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Hybrid Fiber Laser – Arc Welding of High Strength Low Alloy Steel

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Abstract. High strength low alloy (HSLA) grade 80 (HSLA-80) steel with a thickness of 9.1-mm was successfully welded in a single pass using a 5-kW fiber laser combined with an arc welding process. It was found that the maximum gap size in the Y-groove butt joint configuration should remain below 0.2 mm for a 9.1-mm section thickness. Sound butt joints with no welding cracks but some minor microporosity were successfully obtained. However, an underfill defect was frequently observed on the top surface of the weldments. Underfilling could be mitigated through increasing the filler wire addition so as to exceed the groove volume; however, due to significant evaporation and expulsion of the molten pool at the high laser power density, the underfill defect would be difficult to fully eliminate. At slow cooling rates, slight softening in the heat affected zone (HAZ) was present in the mid-thickness region but its elimination is possible at high welding speeds.

Introduction

Developed for structural applications in naval shipbuilding, HSLA-80 has been used as a substitute for HY-80 since 1984. Designed to have better weldability than HY80, HSLA-80 steel has higher strength and toughness at ambient and sub-ambient temperatures, good corrosion resistance, reduced process control and non-destructive testing prerequisites, high productivity, low costs and good weldability with reduced or no preheating requirements [1,2]. In general, HSLA steels are low in C content, Cu-bearing and alloyed with Ni, Mn, Mo, Cr along with microalloying elements Nb, Ti and V, many of which provide hardenability for transformation of the austenite (γ) into fine structures in heavy gauge sections. In HSLA-80 steel, a precipitation-hardened ferritic steel containing ~1% Cu, the high strength is attributed to (1) precipitation of fine coherent ϵ -Cu particles during aging and (2) γ hardening from the microalloying additions that control the γ grain size and on cooling precipitate as carbides and/or carbonitrides (CN) that strengthen the transformed structure of ferrite (α), bainite (α_b) and/or martensite (α').

HSLA-80 steel can be joined by conventional arc welding processes using minimum preheating and interpass temperature, e.g. 60°F for critical and 32°F for general structures [2]. It has been reported that the HAZ softening and microstructural gradients across the weldment that occur can degrade the mechanical properties such as tensile and fatigue strength [3]. Hybrid laser-arc welding (HLAW), a relatively low heat input (HI) joining technology that combines the synergistic qualities of both the high energy density laser beam for deep penetration and the arc for wide fit-up gap tolerance, is especially suited for the shipbuilding industry where long and thick steel plates have been widely used in butt-type joint configuration [4-6]. To date, however, preliminary work related to HLAW of HSLA-80 steel was only reported by Blomquist et al. [7]. In the present work, a HLAW system was applied to fully penetrate 9.1-mm thick HSLA-80 steel butt joints using a single-pass process through optimizing the groove, joint gap, and wire filler rate. The resulting welded joints were characterized in terms of defects, microstructures and microhardness.

Experimental

HSLA-80 steel plates (9.1 mm in thickness), hardened at 904°C for 19 minutes followed by water quenching and precipitation aged at 638°C for 40 minutes followed by air cooling, were sectioned to have dimensions of 100 mm (width) by 200 or 400 mm (length) and welded along the length (parallel to the rolling direction). The HSLA-80 steel plates had a chemical composition as given in Table 1 and tensile properties in the transverse direction of 593 MPa for the yield strength (YS), 652 MP for the ultimate tensile strength (UTS) and 28% for the elongation. The laser equipment consisted of an IPG Photonics 5 kW continuous wave solid-state Yb-fibre laser (YLR-5000) equipped with an ABB robot. A collimation lens of 150 mm, a focal lens of 250 mm and a fiber diameter of 200 μm were employed to produce a nominal focusing spot diameter of 0.33 mm. The laser beam was positioned on the top surface of the work-piece and calibrated at 5 kW. All experiments were conducted at a defocusing distance of -2 mm and the maximum laser power (5 kW) in “laser leading” mode that enabled a maximum welding speed of 1.5 m/min for full penetration. It is noteworthy that the laser head was inclined 5° from the vertical position to avoid any damage to the equipment from laser beam reflection. Due to the lack of matching or compatible filler material for HSLA-80 steel [2,7], AWS ER70S-6 wire with a diameter of 1.14 mm (0.045") was used (Table 1). A DC pulsed Fronius MAG arc was placed on the work-piece top surface with a distance of 2 mm from the laser beam. The angle between the electrode axis and the work-piece surface was 55°. To protect the molten weld pool during welding, the top surface of the work-piece was shielded using a mixture of 96% Ar and 4% O₂ that was fed through a MAG nozzle at a flow rate of 23.6 L/min (50 cfh), while the bottom surface was shielded using a mixture of 50% He and 50% Ar at a flow rate of 9.44 L/min (20 cfh). The HLAW process was developed to achieve full penetration in the 9.1-mm thick HSLA-80 steel plates using a butt joint configuration with a Y-groove. The bevel angle used was approximately 30°. The root size was 6 mm. The joint gap at the root was varied from 0 to 0.3 mm. The nominal wire feed rate (WFR) was calculated based on the groove volume to be filled. To explore the possibility of improving the surface quality, higher WFRs were used to compensate for the metal loss caused by spatter and evaporation. Table 2 lists the main processing parameters used. The HI was calculated using an arc efficiency of 0.84 [4].

Table 1 Chemical compositions of HSLA-80 steel plates and AWS ER70S-6 electrode (wt. %)

	C	Mn	Si	Ni	Cr	Cu	Mo	Ti	N	P	S
HSLA-80	0.04	0.55	0.25	0.85	0.69	1.14	0.20	0.002	0.0086	0.009	0.002
ER70S-6	0.09	1.4	0.95	0.034	-	0.022	-	0.006	-	-	0.013

Table 2 Process parameters used in this study

Weld #	Gap (mm)	Nominal WFR (m/min)	Actual WFR (m/min)	Arc current (A)	Arc voltage (V)	HI (kJ/mm)	t _{8/5} (s)
1	0.0	8.1	9.9	303	26.8	0.47	2.0
2	0.2	10.8	12.6	369	27.7	0.54	2.8
3	0.3	12.2	14.0	393	27.2	0.56	2.9

After welding, preparation for metallographic examination involved extracting the specimens from three sections (transverse, longitudinal and horizontal) in the steady state region of the welds. Each metallographic specimen was then ground, polished and etched in a solution of 0.5% Nital. An inverted optical microscope (Olympus GX71) equipped with AnalySIS Five digital imaging software was used for examination of the microstructures. The microindentation hardness was measured using a 500 g test load and a dwell period of 15 seconds at an indent interval of 0.3 mm with a Vickers micro-hardness machine (Struers Duramin A300).

Results and Discussion

Fig. 1 shows typical transverse weld sections obtained at three joint gaps. All the joints were fully penetrated at a welding speed of 1.5 m/min. Both macro- and micro-cracks were absent. Some minor micropores were occasionally observed. However, underfill defects were apparent on the top surface of each weld. With increasing joint gap, the severity of the underfill defect increased and the possibility of sag defects at the root was noted at 0.3 mm. In this study, the WFRs used were higher than the nominal values calculated based on the groove volume to be filled; i.e. the WFRs were increased by 22.2% at zero gap, 16.7% at 0.2 mm and 14.8% at 0.3 mm. It seems that a higher percent increase in the WFR can decrease the size of the surface underfill defects, which form mainly due to the evaporation and expulsion of molten metal. Specifically, significant vaporization of the metal at a high laser power density and the escaping vapor in the keyhole can exert a large recoil force on the surface of the molten pool. When the vapor recoil force exceeds the surface tension force of the liquid metal, expulsion or ejection of the liquid metal can occur. Both evaporation and expulsion of the molten metal lead to material losses and the formation of the underfill defect, which may be difficult to eliminate despite the use of a filler wire volume greater than the nominal value needed to fill the groove. Moreover, at a wide joint gap (e.g. 0.3 mm), significant variations in the underfill size (i.e. a large underfill depth or a small underfill depth with a sag defect) suggest process instability. The formation of sag defects is due to the wide joint gap and low surface tension at a high temperature as generally encountered in thick plate welding. Hence for 9.1-mm thick HSLA-80 steel, the maximum joint gap should be limited to 0.2 mm for the Y-groove butt configuration used and further work is needed to improve the process stability at wider root gap sizes, which are usually present during assembly of long and thick plates.

Fig. 1 Photomacrographs of typical transverse weld sections at various joint gaps.

Three zones, i.e. the fusion zone (FZ), HAZ and base metal (BM) can be clearly identified in Fig. 1; the typical microstructures in the BM and regions 1, 2, 3 and 4, demarcated in Fig. 1 for the 0.2 mm root gap, are shown in Fig. 2. The BM microstructure of the HSLA-80 steel is ferritic consisting predominately of polygonal (idiomorphic or intragranular) ferrite (α_p) that had roughly equiaxed grains with smooth boundaries and quasi-polygonal ferrite (α_q) with undulating grain boundaries, some of which possibly originating from grain boundary (allotrimorphic) ferrite. Also, fine ϵ -Cu and Nb(CN) precipitates are likely to be present [8]. It is noteworthy that Howell [9] also referred to the irregularly formed ferritic structure as granular bainite or ferrite depending on heat treatment (e.g. as-quenched or aged) and the nature of the second phase particles.

The HAZ is referred to the region heated to temperatures between the lower critical A_1 (the eutectoid temperature) and the peritectic temperature. Generally the microstructure in the HAZ consists of a mixture of acicular ferrite (α_a), Widmanstätten ferrite (α_w), bainite (α_b), α_p and α_q ; the proportion and morphology of these constituents vary with the peak temperature and cooling rate experienced. Typical microstructures are present in Fig. 2 for the sub-regions of the HAZ, i.e. intercritical HAZ (ICHAZ), fine-grained HAZ (FGHAZ), and coarse-grained HAZ (CGHAZ). In CGHAZ, γ grain growth usually occurs at peak temperatures $> 1100^\circ\text{C}$ for HSLA-80 steel due to

the dissolution of Nb(CN) precipitates, which otherwise pin the γ grain boundaries and delay the γ -to- α transformation [8]. The CGHAZ is located near the FZ and the suppressed phase transformation results in the formation of some α_w , but mostly α_b with either an aligned lath or granular morphology that is characterized by the presence of dispersed cementite, retained γ and/or α' [9]. In the FGHAZ, fine α grains are produced within a region that is exposed to a low peak temperature ($< 1100^\circ\text{C}$). Grain refinement is due to the large number of nucleation sites provided by the α grains at the time of heating. The ICHAZ is a result of peak temperature heating between the A_{c1} (low-temperature boundary) and A_{c3} (high-temperature boundary). Below A_{c1} , the subcritical HAZ (SCHAZ) does not involve any microstructural change due to the absence of α' in the BM.

In the FZ, α_a , α_w and α_b are mainly observed. The formation of α_a can improve the toughness of the weld. To investigate the heterogeneity of the microstructures in the FZ, the weld was sectioned along the longitudinal (welding) direction, as shown in Fig. 3. Due to the upward growth direction, columnar grains are clearly observed within the upper width of the Y-groove. In the lower section of the Y-groove, equiaxed prior- γ grains were observed.

Fig. 2 Microstructures at locations indicated in the 0.2-mm gap size weld image shown in Fig. 1.

Fig. 4 shows typical hardness profiles measured at the near-surface regions (top and root) and mid-thickness along a transverse weld section for the 0-mm joint gap condition. As compared to the BM (~ 220 Hv), the FZ was hardened slightly (~ 50 Hv increase) after HLA. HSLA-80 steel has a very low carbon equivalent content. Depending on the chemistry and thermal (e.g. cooling rate, ageing) conditions, hardness values between 225 Hv and 275 Hv have been reported [9], which supports the findings in the present work. Within the HAZ, the hardness at near-surface regions decreases rapidly to that of the BM hardness (Fig. 4), but, at the mid-thickness, a softened zone was apparent (Figs. 4-5). In general, the evolution in the HAZ hardness is attributed to the changes in strengthening precipitates (size and distribution), matrix microstructure (phase constituents and grain size) and dislocation structure. Moreover, the formation of a local softened region in the weldment, which renders a local brittle zone (LBR) with a low fracture toughness, is probably due to the dissolution and overaging of ϵ -Cu precipitates [2, 10], tempering of dislocation structures [10] and grain coarsening [10]. In particular, HAZ softening has been associated with high HI; with increasing HI both the extent of softening and the width of the LBZ are greater due to the longer

cooling times [10]. Presently, the HI restriction imposed for arc welding of U.S. Navy ship structures is 1.77 and 2.17 kJ/mm for HSLA-80 steel with a thickness less than $\frac{1}{2}$ " and $\frac{1}{2}$ " or greater, respectively [10]. In the present work, the much lower HI (0.47-0.56 kJ/mm) limits markedly the time for the softening to occur. However, the slower cooling rate at the mid-thickness caused slight softening ($\sim 10\%$ decrease in hardness) locally. Thus, for a given plate thickness, a combination of higher laser power and welding speed to give a lower HI and higher cooling rate may potentially be effective in eliminating the HAZ softening.

Fig. 3 Microstructures of the FZ in a longitudinal section (0.3-mm joint gap).

Fig. 4 Hardness profiles along a transverse section at zero joint gap.

It has been reported that conventional arc welding processes can be satisfactorily applied if the weld metal cooling rate is above 10°F/s or 5.6°C/s [2]. However, the minimum preheat and interpass temperature for HSLA-80 steel is 60°F (15.6°C) for critical and 32°F (0°C) for general structures [2]. In the present study HLAW was conducted at room temperature; the estimated $t_{8/5}$ is 2-3 s (Table 2) and a cooling rate of $100\text{-}150^{\circ}\text{C/s}$ over $800\text{-}500^{\circ}\text{C}$ was likely achieved. Though preheating prior to HLAW of HSLA-80 steel at room temperature is not needed, it is not clear whether preheating can be avoided with this process during welding at low temperatures. In addition, future work is needed to control the underfill defect and evaluate the mechanical properties including tensile, impact and fatigue.

Fig. 5 Transverse hardness profile on a horizontal section at mid-thickness at zero joint gap.

Summary

1. Hybrid laser-arc welding process conditions led to sound welds with no cracking but minor microporosity. Regardless of the joint gap size, ranging from the 0 mm to 0.3 mm investigated in this work, underfill defects were frequently observed on the crown surface.
2. Underfill defects were mitigated by utilizing a filler wire volume that exceeded the groove volume to be filled, but their complete elimination would be difficult at the high laser power density applied.
3. A wide joint gap is not recommended for thick gauge section welding. For 9.1-mm thick HSLA-80 steel plates, the joint gap at the root should remain below 0.2 mm for Y-type grooves used.
4. Hardness measurements indicated that slight softening ($\sim 10\%$ decrease in hardness) in the HAZ occurred at the mid-thickness of the weld that was attributed to the relatively low cooling rate. The fusion zone hardness was ~ 50 Hv greater than the parent metal value of 220 Hv.

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