Original Article

Influence of dwell time on microstructure evolution and mechanical properties of dissimilar friction stir spot welded aluminum–copper metals

Gaohui Li\textsuperscript{a}, Li Zhou\textsuperscript{a,b,}\textsuperscript{∗}, Weilu Zhou\textsuperscript{b}, Xiaoguo Song\textsuperscript{b}, Yongxian Huang\textsuperscript{a,}\textsuperscript{∗}

\textsuperscript{a} State Key Laboratory of Advanced Welding and Joining, Harbin Institute of Technology, Harbin 150001, China
\textsuperscript{b} Shandong Provincial Key Laboratory of Special Welding Technology, Harbin Institute of Technology at Weihai, Weihai 264209, China

\textbf{ARTICLE INFO}

Article history:
Received 22 June 2018
Accepted 27 February 2019
Available online 24 May 2019

Keywords:
Friction stir spot welding
Aluminum/copper dissimilar metals
Microstructure evolution
Mechanical properties
Fracture path.

\textbf{ABSTRACT}

1060 aluminum–T2 copper dissimilar lap joints were produced by friction stir spot welding (FSSW) with various dwell time. All the joints possess a Cu hook extruded upward from the lower Cu plate into the upper Al plate with intermetallic compounds (IMCs) developed on its interface. Increasing the dwell time produced an increase in the heat input during welding and promoted IMC growth. At short dwell time, the interface was characterized by the interruptedly distributed CuAl\textsubscript{2} layer partially mingled with CuAl phase. However, continuous CuAl\textsubscript{2}–CuAl–Al\textsubscript{2}Cu\textsubscript{6} laminated layer developed at the interface at longer dwell time. IMC formation sequence for CuAl\textsubscript{2}, CuAl and Al\textsubscript{2}Cu\textsubscript{6} were determined by thermodynamic principles. Microhardness was quite different in different zones. Hardness values in the stir zone (SZ) were much higher due to the dispersively distributed IMC particles and the refined grains. Joints with better tensile properties have a higher penetration depth of the hook into the upper Al plate as well as a continuous IMC layer at the hook interface. Results of fracture path analysis indicated that all the fracture initiated at the CuAl\textsubscript{2}–CuAl or CuAl\textsubscript{2}–Al interface and then extended along the IMC layer at the hook interface. The dispersed IMC particles in the SZ provided an alternative path for crack extension when strong metallurgical bonding was achieved at the hook interface.

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1. Introduction

Cu is a common commercial metal that is widely used in the electric power industry due to its excellent conductivity and thermal performance. However, the Cu resources are relatively scarce and its price is expensive. On the other hand, design and manufacturing of electric products with Al alloys are longstanding due to the good conductivity, economic price and low density of Al alloys [1]. In order to economically use and conserve the global Cu resources, Al–Cu composite structure that substituted a part of Cu for Al (like Al–Cu transition pieces) has been widely applied to transmit electricity. Since the mechanical bolted joints between

\textsuperscript{∗} Corresponding authors.
E-mails: Zhouli@hitwh.edu.cn (L. Zhou), huangy@hit.edu.cn (Y. Huang).
https://doi.org/10.1016/j.jmrt.2019.02.015
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Al and Cu can hardly achieve stabilization in electrical conductivity, the joining of Al and Cu by welding has been attracted much attention in the past decade. It is difficult to produce sound Al–Cu dissimilar joints by fusion welding due to their significant difference in chemical, physical and mechanical properties as well as high affinity to form brittle IMCs at high temperatures [2]. Thus, some solid-state joining methods like forge welding, explosive welding and friction welding have been employed to the assembling of Al and Cu. However, some disadvantages such as the lack of versatility and safety of these welding methods have limited their application [3,4].

In recent years, the attempt to joining dissimilar metals by friction stir welding (FSW) has been made on combinations such as Al–steel, Al–Cu and Al–Mg. FSW shows considerable potential in joining of Al–Cu as the low heat input during welding enables its great advantage in IMC control [5,6]. The preliminary experiment in this field was conducted by Murr et al. [7], who reported that it is hard to obtain sound Al–Cu dissimilar joint by FSW because the joint failed easily along the interface. Afterwards, Ouyang et al. [8] studied the interfacial microstructure of FSWed Al–Cu joint and ascribed the poor tensile property to the excessive formation of IMCs at the interface. Xue et al. [9–11] optimized the welding parameters and revealed that a thin and uniform IMC layer at the interface is necessary to obtain sound joint. The relationship between material flow and distribution of IMCs was reported by Galvão et al. [12–14]. As a derivative of the FSW process, friction stir spot welding (FSSW) was later proposed in Japan as an alternative to conventional resistance spot welding (RSW) [15]. Until now, FSSW is mainly applied to the homogeneous welding of Al to Al or Mg to Mg. As for FSSW of Al to Cu, the literatures are relatively limited. Al–Cu dissimilar FSSWed joints were initially produced by Heideman et al. [16], but the tensile properties of the joints were poor with the tensile shear failure loads generally lower than 2 kN. Shiraly et al. [17] reported that joints with higher tensile properties own a Cu ring extruded upward by the pin tool from the lower Cu plate into the upper Al plate that promoted mechanical interlocking between the plates. However, the effect of the Cu ring’s configuration on tensile properties of the joints were not reported. Mubiayi et al. [18] revealed the presence of some IMCs at the Al–Cu interface. The presence of various IMCs was thought to contribute to the high hardness values around the interface. In [19,20] it was reported that tensile shear failure load increased with increasing shoulder plunge depth with all the joints exhibited a nugget pull-out failure mode. Besides, optimization of process parameters was also conducted by Siddharth et al. [21,22] to obtain the FSSWed Al–Cu joint with reduced interface hardness and increased tensile shear failure load using response surface methodology. The optimized operating windows can act as reference maps for future welding engineers in selecting proper parameters to obtain sound joints.

These aforementioned literatures [7–22] have yield some basic information on Al–Cu FSSW, however, the interfacial microstructure evolution is not comprehensively analyzed as well as the relationship between tensile property and fracture mode is not established. In the present study, Al–Cu dissimilar joints were produced by FSSW at various dwell time, the purpose of this study is to elucidate the correlation between microstructure and mechanical properties in dissimilar FSSWed joints.

2. Experimental procedure

Commercial 1060 Al and T2 Cu plates with dimensions of 100 × 40 × 2 mm were selected as base metals. The plates were overlapped by 40 mm with the Al plates on the top and FSSW was conducted in the center of the overlapped area. The welding tool consisted of a concave shoulder with a diameter of 14 mm and a right-hand thread pin with a major diameter of 4.6 mm and a length of 2.85 mm. The FSSW was performed with dwell time ranging from 1 to 9 s at a constant rotational speed of 2250 rpm and a shoulder plunge depth of 0.1 mm. The K type thermocouples were embedded at the faying surface of the Al–Cu plates, 4 mm (point A) and 8 mm (point B) away from the weld centerline respectively to keep track of welding temperature, as indicated in Fig. 1a.

The specimens for microstructure examination were prepared by the standard metallographic procedure and etched with Keller reagent. Macroscopic cross-sections of the joints were observed by a DSXS510 optical microscopy (OM) equipped with quantitative image analysis software. Microstructural observation was performed on a Zeiss-MERLIN Compact scanning electron microscope (SEM). For characterizing phases formed at the interface, energy dispersive spectroscopy (EDS) and X-ray diffraction (XRD) analysis were carried out. The hardness testing was operated on a MICRO-586 hardness tester with a load of 200 g for 10 s dwell time. Microhardness profiles were measured every 0.5 mm spacing along three different lines on the transverse cross section. These lines were 1, 2 and 3 mm away from the lower surface of the Cu plate, as shown in Fig. 1b. Tensile tests were carried out triplicate for each processing parameter at room temperature using an Instron-1186 mechanical tester with a crosshead displacement speed of 1 mm/min.

![Fig. 1 – Schematic illustration for: (a) location of thermocouples; (b) distribution of hardness test points.](image-url)
Table 1 – List of welding parameters with the measured variables during welding.

<table>
<thead>
<tr>
<th>Dwell time</th>
<th>Plunge depth (mm)</th>
<th>Rotational speed (rpm)</th>
<th>Torque (N,m)</th>
<th>Plunging force (kN)</th>
<th>A peak temperature (°C)</th>
<th>B peak temperature (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1 s</td>
<td>0.1</td>
<td>2250</td>
<td>19</td>
<td>6.25</td>
<td>501.4</td>
<td>355.5</td>
</tr>
<tr>
<td>5 s</td>
<td>0.1</td>
<td>2250</td>
<td>16</td>
<td>5.99</td>
<td>574.2</td>
<td>411.9</td>
</tr>
<tr>
<td>9 s</td>
<td>0.1</td>
<td>2250</td>
<td>9</td>
<td>5.57</td>
<td>587.3</td>
<td>422.1</td>
</tr>
</tbody>
</table>

Fig. 2 – Thermal cycles produced by different dwell time at different position (a) point A; (b) point B.

3. Results and discussion

3.1. Thermal history and weld formation

It is important to investigate the thermal history during welding because the elevated temperature is the precondi-
tion for material softening, heterogeneous mixing as well as nucleation and growth of IMCs, since their evolution is thermally activated [23,24]. All the measured variables during welding are listed in Table 1 with thermal cycles of points A and B at various dwell time shown in Fig. 2. It can be found that an increasing dwell time leads to a higher peak temperature but a lower plunging force and torque due to the higher softening degree of the metals and the resultant lower material flow stress. When the dwell time was 1s, peak temperature in points A and B was 501.4 °C and 355.5 °C, respectively. As dwell time of 9s was used, peak temperature in points A and B increased to 587.3 °C and 422.1 °C, respectively. All the curves in Fig. 2 have a similar variation pattern, i.e. the ascent stage are all characterized by two steep rises connected by a flattening rise. When the subface of stir pin contacted the upper surface of the Al plate, drastic friction occurred between the two surfaces and thus a large amount of friction heat generated, which could be responsible for the first steep rise. However, when the temperature exceeded a threshold value, materials were softened and began to flow around the stir pin [15], leading to the decrease of friction resistance between the pin tool and the metals, thus, the rate of the temperature rise decreased. As the pin tool continuously plunged into the plates, the shoulder of the pin tool contacted the plates, which increased the frictional area significantly and brought about the second steep rise of the temperature.

Cross-section morphologies of the joints welded at various dwell time are shown in Fig. 3. At dwell time of 1s, metals were not fully softened and thus plastic flow only occurred in a small scope, resulting in a small stir zone (SZ). As dwell time increased, size of the SZ enlarged obviously (Fig. 3a–c). In the SZ, a Cu hook that was extruded upward from the lower Cu plate into the upper Al plate due to the extrusion and stirring exerted by the pin tool can be detected. The shape of the hook varied depending on the dwell time. For a brief discussion, the geometric parameters of the hook are indicated in Fig. 3d. Among them, hook height (HH) represents the distance from the tip of the hook to the faying surface of the Al-Cu plates. Fully bonded region (FBR) signals the horizontal distance between the tip of the hook and the margin of the exit hole. The ratio of hook height to FBR is defined as HH/FBR, joints with larger HH/FBR own a steeper Cu hook penetrated into the Al plate and thus a stronger mechanical interlocking between the Al–Cu plates. All geometric parameters of the hook at different dwell time were quantitatively analyzed with the results listed in Table 2. As dwell time increased from 1 to 5s, all the geometric parameters show an increase due to the enhanced heating and heterogeneous mixing. However, when the dwell time further increased to 9s, the HH/FBR dropped dramatically due to the curling of the hook (Fig. 3c).

3.2. Interfacial microstructure evolution

SEM images for the hook interface facing the exit hole at different dwell time are shown in Fig. 4 with the correspond-
ing EDS results listed in Table 3. The formations of the IMCs are closely related to the temperature, which is decided by the dwell time in the present study. At the shortest dwell time of 1s, the interface was characterized by the 1.9μm thick laminated IMC layer with three sub-layers (Fig. 4a and c), the EDS results indicated that the chemical compositions at points A, B and C in Fig. 4c are approximately equal to that of the CuAl2, CuAl and Al4Cu3. Meanwhile, obvious characteristic diffraction peaks of CuAl2, CuAl and Al4Cu3 were observed from the XRD result of the interface (see Fig. 4g),
which along with the EDS results indicated the existence of these Al–Cu IMCs. As dwell time increased to 9 s, the composition of the laminated IMC layer retained with the thickness of the IMC layer showed a 1.2 µm growth (Fig. 4b and d) due to the enhanced reaction between Al and Cu at higher temperature [8]. This result is consistent with the phenomenon reported by Xue et al. [9,10] in FSW of Al to Cu. Moreover, an increase in the thickness of the IMC layer at the Al–Cu interface was also observed under higher heat input. Notably, microcracks are formed at the boundary between CuAl2 sub-layer and Al matrix at dwell time of 9 s (Fig. 4b), which could be attributed to the dissimilarity in thermal expansion coefficients between CuAl2 and Al and the resultant incongruous deformation during cooling process [25]. Microstructure in the SZ in vicinity of the interface is presented in Fig. 4e and f with the streamlines of the material indicated by yellow arrows. Along the streamlines, plenty of IMC particles dispersively distributed with some coarse particles showing a multiphase layered structure (Fig. 4f). Most of the IMC particles became finer at higher dwell time due to the intensified mechanical stirring effect. Further, the higher softening degree of the metals allows Cu pieces with even large sizes to be detached from the Cu bulk and finally evolved to the multiphase layered structure in Fig. 4f through Al–Cu interaction [6,10].

With respect to the hook interface back to the exit hole, it can be seen from Fig. 5a and b and Table 4 that dwell time has strong effect on bonding conditions of the interface. At the shortest dwell time of 1 s, the interface was characterized by the interruptedly distributed CuAl2 layer partially mingled with CuAl phase. The average thickness of the IMC layer was estimated to be 0.8 µm. As dwell time reached 9 s, a continuous CuAl2–CuAl–AlCu9 laminated layer developed at the interface as well as the thickness of IMC layer increased to 1.8 µm. Fig. 5c shows the river pattern formed in the hook tip at dwell time of 5 s. The river pattern shows a multiphase structure that consisted of CuAl2, Cu3Al2 and AlCu9 according to the EDS results in Table 4. These IMCs are the resultants of the
Al–Cu reaction at different stages [23]. It can be deduced that some Al mixed into the Cu during the upward movement of the hook and then reacted with the Cu matrix at the interface, resulting in the formation of the river pattern. A higher magnification of the hook tip is presented in Fig. 5d, where a microcrack could be detected between CuAl2 layer and Al matrix.

3.3. Interface formation mechanism

In this section, IMC formation sequence and interface formation mechanism are discussed. Although [8,9,11–13] have described the microstructural characteristic at the Al–Cu interface in the FSWed joint, the IMC formation sequence was not investigated. The knowledge about this sequence allows
Table 3 – EDS result for the hook interface facing the exit hole.

<table>
<thead>
<tr>
<th>Location of the EDS analyses</th>
<th>Element content (at.%)</th>
<th>Composition</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Al</td>
<td>Cu</td>
</tr>
<tr>
<td>A</td>
<td>66.75</td>
<td>33.25</td>
</tr>
<tr>
<td>B</td>
<td>54.46</td>
<td>45.54</td>
</tr>
<tr>
<td>C</td>
<td>37.26</td>
<td>62.74</td>
</tr>
<tr>
<td>D</td>
<td>73.43</td>
<td>26.57</td>
</tr>
<tr>
<td>E</td>
<td>49.57</td>
<td>50.43</td>
</tr>
<tr>
<td>F</td>
<td>39.52</td>
<td>60.48</td>
</tr>
<tr>
<td>G</td>
<td>83.56</td>
<td>16.44</td>
</tr>
</tbody>
</table>

Although the effective concentration of the elements at the growth interface cannot be calculated directly, it is known that the smaller the activation energy is for solid-state interdiffusion, the greater the mobility of the interfacial atoms. Also, the activation energy is proportional to the melting point of the solid. Therefore, the effective concentration is chosen to be the composition at the lowest eutectic temperature of the binary system [27]. The lowest temperature of the eutectic of the Al–Cu binary system forms is at 548 °C, with the concentration of Cu at 17% [28]. Meanwhile, component concentration of Cu in CuAl₂, CuAl and Al₂Cu₃ are all larger than 17%. Therefore, Cu is determined as the limiting element. So far, ΔH′ of CuAl₂, CuAl and Al₂Cu₃ can be calculated by referring to the standard thermodynamic data in Ref. [29]. For example, ΔH′ (273 K) of CuAl₂ is 17%×(−13.44) = −6.86 kJ (mol at.)⁻¹. The calculated ΔH′ for other Al–Cu IMCs are listed in Table 5, by which, IMC formation sequence for CuAl₂, CuAl and Al₂Cu₃ is determined.

Schematic diagrams for interface formation are shown in Fig. 6. At the shortest dwell time of 1 s, the alloys plasticized little due to the insufficient heat input. Therefore, the lower Cu plate was slightly extruded upward by the pin tool into the upper plate. Meanwhile, only few fine Cu pieces were detached from the Cu bulk and flowed counterclockwise on the vertical plane by the effect of the threaded pin, which finally evolved to the Al-Cu IMCs through the interaction with the
Table 4 – EDS result for the hook interface back to the exit hole.

<table>
<thead>
<tr>
<th>Location of the EDS analyses</th>
<th>Element content (at.%)</th>
<th>Composition</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>68.65</td>
<td>CuAl</td>
</tr>
<tr>
<td>B</td>
<td>50.16</td>
<td>CuAl</td>
</tr>
<tr>
<td>C</td>
<td>68.64</td>
<td>CuAl</td>
</tr>
<tr>
<td>D</td>
<td>46.65</td>
<td>CuAl</td>
</tr>
<tr>
<td>E</td>
<td>23.28</td>
<td>Al₄Cu₉ + (Cu)</td>
</tr>
<tr>
<td>F</td>
<td>67.19</td>
<td>CuAl</td>
</tr>
<tr>
<td>G</td>
<td>67.54</td>
<td>CuAl₉</td>
</tr>
<tr>
<td>H</td>
<td>40.85</td>
<td>Cu₄Al₉</td>
</tr>
<tr>
<td>I</td>
<td>36.09</td>
<td>Al₄Cu₉</td>
</tr>
<tr>
<td>J</td>
<td>71.39</td>
<td>(Al) + CuAl₂</td>
</tr>
</tbody>
</table>

Al matrix. Owning to the insufficient heat input, the CuAl₂-CuAl-Al₄Cu₉ laminated layer formed discontinuously at the hook interface facing the exit hole. Moreover, it was more difficult to match the thermodynamic conditions for nucleation of Al₄Cu₉ at the hook interface back to the exit hole, resulting in formation of discontinuous CuAl₂ layer partially mingled with CuAl. In this case, the metallurgical bonding is weak, as shown in Fig. 6a. As dwell time increased to 5s, the copper plate was more extruded into the upper aluminum plate due to the higher softening degree of metals. During the upward movement of the hook, some Al was mixed into the hook tip and metallurgically combined with the Cu hook, as shown in Fig. 6b. Besides, it had matched the thermodynamic conditions for the formation of IMCs at the whole hook interface, leading to the formation of continuous CuAl₂–CuAl–Al₄Cu₉ laminated layer at the hook interface facing the exit hole and continuous CuAl₂–CuAl laminated layer at the hook interface back to the exit hole. When the heat input was excessive at dwell time of 9s, the metals were greatly softened, resulting in curling of the hook, which compromised the mechanical interlocking between the plates. Meanwhile, Cu pieces with larger sizes could be detached from the Cu bulk, which finally transformed into the coarse multiphase layered structures in the SZ. Owning to the high heat input, CuAl₂–CuAl–Al₄Cu₉ laminated layer was developed at the whole hook interface accompanied by microcracks formed at the boundary between CuAl₂ sub-layer and Al matrix, as shown in Fig. 6c.

3.4. Microhardness

Fig. 7 shows the hardness profiles of the joints at different dwell time. The hardness profiles were symmetrical about the centerline of the joint. Owning to the existence of the exit hole, hardness profiles of the upper and middle part of the joint were discontinuous. For the upper part, the hardness

Table 5 – Standard thermodynamic data of the Al–Cu IMCs with the calculated ΔH.

<table>
<thead>
<tr>
<th>Formation</th>
<th>Component concentration</th>
<th>ΔH° (298 K) (kJ (K mol))</th>
<th>ΔH° (773 K) (kJ (K mol))</th>
<th>ΔH (298 K) (kJ (K mol))</th>
<th>ΔH (773 K) (kJ (K mol))</th>
</tr>
</thead>
<tbody>
<tr>
<td>CuAl₂</td>
<td>Cu₆₃Al₉₆₇</td>
<td>−13.44</td>
<td>−13.05</td>
<td>−6.86</td>
<td>−6.66</td>
</tr>
<tr>
<td>CuAl</td>
<td>Cu₆₃₅Al₉₅</td>
<td>−20.12</td>
<td>−19.52</td>
<td>−6.84</td>
<td>−6.63</td>
</tr>
<tr>
<td>Al₄Cu₉</td>
<td>Al₃₃₇Cu₀₶₉₁</td>
<td>−23.1</td>
<td>−21.69</td>
<td>−5.67</td>
<td>−5.61</td>
</tr>
</tbody>
</table>

Fig. 6 – Schematic diagram for interface formation at various heat input (a) insufficient heat input; (b) proper heat input; (c) excessive heat input.
increased progressively before reaching a maximum within the SZ. Hardness is closely related to the grain size and distribution of IMC particles [34]. The larger hardness in the SZ can be attributed to the grain refinement induced by both the comprehensive effect of heat and deformation during FSSW and the Orowan strengthening due to the dispersion of the hard IMC particles. Hardness profiles of the middle part (measured along the faying face of Al–Cu plates) show a similar pattern. Notably, hardness values fluctuated drastically in the region far away from the exit hole, where the Al–Cu interface was weak-bonded due to insufficient heat input. Therefore, hardness varied with the bonding state at different locations. For the lower part, W-shaped hardness profiles with the highest points located in the center were observed. The center part of the lower plate was directly stirred by the tip of the stir pin and interacted with the upper plate through the plastic flow. Thus, continuous dynamic recrystallization occurred and grains were greatly refined in this region [30], which can be responsible for the higher hardness than that of the base metal. The lowest points in the hardness profiles are located in the region that is 4 to 4.5 mm away from the centerline, where materials were softened most by the welding heat. As dwell time increased, a slight decrease in hardness in most regions of the joint can be observed due to the increased heat input and the resultant higher softening degree and bigger softening area of the alloys.

3.5. Tensile behavior and fracture mode

Tensile behavior of the FSSW joint and its corresponding influencing factors are shown in Fig. 8. The tensile shear failure loads of the joints increased with increasing dwell time in the range of 1–5 s with the maximum value of 4.304 kN reached at 5 s. As dwell time further increased to 9 s, the tensile shear failure load decreased to 2.236 kN. The FSSW joint was combined both mechanically by the Cu hook and metallurgically by the IMC layer. A uniform interfacial IMC layer with a proper thickness indicates a good metallurgical bonding, while a

![Fig. 7 - Microhardness profiles along (a) top, (b) middle and (c) bottom lines on the cross-section.](image)

![Fig. 8 - (a) Tensile shear failure load with HH/FBR; (b) thickness of IMC layers.](image)
thicker IMC layer tends to lead to the formation of microcracks and thus compromise the tensile property of the joint \cite{9,10}. IMC layer formed discontinuously at the hook interface under dwell time of 1 s, results in a poor metallurgical bonding. As dwell time increased to 5 s, continuous IMC layer developed at the interface as well as larger HH/FBR was obtained. Therefore, the joint performed well during tensile shear test. At longer dwell time of 9 s, thicker IMC layer formed at the interface.
which increased the brittleness of the joint and led to the formation of microcracks along the IMC layer (Fig. 4b). Besides, the HH/FBR decreased due to the curling of the hook, which weakened the mechanical interlocking between the plates, further deteriorated the tensile property of joint. Therefore, the tensile shear failure load dropped to a low value.

The fracture paths of the joints welded at different dwell time are shown in Figs. 9–11 with the corresponding EDS results summarized in Table 6. Fracture path differed with the interface bonding conditions that were decided by the dwell time in this study. At shorter dwell time of 1 s, it can be seen from Fig. 9a that the fracture extended along the hook interface until the joint cracked because the interface was weak bonded through the interruptedly distributed IMCs (Fig. 5a). The crack seems to have initiated at the boundaries between CuAl2 and CuAl phases located at the hook interface back to the exit hole. This occurred since only CuAl and Cu solid solution (CuAl) were left on the fracture surface in the hook side (Fig. 9b). Followed by which is a crack extension across the CuAl–CuAl2 interface (Fig. 9c), leaving the saw-toothed gap between CuAl and CuAl2 in the fracture section, as indicated by the yellow dot lines in Fig. 9d. The fracture occurrence across the CuAl–CuAl2 interface can be explained by the low interfacial strength between the two phases due to their big different in lattice structures (CuAl2 has tetragonal lattice while AlCu has either monoclinic or orthorhombic lattice) and the resultant large misorientation between the two phases [13]. Xue et al. [11] also reported that the bonding between CuAl and CuAl2 phases are weak. Consequently, the boundaries between these two phases in the FSWed Al–Cu dissimilar joint tend to act as propagation path of crack during tensile test.

<table>
<thead>
<tr>
<th>Location of EDS analyses</th>
<th>Element content (at.%)</th>
<th>Composition</th>
</tr>
</thead>
<tbody>
<tr>
<td>CuAl</td>
<td>16.80 83.20</td>
<td>Cu + (Al)</td>
</tr>
<tr>
<td>CuAl</td>
<td>48.54 51.46</td>
<td>CuAl</td>
</tr>
<tr>
<td>CuAl</td>
<td>47.19 52.81</td>
<td>CuAl</td>
</tr>
<tr>
<td>CuAl</td>
<td>38.96 61.04</td>
<td>Al2Cu3</td>
</tr>
<tr>
<td>CuAl</td>
<td>48.18 51.82</td>
<td>CuAl</td>
</tr>
<tr>
<td>CuAl</td>
<td>55.70 44.30</td>
<td>CuAl + CuAl2</td>
</tr>
<tr>
<td>CuAl</td>
<td>47.52 52.48</td>
<td>CuAl</td>
</tr>
<tr>
<td>CuAl</td>
<td>77.58 22.42</td>
<td>(Al) + CuAl2</td>
</tr>
</tbody>
</table>

At longer dwell time of 5 s, the fracture also extended along the boundary of CuAl and CuAl2 at the hook interface back to the exit hole in the initial stage of fracture. Notably, when the fracture extended to the hook tip, the fracture path bifurcated
with a branch of the fracture turned to propagate along the river pattern in the hook tip and then ended in the hook, leaving CuAl, CuAl2 phases on the fracture surface (Fig. 10b and c). Since good metallurgical bonding was achieved at the hook interface facing the exit hole through the uniform and continuous IMC layer, the dispersed IMC particles in the SZ provided an alternative path for crack propagation. It can be seen from Fig. 10d that when the other branch of the fracture extended to the hook interface facing the exit hole, the fracture turned to propagate along the IMC particles in the SZ rather than the IMC layer at the hook interface. As dwell time reached 9 s, a large amount of CuAl2 particles dispersed around the hook interface due to the high heat input and enhanced plastic flow (Fig. 4d and f), and the crack was found initiated at the boundaries of the dispersed CuAl2 particles adjacent to the hook interface back to the exit hole (Fig. 11a). These CuAl2 particles can constrain the plastic deformation in the Al matrix, which undermines the deformation consistency of this part of material [25]. Accordingly, microcracks formed easily at the boundaries between the CuAl2 particles and Al matrix under the effect of tensile load (Fig. 11b) and propagated along these CuAl2 particles until reaching the IMC layer. Afterwards, the fracture extended along the IMC layer at the hook interface until the joint failed (Fig. 11c).

4. Conclusions

1 A longer dwell time can lead to a higher peak temperature but also to a lower plunging force and torque, which is an indication of higher softening degree of the metals as well as lower material flow stress during welding. All the joints possess a Cu hook extruded upward by the pin tool from the lower Cu plate into the upper Al plate with IMCs developed on its interface.

2 Bonding conditions at the hook interface varied with dwell time. At shorter dwell times, discontinuous CuAl2 layer mingled with CuAl formed at the interface due to the insufficient heat input. While at longer dwell times, the interface was characterized by the continuous CuAl2–CuAl–Al4Cu9 laminated layer accompanied by microcracks formed at the boundary between CuAl2 and Al matrix. IMC occurrence sequence for CuAl2, CuAl and Al4Cu9 can be predicted by the EHF model.

3 Microhardness was quite different in distinct zones of the FSSW joint. Microhardness in the SZ was higher due to the dispersion of IMC particles and grain refinement induced by the combination effect of welding heat and mechanical stirring. A slight decrease in hardness could be observed at longer dwell time.

4 FSSW joints were both metallurgically and mechanically bonded. Joints performed well during tensile test owing to a continuous IMC layer with a thickness no more than 2.3 μm and a large HHI/FBR.

5 All the fractures were found initiating at the CuAl2–CuAl or CuAl2–Al interface and tended to extend along the brittle IMC layer at the hook interface. The dispersed IMC particles in the SZ could provide an alternative path for crack extension when good metallurgical bonding was achieved at the hook interface.

Conflicts of interest

The authors declare no conflicts of interest.

Acknowledgements

The research was sponsored by the National Natural Science Foundation of China (No. 51205084) and Natural Science Foundation of Shandong Province (ZR2016EM043).

REFERENCES


