Microstructures and mechanical properties of friction stir welded brass/steel dissimilar lap joints at various welding speeds

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Abstract

The dissimilar lap joints of commercially brass (Cu–40Zn) to plain carbon steel (S25C) were fabricated using friction stir welding. The relationship between welding speed and heat input during the friction stir welding processes was discussed. The effect of welding speed on microstructures and mechanical properties of brass/S25C dissimilar joints was investigated. The grain size, Vickers hardness at the stirred zone and tensile shear fracture load of the joints varied significantly with the change of welding speed. The transmission electron microscope observations and energy dispersive X-ray spectrometry line analyses at the interface revealed that a mutual diffusion zone of Cu, Zn and Fe, the dominant elements of each plate was formed at the interface. The microstructure evolution and the joining mechanism of the brass/S25C joint are systematically discussed.

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1. Introduction

According to the new manufacturing concept of multi-materials design, applications involving the welding of dissimilar metals are becoming increasingly widespread. Compared to pure copper, brass (a Cu–Zn alloy) has advantages such as higher plasticity, strength, hardness and corrosion resistance; therefore it has been used as a structural material in various industries. In recent years, brass has been used to produce the sliding parts of hydraulic equipment to obtain good tribological characteristics. For sliding parts, hybrid structures of brass/steel dissimilar metal joint are preferred as it requires higher structural strength at a lower cost. However, it is very difficult to clad brass to steel by fusion welding because of the strength loss in the fusion zone due to the evaporation of Zn, as well as the large differences in the heat physical properties between brass and steel, such as melting point, heat conductivity, and heat expansion coefficient.

To solve this problem, solid-state welding process has been tried in recent years. Kimura et al. [1,2] achieved the brass/low carbon steel dissimilar metal welding by friction welding and studied the effect of post-weld heat treatment on joint properties of the brass and low carbon steel friction welded joints. However, the friction welding has a large limitation in shape of the joint, which should be a body of rotation such as a pipe and a rod-type joint.

Friction stir welding (FSW), a solid-state welding process developed by The Welding Institute (TWI), can weld plate-shaped materials as butt or lap joints with large applicability in the dimension of the joint. Since the development of FSW, it has gained considerable interest because it can prevent solidification problems associated with conventional fusion welding techniques [3–5], and has been successfully applied to low melting metals such as Al alloys and Mg alloys [6–12].
In recent years, FSW of dissimilar metal joints has been researched extensively, such as Al alloy/steel [13], Mg alloy/steel [14], Al alloy/copper [15] and Ti/steel [16], but few researches have been already conducted on the FSW of brass/steel dissimilar joint. In previous work [17,18], the FSW process window for joining brass to steel with a constant tool load has been made clear and the effect of tool design on the brass/steel dissimilar joint has been studied. But until this paper, the effect of FSW welding speed or rotation speed on the brass/steel dissimilar joint has not been discussed.

Many parameters can influence the heat input during the FSW processing, such as tool shape, tool tilt angle, welding speed, rotation speed and the load of tool. Optimizing the heat input is very important for FSW joint. The too low heat input may cause many joint defects such as channel defects and unjointed interfaces like kissing bond defects [19]. Also too much high heat input can lead to some defects increasing like big burr, hooking defects and top sheet thinning for lap joints [20].

The total heat input, \( Q \), generated by the FSW can be simply expressed through the following equation (with the constant load).

\[
Q = \frac{R}{V}
\]

(1)

where \( R \) is the tool rotation speed (rpm), \( V \) is the welding speed (mm/min) and \( Q \) refers to the generated heat (J/mm). Eq. (1) can indicate the effect of rotation speed and welding speed on the heat input during FSW processing [21]. It means that increasing the rotation speed at a constant welding speed or decreasing the welding speed at a constant rotation speed can lead to a higher heat input. Although both of rotation speed and welding speed can affect the heat input during FSW processing and that affect the FSW joint properties, but it should be noticed that from an economical point of view, using higher welding speed is a favourable industrial demand [22].

The aim of this study is devoted to determine the effect of welding speed on the microstructure (by the OM, SEM, EDX and TEM) and mechanical properties (by the Vickers hardness at the stirred zone and tensile shear fracture load) of the brass/steel dissimilar joints and the joining mechanism of brass/steel dissimilar joint was discussed.

2. Experimental

Plates of commercially available brass (3 × 100 × 200 mm³), and plain carbon steel (S25C, 5 × 100 × 200 mm³) were lap joined by FSW; with brass used as the top plate and S25C as the substrate plate, as shown in Fig. 1. Table 1 shows the nominal chemical compositions of these materials (data of the inspection certificate of materials). The WC-Co based FSW tool comprises of a concave shoulder of 15 mm in diameter, and probe diameter of 6 mm with a probe length of 2.9 mm, which is 0.1 mm shorter than the thickness of the top brass plate (3 mm). The effect of welding parameters on the brass/steel dissimilar joint was studied with different welding speeds (250, 400, 500 and 600 mm/min) and rotation speeds (1000, 1250 and 1300 rpm) but with the constant tool load of 9.8 kN and tool tilt angle of 3°. During the FSW process, the temperature rise was measured by thermocouples (K type, sheet type) placed at the interface, 3 mm away from the centre of the stirred zone (SZ).

The FSW joints were cross sectioned perpendicular to the welding direction using an electrical-discharge cutting machine for metallographic analysis.

<table>
<thead>
<tr>
<th>Welding speed (mm/min)</th>
<th>Appearance</th>
<th>Cross section</th>
</tr>
</thead>
<tbody>
<tr>
<td>Large burr 250</td>
<td>AS</td>
<td>15mm</td>
</tr>
<tr>
<td>Good lap joint 500</td>
<td>AS</td>
<td>15mm</td>
</tr>
<tr>
<td>Lap joint was not formed 600</td>
<td>AS</td>
<td>Surface of S25C 15mm</td>
</tr>
</tbody>
</table>

Fig. 3. Surface appearances and cross-sections of joints welded with different welding speeds of 250, 500 and 600 mm/min.
The specimens were prepared for optical microscopy by polishing and then etching with an acid solution of 20 mL HCl, 100 mL ethanol and 5 g FeCl₃. The microstructures of the joints were studied using optical microscopy (OM), scanning electron microscopy (SEM) and transmission electron microscopy (TEM). The grain size of the α-phase in the brass SZ near the interface was measured using the line-intercept method, where the mean grain widths of the α-phases on all the 10 lines drawn onto the SEM image magnified Fig. 4.

Microstructures of sound joint welded at a welding speed of 500 mm/min.

![Microstructures of sound joint welded at a welding speed of 500 mm/min.](image)

**Fig. 4.** Microstructures of sound joint welded at a welding speed of 500 mm/min.

<table>
<thead>
<tr>
<th>Welding speed</th>
<th>4000 mm/min</th>
<th>5000 mm/min</th>
<th>6000 mm/min</th>
</tr>
</thead>
<tbody>
<tr>
<td>Macrostructure of SZ (OM)</td>
<td><img src="image" alt="Image" /></td>
<td><img src="image" alt="Image" /></td>
<td><img src="image" alt="Image" /></td>
</tr>
<tr>
<td>Microstructure of SZ (SEM)</td>
<td><img src="image" alt="Image" /></td>
<td><img src="image" alt="Image" /></td>
<td><img src="image" alt="Image" /></td>
</tr>
<tr>
<td>Mean grain size of α-phase in SZ (μm)</td>
<td>3.5</td>
<td>2.4</td>
<td>2.0</td>
</tr>
</tbody>
</table>

**Fig. 5.** Interfaces of the joints welded at welding speeds of 400, 500 and 600 mm/min.

Fig. 5. Interfaces of the joints welded at welding speeds of 400, 500 and 600 mm/min.

![Graph showing hardness profiles at different welding speeds](image)

**Fig. 6.** Typical hardness profiles of the cross section at the brass side 0.5 mm from the interface of the lap joints welded with different welding speeds.

![Graph showing relationship between hardness and grain size](image)

**Fig. 7.** Relationship between the hardness and the (grain size)⁻¹/² of the α-phase in the SZ.
by 3000× for each welded joint are measured. The Vickers hardness profile measurements were conducted on the cross-section of the lap joints using a micro-hardness tester under a load of 0.1 N and a holding time of 15 s at the brass side 0.5 mm from the interface. Element distributions and chemical compositions of the joint interfaces were analysed using an energy-dispersive X-ray spectrometer (EDX) equipped with a SEM or a TEM. In addition, specimens 15 mm wide and 160 mm long were machined from the lap joint samples, and shear tensile test was performed at room temperature with a constant crosshead speed of 1 mm/min. The fractography of the tensile fracture surface is discussed using SEM or EDX investigations.

3. Results and discussion

Fig. 2 shows the effect of tool rotation speed and welding speed on the brass/steel dissimilar lap joint with the constant tool load of 9.8 kN. The range for the good lap joints is very narrow. Because the effect of welding speed and rotation speed on the heat input and joint properties is very similar, in this study, four points in Fig. 2 (1000 rpm; 250, 400, 500 and 600 mm/min) were used to study the effect of welding speed on the brass/steel dissimilar lap joint.

Fig. 3 shows the three typical surface appearances and cross-sections of the joints welded at welding speeds of 250, 500 and 600 mm/min with the constant rotation speed of 1000 rpm in Fig. 2. With excess heat input, the joint welded at welding speed of 250 mm/min acquired a large burr on the retreating side (RS) and a surface groove defect on the advancing side (AS). At the interface of the joint, a part of the S25C was stirred into the brass side by contacting a rotating probe tip, and a big inner cavity was observed. Good joints with a regular surface and a smooth interface were obtained at welding speeds of 400 and 500 mm/min; because their surface appearances are similar, only the joint welded at 500 mm/min was displayed. Although the joint welded at welding speed of 600 mm/min also had a clean and regular surface, but the lap joint was not formed for the insufficient heat input during the welding process by the too fast welding speed.

Fig. 4 shows a typical microstructure in the cross-section of the good joint welded with a welding speed of 500 mm/min. The grain size in the SZ is significantly smaller than that of the base metal (Fig. 4(g)) because of dynamic recrystallization during FSW [23]. At the top of the SZ (Fig. 4(b)), the grain size is very small and uniform, the middle region of SZ (Fig. 4(c)) shows a banded structure, and an onion ring structure is observed at the bottom of the SZ (Fig. 4(d)). The boundary of the SZ at the AS (Fig. 4(e)) is significantly clearer than that at the RS (Fig. 4(f)), and this result can be explained by the higher temperature at the RS of the FSW joint [24]. Fig. 5 shows the macrostructure (OM) and microstructure (SEM, 3000×) of the SZ near the interfaces of the joints welded at welding speeds of 400, 500 and 600 mm/min. The mean grain sizes of the α-phases of brass SZ in the joints are shown at the bottom of each image. The grain size of the α-phase in the brass SZ near the interface was measured using the line-intercept method which was introduced in the Experimental section. The brass metal has a microstructure that consists of two phases, α and β, which appear as convex and concave phases in the SEM photos, respectively. As the welding speed increases, the grain size of the α-phase became smaller. This suggests that with the increase in welding speed, the heat input per unit length of welded joint decreased during FSW processing. Because

Fig. 8. Element mapping at the interface of the joint welded at a welding speed of 500 mm/min, (a) TEM bright field image of the interface, (b) Fe, (c) Cu, (d) Zn area maps.
of the insufficient heat input, the dissimilar lap joint was not formed at the welding speed of 600 mm/min.

Vickers hardness distributions across the joints welded at different welding speeds are shown in Fig. 6. The hardness in the SZ is higher than that in the brass base metal in each case by the reason of grain refinement. As the welding speed increases, the heat input decreased and the grain size became smaller. For conventional polycrystals, with the grain size ranging from several tens to hundreds of micrometres, the dependence of hardness and the strength on the mean grain size can be

Table 2
Diffusion coefficients of Cu and Fe in α-Fe and Cu at 1000 K.

<table>
<thead>
<tr>
<th>Matrix</th>
<th>Diffuse element</th>
<th>D (m²·s⁻¹)</th>
<th>Do (m²·s⁻¹)</th>
<th>Q (kJ·mol⁻¹)</th>
</tr>
</thead>
<tbody>
<tr>
<td>α-Fe</td>
<td>Cu</td>
<td>8.42 × 10⁻¹⁸</td>
<td>4.7 × 10⁻⁵</td>
<td>244</td>
</tr>
<tr>
<td>Cu</td>
<td>Fe</td>
<td>7.55 × 10⁻¹⁶</td>
<td>1.01 × 10⁻⁴</td>
<td>213</td>
</tr>
</tbody>
</table>
described by the Hall–Petch relationship:

\[ H = H_0 + k'd^{-1/2} \]  
\[ \sigma_y = \sigma_0 + kd^{-1/2} \]

where \( H \) is the hardness; \( \sigma_y \) is the yield stress; \( d \) is the average grain size and \( H_0, \sigma_0, k \) and \( k' \) are the material constants [25]. The description of the mechanical behaviour of materials with smaller grains using various models is still a topic of discussion. Some studies have revealed material softening with decreasing grain size (negative slope of Hall-Petch-like relationships) [26–27], while the usual relationship (positive slopes) has been observed in others [28–30]. In the present work, the grain size of the \( \alpha \)-phase fell on the straight line representing the positive slopes as shown in Fig. 7.

In order to clarify the formation mechanism of the joint interface between brass and S25C, a TEM sample of the joint interface was prepared using a focused ion beam (FIB) instrument. Fig. 8(a), (b), (c) and (d) shows the microstructure and area distributions of elements Fe, Cu and Zn, respectively, measured by EDX at the interface of the joint welded at welding speed of 500 mm/min, respectively. Fig. 8(b) and (c) indicates that some small pieces of S25C were cut and stirred into the brass side over the range of approximately 2 \( \mu \)m from the interface. On the S25C side, a grain-refined zone of S25C with a width of approximately 1 \( \mu \)m was observed, which was stirred and grain refined by the dynamic recrystallization [31]. This result can prove that the tip of the rotating probe touched directly to the steel surface and eliminated the contaminants and oxide layer. Some very small inner cavities of approximately 300 nm in diameter were observed at the interface, but were too small to be observed by OM.

Fig. 9(a) and (b) shows the TEM bright field image (BFI) of the joint interface in Fig. 8 and a higher magnification micrograph of the square position in (a), respectively. The result of line analysis of elements is shown in Fig. 9(c), which shows the existence of a mutual diffusion zone of Cu and Zn as major elements of brass, over a very narrow range of approximately 70 nm, and Fe from S25C over a larger range of approximately 120 nm. No intermetallic compound was observed and detected at the joint interface. This result is similar to the report of Luo et al. [32]. Fig. 10 shows the result of temperature measurement of the joint. The maximum temperature at the interface was 420 °C (693 K). However, it should be noted that, for obtaining a stable temperature data, the measured position was not at the centre of the SZ, but 3 mm away from the centre. Therefore, it is expected that the true maximum temperature at the centre of the SZ during FSW processing was higher than 693 K. According to Fick’s first law of diffusion, the diffusion coefficient \( D \) is an important physical quantity, expressed as

\[ D = D_0 \exp\left(-\frac{Q}{RT}\right) \]

where \( D_0 \) is the diffusion constant, \( Q \) is the activation energy of diffusion, \( R \) is the gas constant (8.314 J·K\(^{-1}\)·mol\(^{-1}\)), and \( T \) is the thermodynamic temperature. Table 2 shows the diffusion coefficients of Cu and Fe in \( \alpha \)-Fe and Cu at 1000 K [33]. This shows that the diffusion coefficient of Cu diffusing to \( \alpha \)-Fe is much smaller than that of Fe diffusing into Cu. Thus, the length of Fe diffusion into brass is larger than that of Cu into S25C. This result demonstrates that the joint formation mechanism of brass/S25C lap joint is the mutual diffusion of the constituent elements of each material at the joint interface. The observed results by the TEM are similar to the results of previous work, the intermetallic compound was also not observed at the joint interface by the diffraction patterns [18].

Fig. 11. Tensile shear fracture load of the lap joints with different welding speeds.

Fig. 12. Fracture surfaces of the tensile test sample welded at a welding speed of 400 mm/min.
Fig. 11 shows the results of the tensile shear test for the joints welded at welding speeds of 400 and 500 mm/min. As welding speed decreases, the tensile shear fracture load of the joints increased remarkably. Fig. 12 shows the fracture surfaces of the brass and S25C sides of the joint welded at 400 mm/min. After the tensile shear test, the colour of brass can be observed clearly on the fractured surface of the S25C side, as shown in Fig. 12(a). This indicates that a thin layer of brass attached on the surface of S25C and the fracture occurred at the brass side but near the interface. Fig. 12(d) and (e) is the higher magnification SEM photos of the square area in (b) and (c) for the brass and S25C side, respectively, which show a fine and uniform dimple pattern in each case, and this indicates the ductile fracture of the SZ of the brass. Fig. 13(a-1) to (a-4) and (b-1) to (b-4) shows the distributions of Cu, Zn and Fe on the fractured surfaces of the brass and S25C sides, respectively, which correspond to Fig. 12(b) and (c). No significant amount of Fe was detected on the fractured surface on the brass side. On the contrary, Cu and Zn as major elements of the brass were observed clearly on the fractured surface on the S25C side. From these results, it was concluded that the fracture occurred at the SZ of the brass and near the interface. Thus, it suggests that both the joint area and the strength of the SZ of brass impacted the joint strength. The joint widths measured as the width of the brass-adhered area on the S25C fractured surface were 5.0 and 4.9 mm for the joints welded at 400 and 500 mm/min, respectively. The joint width did not change much with the change at welding speed. With the decrease of welding speed, the grain size of the α-phase in the brass SZ became larger and the hardness decreased. This indicates that the tensile shear fracture load of the joint should decrease with a decrease in the welding speed. However, the tensile shear fracture load of the joint increased significantly as already shown in Fig. 11. A possible explanation for these results could be that the decrease in welding speed increased the heat input and thus promoted the mutual diffusion of elements at the interface, and the small micro-cavities (Fig. 8(a)) that appeared at the interface of the joint welded at higher welding speed would disappear with the increase of heat input by the lower welding speed.

4. Conclusions

Friction stir welding was performed to accomplish the dissimilar lap joints of brass/S25C. At a constant tool load of 9.8 kN and tool rotation speed of 1000 rpm, the welding speed was varied from 250 to 600 mm/min to investigate the effect of welding speed on the microstructures and mechanical properties of the dissimilar joints. The results can be summarized as follows.

1. The range of welding parameters for sound brass/steel dissimilar lap joints is very narrow. The excess heat input caused larger burrs and groove defect on the surface of the joint and the joint was not formed by the insufficient heat input.

2. The grain size, Vickers hardness at the stirred zone and tensile shear fracture load of the joints varied significantly with the change of welding speed.

3. The transmission electron microscope observation and energy dispersive X-ray spectrometry line analyses at the interface of the joint welded at welding speed of 500 mm/min revealed that a mutual diffusion zone of dominant elements of each plate material was formed at the interface, but no intermetallic compound was observed.

Fig. 13. The distribution of elements of the fractured surfaces of the joint with a welding speed of 400 mm/min. (a-1) and (b-1) SEM micrograph of the fractured surfaces of brass side and S25C side. (a-2) and (b-2) Fe. (a-3) and (b-3) Cu and (a-4) and (b-4) Zn area maps.

References


