Friction stir welding of carbon steels

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Abstract

In order to determine the effect of the carbon content and the transformation on the mechanical properties and microstructures of the FSW carbon steel joints, three types of carbon steels with different carbon contents (IF steel, S12C, S35C) were friction stir welded under various welding conditions. Compared with IF steel, the microstructures and mechanical properties of the carbon steel joints are significantly affected by the welding conditions. The strength of the S12C steel joints increases with the increasing welding speed (decreasing the heat input), while the strength of the S35C steel joints shows a peak near 200 mm/min. This can be explained by the relationship between the peak temperature and the A1 and A3 points. When friction stir welding is performed in the ferrite–austenite two-phase region, the microstructure is refined and the highest strength is then achieved.

Keywords: Friction stir welding; Steel; Carbon; Phase transformation

1. Introduction

Friction stir welding (FSW) is a welding technique, patented in 1991 by TWI [1], that has been widely investigated for mostly low melting materials, such as Al, Mg and Cu alloys [2–4]. The application of FSW to steels and other high temperature materials has been limited due to the absence of a suitable tool material, which is required to remain intact at temperatures higher than 1000 °C. Several previous studies [5–8] have reported the friction stir welding of carbon steels. These studies [5–7] have reported that FSW achieves grain refinement in the stir zone of the carbon steel, similar to Al alloys. Additionally, complex phase transformations also occurred in the FSW process. Accordingly, the mechanical properties were improved compared to the base metals. However, in these papers, important steel features, such as the effect of the carbon content and the transformation during the friction stir welding of carbon steels was not systematically classified. For example, Thomas et al. [5] only used one type of carbon steel (EN10083, 0.10%C–0.16%Si–0.69%Mn–0.008%P–0.023%S) and one kind of welding condition, and therefore, no comparison of the weld properties or microstructure was performed. Lienert et al. [6] suggested that the peak temperature of the stir zone exceeded 1100 °C and surpassed 1200 °C for the 1018 steel (0.18%C–0.01%Si–0.82%Mn–0.011%P–0.006%S) at a 25 mm/min welding speed. Also in their paper, no attempt was made to associate the change in the weld properties with the variations in the FSW parameters. In the paper by Reynolds et al. [7], the DH36 steel, a kind of C–Mn steel (0.18%C–0.10–0.50%Si–0.9–1.6%Mn–0.035%P–0.035%S) was investigated under four kinds of welding parameters. However, the peak temperatures under all the welding conditions exceeded the A3 temperature, and therefore, the effect of the phase transformation on the microstructure was not completely investigated. Fujii et al. [8] reported the friction stir welding of interstitial free steel (IF steel: 20 ppmC) and ultrafine grained steel. The microstructure and mechanical properties do not significantly change due to the ultra-low carbon content, although a series of welding condition was used. Accordingly, the objective of this study is to determine the effect of the carbon content and the transformation on the mechanical properties and microstructures of the FSW carbon steel joints. Three types of carbon steels
with various carbon contents were friction stir welded under different welding conditions including low temperature conditions.

2. Experimental procedure

Butt-welding of an ultra-low-carbon IF steel, and two kinds of plain carbon steels with different carbon contents—JIS S12C (equivalent to UNS G10120, SAE-AIEI 1012, 0.12 wt.%C) and JIS S35C (equivalent to UNS G10350, SAE-AIEI 1035, 0.34 wt.%C) was performed. The plates are 1.6 mm thick, 30 mm wide and 300 mm long. The chemical composition and mechanical properties of these materials are listed in Table 1. The compositions of the three kinds of carbon steels are located in a section of the Fe–Fe₃C phase diagram as shown in Fig. 1. This figure suggests that the IF steel features a ferrite single-phase over the entire range of temperature below 910°C, though an austenite–ferrite transformation occurs at the A₃ temperature (about 910°C). For both the S12C and S35C, the austenite–ferrite transformation occurs between A₃ (about 860 and 800°C, respectively) and A₁ (about 723°C), and the eutectoid reaction at A₁. Therefore, S12C would show a mixture of proeutectoid ferrite and pearlite when it is slowly cooled from austenite. The volume fraction of the pearlite is expected to be 13%. Slowly cooled S35C would also show a ferrite + pearlite structure in which the ferrite and pearlite have nearly equal volume fractions (57% and 43%, respectively). The microstructures of the base metals are shown in Fig. 2. The IF steel consists of only ferrite, and the S12C steel has the typical ferrite + pearlite structure. The S35C steel is characterized by the ferrite + globular cementite two-phase structure because spheroidizing annealing was performed in order to improve the machinability.

The welding experiments were performed using a load-controlled FSW machine. The welding tool, made of a WC-based material and equipped with a columnar probe without threads [9], was tilted 3° from the plate normal direction. The rotation speed and the welding speed are shown in Table 2. A K-type thermocouple was placed on the material’s bottom surface at the centerline to measure the change in temperature during the FSW.

Optical microscopy (OM) and SEM observations of each FSW joint were carried out. The metallurgical inspections were performed on a cross-section of the joint after polishing and etching with nitreagent. In order to clarify the crystallographic features of the microstructures, an electron back-scattering pattern (EBSP) technique was also used in a field emission type scanning electron microscope (FE-SEM) operated at 25 kV. The cross-section was cut perpendicular to the welding direction and then electrolytically polished in 20 ml HClO₄ + 180 ml CH₃COOH solution at 248 K for the EBSP analysis. The average grain size was determined from the EBSP orientation maps by the mean-linear-intercept method using the high angle grain boundary in which the grain misorientations are larger than 15°.

The tensile properties of each joint were evaluated using three tensile specimens cut perpendicular to the welding direction from the same joint. All tensile tests were performed using an initial strain rate of 3×10⁻³ s⁻¹. The Vickers hardness profile of the weld was measured on the cross-section perpendicular to the welding direction with a 0.98 N load for 15 s.

3. Results and discussion

3.1. Mechanical properties

Fig. 3 shows the tensile properties of the carbon steel FSW joints welded at different welding speeds. The heat input (energy input per unit weld length) decreases with the increasing welding speed at a constant rotation speed [10]. Because all the tensile specimens involving the entire joint shown in Fig. 3(a) fractured

<table>
<thead>
<tr>
<th>Type</th>
<th>Chemical composition in mass%</th>
<th>Mechanical property</th>
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<tr>
<td></td>
<td>C</td>
<td>Si</td>
</tr>
<tr>
<td>IF</td>
<td>0</td>
<td>0.1</td>
</tr>
<tr>
<td>S12C</td>
<td>0.12</td>
<td>0.20</td>
</tr>
<tr>
<td>S35C</td>
<td>0.34</td>
<td>0.21</td>
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</table>

Fig. 1. Schematic illustration of the different carbon steels in the Fe–Fe₃C phase diagram.
Fig. 2. SEM microstructures of carbon steel base metal: (a) IF steel, (b) S12C, and (c) S35C, the arrows in (c) means spheroidized Fe₃C.

Table 2

<table>
<thead>
<tr>
<th>Tool size (mm)</th>
<th>Welding parameters</th>
</tr>
</thead>
<tbody>
<tr>
<td>Shoulder diameter</td>
<td>Probe diameter</td>
</tr>
<tr>
<td>12</td>
<td>4</td>
</tr>
</tbody>
</table>

in the base metals, small tensile specimens of which the gauge length was covered by the weld nugget shown in Fig. 3(b) were used to characterize the tensile properties of the stir zone. In this case, the strength of the IF steel joints slightly increased with the increasing welding speed (decreasing the heat input), while the dependence of the welding speed for S12C and S35C are greater than that of the IF steel. For the S12C steel, the strength of the joints increased with the increasing welding speed (decreasing the heat input), while for the S35C steel, the strength of the joints show a peak near 200 mm/min. The reasons for these characteristic changes will be discussed later.

Fig. 3 shows typical Vickers hardness profiles of the carbon steel FSW joints welded at different welding speeds. Basically the hardness within the weld regions is higher than the base metals, though the hardness profile is different depending on both kinds of steels and welding conditions. For the IF steel, the welding speed does not significantly affect the hardness profiles of the joints, while for the S12C steel, the higher the welding speed (the smaller heat input), the greater the hardness. However, for S35C, the greatest hardness was obtained with the medium heat input (near 200 mm/min), and a significant decrease in the heat input induced the decreasing hardness.

3.2. Temperature cycle

Fig. 5 shows typical changes in temperature of the workpiece during the process. Thermocouples were embedded on the bottom surface at the centerline. The peak temperatures increase with the decreasing welding speeds for all three types of steels. Additionally, the heating and cooling rates of the three types of steels decrease with the decreasing welding speeds. It is noted that the peak temperatures at the 400 mm/min welding speed are about 650 °C for all the IF steel, S12C and S35C, which are all below A₃. It can be stated that the temperatures much lower than the peak temperatures reported in the previous papers [5–7] were controlled in this study. On the other hand, the temperature could exceed the A₁ or A₃ temperature under the conditions of a high heat input.

Fig. 4 shows typical Vickers hardness profiles of the carbon steel FSW joints welded at different welding speeds. Basically the hardness within the weld regions is higher than the base metals, though the hardness profile is different depending on both kinds of steels and welding conditions. For the IF steel, the welding speed does not significantly affect the hardness profiles of the joints, while for the S12C steel, the higher the welding speed (the smaller heat input), the greater the hardness. However, for S35C, the greatest hardness was obtained with the medium heat input (near 200 mm/min), and a significant decrease in the heat input induced the decreasing hardness.
Fig. 4. Microhardness profile of friction stir welded carbon steels for: (a) IF steel, (b) S12C and (c) S35C.

3.3. Microstructure

The basic properties of the FSW joints for the three kinds of steels processed under the various conditions were clarified in the present study. In this section, the characteristic features of each steel are systematically discussed on the basis of their microstructures.

3.3.1. IF steel—an ultra low carbon steel with a ferrite one-phase structure

Fig. 5 shows the EBSP color maps in the stir zones of the FSW joints of the IF steel. In the maps, the black lines correspond to the high angle boundaries having misorientation angles higher than 15°, while the gray lines correspond to the low angle boundaries having misorientation angles between 2° and 15°.

All maps in the stir zones exhibit nearly uniform grain structures composed of equiaxed grains with sub-structures. It is found that the grain sizes at the center of the stir zone welded under the highest and lowest heat input conditions are 6 and 5 μm, respectively, which are much smaller than that of the base metal, 24 μm. On the other hand, the bottom surface temperatures under the greatest and smallest heat input conditions reached 839 and 643 °C, respectively. It was revealed that all of the IF steel were welded in the ferrite single-phase region, and no transformation occurred. Therefore, the grain refinement is the major hardening factor of the material. The reason why the fine-grained structure was obtained is that FSW is a kind of severe plastic deformation (SPD). The fine-grained structure including substructures, like subboundaries and dislocations, is a typical feature of severely deformed materials [11]. Since the temperature of the materials significantly increases during FSW, the recovery and grain growth are enhanced during simultaneous processing which results in the equiaxed structure. Such a microstructural evolution can be considered as a continuous recrystallization. It is noteworthy that the process occurring in the FSW of the IF steel is mainly heavy deformation and recrystallization without phase transformation.

Because the grain size did not significantly decrease with the decreasing heat input, the strength only slightly increased by less than 4%, as shown in Fig. 3. Similar results are obtained for the aluminum alloys or copper alloys, etc., in which no transforma-
tion occurs during the welding, namely, the strength of the joints slightly increases with the decreasing heat input [4,12–15].

3.3.2. S12C—a carbon steel with small proportion of pearlite

Fig. 7 shows the microstructure in the stir zones of the friction stir welded joints of the S12C steel. It is found that ferrite–pearlite structures with a limited amount of pearlite were obtained under all the welding conditions. Under the higher heat input condition (400 rpm, 100 mm/min), almost the entire microstructure consists of equiaxed ferrite grains while several large pearlite clusters were found in limited areas. However, under the lower heat input conditions (400 rpm, 400 mm/min), small pearlite colonies and carbides were uniformly distributed in the ferrite matrix.

The peak welding temperature measured on the bottom surface was 790 °C under the higher heat input conditions (400 rpm, 100 mm/min), which reveals that the S12C steel welding was performed in the ferrite–austenite two-phase region at the bottom, and the transformation then occurred during cooling after welding. In this case, because only the austenite with the small proportion (about 13%) was transformed into the pearlite below Ac3, the pearlite distribution is not uniform. The ferrite grain size is fairly small (about 3 μm), because heavy deformation in the ferrite + austenite two-phase region causes fine recrystallized microstructures.

On the other hand, the bottom surface temperature was about 640 °C under the lower heat input condition (400 rpm, 400 mm/min). In this case, FSW is performed below the Ac1 temperature and therefore, the transformation does not occur during cooling after welding. Accordingly, the pearlite in the base metal was stirred and fractured into small pieces, so that the carbides were uniformly distributed in the ferrite matrix. Similarly, the breakup of constituent particles is also present in the friction stir weld of the metal matrix composites (MMC) [16] and the cast Al–Si alloy [17,18], resulting in the increased hardness of the welds. The ferrite matrix is, of course, refined by a kind of SPD (FSW). It is interesting that the total amount of carbides seems very small in this material, which suggests that a dissolution of carbides may occur during the FSW as was reported in the heavily drawn pearlite wire [19]. This might be another reason why the fine carbides are scattered in the fine-grained ferrite structure. It is known that the matrix grain size and the dispersion of the strengthening precipitates (carbides) significantly contributes to the strength of the plain carbon steels [20]. The area having uniformly distributed carbides increases with the increasing welding speed, while the ferrite grain size does not change significantly. This is because the temperature near...
the sample surface is higher, and therefore the area where the temperature exceeded the A1 point decreases with the increasing welding speed.

Fig. 8 shows the microstructures in different regions of the stir zones of the S12C friction stir welded joints. Under the lower heat input condition (400 rpm, 400 mm/min), the ferrite grain size and the pearlite morphology are similar in all the regions. However, under the higher heat input condition (400 rpm, 100 mm/min), the ferrite grain size (about 3 μm) at the bottom is much smaller than that in the top or middle (about 5 μm) of the stir zone. This can be explained by the austenite–ferrite transformation and the difference in the cooling rate of the welds. Because the measured bottom temperature reached 790 °C, the transformation should occur. However, it is considered that the temperature in the top and middle parts should also exceed A3. Namely, the top and middle regions were friction stir welded in the austenite single-phase region, and then cooled to form the microstructures shown in Fig. 8. Though the austenite should be refined by dynamic or/and static recrystallization during the FSW, the final grain size is significantly affected by the temperature. When the temperature rises much above A3, grain coarsening would occur. This is the reason why the resulting ferrite grains of the top and middle regions under this condition are relatively coarser. The bottom region was friction stir welded in the ferrite-austenite two-phase region. At the beginning of the friction stir welding, the ferrite phase is deformed, recrystallized and refined below A1. Austenite is formed at the boundaries of the refined ferrite phases above A1 and finely distributed, and austenite is also deformed and recrystallized. When the temperature during the friction stir welding remains in the ferrite–austenite two-phase region, the grain growth of the austenite phase is significantly restricted due to the lower temperature and the existence of the ferrite phase which inhibits the grain growth of the austenite. Furthermore, the refined austenite is transformed into the refined ferrite and pearlite phases during cooling. This is the reason why a finer structure was observed in the bottom region. The total amount of carbides in the joints by FSW, a kind of severe plastic deformation, is much smaller than that under equilibrium conditions, while their distribution is not significantly affected by the relationship between the peak temperature and the A3 point.

table

<table>
<thead>
<tr>
<th>position</th>
<th>400rpm, 100mm/min</th>
<th>400rpm, 400mm/min</th>
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<tbody>
<tr>
<td>top</td>
<td><img src="image1.png" alt="SEM microstructures" /></td>
<td><img src="image2.png" alt="SEM microstructures" /></td>
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<tr>
<td>middle</td>
<td><img src="image3.png" alt="SEM microstructures" /></td>
<td><img src="image4.png" alt="SEM microstructures" /></td>
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<tr>
<td>bottom</td>
<td><img src="image5.png" alt="SEM microstructures" /></td>
<td><img src="image6.png" alt="SEM microstructures" /></td>
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</table>

Fig. 8. SEM microstructures of the stir zone in S12C FSW joints.
On the other hand, because the welding was performed below the A1 point under the lower heat input condition, the transformation did not occur. Accordingly, the grain size does not significantly change throughout the sample. Similar results are obtained in many alloys, such as aluminum alloys, in which no transformation occurs during the FSW [4,12–15].

3.3.3. S35C—a medium carbon steel with large proportion of pearlite

Fig. 9 shows the microstructures in the stir zones of the friction stir welded joints of the S35C steel. It was found that the ferrite–pearlite structures were mainly obtained in the stir zones under almost all the welding conditions. The peak welding temperatures (on the bottom surface) at the welding speeds of 100, 200 and 400 mm/min are 873, 741, and 653 °C, respectively. These temperatures correspond to the austenite single-phase region, the ferrite + austenite two-phase region and the ferrite + cementite two-phase region, respectively. The microstructure under the lowest heat input condition (400 rpm, 400 mm/min) shows a fine ferrite + partially deformed pearlite in the upper part and a ferrite + globular cementite structure in the lower 1/3 part, which is similar to the base metal.

Because the bottom surface temperature under the lowest heat input condition (400 rpm, 400 mm/min) of 653 °C is below A1, the S35C steel welding was performed in the ferrite + cementite two-phase region on the bottom, and therefore, the transformation should not occur during cooling after welding. Consequently, the microstructure is almost the same as that in the base metal while the ferrite grain size slightly decreased from the base metal due to the recrystallization. This result indicates that despite the stirring by the tool, the microstructure was not significantly changed under this condition. However, the temperature on the upside of the stir zone seems to exceed A1. Consequently, the partially deformed pearlite structure was formed in the upper part of the stir zone. It is deduced that, although the peak temperature slightly exceeded A1, the temperature soon decreased to below A1, just after the tool went through the position, but the metal is still being stirred.

Because the ferrite + globular cementite structure, which is softer than the ferrite + pearlite structure, exists in the lower part, the tensile strength of the entire weld joints decreased when the welding speed exceeded 200 mm/min, as shown in Fig. 3. Under the medium heat input condition (400 rpm, 200 mm/min), the ferrite + pearlite structure is the finest and the joint strength is the highest. This is because the peak temperature was in the ferrite–austenite two-phase field, as mentioned in Section 3.3.2.

4. Summary

In this study, the effect of the carbon content and the transformation on the mechanical properties and microstructures of the FSW carbon steel joints was investigated. The low temperature friction stir welding of steels was successful at around 650 °C, which should be the first example of welding general steels without any transformation. In addition, the control of the temperature enabled the steels to be welded in various regions, such as the α + γ two-phase region and the γ single-phase region. As a result, the following conclusions were achieved.

(1) Compared with the IF steel, the microstructures and mechanical properties of the carbon steel joints are significantly affected by the welding conditions.
(2) The strengths of the S12C steel joints increased with the increasing welding speed (decreasing the heat input), while the strengths of the S35C steel joints show a peak near 200 mm/min.
(3) This can be explained by the relationship between the peak temperature and the A1 or A3 point. When the friction stir welding is performed in the ferrite–austenite two-phase region, the microstructure is refined and the highest strength is then achieved.
Thus, during the welding of carbon steels, both the temperature and composition significantly affect the microstructure evolution. Friction stir welding enables us to control these factors and then produce higher strength joints.

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References