Transformation in Stir Zone of Friction Stir Welded Carbon Steels with Different Carbon Contents

Ling CUI,1) Hidetoshi FUJII,1) Nobuhiro TSUJI,2) Kazuhiro NAKATA,1) Kiyoshi NOGI,1) Rinsei IKEDA3) and Muneo MATSUSHITA3)

1) Joining and Welding Research Institute, Osaka University, 11-1 Mihogaoka, Ibaraki, Osaka 567-0047 Japan. E-mail: fujii@jwri.osaka u.ac.jp 2) Department of Adaptive Machine Systems, Osaka University, Suita, Osaka 565-0871 Japan. 3) Steel Research Laboratory, JFE Steel Corporation, 1 Kawasaki-cho, Chuo-ku, Chiba 260-0835 Japan.

(Received on September 15, 2006; accepted on November 10, 2006)

Five types of ferrite-pearlite structure carbon steels with different carbon contents (IF steel, S12C, S20C, S25C, S50C) were friction stir welded under various welding conditions, and the mechanical properties and microstructures of the FSW carbon steel joints were evaluated. Compared with IF steel, the microstructures and mechanical properties of the carbon steel joints are significantly affected by the welding conditions. When the carbon content is less than or equal to 0.12 mass%, the welding produces ferrite-pearlite structures, and the strength slightly increases compared to the base metal due to the refined microstructure; when the carbon content is above 0.2 mass%, the welding produces ferrite-pearlite plus harder phases like the martensite and bainite microstructures, resulting in a significantly increased strength of the joints. These are dependent on each of the thermal-mechanical cycles.

KEY WORDS: friction stir welding; carbon steel; carbon content; phase transformation.

1. Introduction

Friction stir welding (FSW)1) has been widely investigated and even been commercialized mostly for low melting materials, such as Al, Mg and Cu alloys.2–4) After the initial trial for the higher temperature materials was successful by Thomas et al.,5) several previous studies have reported the friction stir welding of interstitial free steel (IF steel: 20 ppm C),6–8) carbon steels,9–15) alloy steels,16) and stainless steels.17–22) Lienert et al.9) and Reynolds et al.10) studied one kind of carbon steel—the 1018 steel (0.18%C–0.01%Si–0.82%Mn–0.011%P–0.006%S) and a C–Mn steel (0.18%C–0.10–0.50%Si–0.9–1.6%Mn–0.035%P–0.035%S) respectively. They reported that the FSW achieves grain refinement in the stir zone of the carbon steel, similar to Al alloys. Additionally, complex phase transformations also occurred during the FSW process. Therefore, the hardness or strength was improved compared to the base metals. However, the peak temperature of the stir zone reported in these studies was around 1 100 to 1 200°C thus exceeding the A3 temperature, and therefore, the effect of the phase transformation on the microstructure was not completely investigated. In the authors’ previous study,15) 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) were studied. 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. However, compared to the S12C base metal having the typical ferrite+pearlite structure, the S35C base metal was characterized by the ferrite+globular cementite two-phase structure because spheroidized annealing was performed in order to improve the machinability of the middle and high carbon steels for practical use. Accordingly, in the strict sense of the word, the effect of the carbon content was not completely clarified. Therefore, in this study the same typical ferrite+pearlite structure is standardized for all carbon steel base metals, prepared by annealing the commercial carbon steels, which have ferrite+globular cementite structures, several minutes at about 900°C. The IF steel and four types of carbon steels with various carbon contents were friction stir welded under different welding conditions, and then the mechanical properties and microstructures of the joints were evaluated. Based on these experimental results, the effect of the carbon content on the mechanical properties and microstructures of the FSW carbon steel joints is discussed.

2. Experimental Procedure

An ultra low-carbon IF steel, and four kinds of plain carbon steels with different carbon contents—JIS S12C (equivalent to UNS G10120, SAE-AIEI 1012, 0.12 wt% C),
JIS S20C (equivalent to UNS G10200, SAE-AIEI 1020, 0.21 wt% C), JIS S35C (equivalent to UNS G10350, SAE-AIEI 1035, 0.34 wt% C) and JIS S50C (equivalent to UNS G10500, SAE-AIEI 1050, 0.50 wt% C) were utilized in this study. The chemical composition and mechanical properties of these materials are listed in Table 1. The compositions of the five kinds of 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 temperatures below 910°C, although an austenite–ferrite transformation occurs at the A₃ temperature (about 910°C), and above 910°C featuring an austenite single-phase. For the S12C, S20C, S35C and S50C, the austenite–ferrite transformation occurs between A₃ (about 860, 840, 798 and 770°C respectively) and A₁ (about 723 °C), and the eutectoid reaction at A₁. Therefore, the carbon steels would show a mixture of proeutectoid ferrite and pearlite when they are slowly cooled from austenite. The volume fraction of the pearlite is expected to be about 13%, 24%, 43%, 64% respectively, increasing with the increasing carbon content, and these are quite consistent with the microstructures of the base metals for the four kinds of carbon steels shown in Fig. 2. Plates of 300 mm × 30 mm × 1.6 mm were cut and the joint butt surface wasmachined. All the plates were degreased before welding.

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, was tilted 3° from the plate normal direction. A constant rotation speed of 400 rpm (6.7 s⁻¹) and the various welding speeds in a range from 25 mm/min (0.42 mm/s) to 400 mm/min (6.7 mm/s) were utilized as 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 as shown in Fig. 3.

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 a nitreagent. 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 4×10⁻³ 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

3.1. Temperature Cycle

Figure 4 shows the typical temperature cycles of the workpiece during the process for the S50C steel. Thermocouples were embedded on the bottom surface at the centerline, therefore the measured temperature corresponds to the value at the bottom part rather than the whole stir zone. The temperature in the center of the stir zone should be higher, however, the difference is less than 100°C, judging from the microstructures, as discussed later. The measured peak temperatures decrease with the increasing welding speed for all five types of steels. Additionally, the heating and cooling rates of the five types of steels increase with the increasing welding speed. Here the welding cooling rate is defined by the average cooling rate from 800 to 500°C. However, when the peak temperature ($T_{\text{max}}$) was below 800°C, the cooling rate is defined by the average cooling rate from the peak temperature to 500°C.

The peak temperature and cooling rate of the workpiece during the process is shown in Fig. 5. Independent of the kind of steel, with the increasing welding speed, the peak temperatures decrease, while the welding cooling rates increase. It is noted that the peak temperatures at the 400 mm/min welding speed are all below 700°C for all the carbon steels, which are all below $A_1$, and therefore, the transformations does not occur. It can be stated that the temperatures, much lower than the peak temperatures reported in previous papers, were controlled in this study. On the other hand, by controlling the welding conditions (e.g., increasing the rotation speed or decreasing the welding speed), the temperature can exceed the $A_1$ or $A_3$ temperature, and therefore, the welding can be performed in the ferrite–austenite two-phase region or austenite single-phase region, and the transformation then produces various microstructures due to the different cooling rates after welding.

3.2. Mechanical Properties

No volumetric defect was observed for all the joints. Because all the tensile specimens referenced to JIS Z2201, which involve the entire joint, fractured in the base metals, small tensile specimens of which the gauge length was covered by the weld nugget shown in Fig. 6 were used to characterize the tensile properties of the stir zone. Figure 7 shows the tensile properties of the carbon steel FSW joints welded at the constant rotation speed of 400 rpm, but at different welding speeds. The heat input (energy input per unit
weld length) decreases with the increasing welding speed at a constant rotation speed. In this case, the strength of the IF steel joints slightly increases with the increasing welding speed (decreasing the heat input), while the dependence of the welding speed for carbon steels are greater than that of the IF steel. Compared with the IF steel, the strength of all the carbon steel joints significantly increases with the increasing welding speed. In greater detail, compared with the S12C steel, the strength of the joints simply increases with the increasing welding speed (decreasing the heat input), the strength of the joints for the S20C, S35C and S50C steels show a significant increase at a definite welding speed. The definite speed for the S20C, S35C steels is between 50 mm/min and 100 mm/min. For the S50C steel, it is between 25 mm/min and 50 mm/min. The reasons for these characteristic changes will be discussed later.

Figure 8 shows typical micro-Vickers hardness profiles of the carbon steel FSW joints welded at the constant rotation speed of 400 rpm, but different welding speeds. Basically, the hardness within the weld regions is higher than the base metals. At the constant rotation speed, for the IF steel, the welding speed does not significantly affect the hardness profiles of the joints, while for the carbon steels, the hardness profile is significantly changed depending on both the kinds of steel and the welding conditions. For the S12C steel, the hardness within the weld regions appears constant under all the welding speeds and increases with the increasing welding speed; while for the S20C, S35C and S50C steels, the welding hardness is significantly changed at the higher welding speeds (equal or larger than 100 mm/min) except for those under the lower welding speeds. These microhardness features were thought to correspond to the microstructures, i.e., the significant changed hardness profile implies that the complex transformation occurred. In order to evaluate the average Vickers hardness of the complex microstructures, five measurements under a load of 19.6 N for 15 s at the center of the SZ were also carried out. These average hardness results are quite consistent with the trend in the tensile results.

3.3. Microstructure

Figure 9 shows the microstructure at the center of the stir zones of the friction stir welded S12C steel joints under several typical welding conditions. It was found that compared with the base metal, the refined ferrite (dark area) with a limited amount of pearlite (light area) is obtained under all the welding conditions. The average intercept grain size of ferrite in the center of the SZ at 25 mm/min, 100 mm/min and 400 mm/min welding speeds are 6.6 μm, 5.1 μm, 4.5 μm, respectively, compared with 13.4 μm for
the base metal. It was revealed that the degree of ferrite refinement increases with the increasing welding speed (decreasing the heat input). This is because the temperature increases with the decreasing welding speeds, as shown in Fig. 5. However, for pearlite, under the higher heat input conditions (400 rpm, 25 mm/min and 100 mm/min), several large pearlite clusters are found in limited areas, while under the lower heat input conditions (400 rpm, 400 mm/min), small pearlite colonies or carbides are uniformly distributed in the ferrite matrix, as shown in Fig. 9(f). This is because in the former case, austenite is formed once above $A_1$, while in the latter case, the transformation does not occur. Under the higher heat input conditions, only the austenite with the small proportion (about 13%) is transformed into the pearlite below $A_1$, and consequently, the pearlite distribution is not uniform. For the lower heat input condition (400 rpm, 400 mm/min), on the other hand, the pearlite in the base metal is stirred and fractured into small pieces, so that the carbides are uniformly distributed within the ferrite matrix. In addition, according to the equilibrium phase diagram, the total amount of carbides appears very low in this material, which suggests that the dissolution of carbides may occur during the FSW, as was reported in the heavily drawn pearlite wire.25)

**Figure 10** shows the microstructures in the stir zones of the friction stir welded S20C steel joints under several welding conditions. Refined ferrite (dark area) + pearlite (light area) structures are mainly obtained in the stir zone at the 25 mm/min welding speed and ferrite (dark area) + pearlite–bainite (light area) structures are mainly obtained at the welding speeds from 100 mm/min. On the bottom of the stir zone, the refined ferrite (dark area) due to recrystallization and pearlite–bainite (light area) are mainly formed. However, at the 400 mm/min welding speed, the bottom of the stir zone (Fig. 10(f)) is ferrite (dark area) + globular cementite (light area), which is usually produced by annealing below or near the $A_1$ temperature after deformation,26) while the middle of the stir zone (Fig. 10(c)) is similar to that of the bottom of the 100 mm/min weld (Fig. 10(e)). In the entire SZ at the welding speeds from 100 mm/min to 300 mm/min, and at the upside of the SZ at the welding speed of 400 mm/min (Fig. 10(c)), a several percent amount of martensite is also observed, as shown by the arrows in Figs. 10(b), 10(c), 10(e). This is consistent with the micro-hardness results, , when the welding speed exceeds 100 mm/min, the hardness profile significantly changes due
to the significant difference in hardness between the ferrite–pearlite and martensite as shown in Fig. 8(b).

**Figure 11** shows the microstructures in the stir zones of the friction stir welded joints of the S35C and S50C steels under several typical welding conditions. At the welding speed of 25 mm/min (Fig. 11(a)), a small amount of pro-eutectoid ferrite, which is formed on the former austenite grain boundaries (elongated dark area), and a large amount of pearlite colonies (light area) are obtained in the stir zones of both steels. However, when the welding speed exceeds a specific value, martensite (large dark area) is also found in the stir zones. The critical welding speed is between 50 and 100 mm/min (Fig. 11(b)) for S35C and between 25 (Fig. 11(d)) and 50 mm/min for S50C. This difference can be explained by the higher carbon content improving the hardenability. In addition, with the increasing welding speed, the structure refinement is promoted and the martensite fraction increases.

### 4. Discussion

#### 4.1. Effect of Welding Conditions

The design of the welding conditions mostly depends on the influence of the peak temperature, which in turn, affects the heat input. However, for the material with a transformation like carbon steel, the cooling rate can also significantly change the resulting microstructures although the peak temperature is still important. Therefore, both the peak temperature and the cooling rate must be considered. In the present study, the welding speed from 25 to 400 mm/min was changed at the constant rotation speed of 400 rpm in order to change the heat input. When compared with the rotation speed, the welding speed is very low (below 3% in this study), therefore, the influence of the welding speed on the strain rate should be negligible. Accordingly, in this study, the welding conditions, which change the welding speed at a constant rotation speed, only varied the temperature at a constant strain rate. As a result, the temperature cycles were obtained, as shown in Fig. 4. It is noted that the temperature of the workpieces increases from room temperature, just before the tool reached the measuring position, to the peak temperature in about several dozens seconds and then generally decreases to below 500°C within ten seconds. Therefore, the peak temperature and welding cooling rate of the workpiece are shown in Fig. 5. Note that the peak temperature and the cooling rate are independent of the type of steel. This can be related to the motor torque of the process being similar for all the carbon steels under the same welding conditions, as shown in **Fig. 12**. Therefore, the flow resistance of the material near the welding temperature would not be significantly dependent on the carbon content. This result is consistent with a report that in the single-phase region, the carbon content dependence was very low for the flow stress of the carbon steels.27)

When the process temperature is controlled below A_{1}, the transformation does not occur, and the ferrite is refined by recrystallization, similar to the Al alloys. The pearlite in the base metal is stirred and fractured into small pieces, so that the carbides are uniformly distributed in the ferrite matrix, as shown in the 400 mm/min joint of Fig. 9(f) and in the bottom area of the 400 mm/min joint in Fig. 10(f).

On the other hand, when the peak temperature is controlled above A_{1}, the SZ structures are significantly dependent on the cooling rate. This is because, according to Fig. 5(a), the transformation should occur during cooling after welding under all the conditions except for the lowest heat input condition (400 rpm, 400 mm/min). However, because the welding cooling rate increases with the increasing welding speeds, as shown as Fig. 5(b), the martensite is formed under the welding speed above a specific value.
Therefore, from Fig. 5(b), it can be stated that the lower critical cooling rate for S20C and S35C in the FSW process is 40 to 80°C/s, and for S50C, it is 20 to 40°C/s. Consequently, the S20C and S35C steel joint strengths have a significant increase between 50 mm/min and 100 mm/min, while that of S50C is between 25 mm/min and 50 mm/min, as shown as Fig. 7. Thus, for the S20C, S35C and S50C steels, not only the refined structures as in the IF and S12C steels, but also the martensite hardening effect at the welding speed exceeding the lower critical cooling rate, affect the joint strength, hence, the results are more complicated than that of the IF and S12C steels.

4.2. Effect of Carbon Content

Figure 13 shows the effect of the carbon content on the hardness in the center of the SZ of the joints with a 19.6 N load for 15 s. This result well corresponds to that of the tensile test shown in Fig. 7. When the carbon content is less than or equal to 0.12%, the hardness in the center of the SZ slightly increases from the base metal along with a lower dependence of the welding conditions. While the carbon content is equal or larger than 0.2%, the hardness in the center of the SZ significantly increases with the increasing carbon content along with a significant dependence on the welding conditions. When the carbon content is less than 0.2%, because the fraction of the equaxed ferrite is high, the ferrite grains refinement is the dominant factor. However, the carbon content is equal to or larger than 0.2%, the matrix ferrite fraction is low, while the quenched microstructure (martensite) fraction significantly increased, so that the quench hardening effect also significantly affects the joint strength. The volume fractions of the martensite, which are quantified by a computer graphics treatment, are shown in Fig. 13(b). As a reference, the hardness of the martensite itself is also shown in Fig. 13(b). Therefore, under the martensite forming conditions, namely, at the welding speeds between 100 mm/min and 400 mm/min (except the bottom at 400 mm/min), both the martensite fraction and the hardness of the martensite affect the hardness of the SZ. As a result, the hardness shows a significant increase from a 0.2% carbon content at the higher welding speeds.

Figure 13(c) shows the influence of the carbon content on the hardness of the SZ under the conditions when the peak temperature is below A₁ (on the bottom at the welding speed of 400 mm/min). It was revealed that under the welding conditions in which no transformation occurs, the joint hardness shows a much lower dependence on the carbon content along with a small increase in the refined microstructure. This should be similar to the case when the welding speed is low (25 mm/min) enough to form the refined ferrite–perlite structure. In this case, while the martensite is formed when the initial structure is the ferrite–perlite, the initial structure is the ferrite–perlite when the initial structure is the globular cementite structure. Martensite is hardly observed in this case, while the martensite is formed when the initial structure is the ferrite–perlite. This difference can be explained by the effect of the cementite size on the transformation and hardenability, because the size of the globular cementite in the previous study was much larger than that of the pearlite in this study. Also, the FSW process is a kind of short time process, including both the quick heating and cooling (both dozens of seconds), as shown in Fig. 4. Accordingly, the coarse cementite cannot completely dissolve into the γ phase, therefore, the γ phase with a lower carbon content decreases the hardenability.

When the initial structure is the ferrite–globular cementite, the joint strength decreases from 300 mm/min with the increasing welding speed. This is due to the influence of the soft initial structure because some parts (lower part) of the joint are not transformed. When the initial structure is controlled to be the ferrite–globular cementite, the martensite formation during the FSW decreases so that the joint’s mechanical properties, such as ductility can be improved.

5. Summary

Five types of ferrite+pearlite carbon steels with different
carbon contents (IF steel, S12C, S20C, S35C, S50C) were friction stir welded under various welding conditions, and the joint properties were systematically investigated. Consequently, the effects of the carbon content and the welding conditions on the microstructure and mechanical properties of the FSW joints were clarified. Additionally, the friction stir welding was successful below A1. As a result, the following conclusions were achieved.

(1) All the carbon steels in this study increased in the strength of the FSW joints from the normal structured (ferrite) base metal to that of the base metals due to the refined microstructures, independent of the carbon content.

(2) The peak temperature decreases and the cooling rate increases with the increasing welding speed. These values are independent of the types of steels.

(3) When the carbon content is less than or equal to 0.12 mass%, the welding produces a ferrite–pearlite structure, and the strength slightly increases with the increasing welding speed (the decreasing heat input) due to the refined microstructure.

(4) When the carbon content is equal to or above 0.2 mass%, under the conditions exceeding the lower critical cooling rate, martensite is formed, resulting in a significant increase in joint strength. Below the critical cooling rate, the joint strength also increases with the increasing welding speed (the decreasing heat input) due to the refined microstructures.

(5) When the peak temperature is controlled to be below A1, friction stir welding can be performed without any transformation. In this case, the joint strength increases from that of the base metals to the refined microstructures, independent of the carbon content.

(6) When the initial structure is controlled to be the ferrite + globular cementite, the martensite formation during FSW can be decreased, so that the joint’s mechanical properties, such as ductility, can be improved.

These conclusions enable us to design the friction stir welding conditions for each carbon steel in order to obtain the required structure and mechanical properties without any post heating treatment.

Acknowledgements

The authors wish to acknowledge the financial support of the Toray Science Foundation, ISIJ Research Promotion Grant, a Grant-in-Aid for the Cooperative Research Project of Nationwide Joint-Use Research Institutes on Development Base of Joining Technology for New Metallic Glasses and Inorganic Materials, “Priority Assistance of the Formation of Worldwide Renowned Centers of Research—The 21st Century COE Program (Project: Center of Excellence for Advanced Structural and Functional Materials Design) “from the Ministry of Education, Sports, Culture, Science and Technology of Japan and a Grant-in-Aid for Science Research from Scientific Research from the Japan Society for Promotion of Science.

REFERENCES


