Microstructural Characterization of Submerged-arc and Gas–Metal-arc Weldments in HY-130 Steel

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The microstructure of two HY-130 steel weldments prepared by submerged-arc welding (SAW) and gas–metal-arc welding (GMAW) processes using identical filler wires were investigated. Light and electron microscopy along with microhardness traverses were used to characterize and compare the microconstituents present in the base metal, weld metal, and heat affected zone (HAZ) of both weldments. The base metal consisted of a bainite–martensite microstructure and was uniform through the plate thickness. The weld metal and HAZ structures consisted of a mixture of lath and twinned martensite, bainite, autotempered martensite, and retained austenite. The GMAW weld metal exhibited a finer lath martensite with more interlath retained austenite and transformation microtwins. The SAW weld metal structure was more bainitic and contained higher inclusion densities as a result of the flux used to make the joint. Microhardness traverses exhibited higher hardness values in the GMAW weld metal compared to the SAW. The HAZ microhardness followed a similar trend for both weldments and showed a maximum value in the fine-grained region near the base metal.

KEY WORDS: HY-130; gas–metal-arc welding; submerged-arc welding; heat-affected zone; weld metal; microstructure; microhardness.

1. Introduction

HY-130 is a low-carbon 5NiCrMoV steel developed for the U.S. Navy for applications requiring yield strength of 130 ksi (896 MPa), high toughness at low temperatures, and good weldability.¹⁾ The combination of high strength and resistance to brittle fracture is based on the development of a fine tempered martensitic–bainitic microstructure produced after a quench and temper treatment. The HY-130 system development includes specifications for rolled plate, forged, extruded and cast product forms, and the required filler metal and welding procedures for fabrication of high strength and high toughness structures.²⁾ Recently the alloy has been considered as a replacement for the lower strength HY-80 grade (yield strength 552 MPa) for the construction of submarine hulls. HY-130 steel can provide an increased strength-to-weight ratio and it is anticipated that weight savings of about 40 to 50 t will be achieved in certain submarine applications.³⁾

One of the major difficulties in designing steel structures is the lack of adequate knowledge of properties of the welded joints. In particular, the structures that are used at low temperature or those that are cyclically loaded are the most difficult to design, because brittle and/or fatigue cracking can initiate at localized regions in weld joints.^{4–6)} Thus, when crack initiation is the criterion used to prevent fracture, control of the microstructures

of the weld metal and heat affected zones can be critical.

The initial structure and composition of the base material, filler metal, and welding processes combine to determine the characteristics of each microstructural zone in welded joints. Although the microstructural changes occurring within the HAZ during cooling are similar to those taking place during hot rolling or heat treatment of steels, the conditions created by the welding process are significantly different from those employed in steel production. Very rapid local heating creates steep temperature gradients leading to non-equilibrium conditions. The unmelted base metal near the fusion line is strongly affected by the temperature gradient and the peak temperature at the fusion line. The temperature decreases sharply with increasing distance from the fusion line creating microstructurally distinct regions.

In the present work, the microstructures developed in the base metal, weld metal, and HAZ of two HY-130 steel weldments produced by multipass SAW and GMAW processes are discussed. The microstructures were characterized by light and electron microscopy and the mechanical properties across the welded joints were assessed by microhardness testing.

2. Experimental Procedure

2.1. Material and Weld Preparation

Two HY-130 steel plates approximately 25 mm thick

Table 1. Chemical compositions of the base plate, weld metal, and filler metal. (mass %)

Element	Base metal	Weld metal GMAW	Weld metal SAW	Filler metal
C	0.106	0.074	0.055	0.091
Mn	0.87	1.46	1.44	1.67
Si	0.22	0.3	0.37	0.39
P	0.007	0.006	0.014	0.005
S	0.002	0.005	0.001	0.01
Ni	4.4	2.39	2.4	2.23
Mo	0.42	0.62	0.62	0.63
Cr	0.51	1.05	0.99	1.05
V	0.058	0.014	0.02	0.011
Al	0.019	0.003	0.014	0.004
Ti	0.002	0.001	0.004	0.01
Zr	0.001	0.001	0.001	0.003
Cu	0.057	0.046	0.046	0.045
O	0.006	0.019	0.028	0.003
N	0.0008	0.0046	0.005	0.003
B	0.004	0.005	0.003	0.004
H	1.4 ppm	1.4 ppm	0.4 ppm	2 ppm

Table 2. Welding conditions employed.

	SAW	GMAW
Preheat/interpass temperature (°C)	94	121
Voltage (V)	34	27
Current (A)	300	340
Travel speed (mm/s)	7.4	4.6
Heat input (kJ/mm)	1.38	1.97
Shielding gas		98 % argon + 2 % oxygen
Flux	Operlikon* OP12177	—

* Basic Flux—Basicity Index 3.1.

were welded by standard multipass gas-metal-arc welding (GMAW) and submerged-arc welding (SAW) processes. The chemical composition of the base plate, the filler metal and the weld metal after each process is given in **Table 1**. The weld preparation for each plate was a single beveled 60 degree V joint with a backing plate tack-welded to the base of each plate at the joint. Both welds were made using the same electrode material. The welding conditions for each process are summarized in **Table 2**. The welded plates were sectioned perpendicular to the weld direction, surface ground, polished, and etched in 2 % Nital for 10 min to reveal details of each weld pass as illustrated in **Fig. 1**.

2.2. Microhardness Measurements

Samples were cut, mounted, polished and etched to locate regions of interest. To facilitate microhardness measurements, lines were scribed across each selected region with a diamond scribe (lines S1 through S6 in **Fig. 2**). A Buehler Micromet microhardness tester with a diamond indenter and a 300 g load was used to take the measurements. Indentations were made at 0.2 mm intervals along the linear scribe, with smaller (0.1 mm) intervals in the vicinity of the fusion line.

Fig. 1. Macrosections of the (a) SAW and (b) GMAW weldments showing the various weld passes.

Fig. 2. Schematic representation of (a) SAW and (b) GMAW weldments showing locations of microhardness traverses.

2.3. Microstructural Characterization

The microstructure was characterized by light and transmission electron microscopy (TEM). Specimens for light microscopy were prepared by standard metallographic techniques and etched in 2 % Nital. Thin foils for TEM were prepared by cutting slices approximately

0.3 mm thick perpendicular to the welding direction and thinned to about 0.15 mm by gentle abrasion on wet silicon carbide paper. Discs of 3 mm diameter were then punched and ground to 0.07 mm thickness, followed by jet polishing in a 10% perchloric acid–90% glacial acetic acid solution. The foils were examined with a JEOL-100CX transmission electron microscope operated at 120 kV.

3. Results

3.1. Microhardness Measurements

The microhardness profiles through the weld metal, Fig. 3, indicated that the GMAW was generally harder than the SAW weld metal. The average hardness of the last weld pass was 371 VHN for the GMAW and 333 VHN for the SAW weldment. Hardness drops of 20 to 60 VHN typically occurred as the traverses crossed the weld heat affected zone (points a, a' through c, c') of one pass to a previous pass.

The microhardness traverses across the HAZ, Fig. 4,

Fig. 3. Microhardness traverses across the center of the weld metal. Points a, b, c, a', b', c' are the fusion boundaries between weld passes.

Fig. 4. Microhardness profiles across the HAZ of a single weld pass. Regions A, A': weld metal; B, B': coarse HAZ; C, C' and D, D': fine HAZ; E, E': base metal. Weld metal is shown in negative side; HAZ is between 0 and 4 mm from the fusion zone.

showed that both processes resulted in the same hardness trend in the similar single-pass region. The hardness drops at the fusion line for the GMAW, and then sharply increases through the HAZ displaying two distinct regions. The first region (between 0 and 1 mm from the fusion zone) corresponds to the coarse HAZ, while the second region (1 to 3.4 mm from the fusion zone) reveals the fine HAZ. It is important to note that hardness reaches a maximum value for both processes in the fine

Fig. 5. Microhardness profiles across the HAZ affected by multiple weld passes. Weld metal is shown in negative side; HAZ is between 0 and 4 mm from the fusion zone.

Fig. 6. TEM micrographs of the base plate showing: a) dislocated lath martensite and bainite; b) autotempered martensite.

HAZ, sharply decreases at the end of the visible HAZ, and reaches a value of approximately 317 VHN in the base metal. The GMAW weldment is slightly harder throughout the HAZ reaching a maximum value of 406 VHN compared to a maximum of 385 VHN for the SAW process. It is coincidental that the width of the HAZ is approximately 3.4 mm for both processes at the locations marked by the S2 and S5 scribes. In general, the width of the HAZ varies with location in the HAZ and the angle of the scribe.

Microhardness traverses across the scribes S3 and S4 are shown in Fig. 5. The hardness in both processes decreases at the fusion line, increases at about the same slope throughout the HAZ, and decreases to a nominal value of 316 VHN in the base metal. The maximum hardness in the fine-grained HAZ was 397 VHN for the GMAW and 367 VHN for the SAW process.

Four regions characterized by the microhardness trends of Figs 3 and 4 were selected for microstructural investigation. These regions are:

- weld metal (weld beads #21 and #40, Fig. 1; regions A, A', Fig. 4)
- coarse HAZ (regions B, B', Fig. 4)
- fine HAZ (regions C, C', and D, D', Fig. 4)
- base metal (regions E, E', Fig. 4)

3.2. Microstructural Characterization

Light microscopy did not provide sufficient information to characterize the microstructure of the weldments as the prior austenite grain size, inclusion distribution,

laths and interlath elements were not clearly resolved. The lath morphology and microconstituents present in the base metal, weld metal, and HAZ were more closely compared using the TEM. Representative micrographs are shown in Figs. 6 to 11.

The microstructure of the base plate away from the fusion line consisted of lath martensite, bainite, and autotempered martensite as shown in Fig. 6. The dislocation density within the fine martensite laths were high although the steel was tempered in the 520 to 600°C temperature range. The micrograph presented in Fig. 6(b) highlights the temper carbides within a large martensite lath and coarse cementite particles (C) precipitated along the bainite-martensite boundary.

Thin foil electron micrographs from the coarse-grained HAZ adjacent to the fusion line were very similar for both the SAW and GMAW weldments, Fig. 7. The microstructure was predominantly lath martensite, with large amounts of bainite and autotempered martensite. The martensite consisted of packets of heavily dislocated parallel laths separated by low-angle boundaries; the laths appeared to be finer in the GMAW weldment. Retained austenite was often observed between the laths. The bainitic structure, Figs. 7(b) and 7(d), contained a single variant of the cementite particles, as well as a single variant of the laths.

Figures 8(a) and 8(b) show similar microconstituents observed in thin foils of the GMAW and SAW weldments located approximately 1.2 and 1.6 mm from the fusion line (regions C, C' in Fig. 4), respectively.

Fig. 7. TEM micrographs of the coarse-grained HAZ showing: a) and c) dislocated lath martensite; b) and d) bainite. (a) and (b) GMAW; (c) and (d) SAW.

Fig. 8. TEM micrographs of the fine-grained HAZ at (a) 1.2 mm from the fusion line (GMAW) and (b) 1.6 mm from the fusion line (SAW) showing autotempered martensite and coarse cementite particles between some of the laths.

The predominant autotempered martensite structure is readily visible with coarse cementite particles precipitated between some of the laths. Figures 9(a) through 9(c) show the structure of the two weldments located approximately 3 mm from the fusion line (regions D, D' in Fig. 4). The GMAW weld exhibits a packet of fine parallel laths, Fig. 9(a), and autotempered martensite containing intralath twins (T), Fig. 9(b). The SAW weld, on the other hand, exhibits an ill-defined coarse lath structure with low dislocation density, Fig. 9(c).

Figures 10 and 11 are representative TEM micrographs taken from the center of the last weld pass (weld beads #21 and #40) of the GMAW and SAW weld metal, respectively. Similar weld microstructures were observed in the weld beads #19 (GMAW) and #36 (SAW), regions A and A' in Fig. 4. The GMAW weld metal predominantly revealed a dislocated lath martensite structure, significant amounts of autotempered martensite, and bainite, Fig. 10(a). Twins were also observed within some of the laths, Fig. 10(b). The laths were usually decorated with retained austenite, Figs. 10(c) and 10(d), although at some locations other carbon rich constituents were also observed. In contrast with the GMAW weld metal, the SAW weld metal microstructure was coarser, predominantly bainitic, and contained a higher density

Fig. 9. TEM micrographs of the fine-grained HAZ at 3 mm from the fusion line showing: a) a packet of dislocated martensite laths; b) autotempered martensite with microtwins (T); and c) coarse ill-defined bainite laths. (a) and (b) GMAW; (c) SAW.

of non-metallic inclusions, Figs. 11(a) and 11(b). The inclusions size ranged from 2 to 6 μm and they were not uniformly distributed throughout the matrix.

4. Discussion

Microhardness traverse measurements served not only as an indication of the mechanical property gradient through the weld metal and HAZ, but also as a guide to choose locations for microstructural study and comparison. The higher hardness values observed for the GMAW weld metal over the SAW weld metal can be attributed to the finer microstructure of the former process. Furthermore, in the as-deposited condition, the GMAW weld metal had a slightly higher carbon content (Table 1), which may also contribute to the higher

Fig. 10.

TEM micrographs of the GMAW weld metal showing: a) fine dislocated lath martensite and autotempered martensite; b) twinned martensite; c) retained austenite between martensite laths; d) dark-field micrograph of austenite reflection; e) selected area diffraction pattern showing austenite and ferrite spots—arrow points to the $(002)_\alpha$ spot used to illuminate d).

hardness. The hardness plot through the weld metal was not smooth and significant drops were observed at locations near the weld HAZ due to heavy tempering by subsequent weld passes.

The hardness traverses across the HAZ of a single weld pass, Fig. 4, was similar for the two processes. This is in agreement with the similar structures observed. The microstructure of the HAZ in both processes consisted of a bainite-martensite duplex structure with larger amounts of autotempered martensite. The presence of heavily dislocated lath martensite in the HAZ near the base metal for the GMAW compared to the SAW may explain the higher hardness in location D'. The HAZ adjacent to the base metal has low hardness (*i.e.*, 300 VHN) due to overtempering caused by thermal cycling this region to just below the AC_1 (732°C) which is considerably above the tempering temperature specified for this steel.

The hardness across the region of repeated thermal

cycling did not show significant differences between the two processes, Fig. 5. Although a more thorough study comparing the effect of multiple thermal cycling on the microstructure of the two weldments may further explain this observation, it is reasonable to expect that the repeated thermal cycling would reduce the effect of the initial structure on the hardness, through increasing the amount of tempering.

Transmission electron micrographs suggest that the GMAW weld metal may contain a larger amount of retained austenite between the laths and a larger amount of twinned martensite. During cooling of austenite as ferrite forms, the carbon content of the remaining austenite is increased, promoting the formation of twinned martensite and/or the retention of some high carbon austenite. Twinned martensite is unlikely to form unless the carbon content of the austenite exceeds 0.2%. The observation of twins in the structure may indicate that some inhomogeneities in carbon content existed

Fig. 11. TEM micrographs from the SAW weld metal showing: a) coarse bainite ferrite laths and b) coarse non-metallic inclusions.

within the structure, prior to austenite transformation. These inhomogeneities are not uncommon in weld metals due to the short and insufficient time the weld is at or above the austenitizing temperature. Twinned martensite is the most susceptible microstructure to hydrogen-assisted-cracking (HAC); however, this phase represents only a small fraction compared to other microconstituents present in the weldments.

The overall effect of retained austenite on the properties depends mainly on the morphology. Stable retained austenite is expected to enhance the fracture toughness but there is some controversy about its role in preventing HAC. Ritchie *et al.*,⁷⁾ suggest that retained austenite is beneficial to the hydrogen resistance of lath martensitic steels. Based on the higher solubility and lower diffusivity of hydrogen in the fcc austenite, they suggested that the austenite might act as a sink for hydrogen to retard its migration to the crack front. On the other hand, the results of Kim *et al.*,⁸⁾ suggest that the presence of retained austenite increases susceptibility to HAC. They noted that the interlath retained austenite were unstable and underwent a strain-induced transformation to martensite ahead of a crack tip. The strain associated with the transformation imposed a tensile stress across the lath boundary that promoted the fracture. The authors also stated that the transformed austenite may also act as a source of hydrogen that contributes to the embrittlement of the lath boundaries. Studies by Thompson and co-workers^{9,10)} indicated that

there are strict conditions under which the austenite can be beneficial. For example, the austenite has to continuously surround the plates of ferrite if it is to hinder the diffusion of hydrogen significantly.

The higher oxygen content of the SAW weld metal is assumed responsible for the higher amount of non-metallic inclusions observed in the SAW weld metal as compared to the GMAW weld metal. In the SAW process, the main source of oxygen in the weld metal is the flux which typically includes oxides such as silica, manganese oxide, or iron oxides.¹¹⁾ Although large amounts of oxygen can be removed from the weld metal due to chemical reactions with deoxidizing elements, removal of the oxidation products is rarely complete and entrapment of the solid oxide particles occurs. The size, shape, and distribution of these inclusions are known to affect the austenite to ferrite phase transformation of low-carbon HSLA steels by promoting intragranular nucleation of acicular ferrite.¹²⁾ These non-metallic inclusions may also affect the weld microstructures of higher strength steels by pinning the austenite grain boundaries resulting in smaller austenite grains and finer bainite/martensite packets.¹³⁾ In the present work, the density of non-metallic inclusions was not sufficient to have any significant influence on the weld microstructure of the weldments.

5. Summary

The effect of two welding processes, GMAW and SAW, on the microstructure and the microhardness variations of 25 mm thick HY-130 weldments was compared. The comparison was made in four main regions, *i.e.*, base metal, coarse-grained HAZ, fine-grained HAZ, and the weld metal. The structure and properties of the base metal away from the fusion line were very similar for the two processes. Throughout the HAZ, the overall microhardness value was higher for the GMAW weldment. In the coarse HAZ, the higher hardness of the GMAW weldment was associated with a finer lath martensitic structure observed. In the fine HAZ, the structure of the GMAW contained packets of finer lath martensite, was more heavily dislocated, and contained transformation twins. The features all concur with the overall higher hardness observed. In the weld metal, the GMAW's higher hardness value was attributed to an apparently higher volume fraction of highly dislocated martensite, the presence of carbon-rich microconstituents and the observation of twinned martensite. The coarser structure of SAW weld metal was predominantly bainitic and contained a higher density of non-metallic inclusions. The inclusion density in either of the weldments seemed to be insufficient to influence the transformation kinetics of austenite.

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