Fracture Toughness of HY-130 Steel Weld Metals

Welding process and selection of welding conditions can significantly affect the fracture resistance of HY-130 steel weld metals

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ABSTRACT. The fracture toughness of HY-130 steel weld metals and base metal was investigated using 1T compact tensile tests. The purpose of the study was to determine the effect of various welding procedures on the fracture toughness of as-deposited HY-130 steel weld metals compared to rolled base metal.

Mechanical property, Charpy V-notch energy, hardness and J_{IC} computer interactive fracture toughness tests along with metallography and scanning electron microscopy were performed on gas metal arc (GMA) welds and gas tungsten arc (GTA) welds and also on rolled base metal.

The HY-130 weld metals deposited by GMAW had lower crack initiation energy, J_{IC}, values and tearing resistance, T, than the HY-130 steel base metal. This was attributed primarily to the higher oxygen content of the weld metals. The microstructure of the weld metals deposited by GMAW also had an effect on toughness. The HY-130 weld metals deposited by GTAW had superior fracture toughness compared to HY-130 base metal. Low heat input GTA welding conditions produced the highest fracture resistance. This was attributed to the repeated reheating, refining and tempering of the weld metal microstructure during the fabrication of these welds.

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Introduction

The need for high strength, high toughness and stress corrosion cracking resistant marine structural steels is of current interest (Ref. 1). Candidate structural steels must also exhibit suitable weldability. HY-130, a 5Ni-Cr-Mo-V quenched-and-tempered steel, is currently under investigation as a steel which could meet these requirements. Initial development and weldability of this steel has been reported by Rathbone *et al.* (Ref. 2).

Recently, Zanis *et al.* (Ref. 3) reported on the seawater subcritical cracking behavior of HY-steel weldments. They found that welding parameters, which effectively reheat, refine, and temper the weld metal microstructure, also improve the resistance to susceptibility to subcritical cracking (SCC) in seawater. In particular, a weldment fabricated by the gas tungsten arc (GTA) process using a low deposition low heat input procedure produced a weld metal with superior SCC resistance compared to gas metal arc (GMA) weldments and high deposition, high heat input GTA weld metals. It is also of interest to ascertain whether welds produced by this process exhibit a similar improvement in the fracture toughness.

The objective of this paper is to present the results of an investigation to determine the J_{IC} fracture toughness parameter for HY-130 steel weld metals deposited by GTA and GMA welding and HY-130 base metal. Since the base and weld metal for HY-130 behave in an elastic-plastic manner for the material thickness under investigation, the J_{IC} fracture toughness criteria was selected.

Experimental Procedure

Materials

The HY-130 steel base metal was from a 1 in. (25 mm) thick electric furnace (EF) plate. The plate was in a quenched-and-tempered condition. The microstructure is tempered martensite with a prior austenite average grain size of 12.5 μm. The chemical composition of the base metal (designated as HY-130/EF) is given in Table 1.

The chemical compositions of the HY-130 weld metals which were fabricated from standard filler metals are also listed in Table 1.

Weldment Processes

The welding conditions used to make butt joints featuring 1.5 in. (37.5 mm) thick, balanced double V-groove welds are listed in Table 2. The GMA welds, designated HY-130/D and HY-130/C, were fabricated using standard and high heat input procedures, respectively. The GTA weld, designated HY-130/AB, was fabricated at a high heat input with about the same number of weld beads as the GMA weld HY-130/D. The GTA weld, designated HY-130/Y, was manufactured using low heat input, low deposition procedures (*i.e.*, 111 weld beads) to achieve the best weld metal mechanical properties (strength and toughness) possible in the HY-130 system. All weldments were radiographed to ensure lack of significant discontinuities.

Test Methods

Duplicate standard 0.252 in. (6.3 mm)

Table 1—Chemical Analysis of Materials

Material code	Condition[b]	Composition, %[a]												
		C	Mn	P	S	Si	Cu	Ni	Cr	Mo	V	Ti	Al	N	O
HY-130/EF	Base Metal[c]	0.11	0.76	0.005	0.004	0.31	0.02	5.00	0.42	0.53	0.043	0.008	0.021	0.011	0.0032
HY-130/D	GMAW-46 kJ/in.	0.11	1.32	0.004	0.003	0.32	0.04	2.81	0.60	0.84	0.010	0.016	0.006	0.006	0.0167
HY-130/C	GMAW-81 kJ/in.	0.11	1.32	0.004	0.003	0.32	0.04	2.81	0.60	0.84	0.010	0.016	0.006	0.006	0.0167
HY-130/AB	GTAW-65 kJ/in.	0.10	1.77	0.006	0.005	0.36	0.06	2.56	0.98	0.60	0.021	0.012	0.012	0.008	0.0024
HY-130/Y	GTAW-34 kJ/in.	0.11	1.47	0.007	0.005	0.33	0.05	2.78	0.82	0.60	0.030	0.012	0.012	0.007	0.0017

(a) Average of duplicate analyses.
(b) For welding parameters see Table 2.
(c) From published sources (Refs 10, 11).

Table 2—Welding Conditions

Material code	Process	Heat input, kJ/in.	Filler metal feed rate, ipm	Voltage, V	Current, A	Travel speed, ipm	Shielding gas	Preheat interpass temperature, °F	No. of beads
HY-130/D	GMAW	46	–	28	325	12	Ar + 2% O_2	275-300	36
HY-130/C	GMAW	81	–	30	360	8	Ar + 2% O_2	375-400	15
HY-130/AB	GTAW	65	73	17	380	6	75% Ar + 25% He	175-200	31
HY-130/Y	GTAW	34	19	14	240	6	75% Ar + 25% He	275-300	111

diameter all-weld-metal tensile specimens and 0.505 in. (12.5 mm) diameter transverse base metal specimens were tested at room temperature. Triplicate transverse Charpy V-notch (CVN) tests at 0° F (−18° C) were conducted on all weld metals and on T-L orientation base metal specimens. Rockwell C hardness measurements were taken on all specimens.

All J_{IC} fracture toughness tests were performed utilizing the computer interactive unloading compliance test procedure of Joyce and Gudas (Ref. 4). Their approach allows for on-line, real time collection and analysis of digitized load and displacement data. J-integral tests are carried out by performing a series of approximately 10% unloadings during the course of a fracture mechanics type test. From compliance measurements, instantaneous values of crack length and change in length are determined. J_i is calculated according to the expression (Ref. 5):

$$J_{(i+1)} = \left[J_i + \left(\frac{\eta}{b} \right)_i \frac{A_{i,\, i+1}}{B_N} \right] \left[1 - \left(\frac{\gamma}{b} \right)_i (a_{i+1} - a_i) \right] \quad (1)$$

where $\eta = 2 + (0.522)\, b/W$ for compact specimens, W = specimen width, $\gamma = 1 + (0.76)\, b/W$, b_i = instantaneous length of remaining ligament, B_N = minimum specimen thickness, a_i = instantaneous crack length, and $A_{i,\, i+1}$ = area under the load vs. load line displacement record between lines of constant displacement at points i and i + 1.

The compliance formula used for the calculations in this procedure was that of Saxena and Hudak (Ref. 6). All tests were performed at conventional loading rates, with a maximum crosshead displacement rate of 0.25 mm/min (0.01 ipm).

The specimens were modified 1T compact tensiles (1TCT) with the geometry given in Fig. 1. The weld metal specimens were side grooved 20%. The 20% side groove value is that necessary to guarantee saturation of the tearing modulus as determined by Gudas and coworkers (Ref. 7). All specimens were fatigue precracked under conditions specified in ASTM E399 (Ref. 8) to an initial crack length to specimen width ratio (a/W) of 0.65 to 0.70.

For 1TCT specimens, J_{IC} values were computed from the intersection of the crack opening stretch line ($J = 2\sigma_F \cdot \Delta a$) with the least-squares fit of data points which fell at least 0.15 mm (0.006 in.) beyond the blunting line and did not exceed 1.5 mm (0.06 in.) in crack growth from that point. Tearing moduli were calculated according to the following expression (Ref. 9) using the same range of crack extension:

$$T = \frac{dJ_I}{da} \cdot \frac{E}{\sigma_F^2} \quad (2)$$

where T = tearing modulus = dimensionless, dJ_I/da = least squares linear regression slope of J_I vs. crack extension curve, E = elastic modulus = 29×10^6 psi, and σ_F = flow stress = $(\sigma_{YS} + \sigma_{UTS})/2$ with YS and UTS standing for yield strength and ultimate tensile strength, respectively.

At the conclusion of testing, specimens were heat-tinted at 698° F (370° C) for 30 minutes (min) to mark the extent of crack growth. After breaking open at liquid nitrogen temperature, the crack length and crack extension were measured at nine equally spaced points across the crack front, including the two surfaces. An average value of the two surface measurements was used as a single point in computing crack extension and crack growth.

Fractography

The specimens were also prepared for macrofractographs and scanning electron microscopy (SEM) microfractography. All SEM fractography was performed using stereo pairs.

Results and Discussion

Chemical Composition

Table 1 presents the results of chemical analysis on the HY-130 steel base metal and weld metals. The base metal data are from the literature (Refs. 10, 11). It is noted that the compositions of the weld metals differ slightly from that of the 5Ni-Cr-Mo-V base metal.

The weld metals were essentially low carbon, low alloy steels, which use Ni, Mn, Cr and Mo for strengthening and to control solid state transformations. Nickel is a moderate weld metal strengthener and is added primarily to increase toughness. Manganese is reportedly a good hardenability addition, which improves toughness and is a mild deoxidant (Ref.

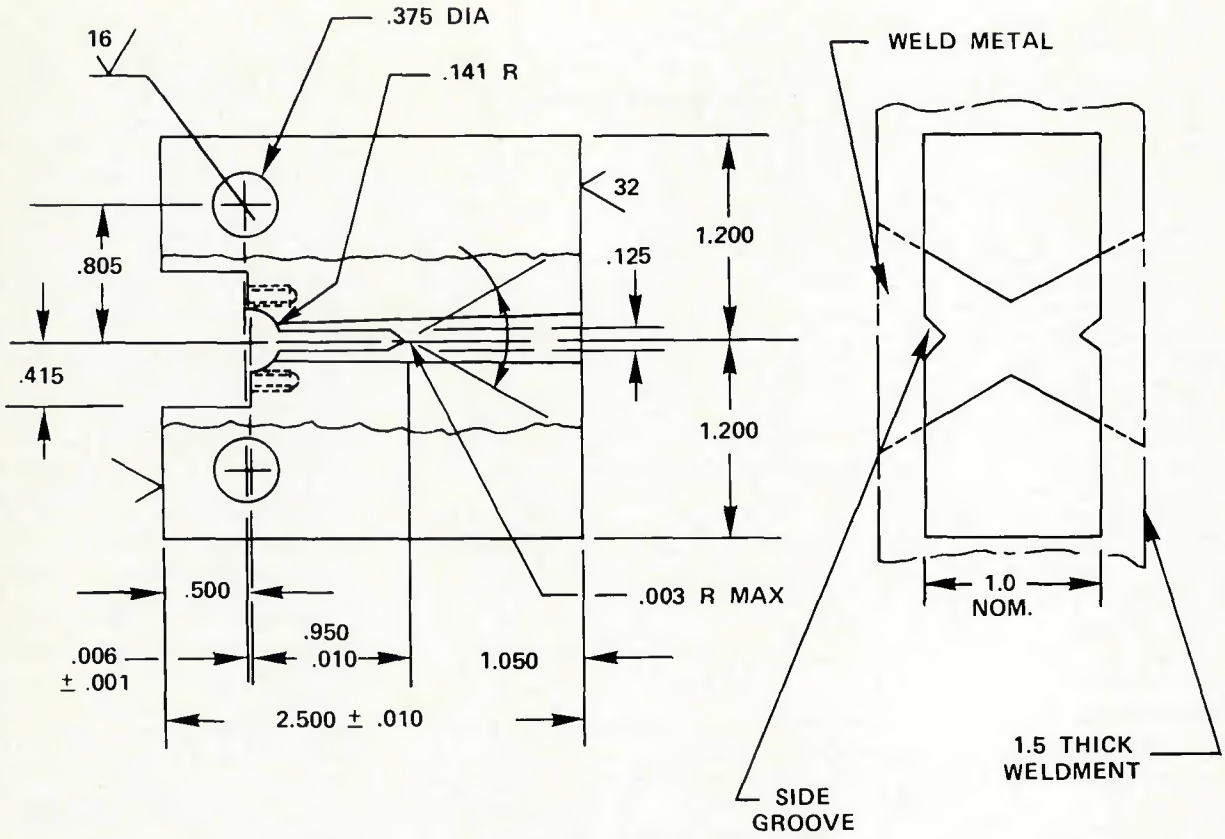

Fig. 1—Modified compact specimen (1TCT) for J-integral testing. Dimensions are shown in inches (1 in. = 25.4 mm)

12). Chromium and molybdenum increase hardenability and prevent softening of weld metals by subsequent passes. The impurity elements (P and S) are low in the base metal.

These results are reflected in favorable mechanical properties—Table 3. As expected, there was a lower total oxygen concentration in the GTA welds. This is attributed to the replacement of oxygen in the shielding gas with helium—Table 2.

Mechanical Properties

The results of tensile, CVN and hardness tests on the base metal and weld metals are presented in Table 3. The data for plate are from earlier work (Refs. 10, 11).

The strengths of all the weld metals listed in Table 3 are equal to or greater than the base metal. The tensile elongation of the weld metals was generally lower than that for the base metal. However, reduction of area values were comparable. The CVN impact energy at 0° F (−18° C) (upper shelf based on macrofractographic observations) of the GMA weld metal was comparable to the base metal. It was noted that weld metal deposited by high heat input GMAW (HY-130/C) had lower strength and a higher CVN impact energy than the other weld metals deposited by GMAW.

As expected, the weld metals deposited by low oxygen GTAW exhibited higher CVN impact energy values than weld metal deposited by GMAW. It was noted that the weld metal deposited by fine bead low deposition GTAW (HY-130/Y) had the highest CVN impact energy values and exceeded GMA welds by almost a factor of two.

Microstructures

The microstructure of the HY-130 plate, as mentioned previously, was primarily tempered martensite, with a prior austenite grain size of 12.5 μm (Refs. 10, 11). It has been reported that the HY-130/D weld metal contained a mixture of martensite and bainite with substantial twinned and untempered martensite and some retained austenite (Ref. 13).

In the tests described in this paper, the microstructure of the weld metal deposit-

Table 3—Mechanical Properties of HY-130 Base Metal and Welds[a]

Material/ code	Condition	0.2% YS,[b] ksi (MPa)	UTS,[b] ksi (MPa)	Elongation, % in 4D	Reduction of Area, %	CVN[b] at 0° F (−18°C), ft-lb (J)	Hardness range, RC
HY-130/EF[c]	Base metal	130 (909)	139 (958)	27	63	75 (102)	31
HY-130/D	GMAW, 46 kJ/in.	144 (992)	153 (1054)	19	68	66 (90)	32-38
HY-130/C	GMAW, 81 kJ/in.	126 (847)	140 (965)	22	70	78 (106)	28-34
HY-130/AB	GTAW, 65 kJ/in.	145 (999)	152 (1047)	20	68	92 (125)	29-37
HY-130/Y[d]	GTAW, 34 kJ/in.	147 (1014)	150 (1034)	22	74	141 (191)	32-37

(a) Average of 3 tests.
(b) YS—yield strength; UTS—ultimate yield strength; CVN—Charpy V-notch.
(c) T-L orientation (Refs. 10, 11).
(d) Upper yield point—149 ksi (1027 MPa).

ed by high heat input GMAW (HY-130/C) was predominantly bainite with some martensite. The coarse bead HY-130 GTA weld (HY-130/AB) was macroscopically similar to the GMA welds and had a predominantly coarse martensitic structure. The macrostructure of the fine bead weld metal deposited by GTAW (HY-130/Y) consisted of "eyebrows" due to the numerous passes required to fabricate these low deposition rate welds. "Eyebrows" are dark etching boundaries which are produced as a heat effect in weld metal (Ref. 3); at the same time, eyebrows are also regions which represent intercritically reheated (partially austenitized) and/or heavily tempered heat-affected zones in the weld metal.

The microstructure of the bulk of the HY-130/Y weld metal was extremely fine and was identified as bainite and ferrite with patches of retained austenite (Ref. 13). There was no martensite except at the nonrefined surface and root locations.

J_{IC} Fracture Toughness

Table 4 summarizes the J_{IC} fracture toughness results of the 1TCT specimens. The flow stress, σ_F, is also presented. The data for the plate material are from published sources (Refs. 10, 11).

Crack extension, as explained in Test Methods, was an arithmetic average (*i.e.*, mean) of the actual crack extension. The crack extension was irregular in the GMA welds (Figs. 2A and B) and fairly uniform in the GTA welds (Figs. 2C and D). Areas where the crack extension protruded ahead of the uniform crack extension, as seen in the HY-130/D weld (Fig. 2A), could be due to residual stresses. The uniformity of crack extension in the coarse bead GTA weld (HY-130/AB) was the same as that observed in HY-130 plate (Ref. 10). Although there are presently no published standards on allowed variation in the crack extension, the present test is comparable to tests of plate material, with the possible exception of the HY-130/D specimen.

The Paris specimen thickness criterion values for a valid J_{IC} test (Ref. 14) also appear in Table 4 and are defined by:

$$B > 40\left(\frac{J_{IC}}{\sigma_F}\right) \qquad (3)$$

Since the specimen thickness was normally 1 in. (25 mm) for the base metal and 0.8 in. (20 mm) for the side grooved weld metal specimens, the Paris criterion was satisfied and all the results in Table 4 are valid.

In general, the data on Table 4 show that the crack initiation energy, J_{IC}, and resistance to tearing propagation T, of the weld metals deposited by GMAW was lower than that of HY-130 base metal. This is attributed to both the significantly higher total oxygen content of these weld metals (Table 1) and to microstructural differences. It has been reported that high oxygen content (150–200 ppm) can have a significant negative effect on the fracture toughness of weld metals (Ref. 15).

With regard to microstructure, the presence of significant twinned and untempered martensite in the HY-130/D weld metal may also be a contributor to the lower fracture toughness of this material compared to base metal. It was noted that the weld metal deposited by high heat input GMAW (HY-130/C) with a predominantly bainitic microstructure had comparable crack initiation energy to the HY-130 base metal. However, the tearing resistance of this weld metal was less than the base metal.

Although the weld metals deposited by GTAW had the same oxygen content as the base metal (Table 1), both materials (HY-130/AB and HY-130/Y) exhibited higher crack initiation energy. The tearing resistance, T, was comparable to base metal for the 65 kJ/in. GTA weld metal,

Table 4—J_{IC} Fracture Toughness Properties of HY-130 Base Metal and Welds

Material/code	Condition	σ_F ksi (MPa)	J_{IC} in-lb/in.^{2} (KPa/m)	T	B, in.
HY-130/EF[a]	Base metal	134.5 (927)	999 (175)	29.3	0.30
			919 (161)	27.7	0.27
HY-130/D	GMAW, 46 kJ/in.	148.5 (1024)	525 (92)	8.4	0.14
HY-130/C	GMAW, 81 kJ/in.	133 (977)	893 (156)	19	0.27
			1031 (181)	24	0.31
			782 (137)	28	0.24
HY-130/AB	GTAW, 65 kJ/in.	148.5 (1024)	1252 (219)	29	0.34
			1175 (203)	23	0.31
HY-130/Y	GTAW, 31 kJ/in.	148.5 (1024)	1990 (348)	32	0.54
			1583 (277)	36	0.43
			1710 (299)	44	0.46

(a) Data from published sources (Refs. 10, 11).

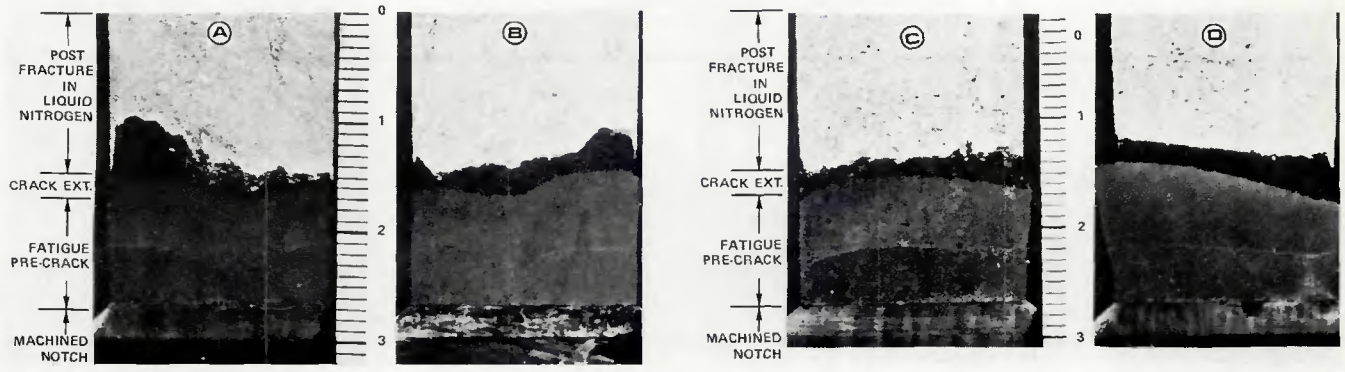

Fig. 2 – Typical macrofractographs of J_{IC} fracture toughness specimens: A – GMAW, 46 kJ/in.; B – GMAW, 81 kJ/in.; C – GTAW, 65 kJ/in.; D – GTAW, 34 kJ/in.

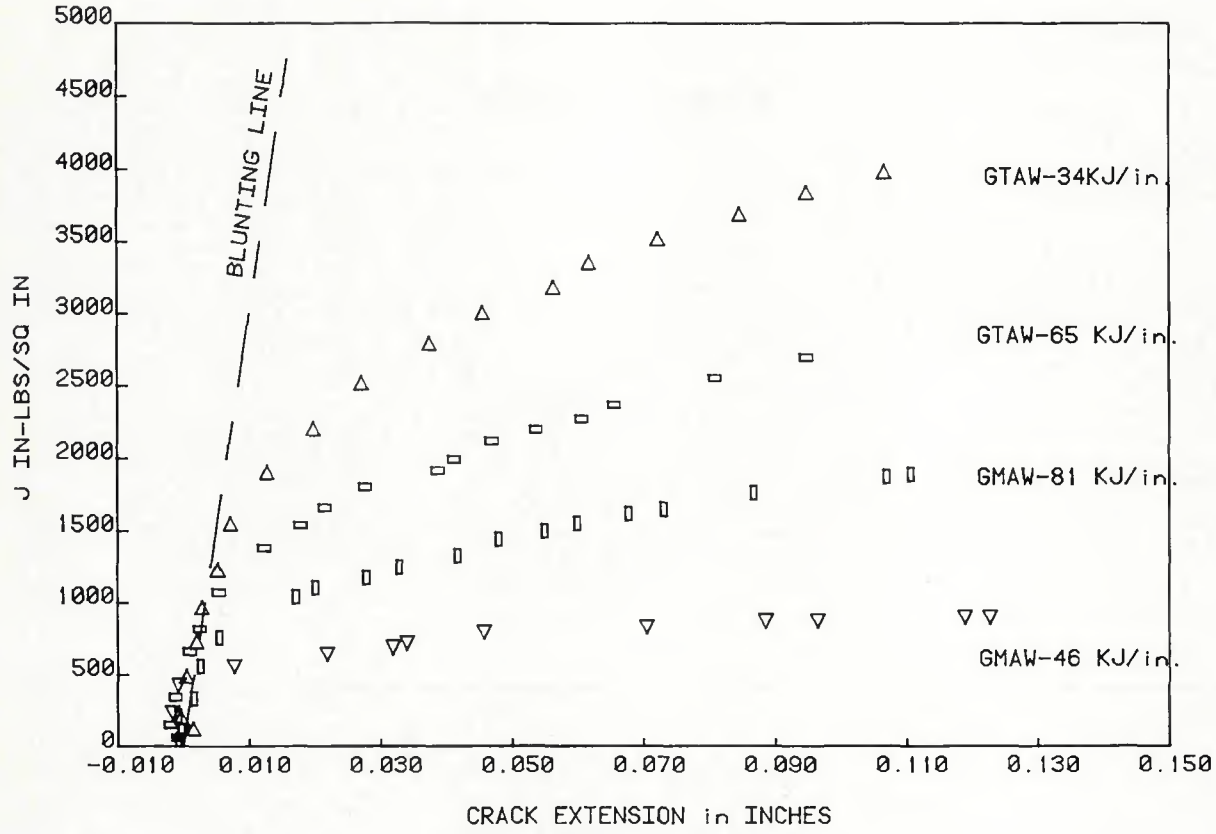

Fig. 3—J vs. crack extension of HY-130 welds

while the low heat input GTA weld (HY-130/Y) tearing resistance is greater than for base metal. A representative R-curve of J_I vs. crack extension for each process and heat input is illustrated in Fig. 3.

Figure 3 provides a direct comparison between the various HY-130 welding processes and procedures with regard to ductile fracture toughness. It also illustrates that both the fracture toughness and tearing resistance of HY-130 weld metal is significantly altered by welding process and procedure selection.

As indicated in Table 4, the fine bead, low heat input HY-130/Y weld metal had significantly higher fracture resistance than the coarse bead, higher heat input HY-130/AB weld metal. The superior fracture toughness properties of the HY-130/Y weld metal are attributed to the low heat input during welding (*i.e.*, amperage and filler metal feed rate).

The welding conditions for HY-130/Y, listed in Table 2, produced a higher heat affected zone-to-fusion zone area ratio than in the HY-130/AB weld metal. This resulted in a repeated reheating, refinement and tempering of the microstructure, and virtual elimination of coarse columnar grain structure and coarse untempered martensite in this material.

Figure 4 illustrates the effects of welding conditions on the microfracture of the weld metals deposited by GTAW. The high heat input HY-130/AB weld metal exhibited patches of low energy fractures. These regions of low energy fracture were not apparent in the low heat input HY-130/Y weld metal, which exhibited ductile fracture.

Conclusions

1. The HY-130 weld metals deposited by GMAW had lower crack initiation energy, J_{IC}, values and tearing resistance, T, than the HY-130 steel base metal. This was attributed primarily to the higher oxygen content of these weld metals. The microstructure of the weld metals deposited by GMAW also had an effect on toughness.

2. The HY-130 weld metals deposited by GTAW had superior fracture toughness compared to HY-130 base metal.

3. Low heat input GTA welding procedure produced the highest fracture resistance. This was attributed to the repeated reheating, refining and tempering of the weld metal microstructure during fabrication of these welds.

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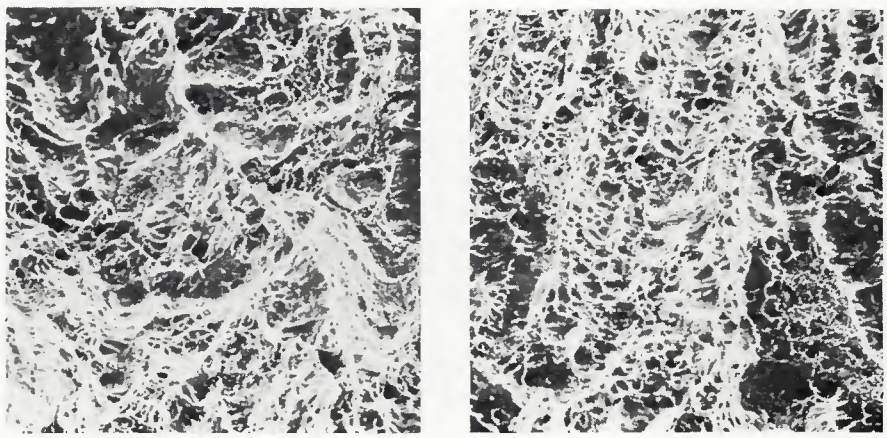

Fig. 4—SEM fractographs of GTA weld metals: Left—HY-130/AB; Right—HY-130/Y

gram which served as a basis for this work. The laboratory support of Mr. W. Umlandt is also acknowledged.

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