



Microstructure and mechanical properties of dissimilar Al–Cu joints by friction stir welding

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Abstract: Dissimilar friction stir welding between 1060 aluminum alloy and annealed pure copper sheet with a thickness of 3 mm was investigated. Sound weld was obtained at a rotational speed of 1050 r/min and a welding speed of 30 mm/min. Intermetallic structure formed at the crown and Cu/weld nugget (WN) area promotes interfacial diffusion and metallurgical bonding of aluminum and copper. However, corrosion morphology reveals the weak bonding mechanism of internal interface, which causes the joint failing across the interface with a brittle–ductile mixed fracture mode. The tensile strength of the joint is 148 MPa, which is higher than that of the aluminum matrix. Crystal defects and grain refinement by severely plastic deformation during friction stir welding facilitate short circuit diffusion and thus accelerate the formation of Al_4Cu_9 and Al_2Cu intermetallic compounds (IMCs). XRD results show that Al_4Cu_9 is mainly in Cu/WN transition zone. The high dislocation density and formation of dislocation loops are the major reasons of hardness increase in the WN.

Key words: aluminum alloy; copper; friction stir welding; dissimilar material; microstructure; mechanical properties

1 Introduction

Copper and aluminum have been widely applied as engineering structure materials due to their good comprehensive properties such as excellent corrosion resistance, ductility, heat and electric conductivity [1,2]. Recently, joining dissimilar materials such as aluminum and copper is of great interest in engineering applications because of their technical and beneficial advantages. Composite structure composed of aluminum and copper will contribute to mass and cost reduction by decreasing the amount of precious metals usage. However, due to great difference in their physical and chemical properties, the dissimilar combination of copper and aluminum is generally more difficult. Various welding methods, including fusion welding, braze welding and pressure welding, have been applied to joining Al–Cu dissimilar materials but many problems occurred such as oxidation, cavities and cracks [3,4].

Friction stir welding (FSW), a solid-state welding technique, is an innovational candidate for joining dissimilar materials with very different physical and

mechanical properties, such as Al–Mg, Al–Ti, Al–steel, Al–Cu [5–12]. For Al–Cu dissimilar FSW, AKBARI et al [13] and LEE et al [14] reported that sound defect-free dissimilar Al–Cu joint could be produced by inserting the stir pin with distance away from the faying surface towards the aluminum side, which is consistent with most scholars' experimental results. However, whether the pure copper sheet should be placed on the advancing or retreating side in order to obtain sound weld surface is controversial. Additionally, XUE et al [15] and SAEID et al [16] suggested that a thin, uniform and continuous layer of IMCs was necessary to achieve high quality Al–Cu joints. They suggested that if it is favorable or detrimental strongly depended on its amount, form and distribution. So far, though some achievements have been made, the bonding mechanism of Al–Cu dissimilar joints is not fully understood up to now and a deep understanding of the relationship between the microstructure and mechanical properties is still lacking.

In this work, dissimilar FSW of 1060 aluminum and pure copper sheets was carried out, and the sound dissimilar Al–Cu joint was achieved under suitable welding parameters. The relationship between

microstructure and mechanical properties was investigated in detail.

2 Experimental

Commercial pure copper (99.9%) and 1060 aluminum alloy plates with a thickness of 3 mm were jointed by FSW. The plates were cut into pieces with dimensions of 300 mm × 100 mm. The pure copper was annealed at about 650 °C, holding for 1 h, and then cooled in the air. The surfaces of the sheets were ground with grit paper to remove the oxide film and then cleaned with acetone. Before welding, the plates were placed on a backing plate and clamped rigidly by an anvil along the welding direction to prevent relative movement. Dissimilar FSW was carried out on an FSW-LM-5025 welding machine at a rotational rate of 1050 r/min and a welding speed of 30 mm/min which were the optimized welding parameters in our previous study. A cylindrical tool made of H13 steel with a shoulder of 12 mm in diameter and a pin of 4.5 mm in diameter and 2.8 mm in length was applied. The tilt angle was 2.5° from the normal surface of the plates. Unlike conventional FSW, the stir pin was mostly inserted on the aluminum side in this study and the configuration of base metals is shown in Fig. 1. During the welding process, the plunge depth of shoulder was controlled manually in order to modify the quality of weld formation.

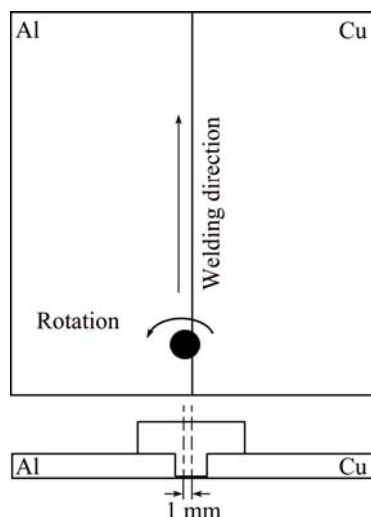


Fig. 1 Schematic illustration of Al–Cu dissimilar FSW

The metallographic specimens were machined perpendicular to the direction of welding. A solution of 5 mL H₂O₂ + 45 mL NH₄OH was used for obtaining the weld microstructure. Microstructural characterization was carried out by optical microscopy, and scanning electron microscopy (SEM) complemented by energy dispersive spectroscopy (EDS). Then, the test specimen was soaked in a solution of 3.5% NaCl for 24 h to reveal

the interfacial corrosion morphology by immersion test and SEM. The phase component in the WN was identified using X-ray diffraction (XRD) and transmission electron microscopy (TEM). Vickers hardness measurements were performed on the cross section perpendicular to the welding direction with a load of 50 g for 15 s using an FM-700 Vickers microhardness tester. The tensile specimens with a gauge length of 117 mm and a width of 15 mm were machined perpendicular to the direction of welding, according to GB/T 2652–2008 standard (equivalent to ISO5718:2001). The tensile test was carried out using a testing machine (CONROLLER).

3 Results and discussion

3.1 Macrostructure of Al–Cu joint

The surface morphology of Al–Cu FSW joint is shown in Fig. 2. No obvious groove and hole-type surface defects can be found, indicating that excellent weld surface appearance can be achieved when pure copper is placed on advancing side with a pin-off technique. However, poor surface morphology is obtained when pure copper is placed on the opposite side in this study, which is attributed to the distinct difference in their melting points and thermal conductivities. Aluminum possesses better plastic flowability and fillability at the same processing temperature, which ensures that the material on retreating side can be transferred to the advancing side continuously [17]. Furthermore, it is interesting to note that the tool shaking and stir pin adhesion problems can be modified by the superposition effect of annealing softening of pure copper before FSW combined with stir pin-off technique, which can reduce the resistance that impedes the tool forward, and thereby modify the weld formation.

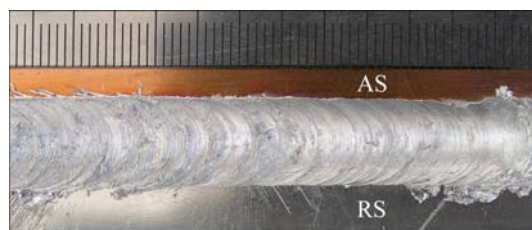


Fig. 2 Surface morphology of Al–Cu FSW joint

Figure 3 shows the thorough macrostructure of the Al–Cu FSW joint. It is noteworthy that the horizontal morphology (Fig. 3(a)), rarely studied in previous study, can reflect the material's flow and evolution better. The nugget zone is composed of a mixed structure of these two materials, i.e., so called intercalation swirls [18] and vortex-like patterns [19]. An obvious boundary can be found at the Cu/weld nugget (WN) interface. Since the tool

is mostly inserted in the soft material, the bimetal can flow smoothly with the tool and the aluminum content dominates the large fraction in the WN, whereas Cu fragments with various sizes and swirled laminates that are scraped from the Cu matrix under the agitation action of the tool are heterogeneously distributed in the nugget zone (Fig. 3(b)).

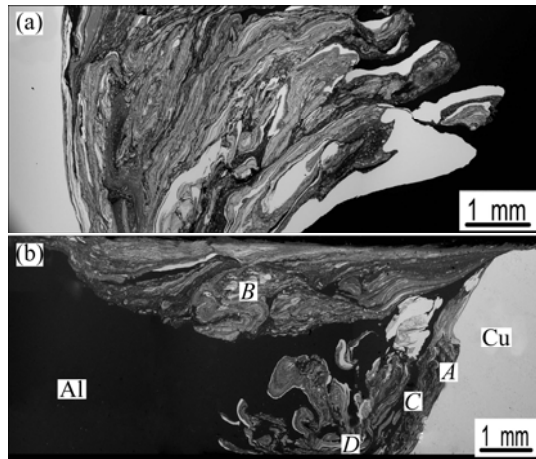


Fig. 3 BEI macrographs showing structure of Al–Cu FSW joint: (a) Horizontal morphology; (b) Cross-sectional morphology

3.2 Microstructure of Al–Cu joint

It is noteworthy that the quality of Al–Cu dissimilar joints is judged from not only the macrostructure but also the interior quality of the weld. Figure 4 shows the detailed microstructure of different regions marked in

Fig. 3(b). Different from FSW of homogenous materials, intercalation is the typical structure distributed widely in the WN which involves lamellar alternating patterns [3]. The sharp boundary appearing on the copper side which is shown in Fig. 4(a) may be the main cause of the low mechanical properties on the advancing side (Details on it are discussed on latter sections). Moreover, similar lamellar alternating structures of stacked aluminum and copper are distributed widely at the crown and Cu/WN interface, which can be observed in Figs. 4(b) and (c). The width of each lamella is not more than several micrometers and such a characteristic of blended structure demonstrates that apparent mechanical mixing and a certain level of metallurgical bonding occur between these two materials. Furthermore, as shown in Fig. 4(d), the elongated copper strips distributed at the bottom of the WN indicate that severe plastic deformation has taken place during FSW.

Further, Fig. 5 shows the element distribution taken from the weld nugget. Distribution of aluminum (red) and copper (green) can be seen in the EDS map. The results show that these two materials represent a good mixing, although some large Cu fragments can still be seen. Also, the slight variation of color contrast marked by the arrows in Figs. 4(b) and (c) illustrates that interfacial diffusion and reaction occur during Al–Cu dissimilar FSW, which may also give rise to the corresponding IMCs in these places.

TEM analysis was also performed to reveal the formation mechanism in the WN. Figure 6 shows TEM

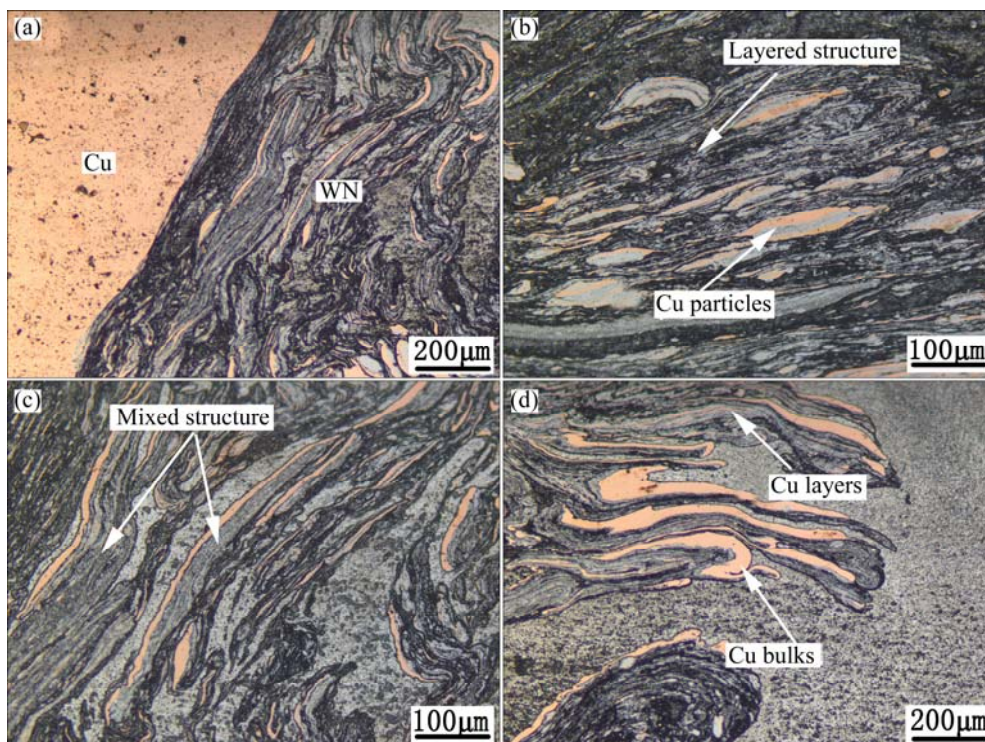


Fig. 4 Microstructures of WN: (a–d) Magnified microstructures of different regions marked A–D in Fig. 3, respectively

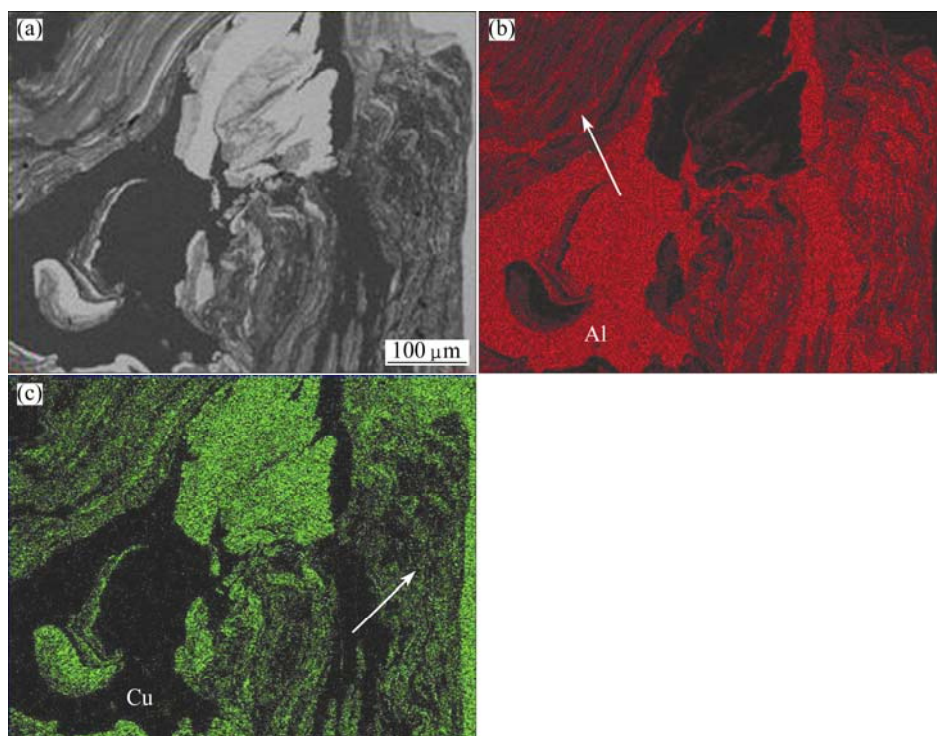


Fig. 5 EDS maps of Al-Cu FSW joint in WN: (a) Element distribution of WN; (b) Element distribution of Al; (c) Element distribution of Cu

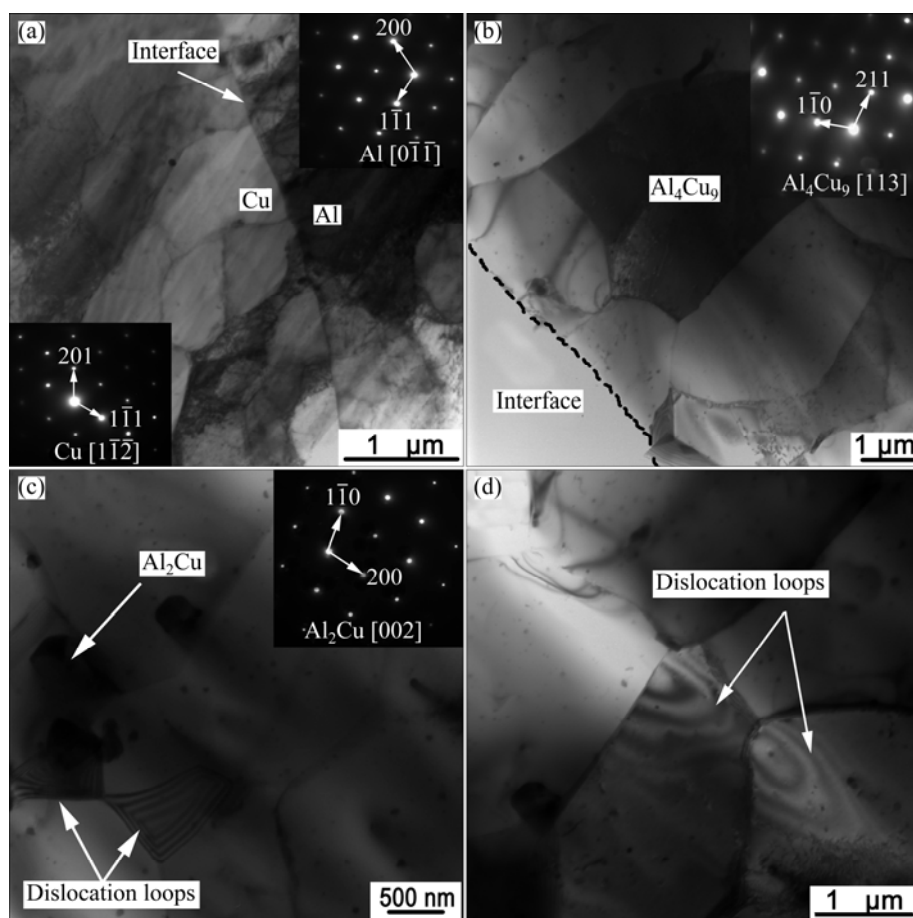


Fig. 6 TEM micrographs showing morphologies of nugget zone (a), Al_4Cu_9 particles (b), Al_2Cu particles (c) and dislocation loops (d) in WN

bright field micrographs and the corresponding SAED patterns taken in the WN. As shown in Fig. 6(a), it can be seen that the grains of both materials are refined because of dynamic recrystallization. The formation of straight grain boundary, deriving from the atomic diffusion under the effect of thermodynamical driving forces, further illustrates that metallurgical bonding between copper and aluminum has taken place. According to the SAED patterns in Figs. 6(b) and (c), the Al_4Cu_9 and Al_2Cu phases which are two common IMCs in previous studies are found, although the peak temperature in FSW is quite low and the holding time of weld at higher temperature is short. Preliminary analysis shows that the involved intense plastic deformation during FSW may give rise to a lot of crystal defects such as vacancy and dislocations, which would attribute to short circuit diffusion and thereby increase the possibility of IMCs formation. Beyond that, the grain refinement in the WN can simultaneously increase grain boundary and shorten the diffusion distance, which may facilitate the short circuit diffusion to some degree. Furthermore, the extreme plastic deformation in FSW is combined with the dislocation multiplication, which accelerates the formation of dislocation loops (Figs. 6(c) and (d)). The nugget zone is thus strengthened and hardness in the nugget is also enhanced.

3.3 Interfacial formation mechanism

The interface is a key to dissimilar joints, because it is often the position where the dissimilar joints fail. It is well documented that the bonding theory includes mainly three-stage processes, i.e., physical contact, activation of surfaces in connection and interaction with each other [20]. The backscattered electron image of the Cu/WN interface is shown in Fig. 7(a). The interface is extremely obvious and plenty of streamline structure concentrated around the interface. From the EDS analysis, the light and dark areas are mainly composed of copper and aluminum, respectively. Anyway, the intense agitation action facilitates atomic diffusion and the formation of the corresponding IMCs. Figure 7(b) shows the magnified view marked in Fig. 7(a). A thin, uniform and continuous IMCs layer, with no cracks that can be seen, has a positive effect on metallurgical bonding between aluminum and copper. Also, this can increase the bonding strength of the interface to some degree.

Correspondingly, the corrosion resistance of the interface, rarely mentioned in previous study, is another crucial factor for a comprehensive understanding of the interface bonding state in dissimilar FSW. As shown in Fig. 7(c), it is obvious that serious corrosion occurs at the interface. Preliminary analysis suggests that the inner of interface is not continuity or discontinuity, leading to a loosely-bonded interface mechanism. Hence, although

copper and aluminum exhibit excellent corrosion resistance, the corrosion is relatively active at the Cu/WN interface. Also, this goes some way to explain the reason why the crack tends to initiate and propagate at the bonding interface in the tensile test.

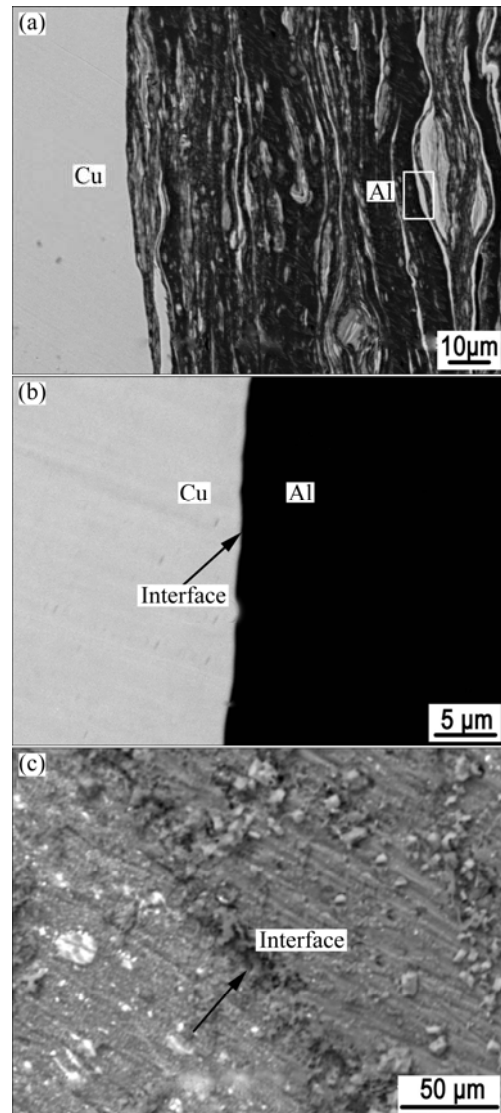


Fig. 7 BEI images showing Cu/WN interfacial region: (a) Microstructure of interface; (b) Magnified view of region marked in Fig. 7(a); (c) Corrosion morphology of interface

3.4 Mechanical properties

3.4.1 Tensile strength and fracture behavior

The fracture position of the Al–Cu FSW joint in this study is located in the transition zone of the advancing side (Cu side), where is exactly the loosely-bonded interface revealed by the aforementioned corrosion morphology. Table 1 shows the tensile properties of the Al–Cu dissimilar joint. The ultimate tensile strength (UTS) of the joint is higher than that of aluminum matrix but far less than that of the copper bulk, which is mainly caused by the inhomogeneous and chaotic structure formed in the WN. Furthermore, the extremely low

elongation rate is principally concerned with the formation of hard and brittle IMCs in Al–Cu FSW joints.

Table 1 Tensile properties of Al–Cu FSW joints

Material	UTS/MPa	Elongation/%
1060 Al alloy	130	18
Copper (T2)	255	22
Al–Cu joint	148	4

Figure 8(a) shows the SEM image of tensile fracture surface at low magnification. The appearance of tensile fracture varies appreciably with the locations across the WN due to their difference in microstructures. Figures 8(b)–(e) show the magnified views of the location marked in Fig. 8(a). As shown in Figs. 8(b) and (c), quantities of cleavage planes and small dimples can be seen clearly in these regions, which illustrates that a brittle-ductile mixed fracture has taken place. While large numbers of dimples with various sizes and depths are observed in Figs. 8(d) and (e). The heterogeneity of these dimples is mainly caused by the asymmetrical mixing of these two materials, which is also a significant reason for the poor mechanical properties.

3.4.2 Microhardness and IMCs distribution

Figure 9 shows the representative transverse cross sectional Vickers hardness along the top, the middle and the bottom of the welded sheets, respectively. It can be seen that the hardness profiles in the weld nugget are obviously higher than the base metals. As we mentioned earlier, the high dislocation density and grain refinement may be the prime reasons for microhardness increase in the WN. Furthermore, the existence of corresponding IMCs leads to a high fluctuation in hardness values. The average hardness at the top of the nugget is generally higher than that in the middle and bottom regions. However, XUE et al [15] found that the hardness in the bottom region was higher than that in the upper region. The fact that the extrusion and friction force of the shoulder are higher in the upper layer in the WN, which is conducive to the material mixing and mutual diffusion, is the main reason for the increase of corresponding microhardness. In this study, it is noteworthy that the hardness profiles on advancing side (Cu/WN side) are higher than those on retreating side in the WN. This may be attributed to lots of intercalation layers concentrating at the Cu/WN interface (Fig. 4(a) and Fig. 7(a)). That is

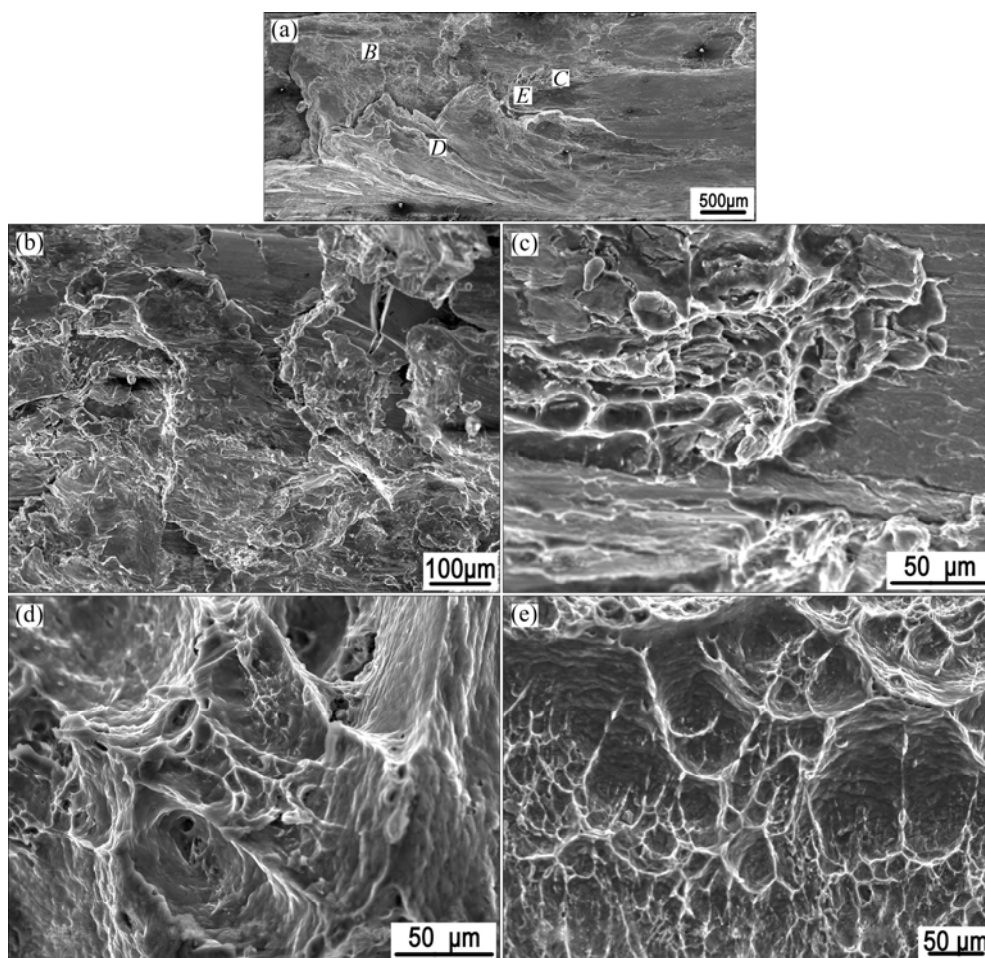


Fig. 8 SEM images showing tensile fracture surface of joint: (a) Macrograph of tensile fracture surface; (b–e) Magnified views of regions B–E in Fig. 8(a), respectively

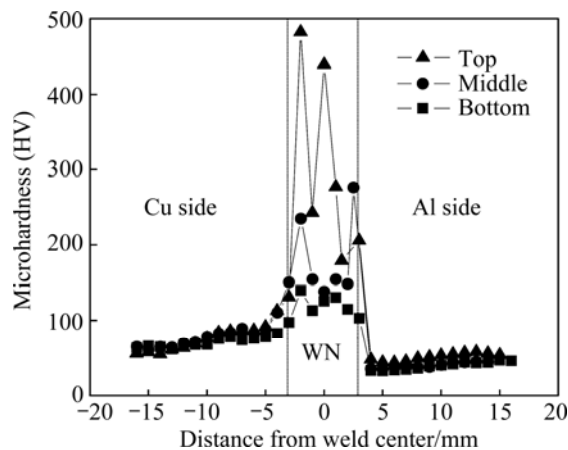


Fig. 9 Transverse cross-sectional hardness of Al-Cu FSW joint

to say, the IMCs may be mainly located in this region, resulting in the increase of hardness profiles.

Unlike the IMCs formed in traditional fusion welding, it is acknowledged as a more complicated and diffusion related process in FSW [18]. A good metallurgical bonding is often expected when small quantities of IMCs are formed at the interface [15]. In this study, the WN was divided into three cross-sectional locations: WN centerline (spectrum 1), Cu/WN side and WN/Al side both 3 mm from the weld centerline (spectra 2 and 3). The results of the XRD analysis performed in different regions of the nugget are shown in Fig. 10. Only a weak diffraction peak of Al_2Cu phase is detected at weld centerline, whereas diffraction peaks of Al_4Cu_9 and Al_2Cu phases with similar intensities are clearly observed at Cu/WN interface, which is in accordance with the TEM analysis. However, no IMCs can be found near the WN/Al interface. Thus, the emergence of the corresponding IMCs and their location and quantity explain why the hardness at the Cu/WN interface is higher than that in other regions of the nugget.

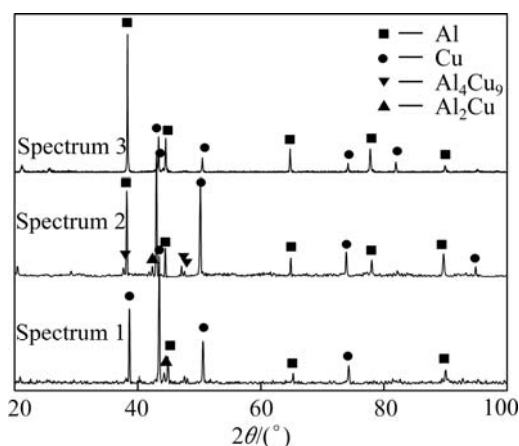


Fig. 10 X-ray diffraction patterns of dissimilar Al-Cu FSW joint

4 Conclusions

1) Pure copper and 1060 aluminum alloy are jointed successfully by FSW at a rotation rate of 1050 r/min and a welding speed of 30 mm/min with a configuration where copper is located on advancing side, and most of the tool pin is inserted on the aluminum side.

2) Rising dislocations and grain boundaries in the WN facilitate short circuit diffusion, and thus accelerate the formation of Al_4Cu_9 and Al_2Cu phases. Moreover, Al_4Cu_9 phase is mainly distributed in the Cu/WN area, which is consistent with the intercalation patterns formed within this region.

3) Intercalation structure formed in the crown and Cu/WN areas promotes metallurgical bonding of aluminum and copper. However, the weak connection of internal interface should be responsible for the crack initiation and propagation in tensile test. The ultimate tensile strength of the joints is 148 MPa, failing across Cu/WN interface with a brittle-ductile mixed fracture mode.

4) The average hardness values in the WN are higher than those of the base metals due to a high dislocation density during the involved extreme plastic deformation in FSW. Furthermore, the fluctuation of hardness values in the WN is wild because of the formation of corresponding IMCs.

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铝–铜异种金属搅拌摩擦焊接头组织及力学性能

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摘 要: 对 3 mm 厚的 1060 工业纯铝和退火纯铜异种金属搅拌摩擦焊进行研究。在搅拌头旋转速度为 1050 r/min、焊接速度为 30 mm/min 时, 获得性能良好的铝/铜接头。交替片层结构主要分布于焊核区顶部与铜/焊核区交界附近, 促进了铝、铜两种材料的界面扩散及铝和铜之间的冶金键合。然而, 界面腐蚀形貌揭示界面内部的弱连接机制, 导致拉伸试验中裂纹沿界面区域萌生和扩展, 断裂方式为韧–脆混合型断裂。接头的抗拉强度为 148 MPa, 高于工业纯铝母材的抗拉强度。搅拌摩擦焊强烈塑性变形引起的晶体缺陷和晶粒细化加速原子间的短程扩散, 从而促进金属间化合物 Al_4Cu_9 和 Al_2Cu 的生成。XRD 结果显示, 金属间化合物 Al_4Cu_9 主要位于铜/焊核区过渡区域。焊核区较高的位错密度和位错环的形成是导致该区域硬度明显升高的主要原因。

关键词: 铝合金; 铜; 搅拌摩擦焊; 异种材料; 显微组织; 力学性能

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