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Enhanced mechanical properties in friction stir welded low alloy steel joints via structure refining

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ABSTRACT

Submerged arc weld metal of a high strength low alloy steel was subjected to friction stir welding (FSW) at a higher rotation rate of 400 rpm (FSW-a) and a lower rotation rate of 125 rpm (FSW-b), respectively. The microstructures and mechanical properties of three typical phase structures, coarse bainite phase in the weld metal, refined bainite phase and ferrite phase in the nugget zones (NZs) of FSW joints were investigated. Compared to the weld metal, enhanced mechanical properties were achieved in the NZs of both FSW joints. Large cracks apparently propagated along the bainite lath in the coarse grains of the weld metal, which would cause the brittle quasi-cleavage fracture. However, large crack propagation was inhibited in the refined bainite phase structure in the NZ of FSW-a joint, and enhanced strength and toughness with dimple fracture were achieved. Meanwhile, enhanced mechanical properties, including strength and ductility, as well as toughness, were obtained in the NZ of FSW-b joint, because of the refinement of the ductile ferrite structure.

1. Introduction

In recent years, high strength low alloy (HSLA) steels have attracted considerable interest in various industrial areas, such as oil and gas transportation, ship building, and automotive industries [1–3]. In the fabrication of HSLA steel structures, the integrity and reliability of structures are extremely dependent on the mechanical properties of the welded joints. However, the growth of prior austenite grains during fusion welding process will lead to poor mechanical properties, especially when the formation of coarse bainite and martensite phases occurs. Therefore, the control of the microstructure in the welded joints for obtaining sound strength-toughness synergy should be the key issue for the actual application of HSLA steels [4,5].

It is well accepted that grain refining is an essential strengthening method that enhances both the strength and the toughness of the metal materials simultaneously. Therefore, various grain refining methods have been used for refining the weld structure of HSLA steel joints. Inclusion assisted microstructure control is the most encouraging method to refine the weld metal, with the formation of fine intragranular acicular ferrite (AF) [5–8]. Though a fine interlocking AF structure with the average grain size of less than 5 μm can be produced, some coarse grains larger than 10 μm usually form in the weld metal, especially at the prior austenite grain boundaries [5,9,10]. Besides, in order to increase the strength without significant loss in toughness, bainite phase structure is desirable in the weld metal. However, very large bainite laths usually form in the coarse prior austenite grains, which may result in the easy propagation of large cracks along the lath boundaries [11,12]. In this case, other methods are still needed for obtaining a uniform refined microstructure in the HSLA steel joints.

Friction stir welding (FSW) is a solid state technology which could eliminate the melting and solidification associated problems, such as liquation and solidification cracking [13,14]. Since invented in 1991, FSW has been successfully applied to various types of materials, such as Al, Mg, Cu alloys [13–17]. Recently, FSW of high melting point alloys including Ti, Ni-based alloys, and also steels, has been increasingly drawing attention [18–24]. Though FSW is characterized as a solid-state process, the peak temperature is usually higher than A_{C1}, i.e. the α–γ phase transformation temperature, in FSW of steel [19]. Therefore, transformed phase structures, martensite and bainite phases usually appear in the nugget zone (NZ). However, due to the intense plastic deformation, the prior austenite grain size should be greatly refined compared to that in the conventional fusion welding [19–24]. Then, the mechanical properties may be enhanced in the NZ of FSW joint, especially when the bainite forms.

On the other hand, the heat input could be reduced greatly if a very low rotation rate is adopted during FSW, and the peak temperature could be controlled to below A_{C1} [25,26], bring about the fine ductile ferrite structure. Therefore, if we control the rotation rates, different peak temperatures of above and below

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A_{C1} can be achieved during FSW. Then, various refined microstructures with/without phase transformations will be obtained in the NZs, and this will inevitably bring different mechanical properties. It is interesting to compare the mechanical performances of these special microstructures with the weld metal in fusion joints; however, related investigations are still lacking.

In this study, a submerged arc weld metal was subjected to FSW at a higher rotation rate of 400 rpm and a lower rotation rate of 125 rpm, resulting in different peak temperatures of above A_{C3} and below A_{C1} in the NZs, respectively. In this case, three different typical phase structures, coarse bainite phase in the weld metal of fusion joint, fine bainite phase and ferrite phase in the NZs of FSW joints were achieved. The investigations of the microstructure evolution and mechanical properties in the NZs of the FSW joints, as well as the relationship between the microstructure and mechanical behaviors of the three typical phase structures were carried out in this study.

2. Experimental procedures

The materials used in this study were HSLA steel weld plates fabricated by submerged arc welding (SAW), and the chemical composition of the weld metal is shown in Table 1. In order to investigate the intrinsic differences of the mechanical properties between the fusion weld metal and the NZs of FSW joints, FSW was performed on the SAW weld metal, and thus the chemical compositions should be the same. After SAW, the weld plates were machined to 300 mm length, 100 mm width and 2 mm thickness blocks. FSW experiments were performed using a load-controlled FSW machine, and an argon (Ar) shielding gas was used to prevent the oxidation of the weld surface during FSW. Two FSW parameters of 400 rpm–200 mm/min and 125 rpm–100 mm/min were used in this study, designated as FSW-a and FSW-b, respectively. The welding tool was made of a tungsten carbide (WC) based material and equipped with a columnar pin without threads. The tool size and FSW parameters are shown in Table 2.

Microstructural observations were conducted by optical microscopy (OM), electron backscatter diffraction (EBSD), and scanning electron microscopy (SEM). The specimens for OM and SEM observation were machined perpendicular to the welding direction, and etched with a 4% nital solution. EBSD specimens were prepared by electro-polishing at room temperature using a solution of 92% acetic acid and 8% perchloric acid under a potential of 30 V.

Vickers hardness tests of the FSW joints were measured on the cross-section perpendicular to the welding direction at the center of the NZ. For each measurement, 1000 g load was applied for 15 s at intervals of 0.5 mm. The dog-bone-shaped tensile specimens with a gauge of 3 mm length, 1.4 mm width and 0.7 mm thickness were machined perpendicular to the welding direction from the weld metal and NZs, as schematically shown in Fig. 1. Uniaxial tensile tests were conducted at an initial strain rate of $1\times10^{-3}\,s^{-1}$. The fracture surfaces were examined using SEM.

Mini-specimens for Charpy impact tests were cut from the weld metal and NZs in a direction perpendicular to the welding direction to dimensions of 20 mm length, 0.5 mm thickness, and 0.5 mm width, and all samples were given a 0.1 mm notch (Fig. 1). A miniaturized Charpy impact machine, developed earlier [27], was employed to determine energy absorption under dynamic fracture. The samples were impacted with a punch driven by a pressurized gas cylinder, with impact load measured by a load cell placed at the top of the punch. Impact testing was carried out at the room temperature, with the impact deformation rate (i.e. the speed of the punch) of 1 ms^{-1}.

3. Results

Fig. 2 shows the macrostructures of the SAW weld metal and FSW joints. The weld metal exhibited a typical fusion welded structure, with many coarse columnar grains, as shown in Fig. 2a. Defect-free NZs were successfully achieved by both FSW parameters, and the microstructures were refined apparently (Fig. 2b and c). The macrostructure of the FSW joint revealed a “bowl” shape, indicating an intense plastic deformation occurrence during FSW owing to both shoulder and pin effect.

Fig. 3 shows the detailed microstructure of the weld metal. The weld metal was characterized as a typical bainite structure, with many large bainite laths in the coarse grains. Moreover, coarse grain boundary ferrite (GBF) phase formed at the prior austenite grain boundaries during the phase transformation of the welding process (Fig. 3a). From the SEM microstructure shown in Fig. 3b, many carbide strips between the bainite laths could be observed.

After FSW, the microstructure was greatly refined, and equiaxed grains were achieved in the NZs, as shown in Fig. 4. Bainite phase structure was obtained at a high rotation rate of 400 rpm (FSW-a). Compared to the weld metal, the bainite grain size was significantly decreased, and the largest grain size was about 10 μm. The distributions of various paralleled bainite laths in the fine equiaxed grains, and carbides between the bainite laths were confirmed (Fig. 4a and b). However, uniform equiaxed ferrite grains formed at a lower rotation rate of 125 rpm (FSW-b), indicating a peak temperature of below A_{C1} during FSW process. It is obvious that the grains in the NZ were further refined compared to that in FSW-a joint, with an average grain size of ~3 μm. Moreover, many ultrafined carbide precipitations distributed in the fine ferrite grains or the grain boundaries (Fig. 4d).

Table 1
Chemical composition of the weld metal (in wt%).

Element	C	Si	Mn	P	S	Cu	Ni	Cr	Ti	V	Nb	B	Al	O	N
Content	**0.057**	**0.30**	**1.51**	**0.010**	**0.005**	**0.18**	**0.11**	**0.12**	**0.017**	**0.03**	**0.023**	**0.0030**	**0.034**	**0.027**	**0.0045**

Table 2
Summaries of the welding conditions.

No.	Welding parameters			Tool size		
	Rotation rate (rpm)	Welding speed (mm/min)	Tool tilt	Shoulder diameter (mm)	Pin diameter (mm)	Pin length (mm)
a	400	200	3	12	4	1.8
b	125	100	3			