Friction Stir Welding of Aluminum Alloy AA5052 and HSLA Steel

Mechanical and microstructural characterization of dissimilar friction stir welded butt joints

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ABSTRACT

The as-rolled condition 3-mm-thick Al Alloy AA5052 H32 and HSLA steel IRS M-42 were successfully joined using the friction stir welding (FSW) process in a butt-joint configuration. The effect of tool rotational speed and tool tilt angle on the mechanical and metallographic characteristics of the joint was investigated. The microstructure at the joint interface region was investigated by optical microscopy, scanning electron microscopy, and EDS analysis. The EDS analysis suggested, in all cases, the intermetallic compound layer is formed at the joint interface. The range of tool rotational speed and tool tilt angle that could produce joints with good tensile strength was found to be very narrow. The highest joint strength of about 94% of the ultimate tensile strength of the base Al alloy was obtained at a tool rotational speed of 450 rev/min, a welding speed of 45 mm/min, an axial load of 7 kN, and a tool tilt angle of 1.5 deg. The distributed steel flakes and intermetallic compounds strengthened the stir zone (SZ) at the Al alloy side and, up to a certain extent, alleviated the problem of the characteristic softening in the SZ of 5xxx series Al alloys.

KEYWORDS

- Welding
- Friction Stir Welding
- Dissimilar Joint
- Microstructure
- Mechanical Properties
- High-Strength Low-Alloy (HSLA) Steel
- Aluminum Alloy 5052

Introduction

The motive for efficient joining of aluminium (Al) alloy and steel mainly arose from the need and importance of combining steel's high strength, good creep and fatigue resistance, and formability with Al alloy's low density, high thermal conductivity, good low-temperature strength, and high corrosion resistance (Ref. 1). Typical applications of such dissimilar combinations of steel and Al alloys are in transportation systems (improved strength-to-weight ratio), process industries (high corrosion resistance), and in cryogenics (good strength at low temperatures) (Refs. 1, 2). In the automotive sector, it is important to reduce the weight of vehicles in order to improve their overall performance and fuel economy in view of the quickly depleting fossil fuel reserves and their adverse impact on the environment. But, the main obstacle in the use of Al alloy together with steel is the difficulty of joining these materials by fusion welding techniques; this is because of the large difference in their thermomechanical properties and their strong affinity for forming brittle intermetallic compounds (IMCs) (Ref. 3). Recent studies show that friction stir welding (FSW), a solid-state joining technique successfully applied in cases of similar and dissimilar low-temperature materials, will be a potential candidate for the dissimilar joining of Al alloys and steels (Ref. 4).

With regard to dissimilar FSW of high-temperature materials such as steel and low-temperature Al alloys, the investigations are ongoing and industrial-level implementation is yet to take shape. The bonding mechanism in FSW of dissimilar Al alloy and steel is very different from FSW of similar and dissimilar materials with comparable thermomechanical properties. Watanabe et al. (Ref. 5) testified that the rubbing action of the tool pin mechanically removes the oxide film from the steel faying surface, and the Al alloy, in a fluid-like plastic state, adheres to the activated faying surface of the steel, so that joining between steel and aluminium is achieved. It has been reported that the Al alloy/steel interface with a layer of IMC is important to the weld strength, but the joint strength decreases if the IMC layer is too thick (Refs. 6, 7). Dehghani et al. (Ref. 8) reported that in FSW of dissimilar Al alloy and steel, the inhomogeneous movement of material at the stir zone (SZ) might cause poor joint performance. Springer et al. (Ref. 7) reported that in both solid/solid and solid/liquid interdiffusion experiments with low-carbon steel and Al, a thick IMC layer is formed in the temperature range of 600°–800°C.

Many researchers have described the scientific heat generation mechanism in the FSW process. According to
Schmidt et al. (Ref. 9), the heat generated in the FSW process consists of two components: the heat generation due to friction, and the heat generation due to visco-plastic deformation. In the case of similar and dissimilar low-melting-temperature materials, for a given combination of FSW parameters, the frictional heat generation will be nearly constant and the deformation heat generation will be negligible. But, in the case of FSW of dissimilar Al alloy and steel, because the steel at one side of the joint has very high flow stress relative to the Al alloy, and at a temperature around 500°C, the deformation heat generation is significant (Refs. 9–11). It is reported that an increase in pressure and material strain rate can cause the formation of an IMC layer at relatively lower temperatures (Ref. 12).

With regard to dissimilar FSW of 5xxx series Al alloys and steels, only a few works are reported in the literature. Watanabe et al. (Ref. 5) friction stir butt-joint welded 2-mm-thick Al alloy AA5083 and SS400 cold-rolled steel, and reported they could obtain a joint tensile strength of 86% of the base Al alloy. The authors used a FSW tool made of tool steel that had a straight cylindrical (SC) pin profile with a 15-mm shoulder diameter and 2-mm pin diameter. The highest joint efficiency was obtained at a tool pin offset of 0.2 mm (on pin diameter) into steel and the authors observed formation of an FeAl and FeAl3 IMC layer at the interface, toward the top of the joint. They further reported that joining was not achieved when the Al alloy was placed in the advancing side and steel at the retreating side of the joint. In 2013, Dehghani et al. (Ref. 8) friction stir butt-joint welded 3-mm-thick Al Alloy AA5186 and mild steel St52. They used three different FSW tools with shoulder diameter of 18 mm and pin geometry of 3 and 4 mm SC, and 3 mm SC threaded, but they did not report the tool material used or the joint efficiency obtained. Kimapong et al. (Ref. 13) successfully lap welded dissimilar Al Alloy AA5083 and SS400 steel (both 3 mm thick) using a FSW tool made of JIS-SKH57 tool steel. The tool had a 20-mm-diameter shoulder and a 5-mm-diameter SC pin. In 2013, Movahedi et al. (Ref. 14) lap welded 3-mm-thick Al alloy AA5083 with 1-mm-thick St 12 steel. Movahedi et al. used an FSW tool made of H13 tool steel that featured a 16-mm shoulder diameter and taper cylindrical (TC) pin profile with a pin tip diameter of 3 mm and a taper angle of 26.5 deg. In all the above-mentioned investigations using tool steel as FSW tool material, it is reported that the tool pin wore out in welding run lengths of 100 to 200 mm.

5xxx series Al alloys used for structural application, such as 5052 and 5083, are popularly called workhorses from the structural standpoint. Magnesium in aluminum promotes the formation of brittle IMCs in Al alloy/steel joints (Ref. 15). As a result, joining 5xxx series Al alloy to steel will show more sensitivity to the welding parameters as the joints likely have thicker IMC formation in the weld zone due to the presence of Mg. Al Alloy AA5052 is applied in the manufacture of truck and trailer components, aircraft components, and boat hulls. High-strength low-alloy
(HSLA) steel is one of the most commonly used structural materials for the fabrication of automotive chassis, rail bogies, and ship hulls. Therefore, to reduce weight in automotive and marine sectors, it is necessary to efficiently join the above grades of Al alloy and steel. Although FSW is reported to be a viable method for the joining of dissimilar metals and alloys, and despite its importance from the application point of view, no investigation on FSW of dissimilar Al Alloy AA5052 and HSLA steel is found reported in the literature. Therefore, in the present work an attempt is made to friction stir dissimilar butt-joint welds and characterize Al Alloy AA5052 and HSLA steel.

**Experimental Procedures**

The base materials used in this study were a 3-mm-thick rolled sheet of Al Alloy AA5052 H32 and a hot-rolled sheet of HSLA steel IRS M-42-97. The chemical composition of both the materials is given in Table 1. The as-rolled condition sheets were cut to $100 \times 50 \times 3$-mm size so that the 100-mm length lay in the rolling direction. The faying surfaces were then mechanically polished using 300-grit emery paper to ensure gap-free contact between the faying edges, and the pieces were thoroughly cleaned using acetone. In order to reduce the tooling cost, a compound FSW tool featuring oil-hardened EN31 steel shank and tungsten carbide (WC) pin and shouder was used in the trials. An assembled view of the FSW compound tool used is shown in Fig. 1.

Extensive preliminary experimental trials were carried out to find the best tool pin profile and the best tool axis offset with respect to the joint faying surface. The ultimate tensile strength (UTS) of the joint was considered as the performance criterion. Friction stir welding tools with three different taper cylindrical (TC) tool pin profiles and taper angles of 10, 20, and 30 deg, and one straight cylindrical (SC) tool pin profile with a pin tip diameter of 4 mm and pin length of 2.7 mm were used in the preliminary investigation. The results of the investigation revealed that the WC TC tool pin with a 10-deg taper angle is the best profile at a tool axis offset of 2 mm toward the Al alloy. Therefore, the FSW trials were carried out using the compound FSW tool with the best TC tool pin profile (20-mm shoulder diameter at a tool axis offset of 2 mm toward the Al alloy from the faying surface). A schematic of the tool axis offset used in the FSW trials is shown in Fig. 2.

The various FSW parameter combinations that were experimented are given in Table 2. A large number of initial FSW trials was carried out to fix the constant FSW parameters and the range of tool rotational speeds and tool tilt angles that could produce visually defect-free FSW joints.

Tensile test and metallographic specimens were cut perpendicular to the line of the joint using a conventional milling process. The tensile specimens prepared as per the ASTM 8/E8M-11 standard were tested using a Tinius Olsen (UK) H25KT0125 universal testing machine with a 25 kN capacity at a cross-head speed of 3.5 mm/min. The metallographic specimens were mechanically polished following standard procedures, etched with Nitral and Keller’s reagent, and observed under an Olympus BX51M optical microscope. The joint interface microstructure and fracture surface of the tensile specimens were observed using a field emission scanning electron microscope (FE-SEM), model SIGMA HV manufactured by Carl Zeiss, Germany, with a Bruker Quantax 200–Z10 EDS Detector.

**Table 1 — Chemical Composition of Base Materials**

<table>
<thead>
<tr>
<th>Material</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Al</th>
<th>Cu</th>
<th>Cr</th>
<th>Mo</th>
<th>Ni</th>
<th>Nb</th>
<th>Ti</th>
<th>V</th>
<th>Fe</th>
<th>Mg</th>
<th>Zn</th>
<th>Pb</th>
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<tr>
<td>IRS M-42-97</td>
<td>0.11</td>
<td>0.42</td>
<td>0.32</td>
<td>0.10</td>
<td>0.01</td>
<td>0.029</td>
<td>0.31</td>
<td>0.54</td>
<td>0.001</td>
<td>0.22</td>
<td>0.001</td>
<td>0.002</td>
<td>0.002</td>
<td>Bal</td>
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<td>—</td>
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<tr>
<td>AA5052 H32</td>
<td>—</td>
<td>0.08</td>
<td>0.14</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>0.18</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>0.02</td>
<td>—</td>
<td>0.29</td>
<td>2.33</td>
<td>0.004</td>
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</table>
microhardness at the middle, along the thickness across the joint, was observed using a Mitutoyo MVK-H1 hardness testing machine at an applied load of 25 gf for 15 s.

**Results**

Figure 3 shows one set of joints welded as per the FSW parameter combinations given in Table 2. Within the range of parameters experimented, all the joints were free from visible external defects. The optical macrographs of the joint interface at 400, 450 (highest strength), and 600 rev/min are shown in Fig. 4. The macrographs reveal that the joints were free from internal macro defects such as voids, piping, and cracks, and at 400 and 450 rev/min distinct onion ring formations were observed at the stir zone (SZ) of the Al side. The optical macrographs of the joint interface at tool tilt angles of 0.5, 1.5 (highest strength), and 2.5 deg are shown in Fig. 5. Within the experimented range of tool tilt angle, all the joints were free from internal macro defects and clear onion ring formation was observed at tool tilt angles of 0.5 and 1.5 deg. The onion ring formation is a geometric effect caused by cylindrical sheets of material that are extruded during each revolution of the tool. The cutting through the section of the material produces apparent onion rings. Too low or too high heat generation and increased material mixing at the SZ results in the dissappearance of onion ring formations.

<table>
<thead>
<tr>
<th>Tool Rotational Speed (rev/min)</th>
<th>Welding Speed (mm/min)</th>
<th>Axial Force (kN)</th>
<th>Tool Tilt Angle (degree)</th>
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<tbody>
<tr>
<td>400</td>
<td>45</td>
<td>7</td>
<td>1.5</td>
</tr>
<tr>
<td>450</td>
<td>45</td>
<td>7</td>
<td>1.5</td>
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<tr>
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<td>7</td>
<td>1.5</td>
</tr>
<tr>
<td>600</td>
<td>45</td>
<td>7</td>
<td>0.5</td>
</tr>
<tr>
<td>500</td>
<td>45</td>
<td>7</td>
<td>2.0</td>
</tr>
<tr>
<td>500</td>
<td>45</td>
<td>7</td>
<td>2.5</td>
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</table>

<table>
<thead>
<tr>
<th>Joint Description</th>
<th>Norm. Mass Percent (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
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</tr>
<tr>
<td>At 400 rev/min</td>
<td>—</td>
</tr>
<tr>
<td>At 450 rev/min</td>
<td>0.52</td>
</tr>
<tr>
<td>At 600 rev/min</td>
<td>1.53</td>
</tr>
<tr>
<td>At 0.5 deg</td>
<td>1.87</td>
</tr>
<tr>
<td>At 1.5 deg</td>
<td>3.50</td>
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<tr>
<td>At 2.5 deg</td>
<td>0.90</td>
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**Microstructure and Joint Interface Characteristics**

**Optical Microscopy**

The optical micrographs of the base materials and FSW joints at 400, 450 (highest UTS), and 600 rev/min are shown in Figs. 6 and 7, respectively. There was a narrow zone, about 10–25 μm thick, at the steel side nearer to the joint interface where the grains were slightly deformed and elongated. This may be due to the forging effect created by the tool pin during the FSW process. There was no such distinct thermomechanically affected zone (TMAZ) or heat-affected zone (HAZ) observed on the steel side, probably due to the fact that the temperature experienced was less than 600°C. In the SZ of Al alloy side, the grain was refined due to the well-known dynamic recrystallization effect. Next to the SZ, there was a little bit of deformation and refinement of grains when compared to the base Al, and, thus, a finite TMAZ exists on the Al side. There was no such distinct TMAZ and HAZ observed on the steel side, but grain-refined SZ and distinct TMAZ exist on the Al side. Referring to Fig. 8A, B, at a 0.5-deg tool tilt angle, scraped-off steel flakes with layer formations around were severely distributed in the SZ, particularly toward the top of the joint (refer to Fig. 10A also). As shown in Fig. 8C, D, at a tool tilt angle of 1.5 deg, only a very few steel scrapings were found distributed in the SZ, and there was no layer formation around (refer to Fig. 10B also). From Fig. 8E, F it can be seen that, at a tool tilt angle of 2.5 deg, relatively larger steel segments without layer formation around are distributed in the SZ of Al side as a result of increased material mixing at relatively lower temperatures (refer to Fig. 10C also).

**Scanning Electron Microscopy**

The SEM-BSE images of the joint interface at 400, 450 (highest UTS), and 600 rev/min are shown in Fig. 9, and the results of the EDS analysis at the positions of windows marked in Fig. 9A–C are given in Table 3. The EDS spectra corresponding to the results of the EDS analysis given in Table 3 are shown in Fig. 11A–C. The results of the EDS analysis suggest...
that the layer formations at the joint interface are IMCs with the average chemical composition (wt-%) of 72% Al/28% Fe, 76% Al/24% Fe, and 36% Al/64% Fe at 400, 450, and 600 rev/min, respectively. From the chemical composition, and also referring to the Fe-Al phase diagram, the IMCs formed at the interface are consistent with FeAl₃, FeAl₃, and FeAl at 400, 450, and 600 rev/min, respectively.

At 400 rev/min, the IMC layer formed at the interface was very thin, and no IMC layer was formed toward the bottom of the joint. The highest IMC layer thickness (about 0.7 μm) was observed toward the top of the joint, where the tool shoulder rubbed the workpiece. At the middle, the IMC layer thickness was about 0.2 μm, as shown in Fig. 9A. At 450 rev/min, the thickness of the IMC layer formed at the interface was almost uniform throughout the cross section of the joint, with the greatest thickness being about 1.5 μm toward the top of the joint, and the smallest being about 0.25 μm toward the bottom of the joint. As shown in Fig. 9B, at the middle of the joint the IMC layer thickened to about 1 μm. The overall interface microstructure is a clear indication of the near-optimal heat generation and almost-uniform distribution of the generated heat across the joint cross section. At 600 rev/min, the IMC layer formed at the interface was very thick; the thickest point was about 9 μm toward the top of the joint, and the thinnest point was about 2 μm toward the bottom of the joint. At the middle of the joint, as shown in Fig. 9C, the IMC layer thickness was around 4 μm. Also, the IMC layer formation at the interface had microfissures and no distinct linear boundary on the steel side toward the top. All these observations are results of the higher temperature experienced at the weld zone as an effect of the increased heat generation at 600 rev/min.

The SEM-BSE images of the joint interface at tool tilt angles of 0.5, 1.5, and 2.5 deg are shown in Fig. 10, and the results of the EDS analysis at the positions of windows marked in Fig. 10A–C are given in Table 3. Figure 11D–F depicts the EDS spectrums corresponding to the results of the EDS analysis given in Table 3. The EDS analysis suggests that the average chemical composition (wt-%) of the IMC layer formation at the joint interface is 48% Al/52% Fe, 62% Al/38% Fe, and 61% Al/39% Fe at tool tilt angles of 0.5, 1.5, and 2.5 deg, respectively. Based on the chemical composition and from the iron-Al phase diagram, the IMCs formed at the interface are consistent with FeAl₂, FeAl₃, and FeAl at 0.5, 1.5, and 2.5 deg, respectively.

At a tool tilt angle of 0.5 deg, the IMC layer formed at the interface was relatively thick — about 4.6 μm toward the top of the joint and about 1 μm toward the bottom of the joint. For most of the joint cross section, the IMC thickness was about 2.5 μm, as shown in Fig. 10A. At 1.5 deg, the thickness of the IMC layer was almost uniform throughout the cross section of the joint. The thickest point was about 2 μm toward the top of the joint and the thinnest was about 0.35 μm toward the bottom of the joint. As shown in Fig. 10B, at the middle of the joint, the IMC layer thickness was about 1 μm. At 2.5 deg, the IMC layer formed at the interface was relatively thin, with the greatest thickness, about 1.5 μm, observed toward...
the top of the joint, and an extremely thin or nonexistent IMC layer toward the bottom. At the middle of the joint, the IMC layer thickness was around 0.3 μm, as shown in Fig. 10C.

**Microhardness across the Joint Interface**

The microhardness profiles across the joint interface at the middle along the thickness at tool rotational speeds of 400, 450, and 600 rev/min and tool tilt angles of 0.5, 1.5, and 2.5 deg are shown in Fig. 12A, B, respectively. The hardness measurements were made such that the micro intender penetrates directly over the IMC layer at the joint interface. The base Al alloy and HSLA steel have hardness values of 60±4 HV and 234±5 HV, respectively. In all the cases, the hardness at the joint interface is considerably higher than that at the rest of the joint.

At 600 rev/min, the microhardness at the steel side, nearer to the interface, was slightly more than that of the base steel. Because of HSLA steel’s high hardenability, this higher hardness was probably due to the higher temperature experienced nearer to the interface as a result of higher heat generation at 600 rev/min. Both at 600 and 450 rev/min, the microhardness in the SZ of Al alloy was higher than that of the base Al alloy, and it also showed a zigzag-type variation. This type of hardness variation is attributed to the presence of steel flakes with and without IMC and segments of IMCs distributed in the SZ at the Al side. But, both at 400 and 450 rev/min, in the SZ of the Al side, away from the interface, the hardness falls slightly below that of the base Al alloy. At 400 rev/min, the hardness at the steel side, up to 0.5 mm from the interface, is slightly higher than that of the base steel; this is probably due to the strain hardening effect as a result of the stirring action of the tool pin at a lower temperature.

At all tool tilt angles, the hardness in the SZ of the Al side is higher than that of the base Al alloy, but the zigzag-type variation is more predominant only in respect to the 2.5-deg tool tilt angle. This is due to the fact that at low tool tilt angles the severity of stirring and mixing will be low, and, as a result, fewer steel scrapings will be distributed toward the middle and bottom of the joint cross section. But, at higher tool tilt angles, the increased severity of material mixing causes the steel scrapings to be distributed throughout the cross section of the joint. At a 2.5-deg tool tilt angle, similar to 400 rev/min, the hardness at the SZ of the Al side away from the joint interface is slightly less than that of the base Al alloy.

**Joint Tensile Strength**

The ultimate tensile strength and % elongation of the joints at various tool rotational speeds and tool tilt angles are illustrated in Figs. 13 and 14, respectively. Referring to Fig. 13, the highest UTS (196 MPa, about 94% of the UTS of base Al alloy) was obtained at a tool rotational speed of 450 rev/min. The UTS plot shows a decreasing trend on both sides of 450 rev/min and the UTS is more sensitive to tool rotational speed beyond 500 rev/min. Also, it can be seen that the range of tool rotational speeds that could produce good joint strength is very narrow. The highest value of % elongation (7.8%) corresponds to the highest UTS, and a type of variation similar to that observed in the UTS of the joint is observed on either side of the peak value of elongation. With regard to the effect of tool tilt angle, as shown in Fig. 14, the highest UTS (187 MPa, 85% of the base Al alloy) was obtained at a 2.5-deg tool tilt angle.
MPa, about 90% of the UTS of base Al alloy) was obtained at a tool tilt angle of 1.5 deg. On either side of the maximum, the UTS decreased, but the rate of decrease was gradual. The peak value of % elongation (6.31%) was obtained at a 1.5-deg tool tilt angle where the UTS was the maximum. The % elongation also followed a variation similar to that of the UTS on either side of the maximum. The base Al alloy had an elongation of 14±2%.

Photographic images of one set of fractured tensile specimens are shown in Fig. 15A, B. Referring to Fig. 15A, it can be seen that all specimens except the ones welded at 450 rev/min and 500 rev/min were fractured at the joint interface. The specimen welded at 450 (highest UTS) was fractured on the Al side away from the joint interface at the SZ, nearer to the TMAZ. In the case of the specimen welded at 500 rev/min (tool tilt of 1.5 deg), the location of fracture is partially at the SZ (toward the bottom of the joint), and partially at the interface (toward the top of the joint), and it is the second best joint in terms of UTS. Referring to Fig. 15B, all the specimens except the one welded at a tool tilt angle of 1.5 deg are fractured at the joint interface. As discussed previously, the specimen welded at a tool tilt angle of 1.5 deg is fractured partially at the interface and partially at the stir zone (toward the bottom).

Fracture Surface Analysis

The SEM images of the fracture surface of the tensile test specimens of the joints produced at 400, 600, and 450 rev/min (highest UTS) are shown in Fig. 16. Referring to Fig. 16A, B, it can be seen that the specimens at 400 and 600 rev/min are fractured mostly caused by the brittle fracture mechanism. In the case of the specimen welded at 600 rev/min, the interface brittle fracture can be attributed to the formation of a very thick and brittle IMC layer at the joint interface, because segments of the IMC layer are visible at the fractured surface. But, in the case of the specimen welded at 400 rev/min, the cause for the interface brittle fracture might be a lack of proper interface bonding and strain-induced brittleness as a result of lower heat generation. For the specimen welded at 450 rev/min, the presence of dimples and brittle cleavages clearly indicate the specimen was fractured by a combination of brittle and ductile fracture mechanisms. Toward the top of the welded joint, where the tool shoulder rubbed, the fracture was almost completely caused by the brittle fracture mechanism. But, in the rest of the joint interface, however small it was, a fracture was also associated with the ductile mechanism.

Referring to Fig. 17A, B, it can be seen that at both tool tilt angles of 0.5 and 2.5 deg, the fracture is almost completely caused by the brittle fracture mechanism. The presence of segments of IMC toward the top of the fracture surface (joint interface) at a tool tilt angle of 0.5 deg clearly indicates the brittle fracture mechanism.
nature of the fracture is due to the formation of a thicker IMC layer at the joint interface. But, at a 2.5-deg tool tilt angle, the joint was detached at the interface without any sign of deformation, which indicates a lack of proper bonding. Referring to Fig. 17C, at a tool tilt angle of 1.5 deg, toward the top and bottom of the joint, the fracture is almost completely caused by the brittle fracture mechanism, but for most of the joint, the fracture is a blend of ductile and brittle mechanisms.

Discussion

The tool rotational speed influences heat generation, material strain rate, and material coalescence in the FSW process. When the tool rotational speed increases, both the heat generation and material strain rate increase, and the material coalescence slightly decreases, and vice versa. Using a TC pin tool in the FSW process, the heat generation is the sum of heat generation due to both the frictional and deformation components at the tool shoulder, tool pin side surface, and tool pin tip surface. Assuming perfect slip condition, the heat generation computed as per Ref. 10 corresponding to 400, 450 (highest UTS), and 600 rev/min are 820, 923, and 1231 J/mm, respectively. For a given time duration (welding speed) and FSW axial pressure, the thickness of the IMC layer formed at the joint interface is directly related to the magnitude of heat input (peak temperature) and material strain rate. The variation in IMC layer thickness with respect to tool rotational speed observed at the joint interface clearly substantiates the previously stated influence of tool rotational speed on the heat generation and material strain rate and ultimately on the thickness of the IMC layer formed at the interface.

Appropriate tool tilt angle is required for the tool shoulder to bring the plasticized material from the advancing side and front of the tool pin to the retreating side and back of the tool pin. This process fills the cavity created due to the extrusion and stirring of the material by the tool pin. Correct tool tilt angle provides better stirring of material, allows reduced shoulder plunge, and improves the material coalescence by increasing the forging pressure at the trailing side.

But when the tool tilt angle increases beyond a certain value, it will decrease the heat generation by effectively reducing the frictional coefficient and also decrease the material coalescence at the trailing side. Dehghani et al. (Ref. 8) reported that in FSW of dissimilar Al Alloy AA5186 and mild steel using a SC pin tool, an increase in tool tilt angle from 3 to 5 deg increased the weld heat input and downward forging force and removed the tunnel defect, but formed a thicker IMC layer. However, Seighalani et al. (Ref. 16) found the formation of a tunnel defect when the tool tilt was increased from 1 to 3 deg as a result of a decrease in heat generation due to the increase in gap between tool shoulder and workpiece. In fact, in FSW of dissimilar Al alloy and steel, as steel with higher strength on one side of the joint, an increase in tool tilt angle beyond a certain value decreases the effective contact between the tool shoulder and work, and reduces the heat generation significantly. The observed increase in IMC layer thickness at lower tool tilt angles and decrease in IMC layer thickness at higher tool tilt angles substantiates the previously stated influence of tool tilt angle on the heat generation and eventually on the IMC layer thickness.

According to Rathod et al. (Ref. 17), the hardness of IMCs such as FeAl and FeAl₆ are 470 and 982 HV, respectively. Therefore, the higher values of microhardness observed at the interface confirm the presence of an IMC layer at the interface. At tool rotational speeds of 400 and 450 rev/min and a tool tilt angle of 2.5 deg, the reduction in microhardness at the SZ nearer to the TMAZ may be attributed to the typical softening in the SZ of nonheat-treatable and strain-hardenable Al alloys, owing to the reduction in dislocation density (Ref. 18). At 450 rev/min (highest strength), the observed location of the fracture of the tensile specimen at the SZ nearer to the TMAZ is probable due to this characteristic softening in the SZ. But, when the heat generation is high, the softening at the SZ is not present; this is probably due to the fact that the steel scrapings and IMCs distributed in the SZ of the Al side act as reinforcements and effectively alleviate the typical softening at the SZ. At a tool tilt angle of 1.5 deg, the location of the fracture of the tensile specimen was observed partially at the SZ (toward the bottom) and partially at the joint interface (toward the top), and this indicates the SZ is weak toward the bottom and the interface bonding is not strong enough toward the top. This observation also implies that the typical softening of the SZ might persist toward the bottom. In all other cases, the location of the fracture at the joint interface indicates the weakest zone is the joint interface as a result of either the formation of a thicker IMC layer at the interface, or a lack of complete bonding at the interface.

As stated previously, at higher tool rotational speeds, the material strain rate and heat generation both increase and both these factors augment the formation of a thick IMC layer at the joint interface. Through better bonding results, the increased brittleness and development of microcracks in the thicker IMC layer leads to a sudden fall in the UTS of the joint. The sudden decrease in UTS beyond a tool rotational speed of 500 rev/min may be attributed to the previously stated cumulative influence of higher heat generation and material strain rate on the formation of the IMC layer. At lower rotational speeds, both the reduced heat generation and material strain rate suggest a very thin or no IMC layer formation and poor bonding at the joint interface. As relatively higher temperatures, and hence better bonding, exist toward the top of the joint where the tool shoulder rubs, the decrease in joint UTS is gradual at lower tool rotational speeds. The fact that the highest UTS obtained for a joint produced at 450 rev/min with a near-uniform and thin IMC layer at the joint interface is consistent with the reported literature (Refs. 6, 7) that state a thin layer of IMC at the interface is important for the weld strength. At tool rotational speeds of 450±50 rev/min, the joint strength observed was more than 80% of UTS of the base Al alloy. Therefore, for producing good joints, any tool rotational speed between 400 and 500 rev/min under constant welding speed of 45 mm/min, axial force of 7 kN, and tool tilt angle of 1.5 deg is rational.

As discussed previously, at lower tool tilt angles, the increased amount of heat generation due to better contact of the tool shoulder with the workpiece without significant decrease in material strain rate and the consequent formation of a thicker IMC layer at the joint
interface might be the cause of the decreased joint strength. But, when the tool tilt angle increases, though there is marginal increase in material straining, the effective reduction in contact between the tool shoulder and work material reduces the heat generation, and, hence, results in a lack of proper bonding at the interface. The highest value of UTS at a tool tilt angle of 1.5 deg indicates the heat generation, material strain rate, and coalescence are near optimal at this parameter combination. Also, the joint strength observed over the entire experimented range of tool tilt angles (0.5–2.5 deg) was more than 80% of the UTS of the base Al alloy. This observation indicates that within the experimented range of tool tilt angles, the joint strength is less sensitive to the tool tilt angle. Any value of tool tilt angle between 0.5 and 2.5 deg under constant tool rotational speed of 500 rev/min, welding speed of 45 mm/min, and axial force of 7 kN can produce fairly good joints.

The yield strength of the base HSLA steel and Al alloy was 455 and 176 MPa, respectively. As the values of joint strength attained were far below the yield strength of the base steel, the experiment needs to be carried out at different tool rotational speeds and tool tilt angles by keeping the other parameters constant. From the mechanical and metallurgical characterization of the welded joints, the following conclusions were made:

- Within the range of parameters experimented, the tensile strength of the joint is more sensitive to change in tool rotational speed when compared to the tool tilt angle. The highest joint strength obtained was about 94% of the UTS of the base Al alloy at 450 rev/min under constant welding speeds of 45 mm/min, axial force of 7 kN, and tool tilt angle of 1.5 deg.
- The range of tool rotational speed and tool tilt angles that could produce a sound joint is relatively narrow. For a joint UTS of at least 80% of the UTS of the base Al alloy, it was identified that any value of tool rotational speed from 400 to 500 rev/min (under constant welding speed of 45 mm/min, axial force of 7 kN, and tool tilt angle of 1.5 deg) and any value of tool tilt angle from 0.5 to 2.5 deg (under constant tool rotational speed of 500 rev/min, welding speed of 45 mm/min, and axial force of 7 kN) is appropriate.
- The results of the EDS analysis were consistent with the formation of intermetallic compounds (IMC) such as FeAl, FeAl2, and FeAl3 at the joint interface, and it was inferred that the strength of the joint greatly depends on the thickness of the IMC layer formed at the interface.
- The scraped-off steel flakes and IMCs distributed in the stir zone (SZ) of the Al alloy side act as reinforcements, and the characteristic softening in the SZ of 5xxx series Al alloys due to reduction in dislocation density was alleviated up to certain extent.

Conclusions

In the present work, 3-mm-thick dissimilar aluminum (Al) Alloy AA5052 H32 and HSLA steel IRS M-42 were successfully welded in a butt joint using a FSW technique. The welding trials were carried out at different tool rotational speeds and tool tilt angles by keeping the other parameters constant. From the mechanical and metallurgical characterization of the welded joints, the following conclusions were made:

- With the range of parameters experimented, the tensile strength of the joint is more sensitive to change in tool rotational speed when compared to the tool tilt angle. The highest joint strength obtained was about 94% of the UTS of the base Al alloy at 450 rev/min under constant welding speeds of 45 mm/min, axial force of 7 kN, and tool tilt angle of 1.5 deg.
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References