel Pressure Vessels: Material Property Considerations*, Scientific Paper 69-1D7-BFPWR-P1, Westinghouse Research Laboratories, Feb. 24, 1967.

6. E. T. Wessel and T. R. Mager, "The Fracture Mechanics Approach to Reliability in Nuclear Pressure Vessels," ed. by M. S. Wechsler, *The Technology of Pressure-Retaining Steel Components*, pp. 119-43, The Metallurgical Society of the American Institute of Mining, Metallurgical and Petroleum Engineers, Inc., New York, July 1970.

Fig. 2.13. Fatigue crack growth rate for material C as a function of ΔK for a 9-in.-thick 22 Ni Mo Cr 37 forging ring (points) compared with a 12-in.-thick A 533, grade B, class 1 steel plate (upper and lower scatter band). (Ref. 3, curve 587406-A.)

Fig. 2.14. Comparison of Charpy V-notch test results and brittle fracture for 22 Ni Mo Cr 37 forgings and A 533, grade B, class 1 steel. 12-in. HSST plate 2. (Ref. 4, curve 588559-A.)

Some of the scatter for the forging may result from the variation in forging work or variation in the cooling rate during heat treatment. The results from material B and previously reported HSST program results infer that the cooling rate has the higher influence in the region of the transition temperature. The values at the surface are independent of the wall thickness, generally falling close to the curve for material C.

The values for reference material D correspond very well with the values for material C. The percentage brittle fracture and the NDT temperature for the two materials are also reasonably close. Since material A is rejected forging having quality deficiencies, a valid comparison cannot be made with the other materials. Only results from material C are available for comparison with the crack growth data obtained on the HSST program. The coincidence in values with reference

material D definitely indicates that the behavior is identical.

As seen from Fig. 2.15, the yield strengths of materials B and C both fall very close to the values for reference material D. Material A, however, has clearly lower values. In reviewing the comparison of the fracture toughness data shown in Fig. 2.15, it should be noted that the lower limit curve is used for reference material D and the best-fit curves for the other materials.

As expected, material A shows the lowest K_{Ic} values, which, however, are only slightly below the lower limit curve for material D. The best-fit results of materials B and C yield higher K_{Ic} values. Single values fall nearly at the lower limit curve of reference material D.

The K_{Ic} values of all materials coincide at -321°F . From this it is evident that it is dangerous to extrapolate upward from low-temperature data.

Fig. 2.15. Comparison of K_{Ic} fracture toughness and yield strength for 22 Ni Mo Cr 37 forgings and A 533, grade B, class 1 steel, 12-in. HSST plate 2. (Refs. 4, 5, curves 588560-B and 593710-A.)

Generally for comparable parameters the tested forgings from 22 Ni Mo Cr 37 show essentially the same properties as the reference material A 533, grade B, class 1.

FRACTURE MECHANICS INVESTIGATIONS

J. G. Merkle

The objectives of the fracture mechanics investigations are (1) to establish the applicability of fracture

mechanics methods as currently understood to the low-alloy steels being used in reactor pressure vessels, (2) to extend the applicability of this knowledge to cover the tougher behavior at higher temperatures, and (3) to develop new methods specifically applicable for predicting the fracture behavior of steels which undergo gross plastic strains before fracture. These objectives are being pursued primarily through a series of subcontracts, with related tasks (described elsewhere) contributing significantly to the developments. The activities under the subcontracts are discussed here.

Gross Strain Crack Tolerance of Steels⁷

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The principal effort during this report period has been a study of the effects of crack size on the values of gross strain at maximum load and as a function of test temperature. Several experimental problems encountered earlier were worked out in the current program. Extensometers were redesigned to improve their survival rate when the specimens broke. Gross strains were measured by the use of extensometers mounted in the gross section above and below the crack as well as in the net section. As a result of earlier difficulties, a careful survey was made of the variation of tensile properties through the thickness and lengthwise of the material received for these tests. In addition to the usual properties, measurements were made of the extent of the yield plateau and the strain at maximum load. Studies of the effects of shape of specimen cross-section and net-section effects were made. The main effort, however, was a study of the effects of crack size on gross strain crack tolerance of the material as a function of temperature. Three crack sizes were studied, in specimens that were geometrically similar.

Material

The material furnished at the beginning of the program was obtained from the lower end of plate 02 – pieces 02GX, Y, and Z as shown in Fig. 2.16. To avoid

the inhomogeneity in yield properties that was found in the first year's work, which required classification of specimens as either "mid-plate" or "bottom-block" material, these pieces were cut 3.5 in. from either plate surface. To check for variation of the yield point phenomenon through the thickness of the blocks and along their length, twenty 0.505-in.-diam tensiles were prepared from locations that provided two thickness traverses and a longitudinal traverse across the material. They were tested at -20°F because most of the tests of gross strain crack tolerance were expected to be at subzero temperatures.

There was very little variation of yield or ultimate strength with position in the block. The range of ultimate strength values was about 2 ksi and of lower yield point values less than 4 ksi. The shape of the stress-strain curves in the yield region was uniform for all locations in the block – the yield plateau extended to about 1.4% total (elastic-plus-plastic) strain. Strain to maximum load was about $12 \pm 1\%$, which means that the upper limit of ϵ_{gnc} , critical gross strain in the net section, is about 12% for this material at -20°F . Consequently, only average results are shown in Table 2.5.

Elongation and reduction of area values were reduced by the presence of slag inclusions in the specimen and by another form of inhomogeneity that produced "fisheyes" in the fracture surface. The inclusions were most evident in the central part of the block, that is, in the middle third of the plate. In the specimen that was most affected, elongation was only 5.5% and reduction of area 20.8%; the ultimate strength was only 66.5 ksi. Four specimens were deemed to be outliers, as indicated in the table.

7. Work performed under UCCND Subcontract No. 3134 between Union Carbide Corporation and TRW, Inc.

Table 2.5. Effects of temperature and location in plate on tensile properties

	$+75^{\circ}\text{F}$		-20°F		-100°F	
	02G	02A	02G	02A	02G	02A
Ultimate strength, ksi	85.4	91.6	90.9 ^a	98.1	99.5	105.3
Upper/lower yield point, ksi	65.4/65.2	71.6/70.5	70.1/63.3 ^a	75.1/73.7	77.0/76.3	83.8/79.4
Yield point elongation, %	1.2	1.0	1.4 ^a	1.2	1.8	1.6
Strain to maximum load, %	11.1	10.2	12.3 ^a	11.6	13.4	11.8
Percent elongation	26.8	24.0	26.0 ^b	23.3	26.8	26.2
Reduction of area, %	60.2	55.7	53.9 ^b	51.2	49.6	53.8

^aOnly 19 specimens were used in computing the average because slag inclusions caused one specimen to be a distinct outlier.

^bOnly 16 specimens were used in computing the average.

Fig. 2.16. Layout of specimens in plate 02. All blocks were cut with 5 in. thickness centered on the plate midline. Both the gross strain and the smooth tensile specimens were longitudinal.

The trend in tensile properties with temperature was also measured by testing one pair of specimens at $+75^{\circ}\text{F}$ and another pair at -100°F , as shown in Table 2.5. The yield point phenomenon was more pronounced at low temperatures. Tensile instability strain also increased slightly. Trends in the elongation and reduction of area values are hard to detect because many of them were reduced by the presence of slag in the vicinity of the fracture. By comparing the higher

values at each temperature, assuming that they were least affected by the inclusion, it may be concluded that the effect of temperature on these ductility values was small.

Radiographs of the largest inclusions in the machined specimens, taken normal to the plate surface, had the appearance of a cloud of low-density particles. The largest dimension of the larger clouds was between 0.5 and 1.0 in. A study of the radiographs made as a result

of the findings from the tensile tests led to the conclusion that at least one-third of the test results for individual gross strain specimens might be affected by the slag inclusions. The specimen might fail at the inclusion instead of at the fatigue crack, the size of the fracture origin might be put in doubt by the presence of "fisheyes" adjacent to the fatigue crack, delamination might reduce constraint below the crack, or gross strain values might be affected by local strain around an inclusion. Consequently, replacement material, pieces 02AE and 02AF (Fig. 2.16), was furnished from the upper end of plate 02. A similar set of tensile specimens was run on the new material with good results, that is, no significant variation through the thickness or diagonally across the blocks furnished. As shown in Table 2.5, ultimate strength was about 6 ksi higher, yield strength was about 5 ksi higher, and elongation was slightly lower in the material from piece 02A.

Test techniques

General. As described in the report on earlier work the test specimens were rectangular bars cut from the plate with their axis in the longitudinal direction as shown in Fig. 2.16. They were surface cracked at midlength (one crack only). The surface cracks were grown in bending fatigue. The crack starter was a ground slot 0.025 in. wide and two-thirds as long as the crack to be grown. Slot depth was about one-third of the expected crack depth. The 0.11-in.-deep cracks were started in reversed bending at ± 39 ksi. When a crack was visible to the full length of the slot, the

loading was changed to zero-to-tension at 50 ksi. The two larger crack sizes, which permitted a larger slot, were cracked in zero-to-tension loading at 50 ksi throughout. The 0.34-in.-deep cracks required about 80,000 cycles of loading, and the remainder required 150,000 to 250,000 cycles to complete. Cracks were grown to a chosen length that was measured with the aid of a binocular microscope. The length needed to obtain the desired crack depth was known from previous experience.

All specimens were loaded in tension to propagate the crack in the plate-thickness direction from front to back of the specimen. Specimen sizes are given in Fig. 2.17 and Table 2.6.

Loading was effected by means of separate wedge grips that could be adjusted for axially. The extensometers used for the ϵ_{gg} measurement were first used to indicate bending by connecting them to measure the strain difference, front to back. Thus the loading

Table 2.6. Gage lengths used for each specimen size

Specimen dimension (in.)			Gage length (in.)			Dimension "A" Fig. 2.17 (in.)
Crack depth	Thickness, B	Width, W	COD	ϵ_{gn}	ϵ_{gz}	
0.10	0.400	4.00	0.20	5.12	1.60	1.75
0.13	1.200	1.32	0.20	5.12	1.60	1.75
0.11	0.300	1.32	0.20	1.60	0.50	1.75
0.22	0.600	2.64	0.20	3.20	1.00	1.75
0.34	0.960	4.22	0.20	5.06	1.60	1.75

Fig. 2.17. Location of gage marks for displacements ϵ_{gn} , ϵ_{gz} , and COD.

condition may be described as one of uniform tension on the gross section ($\pm 5\%$) up to yield; beyond that it is not known. Asymmetric yielding occurred in the net section because the specimen was cracked on one face, and some lateral shift must have accompanied yielding.

Cooling of the specimen and grips by the vapors from liquid nitrogen was as described earlier. The grips were wrapped to permit the individual control of temperature needed to obtain uniformity. The wrappings and the insulated box that surrounded both grips and specimen are shown in Fig. 2.18.

Strain measurement. One result of the first year's work was the conclusion that critical values of gross strain should be measured in the gross section above and below the crack (ϵ_{ggc}) as well as in the net section spanning the crack (ϵ_{gnc}). The ϵ_{ggc} values had been obtained post test from micrometer readings that gave the reduction in width and thickness, using the technique described earlier. For research purposes the records of load vs displacement provide opportunity for measurement of energy values and for fuller observation of the deformation process. Hence, pairs of extensometers (back to back) were provided for measurement of average values of ϵ_{gg} as well as ϵ_{gn} , and a single extensometer was used for the crack opening displacement (COD).

The seven extensometers shown in Fig. 2.18 utilize the conventional strain-gaged flexure principle with special fingers to contact the specimen. Many gage lengths are required in a size-effect study of this kind, but for reasons of economy and of survivability when the specimen breaks, all extensometers are of one size — as small as possible. The gage points on the specimen are spot welds, which attach adapters made of wire and threaded rod, the length of which can be adjusted fairly readily to fit the extensometer to any gage length. Typical configurations are shown in Fig. 2.18. For size effect studies, gage lengths must be scaled the same as other dimensions. Figure 2.17 and Table 2.6 show the dimensional requirements for placement of gage points. A typical set of records traced from two charts obtained from two-pen X-Y plotters is shown in Fig. 2.19. The dip in load evident in the curve for ϵ_{gn} was caused by sudden yielding elsewhere in the specimen. While not evident in this record, the pens traverse the yield plateau at different times during the test.

Test results

Size and shape of specimen. Earlier studies had revealed that reducing specimen width relative to the crack length did not lower the transition temperature as

expected by reason of reduced constraint, but raised it. This was attributed to a net-section effect, that is, a constraint that was not relieved until yielding occurred in the gross section. Making the specimen narrower decreased the area ratio, A_{net}/A_{gross} , which increased the stress ratio, $\sigma_{net}/\sigma_{gross}$, with the result that a higher strain level was required in the net section before yielding occurred in the gross section. To test this early conclusion, three groups of specimens from piece 02G, all with 0.11-in.-deep cracks (nominal), were tested. In the first two groups, crack area was 1.5 to 2.0% of the gross area, but the specimen cross section was 4.00 X 0.400 in. in one case and 1.32 X 1.20 in. in the other. The third group was 1.32 X 0.300 in., which made the crack area about 7% of the gross area.

Figures 2.20a-c show ϵ_{ggc} , ϵ_{gnc} , and COD, respectively, plotted against temperature. Table 2.7 lists the actual specimen dimensions, and Table 2.8 gives the numerical test results.

By comparing the first two groups of specimens, it may be seen that the shape of specimen cross section had little effect on the transition temperature. For ϵ_{ggc} , the foot of the transition was at -83°F or below in both cases. The reason for the scatter is that these specimens, which have a net-section stress only 1.5 to 2.1% greater than the stress on the gross section, often yielded first at the fillet at one end of the specimen. The Luders' bands spread toward the crack, causing the extensometer in the gross section to indicate higher strain than that in the net section or in the other gross section gage length until the yielded region covers the entire reduced section.

Comparing results for the third group of specimens (Fig. 2.20 also) with the first two, we see a net-section effect amounting to about 20°F ; that is, the transition in ϵ_{ggc} was about -62°F , some 20° above that for the same crack size in larger specimens. Note also the reduced scatter in the small specimen. A crack area/gross area ratio of 7% would appear to give a more practical specimen from this standpoint as well as from considerations of specimen size and load requirements. Furthermore, it gives results that should be conservative when applied to the same crack in full-scale hardware. The amount of conservatism may be a function of crack size, but this is not known.

Comparison of material. Material variation from bottom to top in the plate can be examined in terms of gross strain crack tolerance for 0.11-in.-deep cracks by comparing the results of the 1.32- by 0.300-in. specimens from the net-section study with those from the size-effect study. This comparison is given in Fig. 2.21.

Fig. 2.18. Spot-welded wire adapters for use of extensometers of a single span (top); grips wrapped for better temperature control (bottom).

Fig. 2.19. Typical load-displacement records.

Fig. 2.20. Effects of cross-sectional shape and size relative to crack size on gross strain crack tolerance as a function of temperature.

Fig. 2.21. Effects of location in the plate on ϵ_{ggc} , ϵ_{gnc} , and COD for specimens of 1.32×0.300 gross section containing 0.11-in.-deep cracks.

Table 2.7. Specimen dimensions

Specimen No.	Test temperature (°F)	Specimen Dimensions			Crack Size			Specimen net area (in. ²)	$A_{\text{crack}}/A_{\text{gross}}$ ratio (%)
		Width (in.)	Thickness (in.)	Gross area (in. ²)	Length (in.)	Depth (in.)	Area (in. ²)		
103	+75	4.005	0.403	1.612	0.300	0.100	0.0235	1.588	1.46
102	-50	4.002	0.397	1.588	0.302	0.102	0.0242	1.564	1.52
104	-60	4.001	0.400	1.601	0.302	0.101	0.0239	1.577	1.49
106	-83	4.003	0.402	1.610	0.347	0.105	0.0285	1.582	1.77
101	-90	4.002	0.398	1.594	0.300	0.103	0.0242	1.570	1.52
105	-112	4.000	0.396	1.583	0.304	0.096	0.0229	1.560	1.45
109	-50	1.328	1.200	1.593	0.300	0.128	0.0301	1.563	1.89
112	-72	1.328	1.199	1.592	0.304	0.130	0.0310	1.561	1.95
110	-83	1.328	1.201	1.593	0.321	0.153	0.0386	1.554	2.42
111	-84	1.328	1.201	1.594	0.304	0.123	0.0294	1.565	1.84
114	-104	1.327	1.200	1.592	0.304	0.114	0.0272	1.565	1.71
113	-118	1.328	1.201	1.594	0.311	0.130	0.0317	1.562	1.99
118	-25	1.325	0.3014	0.399	0.331	0.108	0.0280	0.371	7.02
117	-41	1.325	0.3004	0.398	0.331	0.106	0.0276	0.370	6.94
119	-50	1.326	0.3012	0.399	0.333	0.104	0.0272	0.372	6.82
120	-55	1.325	0.3010	0.399	0.333	0.104	0.0272	0.372	6.82
116	-63	1.325	0.2990	0.396	0.342	0.112	0.0301	0.366	7.61
115	-65	1.325	0.3001	0.398	0.338	0.112	0.0297	0.368	7.47
121	-71	1.325	0.2992	0.396	0.336	0.114	0.0301	0.366	7.60
122	-93	1.325	0.3006	0.398	0.337	0.104	0.0275	0.370	6.92
210	-16	1.325	0.3033	0.402	0.331	0.110	0.0286	0.373	7.11
206	-27	1.324	0.3015	0.399	0.331	0.110	0.0286	0.370	7.17
209	-37	1.325	0.3014	0.399	0.329	0.108	0.0279	0.371	6.99
205	-49	1.324	0.2997	0.397	0.333	0.112	0.0293	0.368	7.37
203	-69	1.325	0.3017	0.400	0.333	0.112	0.0293	0.371	7.32
204	-110	1.325	0.3022	0.400	0.331	0.110	0.0286	0.371	7.15
212	0	2.640	0.6013	1.587	0.679	0.216	0.115	1.472	7.25
213	-12	2.642	0.6006	1.587	0.679	0.213	0.114	1.473	7.18
216	-18	2.641	0.6009	1.587	0.675	0.213	0.113	1.474	7.12
217	-21	2.640	0.6012	1.587	0.679	0.218	0.118	1.469	7.44
215	-30	2.642	0.5993	1.584	0.679	0.213	0.114	1.470	7.19
218	-49	2.643	0.6015	1.590	0.675	0.216	0.114	1.476	7.20
222	+32	4.216	0.9611	4.052	1.06	0.33	0.274	3.778	6.77
230	+23	4.216	0.9589	4.023	1.06	0.34	0.283	3.740	7.00
224	+22	4.215	0.9611	4.052	1.06	0.34	0.283	3.769	7.00
228	+11	4.216	0.9607	4.050	1.06	0.34	0.283	3.767	7.00
229	+9	4.215	0.9612	4.052	1.05	0.325	0.268	3.784	6.61
219	-10	4.215	0.9607	4.050	1.06	0.34	0.283	3.767	7.00
221	-55	4.216	0.9605	4.049	1.07	0.34	0.283	3.766	7.00

It can be seen that material from piece 02A from the top of the plate exhibited a transition temperature about 25° above that for the bottom of the plate. In terms of ϵ_{gnc} values at, say -70°F, below the foot of the lower transition, the two materials are comparable.

Crack-size study. Justification for a crack-size study requires little beyond a reminder that a basic objective of the gross strain approach is to treat flaw size

quantitatively. Application to heavy sections requires considerable extrapolation for laboratory-size specimens. To gain an understanding of the effects of crack size on gross strain crack tolerance as a function of temperature, three crack depths were investigated: 0.11, 0.22, and 0.34 in. Specimen width and thickness and the gage lengths for strain measurements, ϵ_{gn} and ϵ_{gg} , were varied in proportion, as shown in Table 2.6,

Table 2.8. Test results

Specimen No.	Test temperature (°F)	Specimen size, nominal (in.)	Maximum load (kips)	Gross stress (ksi)	Net stress (ksi)	Critical gross strain and displacement				
						ϵ_{gpc} , top (%)	ϵ_{gpc} , bottom (%)	ϵ_{gnc} (%)	Displacement in ϵ_{gn} gage length, spanning crack (mils)	COD (mils)
103	+75 ^a	4.00 x 0.400	132.0	81.9	83.1	12.4	9.8	>5.2 ^b	>370	70
102	-50		141.8	89.3	90.7	13.1	12.9	16.0	820	
104	-60		128.2	80.0	81.2		4.1	3.8	19 ^c	30
106	-83		117.5	73.0	74.3	2.5	0.25	1.0	52	7 ^c
101	-90		120.3	75.5	76.7	0.25	0.25	0.33 ^d	17	7
105	-112		119.8	75.7	76.8	2.9	0.25	1.3	68	6
109	-50	1.32 x 1.200	143.1	89.8	91.5	13.6	11.8	12.8	658	89
112	-72		120.7	75.8	77.3	2.6	2.4	2.4	124	15
110	-83		112.8	70.8	72.5	0.69	0.37	0.66	34	10
111	-84		128.8	80.8	82.2	2.6	3.6	3.70	189	16
114	-104		118.8	74.6	75.9	2.2	0.25	1.11	57	9
113	-118		123.5	77.5	79.0	0.26	0.26	0.44	22	6
118	-25	1.32 x 0.300	34.8	87.2	93.8	9.2	7.2	>7.0	>100	>44.0
117	-41		35.0	87.9	94.6	7.7	6.6	7.8	125	74.0
119	-50		34.4	86.2	92.5	6.4	6.0	6.7	108	42.0
120	-55		33.1	83.0	89.0	4.0	4.0	4.9	79	29.5
116	-63		27.4	69.2	74.8	0.23	0.23	0.86	14	8.7
115	-65		29.1	73.1	79.1	0.24	0.24	0.52	8	9.0
121	-71	1.32 x 0.300	25.6	72.2	78.2	0.24	0.24	0.97	15	6.8
122	-93		29.7	74.7	80.2	0.25	0.25	0.54	10	6.4
210	-16		37.6	94.0	101.4	7.7	8.6	9.4	150	70.0
206	-27		37.6	94.2	101.6	6.8	7.0	9.0	144	80.0
209	-37		29.8	74.7	80.3	0.25	0.25	1.2	19	11.0
205	-49		29.7	74.8	80.8	0.25	0.25	1.1	18	13.0
203	-69	0.60 x 2.64	30.2	75.5	81.3	0.25	0.25	0.73	11.7	6.5
204	-110		30.4	76.0	81.9	0.25	0.25	0.34	5.5	3.5
212	0		134.2	84.7	91.2	3.8	3.7	4.0	128	44
213	-12		119.2	75.2	80.9	1.9	1.8	1.9	62	27
216	-18		116.0	73.1	78.7	0.25	0.25	1.1	35	27
217	-21		114.0	71.8	77.7	0.25	0.25	0.87	77.5	22
215	-30	0.96 x 4.22	113.3	71.5	77.2	0.25	0.25	0.43	13.8	10
218	-49		114.0	71.7	77.3	0.25	0.25	0.35	12.1	8.5
222	+32		359.5	88.8	95.4	5.81	5.77	7.24	366	170
230	+23		316.0	78.6	84.5	2.19	2.13	2.27	115	44
224	+22		290.0	71.6	76.9	0.25	0.25	0.63	32	23
228	+11		289.0	71.4	76.7	0.25	0.25	0.55	28	25
229	+9	0.96 x 4.22	296.0	73.0	78.5	0.25	0.25	0.67	34	24
219	-10		281.5	69.5	74.8	0.25	0.25	0.37	18	14
221	-55		294.0	72.5	78.0	0.25	0.25	0.28	14.0	7.5

^aBroke at slag inclusion, not at crack.^bRan out of extensometer travel.^cEstimated (extensometer failed).^dLooks low, but Luders' band pattern confirms that specimen 101 strained much less than 105 or 106.

to make this a study of size effect in specimens that were geometrically similar in all critical dimensions. Thickness was 2.7 times crack depth, enough to provide full constraint according to earlier findings. Specimen widths were 4.0 times the crack length, enough to remove the free surfaces at the edge from proximity to the crack tip. The ratio $A_{\text{crack}}/A_{\text{gross}}$, being about 0.07, gave a small net-section effect which would be on the conservative side if these test data were applied to full-scale hardware problems. Finally, specimen sizes were selected to produce breaking strengths consistent with available machine capacity. Six or seven specimens of each size were tested to establish the effects of temperature in and below the transition range.

Variation of the three critical strains (displacements) COD, ϵ_{gnc} , and ϵ_{ggc} with temperature are plotted in Fig. 2.22a-c for the three crack depths 0.11, 0.22, and 0.34 in., respectively, to reveal any differences in the general pattern. One difference is apparent. The transitions in COD, ϵ_{gnc} , and ϵ_{ggc} for each group of specimens should occur at the same temperature. This is true for the largest and the smallest crack sizes, but for the 0.22-in.-deep cracks the foot of the transition (judging by the point of maximum change in slope) appears to be some 12° lower for ϵ_{gnc} and COD than for ϵ_{ggc} . One factor that affects the congruency of the plots of Fig. 2.22 is the scatter in values of critical local strain capability from point to point within the material. The two specimens tested at -18 and -21°F possibly represent samples of high local strain capability, with the result observed - high values of COD and ϵ_{gnc} even without the beneficial influence of gross-section yielding. It should be recalled that earlier results link the occurrence of the transition in COD and ϵ_{gnc} to the onset of yielding in the gross section above and below the crack. These tests fit the same pattern. As shown in Fig. 2.23, gross stress reaches yield at the temperature of the transition in ϵ_{gnc} , ϵ_{ggc} , and COD. Thus ϵ_{gnc} at the foot of the transition should be the strain coincident with a stress value of 1.07 σ_{yp} for these specimens. Simple statics requires that this be so. From the stress-strain curves obtained in this program, that strain value is about 2.0%. It is rather large, because the yield plateau alone is 1.3% strain. In none of the specimen sizes did the strain approach this value, although those with 0.11-in.-deep cracks approached it more closely than the others. The only plausible explanation is that strain is not uniform within the gage length used for measuring ϵ_{gn} . Its length, 1.2 W, was chosen to definitely include all the plastic zone emanating from the crack as Luders' bands prior to gross section yielding. An average strain value is quoted, and

strain intensity must be higher on a shorter gage length. Further evidence on this point, given in Fig. 2.24, is the observation that the ratio of the critical displacement in the ϵ_{gn} gage length to COD ranges from 1.7 for the small specimen to 1.3 for the large ones. By comparing displacements instead of strain, the averaging effect is removed. We must conclude that something about the deformation pattern does not scale uniformly. The evidence of Fig. 2.24 suggests the equivalent of a shear lag between the end of the crack and the edge of the specimen that has a greater effect in larger specimens, although they are geometrically similar. The phenomenon needs further investigation.

Results of these three groups of tests are replotted in Fig. 2.25 to show the effects of crack size on gross strain crack tolerance. In each case, the qualitative effect of increasing crack size is to shift the curves toward higher temperatures and lower strain levels. The curves for the three crack sizes were not perfectly congruent, however. Yet there is a similar trend to be seen in COD vs temperature, as in ϵ_{gnc} . A check of the test records revealed no reason to suspect one value more than others. The extent of Luders' band formation on the surfaces of the specimens seemed consistent with the strain values that were recorded.

The prime feature of the plot of ϵ_{ggc} vs temperature (Fig. 2.25a) is the transition from a condition of no measurable plastic strain in the gross section to a strain equal to the yield-point elongation or higher. This transition shifted upward 18°F for a factor of 2.0 shift in crack size from the 0.11- to the 0.22-in.-deep cracks. It shifted upward 40°F for a factor of 1.6 shift in crack size from 0.22- to 0.34-in.-deep cracks. Before deciding that this means that crack size becomes more significant at the larger values, the other measures of crack tolerance are examined.

The second measure of the effects of crack size (Fig. 2.25b) illustrates the trend in ϵ_{gnc} with temperature for the three sizes of specimens. The sharp increase in slope of the curves within the range of strain values that are of greatest interest places great importance on the choice of temperature at which to compare gross strain crack tolerance values and vice versa.

The chosen strain level shown is 1.5%, just above the foot of the transition for the smallest crack size that we tested. It is also about the maximum expected strain level in hardware. The temperature shift at 1.5% strain (ϵ_{gnc}) was 22°F between the groups of specimens having cracks 0.11 and 0.22 in. deep, and the temperature shift was 38°F between cracks 0.22 and 0.34 in. deep. These values are comparable with the shift observed in the temperature required for gross section

Fig. 2.22. Comparison of the pattern of critical strain and COD values for three crack sizes as a function of temperature.

Fig. 2.23. Effects of crack size and temperature on gross stress at maximum load.

Fig. 2.24. Relationship of displacement measurements for ϵ_{gnc} and COD, as a function of crack size and temperature.

Fig. 2.25. Effects of crack size on ϵ_{ggc} , ϵ_{gpc} , and COD, as a function of temperature.

yielding to precede fracture. Both measures of the effect of crack size are plotted in Fig. 2.26. Also shown are cross plots of Fig. 2.25b for other values of ϵ_{gnc} . The differences between them are slight until strain levels below the foot of the transition are reached. In terms of strain tolerance at -40°F , Fig. 2.25b shows a downward shift from 1.2 to 0.40 to 0.30% for the three crack sizes, 0.11, 0.22, and 0.34 in., respectively. The lack of a smooth trend in these strain values is attributed to the scatter produced by the heterogeneous nature of the strain at yield in this material. All three values are within the yield plateau.

Why should crack size affect the transition temperature and the coincident strain levels? In the upcoming technical report on this subject, evidence is presented showing that the relationship of crack tip local strain to gross strain in the net section, ϵ_{gnc} , is directly dependent on crack size. Furthermore, lower values of critical local

strain are to be expected for larger crack sizes because the crack tip region is embedded more deeply and the plastic zone does not surface so readily; that is, the deeper crack provides more constraint. Therefore a shift to higher temperatures is required to raise the critical local strain to the value required to develop gross section yielding and the transition in critical gross strain values.

From the foregoing explanation, tying the effect of temperature on ϵ_{gnc} to its effect on flow stress, there follows the argument that the shift in the transition temperature with crack size should depend on the rate of change of flow stress with temperature in the temperature region in question. It is seen from Fig. 2.23 that the curve is concave upward, the slope becoming less at higher temperatures. Thus it follows that the effect of crack size on transition temperature should increase as crack size increases, which was observed.

Fig. 2.26. Effect of crack size on the temperature required to achieve various levels of gross strain before fracture.

Extrapolation of the data plotted in Fig. 2.26 to larger crack depths would therefore seem justified, although it is done with some trepidation. In the presence of a surface crack 0.50 in. deep by about 1.6 in. long, this material would be predicted to require a temperature of about +75°F to develop a gross strain crack tolerance whereby ϵ_{gnc} was 1.5% and yielding in the gross section preceded fracture in specimens similar to those described here.

Conclusions

Throughout the program, an endeavor has been made to decide how gross strain crack tolerance should be measured and defined for application of test results to hardware. The experimental work has been complicated by the presence of a yield plateau in the stress-strain curve of the material, which exceeded 1% strain. Any reported strain value between 0.25 and 1.00%, a region of interest in design, must be regarded as an average of strains within the gage length that were either less than 0.25 or greater than 1.00%, depending on the presence or absence of a Luders' band at that location in the gage length. Increasing the specimen cross-sectional area relative to the crack area did not solve the problem. Those test results did indicate, however, that application of results from specimens to full-scale hardware having identical cracks would be conservative.

Earlier test results had shown that the presence of the yield plateau also contributes to the sharpness of the transition. Our response to this situation, which is unavoidable in the type of material tested, is to use a transition-temperature approach in stating our results. We stated the effects of crack size in terms of the temperature at which gross strain in the net section exceeds 1.5% or the temperature at which gross-section yielding occurs. The two temperatures were essentially the same. For material that does not exhibit a yield point or a sharp transition in critical strain values, the results would be more useful when expressed in terms of tolerance for strain in the presence of a crack of a given size at the temperature of interest. Test procedures and the graphical presentation of data are the same for both kinds of material.

Tests of three sizes of surface cracks in geometrically similar specimens have shown that increasing the crack size necessitates significantly higher temperatures to meet the gross-strain criteria given above. Tests at still larger crack sizes are planned. The presence of strain gradients in either the axial or transverse direction is expected to have significant effects on gross strain crack tolerance as a function of temperature. An investigation is planned.

Strain Rate and Crack Arrest Studies⁸

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The selection of specimen blanks for evaluating the mechanical properties of a material to be used in a structure such as a pressure vessel inherently involves the sampling of representative material. Fracturing in service, however, would not be expected to originate in such "typical" material, but rather in a weak region where the fracture toughness is less than this representative value. Such accidentally formed weak regions might be weld defects, hard spots, or sections of unanalyzable stress concentration. The relevant toughness measurement under such a situation would be the condition for arrest of the crack as it emerges from the inhomogeneous region, that is, the crack arrest toughness K_{Ia} . The latter is defined as the stress intensity factor at which a running crack arrests.

The work at the Materials Research Laboratory (MRL) on pressure vessel steels^{9,10} as well as the dynamic data of Shabbits¹¹ showed that high strain rates lower fracture toughness. A running crack establishes a straining rate even in excess of that obtainable with dynamic loading of a stationary crack; therefore the crack arrest toughness would be expected to be the most conservative toughness measure. Moreover, the high straining rate, through its effect in elevating yield stress, should limit the size of the plastic zone and thereby promote the development of plane strain at the crack front. Hence, special emphasis has been given to the measurement of K_{Ia} at MRL.

Previous test results at -70 and 0°F support the following generalizations:

1. The toughness values obtained both statically and dynamically covered a range which agrees well with the values reported by Shabbits et al.^{11,12}

8. Research performed under UCCND Subcontract No. 3152 between Union Carbide Corporation and Materials Research Laboratory, Inc.

9. P. B. Crosley and E. J. Ripling, *Crack Arrest Fracture Toughness of A 533 Grade B Class 1 Pressure Vessel Steel*, HSSTP-TR-8 (March 1970).

10. P. B. Crosley and E. J. Ripling, "Crack Arrest Toughness of Pressure Vessel Steels," to be published in *Nuclear Engineering and Design*.

11. W. O. Shabbits, *Dynamic Fracture Toughness Properties of Heavy Section A 533 Grade B Class 1 Steel Plate*, HSSTP-TR-13 (December 1970).

12. W. O. Shabbits, W. H. Pryle, and E. T. Weasel, *Heavy Section Toughness Properties of A 533 Grade B Class 1 Steel Plate and Submerged Arc Weldments*, HSSTP-TR-6 (December 1969).

2. At each temperature a single level of arrest toughness was measured, independent of the preceding initiation toughness.
3. The difference between the initiation and the arrest toughness tended to be greater at higher temperatures and lower loading rates. Tests of SEN tension specimens are being planned to further explore the invariance of the crack arrest toughness. With these tests, in contrast to the DCB specimens, the stress intensity factor increases as the crack extends.

The invariance of the crack arrest toughness with respect to the preceding initiation value and the occurrence of greater differences between the initiation and the arrest toughness at higher temperatures suggested the possibility of making crack arrest toughness measurements at temperatures where initiation toughness could not be measured. To do this, the initiation toughness would be artificially reduced in order to get the crack running. Three techniques for reducing the initiation toughness have been employed:

1. preloading fatigue crack specimens at 500 to 550°F followed by aging a week or more prior to final testing,
2. imposing a temperature gradient along the specimen so as to initiate fracture at a low temperature and arrest the crack at a higher one,
3. austenitizing the front end of the specimen with the back end immersed in water and water quenching so as to produce a brittle region for crack initiation.

A plot of crack arrest toughness vs temperature is shown in Fig. 2.27. Some of the values above 0°F and, specifically, all the values above room temperature were obtained with the help of the techniques listed above. So far they have enabled measurements up to about 120°F. Measurements might be made at still higher temperatures with combinations of these procedures and/or with the use of brittle weld crack starters.

The crack arrest toughness data above 0°F exhibit more scatter than at lower temperatures, and it is important to determine whether the upper or lower points are more representative of K_{Ia} . The static data of Shabbits et al.¹² suggest that a failure to meet size requirements can lead to lower toughness values. Some of the apparent lowering might be attributable to yielding which, on applying the secant offset method, could be wrongly interpreted as crack extension. The crack arrest data plotted in Fig. 2.27, however, are based on load values following an abrupt instability,

Fig. 2.27. Crack arrest toughness as a function of temperature. Data from all tests.

and there is no reason to expect that such values would be abnormally low.

Other considerations suggest that the lower values plotted in Fig. 2.27 may be the more representative. The higher values are often associated with a wandering of the crack from the minimum section defined by the face grooves which entails a shear separation at the edges. Figure 2.28 is an attempt to express this observation in a more quantitative way. For the crack arrest measurements above 20°F, an index of crack arrest toughness is plotted against the distance by which the crack surface has departed from the minimum section at the arrest position as measured on the broken specimens. The index of arrest toughness is the fractional excess of each data point above a line representing the bottom of the crack arrest scatter band. The plot shows a definite tendency for cracks that have remained flat and in the minimum section to be associated with lower crack arrest toughness values. It is reasonable to set a limiting deviation from the minimum section as a testing requirement. For instance, considering only those cases where the crack plane at arrest is less than 0.10 in. from the minimum section, the curve shown in Fig. 2.29 results. Clearly, it emphasizes the lower range of the scatter band of Fig. 2.27.

With these values of K_{Ia} , it is possible to compare static, dynamic, and arrest toughness as a function of test temperature. The scatter band of static and dynamic toughness obtained by Westinghouse and

Fig. 2.26. Elevation of crack arrest toughness with departure of crack plane from minimum section of specimen.

Fig. 2.29. Crack arrest toughness as a function of temperature. Data from tests where crack front was within 0.010 in. of minimum section.

MRL, along with the arrest toughness data from Fig. 2.29, are shown in Fig. 2.30. Westinghouse used compact tension specimens, while MRL used contoured DCB specimens. It is apparent that these two types of tests yield similar data. At low temperatures the arrest

Fig. 2.30. Comparison of static, dynamic, and arrest toughness.

and dynamic toughnesses are also similar. The abrupt increase in both static and dynamic toughness at temperatures above 0°F, however, does not appear to occur for the arrest toughness values measured to date. A summary of conclusions based on the work completed to date follows.

1. Fracture toughness decreases with increasing strain rate. Dynamic initiation toughness is lower than static initiation toughness; crack arrest toughness is lower than dynamic initiation toughness.
2. The effect of strain rate becomes more pronounced with increasing temperatures.
3. Dynamic initiation toughness can be measured at temperatures where static measurements cannot be made; crack arrest toughness can be made at temperatures where dynamic measurements can no longer be made.
4. Both static and dynamic initiation toughness values show rapid increases with increasing temperature in specific temperature ranges. Measurements of crack arrest toughness up to 120°F have not yet shown a similar abrupt increase.
5. How far the temperature range for crack arrest measurement can be extended is not yet known. The arrest toughness could begin to show an abrupt upturn as do the initiation toughness parameters; it could equally well begin to level off as do the energy absorption values in Charpy impact and drop-weight tests.

6. Should the crack arrest toughness exhibit a stable upper-shelf behavior, the importance of the abrupt increase of initiation toughness as a guarantee of fracture safety may have to be reevaluated.

Three-Dimensional Elastic-Plastic Stress and Strain Analysis for Fracture Mechanics¹³

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In this report we present the numerical solution for the stress-strain distributions in a specimen with a part-through semielliptic crack.

The stress intensity factor K along the minor axis of the crack is calculated from the elastic solution. The significance of the induced moment by the uncracked region of the plate is exhibited by comparing the results with those for a notched plate.

For the elastic-plastic regime, we completed the solutions for two cases: one for an elastic, perfectly plastic material and a second for an elastic strain-hardening material. In the latter case the strain-hardening properties are taken from the tensile stress-strain curve of the actual material used in the tests carried out by the Southwest Research Institute (SwRI).¹⁴ In this report, however, we present only the solution for the perfectly plastic case. Presently, we are in the process of analyzing the numerical results for the strain-hardening case. These results will be compared with the experimental results from the tensile tests carried out by SwRI.

Status of the three-dimensional finite-element program¹⁵

The reliability of the computer program and the distorted cubic elements used in this study has been already tested and assessed.¹⁶ Further work, however, was needed to improve the efficiency of the program and to reduce the CPU time and the total machine time

to an economically acceptable level for problems in the 5000 to 10,000 degrees of freedom range. The work was aimed toward reducing the number of I/O calls to a minimum, incorporating a direct method for solving the stiffness equations, and improving the speed of solution by the block overrelaxation method. Furthermore, the program has been modified and is being documented to make it more user oriented. Postprocessor programs have been developed to plot the results on a Calcomp.

In this study, the problem solved has 4500 degrees of freedom. The elastic solution requires 15 min total CPU time, while the solution for each subsequent load increment requires 10 to 12 min total CPU time on the IBM 360/91 computer.

Specimen specification. The specimen dimensions and the crack configuration are similar to those in the specimen tested at SwRI: width, 18 in.; thickness, 6 in.; and a semielliptic crack of 6.9 in. major axis and 2.0 in. minor axis. The length of the finite-element model is taken as 30.0 in.

The system of coordinates used in our analysis is as follows. The $z = 0$ plane is the plane of the crack, x axis is in the direction of the width, and y axis is in the direction of the thickness of the specimen (see Fig. 2.31).

The specimen is loaded in tension by imposing a uniform displacement in the z direction on a plane parallel to the plane of the crack at a distance 2.5 times the thickness of the specimen, that is, 15 in. The following material mechanical properties are assumed: yield strength, 66,500 psi; Poisson's ratio, 0.30; and Young's modulus, 30×10^6 psi.

In the plastic range, the material is assumed to be perfectly plastic or isotropically strain hardening, with the stress-strain curve obtained from a uniaxial tensile test.¹⁴

Problem specification. Because of symmetry, only one quadrant of the actual specimen need be considered. The boundary displacements are, therefore, as follows: In the crack plane, all points not within the crack are fixed in the z direction. The points on the plane $x = 0$ are fixed in the x direction. The quadrant is also fixed so as to preclude a rigid-body displacement in the y direction.

The quadrant is divided into nine slices varying in thickness from 0.25 in. adjacent to the crack plane to 5.0 in. adjacent to the plane of loading. Figure 2.31 shows the distribution of the nodal points and elements in the plane of the crack, which is identical for all the nine slices. For the most part, the nodal points are defined by the intersection of rays emanating from the

13. Work performed under UCCND Subcontract 3153 between Union Carbide Corporation and Brown University.

14. J. G. Merkle, private communication on results of Southwest Research Institute test, December 1969.

15. P. V. Marcal and J. R. Rice, *Three-Dimensional Elastic-Plastic Stress and Strain Analysis for Fracture Mechanics*, proposal to Union Carbide Corporation, July 1968.

16. N. Levy and P. V. Marcal, *Three-Dimensional Elastic-Plastic Stress and Strain Analysis for Fracture Mechanics, Phase I: Simple Flawed Specimens*, HSST Technical Report No. 12 (December 1970).

Fig. 2.31. Distribution of elements and node points in crack plane.

center of the coordinate system, with curves defined by

$$\left(\frac{x}{a}\right)^p + \left(\frac{y}{b}\right)^2 = 1. \quad (i)$$

Note that with $p = 2$, $a = 3.45$ in., and $b = 2.0$ in., the curve matches exactly the semielliptic crack contour. On the other hand, when $a = 9.0$ in. and $b = 6.0$ in., the curve can match the outer boundary of the specimen as close as we wish by increasing the value of p . In all, 1152 elements and 1500 nodal points are used.

Solution procedure. The elastic solution corresponding to the load that brings the most highly stressed element to the point of initial yielding is first obtained. To obtain the solution in the plastic range, the load is then incremented by 0.1 of the elastic load. The stiffness for each element is determined by taking into account whether the material is perfectly plastic or strain-hardening.

Results of the elastic solution. The most highly stressed element, element 51 in Fig. 2.31, is the one nearest the crack line along the minor axis direction.

The initial yield load corresponds to an average imposed stress $\sigma_m = 0.617\sigma_Y$, where σ_Y is the yield strength of the material.

Figures 2.32 and 2.33 show the variation of the stress components tangential to the crack line and perpendicular to the plane of the crack as functions of the angle between the radius joining the point on the crack tip and the center of coordinates and the x axis. The stress components σ_{zz} and σ_r are nondimensionalized through division by the average imposed stress σ_m .

Figure 2.34 shows the variation of the degree of strain constraint along the crack tip. The strain constraint is defined as the ratio of the stress component tangential to the crack line to the sum of the stress components normal to the crack tip and perpendicular to the plane of the crack respectively.

In the plane strain state, the strain constraint must be equal to the Poisson's ratio of the material, 0.3, while in the plane stress state the value is zero. The figure shows clearly that the value of the strain constraint equal to 0.3 is reached in the elements in the region where the

Fig. 2.32. Variation of σ_1/σ_m with angle in elements nearest crack tip in the elastic solution.

Fig. 2.34. Variation of the degree of strain constraint with angle at the crack tip.

Fig. 2.33. Variation of σ_{xx}/σ_m with angle in elements nearest the crack tip in the elastic solution.

minor axis meets the crack tip but falls rapidly to zero in the region where the crack meets the free surface.

Figures 2.35 and 2.36 show the variation of σ_{zz}/σ_Y with distance from the crack tip in the elements along the minor and major axes of the semielliptic crack respectively. The value of the ratio $\sigma_{zz}/\sigma_Y = 1.15$ in the element nearest the crack tip along the major axis (see Fig. 2.36) suggests that the mode of yielding in this element is that of an in-plane shear. On the other hand, at the minor axis the values $\sigma_{zz}/\sigma_Y = 1.45$ (Fig. 2.35) and $\sigma_{zz}/\sigma_Y = 2.35$ (Fig. 2.33) exhibit the degree of triaxiality attained in this region.

The stress intensity factor along the minor axis of the crack

Estimation of the stress intensity factor at the midpoint of a part-through crack is of considerable practical importance, since it is in the central section that the breaking through to the far plate surface is imminent. To calculate its value from the numerical results, we use two methods.

Fig. 2.35. Variation of σ_{zz}/σ_y with distance from the crack tip in the elements along the minor axis (elastic solution).

Fig. 2.36. Variation of σ_{zz}/σ_y with distance in the elements along the major axis (elastic solution).

In the first method, we use the expression from the singular analytic solution for the near-crack-tip stress state:

$$K = \sigma_{zz} \sqrt{2\pi r} \quad (2)$$

where K is stress intensity factor, σ_{zz} is stress normal to

the plane of the crack, evaluated at this plane, and r is distance from the crack tip.

The value of $\sigma_{zz} \sqrt{2\pi r}$ may be calculated from the numerical results as a function of r and extrapolated to $r = 0$ to give an estimate of K at the crack tip. Figure 2.37 shows the variation of $\sigma_{zz} \sqrt{2\pi r}/\sigma_0$, with σ_0 being the imposed average tensile stress. The extrapolated value of K/σ_0 at $r = 0$ is found equal to 1.64 in.^{1/2}.

The second method consists in computing the tensile stress averaged over the thickness of the plate $-\sigma_0$, for the elements along the minor axis, and the averaged moment of tensile forces in these elements with respect to an axis parallel to the x axis and passing through the middle of the specimen thickness m_0 .

The stress intensity factor is then computed from the expression suggested by Rice and Levy:¹⁷

$$K = h^{1/2} (\sigma_0 g_f + m_0 g_b) \quad (3)$$

where $g_f(\xi)$, and $g_b(\xi)$ are dimensionless functions of the ratio ξ of the crack depth to the plate thickness.

17. J. R. Rice and N. Levy, "The Part-Through Surface Crack in an Elastic Plate," to be published in the *Journal of Applied Mechanics*.

Fig. 2.37. Variation of $\sigma_{zz} \sqrt{2\pi r}/\sigma_0$ with distance from crack tip in the elements along the minor axis (elastic solution).

These functions are taken from the boundary collocation solution by Gross and Srawley:¹⁶

$$g_I = \xi^{1/2} (1.99 - 0.41\xi + 18.70\xi^2 - 38.48\xi^3 + 53.85\xi^4) \quad (4)$$

and

$$g_{II} = \xi^{1/2} (1.99 - 2.47\xi + 12.97\xi^2 - 23.17\xi^3 + 24.80\xi^4) \quad (5)$$

From Eq. (3) the calculated value for K/σ_y turns out to be 1.58 in.^{1/2}, which agrees with the result of the first method within approximately 4%.

Influence of moment induced by the uncracked region on the stress intensity factor along the minor axis

The reduction of the stress intensity factor due to the induced moment may be estimated by comparing the stress intensity factor K_y with that for a notched plate in plane strain loaded in pure tension, with the crack depth/plate thickness ratio equal to $1/3$. From Rice and Levy,¹⁷ we have

$$K_y/\sigma_y = h^{1/2} g_I(1/3) = 4.485 \text{ in.}^{1/2} \quad (6)$$

Thus the ratio of the actual stress intensity factor to the value of the stress intensity factor for a notched plate is

$$K/K_y \approx 0.36 \quad (7)$$

It is clear that the influence of the load shedding to the noncracked region is very significant. Moreover, using $\xi = 1/3$ and the ratio of crack length to plate thickness (designated by $2a/h$ in Ref. 17) equal to 1.15, Rice and Levy's model predicts that $K/K_y \approx 0.38$, which is within 5% of the value given by the numerical solution. This agreement suggests that the model used in Ref. 17 is a good approximation even for not very shallow cracks as the one used in the analyzed specimen.

Behavior in the plastic range, for an elastic-perfectly plastic material

The behavior in the plastic range is studied by imposing load increments, each equal to 0.1 of the initial yield load. Limit load (in the sense that no solution could be obtained beyond this load even after reducing the load increment by half), is reached at the

sixth load increment. This corresponds to an average imposed stress equal to approximately $0.99\sigma_y$, where σ_y is the yield strength of the material. Thus the limit load of the cracked specimen is less than 1% different from the limit load of the same specimen without a crack.

Figures 2.38 and 2.39 show the extents of the plastic zones at various increments at the midplane and the

18. B. Gross and J. E. Srawley, *Stress Intensity Factors for Single Edge Notch Specimens in Bending or Combined Bending and Tension by Boundary Collocation of a Stress Function*, NASA Tech. Note D-2603 (1965).

Fig. 2.38. Plastic zone front for midplane.

free surface respectively. It can be seen that at the free surface the plastic zone is tilted toward the plane of the crack, while in the midplane an elastic wedge exists even at the limit load.

In Figs. 2.40 and 2.41 the variations of σ_{zz}/σ_Y with distance from the crack tip are shown. It is seen that for the elements along the minor axis this ratio tends to the

Fig. 2.39. Plastic zone fronts on the free surface.

Fig. 2.40. Variation of σ_{zz}/σ_Y with distance from crack tip in the elements along the minor axis at limit load.

Fig. 2.41. Variation of σ_{zz}/σ_Y with distance from crack tip in the elements along the major axis at the limit load.

value of 2.0 as compared with 3.0 in the plane strain crack problem as shown by Levy et al.,¹⁹ using a Prandtl field at the crack tip.

Along the major axis the yielding mode for the first two elements nearest the crack tip is that of an in-plane shear. Farther away from the crack, the elements deform in a purely uniaxial tensile mode of yielding.

Crack opening. The elements used do not allow a discontinuous field of displacements at the crack tip. Thus the crack opening at the crack tip could not be obtained in the manner shown in Ref. 19. However, a measure of the opening at the crack tip is obtained from the separation of the crack surface measured at a point along the minor axis at a distance equal to 0.3 in. from the crack tip. Figure 2.42 shows the variation of the crack surface separation suitably nondimensionalized, as a function of the ratio $(K/K_0)^2$, where K is the current stress intensity factor and K_0 is the stress intensity factor at first yielding. It is clear that a linear relationship between the two variables is obtained up to 5% of the limit load. It is to be noted that such a linear relationship has been shown in Ref. 19 for the plane strain case.

19. N. Levy et al., "Small Scale Yielding near a Crack in Plane Strain: A Finite Element Analysis, to appear in the *International Journal of Fracture Mechanics*.

Fig. 2.42. Crack-free surface separation as function of the ratio $(K/K_0)^2$.

Fracture Mechanics Characterization and Crack-Preparation Studies of HSST Program Materials²⁰

T. R. Mager

Westinghouse Electric Corporation

Testing of compact tension specimens from approximately -50 to approximately 550°F continued this report period. One-inch-thick (1 T-CT) specimens oriented in transverse direction (WR) were tested and compared with test data from specimens oriented in the longitudinal direction (RW). Several larger compact tension specimens were also evaluated. As with the previous data, an equivalent-energy procedure was used to determine the fracture toughness parameter K_{Icd} .

As previously reported, Witt²¹ proposed an equivalent-energy method for relating crack size, temperature, and stress levels at maximum load to fracture and outlined a procedure for obtaining the fracture toughness value K_{Icd} , as follows:

Measure the area under the load-deflection curve up to maximum load of a specimen of width d .

20. This program was performed under UCCND Subcontract No. 3196 between the Oak Ridge National Laboratory and Westinghouse Electric Corporation.

21. F. J. Witt, "Equivalent Energy Procedures for Predicting Gross Plastic Fracture," paper presented at Fourth National Symposium on Fracture Mechanics, Carnegie-Mellon University, Aug. 24-26, 1970 (in publication).

Select any point on the linear portion of the load-deflection curve. Measure the area up to this point and divide this area into the area up to maximum load; call this ratio b . Scale the dimensions of the specimen tested by the factor b and the load by b^2 . Using this load as P_Q , determine K_Q for the scaled-up specimen. The value as determined is unique regardless of the point selected on the linear portion of the curve. This number is K_{Icd} .

Experimental test procedure

Four $\frac{1}{2}$ -in.-thick, ten 1-in.-thick, one 4-in.-thick, and one 10-in.-thick compact tension (CT) specimens were evaluated this report period. The ten 1 T-CT (1-in.-thick) specimens were machined with the crack front oriented in the rolling direction and the load applied transverse to the rolling direction. Thus the 1 T-CT specimens were oriented in the transverse direction (WR).

The four $\frac{1}{2}$ T-CT ($\frac{1}{2}$ -in.-thick) specimens were fabricated from the broken halves of two 1 T-CT specimens oriented in the longitudinal direction (RW). The two 1 T-CT specimens were originally fabricated from HSST plate 02. The 4- and 10-in.-thick compact tension specimens were also machined from HSST plate 02.

All specimens were prefatigue cracked according to the ASTM recommended procedure,²² and a clip gage was placed on the front face of each specimen. The clevises and specimen were placed in temperature-controlled chamber for testing. For temperatures below ambient, liquid-nitrogen vapor was used to cool the test specimen; above ambient, an electrically heated chamber was utilized. The load-displacement readings were read out on standard x-y plotters. After the test the crack lengths at crack initiation were read at five equally spaced intervals, and these were averaged to give the crack length for the test. A time-load behavior was also recorded.

Results

The 1 T-CT specimens oriented in WR direction were tested at nine temperatures from -50 to $+550^\circ\text{F}$. The temperatures were selected to correspond to temperatures for which data from specimens oriented in the RW

22. ASTM Standard E399-70T, *Tentative Method of Test for Plane-Strain Fracture of Metallic Materials*, ASTM Standards, Pt. 31, 1970.

direction were available. Then, using the method previously outlined,²¹ K_{Ic1} (that is, K_{Icd} from a 1-in.-thick specimen) values were calculated at each test temperature. These data are summarized in Table 2.9. These values are a little lower than those previously reported²³ for 1 T-CT specimens oriented in the RW direction. This would indicate that there is not a really significant effect of orientation. As before, these data indicate a very high resistance to fracture at evaluated temperatures.

23. HSST Program Seminar. Prog. Rep. Aug. 31, 1970, ORNL-4653.

The $K_{Ic1/2}$ data are summarized in Table 2.10. The $K_{Ic1/2}$ data from the 0°F tests averaged about 86 $\text{ksi}\sqrt{\text{in.}}$. These data are quite comparable with the 76 $\text{ksi}\sqrt{\text{in.}}$ reported²³ for the 1 T-CT specimen from which the $\frac{1}{2}$ T-CT specimens were fabricated. At +50°F $K_{Ic1/2}$ values of 124 and 132 $\text{ksi}\sqrt{\text{in.}}$ agree even better with the 128 $\text{ksi}\sqrt{\text{in.}}$ value reported²³ for the 1 T-CT specimen from which these $\frac{1}{2}$ -in. specimens were fabricated. The 4 T-CT and 10 T-CT specimens were tested at 550°F. The 10 T-CT specimen failed to fracture at a load of 1,070,000 lb, which was the maximum loading capacity of the test apparatus. However, using the method previously outlined,²¹ a K_{Ic10} was calculated from the load displacement curve.

Table 2.9. Fracture toughness results determined from testing 1 T-CT specimens oriented in the RW direction at temperatures up to 550°F.

Specimen designation	Test temperature (°F)	Crack length (in.)	Some fracture toughness parameters ^a ($\text{ksi}\sqrt{\text{in.}}$)				K_{Icd} ($\text{ksi}\sqrt{\text{in.}}$)
			KD	KQ	KU	KF	
3C21	-50	1.090	35.2	61.4	61.4	17.2	64.7
3C22	25	1.136	32.2	55.1	79.8	18.6	76.0
3C23	75	1.103	54.4	65.6	100.1	17.6	142.7
3C24	100	1.098	54.7	66.2	101.6	17.4	182.8
3C25	150	1.081	54.8	62.5	99.5	16.9	227.4
3C26	200	1.088	52.3	64.6	96.4	17.1	253.0
3C27	250	1.088	53.0	65.7	99.5	17.1	244.6
3C28	350	1.088	35.1	56.9	96.0	17.1	160.2
3C29	550	1.091	33.7	53.7	95.3	17.2	158.5
3C30	550	1.058	48.1	65.2	97.8	17.0	203.6
HS2-28-C3 ^b	550	1.038	50.3	62.5	97.9	15.8	211.6

^a KD = fracture toughness at deviation - load displacement, KQ = fracture toughness at secant offset load; KU = fracture toughness at ultimate load, and KF = fracture during prefatigue precracking.

^bOriented in the RW direction.

Table 2.10. Fracture toughness results determined from testing $\frac{1}{2}$ T-CT specimens oriented in the RW direction

Specimen designation	Test temperature (°F)	Crack length (in.)	Some fracture toughness parameters ^a ($\text{ksi}\sqrt{\text{in.}}$)				K_{Icd} ($\text{ksi}\sqrt{\text{in.}}$)
			KD	KQ	KU	KF	
HS2-7-C13	0	0.515	28.5	47.0	71.5	20.0	87.2
HS2-7-C14	0	0.525	26.3	46.4	73.5	20.6	85.5
HS2-7-C15	50	0.518	23.2	38.9	74.6	20.2	132.2
HS2-7-C16	50	0.525	28.4	46.4	65.2	20.6	123.9

^a KD = fracture toughness at deviation - load displacement, KQ = fracture toughness at secant offset load, KU = fracture toughness at ultimate load, and KF = fracture toughness during prefatigue precracking.

Table 2.11. Fracture toughness results determined from testing a 4 T-CT and a 10 T specimen oriented in the RW direction

Specimen designation	Test temperature (°F)	Crack length (in.)	Some fracture toughness parameters ^a (ksi $\sqrt{\text{in.}}$)				K_{Ic4} (ksi $\sqrt{\text{in.}}$)
			KD	KQ	KU	KF	
4D (4T)	550	4.178	103.6	145.4	193.5	16.3	262.6
2HA-1 (10T)	550	10.250					271.8 *

^aKD = fracture toughness at deviation -- load displacement, KQ = fracture toughness at secant offset load, KU = fracture toughness at ultimate load, and KF = fracture toughness during pre-fatigue precracking.

The K_{Ic4} and K_{Ic10} data are reported in Table 2.11. As before, these data indicate a very high resistance to fracture at 550°F.

Future work will include evaluating 6- and 9-in.-thick specimens at temperatures corresponding to the upper-shelf Charpy V-notch range (200–550°F).

FATIGUE AND CRACK-PROPAGATION INVESTIGATIONS

L. F. Kooistra

The effects of environment and stress state on crack growth rates are being studied. The current activity emphasized crack growth rate in an environment of high-temperature water (without irradiation). All combinations of two water conditions (typical of pressurized-water and boiling-water reactors), four materials (plate, submerged-arc weld, electroslag weld, and shielded-metal-arc weld), and two locations (surface and center for plate, weld metal and heat-affected zones for welds) are being investigated at 550°F with the use of 2 T compact tension specimens. Some characterization studies will also be made at room temperature. The status of this work is discussed below.

Fatigue Crack Growth Characteristics of Nuclear Pressure Vessel Grade Materials²⁴

T. R. Mager

Westinghouse Electric Corporation

Introduction

Using the fracture mechanics approach, the effect of an environment of high-temperature primary reactor-grade water on the fatigue crack growth characteristics of nuclear pressure vessel grade materials is being

studied. The stress intensity concept of linear elastic fracture mechanics provides a parameter K which describes the stress situation at the tip of an existing crack. It has been shown that the cyclic range of the stress intensity factor ΔK is the controlling stress parameter which determines the fatigue crack growth rate. The crack growth rate da/dN of pressure vessel materials is being measured as a function of ΔK_I , the change in the stress intensity at the tip of the crack. The fatigue crack growth data are being generated with 2 T WOL (wedge opening loading) specimens at 550°F in an environment (excluding irradiation) typical of PWRs and BWRs. The 2 T WOL specimen is illustrated in Fig. 2.4¹.

Tests are being performed on A 533, grade B, class 1 steel (HSST plate J2) and weld metal prepared by the submerged-arc, electroslag, and manual-arc processes. Material from the heat-affected zone (HAZ) of the weldments prepared by the submerged-arc and manual-arc processes are also being studied.

A test chamber was designed and fabricated to operate at 550°F and 2000 psi. The test chamber was fabricated from stainless steel. The water was pumped from an auxiliary tank through the test chamber. Typical water chemistry is given in Table 2.12. Cycling rates were from 60 to 600 cpm for a minimum of 100,000 cycles. Initially, the ultrasonic crack growth monitoring technique was utilized to measure the crack growth in the 2 T WOL specimens. Use of ultrasonics as the chosen crack growth rate monitoring technique required that the test specimen be clamped to the lid of the test chamber. However, it was concluded that

24. Work sponsored by HSST program under UCCND Subcontract No. 3290 between Union Carbide Corporation and Westinghouse Electric Corporation.

Fig. 2.43. "2T" WOJL-type compact fracture toughness specimen.

Table 2.12. Typical chemistry of proposed environment (water)

	Pressurized-water reactor	Boiling-water reactor
Oxygen, ppm	<0.1	0.2-0.4
Chloride, ppm	<0.15	<0.2
Fluoride, ppm	<0.15	<0.1
Total suspended solids, ppm	<1.0	<0.2
Boron, ppm	0-4000 (500) ^a	
Solution pH	4.2-10.5 (5.1) ^a	6-6.5
Electrical conductivity, μ mhos/cm	<1-40	<0.1
Hydrogen, cc (STP)/kg	25-35	0.03-0.05
Li(OH), m	0.3×10^{-4} - 3.2×10^{-4}	

^a Average.

because of the constraint of the chamber lid, the "stress intensity K_I " at the crack tip was not well defined. To correct this problem, the experiment was redesigned to utilize compliance (LVDT gage) to measure the crack growth rather than ultrasonics.

Experimental results

Crack length a vs number of cycles N data were generated for twelve 2 T WOJL specimens. In order to express the raw test data (a vs N) in terms of fracture mechanics, the data were converted to the form of crack growth rate, da/dN vs ΔK , the cyclic stress intensity range. The crack growth rate was established by computing the slope between successive data points on the a vs N curve obtained by means of compliance (crack opening displacement). The ΔK values associated

with a particular crack length, and in turn the corresponding crack growth rate, were determined from the following expression:

$$\Delta K = K_{\max} - K_{\min} = Y \frac{P_{\max} \sqrt{a}}{BW} - Y \frac{P_{\min} \sqrt{a}}{BW}$$

where a is the crack length measured from the center line of loading, Y is a compliance constant which depends upon a/W , W is the specimen width (see Fig. 2.43), B is the specimen thickness, and $P_{\max}$ and $P_{\min}$ are the maximum and minimum load per cycle. The results are presented in Figs. 2.44 to 2.47. Figures 2.44 and 2.45 present the 550°F fatigue crack growth rate vs stress intensity factor range relationship for specimens fabricated from the base metal (HSST plate 02). Figures 2.44 and 2.45 represent data generated in PWR water chemistry and in BWR environment respectively. The upper scatter band from similar data generated by Clark²⁵ with A 533, grade B, class 1 base plate is also shown in Figs. 2.44 and 2.45. The fatigue crack growth rate vs stress intensity factor range relationship for specimens fabricated from the weldment materials is shown in Figs. 2.46 and 2.47. Figures 2.46 and 2.47

show that water chemistry (PWR and BWR) has no effect on the fatigue crack growth rate of the materials investigated (base, weld metal, and HAZ metal) for specimens fatigued at 600 cpm. However, for tests performed at 60 cpm in BWR water chemistry, the data fall outside Clark's scatter band at the low ΔK values. This anomalous behavior is probably related to the threshold level, which is outside the realm of this experiment.

Conclusions

To date, the presence of nuclear reactor grade water environment (chemistry and pressure) for PWR and BWR was found to have no effect on the fatigue crack growth of A 533, grade B, class 1 steel base plate and weldments. Comparison of the results obtained for each material test at 550°F and 600 cpm indicated that the upper scatterband of the crack growth rate data for the base plate material at 75°F encompasses all other data.

25. W. G. Clark, Jr., *Fatigue Crack Growth Characteristics of Heavy Section ASTM A 533 Grade B, Class 1 Steel Weldments*, ASME Paper 70-PVP-24.

Fig. 2.44. Crack growth rate as a function of ΔK for A 533, grade B, class 1 steel plate in PWR environment.

Fig. 2.45. Crack growth rate as a function of ΔK for A 533, grade B, class 1 steel plate in BWR environment.

Fig. 2.46. Crack growth rate as a function of ΔK for A 533, grade B, class 1 weldment material in PWR environment.

Fig. 2.47. Crack growth rate as a function of ΔK for A 533, grade B, class 1 weldment material in BWR environment.

3. Investigation of Irradiated Materials

Irradiation is one of the environmental factors that must be considered in safety evaluations of reactor pressure vessels, since the mechanical properties of metals may be modified thereby to a degree that is of considerable engineering significance. Investigations of irradiation effects have been carried out by Hanford Engineering Development Laboratory (HEDL), Westinghouse Electric Corporation, Naval Research Laboratory (NRL), and Oak Ridge National Laboratory. The areas of primary interest to the HSST program are the temperature range 450 to 600°F and the fast-neutron ($E > 1$ MeV) fluence range 1 to 8×10^{19} neutrons/cm².

The work at NRL, Westinghouse, and ORNL has essentially been completed, and final reports are in

preparation. The recent work at HEDL and a cooperative effort with Verein Deutscher Eisenhüttenleute are described here.

IRRADIATION EFFECTS ON THE FRACTURE OF HEAVY-SECTION PRESSURE VESSEL STEELS

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Specimen Irradiation

Irradiation of specimens in the M-3 hot water loop of the Engineering Test Reactor (ETR) was terminated at the end of cycle 109 (December 1970); goal exposures were not obtained on all specimens. Table 3.1 summarizes the irradiated specimens obtained from the

Table 3.1. Summary of specimens irradiated in the ETR during cycles 108 to 110

Specimen type	Number of specimens	Material description	Approximate fluence at $E > 1$ MeV (neutrons/cm ²)
			$\times 10^{19}$
Tensile	9	Transverse (W), A 533 base	2
	8	Transverse (W), A 533 base	5
	3	Weld metal	0.5-1
	4	Weld metal	5
High-strain-rate tensile	12	Longitudinal (RW), A 533	0.5-1
1 in. CT	4	Transverse (WR), A 533 base	2
	4	Transverse (WR), A 533 base	5
	4	Weld metal	5
	4	HAZ	0.5-1
Charpy V	5	Transverse (WR), A 533 base	0.5-1
	5	Transverse (WR), A 533 base	2
	7	Transverse (WR), A 533 base	0.5
	5	Longitudinal (RW), A 533 base	0.5-1
	5	Longitudinal (RW), A 533 base	2
	7	Longitudinal (RW), A 533 base	5

loop during cycles 108 through 109. The fluences reported are approximate, pending analysis of all flux monitors.

The M-3 loop water temperature was controlled at 510°F for all irradiations; tensile and Charpy specimen maximum temperatures are 513°F, and compact tension specimen center temperatures are 540°F maximum.

Irradiation of 4T Specimens

The irradiation of thicker fracture specimens is necessary for the measurement of higher K_{IC} values. The details of 4T CT (4-in.-thick compact tension) specimen irradiations and associated problems were discussed at length in the previous semiannual report.¹ In summary, the basic problems of large specimen irradiations are space, sufficient neutron flux to perform the irradiation experiment in a reasonable time period, and a low enough gamma flux to avoid overheating and yield a low crack line temperature gradient.

The ETR at Idaho Falls is suitable on the basis of space and flux for irradiating two specimens to a fluence of 5×10^{19} neutrons/cm² ($E > 1$ MeV) in less than a year. A patch concept for reducing the crack line temperature of the specimen was previously described;¹ sufficiently reliable gamma flux data are not available for calculation of patch size and gas gap to assure correct experimental temperature control during irradiation.

A 4T CT dummy specimen gamma heat experiment was designed to determine specimen temperature profiles in a reflector position facility of the ETR. The irradiation assembly (Fig. 3.1) has been completed and shipped to the ETR for use. The experimental assembly will occupy eight 3- by 3-in. reflector positions. Upper and lower support boxes position the specimens in desired reactor location and channel reactor coolant water for specimen cooling; they function as a single unit but are built in upper and lower sections for reactor installation. Lower and upper support blocks are keyed to the corners of the support box for vertical support and positioning of the specimens; the blocks must be sized to meet positioning requirements and can be withdrawn for removal of the lower specimen. The

Fig. 3.1. Irradiation assembly used in the reflector position of the ETR for irradiating 4T CT fracture specimens.

support box and the support blocks are designed specifically for irradiation of 4T CT specimens as well as the dummy gamma-heat experiment.

The thermocouple instrumented dummy gamma-heat specimen will occupy the space above the reactor midplane; the specimen is a block of A 533 grade B steel the size of a 4T CT specimen (4 X 9.6 X 10 in.). The spacer blocks are adjusted so that the bottom of the specimen is 1 in. above the midplane and the top of the lower aluminum blank is 1 in. below the midplane. Flux gradients are nearly symmetrical about the reactor midplane, so only the temperature gradients in one specimen need be measured. Fifty thermocouples are placed at critical points to provide profiles throughout the specimen. The thermocouples are ISA type K calibration sheathed in 347 stainless steel and insulated with MgO.

The specimen is to be irradiated at full power reactor operating condition in direct contact with the reactor coolant water. Temperature from each thermocouple within the specimen will be monitored at each power level during reactor startup. Full reactor power temperature profiles will be measured at the beginning, middle, and near the end of the reactor cycle.

1. C. W. Hunter and J. A. Williams, "Irradiation Effects on the Fracture of Heavy Section Pressure Vessel Steels," *HSST Program Semiannu. Progr. Rep. Aug. 31, 1970*, ORNL-4653, pp. 31-36.

Base Material Studies of ASTM A 533 B

High fluence irradiated 1T CT specimens from 12 in. ASTM A 533, grade B, class 1 HSST plate 02 were tested at 200 and 245°F. The additional sensitivity to irradiation embrittlement exhibited by these specimens irradiated to fluences of 8×10^{19} neutrons/cm² ($E > 1$ MeV) is shown in Fig. 3.2. The fracture toughness results of the specimens irradiated to 8×10^{19} neutrons/cm² is given in Table 3.2. Irradiation to 8×10^{19} neutrons/cm² caused an additional shift of about 90°F higher than did irradiation to a fluence of 2×10^{19} neutrons/cm². Irradiation embrittlement at 2 and 8×10^{19} neutrons/cm² caused a fracture toughness of 50 ksi $\sqrt{\text{in.}}$ to exist at respective temperatures of 185 and 274°F higher than for unirradiated material. The trend above the 50-ksi $\sqrt{\text{in.}}$ toughness level at both irradiated conditions is for increasing toughness with increasing temperatures. The rate at which toughness

will increase or the level that may be obtained at higher temperatures cannot be assessed by currently available data of this study.

The fatigue crack preparation stress intensity was cursorily examined as to its effect on K_{Ic} , fracture toughness; these results are shown in Fig. 3.3 and Table 3.3. All fatigue cracks were prepared at room temperature. The fracture toughness of $1/2$ plate thickness position fatigued at 10 ksi $\sqrt{\text{in.}}$ is about equivalent to the earlier data obtained from $3/8$ plate thickness position material fatigued at 17 to 21 ksi $\sqrt{\text{in.}}$. Fatigue crack preparation at 10 ksi $\sqrt{\text{in.}}$ of $1/4$ plate thickness position yielded a 50 ksi $\sqrt{\text{in.}}$ fracture toughness at 40°F higher temperature than the $1/2$ position and 50°F higher than the earlier $3/8$ plate thickness position data. The shift in the 50 ksi $\sqrt{\text{in.}}$ level due to irradiation reported above is given with respect to the fracture toughness of material from $1/4$ plate thickness position with a crack preparation stress intensity of 10 ksi $\sqrt{\text{in.}}$.

Fig. 3.2. Irradiated and unirradiated fracture toughness as a function of temperature for ASTM A 533, grade B, class 1 steel from HSST plate 02. Closed points on the curves are K_Q values; K_{Ic} determinations were invalid based on a or $b < 2.5 (K_{Ic}/\sigma_{ys})^2$. All specimens are from the RW orientation (longitudinal) of the plate.

Table 3.2. Fracture toughness of irradiated ASTM A 533, grade B, class 1 steel from HSST plate 02 steel, RW orientation^a

Specimen identification	Test temperature (°F)	Fluence at $E > 1$ MeV (neutrons/cm ²)	Plate position ^a (in.)	$K_{fatigue}$ (ksi $\sqrt{\text{in.}}$)	Yield strength (psi)	Crack length ^a (in.)	K_Q (ksi $\sqrt{\text{in.}}$)	$2.5 (K_Q/\sigma_{ys})^2$	K_{Ic} (ksi $\sqrt{\text{in.}}$)
		$\times 10^{19}$			$\times 10^3$				
02GA 44	200	8.2	3.4	10.3	111.5	1.005	46.2	0.43	46.2
02GA 43	245	8.3	3.4	10.2	109	1.000	63.3	0.84	63.3

^aFrom plate surface to specimen center.

Fig. 3.3. Fracture toughness of unirradiated ASTM A 533, grade B, class 1 steel from HSST plate 02 at 1/2, 3/8, and 1/4 plate thickness position as a function of temperature for CT fracture specimens. Fatigue crack preparation stress intensity values given are for final crack extension: crack preparation was conducted at room temperature

Table 3.3. Fracture toughness of ASTM A 533, grade B, class 1 steel from HSST plate 02, RW orientation, after crack preparation at low fatigue stress intensity

Specimen identification	Test temperature (°F)	Plate position ^a (in.)	K _{fatigue} (ksi √in.)	Yield strength (psi)	Crack length ^a (in.)	K _Q (ksi √in.)	2.5 (K _Q /σ _{ys}) ²	K _{Ic} (ksi √in.)
x 10 ³								
02GA 17	-184	8.6	10.5	91	1.019	28	0.24	28
02GA 16	-140	8.6	10.3	84	1.007	43.5	0.67	43.5
02GA 19	-100	8.6	10.9	78	1.042	43.7	0.78	43.7
02GA 20	-50	8.6	10.9	72	1.046	51.6	1.28	b
02GA 50	-184	6.0	10.2	91	0.997	36.5	0.40	36.5
02GA 49	-140	6.0	10.6	84	0.988	34.9	0.43	34.9
02GA 51	-100	6.0	10.2	78	1.008	51.5	1.08	b
02GA 52	-50	6.0	10.4	72	1.018	68.6	2.26	b

^aFrom plate surface to specimen center.
^bInvalid by crack length and thickness < 2.5 (K_Q/σ_{ys})²; meets secant offset criterion.

The longitudinal tensile properties of A 533 B for high fluence irradiation of 6 and 8 × 10¹⁹ neutrons/cm² (E > 1 MeV) were evaluated over a temperature range from -320 to 500°F. The results are given in Table 3.4; yield and ultimate strength properties of unirradiated and irradiated A 533 B are shown in Figs. 3.4 and 3.5.

Irradiation at all levels causes general irradiation hardening, resulting in elevation of yield and ultimate

strengths. The irradiation sensitivity of the ultimate strength was less than that of the yield strength; thus the margin between flow and failure is decreased with increasing irradiation.

Figures 3.6 and 3.7 demonstrate, in the range of investigation, that the changes in yield and ultimate strength are directly proportional to the fluence and that no saturation of irradiation strengthening is evident. The slope of the yield strength vs fluence in Fig. 3.6 decreases as the test temperature is increased; this

Table 3.4. Irradiated tensile properties of A 533, grade B, class 1 steel from HSST plate 02, longitudinal specimens

Specimen identification	Test temperature (°F)	Fluence at $E > 1$ MeV (neutrons/cm ²)	Plate position ^a (in.)	Yield strength, 0.2% offset (psi)	Ultimate strength (psi)	Total elongation (%)
		$\times 10^{19}$		$\times 10^3$	$\times 10^3$	
02GA 141	-319	5.5	2.7	177.7	177.7	6
02GA 142	-100	5.8	2.7	127.7	127.8	12.9
02GA 149	80	5.5	2.7	115.5	126.0	12.7
02GA 176	200	6.3	3.4	104.2	116.8	11.0
02GA 150	350	5.8	2.7	100.2	115.9	13.4
02GA 229	350	6.1	5.3	100.8	116.0	12.3
02GA 172	500	6.5	3.4	94.5	114.5	11.5
02GA 231	500	6.5	5.3	91.0	108.8	10.5
02GA 148	-250	8.3	2.7	163.0	163.0	0.1
02GA 146	-100	7.4	2.7	133.6	143.3	12.1
02GA 155	80	8.3	2.7	122.5	131.9	13.1
02GA 154	200	7.4	2.7	110.3	121.8	12.3
02GA 184	350	8.1	4.1	99.2	114.3	9.6
02GA 183	500	7.8	4.1	93.7	112.6	10.2

^aFrom top surface of plate to specimen center.

Fig. 3.4. Yield strength of irradiated and unirradiated ASTM A 533, grade B, class 1 steel from HSST plate as a function of temperature. Longitudinal (R) orientation.

slope is given by

$$S_T = 4.3 - 0.0038T,$$

(1)

probably best described by an expression of the form:

$$A(1 - e^{-\Phi/B}), \quad (2)$$

in which T is temperature in degrees Fahrenheit. The increase in the yield strength at fluences below the lowest experimental result at 2×10^{19} neutrons/cm² is

in which $A = 22$ ksi, $B = 5 \times 10^{18}$ neutrons/cm², and $\Phi =$ fluence, neutrons/cm². Thus the increase in yield strength as a function of fluence and test temperature

Fig. 3.5. Ultimate strength of ASTM A 533, grade B, class 1 steel from HSST plate 02 as a function of temperature. Longitudinal (R) orientation.

Fig. 3.6. Effect of irradiation fluence ($E > 1$ MeV) on the yield strength properties at various temperatures for ASTM A 533, grade B steel from HSST plate 02.

Fig. 3.7. Effect of irradiation fluence ($E > 1$ MeV) on the ultimate strength properties at various temperatures for ASTM A 533, grade B steel from HSST plate 02.

may be calculated from

$$\Delta\sigma_{ys} = 22 [1 - \exp(-\Phi/5 \times 10^{18})] + \Phi/10^{19} (4.3 - 0.00387) \quad (3)$$

This expression quite precisely describes the behavior over the testing temperature range (-250 to $+500^\circ\text{F}$) except at low temperatures and high fluences (e.g. -250°F , 8×10^{19}), under which conditions the true yield stress is exceeded first in the test.

COOPERATIVE EFFORTS BETWEEN THE HSST PROGRAM AND THE PRESSURE VESSEL STEEL IRRADIATION PROGRAM OF VEREIN DEUTSCHER EISENHUTTENLEUTE²

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Introduction

On behalf of Verein Deutscher Eisenhüttenleute (VDEh), Kernforschungsanlage GmbH carries out a pressure vessel steel irradiation program. Three types of low-alloy pressure vessel steels are being investigated. The HSST program has made available to the VDEh two sections of HSST plates 02 and 03 (ASTM A 533, grade B, class 1) as reference material. The VDEh in return will supply the HSST program with irradiation results obtained from this steel.

Extent of VDEh Irradiation Program

Three types of low-alloy Mn-Ni-Mo-V and Ni-Cr-Mo steels are being investigated as base metal, weld metal, and heat-affected-zone (HAZ) metal.

These steels will be subjected to Charpy V-notch impact tests, tensile tests at room temperature and 350°C (662°F), Vickers-hardness tests, and metallography. Specimens will be irradiated in the D₂O-moderated and -cooled DIDO reactor, fuel element position, at fluences (neutrons/cm²) of 10¹⁹, 5 × 10¹⁹, and 10 × 10¹⁹ (*E* > 1 MeV). Temperatures will be 300°C (572°F) and 500°C (752°F).

Sample Preparation of HSST Reference Steel

Figure 3.8 shows the plate sections VDEh received from the HSST program: section A from plate 02 and section J from plate 03. Section A is being used as base metal, and section J has been used for weldment.

Base metal

A thickness of 120 mm (5 in.) was removed from the edge of section A and the section was cut into six slices (A to F) from 1/4, 1/2, and 3/4 thicknesses, as shown in Fig. 3.9. By applying the arrangement shown in Fig. 3.9 a total of 1050 Charpy V-notch and 108 tensile samples were obtained.

2. This work was sponsored by Verein Deutscher Eisenhüttenleute and is reported as a cooperative exchange with the HSST program.

Submerged-arc weldment

The weldment was prepared by Mannesmann Forschungsinstitut GmbH, Duisburg-Huckingen, Germany. Figure 3.10 shows the steps followed in preparing the weldment. Section J was cut into six pieces (A to F), these pieces were placed together to form three blocks, and the blocks were cut longitudinally and then welded together. The isometric in the right part of Fig. 3.10 shows the location of the samples (hatched region). Figure 3.11 shows a photograph of the weld and the location of samples in the welded region. A total of 540 Charpy V-notch and tensile samples were obtained from the weld and the HAZ metals. Table 3.5 contains the welding data and the heat treatment. Ultrasonic inspection was performed between the intermediate and final post-weld heat treatment. There was no indication of failures.

Table 3.5. Welding data and heat treatment of submerged-arc weld

Welding data	Root	Remainder
Electrode		
Size	3.25 mm (0.13 in.)	4 mm (0.16 in.)
Type	EML 5	S 2 NiMo1
Flux	Argonarc	OP 40 TT ^a
Current	120 A	600 A
Polarity	+	+
Arc voltage	20 dc	30 dc
Travel speed		50 cm/min (20 in./min)
Heat treatment		
Preheat: 200–250°C (392–482°F)		
Intermediate post-weld heat treatment (PWHT): 620°C (1150°F), held 15 min, air cooled		
PWHT: 620°C (1150°F) held 12 hr, furnace cooled to 320°C (608°F)		
Cooling and heating speed between 620 and 320°C was 40°C (104°F)/hr		

^aOP 40 TT composition (%): 21.6 CaO, 15.6 SiO₂, 19.0 Al₂O₃, 26.4 MgO, 0.7 TiO₂, 0.34 MnO, 1.64 FeO, 12.9 CaF₂.

Results and Present Status

Figure 3.12 shows the results of Charpy V-notch tests of weld metal and HAZ. In both cases the transition temperatures are approximately -60°C (-76°F) at 30

a) ORIGINAL SECTION J

b) CUT INTO PIECES AND WELDED

c) FINAL CUT AND WELDMENT

Fig. 3.10. Cutting plan for section J, HSST plate 03. Dimensions in millimeters (inches).

Fig. 3.11. Photograph of weld and orientation of samples. Dimensions in millimeters (inches).

ft-lb. Table 3.6 contains the chemical composition of the weld and the results of tensile tests.

The results are satisfactory. The values of the weld reach the ASTM requirements of class 2 steel, and those of the HAZ are well within class 1.

The irradiations have been successfully completed, and postirradiation examinations are under way. They will be completed by October 1971, and the results will be available to the HSST program by the end of 1971.

Fig. 3.12. Ductile/brittle transition of weld and HAZ.

Table 3.6. Chemical composition and mechanical properties of ASTM A 533 B weld

Chemical composition (%)													
C	Si	Mn	P	S	N	Al	B	Co	Cr	Cu	Mo	Ni	V
0.09	0.39	1.36	0.02	0.004	0.01	0.022	0.012	0.041	0.20	0.10	0.51	1.08	0.03
Mechanical Properties													
Tensile test at 20°C (68°F)								Tensile test at 350°C (662°F)					
Yield strength		Tensile strength		Elongation (%)	Reduction in area (%)	Yield strength		Tensile strength		Elongation (%)	Reduction in area (%)		
kp/mm ²	psi	kp/mm ²	psi			kp/mm ²	psi	kp/mm ²	psi				
Weld Metal													
53.6	76219	63.2	89870	27.9	74.3	46.3	65838	58.9	83755	21.7	68.3		
Heat-Affected Zone													
44.3	62946	60.6	86173	18.9	69.5	39.4	56026	57.3	81480	17.3	63.0		

4. Pressure Vessel Investigations

During this reporting period emphasis was placed on the simulated service test activity, although effort was also continued under the complex stress state studies and the specific safety research studies. Activities reported include the development of procedures for flaw sharpening of the simulated service test specimens, the testing of the eight additional large flawed tensile specimens at Southwest Research Institute (SwRI), and an evaluation of the acoustic emission data obtained during the testing of these specimens. Expediting of procurement of six intermediate test vessels from Taylor Forge Company was continued, and a summary of the material properties evaluation of the vessel components led to a retempering of the shell courses.

HYDRAULIC FATIGUING USED FOR FLAW SHARPENING OF MACHINED NOTCHES IN LARGE TENSILE SPECIMENS

R. W. Derby A. A. Abbatiello
K. K. Klindt

After the successful sharpening of the notch in the test block described in the preceding progress report,¹ the hydraulic pressure method was used to sharpen the machined notches in six full-size tensile specimens. The larger volume of the system associated with the full-size specimens required the use of a "pump and dump" arrangement as shown in Fig. 4.1. To implement this arrangement, hydraulic oil pressure at about 17,500 psi is pumped into the lines and voids of the machined notch and then released. A pressure gage in the main line is fitted with a set of electrical contacts so that each time the pressure reaches the set point the solenoid of the pressure relief valve opens and drops the

oil pressure to zero; the oil then recirculates to the supply container. An electric counter records each time the upper set point is energized. Using the 20,000-psi Aminco pump to pressurize the system, the required pressure, 17,500 psi, was reached in about two strokes when cycling only one specimen, with a consequent cycling rate of about 1000 pressure cycles per hour. When two specimens were connected in parallel, the added system volume required three pumping strokes to reach the 17,500-psi pressure level, which reduced the cycling rate to about 700 pressure cycles per hour. After shakedown problems were eliminated, a crack could be grown simultaneously in each of two specimens to the required depth (0.3 to 0.5 in.) in 30 to 36 operating hours, representing 20,000 to 25,000 cycles.

The rate of crack propagation was determined with ultrasonic instruments for four of the six specimens. Based on data previously collected on the two specimens which were sharpened at ORNL in September 1970, and broken at SwRI in October (chronologically Nos. 5 and 6), and which showed crack growths about 0.5 in. deep, there is good assurance that crack depth in the subsequent four specimens just completed will be in the same range.

A system of measuring deflection of the specimen as a function of pressure level and number of cycles was implemented as a complementary method of monitoring crack growth that generally confirmed the ultrasonic data. Difficulties with ultrasonic instrumentation prevented a direct comparison of the two techniques on the same set of specimens. However, the complementary techniques were employed on specimens that were cycled in parallel and consequently were exposed to the same pressure and number of cycles such that equivalent crack growth rates were anticipated. On this basis, it is believed that all four of the specimens delivered to SwRI on January 25, 1971, have cracks from $\frac{3}{8}$ to $\frac{7}{16}$ in. deep.

1. F. J. Witt, *HSST Program Semiannual Progr. Rep. Aug. 31, 1970*, ORNL-4653.

Fig. 4.1. Crack sharpening using controlled pressure.

The general arrangement used to fatigue two specimens simultaneously is shown in Fig. 4.2. Figure 4.3 is a closeup view of the seal showing the condition of the O-ring.

Based on the data obtained to date from sharpening six specimen notches, a useful prediction can be made of the crack size and its growth rate as a function of number of cycles and given pressure level. Experience has shown that for pressure levels of 17,500 psi a deflection of about 0.0025 to 0.0027 in. may be expected at the end of the specimen with respect to its center at the start of cycling, and when this deflection increases about 20% (or to a total deflection of about 0.0035 in.) the crack will have grown approximately 0.4 in. deep, a value considered desirable for this application. A curve showing this deflection for various pressure levels and number of cycles is shown in Fig. 4.4.

An extension of this crack-sharpening technique to the intermediate pressure vessels is being planned, but neither the through-transmission nor the pulse-echo method of ultrasonic crack detection appears to be sufficiently developed for use on large cylindrical sections. The shear-wave method has been proposed and may need to be evaluated. Mechanical sealing of the hydraulic pressure in the notches in the intermediate test vessels apparently presents no major problems. However, past experience has shown that flatness requirements (or actually seal clamp-specimen mini-

mum gap size) to seal the O-ring are quite severe and must be maintained to achieve a successful seal.

The crack grown in specimen 5 (see Fig. 4.5) showed a fine grain structure in the fatigued zone and a uniform growth of about $\frac{1}{2}$ in. along the entire periphery of the machined notch except for the region where the crack breaks out into the sealing surface plane, where it narrows down to about $\frac{1}{8}$ in. It is expected that similar fine cracks of uniform width will be found in specimens 7 through 10 except that No. 9 is predicted to be slightly deeper than the others. The ultrasonic measurements also indicate a flaw indication in No. 9 just below the machined notch. It will be interesting to see how good such evaluations can be in view of the extreme depth (almost 4 ft) of signal transmission.

TECHNIQUE FOR GENERATING SHARP CRACKS IN LOW-ALLOY HIGH-STRENGTH STEELS

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The successful culmination of the investigation of the behavior of the intermediate test vessels in the presence of a sharp flaw is dependent on the preparation of a flaw whose characteristics are acceptable to the application of fracture mechanics. To satisfy that need, a study was initiated on metallurgical principles. Below, we will describe briefly the metallurgical technique employed to produce sharp cracks in ASTM A 533 grade B steel.

PHOTO 73460

Fig. 4.2. Cracks in two large tensile specimens being sharpened at same time.

Fig. 4.3. Seal plate and O-ring plate immediately after opening.

Fig. 4.4. Crack growth vs number of cycles for large HSST tensile specimen.

Fig. 4.5. Fracture surface showing the crack growth of flawed tensile specimen 5, which was tested at 75° F.

The production of cracks in the low-alloy high-strength steel is premised on the role of hydrogen in the formation of cracks in ferritic materials. It is well established that hydrogen is responsible for underbead cracking in the heat-affected zone of ferritic weldments. Further, the cracking is associated with stresses and microstructure. Our study is based on the fact that a suitable microstructure (martensite) can be produced in these steels by rapidly cooling them from the austenitic temperature range. This is easily accomplished by an autogenous weld on the surface of a piece of steel. The second requirement is that of stress. The use of an electron-beam welder assures that the heat is concentrated in a small region. Concentrating the heat results in the residual stresses, which are of yield-point magnitude, being contained in a very narrow band. The third requirement for cracking is the presence of hydrogen. The hydrogen can be "pumped" electrically into the weld metal by masking off all of the test specimen except the weld. The weld is then placed in a 10% H_2SO_4 solution which serves as the electrolyte. A lead bar is immersed in the bath, and the system is wired so that the specimen is at the lower potential. A current density of about 0.5 A per square inch of unmasked area is sufficient to cause cracking in less than 2 hr. Figure 4.6 schematically illustrates the process.

We have successfully cracked ASTM A 533 grade B steel in thicknesses from about $\frac{1}{2}$ to 4 in. The thicker materials are more easily cracked because they provide a better heat sink and, consequently, the optimum microstructure and residual stress pattern.

The cracks that are prepared are quite similar to those produced by fatigue. They propagate through the weld metal and/or heat-affected zone and culminate at the end of the weldment. Figure 4.6 contains some typical cracks. We have successfully applied this process to the formation of cracks ranging from 0.075 to over $1\frac{1}{2}$ in. deep.

This procedure has been used in a study of scaled-down intermediate test vessel experiments. The cracks were placed in 3- and 6-in.-diam vessels with an excellent control over crack depth.

The technique has a number of attractive features. Among these are the possibility of cracking a specimen during testing, thereby simulating a pop-in condition. Further, a double-weld technique may provide a means for simulating a pop-in condition without having to produce the crack during testing. This would involve an electron-beam (EB) weld whose depth was equal to the desired pop-in distance. This first weld would be

post-weld heat treated (PWHT) to produce a material whose notch toughness was superior to that of the as-welded structure but poorer than that of the base metal. A second, considerably less deep weld would be placed within the confines of the weld metal of the first weldment. The second weld would be cracked; the PWHT of the first weld would assure that the crack did not propagate beyond the second weld.

TESTING THE 6-IN.-THICK FLAWED TENSILE SPECIMENS²

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As of March 1, 1971, ten large tensile specimens have been broken. Test parameters and basic data are summarized below.

Notch Design and Notch Sharpening Procedure

Criteria

A flaw from which fast fracture would initiate in a reactor pressure vessel would most likely be a propagating fatigue crack. An investigation was made to find an economical way to generate a fatigue crack or equivalent in the large tensile specimens which could also be used for the pressure vessels to be tested in the simulated service test phase of the HSST program. The basic shape of the experimental flaw was to be produced by machining, and the flaw was to be sharpened with the procedure developed.

The first two specimens tested contained semielliptical notches machined using the EDM (electrical discharge machining) process, with the notch in the second specimen slightly smaller than the notch in the first. The remaining four specimens contained part-circular flaws also machined using the EDM process. Notch dimensions are given in Fig. 4.7.

Stress corrosion

An attempt was made to develop a stress corrosion procedure to initiate sharp cracks at the root of each notch. A procedure involving maintaining a specimen at 250°F under a static tensile load and pouring NH_4NO_3 into the notch was investigated; however, tests of

2. Work carried out under UCCND Subcontract No. 3202 between Union Carbide Corporation and Southwest Research Institute.

PHOTO Y101804

22 X

22 X

100 X

Fig. 4.6. Illustration of the process employed in the production of hydrogen-induced cracks in low-alloy high-strength steels. The photomicrographs were taken from specimens cracked by this technique.

Fig. 4.7. Dimensions of the HSST 6-in.-thick flawed tensile specimen.

compact K_{Ic} specimens sharpened by the same procedure indicated that the static load required to generate the crack was too great; that is, residual stresses at the tip of the crack at the temperature at which the crack was generated affected the load required to fail the specimen in the transition temperature range and lower.

Electron-beam weld

It was found at Oak Ridge National Laboratory (ORNL) that cracks could be created in a small specimen by hydrogen embrittling an electron-beam weld located at the root notch and then applying a static tensile load for a short period of time. (See the previous section.) The resulting crack would pop out of the brittle region and arrest in the base metal at a sufficiently low load. This procedure was abandoned only because a large electron-beam welder was not conveniently available to ORNL for use on the large tensile specimens and intermediate test vessels. Present plans are to use this procedure on a series of one-sixth-scale models.

Fatigue

In order to expedite the tensile tests the first two large specimens were tested at 220°F without sharpened notches, as notch sharpness was not considered to be significant at this temperature. The third and fourth

specimens were sharpened by applying a three-point loading to the specimen for lack of a better method. The ends of the specimens rested on round bars set at a span of 42 in. while a load of 187.5 kips was applied at the center of the specimen, evenly distributed across the width on the unnotched surface. (A tensile stress was created on the notched surface.) Cracks initiated in the notch root after 15,000 cycles of loading and required a total of 35,000 cycles for the third specimen and 46,000 cycles for the fourth specimen to propagate the fatigue cracks to a sufficient depth.

The criterion that the notch sharpening procedure be applicable to the test vessels is not satisfied by the three-point loading method. Notches in the vessels could be sharpened by cycling internal pressure in the vessel, but this would be very time consuming. Therefore, a method of sharpening the notch cavity with very high locally applied pressure (about 20,000 pressure cycles at 15,000 psig were required) was developed at ORNL, and notches in specimens 5 through 10 were sharpened in this manner (see prior discussion in this chapter).

The fatigue cracks produced by pressurizing the notch cavity have proved to be more uniform in depth around the periphery of the notch than the cracks produced by bending the specimen. Also, fewer cycles are required to produce a deep fatigue crack using the pressure method. The fatigue cracks produced in specimens 4

and 5 are shown in Fig. 4.8. The depth of the fatigue crack in specimen 5 is about 0.5 in. The dimensions of the notches after sharpening with a fatigue crack are given in Table 4.1. One will note that the length b of the notch did not change significantly by either method.

Table 4.1 Notch dimensions before and after notch sharpening

Specimen No.	Machined notch dimensions (in.) ^a			Dimensions after sharpening (in.)		Dimensions at ultimate load ^b (in.)	
	a	b	r	a	b	a	b
1	2.375	6.598		Unsharpened			
2	2.160	6.000		Unsharpened		2.86	5.92 ^c
3	2.000	8.000	5.000	2.05	8.37	2.05	8.37
4	2.000	8.000	5.000	2.05	8.20	3.45	8.02 ^c
5	2.000	8.000	5.000	2.53	8.26	2.53	8.26
6	3.000	9.160	5.000	3.37	9.53	4.33	10.25
7	2.000	8.000	5.000	2.43	8.13	2.79	8.74
8	2.000	8.000	5.000	2.42	8.30	2.88	8.30
9	2.000	8.000	5.000	2.41	8.29	3.06	8.29
10	2.000	8.000	5.000	2.38	8.33	2.66	8.33

^a a = depth, b = length, and r = radius of curvature.

^bMeasurements made after fracture.

^cLateral contraction from necking in the net section masked small amount of crack growth that occurred at the surface. Dimensions estimated by paint markings which were applied near or after ultimate load.

Instrumentation and Test Procedure

The first two specimens tested were instrumented with over 100 strain gages, and the number of gages was reduced to about 50 on specimens 3 through 10. Six displacement gages were used on all but the first specimen. Acoustic emissions monitoring was performed on specimens 1, 2, 7, 8, 9, and 10. The acoustic emissions monitoring of the first two specimens was performed by Battelle-Northwest, and the last four were monitored by Southwest Research Institute.

Strain gage and displacement data were recorded on the B&F model 156, 100-channel data logger. This instrument will record on paper tape the channel number and strain to the nearest microinch per inch at a rate of 5 channels per second or to the nearest 10 $\mu\text{in./in.}$ at a rate of 20 channels per second. Therefore the actual recording of data required very little time, and the greatest time spent was pressurizing

the hydraulic jacks and examining data from key gages. Data were obtained in 100-kip increments above 5 million pounds and in 50-kip increments when the approximate failure load was approached. The average test required 2½ hr once the test was started.

Three thermocouples were used to monitor the temperature of the specimen. Two were located on the notched surface and one on the unnotched surface. All three thermocouples were located on the longitudinal axis of the specimen 2 in. from the plane of the notch.

Test Results

The discussion of test results will be limited to the nature of the fracture face, the total load, and the total strain. Considerable strain gage data were gathered during each test and will be presented in full in a technical summary report to be published.

Longitudinal specimens

Seven longitudinal specimens were tested at temperatures from 50 to 220°F. Some of the pertinent data obtained from these tests are presented in Table 4.2, along with the results of small specimen tests which were performed in support of this program to provide answers to questions on size effect and shape of the large tensile specimen. The elongation data presented in Table 4.2 were obtained from displacement gages attached to the full-size specimen with a 42-in. gage length. However, the grip, or thick section, of the specimen sees an insignificant amount of strain, and the gage length was considered to be 24 in. The same method was used on the model specimens as well. A summary of model test data appeared in the previous progress report.¹

The net section at or immediately after ultimate load was determined in a crude but fairly effective manner. If the specimen failed in a brittle manner or failed while the load-strain curve was positive, then crack growth, if any, was easily noted by the appearance of the fracture, and the size of the crack at fracture could be measured. If the load-strain curve began to flatten or a considerable drop in load occurred when pumping was stopped, the project engineer would spray the notch with paint. A paint of a different color would be applied if additional loading proved that ultimate load had not been reached. After failure, the size of the notch was determined by the paint marking. The paint markings on specimen 4 may be easily seen in Fig. 4.8.

Stress-strain curves for the various tests are presented in Fig. 4.9. Stress was computed by the conventional

Fig. 4.8. Results of two methods of notch sharpening used on the large tensile specimen. (a) Fatigue crack at the root of the notch produced by bending (specimen 4). (b) Fatigue crack at the root of the notch produced by bending (specimen 5).

Table 4.2. Tensile test data for 6-in.-thick specimens and model tests

Specimen No.	Test temperature (°F)	HSST plate No.	Gage length (in.)	Width (in.)	Thickness (in.)	Gross area (in. ²)	Net area	Elastic limit load (kips)	Elastic limit stress (ksi)		Ultimate load (kips)	Ultimate stress (ksi)		Average strain at ultimate (%)	Average strain at fracture (%)
									Gross section	Net section		Gross section	Net section		
Longitudinal Specimens															
1	215	03	24	18	6	108	95.69	5720	54.0	59.8	7300	67.5	76.3		8.50
2	220	03	24	18	6	108	97.82	6000	55.5	61.2	7550	69.9	77.2	3.83	9.60
3	50	01	24	18	6	108	95.79	4400	40.7	46.0	5300	49.1	55.4	0.24	0.24
4	100	01	24	18	6	108	95.50	6000	55.5	62.8	7500	69.4	78.5	4.10	7.90
5	75	01	24	18	6	108	92.33	5500	50.9	59.7	5900	54.6	63.9	0.35	0.35
6	100	01	24	18	6	108	81.97	4000	37.0	48.8	4700	43.5	57.4	0.48	0.48
7	100	03	24	18	6	108	92.43	6500	60.2	70.4	7050	65.2	76.2	0.88	0.88
1/6 model 2	210	03	4	3	1	3	2.66	175	58.4	65.7	230.6	78.6	88.9	5.61	10.5
1/6 model 3	50	03	4	3	1	3	2.42	176	58.6	72.7	223.0	74.3	92.2	5.75	10.9
1/12 model 03ED-2003	75	03	2	1.5	0.5	0.75	0.68	45	60.0	66.3	65.0	86.6	95.7	7.5	
1/12 model	75	03	2	1.5	0.5	0.75	Unnotched	49.0	65.4		64.8	86.4		10.2	38.9
1/12 model	200	03	2	1.5	0.5	0.75	Unnotched	47.5	63.0		63.4	84.0		10.0	32.4
Standard 0.505	75	03	2	0.505		0.20		13.5	67.5		17.3	86.6		11.5	31.5
Standard 0.505	200	03	2	0.505		0.20		12.6	63.0		16.5	82.8		10.2	26.0
1/48 model 03ED-2005	75	03	1	0.370		0.0451	0.0408	2.900	64.3	71.2	4.140	91.8	101.5		13.5
1/96 model 03ED-2004	75	03	1	0.176		0.0118	0.0107	0.700	58.2	65.4	0.992	84.1	92.6		13.5
Transverse Specimens															
8	100	03	24	18	6	108	92.55	6400	59.4	69.4	6800	63.0	73.5	0.74	0.74
9	220	03	24	18	6	108	93.30	5000	46.4	54.1	6675	61.8	71.6	2.00	3.60
10	130	03	24	18	6	108	93.15	6200	57.4	66.5	6650	61.6	71.4	1.19	2.69

Fig. 4.9. Stress-strain curves for the longitudinal 6-in.-thick flawed tensile specimens.

method of load divided by the gross section area. Strain was computed by dividing the displacement by the length of the test section as just described.

The data in Table 4.2 and Fig. 4.9 show both the marked reduction in strain with increase in size and the loss of ductility in the full-size specimen at 100°F.

Specimen 6 yielded at a lower load than the other specimens because the notch size was considerably greater. However, there is no explanation at this time for the lower yield strength of specimens 3 and 5.

The fracture surfaces of specimens 2 through 6 are shown in Fig. 4.10. Specimens 4 and 6 were both tested at 100°F but behaved quite differently. Specimen 7 was tested at 100°F in an attempt to determine which test result was typical. The fracture surface of this specimen, shown in Fig. 4.11, indicates that the low strain

fracture is typical. The ductile behavior of specimen 4 is still unexplained, and a study of the grain structures of all the specimens is being performed to see if a significant variation exists. It is noteworthy that specimens 4 and 6 were machined from HSST plate 01 and were only a few feet apart. Specimen 7 was machined from HSST plate 03.

Transverse specimens

Three transverse specimens were cut from HSST plate 03 and tested at 100, 130, and 220°F. The data for these tests are presented in Table 4.2, and stress-strain curves are shown in Fig. 4.12. One may note again the reduction of ductility with temperature. As shown in Fig. 4.12, the transverse specimens are less tough than

Fig. 4.10. Fracture surfaces of the longitudinal 6-in.-thick flawed tensile specimens.

Fig. 4.11. Fracture surface of longitudinal specimen 7 after testing at 100°F.

the longitudinal specimens tested at the same temperature.

The fracture surface of the three transverse specimens is shown in Fig. 4.13. The large shear lips on the 220°F specimen are not quite as predominant as those on the longitudinal specimen, and the 130°F specimen resembles specimen 4 tested at 100°F. The appearance of the fracture surface of specimens 7 and 8, both tested at 100°F, are almost identical.

One may note several expected trends in the data presented: (1) there is a marked reduction in ductility with size, (2) there is a significant temperature effect on ductility, and (3) the performance of the transverse

specimens is inferior to that of the longitudinal specimens.

The degree of effect of size, temperature, and rolling direction is somewhat masked by the variation in material properties within the HSST plates. It is difficult to evaluate the difference in performance between specimens 2 and 9, both tested at 220°F, when the yield strength of specimen 9 was significantly lower than the other two transverse specimens. There is also a possibility that the difference in performance between specimens 4 and 6 may be explained by material property variation; this is being investigated.

Fig. 4.12. Stress-strain curves for the transverse 6-in.-thick flawed tensile specimens compared with curves for longitudinal specimens.

Fig. 4.13. Fracture surfaces of the three transverse 6-in.-thick flawed tensile specimens.

The large variation in yield strength shown in Figs. 4.9 and 4.12 was reflected in strain gage readings on the gross section 9 in. from the plane of the notch. One would expect very little variation in material properties within the same plate as determined by these tests, by virtue of the averaging effect of the large cross section. The variation in yield strength as determined by the 6-in.-thick tensile test is an unexpected phenomenon which warrants further study.

Acoustic Emissions Monitoring

Test procedure and instrumentation

During the testing of specimens 7 through 10, two transducers simultaneously monitored areas near the notch of each specimen as shown in Fig. 4.14. Each transducer consisted of a disk of piezoelectric ceramic,

PZT-5, with a metallic housing acting as an electrical shield. The transducers were operated at a resonant frequency of 200 kHz. Because the piezoelectric efficiency of PZT-5 is nearly a constant at the test temperatures, the sensitivity of the transducers varied negligibly during the tests.

Typical instrumentation for monitoring acoustic emission for these tests is illustrated in Fig. 4.15; Fig. 4.16 is a schematic diagram of the experiment. The output from each transducer was fed into a high-input-impedance low-noise-level preamplifier through a short coaxial cable. A second-stage amplifier was used for each channel to achieve a gain for the overall system which could be varied from 60 to 90 db. Since the amplifiers were designed for general purpose with a wide frequency band, bandpass filters, acting in the frequency range from 100 to 300 kHz, filtered out most mechanical and electrical noises. The electric noise

Fig. 4.14. Two transducers mounted on a large flawed tensile specimen.

at the output of the amplifier was about 10 mV, and the maximum signal-to-noise ratio reached 60 db. The signals were then displayed on the oscilloscope, recorded on magnetic tape, and fed into a digital counter. Only those signals significantly above electric noise level are counted as acoustic emissions. The digital-to-analog converter changed the digital counts into analog signals, which were applied to the Y axis of an X-Y recorder. The X axis was reserved for the information obtained from a strain gage on the tensile machine. The strain gage was calibrated as a function of the load. The output of the strain gage was also recorded on an FM channel of the magnetic tape recorder simultaneously with acoustic emission signals recorded on the analog channels. The tape was then played back, and the number of counts and the amplitude of each acoustic emission pulse were studied as the function of load or other desired related parameters. All experiments were begun in late afternoon, and most of the data were obtained after 5:00 PM, when the laboratory was relatively quiet. Electric noise was also negligibly small. The transducers, operated at the selected frequency,

were normally insensitive to the noise of the large tensile machine with the anticipated exception of a friction force suddenly released from the machine pins. Since the friction force was accompanied by audible sound, the counts due to this force, or noise, can be easily traced and the resulting noise counts subtracted in the final data, as in the case for the data presented in the following section.

Results

Acoustic emission from metals during a tensile test is different for a flawed specimen from that for an unflawed specimen. Usually, the cumulative counts of acoustic emission from an unflawed specimen plotted vs strain resembles an ordinary stress-strain curve.^{3,4} The acoustic emission from a flawed specimen was studied as a function of the load or the stress intensity factor. The cumulative counts for a flawed specimen increase slowly at first, but as the load increases, the counts increase more and more rapidly until the specimen breaks.³ The data from four tensile tests of large specimens are presented in Fig. 4.17, showing plots of cumulative counts of acoustic emission vs load for each test. These data are similar to the curves for small tensile specimens presented in Ref. 3. The data for four different tests appear similar and overlap plotted on a linear scale, but they are quite distinguishable plotted on a logarithmic scale. It is also of interest to study the accumulation of acoustic emission as a function of the stress intensity factor. The stress intensity factor K was calculated in terms of stress and flaw dimensions. The notches were fabricated as a part of a circle with a radius of 5 in. The shape of the whole flaw after fatigue cracking is nearly a semielliptical curve with a semi-minor axis, $a \approx 2$ in., and a major axis, $2b \approx 8$ in. The stress intensity factor for the stress in these tests is given by⁵

$$K = \left[1 + 0.12 \left(1 - \frac{a}{b} \right) \right] \frac{\sigma(\pi a)^{1/2}}{\Phi} \left(\frac{2t}{\pi a} \tan \frac{\pi a}{2t} \right)^{1/2},$$

3. H. L. Dunegan, D. O. Harris, and C. A. Tatro, "Fracture Analysis by Use of Acoustic Emission," pp. 105-22 in *Engineering Fracture Mechanics*, vol. 1, Pergamon, Oxford, New York, London, and Paris, 1968.

4. E. R. Reinhardt and S. P. Ying, "Acoustic Emission from Irradiated Steels," to be presented at 8th Symposium on Nondestructive Evaluation in Aerospace, Weapons Systems, and Nuclear Applications, Apr. 21-23, 1971, San Antonio, Tex.

5. P. C. Paris and G. Sih, "Stress Analysis of Cracks," *Fracture Toughness Testing and Its Application*, ASTM Spec. Tech. Publ. No. 381, pp. 30-81, American Society for Testing and Materials, Philadelphia, 1965.

Fig. 4.15. Electronic equipment of acoustic emission system.

Fig. 4.16. Block diagram of electronics for acoustic emission detection.

Fig. 4.17. Accumulation of acoustic emission as a function of load.

where

$$\Phi = \int_0^{\pi/2} \left[1 - \left(\frac{b^2 - a^2}{b^2} \right) \sin^2 \theta \right]^{1/2} d\theta$$

is a complete cycle elliptical integral, and t is the thickness of the specimen (6 in.). Figure 4.18 shows the cumulative counts as a function of the stress intensity factor for the specimen, where the stress σ is parallel to the rolling direction. Figure 4.19 illustrates test results with σ perpendicular to the direction of rolling.

The cumulative counts N is a linear function of the stress intensity factor in a logarithmic-logarithmic scale, as shown in Fig. 4.18. The curves in Fig. 4.19 are not exactly straight lines but are nearly linear over most portions of the curves. Therefore, the data approximate the equation

$$N = AK^n,$$

where A and n are constants. The slope n of curves in logarithmic-logarithmic scales can thus be calculated

from the experimental results. Table 4.3 summarizes the calculated results.

The values of n for the specimens with transverse loading are plotted vs the test temperatures in Fig. 4.20.

Close examination of the curves in Figs. 4.18 and 4.19 reveals small plateaus in the region of 130 ksi $\sqrt{\text{in.}}$, corresponding to the load of 6×10^6 lb; that is, the counting rates are very small in that region. Further clarity is observed by plotting the counting rate in

Table 4.3. Values for the constant n for different test temperatures and specimen orientations

Direction of tension with respect to rolling direction	Test temperature (°F)	Slope n
Longitudinal or parallel	100	7.4
Transverse or perpendicular	100	2.4
Transverse or perpendicular	130	3
Transverse or perpendicular	220	4

Fig. 4.18. Acoustic emission from flawed specimen in which the direction of tension is longitudinal to the rolling direction.

Fig. 4.19. Acoustic emission from flawed specimens in which the direction of tension is transverse to the rolling direction.

Fig. 4.20. The temperature-dependent effect of acoustic emission.

counts per second vs the load. Figure 4.21 is a distribution of these counting rates which is typical for the four tests. Since a higher amplitude or a longer period of an acoustic emission pulse yields more counts, the counting rate is seen to be directly related to the amplitude and the number of counts of an emission pulse, as well as the number of pulses per unit time. Figure 4.21 also partly represents the number of counts and the amplitude of each acoustic emission pulse as a function of load. This distribution is consistent with the data reported by Jolly⁶ for specimen 2 tested at 215°F.

In general, the curves of cumulative counts vs load for the four tests have a similar shape in comparison with each other. The slope n in the logarithmic scale of each test, however, is distinguishably different and systematically varies for different directions of tensile loading with respect to the rolling direction and for various test temperatures. The nearly linear dependence of n on temperature for these large flawed tensile

6. W. D. Jolly, *Acoustic Emission Analysis-HSST #2*, Battelle-Northwest, Richland, Wash., April 1970.

Fig. 4.21. Distribution of acoustic emission.

specimens is similar to that of the precracked wedge-opening load (WOL) specimens.⁴ Therefore the dependence of acoustic emission behavior on temperature is obviously not accidental, but in common agreement with other flawed specimens.

PROCUREMENT OF INTERMEDIATE TEST VESSELS

C. E. Childress

Fabrication of forged components for the six 6-in.-thick intermediate test vessels is still in progress. These vessels are designed to serve as test specimens for demonstrating the capability to predict safe behavior of pressure-containing equipment having flaws of known dimensions under frangible, transitional, and tough temperature regimes. Each vessel will serve as a go, no-go determination of critical flaw size for a specific set of test parameters.

Most of the components for the six vessels, to be fabricated by the Taylor Forge Company, are completed. It is possible that assembly of some components could already be under way, except for the encounter of unusually high tensile properties in the shell courses. The tensile values for the shell courses ranged from 20 to 25% higher than the tensile values for the other components. It thus became necessary to delay actual fabrication of the vessels until this matter could be resolved. Recent experiments in retempering representative materials at higher temperatures (about 1320°F) indicated that the tensile properties of the shell courses could be lowered to the approximate values of the other components without a significant loss in toughness. This is discussed in more detail in later sections.

The completed vessels will appear essentially as shown in Fig. 4.22. The flat and hemispherical heads were made by Lenape Forge Company, and each of the forging operations used to make these items was described previously.¹ On completion of the forging operations the parts made by Lenape were (1) normalized at $1650 \pm 25^\circ\text{F}$, (2) austenitized at $1560 \pm 25^\circ\text{F}$, (3) quenched in agitated water, and (4) tempered at 1225°F minimum and air cooled. All parts were held at temperature for 1 hr per inch of part thickness.

Test specimens from each of the items were subjected to a simulated heat treatment in Lenape's Data-trak (programmed testing procedure). To obtain the cooling rate for the simulated quenching of test specimens, one each of the flat and hemispherical heads was equipped with a thermocouple and monitored during austenitizing and quenching. The cooling rates thus obtained

were applied to the test specimens during the simulated heat treating cycles.

The chemical analyses for the hemispherical and flat heads are given in Table 4.4. Partial results of the mechanical testing performed by Lenape on specimens from the flat and hemispherical heads are given in Table 4.5.

Table 4.4. Chemical analyses for the hemispherical and flat heads (wt %)

C	Mn	P	S	Si	Cr	Ni	Mo	V
Hemispherical heads, ladle analysis								
0.19	0.78	0.008	0.022	0.23	0.39	0.82	0.63	0.03
Flat heads, ladle and check								
0.19	0.78	0.008	0.022	0.23	0.39	0.82	0.63	0.03
0.197	0.75	0.019	0.023	0.22	0.41	0.87	0.59	0.01

The flat heads were inspected ultrasonically after all machining was completed. The parts were scanned using a Branson model Z-101-A, 1-in.-diam, 2.25-MHz transducer, and SAE 40W oil. The transducer was placed over a smooth flat area on the forging, and the back-reflection signal was adjusted to approximately 75% of full screen height. Scanning was performed from the flat side and from the circumference of each head; each scan path was overlapped 50% to assure complete coverage. Lenape's applicable inspection procedure designated as rejectable "... any indication accompanied by a complete loss of back reflection not associated with or attributable to geometric configuration" The loss of back reflection was no more than 40% in any instance.

The hemispherical heads were also inspected to the same ultrasonic procedure, and each was found to be acceptable.

The applicable inspection procedure specified that all components be subjected to magnetic particle inspection on completion of final machining. Taylor Forge Quality Control Manager recommended against such inspection since Lenape uses the contact prod method for magnetic particle inspection. In lieu of magnetic particle inspection, all parts were subjected to liquid penetrant inspection.

Details of the forging operations and the post-forge heat treatment of the ingots used to make the shell courses were described in Ref. 1. The ingot used to

Fig. 4.22. HSST intermediate test vessel.

make the closure flanges (transition sections) was forged and heat treated similar to the shell ingots, but there were some variations and these will be described.

The material for the closure flanges was melted and forged at National's Irvine plant. The reported ladle and check analyses for this heat, 9-8584, are given in Table 4.6.

The ingot from which the closure flanges was made was essentially a truncated cone and weighed 78,000 lb, including the hot top. The temperature of the ingot on

removal from the furnace (preparatory to forging) was about 2400°F; the temperature had dropped to 2200°F by the time forging began. Forging was completed in 40 min, at which time the temperature of the part was determined (by optical pyrometer) to be 1580°F. The completed forging was a solid round bar, 41½ in. in diameter and 12 ft long.

Shortly after forging was completed, the part was charged into a furnace and held at 800°F for 4 hr. The temperature was then raised at 100°F/hr until the part

Table 4.5. Some mechanical properties obtained on specimens from the flat and hemispherical heads

Lenape ID No.	Ultimate tensile strength (ksi)	Yield strength (ksi)	Reduction in area (%)	Elongation (%) in 2 in. gage length	C _v values at +10° F (ft-lb)
Hemispherical heads					
3	99.3	79.8	68	23	60-60-51
	96	78.7	68	23	47-51-52
4	95	78.7	67	23	74-49-71
	96	79.7	64	22.5	47-44-49
5	95.5	78	62	22.5	41-46-34
	95.6	78.7	62	22.5	28-31-36
6	96.5	80.5	67	22.5	28-31-31
	95	78	65	22.5	32-25-33
Flat heads					
1	90.5	70.9	66	23.5	82-78-65
	90	72	67	23.5	81-85-47
2	89	69.3	47.4	21.5	42-44-40
	92.5	74	62.5	23.5	36-31-37
3	91	71	56	24	38-41-39
	90.5	70.6	64	23.5	51-54-50
4	90	69	63	23.5	45-43-51
	90	70.3	65	23.5	39-35-52
5	90.1	73.3	62	22.5	35-22-38
	87.5	74.8	64.9	21.5	39-37-37
6	91.8	72.8	41.8	19	50-48-52
	89.2	70.4	49.3	20.5	53-54-47

Table 4.6. Chemical analyses for closure flanges (wt %)

	C	Mn	P	S	Si	Ni	Cr	Mo	V
Ladle	0.22	0.69	0.008	0.008	0.27	0.82	0.34	0.63	0.09
Check 1	0.21	0.70	0.008	0.008	0.27	0.82	0.34	0.62	
Check 2	0.20	0.69	0.008	0.008	0.26	0.80	0.33	0.62	

reached 1700°F and held at this temperature for 22 hr. It was then removed from the furnace and air cooled to about 300°F. With a furnace preheated to 650°F, the part was recharged and held at this temperature for 3 hr. It was then heated at a rate of 75°F/hr to 1260°F and held for 44 hr, furnace cooled at 50°F/hr to 1100°F, and then air cooled to ambient temperature. On completion of the post-forging heat treatment, the part was subjected to a preliminary ultrasonic inspection.

The rough forging was now ready for final heat treatment and cutting. The ingot was rough-turned to 39½ in. OD, and a 14½-in.-diam section was trepanned from the center essentially as shown in Fig. 4.23. In this form, the forging was subjected to final heat treatment prior to sectioning.

To determine the temperature profile throughout the length of the forging and the cooling rate during quenching, the part was equipped with nine Chromel-Alumel thermocouples, also shown in Fig. 4.23. Six of these were surface attached and three were imbedded. The couples were attached prior to normalizing, since the heat treatment cycle would continue uninterrupted.

The spread in thermocouple readings during the austenitizing heat treatment is shown in Fig. 4.24. It will be noted that the part was held in the austenitizing range for 1 hr per inch of metal thickness. All the heat treatments applied to this item (and all the parts heat treated at National) were performed in vertical furnaces. Quenching was also performed with the parts in the vertical position. Following the 10-hr hold period, the part was secured by the lifting lug (see Fig. 4.23) from an overhead crane, removed from the furnace, and quenched. About 50 sec were required from the time the furnace top was removed and the part completely immersed.

The inlet water piping to the quench tank was arranged to provide agitation of the quench medium. Additionally, a 4-in. IPS water line was installed along the floor of the tank so that the discharge was directed up through the bore of the forging. Throughout the 88-min quench period, the part was alternately raised and lowered to facilitate cooling. The water temperature, at the start of the quench, was 56°F, reached a maximum of 72°F in 25 min, and had returned to 58°F on completion of the quench.

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Fig. 4.23. Location of thermocouples in the forging used to make the closure flanges.

The cooling curves for the three imbedded couples are shown in Fig. 4.25 (Data for the surface-attached couples are not shown since these couples recorded essentially water temperature after 2 min in the quench.) On removal from the water, the surface temperature of the part was found to be about 150°F. The forging was not recharged for tempering until the furnace temperature had dropped below 400°F. This required about 4 hr. Figure 4.26 shows the spread in thermocouple readings during the tempering cycle.

Material for mechanical test specimens was secured from the prolongation near the bottom ingot end of the forging (see Fig. 4.23). The test material was removed after tempering and was subjected to a stress-relieving

heat treatment of $1125 \pm 25^{\circ}\text{F}$ for 30 hr and cooled per Paragraph N-532.3 of Section III of the ASME Code. The results obtained from the mechanical testing are:

Ultimate tensile strength (ksi)	Yield strength (ksi)	Reduction in area (%)	Elongation (%) in 2 in. gage length	C_v values at $+10^{\circ}\text{F}$ (ft-lb)
86	64	69	25	52-59-47
87	66			33.5-35-30

Six closure flanges will be made from this forging, and National has assigned them numbers 1 through 6, beginning at the bottom ingot end (excluding prolongation). Each piece has also been identified so that the

Fig. 4.24. Spread in thermocouple readings during austenitizing of the closure flange forging.

Fig. 4.25. Cooling curves for the imbedded thermocouples in the closure flange forging.

Fig. 4.26. Spread in thermocouple readings during tempering of the closure flange forging.

bottom ingot end can be distinguished from the top end. These pieces have been shipped to Taylor Forge in Paola, Kansas, who will be responsible for machining the raw stock into finished transition pieces.

The ingots used to make the shell courses (one ingot provided material for three shell sections) weighed about 200,000 lb. Each was poured simultaneously from two furnace heats at National's Erie plant. The ingot configuration was essentially a truncated cone 76 in. at the top end and 65 in. at the bottom by 12 ft long, excluding the hot top. It will be seen later that the shell courses have been numbered 4 through 9, although it was originally intended that the numbering sequence would proceed from 1 through 6. When it became necessary to remake one of the ingots, National decided to drop the originally assigned first three numbers and continue consecutively from 4 through 9. The ladle and check chemical analyses of the ingot comprising the first three shells (4-5-6, heats 3V0913 and IV-3809) are given in Table 4.7.

Details of the post-forge heat treatment, applied to the ingots from which the shells were made, were listed in Ref. 1. The heat treatment cycle was similar in many

respects to that which was applied to the six-piece multiple which was described earlier. On completion of the post-forge heat treatment the forging received a preliminary ultrasonic inspection, after which it was sawed into three pieces of equal length (a 15-in. prolongation was left on each end of each piece).

The parts were then shipped to National's Irvine plant for trepanning, final heat treatment, final machining, and inspection. Final heat treatment provided the shell courses at Irvine essentially consisted of the same procedure applied to the ingot described previously. One shell, National's No. 6, was equipped with six surface-mounted and three imbedded stainless-steel-sheathed Chromel-Alumel thermocouples (prior to normalizing) as shown in Fig. 4.27. The thermocouples were installed to determine the cooling rate during quenching and to determine the temperature profile throughout the length of the part during heat treatment, especially during tempering.

Curves showing the spread in thermocouple readings during the normalizing and austenitizing heat treatments are shown in Figs. 4.28 and 4.29. On completion of the 6-hr austenitizing heat treatment, the part was

Table 4.7. Chemical analyses for first three shells (wt %)

	C	Mn	P	S	Si	Ni	Cr	Mo	V
Ladle	0.26	0.75	0.010	0.014	0.26	0.81	0.45	0.61	0.05
Check	0.27	0.77	0.013	0.015	0.32	0.82	0.43	0.58	0.05
Check	0.27	0.77	0.013	0.013	0.33	0.81	0.44	0.59	0.05

Fig. 4.27. Location of thermocouples on shell 6.

Fig. 4.28. Spread in thermocouple readings during normalizing of shell 6.

lifted from the furnace with the handling device shown in Fig. 4.30. About 50 sec were required to lift the shell from the furnace and completely immerse it. The quenching details are essentially the same as those described previously for the closure-flange multiple. The temperature of the water immediately prior to

quenching was 52°F and on completion of quenching was 74°F.

Curves showing the cooling rate data for the three imbedded couples are given in Fig. 4.31. Again, cooling curves for the surface-mounted couples are not included since they recorded water temperature shortly after

Fig. 4.29. Spread in thermocouple readings during austenitizing of shell 6.

Fig. 4.30. Schematic drawing of handling device for quenching shells.

immersing. It will be noted that the couple imbedded at $\frac{1}{4}T$ (from the OD surface) cooled somewhat faster than the pseudosurface couple (located $\frac{1}{2}$ in. from the ID surface). It was suspected that the numbering of these two couples somehow became transposed during installation. However, National advised that they rechecked the installation after the tempering operation and confirmed that the arrangement and numbering sequence shown in Figs. 4.27 and 4.31, respectively, are correct.

The surface temperature of the part on removal from the water was about 125°F. The furnace cooled to about 300°F before the shell was recharged for tempering. The spread in thermocouple readings for the tempering cycle is shown in Fig. 4.32.

Shell courses bearing National Nos. 4, 5, and 6 were made from this ingot. Mechanical test material from prolongations from these shells was stress relieved at $1125 \pm 25^\circ\text{F}$ for 24 hr and cooled in accordance with Paragraph N-532.3 of Section III of the ASME Code. Test results for vessels 4, 5, and 6 are given in Table 4.8. The ladle and check chemical analyses for the ingot comprising vessels 7, 8, and 9, heats 3V924 and IV3828, are given in Table 4.9. Mechanical test results for specimens from shells 7, 8, and 9 are given in Table 4.10.

It will be noted that the tensile properties listed in the above tables are somewhat atypical of A508, class 2

Fig. 4.31. Cooling curves for the imbedded thermocouples in shell 6.

Fig. 4.32. Spread in thermocouple readings during tempering of shell 6.

materials. In fact, the yield values noted above are about the expected ultimate tensile strength of the welds which will be used in the intermediate test vessels. If the vessels are used in their present condition, it will require higher test pressures (to rupture the vessels) than originally anticipated. In turn, the result-

ant higher test pressures would place an added burden on the test equipment, which would require new hazards evaluations.

As discussed in following sections information was developed at ORNL which indicates that the strength properties of the shells could be downgraded by

Table 4.8. Mechanical properties for the first three shell courses

National ID No.	Ultimate tensile strength (ksi)	Yield strength (ksi)	Reduction of area (%)	Elongation (%) in 2 in. gage length	C_v values at +10°F (ft-lb)	NDTT (°F) P-3 drop weight
4	105.5	85	65.4	22	66-55-51	-20
	107.3	87	66.1	22.5	45-54-54	
5	107	86.5	66.3	22.5	44-60-64	-20
	107.5	87	67	22.5	50-53.5-53.5	
6	109.5	88	65.2	22	57-43.5-51.5	-20
	109.5	88	65.9	22.5	39-63-62.5	

Table 4.9. Chemical analyses for the second three shell courses

	C	Mn	P	S	Si	Ni	Cr	Mo	V
Ladle	0.27	0.87	0.007	0.012	0.32	0.74	0.34	0.64	0.04
Check	0.24	0.84	0.009	0.011	0.35	0.72	0.35	0.66	0.04
Check	0.25	0.86	0.009	0.012	0.35	0.72	0.34	0.65	0.04

Table 4.10. Mechanical properties for the second three shell courses

National ID No.	Ultimate tensile strength (ksi)	Yield strength (ksi)	Reduction of area (%)	Elongation (%) in 2 in. gage length (%)	C_v values at +10°F (ft-lb)	NDTT (°F) P-3 drop weight
7	106.25	85	65.9	23	57-40-44.5	-20
	103.5	82.5	66.5	22.5	48.5-50-48	
8	103	83	68.6	24	78-71-79	-20
	102.75	82	69	24	85-69-67	
9	107	87.5	67.9	23	62.5-60-55	-20
	107.5	88.5	67.7	23	38.5-72-74	

controlled retempering. It will be noted from Fig. 4.32 that the tempering temperature for the shells was about 1280°F. maximum. Hence in consideration of retempering there was concern for the lower transformation temperature (AR_1), which is thought to lie between 1325 and 1350°F. Using materials from shell prolongations and Data-trak testing procedures, the data developed in the next two sections justified retempering a single shell course at about 1320°F, maintaining as little deviation as possible.

Shell 9 was retempered by National Forge on February 2 and 3, 1971. The part was equipped with seven

thermocouples as shown in Fig. 4.33. The heatup period comprised some 14 hr, after which couple 4 (located on the ID surface) was reading about 1295°F; the remaining couples were recording in the vicinity of 1310°F. At the end of the 6-hr hold period each recorded about 1320°F, except couple 4, which recorded a temperature of 1310°F.

On completion of the hold period, the part was cooled in the furnace at about 200°F/hr to 600°F. Material for the mechanical test specimens was removed from the shell prolongation and stress relieved at 1125 ± 25°F for 24 hr. The mechanical test results for the

Fig. 4.33. Location of thermocouples on shell 9 (to be retempered).

retempered shell are given below.

Ultimate tensile strength (psi)	Yield strength (psi)	C _v values at 10° F (ft-lb)
91,500	72,000	66-40-33
92,250	71,500	62.5-79-55.5

The mechanical test results noted above are quite encouraging and tend to substantiate the data discussed in the next two sections. It thus appears feasible that the mechanical properties of the shells can be lowered by retempering. However, before the final order is issued to retemper the remaining shells, additional metallographic studies will be made on specimens from shell 9.

RETEMPERING STUDY OF HSST INTERMEDIATE TEST VESSELS

R. G. Berggren W. J. Stelzman

Tensile and Charpy V-notch impact tests were performed on a portion of a 26-in.-ID (nominal) by 6 5/8-in.-thick wall vessel shell fabricated by the National Forge Company and identified as shell course 5. The vessel material (ASTM A 508, class 2) had been normalized (6 hr at 1725°F), austenitized (6 hr at 1580°F),

water quenched, and tempered (6 hr at 1270°F). An end section of this shell was removed by the vendor, who retained a portion for test purposes and sent the remainder to ORNL. The vendor's portion was stress relieved (24 hr at 1125°F) prior to fabricating their test specimens to simulate the final vessel heat treatment. The ORNL portion was not stress relieved. A part of the material was machined into tensile and Charpy blanks of axial orientation from the one-quarter to one-half thickness levels and retempered primarily in the temperature range of 1300 to 1345°F. Retempering of the blanks was performed in a Research Inc. thermal cycle simulator (Data-trak) at prescribed heating and cooling rates with a hold or soak period of 6 hr. The specimens were then machined with the notches located at the same position as the thermocouples used for recording the thermal history. Table 4.11 and Fig. 4.34 present the tensile results of these tests using 0.505-in.-gage-diam specimens. Vendor and ORNL tensile data have been presented to indicate the conditions existing prior to the retemper. One specimen tested was subjected to a high heating and cooling rate by plunging a blank into a large furnace at the retemper temperature, soaking for 1 hr, and then removing and cooling in still air. The 200°F/hr heating and cooling rates used in the study were determined from thermal data obtained by the vendor at various locations in and along the shell

Table 4.11. Effect of retempering on the tensile properties of intermediate vessel course 5 material (ASTM A 508, class 2)

Depth in vessel wall	Specimen orientation	Retemper treatment ^a				Tensile properties ^b			
		Heating rate (°F/hr)	Hold period		Cooling rate (°F/hr)	Yield stress (ksi)	Ultimate stress (ksi)	Total elongation (%)	Reduction of area (%)
			Time (hr)	Temperature (°F)					
Vendor data									
	Axial	No retemper, original temper 6 hr at 1270°F				86.5	107.0	22.5	66.3
	Axial	No retemper, original temper 6 hr at 1270°F				87.0	107.5	22.5	67.0
ORNL data									
1/4t	Tangential	No retemper, original temper 6 hr at 1270°F				92.2	109.0	19.4	52.2
3/4t	Tangential	No retemper, original temper 6 hr at 1270°F				91.1	108.9	18.1	38.6
1/4t	Tangential	c	1	1300	c	83.9	102.3	22.2	56.8
3/4t	Tangential	200	6	1312	200	73.7	92.9	22.5	57.4
1/4t	Axial	200	6	1596	20	60.0	89.4	24.4	46.2
1/4t	Axial	200	6	1303	200	78.7	96.3	25.0	70.0
1/4t	Axial	200	6	1331	200	69.2	95.8	26.0	66.9
1/4t	Axial	200	6	1300	200	75.6	96.1	25.0	69.7
1/3t	Axial	200	6	1346	200	65.3	96.0	22.5	66.5
1/3t	Axial	200	6	1338	200	66.7	95.1	26.5	65.3

^aRetemper was performed using a Research Inc. thermal cycle simulator.^b0.505-in.-gage-diam specimens.^cAmbient specimen was inserted into a large furnace at 1300°F, held for 1 hr, removed, and allowed to cool on firebrick in still air.

Fig. 4.34. Effect of retempering on the tensile properties of intermediate vessel course 5 material (ASTM A 508, class 2).

course during the tempering operation. The 6-hr hold period coincided with the vendor's hold period. Another specimen was cooled very slowly from the austenite range to produce the lowest possible strength. After a 6-hr soak at 1596°F, the specimen was cooled at 20°F/hr. These conditions should provide coarse-grained material with poor strength properties, thereby providing the lower strength datum for the retemper program. The tensile data in Table 4.11 indicate that both the yield and ultimate strengths show a 15% drop between the vendor's 1270°F temper and the 1310°F retemper, with a more gradual reduction in strength to the 1345°F limit of the study. Extrapolation to the 1596°F conditions would indicate that most of the reduction of strength had occurred before the 1345°F retemper. Ductility properties indicate little change in elongation and decrease in the reduction of area, especially for the axial oriented specimens, as the retemper temperature was increased.

The Charpy V impact results are depicted in Figs. 4.35 and 4.36, which describe the fracture energy, fracture appearance, and lateral expansion variations with the fracture temperature. Two retemper temperatures were investigated, 1315 and 1340°F, using heating and cooling rates of 200°F/hr and a hold period of 6 hr

Fig. 4.35. Effect of retempering on Charpy V-notch impact properties of HSST intermediate test vessel 5.

at the retempering temperature. All specimens were notched perpendicular to the vessel surface to produce crack propagation in the circumferential direction. All three parameters indicate curve shifts to higher temperatures, indicating a loss of impact properties or toughness. It is difficult to evaluate the degree of shift due to the inherent scatter of the data; therefore the same parameters were investigated at three fixed fracture temperatures. Figure 4.37 shows the results of this study. The shift of fracture at a given energy to higher temperatures is again apparent by the decrease in property as the retemper temperature was increased. The shift is exaggerated in the 40°F data since the "knee" of the temperature-energy curve is involved. Small curve shifts in this region cause large increases or decreases in energy depending on the direction of shift. The vendor's Charpy V (C_v) values at 10°F and at two locations after tempering at 1270°F were 44-60-64 and 50-53.5-53.5 ft-lb. These values are within reasonable agreement with the ORNL C_v data for the 1315°F retemper, indicating that some toughness will be sacrificed by retempering this material above 1315 to 1320°F. However, retempering at 1340°F results in properties that still appear acceptable for the intermediate test vessel studies.

Fig. 4.36. Effect of retempering on Charpy V-notch impact properties of HSST intermediate test vessel 5.

Fig. 4.37. Effect of retempering on Charpy V-notch impact properties of HSST intermediate test vessel 5.

THE METALLURGICAL ASSESSMENT OF RETEMPER TEMPERATURE ON THE INTERMEDIATE VESSEL TEST COURSES

D. A. Canonico

The design and test parameters for the cylindrical body forging that are to be used in the fabrication of the intermediate test vessels were based on the mechanical properties typical for ASTM A 508, class 2 steel. This specification requires that the minimum yield strength and ultimate tensile strengths be 50,000 and 80,000 psi respectively. Further, a toughness requirement of an average of 30 ft-lb at +10°F was also imposed on the material. The cylindrical body forgings supplied for the vessels had yield and ultimate tensile strengths in excess of 82,000 and 102,750 psi respectively. Moreover, the nil-ductility temperature of the steel as determined by the drop-weight test was below -20°F. The properties of the steel are acceptable as far as the ASTM A 508, class 2 specification is concerned. However, the high tensile values are cause for concern when one anticipates that the ultimate tensile strength of typical submerged-arc welds may be equal to or below the yield strength of the cylindrical body forging.

A review of the program objectives resulted in the conclusion that the cylindrical body forgings should possess mechanical properties within the following ranges which became the guidelines for the acceptance of the forgings.

Ultimate tensile strength, psi	80,000-95,000
0.2% offset yield strength, psi	50,000-75,000
Elongation in 2 in. (min.),%	18
Reduction in area (min.),%	38
Charpy impact energy at 40°F, ft-lb	
Average of set of three specimens	30
Minimum value of one specimen	25

The objective of this study was to determine what reheat treatment methods could be applied that would provide the cylindrical body forgings with mechanical properties that fall within the values set above.

The first investigation was undertaken to assure ourselves that the tensile properties reported initially were correct. This was done, and our results showed yield and ultimate tensile strengths of about 91,000 and 109,000 psi respectively. Moreover, reheated specimens at 1280°F resulted in no change in the tensile properties. Chemical analyses of the steel conducted by ORNL and a steel company were in agreement with the analysis reported by the vendor for all elements except carbon. The ORNL analysis indicated a carbon range of from 0.29 to 0.30%; the carbon analysis reported by

the steel company was 0.27%. This value agrees with the vendor's carbon analysis and is within the allowable range of the ASTM specification.

A sulfur print of the steel showed it to be acceptable. (The cylindrical shape was machined from the original forge material, and hence the "center-line" stock was removed. This resulted in a uniform sulfur distribution across the 6 $\frac{7}{8}$ -in. wall thickness.)

A reheat treatment study was undertaken in an effort to determine what avenues were available to reduce the mechanical properties to within those ranges previously stated. The original heat treatment of the forgings was as follows: normalize 6 hr at 1700°F, austenitize 7 hr at 1580°F, quench in water, and temper 8 hr at 1270°F.

The heat treatment was witnessed by C. E. Childress as discussed previously. Thermocouples were imbedded in the cylindrical shell forgings; consequently, excellent cooling rate records are available. The cooling rate to half-temperature at the quarter-thickness location was about 2.5°F/sec. Our initial response was to reduce this rate by requeenching the steel in a medium whose H value (a measure of the severity of quench) was lower. This approach was cursorily investigated by placing a $\frac{3}{4}$ -in.-square by 6-in.-long piece of steel in a furnace and heating it to 1600°F. It was held at temperature for 1 hr and the furnace was shut down. The piece cooled in the furnace at a rate of 0.07°F/sec, which is 36 times slower than the rate for the forgings. The hardness value of the slowly cooled steel was 16 R_c vs 20 R_c for the quenched-and-tempered forging. The change in cooling rate did not appear to provide an easy means for reducing the strength. Further, the microstructure of an extremely slowly cooled ($\sim 0.006^\circ\text{F/sec}$) steel is even more atypical of ASTM A 508, class 2 than the structure responsible for the high tensile properties of the forgings.

We then sought to lower the strength by retempering. A heat-treatment study was undertaken, and a change-in-thickness hardness survey was made on the as-received steel. These R_c values ranged from 23 on the surface to 19 at a depth of 4 $\frac{3}{8}$ in. in from the outer surface. The hardness values for the various depth locations are given in Table 4.12. A number of samples were taken from the steel and subjected to various retemper temperatures. Following the heating cycle, the specimens were cooled in air. The results of this study are also given in Table 4.12.

Retempering at 1220°F resulted in no apparent change in hardness. A retemper at 1280°F resulted in a similar observation. At 1300°F, there was a small drop, about 2 R_c hardness points, within the steel, but the surfaces were essentially unchanged. Retempering at

Table 4.12. Effect of heat treatment on the hardness of intermediate test vessel body forging

Through the thickness location ^a 1/16 in.	Hardness, Rockwell C				
	As received ^b	Retemper ^c (°F)			
		1220	1280	1300	1325
1	23				
2	22		23.5		
6			24		
8	22	24	22.5		
10	21			17	
13	21		22.5		26
17			21	21	28
20		22			
24			20		
26	22			17	
28			21		
30	20			19	
34	20		20.5		
36					23.5
38	21		21		31
42		21	21.5		
43	20.5				
47	21.5		20.5		30
51	21.5			17	
66			21	17	
70	19		19.5		
72					25
76	20.5		21		
78					25
80			19.5		
82	20.5			20	
86			20	21.5	25
88	19.5				
90			21		
92	20.5				
95			23		
98			24		
104	23		25		
106				22	

Note: Specimen retempered at 1325°F was reheated to 1280°F. Hardness ranged from 12 to 18 R_c .

^aInner wall, 0 in.

^b1580°F austenitize, water quench, 1280°F temper.

^cSpecimen quenched after retemper.

1325°F resulted in a hardness increase. It is assumed that the steel began to transform to austenite, and upon air cooling, the $\frac{3}{4}$ - by $\frac{3}{4}$ -in. section hardened.

Concurrent with the hardness study, we had an electrical resistivity study made to determine the eutectoid temperature (the temperature at which the steel begins to undergo an allotropic transformation from body-centered cubic to a face-centered cubic

crystal structure). The resistivity measurements indicated that the transformation occurred between 1315 and 1346°F (713 and 730°C). This value is in good agreement with the change in hardness observed after the 1325°F retemper treatment.

We also conducted a study of the hardenability of the body forging steel using the Jominy End Quench Test (JEQT). This test allows us to assess the effect of cooling rate on the hardness of a steel. The results of the JEQT are shown in Fig. 4.38. A study of austenitizing temperature was conducted, and the results show no effect over the 200°F temperature range employed. The JEQT results are typical for this steel, with a carbon level of about 0.25 to 0.30%.

The effect of tempering temperature was investigated by conducting a conventional JEQT on steel that had been austenitized at 1600°F and then tempering the specimens at various temperatures. Three temperatures (1280, 1310, and 1340°F) were studied. These results are also shown in Fig. 4.38. The 1340°F temper resulted in an R_c hardness level of about 20. This value is higher than what can be expected in the $\frac{6}{16}$ -in.-thick forging because the slowest cooling rate in the JEQT is faster than that of the $\frac{1}{4}$ -T location in the $\frac{6}{16}$ -in. forging. This study does indicate, however, that heating the steel above the eutectoid temperature did not result in a dramatic change in hardness properties.

These data served as the basis for a more extensive Data-trak reheat treatment study. This work is reported by Berggren and Stelzman in the previous section. Their data show a limiting lower value of ultimate tensile strength of about 90,000 psi. (This is based on a cooling rate of 20°F/hr, 0.006°F/sec.) Their study included Charpy V-notch as well as tensile properties. Figure 4.39 contains photomicrographs of specimens that were retempered in the Data-trak at the indicated temperatures. Following the temper cycle, the specimens were cooled at 200°F/hr, a rate which simulates an air cool. The microstructures of the as-received specimen and the specimens retempered in the Data-trak at 1298, 1311, and 1318°F are quite similar. Retempering at 1343°F resulted in a structure in which the amount of carbide appears to be reduced. The difference, however, is nowhere near the magnitude exhibited by the surface and $\frac{1}{4}$ -T locations. These microstructures are dramatically different and yet represent a hardness variation of only about three R_c numbers.

This similarity in microstructure, combined with the results of the reheat treatment studies, resulted in our suggestion that the desired mechanical properties could be obtained by retempering the forgings at about

Fig. 4.38. Effect of cooling rate and tempering temperature on hardness of ASTM A 508, class 2 steel.

1320°F. The vendor felt that his furnaces were controllable within $\pm 15^\circ\text{F}$ in the temperature range of interest. Our data indicated that a retemper at $1320 \pm 25^\circ\text{F}$ would provide acceptable mechanical properties. One cylindrical shell forging was tempered at 1320°F as reported in a previous section by Childress. The results of heat treatment on the microstructure of the forging is shown in Fig. 4.40, which includes the microstructure from the original 1280°F temper for comparison. There is very little difference in the two microstructures.

Retempering the ASTM A 508, class 2 forgings (NF heats 3V924, IV3828, 3V913, and IV3809) at $1320 \pm 25^\circ\text{F}$ should provide a steel that is within the range of mechanical properties that have been set for the cylindrical shells to be used in the fabrication of the intermediate test vessel. It appears that the carbon content of the forgings (about 0.27%) sets the mini-

mum tensile values that are possible for this steel. The fastest cooling rate that is possible in these thick sections is rather slow and results in what, for these steels, can be compared with an austemper heat treatment. The slow rate permits the carbides to agglomerate, and hence tempering, in the classical sense, is not too effective. The small effect is shown in the tempered JEQT specimens, wherein a 60°F difference (1280 to 1340°F) resulted in a hardness change of only three R_c numbers. Retempering at a temper above the eutectoid results in the transformation of areas of the steel to austenite. Upon cooling, these areas retransform. The cooling rate, however, which is in the range of 200°F/hr , results in a microstructure not too different from that observed in the center of the thick plate. This conclusion is supported by our Data-trak studies at 1343°F .

Austenitized at 1580°F – Quenched in Water

Retemper Study

Fig. 4.39. Photomicrographs of specimens retempered in the data-trak at specified temperatures.

Austenitized at 1580°F

Tempered at 1270°F

Tempered at 1320°F

Fig. 4.40. Photomicrographs of the steel after retempering at 1320°F compared with the results obtained from the original temperature at 1270°F.

SPECIFIC SAFETY RESEARCH PROBLEMS

This activity includes research oriented toward specific safety problems arising within the nuclear power industry and falling within the general framework of HSST research. Activities included in this category are (1) the emergency core coolant problem (thermal shock problem) and (2) the significance of transition temperature tests to fracture toughness. Item 1 was summarized in Ref. 7, and a final report is being written. The initial portion of item 2 was a study of the correlation of Charpy impact data with fracture toughness. This study has been completed and is summarized below.

Relationship between Material Fracture Toughness Using Fracture Mechanics and Transition Temperature Tests⁸

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For pressure vessel steels, the variation with temperature of K_{Ic} and Charpy V-notch energy (C_v) are similar, consisting of a low-temperature "shelf" region in which the toughness is low and increases modestly with temperature, a transition region in which both K_{Ic} and C_v increase markedly over a narrow range of temperature, and a high-temperature "upper shelf" region where both K_{Ic} and C_v are high. Toughness on the upper shelf may reach a peak between 200 and 400°F and decrease modestly at higher temperatures.

Beyond these similarities in the behavior trends, several differences exist between the two types of tests that make numerical conversion from one toughness value (C_v) to the other (K_{Ic}) uncertain. First, the fracture mechanics K_{Ic} provides a measure of toughness at the onset of rapid crack extension, whereas the Charpy (and dynamic tear) test measures the total energy to fracture the specimen. Second, the crack tip constraint is maintained nearly constant in the K_{Ic} specimen by requiring specimen thicknesses that increase with plastic zone size and toughness, whereas in the Charpy test the specimen thickness is maintained constant. Third, the K_{Ic} test induces a slow strain rate just prior to onset of rapid crack extension where the crack tip strain rate throughout the impact test is high. Finally, fatigue crack sharpness is employed in K_{Ic}

specimens, whereas the Charpy specimen has a blunt notch with a radius $\rho = 0.010$ in.

In the high-temperature (upper-shelf) region, where extensive crack tip plastic deformation precedes fracture, the influence of these differences between tests appears to be minimized. The Rolfe-Novak *upper shelf* correlation between K_{Ic} and C_v was reviewed and appeared favorable in the light of additional data obtained by Westinghouse on the upper shelf for a forging steel. It is noted that all the data employed to generate this K_{Ic} - C_v correlation are for steels with yield strengths above 100 ksi. Further, it is noted that estimates of K_{Ic} on the upper shelf for A 533 B steel (HSST material) made using this correlation are lower than the estimates made by F. J. Witt, using an equivalent-energy method. If K_{Ic} is to be estimated from Charpy V-notch energy values, it is essential that a few valid K_{Ic} measurements be made on the upper shelf for an A 533 B steel.

The low-temperature region and a portion of the transition-temperature region are considered together because both K_{Ic} and C_v data for pressure vessel steels are available in this temperature range.

For unirradiated material, the most constant trend type correlation was found when dynamic fracture toughness K_{Id} was compared with C_v data. A number of the Charpy specimens had fatigue-sharpened notches; however, the trend remained essentially the same for blunt notched Charpy specimens also. For irradiated material, the concept of shifting the K_{Id} curve along the temperature axis by an increment ΔT equal to the shift in the Charpy curve was considered. While no data were found to dispute this procedure, an assumption inherent in this concept, along with a lack of adequate data to completely justify the procedure, is noted.

For unirradiated material, the Barsom-Rolfe transition temperature K_{Ic} - C_v correlation was studied using pressure vessel steel data. A modification given by

$$K_{Ic} = 15,500 \sqrt{C_v}$$

is proposed, where K_{Ic} is in psi $\sqrt{\text{in.}}$ and C_v is in ft-lb. This correlation is limited to the range $5 \text{ ft-lb} \leq C_v \leq 50 \text{ ft-lb}$.

This modified correlation is introduced because in the range 5 to 50 ft-lb, the K_{Ic} pressure vessel steels consistently fall below the Barsom-Rolfe estimate. At higher toughnesses the pressure vessel steels tend to follow the Barsom-Rolfe trend; however, uncertainty exists concerning the accuracy of high Charpy energy values.

7. F. J. Witt, *HSST Program Semiannual Progr. Rep. Aug. 31, 1971*, ORNL-4512.

8. Work carried out under UCCND Subcontract No. 3398 between Union Carbide Corporation and the University of Illinois.

By use of the modified C_v - K_{Ic} correlation, K_{Ic} is estimated for unirradiated pressure vessel steel plates, forgings, and welds. The correlation consistently predicts the K_{Ic} values from C_v data for A 533 B plate material at $1/4$ and $1/2$ T. For welds, and possibly forgings, the estimated K_{Ic} values are lower than the measured values of K_{Ic} . For A 533 B plate surface material, the estimated K_{Ic} values are higher than the measured ones.

For irradiated material (A 533 B plate) the modified K_{Ic} - C_v correlation provides an estimate of K_{Ic} higher than the measured values of K_{Ic} for 10 of 11 data points. Again the surface material produces the maxi-

mum deviation between the estimated and measured K_{Ic} values.

Where K_{Ic} data are not available and K_{Ic} must be estimated from Charpy V-notch energy data, the K_{Ic} - C_v correlations can provide guidance if used cautiously. If such correlations are to be used where important safety questions are involved, considerably more data are required before the correlations and their limitations are established. It is recommended that either the Charpy specimen be modified or another specimen be selected as a basis for toughness measurements with pressure vessel steels.