

# Fracture Toughness of Alloy 600 and an EN82H Weld in Air and Water

W.J. MILLS and C.M. BROWN

The fracture toughness of alloy 600 and its weld, EN82H, was characterized in 54 °C to 338 °C air and hydrogenated water. Elastic-plastic  $J_{IC}$  testing was performed due to the inherent high toughness of these materials. Alloy 600 exhibited excellent fracture toughness under all test conditions. While EN82H welds displayed excellent toughness in air and high-temperature water, a dramatic toughness degradation occurred in water at temperatures below 149 °C. Comparison of the cracking response in low-temperature water with that for hydrogen-precharged specimens tested in air demonstrated that the loss in toughness is due to a hydrogen-induced intergranular cracking mechanism. At loading rates above  $\sim 1000 \text{ MPa}\sqrt{\text{m/h}}$ , the toughness in low-temperature water is improved because there is insufficient time for hydrogen to embrittle grain boundaries. Electron fractographic examinations were performed to correlate macroscopic properties with key microstructural features and operative fracture mechanisms.

## I. INTRODUCTION

ALLOY 600 and its weld, EN82H, are used in commercial primary water reactors (PWR), where they are exposed to water that can lead to environmental cracking. Alloy 600 components that have failed due to environmental cracking include piping, pressurizer heater-sleeve penetrations, instrument nozzles, reactor vessel closure-head penetrations, mechanical plugs, and steam-generator tubes.<sup>[1,2,3]</sup> Failure typically involves crack initiation and propagation as a result of stress corrosion cracking (SCC) or corrosion fatigue, followed by stable and unstable tearing when a critical crack size is reached. This study focuses on the final step in the cracking process, involving stable and unstable tearing.

While nickel-based alloys such as alloy 600 and EN82H welds possess excellent toughness in air, a preliminary study<sup>[4]</sup> indicates that their fracture properties are degraded in low-temperature water. Hence, the objective of this study is to characterize the effects of low- and high-temperature water on the fracture resistance of alloy 600 and EN82H welds. Fracture properties were characterized using elastic-plastic  $J_{IC}$  concepts, due to the ductile response exhibited by these materials. Scanning electron microscopy (SEM) examinations of fracture surfaces were performed to correlate trends in fracture-toughness behavior with operative cracking mechanisms and key microstructural features. The role of hydrogen embrittlement in degrading fracture properties in the low-temperature regime was evaluated by comparing the cracking behavior in low-temperature water with that for hydrogen-precharged specimens tested in air.

## II. MATERIALS AND EXPERIMENTAL PROCEDURE

### A. Test Materials

The alloy 600 test material consisted of a 50.8-mm-thick plate annealed at 982 °C and air cooled. The EN82H welds

were fabricated using a manual gas tungsten arc (GTA) weld process to join 25.4- and 50.8-mm-thick alloy 600 plates or 50.8-mm-thick plates of alloy 600 and EN82-clad A516 grade-70 steel. All plates had a 25 deg weld preparation. Alloy 600-to-alloy 600 welds were full-penetration welds (Figures 1(a) and (b)), and alloy 600-to-clad steel welds were partial-penetration welds (Figure 1(c)). Welds were fabricated by three manufacturers (*i.e.*, “A,” “B,” and “C”) using either a pure argon shield gas (welds A1, A2, B1, C4, and C5) or a 98 pct/2 pct or 95 pct/5 pct argon-hydrogen shield gas (welds C1, C2, and C3) to improve weldability. All welds were tested in the as-welded condition. Chemical compositions and mechanical properties for the test materials are provided in Tables I and II.

Representative microstructures for the test materials are shown in Figure 2. The alloy 600 plate has a duplex grain size, where fine grains with a mean grain diameter of  $\sim 20 \mu\text{m}$  surround isolated unrecrystallized grains with a mean diameter of  $\sim 120 \mu\text{m}$ . Intergranular and intragranular  $\text{M}_7\text{C}_3$  carbides are observed in both the fine- and coarse-grain regions, and there is evidence of coarse MC-type inclusions. This heat also exhibits a pronounced hot-deformation substructure with dislocation subgrains.<sup>[5]</sup>

The EN82H welds have well-defined grain boundaries separating colonies of similarly oriented dendrites. The grain structure is anisotropic, particularly in the weld root region, where columnar grains are aligned in the transverse direction. Grain boundaries are decorated with fine niobium and titanium-rich MC carbides. Small- and intermediate-size titanium nitrides are also observed along grain boundaries and in the interdendritic regions.

### B. Fracture-Toughness Testing

Fracture tests were performed on precracked compact tension (CT) specimens that had a width ( $W$ ) of 30.5 mm and thickness ( $B$ ) of 15.2 mm. All EN82H specimens and most alloy 600 specimens had 20 pct side grooves; a few alloy 600 specimens were not side grooved. Weld specimens were tested in either the longitudinal orientation, with the notch parallel to the welding direction (Figure 1(a)), or in the

W.J. MILLS, Consultant, and C.M. BROWN, Senior Engineer, are with Bettis Atomic Power Laboratory, West Mifflin, PA 15122.  
Manuscript submitted October 17, 2000.

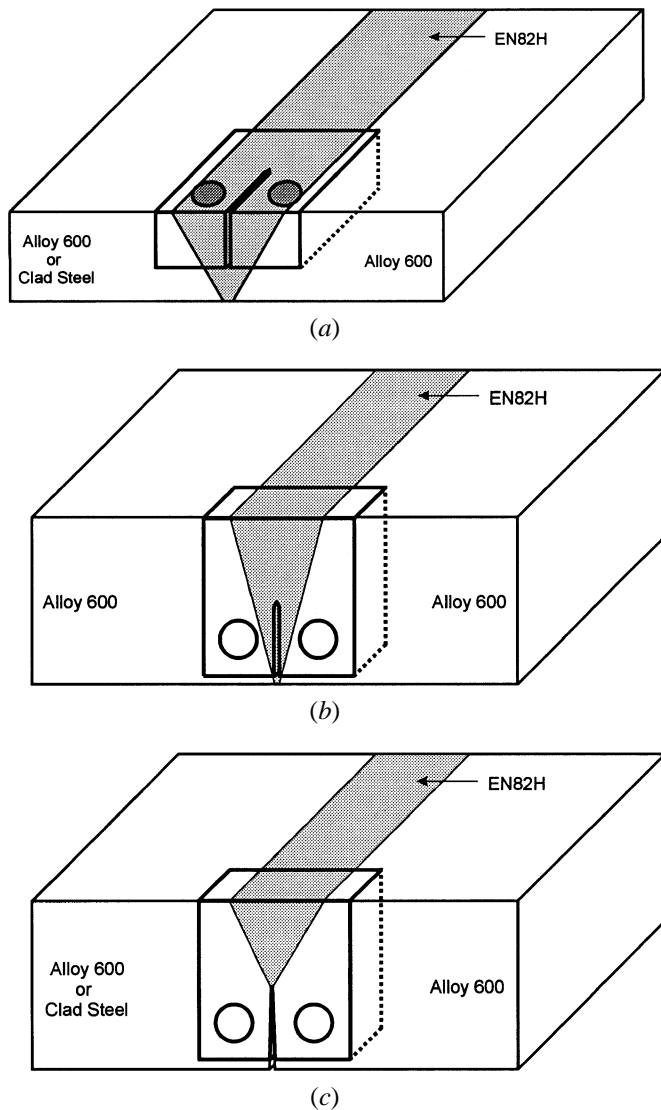


Fig. 1—Orientation of specimens machined from weldments: (a) longitudinal CT, (b) transverse CT, and (c) transverse CT.

transverse orientation, with the notch normal to the welding direction (Figures 1(b) and (c)).

Specimens were tested in air and water using servo-electric feedback-controlled systems that were operated in displacement control. Low loading rates ( $4 \text{ MPa}\sqrt{\text{m/h}}$  at  $54^\circ\text{C}$  and  $0.4$  to  $2 \text{ MPa}\sqrt{\text{m/h}}$  at  $338^\circ\text{C}$ ) were used to allow sufficient time for environmental effects to occur. Loading rates for a few  $54^\circ\text{C}$  water tests were also varied between  $0.4$  and  $26,000 \text{ MPa}\sqrt{\text{m/h}}$ , to evaluate rate effects on environmental cracking behavior.

In accordance with multiple-specimen  $J_{IC}$  test procedures, specimens were unloaded when the target displacement was reached, heat tinted to discolor the crack extension region, and subsequently fatigued apart so the amount of crack extension ( $\Delta a$ ) could be measured by the ASTM E1737-96 nine-point measurement method. Because external load cells were used to measure the test load in autoclave tests, the seal friction load was measured and subtracted from the load measurement at the beginning of the test. External linear variable differential transducers (LVDTs) were used to measure external displacement along the load train ( $v_{LT}$ ), so

measured values included displacements due to the load-train and specimen compliance. To convert measured load-train displacements to load-line displacements ( $v_{LL}$ ), the extraneous load-train displacements (due to stretching of the load train and Brinelling of the specimen load-pin holes) must be subtracted out. This was accomplished by comparing the measured compliance ( $C_{LT} = \Delta v_{LT}/\Delta P$ ) in the linear-elastic region with the theoretical load-line compliance ( $C_{LL} = \Delta v_{LL}/\Delta P$ ). The product of instantaneous load and the difference in measured vs theoretical compliances represents the extraneous load-train displacement, which was then subtracted from the load-train displacement to yield the corresponding load-line displacement. The corrected load vs load-line displacement data were used to compute  $J$  in accordance with ASTM E1737-96.

The multispecimen  $J$ - $R$  curve method was used to establish  $J_{IC}$  values, because the periodic unload/reload cycles associated with the unloading compliance method could possibly affect the environmental cracking behavior in water. The overall  $J_{IC}$  analysis methods were patterned after ASTM E1737-96. The  $J$ - $R$  curves were constructed by plotting  $J$  values against the corresponding crack extension values and fitting the  $J$ - $\Delta a$  data with a power-law regression line. The  $J_{IC}$  toughness was then taken to be the value of  $J$  at the intersection of the power-law curve with the  $0.2 \text{ mm}$  offset blunting line:

$$\Delta a = \frac{J}{4\sigma_f} + 0.2 \text{ (mm)} \quad [1]$$

where  $\sigma_f$  is the flow strength, which is equal to the average of the yield- and ultimate-strength levels (*i.e.*,  $\sigma_f = (\sigma_{YS} + \sigma_{UTS})/2$ ). Note that the blunting line developed for high-strain-hardening materials,<sup>[6]</sup>  $J = 4 \sigma_f (\Delta a)$ , accurately estimates the apparent blunting behavior for both the wrought and weld metals. Values of the tearing modulus ( $T$ ), which represents a dimensionless measure of the tearing resistance after  $J_{IC}$  is exceeded, were calculated from the following equation:<sup>[7]</sup>

$$T = \frac{dJ}{da} \frac{E}{\sigma_f^2} \quad [2]$$

where  $dJ/da$  is the average slope of the  $J$ - $R$  curve (at  $\Delta a \approx 1.3 \text{ mm}$ ) and  $E$  is the elastic modulus. For specimens that exhibited relatively little plasticity prior to the onset of environmental cracking, equivalent critical stress-intensity factors ( $K_{JC}$ ) were computed from experimental  $J_{IC}$  values using the equation<sup>[8]</sup>

$$K_{JC} = \sqrt{E J_{IC}} \quad [3]$$

Although the number and spacing of  $J$ - $\Delta a$  data points do not always conform to ASTM guidelines, the  $J_{IC}$  values obtained herein are judged to be accurate estimates of the material's fracture toughness. In cases where insufficient data preclude multispecimen analysis,  $J$ - $R$  curves and  $J_{IC}$  values were determined using a modified version of the normalization technique proposed by ASTM Task Group E08.08.02. This method estimates the crack extension at various points along a load-displacement curve by assuming a unique relationship between the load-displacement response and instantaneous crack length. A detailed description of the normalization technique is provided in Reference 9. In this article, multispecimen and normalized

**Table I. Material Chemistries (Weight Percent)**

Material	EN82H welds								
	Alloy 600	“B1” long.	“A1” long.	“C1” long.	“A2” trans.	“C2” trans.	“C3” trans.	“C4” trans.	“C5” long.
Heat or weld	NX5853G11	Unk	NX4628DK	NX8602D	NX4628DK	NX7906D	NX8128D	NX8906D	NX8602D
Shield gas	—	Ar	Ar	Ar-2% H <sub>2</sub>	Ar	Ar-5% H <sub>2</sub>	Ar-5% H <sub>2</sub>	Ar	Ar
Element									
C	0.08	0.007	0.03	0.043	0.045	0.04	0.039	0.039	0.04
Mn	0.27	2.4	2.39	2.77	2.69	2.80	2.83	2.78	2.77
Fe	8.61	1.5	2.02	1.18	1.53	1.08	0.83	1.34	1.18
S	0.0003	0.0007	<0.001	0.001	0.001	0.003	<0.001	<0.001	0.001
Si	0.25	<0.1	0.087	0.32	0.15	0.069	0.15	0.092	0.32
Cu	0.11	0.06	NM	0.08	0.06	NM	NM	NM	0.08
Ni	73.6	73.2	73.9	72.6	72.1	73.4	73.7	72.2	72.6
Cr	16.08	18.7	19.51	20.3	20.7	20.19	19.58	20.03	20.3
Al	0.16	0.09	NM	0.04	0.07	NM	NM	NM	0.04
Ti	0.41	0.4	0.39	0.29	0.41	0.41	0.33	0.33	0.29
P	0.011	<0.001	<0.001	<0.001	<0.001	<0.001	<0.001	<0.001	<0.001
Nb + Ta	0.121	NM	2.10	2.39	2.35	2.43	2.43	2.39	2.39

Long. = longitudinal orientation.  
 Trans. = transverse orientation.  
 Unk = unknown.  
 NM = not measured.

**Table II. Summary of Tensile Test Results**

T	Env	Prop	EN82H				Alloy 600			
			$\bar{x}$	H	L	s	$\bar{x}$	H	L	s
338 °C	water	$\sigma_{YS}$ (MPa)	397	445	347	30	307	341	259	43
		$\sigma_{UTS}$ (MPa)	571	622	476	49	661	668	655	7
		pct $\epsilon$	32.0	39.0	18.9	6.9	38.9	39.1	38.6	0.3
		pct RA	39.1	55.5	23.8	10.3	55.2	56.6	53.0	1.9
54 °C	water	$\sigma_{YS}$ (MPa)	474	554	373	74	267	323	199	51
		$\sigma_{UTS}$ (MPa)	601	674	466	70	594	629	527	59
		pct $\epsilon$	18.0	27.4	10.1	6.9	34.7	35.6	34.0	0.8
		pct RA	27.2	37.8	18.8	7.6	54.4	60.7	43.6	9.4
338 °C	air	$\sigma_{YS}$ (MPa)	423	489	358	57	288	327	264	34
		$\sigma_{UTS}$ (MPa)	643	674	599	30	655	666	647	10
		pct $\epsilon$	41.5	54.5	25.9	10.3	38.6	39.5	38.1	0.8
		pct RA	51.5	54.7	46.1	3.2	56.6	61.6	53.2	4.4
54 °C	air	$\sigma_{YS}$ (MPa)	436	525	342	63	319	353	292	31
		$\sigma_{UTS}$ (MPa)	646	679	613	30	654	661	642	10
		pct $\epsilon$	34.3	49.1	24.6	9.5	36.5	37.4	35.7	0.9
		pct RA	51.2	57.1	37.8	6.7	59.7	64.9	56.2	4.6
RT	air	$\sigma_{YS}$ (MPa)	430	483	359	53	337	391	309	47
		$\sigma_{UTS}$ (MPa)	689	710	680	11	709	714	705	5
		pct $\epsilon$	39.6	57.3	23.8	12.9	49.8	50.2	49.5	0.4
		pct RA	55.4	58.8	46.7	4.4	62.8	65.7	60.7	2.6

T—test temperature; ENV—test environment; Prop—material properties;  $\sigma_{YS}$ —0.2% offset yield strength,  $\sigma_{UTS}$ —ultimate strength, pct  $\epsilon$ —percent elongation, and pct RA—percent reduction in area;  $\bar{x}$ —average of all data; H—highest data point; L—lowest data point; and s—standard deviation of data.

J–R curves are represented by solid and broken lines, respectively.

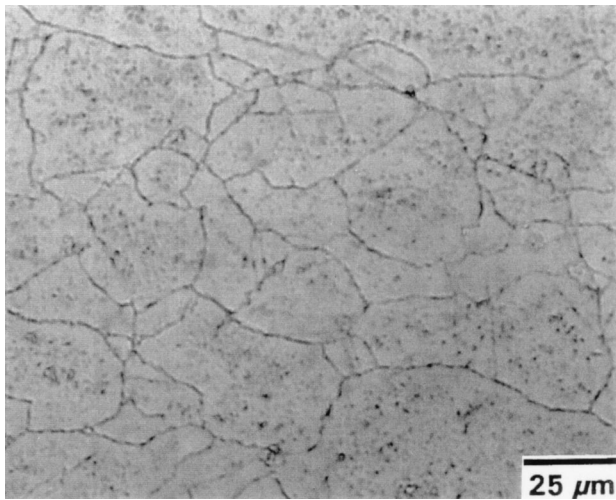
**C. Test Environment**

Flowing autoclaves were used to conduct tests in hydrogenated deaerated water at room temperature, 54 °C, 93 °C, 49 °C, and 338 °C. The room-temperature pH was 10.1 to 10.3, and the oxygen concentration typically ranged

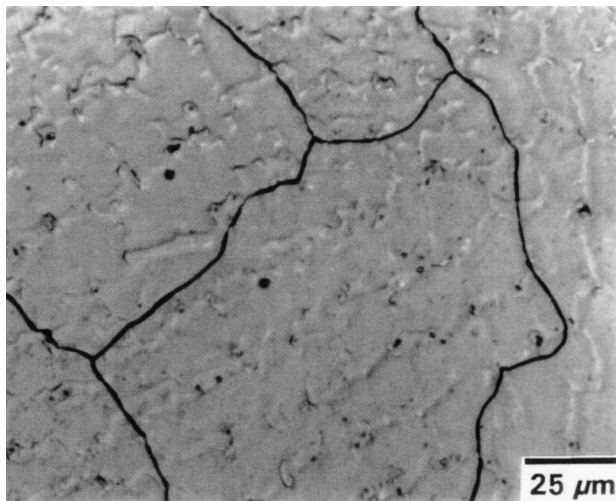
between 3 and 17 ppb. The nominal hydrogen concentration was 150 cc H<sub>2</sub>/kg H<sub>2</sub>O, but ranged from 140 to 180 cc H<sub>2</sub>/kg H<sub>2</sub>O.

**D. Hydrogen Precharging**

Hydrogen precharging was conducted in a high-pressure autoclave using 99.999 pct pure hydrogen. As-fabricated CT specimens were first cleaned in acetone and then rinsed in



(a)



(b)

Fig. 2—Representative microstructures for (a) alloy 600 and (b) EN82H weld.

alcohol before insertion into an autoclave. To assure a uniform hydrogen distribution, precharging runs were performed at 360 °C for 6 weeks. After precharging, specimens were refrigerated below 0 °C to minimize hydrogen loss. Hydrogen concentrations for at least two specimens for each test material were measured after rising load testing was complete. Chips produced by hole drilling were collected for hydrogen analysis. Care was taken not to heat the sample or contaminate the chips. Vacuum extraction analysis on a LECO\* hydrogen analyzer was performed to measure the

\*LECO is a trademark of LECO Corporation, St. Joseph, MI.

hydrogen content. The range of hydrogen concentrations was 42 to 45 ppm for alloy 600 and 55 to 57 ppm for the EN82H weld.

#### E. Fracture-Toughness Testing of Hydrogen-Precharged Specimens

Hydrogen-precharged CT specimens were also tested in 24 °C air and water at a nominal  $\dot{K}$  of 4 MPa $\sqrt{m/h}$ . The resulting load vs load-line displacement curves and maximum stress-intensity factors ( $K_{Pmax}$ ) for these tests are compared with those obtained for nonprecharged specimens tested in water, to determine if hydrogen embrittlement is responsible for the degraded fracture-toughness response in low-temperature water. Values of  $K_{Pmax}$  were computed by substituting the maximum load and initial crack length into the  $K$  solution in ASTM E1737-96. While  $K_{Pmax}$  values are not strictly valid when significant plasticity occurs (*i.e.*, >60 MPa $\sqrt{m}$ ), they provide a relative ranking of susceptibility to low-temperature cracking.

#### F. Fractographic Examinations

Broken specimen halves were examined by SEM to characterize the fracture-surface morphology, in order to compare operative cracking mechanisms for hydrogen-precharged and nonprecharged specimens.

Table III. Summary of Fracture Toughness Values for Welds

Weld	Orientation	Environment	Temperature (°C)	$J_{IC}$ (kJ/m <sup>2</sup> )	$T$
A1,B1	long.	air	54–338	806	364
C5	long.	air	24	731; 891	381; 481
A1	long.	water	54	42; 60; 79	31; 24; 44
B1	long.	water	54	26; 34	18; 15
C1	long.	water	54	14	3
C1	long.	water	93	19	7
C1	long.	water	149	212; 213	84; 118
A1	long.	water	338	679	373
A2	trans.	air	24–338	350	349
C4	trans.	air	24	638; 671	412; 481
A2	trans.	water	54	23; 26; 51	26; 13; 35
C2–C4	trans.	water	54	13	5
C3	trans.	water	149	145; 254	138; 229
A2	trans.	water	338	459	245

Long. = longitudinal orientation.

Trans. = transverse orientation.

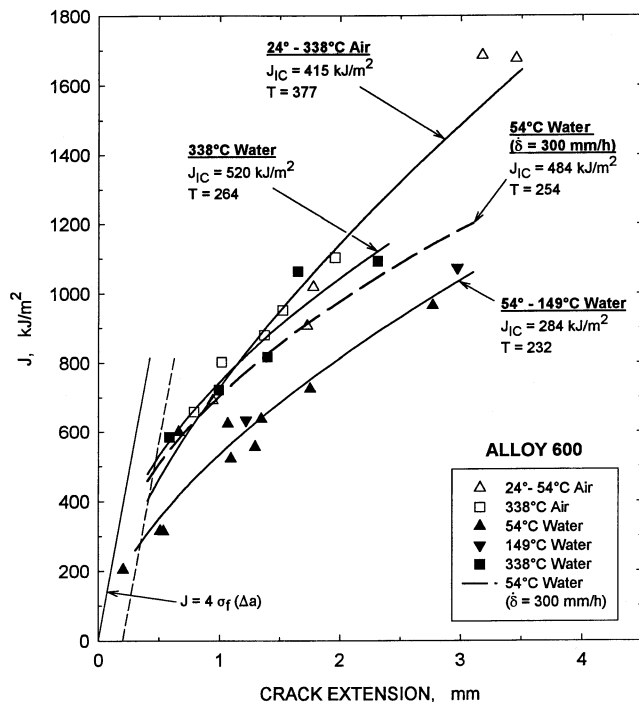


Fig. 3— $J$ - $R$  curves for alloy 600 in air and water. Tests in water were conducted at a displacement rate of 0.05 mm/h ( $K = 4 \text{ MPa}\sqrt{\text{m/h}}$ ) with the exception of one high rate test that was conducted at 300 mm/h ( $K = 26,000 \text{ MPa}\sqrt{\text{m/h}}$ ).

### III. RESULTS AND DISCUSSION

#### A. Fracture of Alloy 600

The cracking behavior of alloy 600 in air and water is summarized in Figure 3. It is seen that the fracture response in air is independent of temperature, so the 24 °C and 338 °C data were combined into a single  $J$ - $R$  curve that yielded a  $J_{IC}$  value of 415 kJ/m<sup>2</sup> and a  $T$  value of 377. High-temperature water has almost no effect on the toughness response, while 54 °C to 149 °C water causes a modest toughness reduction. The exceptionally high values in air and high-temperature water demonstrate that fracture is not an issue for alloy 600 under these conditions. While  $J_{IC}$  in low-temperature water is reduced by 30 pct, sufficient cracking resistance is retained to avoid fracture concerns. In fact, the load-displacement records in Figure 4(a) show that alloy 600 exhibits tremendous plastic deformation capabilities in both low- and high-temperature water.

The dashed  $J$ - $R$  curve in Figure 3 reveals that increasing the displacement rate to 300 mm/h ( $K \approx 26,000 \text{ MPa}\sqrt{\text{m/h}}$ ) eliminates most of the environmental effect, as there is insufficient time to embrittle the crack-tip region. During high-rate testing in 54 °C water, where the time to maximum load is  $\sim 20$  seconds, there is a substantial recovery in toughness as  $J_{IC}$  increases from 284 to 484 kJ/m<sup>2</sup>.

Representative fracture-surface morphologies for alloy 600 samples tested under various conditions are shown in Figure 5. In 54 °C and 338 °C air, the fracture surfaces have a very ductile appearance, with three operative cracking mechanisms (Figure 5(a)), including primary microvoid coalescence, void-sheet formation, and shear-stretch formation. The primary dimples are nucleated by broken MC-type inclusions, and their growth involves considerable plastic

deformation, as evidenced by extensive serpentine slip on the walls of the very deep dimples. Void-sheet regions, associated with a network of very fine dimples,<sup>[10,11]</sup> often surround the primary dimples. This mechanism occurs when high shear stresses in the ligaments between primary dimples cause intragranular carbides to decohere from the matrix, forming an extremely high density of voids that subsequently coalesce to form steep void sheets. Large portions of the fracture surface also exhibit rather nondescript features, termed shear-stretch features.<sup>[4]</sup> This feature also results from high local shear stresses, but occurs when there is a lack of failed particles, so conventional microvoid coalescence is precluded. Well-defined slip offsets are typically observed in the shear-stretch regions, indicating that this mechanism requires extensive plastic deformation. In addition, the few dimples that form in these regions are shallow and elongated, demonstrating that shear stresses play an important role in creating this fracture surface. The overall appearance of the shear-stretch regions is almost identical to conventional stretch zones that develop ahead of fatigue precracks due to crack-tip blunting.

The extensive plastic deformation associated with microvoid coalescence, void-sheet, and shear-stretch mechanisms accounts for the high  $J_{IC}$  fracture toughness. The primary difference between low- and high-temperature air is the presence of considerably more evidence of shear-stretch regions at 640 °F. As shown in Figure 5(b), alloy 600 fracture surfaces generated in high-hydrogen water were indistinguishable from their air counterparts. The identical fracture-surface morphologies in air and water are consistent with the high toughness response observed in both environments.

#### B. Fracture of EN82H Weld

In air and high-temperature water, EN82H welds exhibit very high fracture toughness in both the longitudinal and transverse orientations, as shown in Figures 6 and 7. Values of  $J_{IC}$  and the tearing modulus for the EN82H welds are provided in Table III. For the manufacturer A welds, the cracking response in air is seen to be independent of temperature, so the 54 °C and 338 °C data were pooled to establish common  $J$ - $R$  curves. Longitudinal weld A1 displays exceptionally high toughness, with a  $J_{IC}$  of 806 kJ/m<sup>2</sup> in air and 679 kJ/m<sup>2</sup> in 338 °C water. Although toughness values for the transverse weld A2 are about 30 to 50 pct lower, they are still sufficiently high to preclude fracture concerns. Welds manufactured by vendor C also display exceptionally high fracture toughness, but less anisotropy than the A welds. Values of  $J_{IC}$  for the C welds range from 731 to 891 kJ/m<sup>2</sup> in the longitudinal orientation to 638 to 671 kJ/m<sup>2</sup> in the transverse orientation. These results demonstrate that fracture is not a design or operation concern for EN82H welds in either air or high-temperature water.

In low-temperature water, however, both the longitudinal (A1, B1, and C1) and transverse (A2, C2, C3, and C4) welds show a dramatic toughness degradation, with  $J_{IC}$  and the tearing moduli being reduced by one to two orders of magnitude. In 54 °C water, the lowest-toughness longitudinal (C1) and transverse (C2, C3, and C4) welds have  $J_{IC}$  values of 14 and 13 kJ/m<sup>2</sup> and tearing moduli of 3 and 5, respectively. While increasing the water temperature to 93 °C has little effect on the cracking resistance, further increasing it to 149 °C produces a significant recovery. This demonstrates that

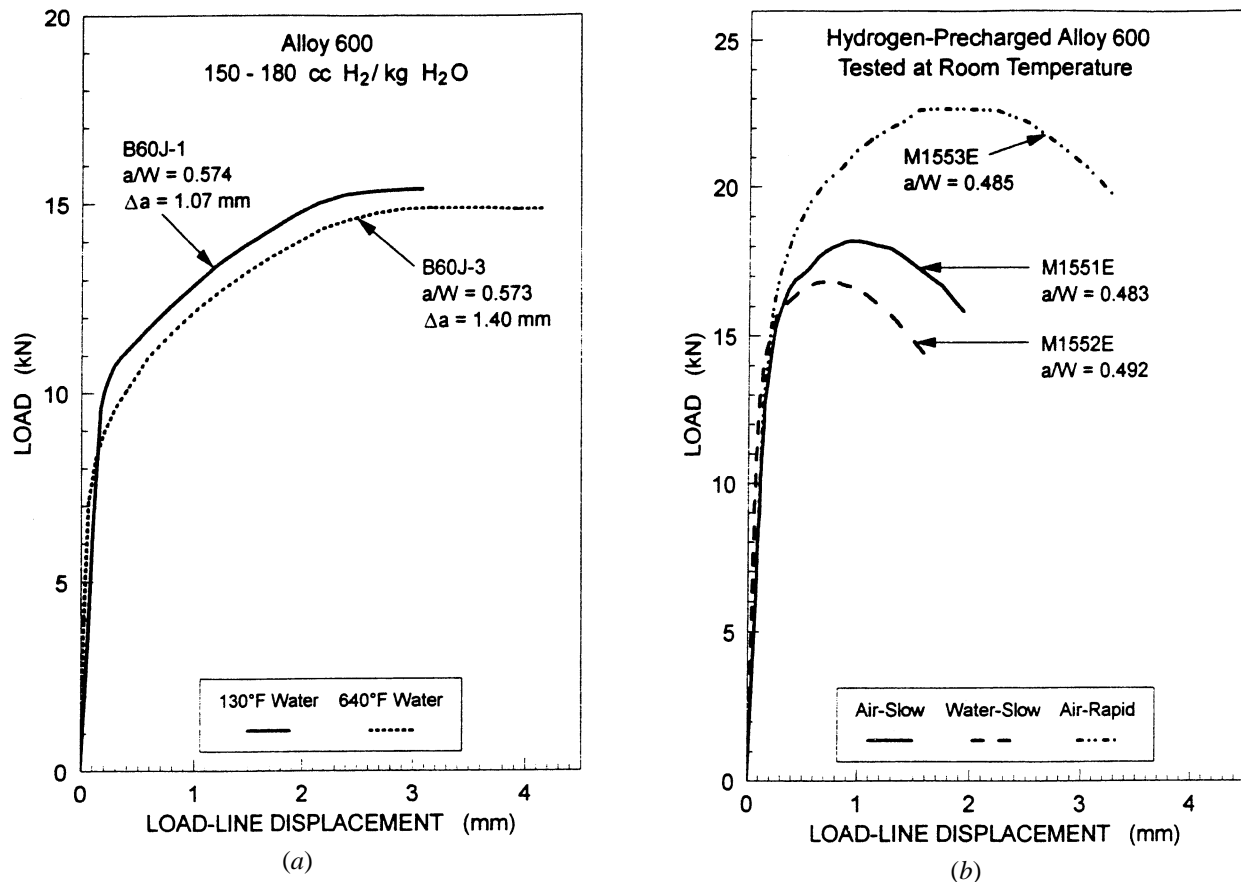


Fig. 4—Load vs load-line displacement curves for alloy 600. (a) Nonprecharged specimens tested in 54 °C and 338 °C water. (b) Hydrogen precharged specimens tested in 24 °C air and water. Rapid testing of specimen M1553E was conducted at a  $K$  of  $4 \times 10^5$  MPa $\sqrt{m/h}$ .

low-temperature crack propagation (LTCP) is only an issue below 149 °C.

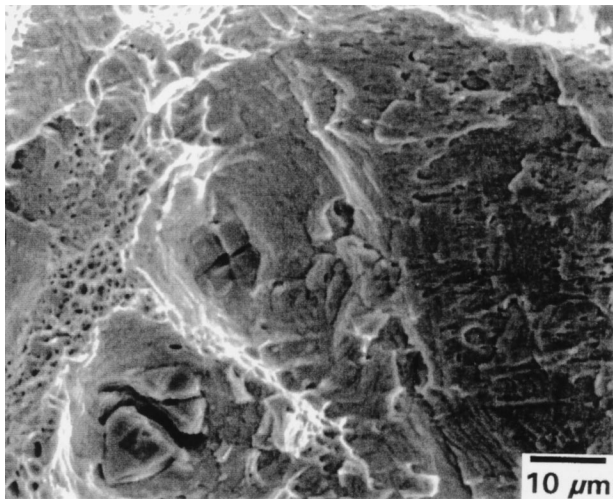
The load-displacement curves in Figure 8(a) illustrate the dramatic environmental degradation in 54 °C water. In contrast with the exceptionally high ductility observed in 338 °C water (which is comparable to that observed in air), specimens tested in low-temperature water display very little ductility, as environmental cracking causes a premature deviation from linear-elastic behavior. Moreover, welds C2, C3, and C4, which cracked under essentially linear-elastic conditions in 54 °C water, show even less ductility than that for specimen M1395 (Figure 8(a)).

Because LTCP in the most embrittled welds initiated in the linear-elastic regime,  $K_{JC}$  values tend to be reasonably consistent with  $K_{Pmax}$  values, although there was considerable scatter in the latter. For longitudinal weld C1, the  $K_{JC}$  value of 54 MPa $\sqrt{m}$  is near the mean of the  $K_{Pmax}$  values, which ranged from 40 to 77 MPa $\sqrt{m}$ . For transverse welds C2, C3, and C4, the  $K_{JC}$  value of 52 MPa $\sqrt{m}$  falls in the lower portion of the  $K_{Pmax}$  range: 48 to 77 MPa $\sqrt{m}$ . The good agreement between  $K_{JC}$  and  $K_{Pmax}$  for the most embrittled welds suggests that LTCP tends to initiate at or slightly before maximum load, and linear-elastic fracture mechanics can be used to quantify the limiting response. Thus, the lower-bound  $K_{Pmax}$  values of 40 and 48 MPa $\sqrt{m}$  can be used in design and operational analyses.

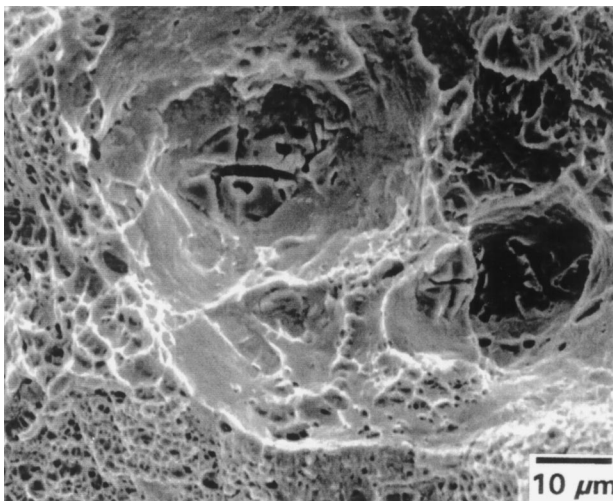
Figure 9, which summarizes the effect of displacement rate on LTCP behavior, shows that the  $J$ - $R$  curve response is essentially rate independent below 1.3 mm/h ( $\dot{K} < 100$

MPa $\sqrt{m/h}$ ). Within this regime, longer exposure times do not further embrittle the crack-tip region. Increasing the rate to 15 mm/h (1300 MPa $\sqrt{m/h}$ ) provides some improvement in toughness, and at 305 mm/h (26,000 MPa $\sqrt{m/h}$ ), the toughness is significantly enhanced. At this rate, there is insufficient time to fully embrittle the crack-tip region. These results indicate that transient loading events with  $\dot{K}$  values greater than about 1300 MPa $\sqrt{m/h}$  will not cause environmental cracking in low-temperature water.

Fractographic studies show that the dramatic toughness degradation in low-temperature water is associated with a fracture-morphology transition from ductile tearing to intergranular cracking. In air and high-temperature water, the ductile tearing behavior in EN82H welds involves a combination of microvoid coalescence and shear-stretch formation regardless of crack orientation, as shown in Figure 10. For the longitudinal orientation, the tremendous amount of plastic deformation required to initiate and grow primary microvoids (Figure 10(a)) and create shear-stretch regions (Figure 10(b)) accounts for the exceptionally high  $J_C$  and tearing modulus. This is evidenced by the extensive serpentine slip offsets found in all areas of the fracture surface. The fracture resistance for the longitudinal weld is higher than its base metal counterpart, because of the lack of large primary MC-type inclusions in the welds. In wrought material, the large inclusions typically fracture during the early stages of plastic straining, thereby creating large microvoids ahead of the crack tip. The absence of these large inclusions in welds suppresses microvoid nucleation. Additional energy is then



(a)



(b)

Fig. 5—Fracture surface morphology for alloy 600. (a) 360 °C air. Large microvoids surrounded by void sheets (left) and shear stretch region (right). (b) 54 °C water. Large microvoids surrounded by void sheets.

required to nucleate microvoids, because the smaller nitrides resist fracture and fail by a decohesion process at much higher strains.

The anisotropy in toughness behavior exhibited by the A welds is due to crack propagation along rows of carbides that form a preferred crack path. For transverse weld A2, the fracture surface exhibits both well-defined dimples and shallow dimples that tend to be aligned in the crack growth direction, as shown in Figure 10(c). The alignment occurs because carbide clusters often form in the interdendritic regions that are aligned in the heat-flow direction during weld solidification. When properly oriented, the rows of carbides serve as effective microvoid nucleation sites that provide a preferred crack path. The alignment is greatest along the columnar grains in the weld root, which is the region tested in specimens machined from weld A2 (Figure 1(c)). While the alignment of carbides in the weld root region accounts for the lower  $J_{IC}$  value in the transverse orientation, the significant plastic deformation required for microvoid growth and shear-stretch formation accounts for the relatively high  $J_{IC}$  value of 350 kJ/m<sup>2</sup> and the exceptionally high

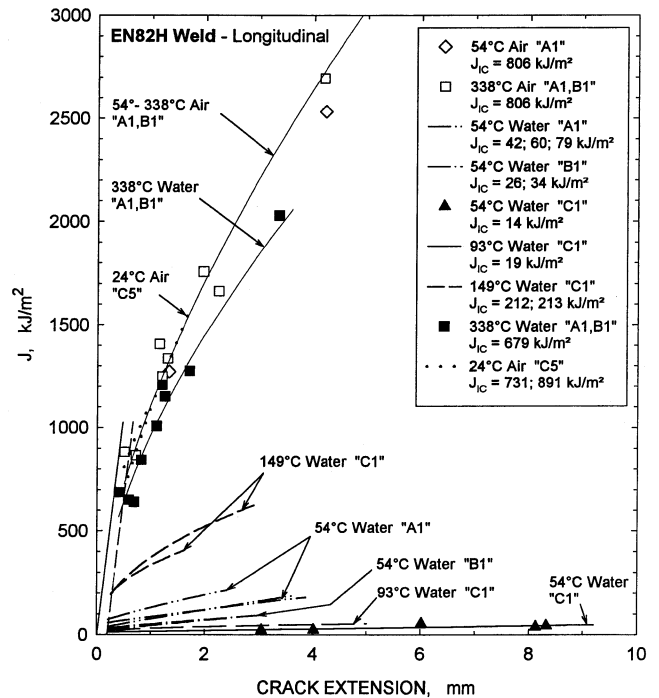


Fig. 6— $J$ - $R$  curves for EN82H welds with longitudinal orientation tested in air and water. Weld identification is provided inside quotation marks.

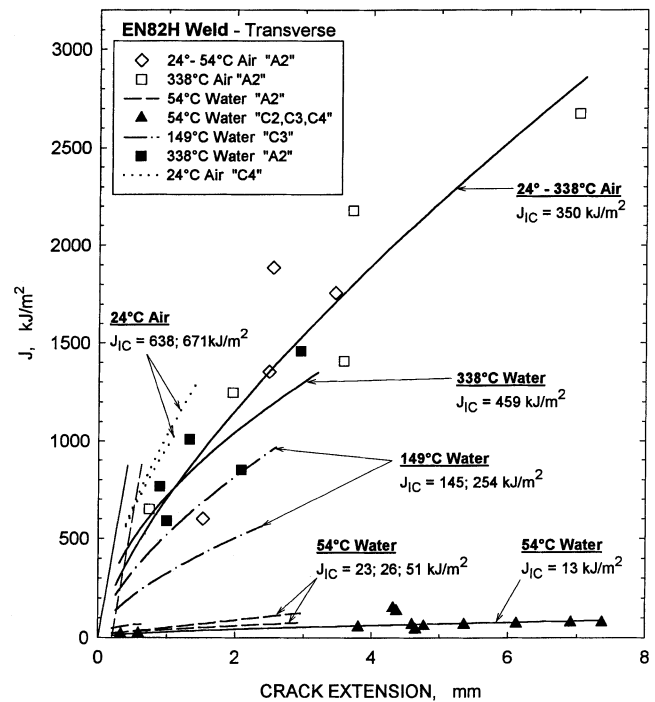


Fig. 7— $J$ - $R$  curves for EN82H welds with transverse orientation tested in air and water. Weld identification is provided inside quotation marks.

tearing modulus of 349. The C welds show less anisotropy, because the center of the weld (Figure 1(b)), rather than the weld root, was tested in transverse-oriented specimens.

In 54 °C water, a fracture-mechanism transition from ductile tearing to intergranular cracking dramatically reduces cracking resistance. Figure 11 shows that intergranular cracking is dominant regardless of specimen orientation.

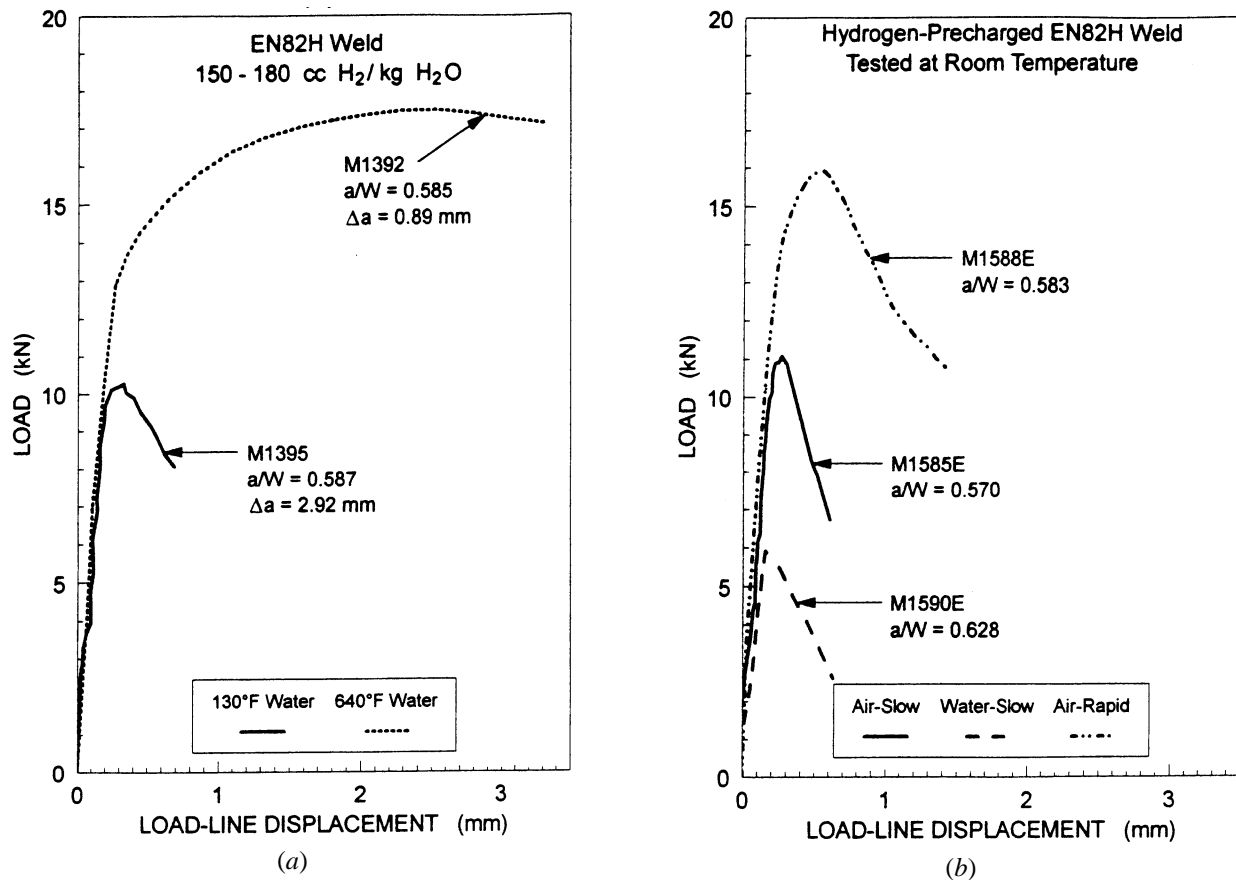


Fig. 8—Load vs load-line displacement curves for EN82H weld. (a) Nonprecharged specimens tested in 54 °C and 338 °C water. (b) Hydrogen precharged specimens tested in 24 °C air and water. Rapid testing of specimen M1588E was conducted at a  $\dot{K}$  of  $4 \times 10^5$  MPa $\sqrt{\text{m/h}}$ .

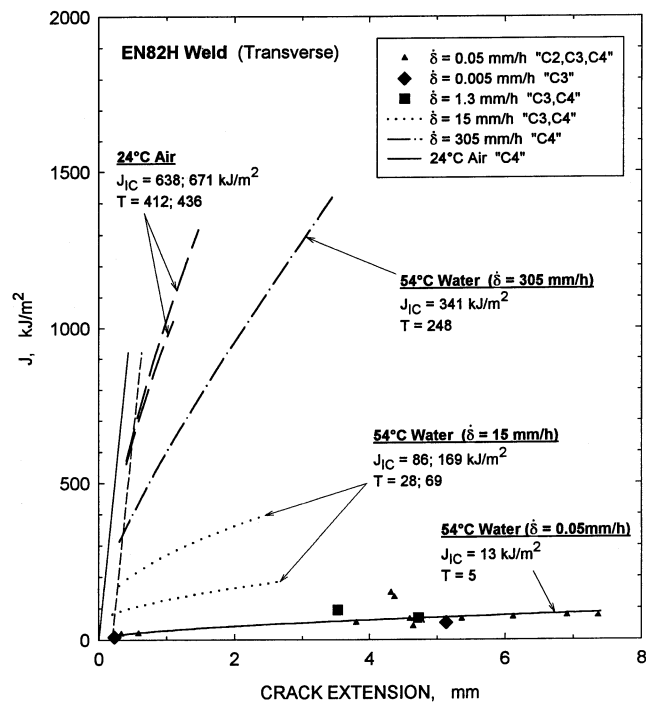


Fig. 9—Effect of displacement rate on cracking resistance of EN82H weld in 54 °C water. Weld identification is provided inside quotation marks.

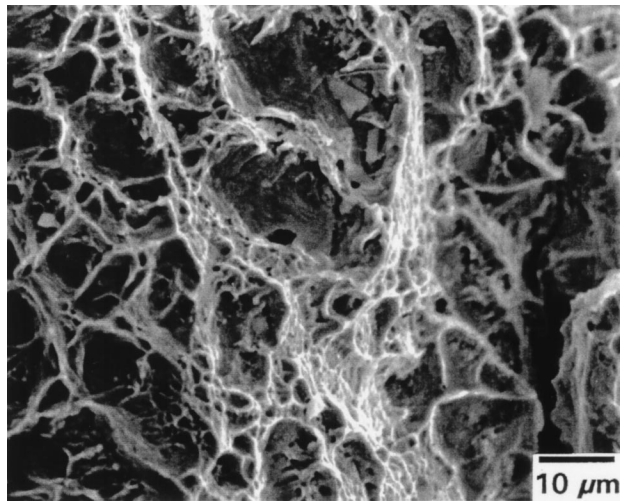
The failed grain surfaces are rather smooth and often have an undulating appearance that reflects the dendritic nature of the weld; intermediate-size nitrides that pinned grain boundaries are apparent in many regions (Figure 11(d)). Within some grains, there is evidence of transgranular faceting along steep walls linking two adjacent intergranular cracks with different elevations. The crystallographic facets are well defined (Figure 11(d)) and sometimes contain coarse slip offsets. As discussed later, both the intergranular and faceted morphologies can be reproduced in hydrogen-precharged specimens tested in air, thereby demonstrating that these cracking modes are associated with hydrogen embrittlement mechanisms.

The overall fracture-surface appearance for specimens tested in 54 °C water suggests that intergranular cracking first occurs along the most susceptible grain boundaries. This leaves unbroken ligaments in the wake of the advancing intergranular crack in regions where grain boundaries resist cracking. When the primary crack extends well beyond these ligaments, local stress intensities become large enough to nucleate an alternate faceted cracking mechanism that causes ligament failure.

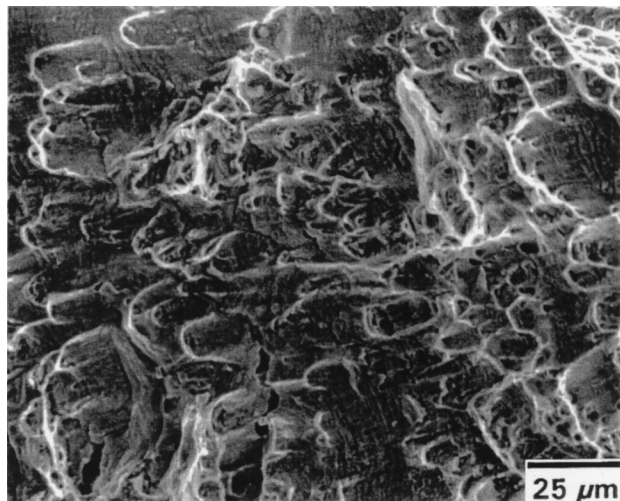
### C. Low-Temperature Cracking Mechanism

The embrittlement displayed by EN82H welds in low-temperature water is very similar to that observed in alloy X-750.<sup>[12]</sup> Because hydrogen embrittlement of grain boundaries was found to be responsible for the rapid cracking of alloy

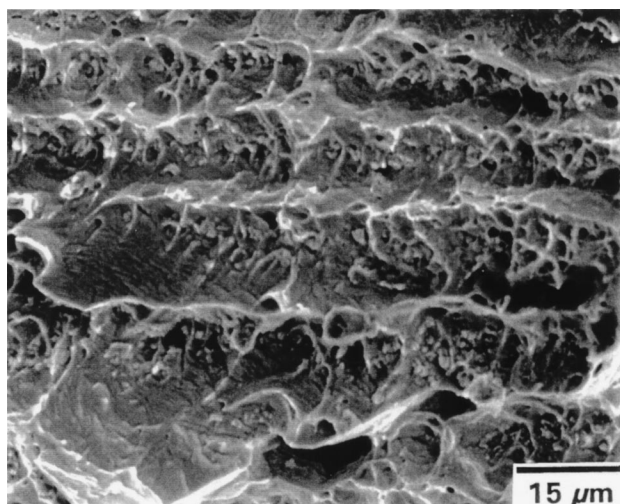




(a)



(b)



(c)

Fig. 10—SEM fractographs of EN82H welds tested in 360 °C water. (a) In longitudinal weld specimens, large primary dimples are well defined and often surrounded by void sheets. (b) Longitudinal weld. Equiaxed dimples (right) and shear stretch region (left and center) with elongated dimples and serpentine slip steps. (c) In transverse weld, dimples are sometimes rather shallow and aligned in rows.

X-750 in low-temperature water, a similar hydrogen-assisted cracking mechanism was suspected for EN82H. To understand the possible role of hydrogen in embrittling the weld metal, CT specimens precharged with 56 ppm hydrogen were tested in room-temperature air in an attempt to reproduce the cracking mechanisms and degree of embrittlement observed in low-temperature water. Most of the tests were conducted at the same loading rate ( $K \approx 4 \text{ MPa}\sqrt{\text{m/h}}$ ) used in conventional rising load tests, to assure that there was sufficient time for fully developing hydrogen enrichment in the peak triaxial stress region ahead of a crack. In addition, a few hydrogen-precharged specimens were rapidly loaded at a  $K$  of about  $4 \times 10^5 \text{ MPa}\sqrt{\text{m/h}}$  (the time to reach maximum load was 1 to 3 seconds), to preclude hydrogen from diffusing and accumulating ahead of the crack. These tests were designed to establish the critical hydrogen level necessary to cause intergranular LTCP. Finally, hydrogen-precharged specimens were tested in 54 °C water with 150 cc  $\text{H}_2/\text{kg H}_2\text{O}$ , to determine if hydrogen from the water combines with hydrogen already present to further embrittle the material. Similar testing of alloy 600 specimens precharged to 44 ppm hydrogen was also performed in an attempt to understand why the wrought metal resists low-temperature embrittlement. Results from the hydrogen-precharged studies are given subsequently.

#### D. Hydrogen-Precharged EN82H

Figure 12 compares  $K$  values at maximum load for hydrogen-precharged and nonprecharged specimens tested in air and low-temperature water. While  $K_{P_{\text{max}}}$  values higher than  $60 \text{ MPa}\sqrt{\text{m}}$  are not valid per linear-elastic fracture-mechanics methodology, they provide a relative ranking of LTCP resistance for comparison purposes.

For nonprecharged weld specimens, testing in 54 °C water is seen to cause a 50 pct reduction in  $K_{P_{\text{max}}}$ , from 97 to 103 to 42 to 60  $\text{MPa}\sqrt{\text{m}}$ . The hydrogen-precharged weld specimen tested in air exhibits a  $K_{P_{\text{max}}}$  of 58  $\text{MPa}\sqrt{\text{m}}$ , which is consistent with the range of values observed for nonprecharged specimens tested in water. Moreover, comparison of Figures 8(a) and 8(b) shows that load-displacement records for the nonprecharged specimen tested in water (M1395) and the precharged specimen tested in air (M1585E) are remarkably similar. The overall agreement in  $K_{P_{\text{max}}}$  values and load-displacement curves strongly implicates hydrogen embrittlement as the cause of the inferior cracking resistance in low-temperature water.

Fractographic observations provide additional evidence for a hydrogen embrittlement mechanism, as fracture-surface morphologies for nonprecharged specimens tested in 54 °C water (Figure 11) are almost indistinguishable from those for the precharged specimen tested in air (Figure 13). Intergranular cracking is dominant in the precharged specimen, and the smooth, undulating nature of the grain-boundary surfaces is consistent with the morphology for LTCP in water. In addition, small islands of transgranular faceting and poorly defined dimples are observed in both environments (Figure 11(c) and 13(b)). Some facets have a sharp crystallographic appearance, while others are poorly defined. Figure 13(c) shows well-defined facets oriented along planar slip bands, which suggests that facet formation is associated with localized failure along planar slip bands.

The presence of periodic slip offsets superimposed on

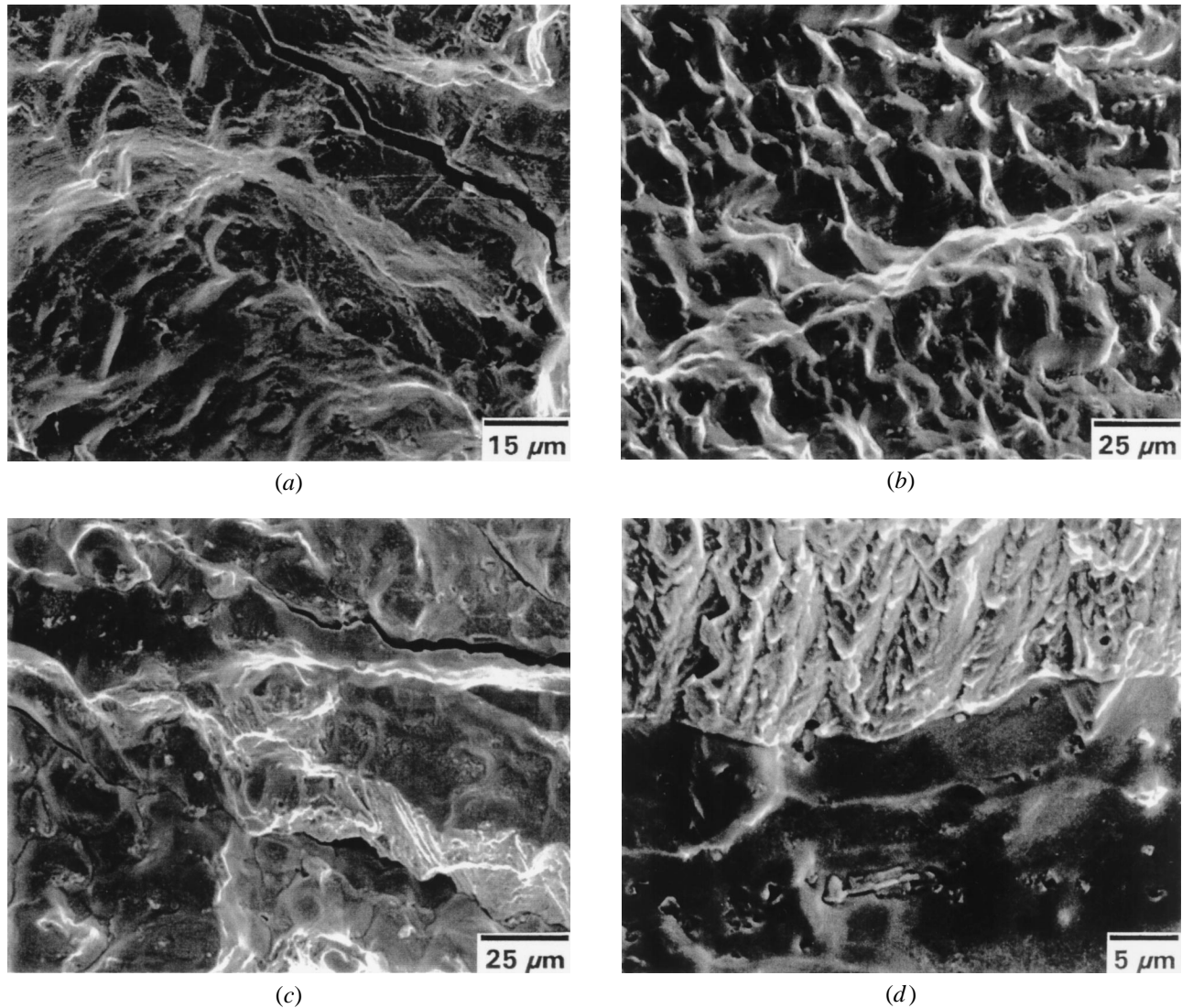


Fig. 11—SEM fractographs of (a) longitudinal and (b) through (d) transverse welds tested in 54 °C water. (a) and (b) Typical intergranular fracture appearance. (c) Typical intergranular cracking with small transgranular region (lower right). (d) High magnification of intergranular fracture (bottom) and transgranular faceting (top).

the transgranular facets in both the hydrogen-precharged specimen and nonprecharged specimen tested in water (Figure 14) provides additional evidence that a hydrogen embrittlement mechanism is active in low-temperature water. This observation indicates that hydrogen in the lattice promotes planar slip. Thus, hydrogen appears to play a dual role in degrading cracking resistance; it reduces grain-boundary cohesive strength and promotes planar slip, which localizes strain concentrations along grain boundaries. In summary, the presence of hydrogen, regardless of its source, appears to promote intergranular cracking, planar slip, and transgranular faceting. While hydrogen-induced planar slip and intergranular cracking have a profound influence on cracking resistance, hydrogen-induced transgranular cracking appears to be less important because this mechanism occurs very late in the cracking process, when the ligaments separating large intergranular cracks fail.

The most severe embrittlement occurs when a precharged weld specimen is tested in low-temperature water. Comparison of the load-displacement records for specimens M1590E and M1585E (Figure 8(b)) shows that testing in water causes

cracking to initiate at lower loads, which results in a further reduction in  $K_{P_{max}}$  to 39 MPa $\sqrt{m}$ . Apparently, hydrogen from the water increases the total hydrogen content ahead of a crack, which further degrades cracking resistance. In this case, hydrogen from the water is added to the precharged hydrogen to severely weaken grain boundaries, fill traps, and interact with dislocations. It is also noted that the presence of hydrogen in the water minimizes the loss of precharged hydrogen from the crack-tip region. The fracture-surface morphologies for the precharged specimens tested in air and water are virtually identical, as intergranular cracking is the dominant mechanism in both cases. The only discernable difference is more crystallographic facets in the transgranular islands of the precharged specimen tested in water.

Rapid loading of the precharged weld restores some ductility and increases  $K_{P_{max}}$  to an intermediate level of 84 MPa $\sqrt{m}$  (Figure 12). Although there is an increase in ductility, the total amount of plastic deformation is much less than that in air or 338 °C water, as shown by comparing the load-displacement records for specimens M1392 and M1588E in Figures 8(a) and 8(b). Rapid loading of the precharged

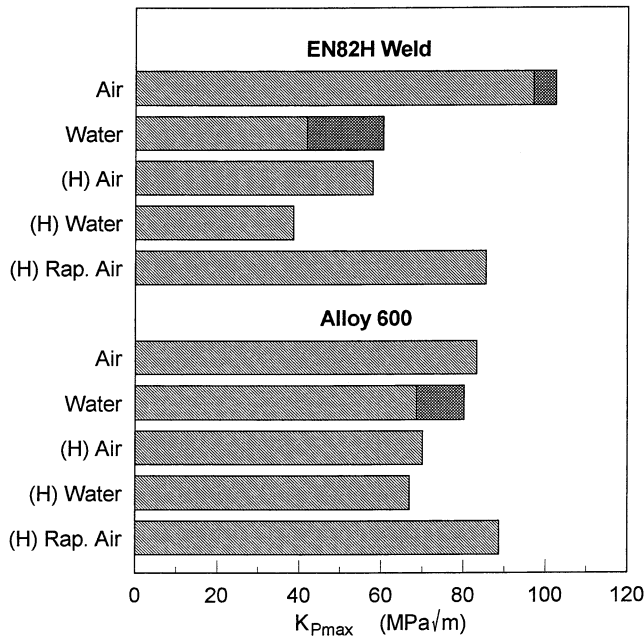


Fig. 12— $K_{Pmax}$  values for EN82H weld and alloy 600. Nonprecharged specimens were tested in 54 °C air or water, while hydrogen-precharged [H] specimens were tested in 24 °C air or water. The short test times for the rapid loading tests [Rap], where maximum load was reached in 1 to 3 s, preclude hydrogen from diffusing to the crack tip; hence, bulk hydrogen levels control fracture behavior. Cross-hatching represents a range of  $K_{Pmax}$  values.

specimens produces a combination of transgranular and intergranular cracking. Significant intergranular cracking is present, but transgranular cracking mechanisms, including dimple rupture and crystallographic faceting, are dominant.

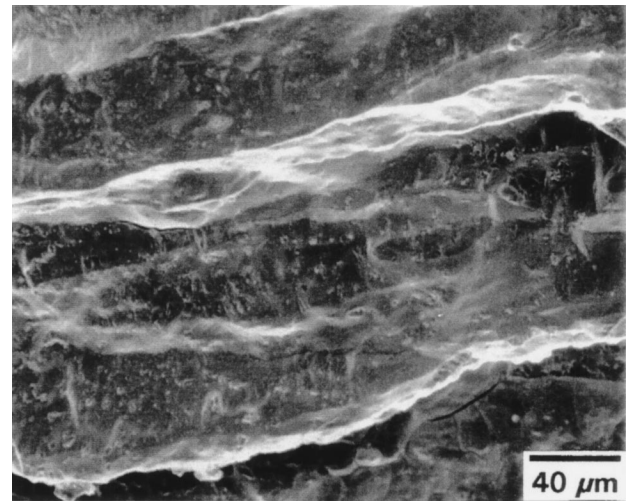
The critical hydrogen concentration required to cause low-temperature embrittlement in water can be estimated by comparing the cracking response in the fast and slow tests on hydrogen-precharged specimens. During the rapid loading test, where the time to maximum load is about 1 to 2 seconds, the local hydrogen concentration in the lattice is taken to be the bulk content (56 ppm), because there is insufficient time for hydrogen to diffuse and accumulate ahead of the crack. Thus, rapid test results for the precharged weld show that a local hydrogen content of 56 ppm causes a substantial decrease in toughness, which is associated with some intergranular cracking and faceting; however, higher hydrogen levels are required to induce predominantly intergranular cracking and reduce toughness levels to those associated with LTCP.

A better estimate of the critical local hydrogen concentration required for LTCP can be obtained from the slow test of the precharged weld specimen. During this test, there is sufficient time for hydrogen to diffuse and accumulate in the peak triaxial stress region ahead of the crack tip. The local hydrogen concentration under equilibrium conditions is estimated using the relationship<sup>[13]</sup>

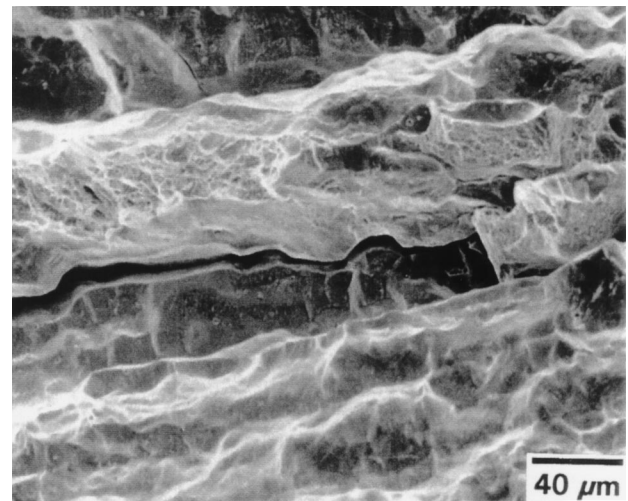
$$\frac{C}{C_0} = \exp \frac{\sigma_H \bar{V}}{RT} \quad [4]$$

where

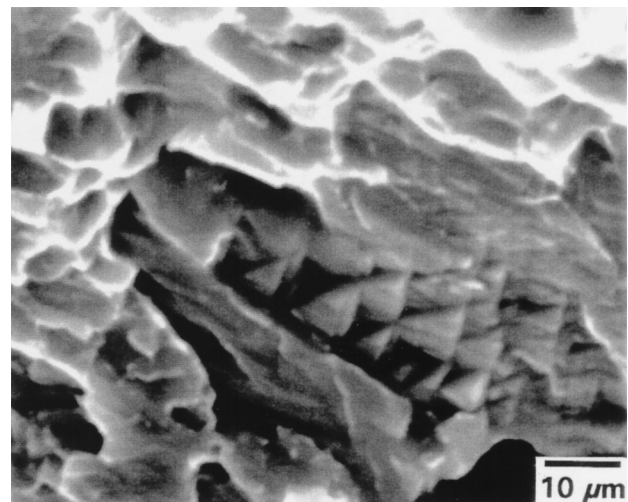
$C$  = local hydrogen concentration,  
 $C_0$  = bulk hydrogen concentration,



(a)

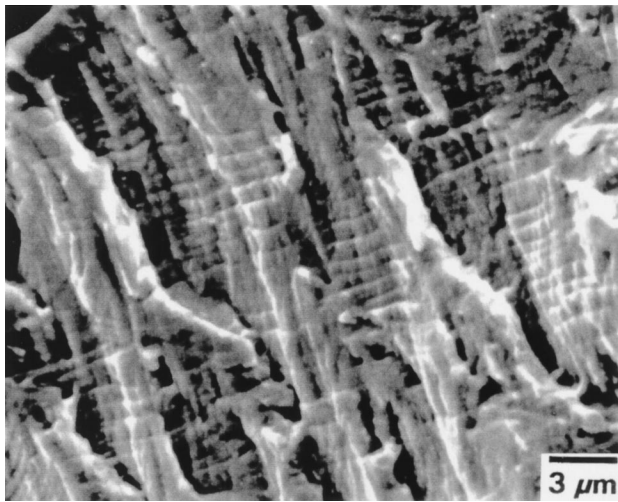


(b)

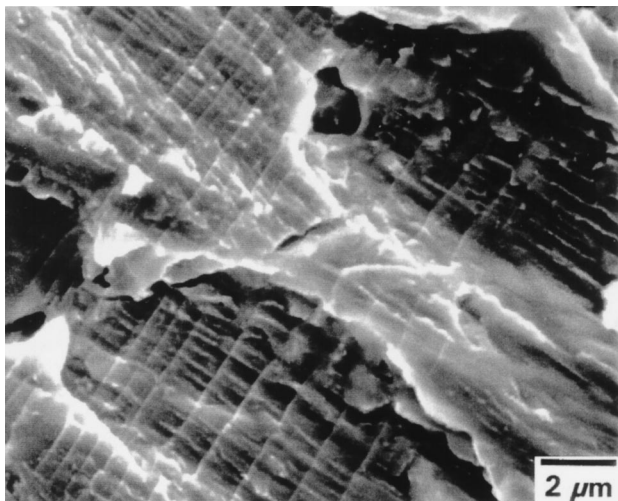


(c)

Fig. 13—SEM fractographs of hydrogen-precharged EN82H weld. (a) Intergranular cracking. (b) Intergranular fracture (top and bottom half) surrounding transgranular island with facets and poorly defined dimples. (c) Crystallographic facets within transgranular region.



(a)



(b)

Fig. 14—SEM fractographs of EN82H welds showing planar slip offsets superimposed on transgranular facets. (a) Nonprecharged specimen tested in 54 °C water. (b) Hydrogen-precharged specimen tested in water.

$\sigma_H$  = hydrostatic stress,  
 $\bar{V}$  = partial molar volume of hydrogen in solid solution,  
 $R$  = universal gas constant, and  
 $T$  = absolute temperature.

The term  $\bar{V}$ , defined as the change in volume produced by the addition of 1 gram-mol of hydrogen into the metal, is estimated to be 1.8 cm<sup>3</sup>/mol based on lattice-parameter measurements for solutions of hydrogen in nickel and other fcc metals.<sup>[14,15]</sup> The peak hydrostatic stress is estimated from References 16 and 17, which reported that the maximum normal stress ( $\sigma_{YY}$ ) inside the plastic zone is essentially independent of  $K_I$ . The effect of increasing  $K_I$  is to expand the peak stress over a wider area and locate the peak stress region farther from the crack tip. For materials with extensive strain-hardening capabilities, the magnitude of the peak stress is about 5 times the yield strength. Assuming plane-strain conditions, the peak hydrostatic stress is about 50 pct of the peak normal stress or 2.5 times the yield strength ( $\sigma_{YS} \approx 450$  MPa). Based on Eq. [4], the peak  $C/C_0$  ratio at 24 °C is 2.3. Thus, the precharged specimen (with 56 ppm

hydrogen) showing grain-boundary embrittlement has a peak hydrogen concentration on the order of 130 ppm.

Based on the rapid test results, susceptibility to intergranular cracking starts when local hydrogen levels reach ~50 ppm, although this concentration does not fully embrittle grain boundaries. Increasing the peak hydrogen concentration to 130 ppm during low-rate testing severely embrittles the grain boundaries and produces the same reduction in  $K_{Pmax}$  that is observed for nonprecharged specimens tested in water. Hence, the critical local concentration required to produce LTCP in EN82H appears to be about 130 ppm, assuming that cracking events nucleate near the peak triaxial stress location. This value is slightly higher than the 50 to 100 ppm hydrogen required to cause LTCP in alloy X-750,<sup>[12]</sup> which indicates that alloy X-750 grain boundaries are slightly more susceptible to hydrogen embrittlement. In alloy X-750, phosphorus and sulfur segregation to grain boundaries is believed to be responsible for the increased susceptibility to hydrogen embrittlement.<sup>[12]</sup>

#### E. Hydrogen-Precharged Alloy 600

Testing of hydrogen-precharged specimens suggests that the superior LTCP resistance of alloy 600, relative to EN82H, is attributed to (1) greater resistance to hydrogen pickup and (2) grain boundaries that resist intergranular cracking. These tests show that alloy 600 is susceptible to hydrogen embrittlement when peak hydrogen concentrations in the lattice reach ~80 ppm; however, the toughness degradation is not as severe as in EN82H and it involves a transgranular cracking mode. Moreover, nonprecharged specimens tested in water show no evidence of this embrittlement mechanism, indicating that alloy 600 tends to pick up relatively little hydrogen from the water. Specific details concerning precharged-specimen behavior are provided subsequently.

While hydrogen precharging has relatively little effect on  $K_{Pmax}$ , as shown in Figure 12, it does significantly degrade the material's ability to plastically deform. Comparison of load-displacement records in Figures 4(a) and 4(b) demonstrates that the precharged specimen tested in air (specimen M1551E) exhibits less total plastic displacement than the nonprecharged specimen tested in 54 °C water (B60J-1). Thus, a bulk hydrogen concentration of 44 ppm, which corresponds to a peak concentration of ~80 ppm ahead of a crack (per Eq. [4]), causes a large reduction in plastic-deformation capabilities. The hydrogen-induced loss in ductility is associated with a transgranular faceting mechanism (Figure 15), rather than intergranular cracking. Thus, alloy 600 grain boundaries are immune to cracking at a peak hydrogen content on the order of 80 ppm for the internal case. On the other hand, this hydrogen level is sufficient to induce a crystallographic faceting mechanism. The absence of faceting on fracture surfaces of nonprecharged specimens tested in 54 °C water suggests that local hydrogen levels were much less than 80 ppm; hence, alloy 600 appears to be more resistant to hydrogen pickup than EN82H welds. This is consistent with the observation that precharging under identical conditions produced bulk hydrogen concentrations of 42 to 45 ppm for alloy 600 and 55 to 57 ppm for EN82H. The apparent higher hydrogen solubility exhibited by the weld metal is believed to be associated with a higher chromium content (16 pct Cr for alloy 600 vs 20 pct Cr for EN82H) and the development of a cold-worked structure

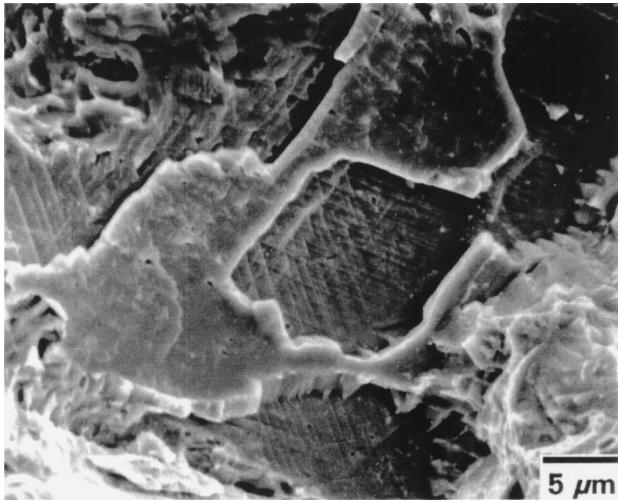


Fig. 15—Fracture surface of hydrogen-precharged alloy 600 showing slip offsets superimposed on crystallographic facets.

during welding. Symons<sup>[18]</sup> found that the hydrogen solubility in Ni-Cr-Fe alloys increases significantly when the chromium content is increased above 15 pct. This finding is consistent with the current observation that the EN82H weld with ~20 pct chromium possessed increased hydrogen solubility in comparison to alloy 600. In another study, Symons *et al.*<sup>[19]</sup> reported that increasing amounts of cold work enhance the hydrogen-trapping energy at carbides in both alloy 600 and alloy 690. It is possible, therefore, that the cold-worked structure induced in EN82H during welding may increase the hydrogen-trapping energy of intergranular carbides, which would also increase the hydrogen solubility.

The facets found in precharged alloy 600 have a very crisp crystallographic appearance and often contain planar-slip offsets (Figure 15). There is limited evidence of poorly defined void sheets, but these shallow voids have a faceted appearance. The combination of crystallographic facets and planar-slip offsets indicates that hydrogen promotes heterogeneous slip and localized failure along the slip bands, consistent with the EN82H behavior. In the welds, this mechanism is usually pre-empted by intergranular cracking, except in isolated regions that resist grain-boundary separation. By contrast, alloy 600 grain boundaries are immune to cracking, at least at ~80 ppm hydrogen, so the crystallographic faceting mode is dominant.

Testing the precharged specimen in water causes a very slight decrease in  $K_{P_{max}}$  from 70 to 66 MPa $\sqrt{m}$  (Figure 12) and reduces the total plastic deformation (Figure 4(b)). The faceted fracture-surface morphologies for precharged specimens tested in air and water are essentially identical, with the exception of fewer and more-poorly defined void sheets in the latter. Apparently, the combined effect of precharged hydrogen and hydrogen from the water is to further promote planar slip and separation along planar slip bands.

Rapid loading of the precharged alloy 600 specimen increases  $K_{P_{max}}$  to 89 MPa $\sqrt{m}$  (Figure 12) and restores significant ductility (specimen M1553E in Figure 4(b)). The high  $K_{P_{max}}$  value is consistent with the enhanced toughness observed in ductile materials when subjected to semidynamic and dynamic loading rates.<sup>[20]</sup> Under rapid loading, the precharged specimen exhibits a ductile tearing morphology, consisting of large primary dimples surrounded by void

sheets, that is essentially identical to that for nonprecharged specimens. Thus, a hydrogen concentration of 44 ppm is insufficient to induce planar slip and a faceted cracking mode. Local hydrogen levels must be increased by diffusion in order to cause a transition to planar slip.

#### IV. CONCLUSIONS

The fracture-toughness behavior of alloy 600 and EN82H welds was characterized in 54 °C to 338 °C air and hydrogenated water. The results of this study are given as follows.

1. Alloy 600 displays exceptionally high fracture toughness in air and high-temperature water. While low-temperature water causes a modest reduction in  $J_{IC}$  and the tearing modulus, the overall tearing resistance remains high. The superior toughness in air and water, regardless of temperature, demonstrates that fracture is not a design or operational issue for alloy 600.
2. The high toughness exhibited by alloy 600 is attributed to the ductile tearing mechanisms that are operative under all test conditions. These mechanisms include microvoid coalescence, void-sheet formation, and shear-stretch formation.
3. The high fracture toughness exhibited by EN82H welds in air and high-temperature water demonstrates that fracture is not a concern under these conditions. Fracture is a concern, however, in low-temperature water, as the  $J_{IC}$  values and tearing moduli are reduced by one to two orders of magnitude. Susceptibility to LTCP is only an issue below 149 °C.
4. For the most embrittled welds, LTCP initiates under linear-elastic conditions and near the maximum load; hence  $K_{P_{max}}$  values can be used to quantify the limiting response in design and operational assessments. The lower-bound  $K_{P_{max}}$  values for longitudinal and transverse welds are 40 and 48 MPa $\sqrt{m}$ , respectively.
5. Increasing the loading rates above 1000 MPa $\sqrt{m/h}$  significantly improves the cracking resistance of EN82H welds in low-temperature water, because there is insufficient time to fully embrittle the material ahead of a crack. At a rate of 26,000 MPa $\sqrt{m/h}$ , the fracture toughness is fully restored.
6. The dramatic reduction in toughness in low-temperature water is associated with a fracture-mechanism transition from ductile tearing to intergranular cracking. In air and high-temperature water, tremendous amounts of plastic deformation are required to initiate and grow primary microvoids and create shear-stretch regions. In low-temperature water, intergranular cracking is dominant, and there are isolated islands of crystallographic faceting.
7. The degree of embrittlement and cracking mechanisms observed in low-temperature water were reproduced in hydrogen-precharged specimens tested in air. This demonstrates that LTCP is a hydrogen-embrittlement mechanism, where hydrogen from the water weakens grain boundaries so they fail at low strain levels. In addition, hydrogen promotes a planar slip mechanism that localizes strain concentrations along grain boundaries. The critical hydrogen content in the lattice to produce LTCP in EN82H welds is on the order of 130 ppm.

## ACKNOWLEDGMENTS

This work was performed under United States Department of Energy Contract No. DE-AC11-98PN38206 with Bettis Atomic Power Laboratory. The authors acknowledge J.R. Suty, J.R. Chalfant, S.A. Derry, and R.D. Wineland for performing environmental tests and B.J. Whitmore and R.K. Ramaley for performing SEM fractographic examinations. Appreciation is also extended to D.M. Symons for the enlightening discussions.

## REFERENCES

1. P. Saint Paul and G. Slama: *Proc. 5th Int. Symp. on Environmental Degradation of Materials in Nuclear Power Systems—Water Reactors*, ANS, La Grange Park, IL, 1992, pp. 39-49.
2. D. Alter, Y. Robin, M. Pichon, A. Teissier, and B. Thomeret: *Proc. 5th Int. Symp. on Environmental Degradation of Materials in Nuclear Power Systems—Water Reactors*, ANS, La Grange Park, IL, 1992, pp. 661-666.
3. D. Buisine, F. Cattant, J. Champredonde, C. Pichon, C. Benhamou, A. Gelpi, and M. Vaindirlis: *Proc. 6th Int. Symp. on Environmental Degradation of Materials in Nuclear Power Systems—Water Reactors*, TMS, Warrendale, PA, 1993, pp. 845-51.
4. C.M. Brown and W.J. Mills: *Corrosion*, 1999, vol. 55, pp. 173-86.
5. D.M. Symons, J.P. Foster, and M.G. Burke: *Proc. 8th Int. Symp. on Environmental Degradation of Materials in Nuclear Power Systems—Water Reactors*, ANS, La Grange Park, IL, 1997, vol. 1, pp. 237-48.
6. W.J. Mills: *J. Testing Evaluation*, 1981, vol. 9, pp. 56-62.
7. P.C. Paris, H. Tada, A. Zahoo, and H. Ernst: *Elastic-Plastic Fracture*, ASTM STP 668, ASTM, Philadelphia, PA, 1979, pp. 5-36.
8. J.A. Begley and J.D. Landes: *Fracture Toughness, Proc. 1971 Nat. Symp. on Fracture Mechanics, Part II*, ASTM STP 514, ASTM, Philadelphia, PA, 1972, pp. 1-20.
9. W.C. Porr and W.J. Mills: "Application of the Normalization Data Analysis Technique for Single Specimen R-Curve Determination," Bettis Atomic Power Laboratory, West Mifflin, PA, Report No. B-T-3269, Feb. 1999.
10. H.C. Rogers: *Ductility*, ASM, Cleveland, OH, 1968, pp. 31-61.
11. T.B. Cox and J.R. Low, Jr.: *Metall. Trans.*, 1974, vol. 5, pp. 1457-70.
12. W.J. Mills, M.R. Lebo, and J.J. Kearns: *Metall. Mater. Trans. A*, 1999, vol. 30A, pp. 1579-96.
13. W.W. Gerberich and Y.T. Chen: *Metall. Trans.*, 1975, vol. 6, pp. 271-78.
14. B.B. Baranowski, S. Majchrzak, and T.B. Flanagan: *J. Phys. F: Metal Phys.*, 1971, vol. 1, pp. 258-61.
15. M.L. Wayman and G.C. Smith: *J. Phys. Chem. Solids*, 1971, vol. 32, p. 103.
16. J.R. Rice and M.A. Johnson: *Inelastic Behavior of Solids*, McGraw-Hill, New York, NY, 1970, pp. 641-72.
17. R.M. McMeeking: *J. Mech. Phys. Solids*, 1977, vol. 25, pp. 357-81.
18. D.M. Symons: *Metall. Mater. Trans. A*, 1997, vol. 28A, pp. 655-63.
19. D.M. Symons, G.A. Young, and J.R. Scully: *Metall. Mater. Trans. A*, 2001, vol. 32A, pp. 369-77.
20. W.J. Mills: *Int. Mater. Rev.*, 1997, vol. 42, pp. 45-82.