Role of welding parameters on interfacial bonding in dissimilar steel/aluminum friction stir welds

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In this study, lap welds between Al5754 to DP600 steel (aluminum plate top, and steel plate bottom) were manufactured by friction stir welding (FSW). The effects of welding parameters (i.e. travel speeds and penetration depth into lower steel sheet) on the interfacial bonding, tensile strength, and failure mechanism were investigated. The results show that intermetallic compound of Fe4Al13 was detected at the Al/Fe interface. The weld strength increases significantly by increasing the penetration depth into the lower steel substrate at all travel speeds. The failure mode under overlap shear loadings is premature failure through the aluminum substrate when the penetration depth is more than 0.17 mm, and shear fracture when the penetration depth is less than 0.17 mm.

ABSTRACT

1. Introduction

The use of friction stir welding (FSW) for joining of dissimilar metals combinations in the automotive and manufacturing industries has been widely studied thanks to the fact that FSW offers a number of advantages for dissimilar materials, including: enhanced mechanical properties (i.e. tensile and fatigue), improved process quality, avoiding consumables, lower health and environmental issues, and reduced operating costs [1–3]. In the automotive industry, the focus on the application of FSW has mainly involved: the joining of extruded parts to form “larger extrusions”, sheet joining for tailor welded blanks, and joining of light-weight materials. FSW offers numerous advantages and potential for cost reductions in each of these cases. However, cost-effective and reliable joints between light-weight materials will demand significant development and further consideration. A compelling example of dissimilar FSW can be found in the 2013 Honda Accord, where this technique has been applied for joining the cast aluminum and stamped steel parts of the engine cradle [4–6]. In this case, a notable innovation is the use of a C-frame linear FSW system which exerts all the axial loading on the tool, thus avoiding the need for an extremely stiff and high load capacity robot and fixture to apply the tool force.

The main advantage common to nearly all the techniques is that solid state processing limits the temperature rise within the weld region. This limits the formation or growth of undesirable and brittle intermetallic compounds (IMCs) within the weld which deteriorate strength. Lower peak temperatures also minimize thermal distortion and residual stresses, which can often lead to the fracture of the joint immediately upon cooling of the weld in the case when intermetallic compounds are present and cracks are formed in the joint. Chen and Kovacevic [7] pointed out that the maximum temperature in dissimilar FSW Al/steel is 631 °C on the steel side, which is drastically lower than that in fusion welding. Nevertheless, local melting of aluminum was observed in the weld, which can promote diffusion rate between the steel and aluminum substrates, thus IMCs tend to be formed in the Al/Fe system [7]. It has been reported that Fe-rich IMCs (i.e. FeAl) are not as detrimental to the mechanical performance of the joint as other Al-rich IMCs (i.e. Fe4Al13), since it has been argued that FeAl is more ductile [8]. Also, an IMCs layer will not drastically deteriorate weld strength when the thickness of which is less than 2 μm [9]. Hence, the mechanical properties of weld can be improved by altering the types, distribution and thickness of IMCs, through selecting welding parameters such as travel speed, and penetration depth. During lap welding of dissimilar alloys, the key parameters to be considered include the tool geometry, rotation speed, and travel
speed, as with all other FSW procedures. However, lap welding of dissimilar alloys also requires careful control of the tool pin length, and its penetration depth into the lower sheet material. For example, when aluminum or magnesium alloys are joined to steel, the pin penetration into the steel will rapidly wear away steel-based tools, and to avoid this one may maintain the pin above the sheet in order to promote diffusion bonding between the sheets [10]. That is, bonding could be promoted by an indirect diffusion joining mechanism while maintaining the tool pin around 0.05–0.1 mm above the surface of the lower steel sheet during A15754/DP600 friction stir lap welding. This maintains the flat interface profile between the sheets, and results in fewer intermetallic compounds at the interface. This approach, however, precludes the contribution of mechanical interlocking between the sheets by deformation of the lower sheet into the upper sheet. It can also be difficult to maintain this small distance between the tool pin and the lower sheet steel surface. However, when a WC-based tool is used, the tool may penetrate into the steel sheet during joining without encountering severe wear. In prior work by Chen and Nakata [11], the influence of tool penetration was considered in Mg/Steel FSW joining, and it was shown that a thin interfacial reaction zone could be promoted when new layer of steel is exposed by the tool. Deformation of the steel sheet during tool penetration will also promote mechanical interlocking, which will contribute to joint strength [8]. However, this will also promote formation of intermetallic compounds when aluminum and steels alloys are joined, which may contain pre-existing cracks, have high hardness, and thus limit joint strength [10,12,13]. It should be noted that in comparison, other works involving diffusion bonding and adhesive bonding have always found that strengths are maximized when the thickness of reaction layer or intermetallic compound regions is minimized. For example, in the case of friction stir spot welding of aluminum to steel joining, it has been shown that bond strength deteriorates drastically once the reaction layer thickness exceeds 1.5 μm [14]. Considering this fine scale, it would appear that the FSW technique presents great potential in achieving the maximum theoretical strength between dissimilar joints, since the low temperatures and rapid speed of the process can be most effective in suppressing the growth of intermetallic compounds.

It is obvious that controlling the structure and phases at the interfacial region of dissimilar joints produced by FSW is very complex due to transient thermal cycles and short diffusion time. Since the influence of welding parameters on the structure and strength of the interfacial region remains unclear, the present work aims to determine the contributions of metallurgical bonding (via diffusion of aluminum and iron in the stir zone) and mechanical interlocking due to deformation of the lower steel sheet during FSW lap joining of AA5754 aluminum and DP600 steel sheets. The contributions of each will be assessed using a combination of microscopy, mechanical testing, and fractography.

2. Experimental procedure

The base materials examined consisted of 2.2 mm thick AA5754 aluminum and 2.5 mm thick DP600 dual phase steel, with the compositions shown in Table 1. A displacement controlled manual milling machine was utilized to fabricate FSW dissimilar joints, where a digital readout was used to control the displacement of the tool with a 0.005 mm precision. The tool material was a WC cermet with a 12 mm diameter shoulder, and a 5.1 mm diameter pin which had a length of 2.1 mm, and 3 flats, whose axis was tilted by 2.5° with respect to the vertical axis of the work piece and keeps constant during the process. The tool rotation speed during FSW was 1800 RPM, while travel speeds of 16 and 45 mm/min were compared, and penetration depth of the tool pin into the lower steel sheet increased up to 0.389 mm. The thermal cycle at the interface was also measured using K-type thermocouples placed directly at the edge of the pin boundary between the sheets, and temperatures were logged at a sample rate of 100 Hz.

Optical microscopy was conducted on samples upon etching by 3% Nital to reveal the DP600 steel microstructure. The detail of AA5754 microstructure has been described by Haghsenas et al. [10]. Scanning electron microscope (SEM) characterization with energy dispersive X-ray (EDX) analysis was conducted on as-polished samples. All chemical compositions measured by EDX spectroscopy are reported as wt%. Wavelength dispersive spectroscopy was used to map the distribution of alloying elements, using a CAMECA SX100 electron probe microanalysis (EPMA) system.

The mechanical properties of the joints were measured during overlap shear testing, as well as microhardness testing. Overlap shear coupons were prepared with dimensions of 140 × 30 mm² with a 30 mm overlapped area and tensile tests were performed at a rate of 1 mm/min by using a Tinius Olsen (H10KT) Tensile Testing machine. All of the strength values were obtained by averaging the strengths of three individual specimens made at the same welding condition. Fracture morphologies of the failure specimens were examined by SEM, and X-ray diffraction (XRD) was used to investigate the phases present at the fracture surface.

3. Results and discussion

3.1. Macro-structural feature and SEM analysis

Analysis of specimens was limited to those which endured water-jet cutting for specimen preparation, and so superficially bonded joints were not considered in this study. Fig. 1a and b show the cross-sections of the welds produced using 45 and 16 mm/min at tool penetration depths of >0.07 mm. Increased tool penetration occurred in the sample produced at the welding speed of 16 mm/min, mainly due to the compliance of the FSW equipment. The lower travel speed produced greater heat and resulted in higher temperature at a constant rotational speed due to the increased processing time at a certain distance of the weld. Therefore, enhanced softening of the materials occurred which allowed the tool to penetrate 0.15 mm further into the DP600 steel sheet. Steel then moves up much more into the aluminum sheet and the height of hook reaches the maximum at 16 mm/min. Meanwhile, flash (material extruded upwards by the tool) formed at the surface of the weld due to the shoulder penetration into the aluminum sheet, especially, at the advancing side. Here the material flow is asymmetric with respect to the weld centerline.

As indicated in Fig. 1c, dynamic recrystallization occurs in the steel under the Al/Fe interface because the steel undergo heavy plastic deformation during FSW process, which exhibits equiaxed fine grains smaller than those in base metal (BM); this has been reported by Cho et al. [15] in the friction stir welded joint of high strength pipe line steels. Furthermore, the grain size increases from the Al/Fe interface to BM due to a decrease in deformation strain rates impose by the tool.

The SEM micrographs of the interfacial locations in the weld produced using 16 mm/min are shown in Fig. 2. The presence of
extensive IMCs is detected in both locations on the Al side which is estimated to be Al-rich IMCs. The layer is formed due to mechanical mixing of Fe and Al during friction stir welding. As shown in Fig. 2a, some cracks and voids are also observed at the corner of the hook because of inadequate material flow. Here, the DP600 material is displaced upwards into the AA5754 alloy to a distance of 920 μm. The average thickness of the intermetallic near the centerline at the Al/Fe interface shown in Fig. 2b is measured to be <10 μm. The EDX quantification at zone A reveals a composition of 34.8% Fe, 63.6% Al, and 1.7% Mg, which is consistent with Fe₄Al₁₃ with a small amount of Mg in solution.

An overview of the FSW dissimilar joint produced using 45 mm/min is shown in Fig. 3a, where a defect-free weld was made with no voids. However, a large amount of IMCs are observed within the stir zone, which can facilitate crack propagation along the Al/steel interface. The composition measured in zone B was 33.3% Fe, 64.9% Al, and 1.8% Mg, which is nearly the same as that measured in zone A and consistent with Fe₄Al₁₃. Meanwhile, the average thickness of intermetallic in the center of Al/Fe is over 170 μm, which exhibits layered structure made of steel, aluminum and intermetallic compounds (see Fig. 3c). As shown in Fig. 3b and c, many steel fragments with different sizes scattered into the aluminum, with the largest fragments near the center. This is due to the fact that the steel at the interface was stirred into the aluminum by the tip of tool pin, and the stirring intensity at the edge is greater than that at the center of the tip of pin due to the higher tangential velocity at the edge. It also should be noted that the boundary between aluminum substrate and the long steel flash can be clearly identified. However, the aluminum and steel were mixed sufficiently within the hook region (see Fig. 3b) because the stirring process is more severe close to the hook region. As indicated in Figs. 2 and 3, lower travel speed does not produce more IMCs, however in the present work it may be likely that both of the travel speeds applied were comparatively low, and hence did not produce a significant difference in this regard. It can be suggested that most of the energy input was consumed by stirring of steel for higher penetration depths.

In order to determine the distribution of Al and Fe, the interfacial region of the joint in Fig. 3a was further analyzed by EPMA, as shown in Fig. 4. The EPMA map shows that the majority of the materials produced around the interface are consistent with the Fe₄Al₁₃ phase with a similar composition across the bonded region. Here, many steel particles fractured and interspersed within this intermetallic. The EPMA map for Al indicates that the steel particles also have a boundary layer (appearing in yellow color), all with a similar small fraction of Al and large Fe content, suggesting these steel particles may be outlined by an Fe-rich intermetallic (other than Fe₄Al₁₃).

Following approximately 10 welding trials with 140 mm length each, and various plunge depths, the tool was examined with a macro microscope, as shown in Fig. 5. The observations of the surface indicate that negligible wear has been imposed on the tool.

![Fig. 1. Optical micrographs of AA 5754/DP600 dissimilar FSW joints produced using (a) 45 mm/min and (b) 16 mm/min, (c) microstructure of steel directly under the tip of pin.](image)

![Fig. 2. SEM micrograph of bonded region in weld produced using 16 mm/min, (a) at the edge of the stir zone, and (b) Al/Fe interface.](image)
pin following dissimilar welding. This suggests that the high temperatures imposed at the interface were sufficient to soften the DP600 steel, and suppress the wear of the WC based pin.

In order to determine the thermal history during the process, temperature measurements were conducted using K-type thermocouples positioned at the interface of the sheets and periphery of the pin. Several thermocouples were positioned, however, most of which were damaged by the deformation induced by the tool pin. Hence, the temperature measured in the present investigation is the temperature at the outer periphery of the weld. The maximum temperature profile successfully detected using a travel speed of 45 mm/min as shown in Fig. 6 (as seen a peak of 424.8 °C was measured). This is consistent with the steel microstructures observed in Fig. 1c, which suggest that no phase transformations in the steel occurred above the Ac1 temperature. The temperature increases to approximately 150 °C at a slow heating rate due to original preheating of the sheets, then immediately to the peak temperature at a fast heating rate, the duration for the temperature higher than 400 °C is approximate 12 s. The heating and cooling rates near the peak are on the order of 6.47 to 2.12 °C/sec, respectively. It should be noted that the maximum stable temperature of the Al/Fe and Fe3Al13 is much higher than 424.8 °C, according to Fe–Al binary phase diagram (see Fig. 7).

3.2. Mechanical responses

The distribution of hardness along the centerline of the weld in the vertical direction is indicated in Fig. 8. As seen the hardness decreases gradually to the minimum (66.4 HV) from the top surface of aluminum sheet to the Al/Fe interface, then increases dramatically to the maximum (349 HV) at the layered structure in the Al/Fe interface, and then drops to another minimum (182.3 HV) in the pin following dissimilar welding. This suggests that the high temperatures imposed at the interface were sufficient to soften the DP600 steel, and suppress the wear of the WC based pin.
heat affected zone (HAZ) of the steel, and then increases to the steel hardness of up to 200 HV.

Being a non-heat-treatable (or work-hardened) aluminum alloy, the mechanical properties of Al5754 are greatly influenced by dislocation contribution (i.e. density) and grain size refinement rather than precipitates in the structure. Therefore, softening in the Al sheet can be attributed to the fact that the recovery occurs and the grain sizes near the Al/Fe interface was coarser than that on the upper surface of the Al [16]. Meanwhile, the variation of hardness in steel also can be attributed to the variation of grain size as a whole (see Fig. 1c), and softening in the HAZ of the steel sheet can be attributed to the tempering of the martensite islands in that base material.

The maximum hardness was measured at the Al/Fe interface due to the formation of intermetallic compound there. In order to investigate the IMCs at the Al/Fe interface in more detail, EDX analysis was conducted at the indents (see Fig. 9 and Table 2, respectively). The hardness at the layered structure in location B is 399 HV, and the composition in location B is 32.5%Fe, 63.62%Al, and 3.88%Mg, which is also consistent with Fe₄Al₁₃. The hardness value here is consistent with 470 HV measured for the intermetallic in prior FSW Al/steel joints in prior work by Kundu et al. [17] In comparison, the average hardness in the steel close to the Al/Fe interface is 290–293 HV, where the composition is mainly Fe (98.31%), with a small amount of Al, due to that aluminum has limited solubility in iron (see Fig. 7). These hardness values in the steel are consistent with the temperatures measured suggesting the stir zone region remained below the steel transformation temperature.

In order to investigate the effects of tool pin penetration into the steel on the strength properties, the overlap shear tests were performed for 30 and 20 mm wide joints produced using different penetrations. As indicated in Fig. 10, the results suggest that the tensile strength decreases and then increases with the increasing of penetration depth for both travel speeds. The maximum failure strength of 236.4 N/mm was obtained at the welding condition of the travel speed of 45 mm/min and penetration depth of 0.389 mm.

Table 2

<table>
<thead>
<tr>
<th>Spectrum</th>
<th>In stats.</th>
<th>Mg</th>
<th>Al</th>
<th>Fe</th>
<th>Total</th>
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<td>A</td>
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<td>1.69</td>
<td>98.31</td>
<td>100.00</td>
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<tr>
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<td>Yes</td>
<td>3.88</td>
<td>63.62</td>
<td>32.50</td>
<td>100.00</td>
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</table>

A weld with comparable strength can be obtained by maintaining the tip of pin approximately 0.1 mm above the Al/Fe interface, since this promotes an interfacial layer with fewer cracks at the Al/Fe interface through an indirect diffusion joining mechanism [10]. As shown in Fig. 11, the aluminum surface is imposed onto the lower steel sheet by the tool at the Al/Fe interface, and the width of the bonded region is consistent with the diameter of pin. However, the surface of steel is rather flat, preventing the formation of mechanical interlocking at the Al/Fe interface (by the displacement of the lower steel sheet into the upper aluminum sheet). The interfacial layer is similar to what has been found by Gendo et al. [18], in which diffusion bonding was formed by diffusion of the coating layer at the steel surface into the aluminum sheet.

As indicated in Fig. 11, chaotic mixed structures with a mass of defects such as cracking and voids, were produced by penetrating the pin a small distance into the steel substrate (less than 0.078 mm), which counteracts the contribution of the mechanical interlocking effect and is responsible for the decreasing of weld strength. However, when the penetration depth reaches 0.092 mm, intermixing of the two sheet was enhanced and metallurgical bonding occurs with fewer defects at the Al/Fe interface within stir zone, thus improving the strength of the weld. It is worth noting that a more IMCs were formed in the interface with intermediate penetration depths (0.092–0.17 mm) than that at higher or lower penetration depth (see Figs. 2 and 3), which tends to deteriorate the weld strength when the crack propagates along the interface. Beyond this penetration depth, much more DP600 material is displaced upwards into upper Al5754 sheet, the elongated steel flash promoted a mechanical interlocking effect at the weld edges [19]. The displaced DP600 material in the aluminum sheet appears to have provided more surface area to disperse the intermetallic compounds, thus resulting in a slightly lower overall thickness. Hence, it can be concluded that the penetration depth plays a crucial role on determining the strength of the weld.

Two types of failure modes were observed during overlap shear loading: either shear fracture occurred through interface, or premature fracture through the aluminum substrate. As shown in Fig. 12a, failure occurred at the Al/Fe interface, when the pin penetrated less than 0.17 mm into the lower steel substrate, which can be attributed to the inferior bonding at the interface. In addition, the steel flashes at the weld edge are not strong enough to preclude the crack from propagating into the interface (see Fig. 11). Furthermore, the formation of brittle IMCs is also a critical factor deteriorating the mechanical properties of dissimilar Al/steel welds. When the penetration depth is higher than 0.17 mm, a weld with significant steel flash or hook-shaped features at the weld edge is formed (see Fig. 1). Hence, when the weld is subjected to external load, the crack initiates at the tip of the long steel flash and propagates along a short distance and finally into aluminum substrate. Therefore, the failure occurred through aluminum substrate (see Fig. 12b). Under such circumstances, the IMCs at the interface scarcely influence the weld strength since the crack does not propagate along the center interface below the pin.

To further investigate the failure mechanism and identify the intermetallic compounds formed at the Al/Fe interface, SEM and
XRD analysis were performed on the failed fracture surfaces. Fig. 13 reveals SEM micrographs of fracture surface from the Al/Fe interface and the Al substrate near the elongated steel flash (see Fig. 12b). Fig. 14 displays the XRD spectrums obtained from the fracture surface of shear fracture and Al substrate. As indicated in Fig. 13a, the fracture surface at the Al/Fe interface, whose location corresponds to the interface in Fig. 2b, is rather brittle. IMC corresponding to Fe₄Al₁₃ was detected at this surface (see Fig. 14a), which partially contributes to the brittle fracture surface. In addition, the presence of a peak corresponding to AlFe was observed at the fracture surface (see Fig. 14b), which is consistent with the Al/Fe interfaces appearing as an intermediate chemistry in Fig. 4, however this is only suggested to be AlFe since only one peak was detected. As a Fe-rich IMC, AlFe is much ductile than Fe₄Al₁₃ [20]. As indicated in Fig. 13b, the fracture surface of Al substrate at the long steel flash is comparatively ductile, as suggested by the boundary between Al substrate and long steel flash clearly identified in Figs. 2a and 3b.

4. Conclusions

The role of welding parameters (penetration depth into lower steel sheet and travel speed) on the interfacial bonding and mechanical performance of friction stir lap welded AA5754 and DP600 were investigated. The following conclusions can be drawn:

1. Weld of Al 5054 plate and DP600 steel plate (Al plate top, steel plate bottom) with excellent mechanical properties was successfully manufactured by friction stir welding.
2. Higher penetration depth resulted in less intermetallic compounds at the Al/Fe interface.
3. Penetration depth into the steel substrate plays a decisive role in determining the weld strength.
4. The micro-hardness distribution across the joint indicates that the micro-hardness in the joint interface is greater than the base materials.
5. There is a correlation between the penetration depth into the lower steel sheet and the failure mode. In other words, premature failure through the Al sheet occurs when the penetration depth is not lower than 0.17 mm into the lower steel substrate. Shear fracture occurs when the penetration depth is lower than 0.17 mm.
6. Intermetallic compound of Fe₄Al₁₃ was detected at the fracture surface, which are responsible for the deteriorated weld strength at lower penetration depth.

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