Microstructure and fracture behavior of friction stir lap welding of dissimilar AA 6060-T5/ Pure copper

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A R T I C L E I N F O	A B S T R A C T
	This study aims to understand the uncertainty about the optimum or best pin position (Dp) for friction stir lap welding (FSLW) of Al-Cu. Tensile shear testing is used to determine the Mechanical strength of FSL welds under static loading. Fracture strength (σLap) corresponding to the maximum load in a test over the sample width is used as the strength value. Interface microstructures differ depending on whether the tool pin penetrates the lapping interface. It has been found that σLap values of the defect free weld samples vary quite significantly and in general are significantly higher than those reported in the literature. When the pin penetration is close to zero no intermetallic layers were formed, hence the value of σLap was zero. When the pin penetration is 0.4mm, the commonly observed a thin Al–Cu interface layer forms and this layer does not grow beyond 3μm. It is shown that the thin interfacial layer can withstand a high tensile-shear load and thus the adjacent Al material shears to fracture. When the pin penetrates more than 0.4mm, the commonly observed mix stir zone (MSZ) forms and values of σLap are lower than that of 0.4mm pin penetration welds but remain quite high.

1. Introduction

Friction stir welding (FSW) is a solid state welding process invented and patented by The Welding Institute (TWI), for butt and lap welding of ferrous and non-ferrous materials and plastics (Thomas et al., 1993). As a good ability of this technique, sound quality weldments for similar Al-Al and dissimilar Al-non Al alloys can be obtained using FSW joining method (Wang et al., 2018; Zhang et al., 2018; Kumbhar & Bhanumurthy, 2012; Lomolino et al., 2005; Torabi et al., 2018; Pao et al., 2001; Zettler et al., 2006; Sun et al., 2013; Aliha et al., 2018; Akbari et al., 2016). In general, it is well known that fusion welding of one metallic alloy to another with considerably higher melting temperatures (referred to as a large ΔT *Melting* couple, in this case it is Al to Cu) is important in many industries but are very challenging due to physical mismatches such as differences in melting temperature, thermal expansion and thermal conductivity can make the joining almost impractical using conventional welding techniques. Formation of thick intermetallic layers (due to high heat input and liquation of aluminum) is known to deteriorate mechanical properties of the joints (Mishra & Ma, 2005; Mishra & Mahoney, 2007; Nandan et al., 2008; Threadgill et al., 2009; Murr, 2010; Lohwasser & Chen, 2009; Rai et al., 2011). Therefore, development of FSW as a solid-state joining process is of high importance from both scientific

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and industrial point of view. Fig. 1 illustrates FSLW during which a section of lapping surfaces of the top and bottom plates is stirred and mixed in the stir zone (SZ) thus forming a weld behind the tool. In FS welding of a large ΔT *Melting* couple, aided by frictional and deformation heat, metallurgical bond is established through diffusion and subsequent formation of interfacial intermetallic. A metallurgical bond is a condition for a quality joint and a metallurgical bond implies low electric resistance, although intermetallic are commonly viewed to affect joint strength adversely. The cited studies on FS Al-to-Cu (including : Galvao et al., 2011, 2012a,b; Elrefaey et al., 2004, 2005; Abdollah-Zadeh et al,. 2008; Saeid et al., 2010; Xue et al., 2010, 2011a,b; Satya Narayana Gupta et al., 2012; Firouzdor & Kou, 2012; Akbari & Behnagh, 2012; DebRoy & Bhadeshia, 2010; Ouyang et al., 2006; Aliha et al., 2018, 2019; Esmaeili et al., 2011 a,b; Liu et al., 2011; Genevois et al., 2011; Sharma et al., 2018) basically converges on three the general aspects;1: Al-Cu couple being FS weld able, 2: it is difficult to achieve sound FSW dissimilar Al-Cu joints due to the formation of brittle interfacial intermetallic layer compound (IMCs) at the interface and 3: it is hard to control the formation of IMCs. Early investigation by Elrefaey et al. (2004, 2005) on Al-Cu FSLW clearly established that the tool pin (0.1 and 0.2 mm) penetrating to Cu are the conditions for a metallurgical joint to be established at the Al-Cu interface, resulting in a good joint strength. Although detailed quantification was not done in their study, it was clear from their micrographs that the interface region of welds made with pin penetration is a highly irregular structure of mix layers. Ouyang et al. (2006) attributed the poor weldability to various brittle IMCs formed in the Nugget zone (NZ). It is well documented that several parameters, such as tool offsetting, rotation rate, and traverse speed, influenced the weld properties of the dissimilar Al-Cu FSW joints by Carlone et al. (2015). Abdollah-Zadeh et al. (2008) and Saeid et al. (2010) reported irregular shape and inhomogeneous distribution of IMCs at the interface during FSLW of Al-Cu with minimum pin penetration. Xue et al. (2010, 2011a,b) study was an attempt to correlate the σLap to the lower rotation rate and larger bonding area, under the condition of pin penetration. Their data shows that increasing the pin diameter results in a larger Al-Cu bonding area exhibited a higher failure load of 268N/mm with the specimen failing in the HAZ on the aluminium side. However, the meaning of the referred larger bonding area results in wide area intermetallics is unclear and difficult to understand. Furthermore Xue et al. (2011a) reported that Al_4Cu_9, and Al_2Cu and formed at the interface region, based on x-ray diffraction analysis. However, assigning the structure of Al_4Cu_9, and Al_2Cu to a mixed region relying on x-ray diffraction analysis is not reliable due to very low intensity. Sharma et al (2018), attributed the effect of different FSW tool pin profiles on the microstructure and the hardness. Firouzdor and Kou (2012) reported weld strength equal to 183 N/mm for conventional Al-Cu FSL weld when the pin penetration at least 0.1mm to Cu. This is a very low weld strength value and from their micrographs a continuous metallurgical bond in Al-Cu interface cannot be confirmed, However, good metallurgical bonding and reasonable tensile properties were ensured when modified lap was used, with increase in weld strength close to 342 N/mm. Satya Narayana Gupta et al. (2012) suggested that better joint properties are obtained in the joints fabricated using straight fluted tool compared with tapered tool and their Micrographs of stir zone are not clear, and Al-Cu interface joint features are not provided. In this paper FSLW of Al-Cu, to explain how interface microstructures affect the fracturing process during tensile-shear testing and thus joint strength. A possible control method for producing Al-Cu welds for a higher joint strength can then be suggested.
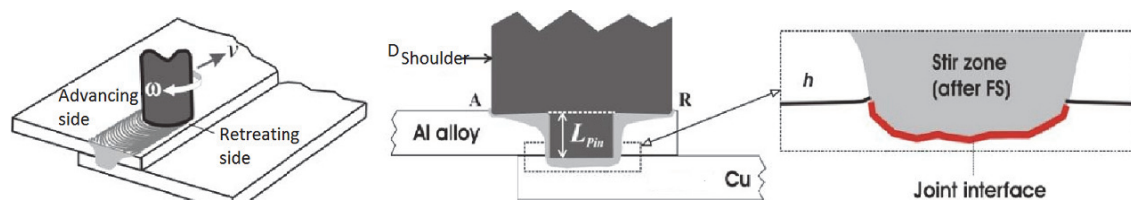

Fig. 1. Schematics illustrations of FSLW of Al-Cu

2. Experimental procedure

All FSLW experiments were conducted using FV200 milling machine and thus the mode of FS was displacement control. Schematic illustration of FSLW process has already been provided in Fig. 1. Fig.

2 shows an actual FSLW experiment. A Lowstir^{TM} device, which is also shown in Fig. 2, was used in each FSLW experiment to monitor the downforce. Al 6060-T5 (300x100x6mm) alloy placed on top of the and pure Cu (300x100x2mm) work pieces were FSL welded. A sufficiently thick top plate is used to avoid fracture at the HAZ during tensile shear testing, and to instead cause fracture along the Al-Cu interface. A major series of experiments were conducted with 1400 rpm, (and 710 rpm is only used to demonstrate the difference in temperature at SZ, Fig 5) as a rotational speed (ω), 60mm/min as a traverse speed (v),3 degrees of tilt angle(Θ) and tool pin penetration Dp= 0 mm and Dp=0.4 mm respectively from the top surface of the bottom plate (Fig.3). Tools were made using H13 tool steel and the left-hand threads of the pins were made with a 1 mm pitch and a 0.6 mm actual depth. The diameter of the concave shoulder was 18 mm and the pin outside diameter was 6 mm. K-type thermocouple was used to measure the FSLW interface temperature.
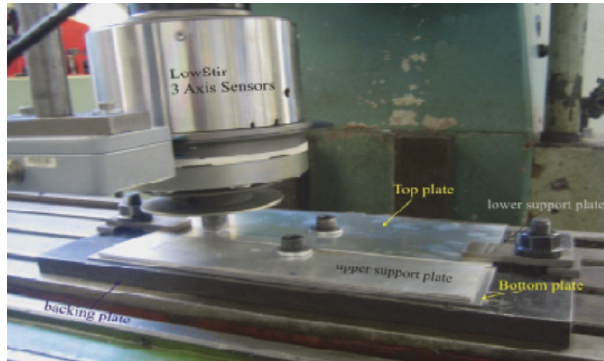

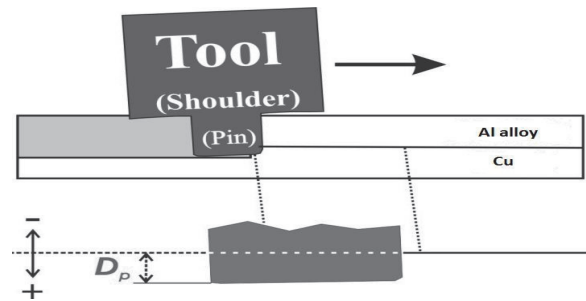**Fig. 2.** FSLW using a FV200 milling machine with a Lowstir^{TM} force measuring device

Fig. 3. Schematic illustration of tool positioning during FSLW showing pin penetration depth

Tensile-shear testing of FS lap welds has been the major method used for evaluating strength of FSL welds in literature. Test samples, 16 mm wide, perpendicular to the welding direction were machined from the welded plates. Fig. 4 illustrates the positioning of a sample together with supporting pieces. Samples were tested at a constant crosshead displacement rate of 3 mm/min using a 50 KN Tinus Olsen tensile testing machine, with a 50-mm extensometer attached. The strength of a lap sample cannot be expressed using the normal load/area, as the stress distribution along the joint area during tensile-shear test is highly uneven. Instead, maximum failure load in a test divided by the width of the sample, σLap, is taken as strength
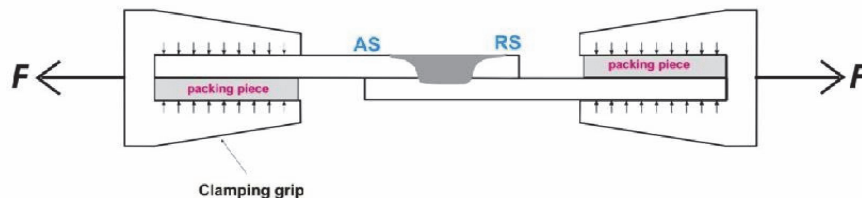

Fig. 4. Schematic illustration of tensile-shear testing

3. Results and discussion

3.1. Temprature measurement

To justify FSW is solid state welding process, (no melting of base metals). K-type thermocouple was placed at Al-Cu interface to monitor the temperature. Traces of temperature at stir zone (SZ) of these welds, during FSLW, are also presented in Fig 5. In each trace, there are disturbances in the peak

temperature region because the thermocouple was pushed slightly by the lower stir flow as the pin approached the thermocouple. The weld made using ω=1400 rpm has obtained higher peak temperature (Tsz =520°C) and spent longer time at the elevated temperatures, compared to the weld made using ω=710 rpm (with Tsz =445°C).
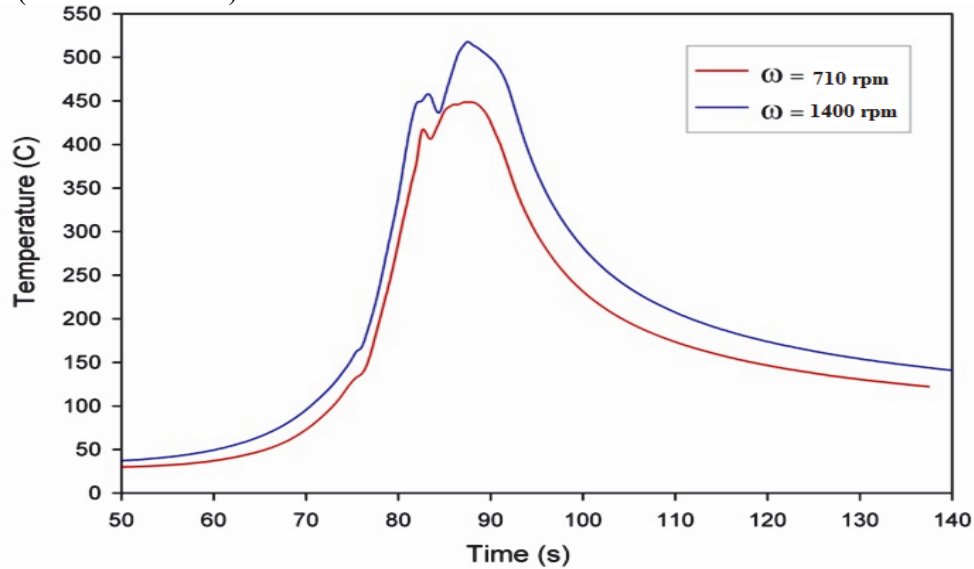

Fig. 5. Measured temperature at SZ of welds made using ω=1400 rpm and and ω=710 rpm with v=60mm/min

3.2 Microhardness distribution in stir zone

As the larger grains have less grain boundaries hence they would impose less restriction to the dislocation movement, resulting in lower hardness and local strength. Therefore, the SZ exhibited the lowest hardness, compare to the other region, as it had the largest grains hardness of SZ for the weld made using ω=1400 rpm is considerably lower than that of the weld made using ω=710 rpm (Fig. 6). This is due to difference in grain size produced at SZ of these welds, as shown in Fig 7. The grains in the weld made using ω=1400 rpm is considerably larger than the grains in the weld made using ω=710 rpm. Traces of temperature at SZ of these welds, during FSLW, are also presented in Fig5. Consequently, the recrystallized grains at SZ of the weld made using ω=1400 rpm could grow more, resulting in larger grains with subsequent reduced hardness.

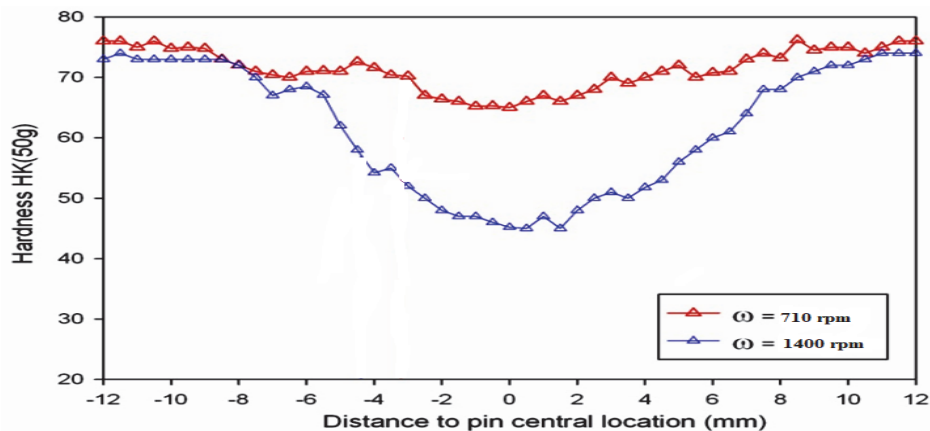**Fig. 6.** Knoop hardness distribution (transverse to welding direction measured for the two welds made using ω =1400 rpm, 710 rpm and v=60mm/min

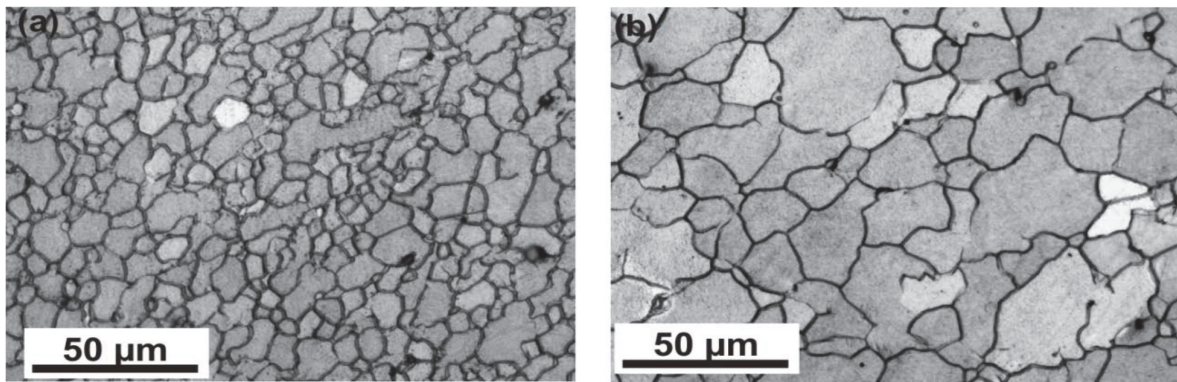**Fig. 7.** Microstructure of SZ for the welds made using ν =60 mm/min: (a) ω =710 rpm and (b) ω =1400 rpm

3.3 Microstructure analysis

Three selected samples are shown here to illustrate the importance of interface microstructures and based on this illustration a suggestion of FSLW control for maximum strength can then be made. Fig. 8(a) is the first example where no intermetallic compound layer (IMC) were formed for Dp≈0, It should be noted that there is no research work available in literature for this condition especially for Al-Cu FSLW. The reason for not forming any intermetallic layer at the Al-Cu weld interface may be because copper has conducted heat faster, since copper has higher thermal conductivity, but for the similar FSLW conditions (pin penetration, Dp≈0) for Al-Steel FSLW, a continuous and thin intermetallic layer has been reported (Shubhavardhan & Surendran 2018). This is just to compare the effect of FSLW pin penetration (Dp) on different large ΔT *Melting* couple, however, FSLW of Al-steel and Al-copper is totally different, cannot compare the results), with maximum value of σLap. When *Dp*≈0.4 as shown in in fig. 8(b), a thin and continuous interface intermetallic layer (up to 3μm in thickness) is observed. When the pin penetration is further increased a non-uniform, irregular intermix MSZ can be observed which is shown in fig 8(c). The area of MSZ largely corresponds to the area of the pin penetrated copper and this zone is a mixture of Al-Cu intermetallic thin pieces embedded in the recrystallized Cu grains.
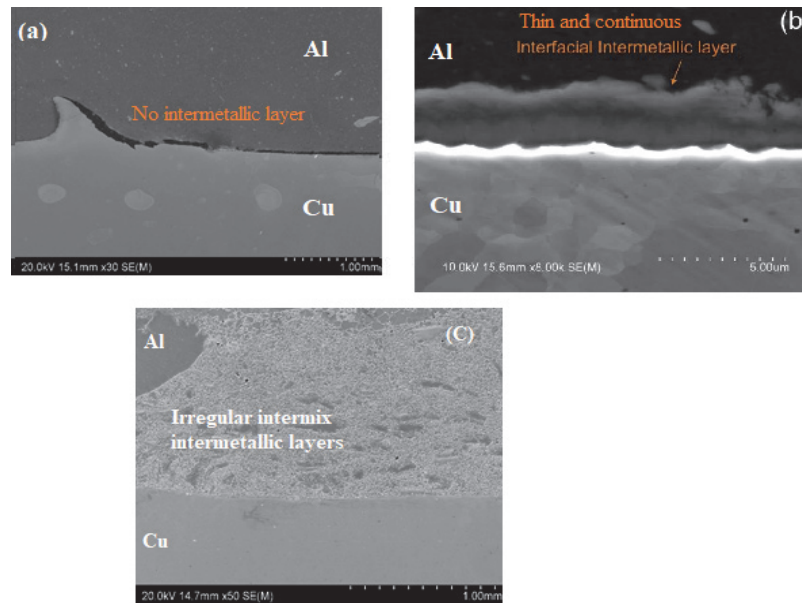

Fig. 8. Cross sectional view of an Al-Cu FSL weld made with ω = 1,400 rpm, *v* = 60 mm/min, (a) *Dp*≈0 with no intermetallic layer formed at the interface, (b) *Dp*≈0.4 mm thin and continuous intermetallic layer formed, and (c) *Dp*>0.4 mm displaying irregular intermix intermetallic layer (MSZ)

In Fig. 8(b) as mentioned earlier a thin and continuous interface intermetallic layer can be observed. A metallurgical bond between Al and Cu is established due to the diffusion of Cu and Al, and thus a slight pin penetration $Dp \approx 0.4$, (referring to Fig. 3) is commonly believed to be the condition for a good weld strength. Naturally, no intermetallic layer or a MSZ cannot form and if $Dp < 0$. However, FS tool can be position controlled so that $Dp \approx 0$. In this case, although there can still be an absence of intermetallic layer and MSZ, but it is totally different in case of Al-steel FSLW. Shubhavardhan and Surendran (2018), because as has already explained the thermal conductivity of steel is lesser than copper, the reason could be steel conducted heat far slower than the copper, forming a thin and continuous intermetallic layer at Al-Steel FSLW interface for $Dp \approx 0$. A thin Al-Cu interface intermetallic compound layer can form when pin penetration Dp is 0.4mm, metallurgical bonding the top and bottom plates together, as demonstrated by an example shown in Fig. 9.
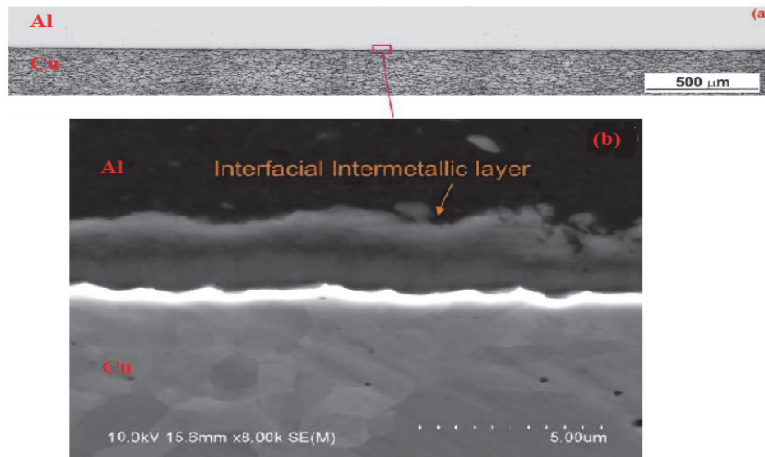

Fig. 9. Cross sectional view of an Al-Cu weld made with $\omega = 1,400$ rpm, $v = 60$ mm/min and $Dp \approx 0.4$ mm, (a) optical macrograph and (b) SEM micrograph of Al-Cu interface displaying no defect but a continuous thin interface layer

Furthermore, formation of intermetallic layers at the Al-Cu interface was further investigated using EDS spot analysis. Typical EDS spectra of intermetallic layers at the Al/Cu interface and inside the stir zone for $Dp \approx 0.4$mm is presented in Fig 10. At point 1 we can notice presence of copper rich phase, and Al rich phase at point 5, point 2 indicate that the thin layer at Al-Cu interface is mainly Al_4Cu_9 intermetallic, according to Al-Cu phase diagram, On the other hand, point 3 and 4 indicate presence of AlCu intermetallic layer, the chemical composition of 61.8at%Al and 38.2at%Cu in Fig 10 at point 6 may indicate that the thin layer within the stir zone is likely to be Al_2Cu intermetallic phase. Some of the observed intermetallic layers here were also found in literature (Elrefaey et al; 2004, Xue et al; 2011a,b). However, due to the analytical spot size being large in SEM/EDS hence assigning the structure of Al_2Cu, AlCu or Al_4Cu_9 to the small size intermetallic layers is not very reliable.
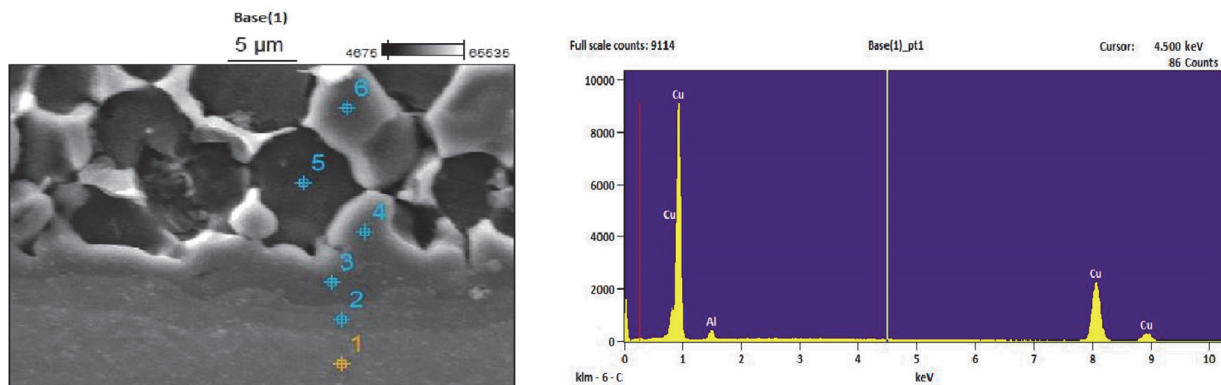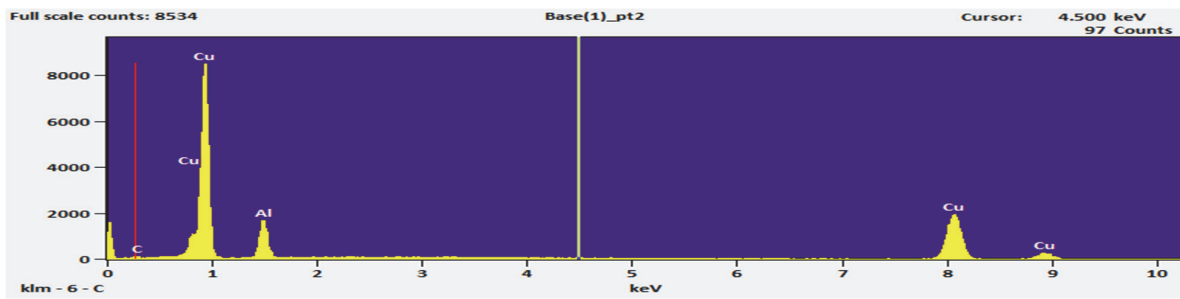
Point 2; $Al_4Cu_9(\gamma_1)$
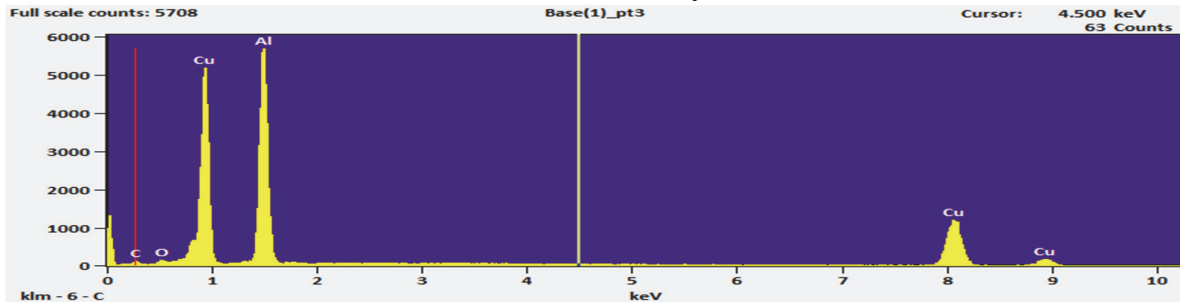
Point 3; $AlCu(\eta_2)$
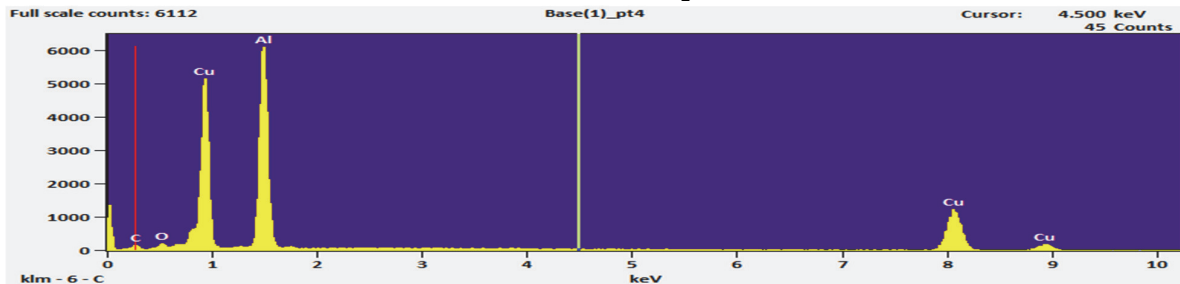
Point 4; $AlCu(\eta_2)$
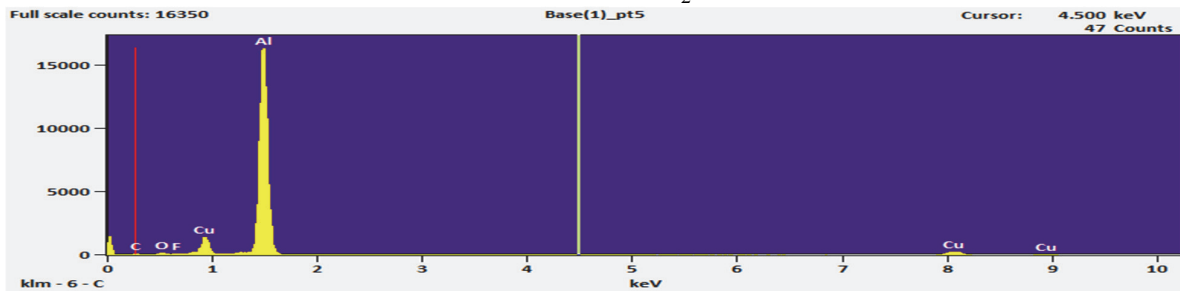
Point 5; Al
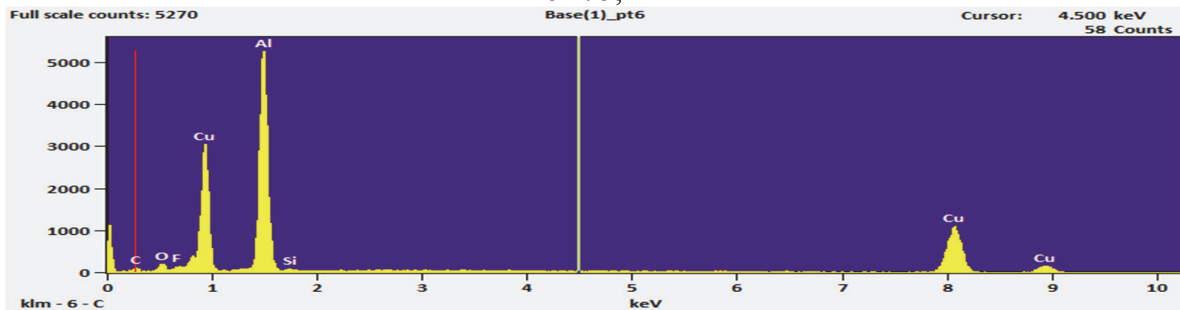
Point 6; $Al_2Cu(\theta)$

Fig. 10. SEM micrograph showing the intermetallic layers at Al/Cu interface, EDS spectra of intermetallic layer at different locations in stir zone are also shown

3.4 Al-Cu Fracture strength

Before discussing the effect of Dp on fracture behavior, the strain measurement should be discussed first. Unlike traditional tensile testing where the sample is elongated axially, in a tensile-shear test, the extension is slightly axially off and there is a degree of local rotation during testing if sufficient plasticity of the sample allows. Thus, the axial strain measurement by the extensometer is slightly distorted due to the slightly local twisting/bending. Furthermore, particularly in the initial stage of elongation (largely within the linear range), the extension values measured by the extensometer were slightly unstable. Thus, determined strain values are not highly accurate but the use of them should be adequate for indicating plasticity. Fig 11 shows selected tensile-shear curves. For the penetrated sample more than 0.4mm, there is a significant amount of fracture energy (Ef), represented by the area under the curve. The weld strength was 302 N/mm, which is obviously much higher than that of Dp ≈0 (0 N/mm, which is not discussed furthermore) but is considerably lower than that of Al-Cu FSL welds (> 450 N/mm), with Dp≈0. 4mm.The weld strength of 302 N/mm is higher than the values of 268 N/mm which is the maximum value obtained for samples using a pin penetration (P Xue et al;2011). In this latter study, when a weld is free of macro- defects the strength equivalent value is close to that maximum value, regardless of what the FS speeds condition.
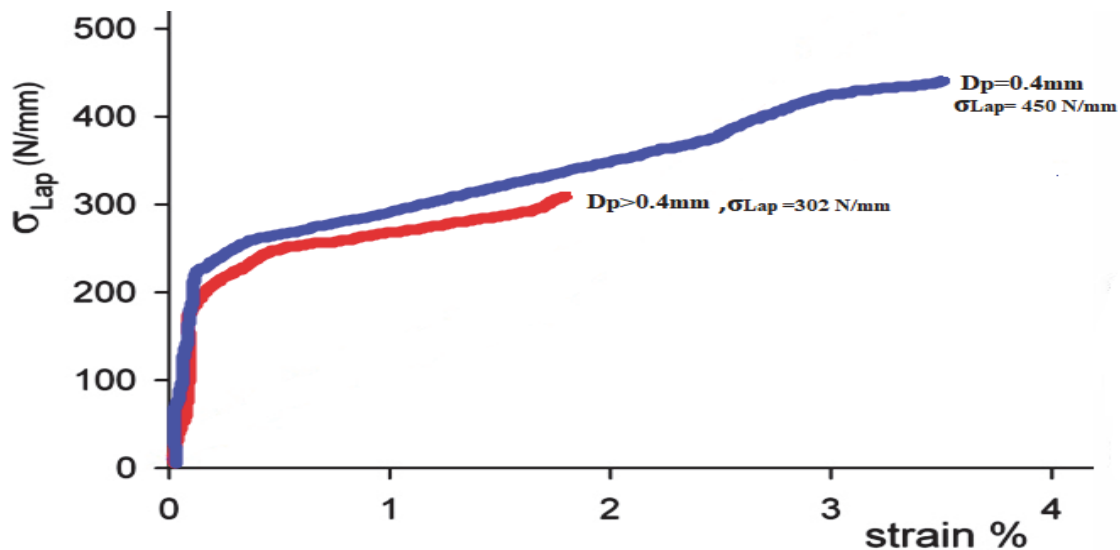

Fig. 11. Tensile-shear curves of two samples of welds made with ω = 1,400 rpm, v = 60 mm/min and *Dp* values as indicated

When Dp ≈0.4 and an interface intermetallic layer was established without MSZ, fracture strength (450 N/mm) is considerably higher than that for the sample with MSZ (302 N/mm, fractured at SZ). The large deformation is accompanied by a corresponding amount of work hardening, as is indicative in the tensile-shear curve in Fig. 11, and thus by a high σLap value of 475 N/mm. It is clear by viewing the tested samples in Fig. 12. For the pin penetrated sample Dp >0.4 mm (Fig12a), the sample having been slightly bent is evident. On the other hand, for the Dp≈0.4mm sample (Fig 12b), a large amount of local deformation and bending before the final fracture at SZ is clearly the feature. The presence of thin and continuous intermetallic layer in the Dp≈0.4mm sample means a different fracture behavior. The large amounts of deformation and fracture energy for this sample means that the thin interface layer is not brittle under tensile-shear condition. From the present results, it can be suggested that careful positioning control for Al-Cu FSL welds is a mean for the optimal weld strength to be obtained.

Fig.12. Tensile-shear tested samples of welds made with ω = 1,400 rpm, *v* =60   mm/min and (a) *Dp>* 0.4 mm and (b) *Dp≈*0.4 mm

4. Conclusion

For Al-Cu FSL welds, fracture strength was found to be very sensitive to pin positioning during FSLW. When Dp ≈0.4 mm, a single continuous intermetallic layer formed along the interface; and thus, continuous metallurgical bonding established. The joint produced by this pin penetrating condition (with a single continuous intermetallic at interface) displayed a higher σLap value (450 N/mm) which was ~ 49% increase in σLap in comparison to the case of Dp>0.4mm condition (with the mixed interface region).The condition of Dp >0.4 mm did not insure a continuous intermetallic layer and instead discontinuous intermetallic outbursts (non-continuous metallurgical bond) formed at the interface, resulting from the early stage of intermetallic growth, formed at the joint interface. Once the pin sufficiently penetrates to copper (Dp > 0.4 mm), the interface region was consisted of intermix and irregular layers of intermetallic and recrystallized Cu grains form. The value of σlap (≈302 N/mm) in pin penetrating condition, although Dp differ for different samples, indicates that σLap is not very sensitive to the size of the interface region (with intermix and irregular intermetallic layers), once it forms. Also, value of σlap (≈302 N/mm, fractured at SZ) for the pin penetrated welds is itself higher than the maximum σlap value (268 N/mm, fractured at the SZ or interface) reported in literature when pin penetrating condition used. In tensile shear loading of Dp>0.4mm samples (with the mixed interface region) failure occurred through crack propagation along the intermetallic layers embedded inside the penetrated region, resulting in a brittle fracture. However, for the sample made using Dp ≈0.4 mm condition (with a single continuous intermetallic layer at interface), a ductile fracture was dominant with plastic deformation preceding failure mostly in aluminum adjacent to and on top of the interfacial intermetallic layer. Therefore, these results indicate that the notion of the presence of intermetallic layers embrittling the joint, is scientifically incorrect and a joint interface with single intermetallic layer can be highly shear fracture resistant.

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