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ARTICLE

Achieving High-Quality Al/Cu T-Lap FSW Joints Using Pre-set Wires and an Eccentric Progressive Pin

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Abstract: In Al/Cu friction stir welding (FSW) T-lap joints, the hook-shaped structure present in the nugget zone (NZ) serves as a crack initiation site, which severely compromises the performance of the joints. In this study, FSW of 6061 aluminum alloy and pure copper in a T-lap configuration was performed employing an eccentric progressive pin. Two types of welding wires with different chemical composition were pre-set at the corner die, and the applicability of three tools for T-lap joints was examined. The research was intended to improve the bonding quality between the hook-shaped structure and the base material to obtain high-quality welded joints. Two optimal welding tools were selected. The influences of these tools on the grain morphology at the Al/Cu interface and the morphology of intermetallic compounds (IMCs) were analyzed, and the mechanical properties of the joints along the skin and stringer directions were evaluated. The results reveal that the tool with a small eccentric pin reduces the thickness of the interfacial IMC layer, simultaneously thinning the tip of the Cu hook and enlarging the volume of the NZ, and achieving metallurgical bonding between the periphery of the Cu hook and the matrix, thus improving the welding quality. For the M800Al welding tool equipped with a small eccentric pin and three cutting surfaces, its periodic shearing action facilitates the formation of a vortex-like interlocking interface. Meanwhile, more fragments scraped from the Cu hook are distributed throughout the NZ. This phenomenon not only refines the grain structure but also promotes discontinuous dynamic recrystallization and geometric dynamic recrystallization. The ultimate tensile strength of the M800Al joint along the stringer direction reaches 176 MPa, marking a 23.9% increment compared with that of the joint produced using a non-eccentric tool. This work provides a feasible solution for improving the performance of FSW T-lap joints of dissimilar metals.

Key words: Al/Cu dissimilar alloy; FSW; T-lap joints; pre-set welding wire; eccentric progressive pin; microstructure; properties

1 Introduction

Aluminum exhibits considerable potential for application in electrical and thermal management components, where it can substitute for copper to improve cost efficiency and reduce mass while maintaining requisite performance levels. As an environmentally benign solid-state joining technique, friction stir welding (FSW) offers inherent benefits for the welding of dissimilar light alloys^[1-5]. However, in the FSW of aluminum-copper systems, the prevalent formation of lamellar composite architectures, coupled with the occurrence of intermetallic compounds (IMCs), exerts a critical influence on the resultant mechanical strength and electrical conductivity of the joints^[6-8]. Strictly controlling the formation type, distribution,

size, quantity, and evolution of IMCs has become a research hotspot in Al/Cu FSW. The type and thickness of IMCs are mainly affected by factors such as Al/Cu chemical composition at the location, local stress and strain, frictional heat generation, plastic deformation heat generation, and cooling rate^[7]. Dispersed, fine IMCs can improve the joint strength, yet their inherent brittleness also makes them prone to crack initiation under stress, thereby significantly affecting the mechanical properties and fracture behavior of the joint. Many researchers have adopted methods such as underwater welding^[9], ultrasonic assistance^[10], or the addition of interlayers^[11] to regulate the FSW thermal cycle or optimize the type of IMCs.

The FSW of Al/Cu T-lap joints is confronted with multiple challenges. Apart from the considerable challenge in

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effectively regulating the IMC layer, defects including cavities and bonding lines frequently occur at the internal T-corner^[12-13], which further increases the difficulty of obtaining high-quality welded joints. Since the T-corner is far from the shoulder, where the maximum heat is generated during welding, and the adjacent corner die features a quick heat dissipation rate, numerous researchers have implemented process measures to avoid defect formation. These measures involve redesigning tool geometry^[14], adjusting the dimensions of the internal corner, and adopting techniques such as multipass welding^[15] or double-pass reverse welding^[16]. To achieve effective material filling at the corner, diverse strategies have been attempted, including pre-placing a skin layer with a specified thickness^[17], inserting a stringer of a designated height^[18], and pre-setting welding wires at the internal T-corner^[19-20]. Relevant studies have demonstrated that during the plastic forming process, temperature and flow velocity act as prerequisites governing the material flow and joint performance, while tool geometry serves as the key factor determining the material flow in local regions^[21]. The redesign of tools for FSW of Al/Cu T-lap joints still entails numerous challenges. Such a redesign must not only take the structural characteristics of T-lap joints into account but also accommodate the distinctive properties of dissimilar materials. For instance, pre-set welding wires undergo intense flow and mixing under the stirring action of the tool. The corner die not only acts as a hindrance to plastic flow but also provides a reaction force for plastic deformation^[20], which further complicates the distribution of welding wires and mixed materials within the nugget zone (NZ). The complex flow behavior between the tool and the local material at the internal corner creates favorable conditions for the formation of massive IMCs. In addition, copper with inferior plastic flowability migrates toward the skin layer, forming Cu hooks at the lap interface. The bonding quality at the periphery of these Cu hooks is inferior, and defects such as cold laps are generated between the Cu hooks and the aluminum base materials. These defects also exert adverse impacts on the mechanical properties of the T-lap joints.

In summary, redesigning tools suitable for Al/Cu FSW offers a feasible approach to fabricating high-quality joints. This redesign must enhance stirring at the internal corner while forming effective mechanical interlocking and a suitable thickness of IMCs. The eccentric structure can increase the material strain in the NZ^[22], particularly in the small pin stirring zone (SPSZ) adjacent to the corner die. This method exhibits great potential in controlling the thickness of interfacial IMCs and improving flow coordination by enhancing material flow. The addition of three cutting surfaces facilitates the scraping of more particles from the inner side of copper into the NZ, thereby eliminating the cold lap defects at the periphery of copper hooks. Therefore, on the basis of the progressive pin, the study of the effects of adding a small pin eccentricity or a combination of small pin eccentricity and three cutting surfaces on the microstructural evolution of the NZ and the mechanical properties of the

joints can provide a theoretical basis for high-quality FSW of dissimilar metal T-lap joints.

2 Experiment

A6061 aluminum alloy with a thickness of 4 mm was selected as the skin, and T2 pure copper of the same thickness was selected as the stringer. The materials to be welded were cut into plates of 200 mm×100 mm and 50 mm×100 mm. Three progressive pins were made of H13 tool steel, with an S-type tool featuring a shoulder diameter of 20 mm, a smooth inner surface with a concave angle of 5°, a large stirring pin of 8 mm, and right-handed threads. Based on the S-type tool shape, a P-type tool was designed with a small pin eccentricity of 0.4 mm. Based on the P-type tool shape, both large and small pins were equipped with three cutting surfaces to form the M-type tool. The geometries and dimensions of the three types of progressive pins are shown in Fig.1. Considering the similar chemical composition of the pre-set welding wires and the base materials, ER6061 and HS210 welding wires with a diameter of 1.6 mm were selected and made tangent to the fillet voids. Given that the proportion of copper in the NZ is relatively low in high-quality Al/Cu joints and the pin length is 2 mm, a boss with a height of 1 mm and a width of 4 mm was machined on the skin to assemble the 4-mm-thick stringer and the boss. This configuration altered the original distance between the workpiece and the tool shoulder, thereby reducing the mixing ratio of Al and Cu in the weld NZ. A gantry-type FSW machine (Model: S-RSW-1510-2D-10T-V1.0) was adopted for the welding experiments. Optimized welding parameters were as follows: rotational speed of 800 r/min, welding speed of 35 mm/min, tool tilt angle of 2.5°, and tool plunge depth of 0.2 mm. The welded samples were designated with abbreviations. For example, M800Al refers to the M-type pin with a rotational speed of 800 r/min and pre-set ER6061 welding wire. Before welding, four holes with a diameter of 1 mm (matching the thermocouple diameter) were machined at specified positions on the corner die (Fig. 1), and K-type thermocouples of the same length as the hole diameter were inserted and fixed. The data acquisition interval was 0.1 s. The specified positions on the corner die were the fillet voids adjacent to the contact surface between the skin and the stringer.

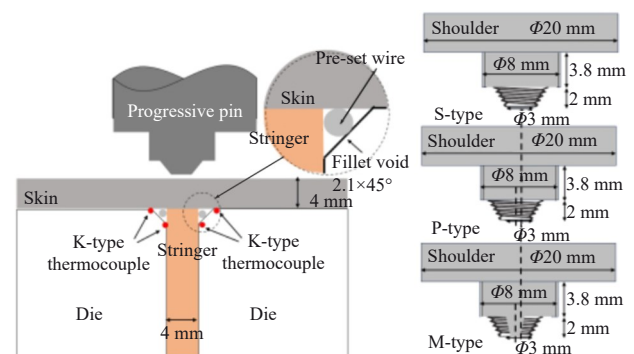


Fig.1 Schematic diagram of setup for welding and progressive pin size

The microstructure was investigated using a Zeiss EVO LS15 scanning electron microscope (SEM) equipped with an energy dispersive spectrometer (EDS), a stereomicroscope, and an HX-500F optical microscope. A solution of acetone: nitric acid:acetic acid=3:2:3 was used to etch the copper side of the metallographic samples, while Keller reagent was used to etch the aluminum alloy side. The mechanical properties were tested using a WDW-200 micro-controlled tensile testing apparatus. Tensile testing was performed at a crosshead speed of $1 \text{ mm} \cdot \text{min}^{-1}$. Microhardness tests were conducted with a load of 0.98 N and a dwell time of 10 s, with an indentation spacing of 0.5 mm. X-ray diffraction (XRD) analysis was conducted on the specific plane, the Al/Cu intersection plane. Electron back-scattered diffraction (EBSD) analysis was also conducted, and the test samples were electropolished in an electrolyte containing *n*-butanol, methanol, and perchloric acid at 20 °C and 25 V.

3 Results and Discussion

3.1 Morphology of the welds

The cross-sectional morphologies of the welds under three progressive pins are displayed in Fig.2, and they all display an onion-ring pattern. Considering the NZ volumes, the M-type tool produces a bigger NZ, while the S-type tool produces the smallest NZ. Compared with those in S800Al, the Cu hooks in P800Al and M800Al become thinner toward the shoulder, with a greater accumulation near the fillet void. Notably, the Cu hook height is significantly smaller with pre-set HS201 wires than with pre-set ER6061 wires at the corner. Pre-set HS201 wires reduce the adhesion at the bottom of the pin and also decrease the amount of Al/Cu mixture adhered by the stirring tool. As a result, the stirring tool greatly reduces the

volume of surrounding material during the forward movement, narrowing the effective NZ. Additionally, M800Cu and S800Cu exhibit small Cu hooks at the advancing side (AS) and retreating side (RS) of inner corners, with bonding-line defects appearing around the Cu hooks. These Cu hooks primarily originate from the pre-set HS201 wires, which have high deformation resistance and inferior fluidity, causing significant obstruction in the fillet void. Compared to that in N800Cu^[19], the Cu hook height in M800Cu, S800Cu, and P800Cu is significantly higher, indicating that the S-type, P-type, and M-type tools have a stronger stirring effect on the local materials in the fillet void, playing a crucial role in improving material fluidity at the corner die.

The assembly method with a certain-height boss platform is analogous to the offset adjustment toward the Al side in butt joints, enabling the adjustment of the Al/Cu mixing ratio in the NZ and thereby achieving the goal of controlling microstructure and properties. Compared to those in S800Al-boss joints, the Cu hooks in P800Al-boss and M800Al-boss joints exhibit a slight increase in height, while the large pin stirring zone (LPSZ) becomes wider. This is attributed to the design of the eccentric small pin, which expands the rotational trajectory of the tool, coupled with periodic compression exerted by the small shoulder in the same region. The left-right asymmetry of the small shoulder causes intermittent horizontal migration of material, leading to periodic alternating retention and compression near the corner die, which promotes upward migration of the Cu hooks. The volume of the SPSZ is constrained by the dimensions of the corner die, hindering further expansion. Since the SPSZ and LPSZ form a continuous NZ, the SPSZ transfers more heat to the LPSZ, thereby increasing its volume. Therefore, the eccentric tool design plays a positive role in promoting material flow within the NZ. The height of the Cu hook of M800Al is higher than that of P800Al. Besides, the height of the Cu hook of M800Al-boss surpasses that of the P800Al-boss. This result indicates that the addition of three cutting surfaces to the tool enhances material migration in the vertical direction.

3.2 Thermal analysis

The material flow behavior in the fillet void is susceptible to tool morphology, thereby influencing defect formation. During FSW, the fillet void can be effectively filled. The better the void-filling effect, the closer the temperature approaching the boundary temperature of the NZ. The difference in pin morphology affects heat generation in the NZ and the interaction time at the Al/Cu interface. Temperature plays a crucial role in the evolution of IMCs, Al/Cu mixing, and Cu hook migration. The fillet void is the closest to the corner die of the T-lap joint, and the temperature variations in this region can affect FSW processes. Additionally, the corner die can be equipped with a reusable thermocouple hole, avoiding errors caused by drilling holes in each workpiece. Fig.3a illustrates the inner corner temperatures under different pin types. The temperature at AS is consistently higher than that at RS, with the highest inner corner temperatures for the

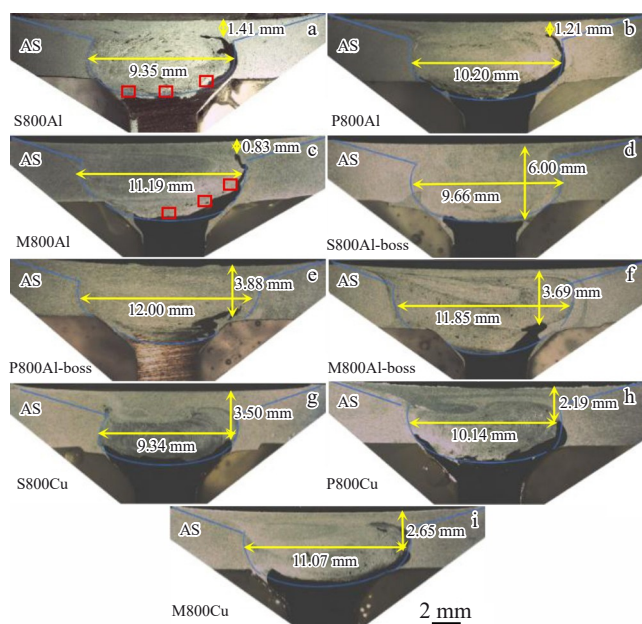


Fig.2 Cross-sectional morphologies of three types of progressive pin joints: (a) S800Al; (b) P800Al; (c) M800Al; (d) S800Al-boss; (e) P800Al-boss; (f) M800Al-boss; (g) S800Cu; (h) P800Cu; (i) M800Cu

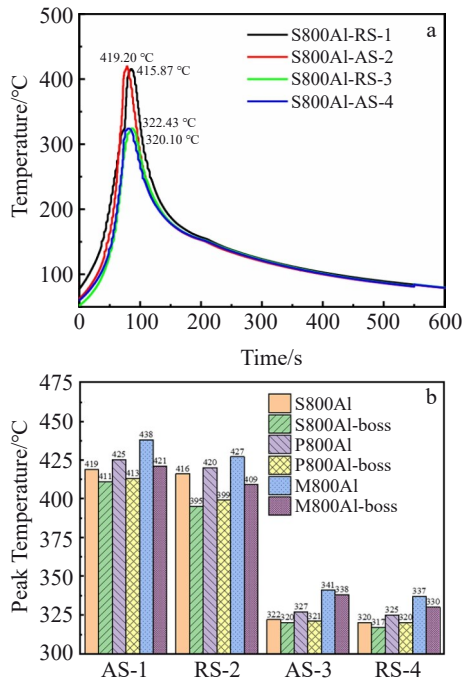


Fig.3 Thermal cycle curves of inner-corner weld of S800Al T-lap joints (a); peak temperatures of different joints (b)

S-, P-, and M-type pins being 419, 425 and 438 °C, respectively. The peak temperatures for the eccentric P- and M-type pins increase by 6 and 19 °C, respectively. Duong et al.^[14] reported inner-corner temperatures for conical threaded, cylindrical smooth, and triangular pins, with the triangular pin generating the highest heat, consistent with the present results. Both M800Al and M800Al-boss exhibit higher inner-corner temperatures, which is consistent with the largest NZ volume observed in Fig.2.

3.3 Microstructures around the keyhole and LPSZ

Fig. 4 shows the microstructures of the horizontal cross-sections of the LPSZ for S800Al and M800Al keyholes, as well as the central cross-section of the P800Al keyhole. Fig.4 shows that the root of the tool exerts a strong stirring effect on the pre-set welding wires, and a significant scraping effect is observed inside the Cu hook on the RS of the corner. In addition, a thread profile is present on the inner edge of the Cu hook, with its end penetrating deep into the pin thread. In other words, the Cu hook is mixed into the Al matrix under the action of shearing and stirring through scraping at the

corner and rotational spiral movement of its end with the pin, and it is distributed throughout the NZ.

On the AS of the keyhole, most material migrates forward with the stir tool; on the RS, the Cu hooks exhibit a wavy, curved morphology. This is due to the upward migration and insertion of Cu from the stringer into the softer LPSZ under the periodic rotational compression of the progressive pin, ultimately accumulating at the RS edge of the tool in a wavy pattern. Under the direct scraping and stirring action of the three-faceted progressive pin, the NZ undergoes intense thermo-mechanical coupling, causing more Cu to be scraped from the inner side of the Cu hooks and mixed into the Al matrix, gradually thinning and tapering the Cu hook tips. The pulsating effect of the three facets causes the scraped Cu to rotate and migrate with the tool, uniformly dispersing Cu throughout the NZ. As shown in Fig.4, the Cu particles at AS are finer than those at RS. This is because the freshly scraped Cu is first elongated and fragmented as stirring continues, and it is enveloped and transported by the highly deformable, fast-migrating Al matrix, undergoing continuous deformation and shearing. Some sheared Cu strips are rotated by the stirring pin to the AS before returning to the RS. Consequently, the total amount of Cu particles is the highest on the RS, with varying sizes, while on the AS, the Cu particles are fewer and finer.

3.4 Grain evolution adjacent to Al/Cu interfaces

Fig. 5a–5c present EBSD images of the AS, weld nugget center (middle), and RS of the Al/Cu interface of S800Al joints, respectively. Fig.6a–6c show EBSD images of the Al/Cu interface at the NZ center of M800Al, the inner side of Cu hook 1 at RS in SPSZ, and the inner side of Cu hook 2 at RS in LPSZ, respectively. Their relative positions within the NZ are marked by small squares in Fig. 2. Regardless of eccentricity, grains at the bottom of the NZ center exhibit smaller sizes than those on the inner side at RS of Cu hooks. This is attributed to the maximum distance from the shoulder and adjacency to highly conductive stringer Cu, promoting refined grain formation. Campanella et al.^[23] reported that the directional thermal conductivity of the backing plate accelerates cooling at the nugget center, generating a comet-shaped temperature distribution where the final solidification point shifts to the boundary between the nugget and heat-affected zone (HAZ). This aligns with our observation of finer grains in the central NZ. In contrast, grains on the inner side

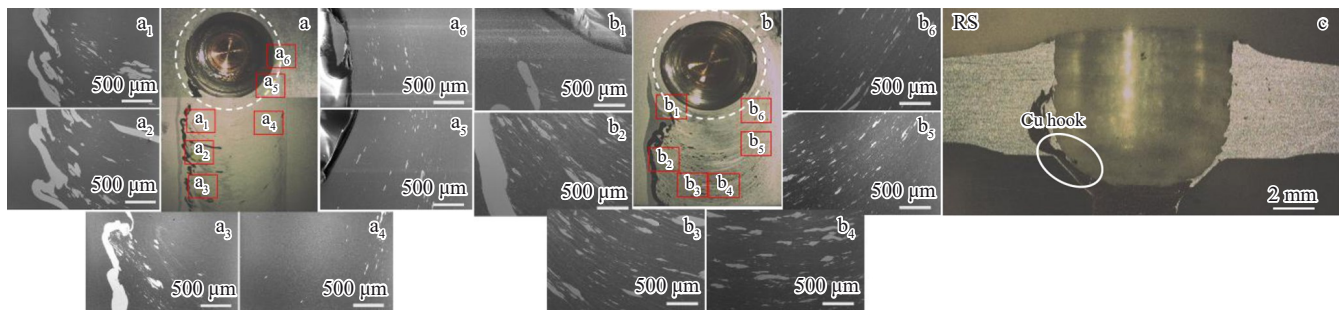


Fig.4 Microstructures around the keyholes of S800Al (a, a₁–a₆) and M800Al (b, b₁–b₆); central cross-section of P800Al (c)

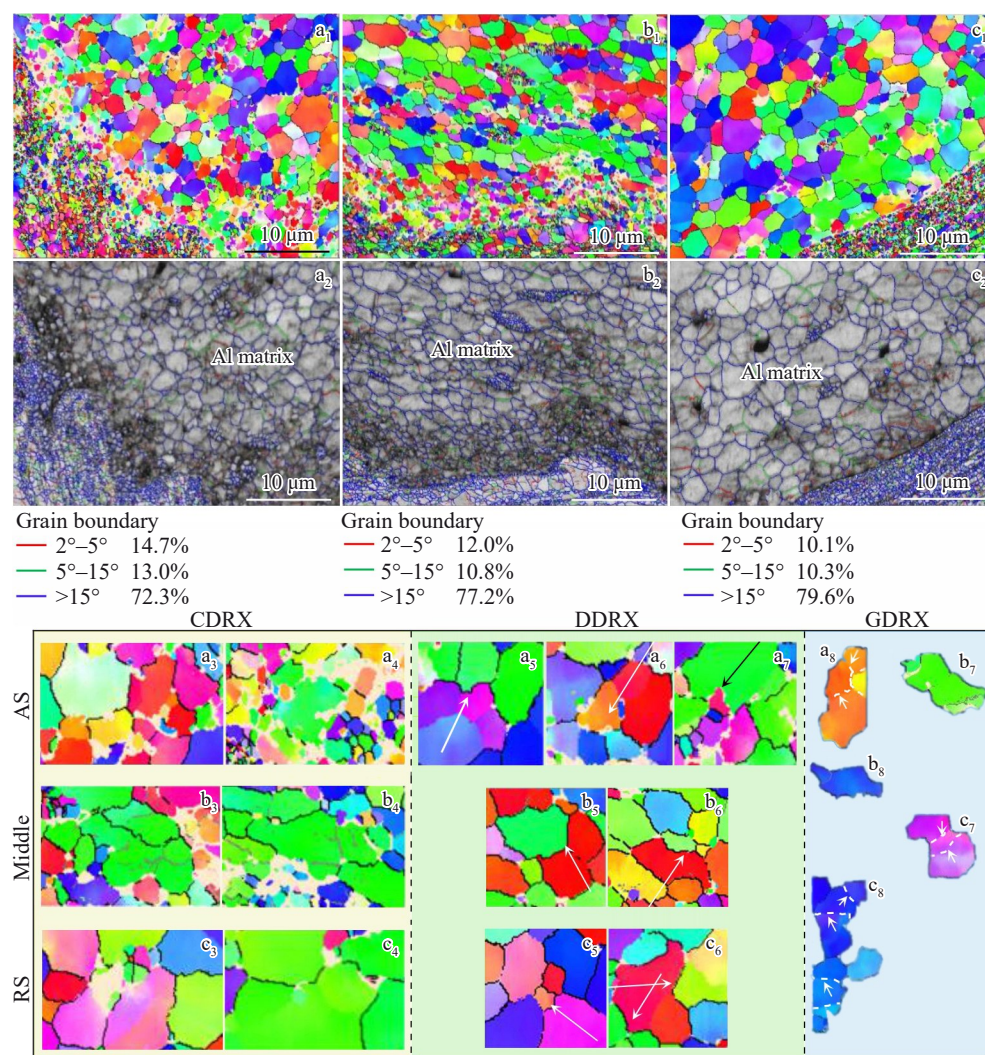


Fig.5 EBSD images of different regions near the Al/Cu interface of S800Al T-lap joints: (a₁–a₈) AS, (b₁–b₈) middle region, and (c₁–c₈) RS

of Cu hooks display elongated morphology with high aspect ratios, resulting from hindered material flow by Cu hooks and subsequent dynamic recrystallization (DRX). Fig.5 indicates that microstructure evolution in the NZ is dominated by recovery. Magnified views (Fig.5a₃–5a₈, 5b₃–5b₈, and 5c₃–5c₈) reveal grain boundary characteristics: most boundaries consist of both low-angle grain boundaries (LAGBs) and high-angle grain boundaries (HAGBs), exhibiting typical continuous dynamic recrystallization (CDRX) features^[24], while bulged regions display discontinuous dynamic recrystallization (DDRX) morphologies. Furthermore, some grains decompose into subgrains, forming LAGBs with small misorientations (indicated by arrows in Fig.5 and Fig.6), suggesting that geometric dynamic recrystallization (GDRX) also contributes to the microstructural evolution^[25]. This observation aligns with the DRX types on the Al side of dissimilar Al/Cu joints reported by You et al^[10].

The introduction of eccentricity in the small pin further enhances the plastic deformation within the NZ, creating favorable conditions for the activation of DDRX and GDRX. Under elevated strain conditions, grain recrystallization

increasingly proceeds via these mechanisms, resulting in a higher relative proportion of DDRX and GDRX events, as illustrated in Fig.6a₃–6a₈, 6b₃–6b₈, and 6c₃–6c₈. As shown in Fig.5b₂, the LAGB fraction in the center region of S800Al is 22.8%, whereas in M800Al, it decreases to 16.5% (Fig.6a₂). This reduction is primarily attributed to the intensified plastic deformation induced by the eccentric small pin, which promotes DRX activity and facilitates the transformation of LAGBs into HAGBs. Moreover, the LAGBs fraction at AS of S800Al is 27.7% (Fig.5a₂), which is significantly higher than that at RS (20.4%, Fig.5c₂). This is because the AS region experiences higher temperatures and more severe deformation, while the RS region is constrained by the obstructive effect of Cu hooks, leading to relatively lower strain accumulation.

Dislocation motion governs the degree of grain refinement, and it is strongly influenced by strain and thermal input. Elevated strain or higher strain rates typically elevate dislocation density, thereby increasing strain-energy storage and enhancing the driving force for recrystallization and grain growth. Shah et al^[26] demonstrated that eccentricity imposes

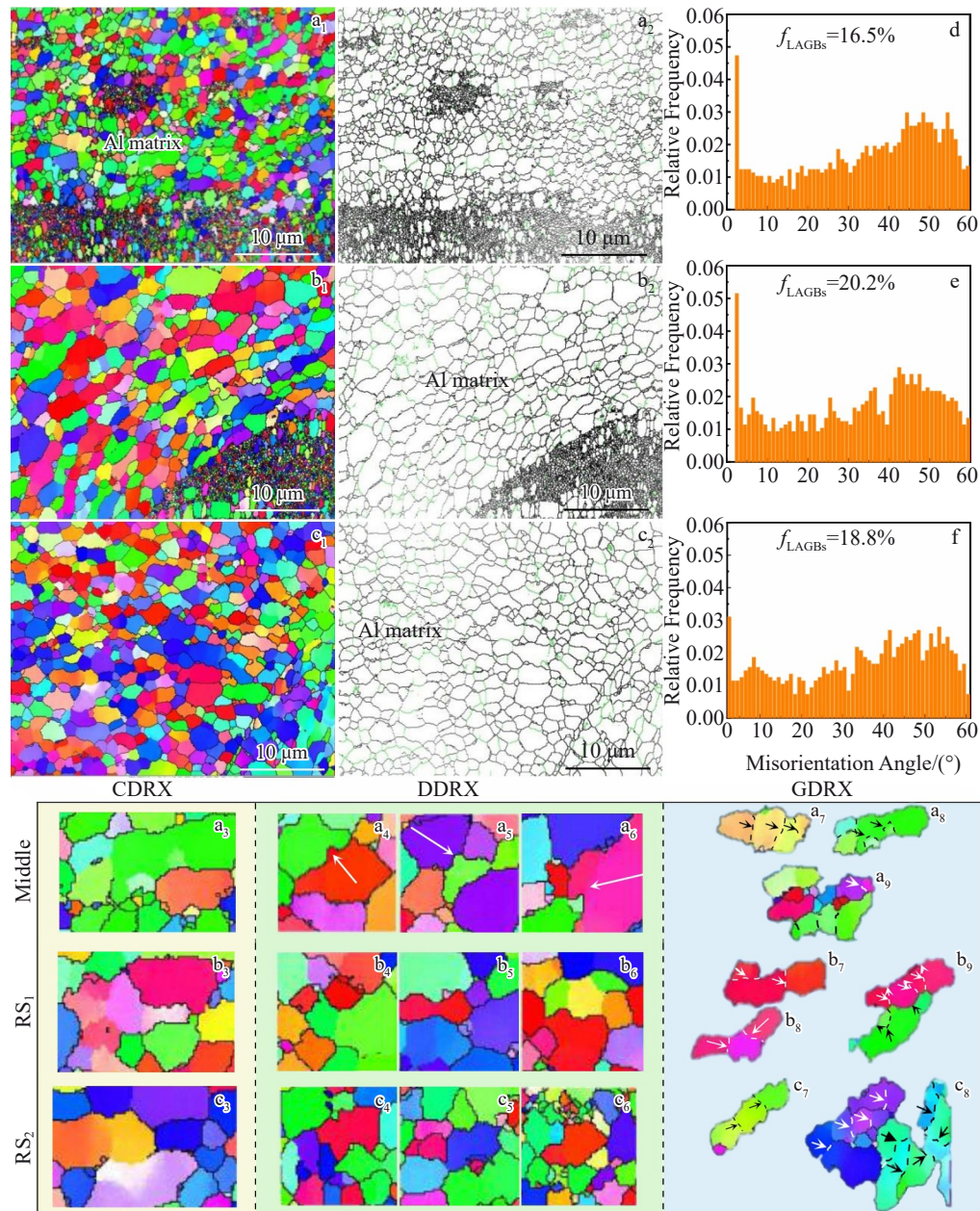


Fig.6 EBSD analysis of different regions near the Al/Cu interface of different T-lap joints: (a₁–a₉, d) M800Al-middle, (b₁–b₉, e) M800Al-RS₁, and (c₁–c₈, f) M800Al-RS₂

additional shear deformation on the stirred-zone material, which promotes further refinement of Cu particles, resulting in a reduction in their size. The average grain size of the Al matrix at the center of the NZ bottom, near the Al/Cu interface, decreases from 2.56 μm to approximately 1.06 μm , while the average grain size within the inner region of the Cu hooks at RS reduces from 4.13 μm to approximately 1.26 μm , as shown in Fig. 7. This significant grain size reduction confirms that the eccentric pin effectively promotes grain refinement within the stirred zone.

In FSW of dissimilar materials, the impact of IMCs on joint performance depends on their type, quantity, size, and distribution^[7]. In this study, the interfacial structure, particularly the composition and thickness of IMC layers, is a

decisive factor.

Compared to ductile IMC phases or thick IMC layers, thinner and more continuous IMC layers are generally associated with suppressed crack initiation at the transition zone, thereby enhancing the mechanical properties. To compare the thickness, size, and distribution of layered IMCs under two progressive pin configurations, the microstructural morphologies of the interfacial IMCs in both S800Al and M800Al joints were observed. The S800Al joint exhibits a bilayer interfacial structure, with IMC layer thickness ranging from 0.82 μm to 2.70 μm and presenting a relatively uniform, continuous distribution, as shown in Fig. 8. The atomic concentrations of Al and Cu within the transition layer, as quantified by EDS, are listed in Table 1 (the date in Table 1

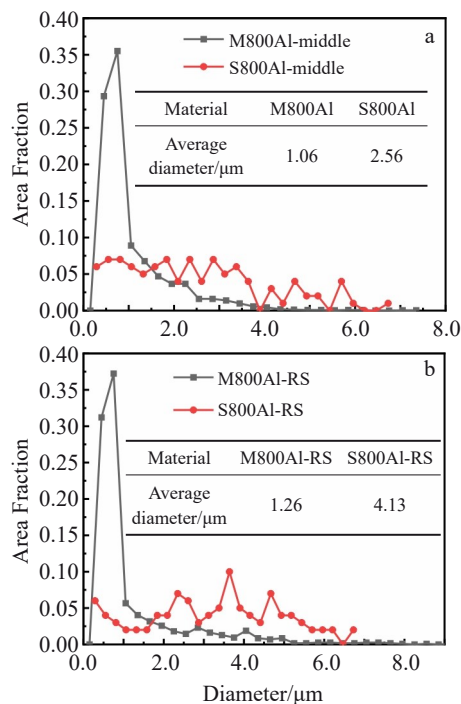


Fig.7 Grain sizes in middle region (a) and at RS (b) of M800Al and S800Al joints

have been normalized). Consistent with the findings of Xue et al^[27], optimal joint performance is achieved when the IMC layer thickness remains below 2.5 μm.

The eccentric M-type tool primarily governs material flow within the SPSZ. The periodic interaction between the eccentric pin and the corner die induces repeated extrusion and scraping, significantly enhancing material fluidity and promoting more intimate mixing of Al and Cu. This phenomenon results in a distinct vertical flow pattern at the Al/Cu interface, as illustrated in Fig.9. Such a flow morphology not only facilitates metallurgical bonding but also establishes a robust mechanical interlock, thereby improving joint mechanical performance. The sinusoidal motion of the eccentric pin generates a heterogeneous strain field, causing Cu particles to fracture from the stringer edges and to disperse

throughout the NZ. This process directly contributes to the formation of vertical flow structures at the interface and increases the dispersion of Cu particles within the Al matrix. As shown in Fig.9, the IMC layer thickness is reduced to 0.81–1.15 μm, which is thinner than that of the S800Al, and its structure evolves from a bilayered shape to a multilayered configuration, exhibiting significantly enhanced interfacial topography.

Pankaj et al^[28] demonstrated that the thickness of IMC layers formed at the interface can be expressed as:

$$\delta = K_0 t \exp\left(\frac{-Q}{RT}\right)$$

(1)

where δ is the thickness of IMCs; K_0 is the pre-exponential factor; t is the reaction time; Q , R , and T stand for the effective activation energy, universal gas constant, and temperature, respectively. According to Eq. (1), the higher the reaction temperature (T), the longer the reaction time (t); the lower the activation energy (Q), the easier the IMCs formation and the larger the thickness. Fig.3 shows that the reaction temperature under the M-type pin is higher than under the S-type pin. Consequently, the IMC thickness under the M-type pin should be larger. However, the calculated IMC thickness under the M-type pin is smaller. This result indicates that a single thermodynamic factor cannot fully explain the experimentally observed results. In fact, the complex deformation behavior of materials during FSW modifies the effective activation energy required for IMC formation. Rathod et al^[29] proposed that the factors influencing the activation energy can be expressed as:

$$Q_f = E_w + E_r + E_e$$

(2)

where Q_f is the activation energy; E_w , E_r , and E_e are the intrinsic activation energy for local fluctuation of the constituent, the activation energy required for lattice reconstruction of IMCs, and the deformation energy of the matrix surrounding the nucleation sites, respectively. Therefore, an increase in deformation will raise the effective activation energy for IMC formation, thus hindering the formation of IMCs. Furthermore, the IMCs at the joint interface undergo continuous plastic deformation under the shearing action of the pin. When the strain rate is relatively high, microcracks or even fracture are induced in the IMCs^[30].

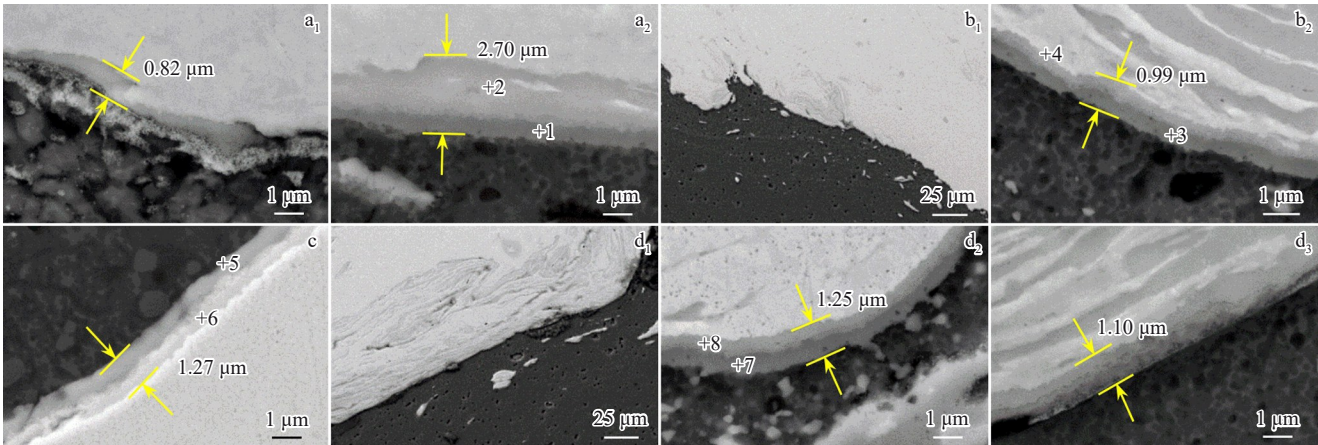


Fig.8 SEM morphologies of middle region (a₁–a₂), AS (b₁–b₂), and RS (d₁–d₃) of S800Al joints; SEM morphology of periphery of Cu hook (c)

Table 1 EDS analysis results of the positions marked in Fig. 8, Fig.9, and Fig.14 (at%)

Point	Al	Cu	Possible phase
1	66.05	33.95	Al ₂ Cu
2	35.09	64.91	Al ₄ Cu ₉
3	69.16	30.84	Al ₂ Cu
4	43.88	56.12	Al ₂ Cu ₃
5	68.27	31.73	Al ₂ Cu
6	51.43	48.57	AlCu
7	74.38	25.62	Al ₂ Cu
8	42.53	57.47	Al ₂ Cu ₃
9	64.25	35.75	Al ₂ Cu
10	37.04	62.96	Al ₄ Cu ₉
11	77.34	22.66	Al ₂ Cu
12	48.08	51.92	AlCu
13	79.78	19.22	Al ₂ Cu/Al ₄ Cu ₉
14	42.68	57.32	Al ₂ Cu ₃
15	54.23	45.77	AlCu
16	96.49	3.51	Al matrix

Kipp et al.^[31] proposed the relationship between the size of IMC fragments and strain as follows:

$$d = \left(\frac{12G_c}{\rho \dot{\epsilon}^2} \right)^{1/3} \quad (3)$$

where d is the size of IMCs; G_c , ρ and $\dot{\epsilon}$ are the fracture energy, density, and strain rate of the material, respectively. It can be observed that the size of IMCs decreases with the increase in strain rate. Given that the strain rate in M800Al is higher than that of S800Al, the interfacial IMCs in M800Al are more prone to fragmentation. Based on the aforementioned two mechanisms and the experimental results, the influence of strain rate dominates under the experimental conditions.

In addition, the periodic movement of the eccentric M-type pin reduces the flow-stress difference between Al and Cu, promotes coordinated deformation, and thus results in a multilayered distribution of IMCs. The enhanced deformation compatibility and extended flow path further facilitate the diffusion of more Cu particles into the Al matrix. The slightly

higher peak temperature at the inner corner of the M800Al prolongs the plastic flow time, contributing to the formation of a more complex multilayered IMC structure. However, the thickness of IMCs does not increase significantly due to the insufficient time for their full growth. Moreover, the pulsating effect of the three cutting surfaces strengthens the scraping effect on the inner side of the Cu hook, improves the degree of plastic flow in the weld NZ, and also favors the formation and development of the multilayered IMCs structure.

Considering the thickness of IMCs at the Al/Cu interface within the NZ, it is also necessary to examine the bonding state between the outer area of the large Cu hook and the Al matrix. The clogged wire is located at the periphery of the large Cu hook, far from the center of the NZ, and it undergoes DRX under extrusion. The periphery of the Cu hook is prone to the formation of bonding-line defects with the pre-set wire, which are the potential sites for stress concentration and crack initiation. High-magnification morphology reveals that the IMC layer in this region also exhibits a bilayered structure, as shown in Fig. 8c and Fig. 9c. The IMC thickness at the Cu hook periphery in S800Al is 1.27 μm , and it reduces to 0.99 μm in M800Al. This reduction is primarily attributed to the wider NZ in M800Al, which positions the Cu hook periphery farther from the heat source and closer to the highly conductive corner die, thereby lowering the local thermal input and suppressing interdiffusion and hindering the IMC growth between Al and Cu. Beygi et al.^[32] reported that in a dual-pass process, partially for the penetrated FSW T-lap joint of steel/Al, bonding-line defects formed at the Al/steel interface regions are not in contact with the tool pin. Besides, no IMCs were detected within these bonding-line defects. In contrast, under the present conditions, a thin (thickness <2.5 μm) and continuous bilayered IMC structure is formed at the non-stirred Cu hook periphery. The peripheral region of the Cu hook, uninvolved with the stirring pin, exhibits a thin, continuous IMC bilayer, indicating an optimal temperature regime that enables sustained atomic diffusion between the pre-set wire and the Cu substrate. The observed IMC layer thickness, consistently below 2.5 μm ^[27], signifies the establishment of a fine metallurgical bond, in alignment with

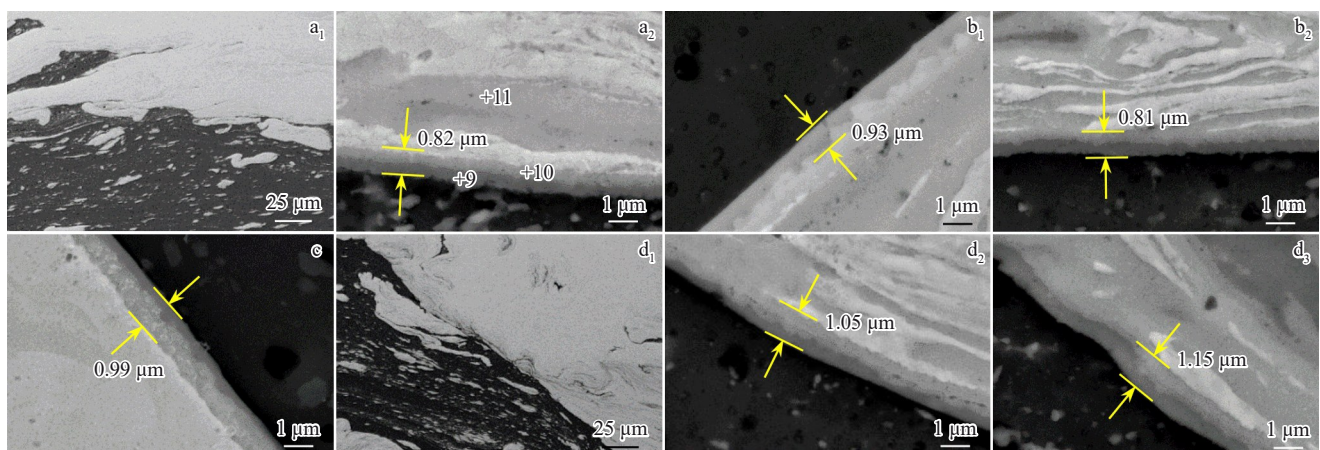


Fig.9 SEM morphologies of middle region (a₁-a₂), AS (b₁-b₂), and RS (d₁-d₂) of M800Al joints; SEM morphology of periphery of Cu hook (c)

the conclusions. Furthermore, the strong interfacial bonding between the Cu hook periphery and the Al matrix effectively mitigates stress concentration at the inner corner, contributing to superior mechanical performance along the stringer direction of the joint.

XRD tests were conducted on a certain section of the three types of joints, and the results are shown in Fig.10. Diffraction peaks of the AlCu phase and Al₄Cu₉ phase are observed, which is consistent with the point scan results in Table 1.

Fig. 11 presents the microhardness distributions of FSW T-lap joints fabricated with three progressive pins. Along the skin direction, the joint featuring a boss feature exhibits the lowest microhardness, attributable to the protrusion of 1 mm on the skin, which impedes the ingress of Cu into the NZ. In contrast, the P800Al and M800Al joints demonstrate significantly higher microhardness values than S800Al, owing

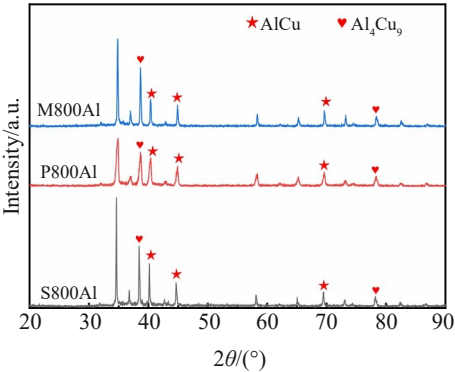


Fig.10 XRD patterns of cross-sections of different Al/Cu FSW T-lap joints

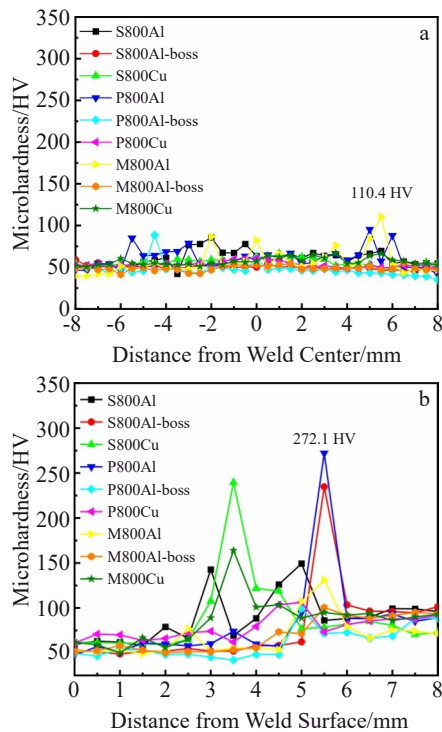


Fig.11 Microhardness distributions along skin (a) and stringer (b) of different T-lap joints

to intensified plastic deformation and DRX induced by the eccentric tool geometry, resulting in pronounced grain refinement within the NZ. Elevated microhardness peaks along the stringer direction are correlated with the formation of brittle, high-hardness IMCs at the Al/Cu interface, driven by interdiffusion under thermal cycling.

Fig.12 illustrates the ultimate tensile strength of the joints along both skin and stringer directions for different T-lap joints. In all cases, the ultimate tensile strength along the stringer direction exceeds that along the skin direction. Notably, the M800Al joint achieves a 23.9% increment in ultimate tensile strength (176 MPa) along the stringer direction relative to S800Al. This enhancement is attributed to the vortex-like interlocking microstructure generated by the eccentric tool, which promotes homogeneous material flow and improves mechanical interlocking between the deformed zones, thereby increasing the ultimate tensile strength of the joint.

The ultimate tensile strength of Al-Cu FSW T-lap joints with a protrusion of 1 mm on the skin plate is significantly higher than that of joints with a flat interface. The protrusion modifies the original position of the stringer Cu within the NZ, significantly reducing the amount of Cu hooks migrating to the periphery of the NZ during tool rotation. This alteration changes the Al/Cu mixing ratio in the NZ, increasing the effective bonding width and height, and improving joint performance. Therefore, adjusting the relative position of the original faying surfaces to modify the distribution of Cu in the NZ and the Al/Cu mixing ratio in the mixed zone is an effective approach to enhancing the ultimate tensile strength in the stringer direction.

The variation in the contact position between the pre-set ER6061 wire with high fluidity and the original surface promotes vertical material flow in the fillet void, thereby enhancing the mechanical properties in the stringer direction. The flow behavior of the fillet void is constrained by the pre-set wire, and it is also influenced by the movement of the surrounding material. When the skin plate has no protrusion, the high-fluidity ER6061 is obstructed by the rib Cu, preventing timely filling of the fillet void. Even if the high-fluidity ER6061 adequately fills the fillet void, it may stagnate. Consequently, the material flow state in this region is

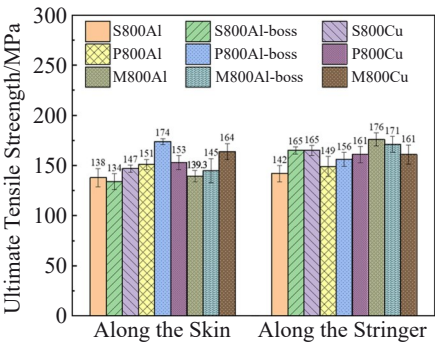


Fig.12 Ultimate tensile strength along the skin and stringer of different T-lap joints

poorer, leading to the formation of bonding-line defects. However, when a protrusion of 1 mm is introduced on the skin plate, the Al-based protrusion facilitates better bonding with the ER6061 filler wire, thereby improving the flow into the fillet void. Additionally, the rapid cooling at the corner die means that the combination of the pre-set ER6061 wire and the protruding skin plate can more easily maintain the high-temperature dwell time in the internal corner, enabling more complete material flow in the fillet void.

All fractures along the stringer direction occur precisely at the Al/Cu interface. Therefore, joints with representative fracture characteristics are selected to elucidate their fracture behavior. Fig. 13 illustrates the fracture locations and nearby deformation characteristics of S800Al and M800Al joints. Joints with processed bosses exhibit limited elongation along the stringer direction, whereas the skin of both S800Al and M800Al joints undergoes substantial plastic deformation prior to fracture. Notably, the deformation on the AS is far more pronounced than that on the RS (Fig. 13c–13d). This indicates that under continuous tensile stress, the bonding quality at the Al/Cu interface is superior to that on the AS of the NZ. Plastic deformation initially occurs in the Al matrix on the AS, accompanied by work hardening in this region. Simultaneously, microcracks are initiated at the IMCs along the interface. These phenomena demonstrate that the joint possesses a certain level of toughness in the stringer direction. It is worth noting that a spalled Al-based composite structure is observed near the RS of the M800Al-boss joint, as presented in Fig. 13b. This phenomenon verifies that the bonding strength of the interfacial IMCs is approximately equivalent to the strength of the Al matrix in the NZ. This finding is consistent with the results reported by You et al.^[10], who proposed that Cu particles dispersed in the Al matrix provide favorable propagation paths for cracks.

Fig. 14 presents the tensile fracture morphologies along the skin and stringer directions under two progressive pin conditions. Fig. 14a₁ reveals small, shallow equiaxed dimples, indicative of ductile fracture occurring within the Al matrix. Fig. 14a₂ and 14a₃ show SEM images in backscattered and secondary electron modes of the Al/Cu interface in S800Al at

the same location, respectively, where the Al matrix is enveloped by a significant number of IMCs with uneven particle sizes. Fig. 14b₁–14b₃ show fractures along the stringer direction. In contrast to the dimples observed in the skin direction, Fig. 14b₁ exhibits a smooth fracture surface with brittle fracture characteristics. The magnified view of the yellow-boxed area in Fig. 14b₁ (Fig. 14b₂) reveals elongated dimples of varying sizes, while further magnification of the yellow-boxed area in Fig. 14b₂ (Fig. 14b₃) shows particles within the dimples.

Fig. 14c₁–14c₆ present the fracture morphologies of M800Al along the stringer direction. The undulating Al/Cu interface effectively enhances bonding strength, particularly where the layered banded composite structure is disrupted. A magnified view reveals the particles embedded within shallow dimples. The point-scan results of the particles are shown in Table 1, which are identified as Al₂Cu or Al₄Cu₉ phases. Larger particles embedded in the banded structure exhibit microcracks under sustained tensile loading (Fig. 14c₄), attributed to the insufficient toughness of the brittle IMCs formed by encapsulation. The gradient distribution of these IMCs at fracture sites indicates a mixed ductile-brittle fracture mode in the joint. The intergranular fracture of Al₄Cu₉ (Fig. 14c₅) reveals granular grain boundaries, with microcracks further demonstrating the role of inferior toughness in stress transfer. These findings highlight the contribution of microcrack initiation and propagation to tensile strength, corroborating the mechanical and metallurgical improvements achieved by the eccentric small pin.

3.5 Discussion

Under the pulsating effect of the eccentric small pin, the brittle, hard IMCs are more prone to migration into the Al matrix and fragmentation, and Cu particles are also more easily broken. These fragments provide new sites for interfacial mutual diffusion and the continuous formation of IMCs^[22]. As the tool completely exits the plastic deformation zone, smaller Cu particles and IMCs are distributed throughout the NZ. The three-flat-face tool increases the strain in the NZ, scraping off and refining more Cu particles from the inner side of the Cu hook. The sinusoidal variation in the interaction between the eccentric pin and the workpiece generates an inhomogeneous strain field, promoting the fragmentation of Cu particles. Simultaneously, repeated shear action accelerates the mutual diffusion between Al and Cu. The eccentric pin primarily alters the width of the large Cu hook at the lap interface and reduces the size of Cu fragments, without focusing on its regulatory role in interfacial phases. The dual effects of mechanical refinement and metallurgical bonding by the eccentric pin contribute to achieving high-quality welds.

An eccentricity of up to 0.3 mm is considered to limit the formation of hooks in lap joints^[33]. Shah et al.^[34] emphasize that eccentricity results in the plastic material, which is incompressible in fluid mechanics, being easier to flow around the tool. Under the action of a small eccentric pin, the gap between the progressive pin and the corner die undergoes

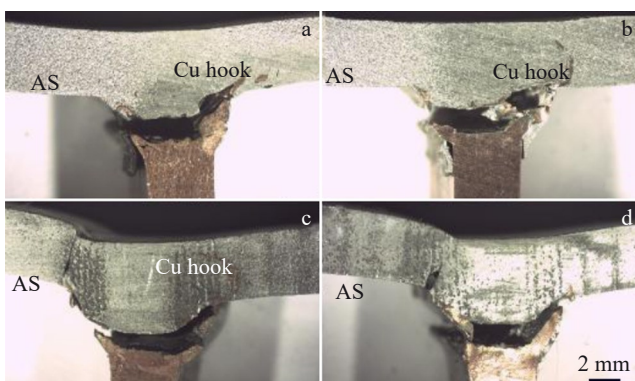


Fig. 13 Fracture locations of different Al/Cu FSW T-lap joints: (a) S800Al-boss; (b) M800Al-boss; (c) S800Al; (d) M800Al

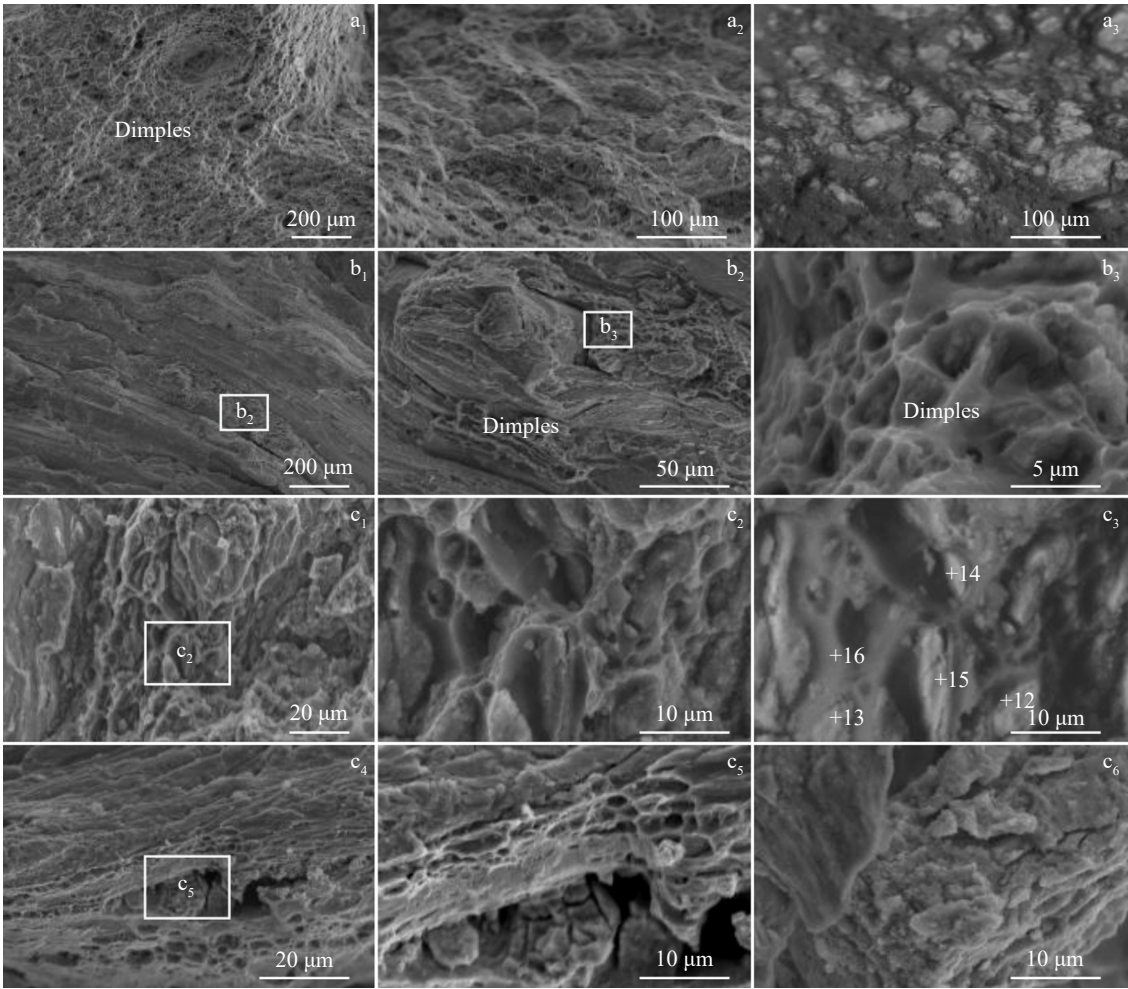


Fig.14 SEM morphologies of fracture surfaces in different T-lap joints: (a₁–a₃) along the skin direction of S800Al; (b₁–b₃) along the stringer direction of S800Al; (c₁–c₆) along the stringer direction of M800Al

sinusoidal variation (as shown in Fig. 15), creating an inhomogeneous stress field that reduces the sustained squeezing action of Cu hooks and thereby decreasing the

amount of Cu hooks leaving in the LPSZ. Fonda et al^[35] proposed a reverse shear forming mechanism caused by eccentric motion. When the eccentric pin is pressed forward in the

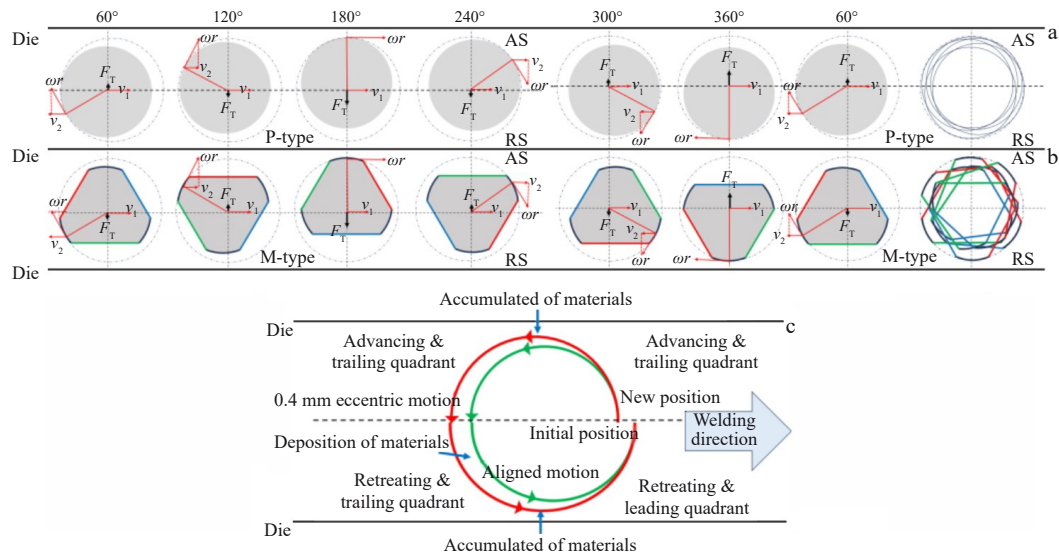


Fig.15 Schematic diagrams of relationship among the angular velocity, eccentricity, and tool velocity: (a) circular pin without a facet; (b) circular pin with three facets; (c) detailed explanation of eccentric motion

welding direction, the material is sheared in front of the progressive pin and extruded around it, depositing in the weak zone of the pin. The relationship among the angular velocity, eccentricity, and tool velocity in this study is illustrated in Fig. 15 (F_T refers to the reaction force exerted by the corner die), and they can be expressed by Eq. (4–6), where v_1 is the welding speed, r is the eccentricity, and ω is the rotational linear speed. Since the tool tilt angle is 2.5° , $\cos\alpha$ is approximately 1, and the resultant velocity is the sum of v_1 and v_2 . At an eccentricity of 0.4 mm, the maximum and minimum velocities are 34.08 and -32.91 mm/s, respectively. Fig. 16 shows a significant periodic variation in velocity. The negative minimum velocity suggests that material migration occurs in the opposite direction to the welding advance on the RS.

$$v_1 = 35 \text{ mm/min} \approx 0.583 \text{ mm/s} \quad (4)$$

$$v_2 = \omega r \cos\alpha \cos\omega t \quad (5)$$

$$v_1 + v_2 = v_1 + \omega r \cos\alpha \cos\omega t \quad (6)$$

The eccentric small pin used in this study features a considerable taper angle, which interacts with the corner die to induce alternating material flow behavior. When the short axis of the pin is in contact with the corner die, shear extrusion occurs, pushing material outward. However, when the long axis is engaged, the material is retained in the opposite shear direction. These two localized regions subject the material to cyclic shear strains in opposing directions, periodically compressing the stringer Cu and reducing the thickness at the Cu hook tip. When the gap between the short axis and the AS is small, the RS experiences a larger gap. When the gap between the minor axis and the AS is small, and the gap on the RS is large, a portion of the stringer Cu changes its flow direction under the forging action of the major axis (Fig. 17a–17b). The downward-extruded aluminum alloy matrix, under the effect of the small gap, is instead squeezed upward. Conversely, when the short axis approaches the RS with a reduced gap, the AS gap widens, causing the Al matrix to extrude downward, while Cu on the RS is pushed toward the outer periphery of the large pin (Fig. 17c–17d). This periodic reversal of local material flow direction within the stir zone enhances the interaction between the stirring tool and the plasticized material, resulting in a distinctive morphological

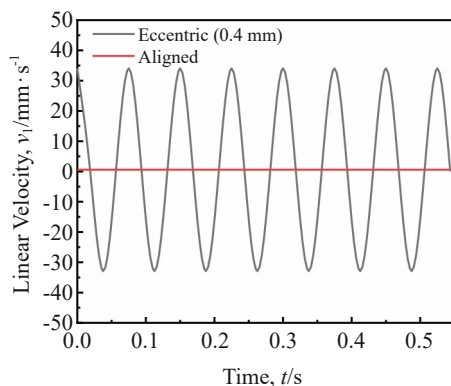


Fig.16 Time-dependent linear velocity of aligned and eccentric small pins

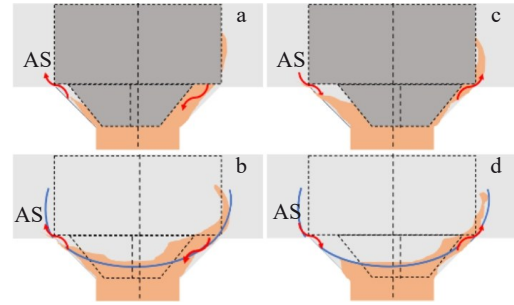


Fig.17 Schematic diagrams of Cu hook evolution under eccentric small pin: (a–b) small gap between minor axis and AS; (c–d) large gap between minor axis and AS

feature of the Cu hook: fine and curved at the tip, yet accumulated at the bottom. Additionally, conventional stirring tools tend to adhere a significant amount of Al/Cu mixture during rotation^[36], thereby increasing the effective rotational volume of the tool. However, in the case of the eccentric small pin, when the gap between the pin and the corner die is minimal, the adhered mixed material is periodically scraped off by the corner die. This cyclic scraping not only prevents excessive buildup but also disrupts the continuity of local material flow around the corner die, exerting a critical influence on the migration behavior of the Cu hook.

The M-type pin with an eccentric three-faceted structure not only enhances the pulsation effect but also alters the gap between the stirring pin and the corner die (Fig. 17b), thus leading to a change in the stirring behavior. The three-faceted structure is an additional feature based on the P-type pin. Comparing the gaps between the corner die and the P-type/M-type pins during 360° rotation shows that even when the gap of the M-type pin is relatively small, it is still larger than that of the P-type pin, which reduces the scraping amount of the corner die. The reduction magnitude is not only related to the rotation speed and welding speed, but also associated with the depth of the three-faceted structure and the pulsation effect. As shown in Fig. 15, F_T refers to the reaction force exerted by the corner die. It is revealed that compared with the P-type pin, the M-type pin exhibits a higher direction change frequency of F_T , and its three facets can store more metal, which in turn affects the scraping amount. In addition, the addition of three facets to the small pin slightly increases the size of its small shoulder, which changes the material flow direction in the stirring zone of the small pin, thereby further weakening the upward extrusion effect on Cu. Under the pulsation effect induced by the three-faceted structure of the large pin, the massive Cu hook scrapes off more Cu particles, which mix into the Al matrix, as indicated by the slender fractured Cu phases observed in the onion ring banded structure. Compared with the P-type pin, even if the intermixing ratio of Al-to-Cu remains unchanged, the M-type pin is more conducive to the mixing of more Cu fragments into the Al matrix. It promotes the uniform distribution of Cu particles in the banded Al matrix and reduces the thickness of the Cu hook, thereby improving the bonding quality between

the Cu hook and the thermo-mechanical affected zone (TMAZ).

Eq. (7–8) show that the eccentric motion increases heat input^[37–38]. As shown in Fig.3, the eccentric small pin generates more heat. Due to the restriction imposed by the corner die, the enlarged volume of the SPSZ cannot fully accommodate the increased heat, transferring excess thermal energy to the LPSZ. This phenomenon promotes the expansion of the LPSZ volume and elevates the temperature in the NZ, facilitating the migration of the Cu hook into the softer LPSZ region and forming a higher Cu hook profile. However, the thickness of the Cu hook is significantly lower than that produced by the S-type pin under non-eccentric conditions with lower heat input. This result suggests that the periodic forging effect induced by the alternating long and short axes of the eccentric pin on the SPSZ is weakened, which in turn reduces the vertical movement of the Cu hook while enhancing its horizontal displacement. Meanwhile, the corner die plays an auxiliary role in the migration of the Cu hook. The sinusoidal horizontal fluctuation of the eccentric small pin results in a correspondingly periodic reaction force from the corner die, which suppresses the upward migration of the Cu hook. Conversely, the expansion in the horizontal direction favors the intermixing of Al and Cu. Both effects contribute to improved welding quality.

$$Q_{\text{Total}} = \frac{2}{3} \omega \tau_{\text{contact}} \left[\frac{R_s^3 - R_p^3}{\cos \alpha} + 3H_p(R_p + e)^2 + (R_p + e)^3 - e^3 \right] \quad (7)$$

$$\frac{T_{\text{MAX}}}{T_s} = 2 \times 10^{-4} \times \frac{Q_{\text{Total}}}{v} \times \frac{H_p}{t} + \frac{0.5R_p}{(R_p + e)^2} \quad (8)$$

The tensile test results of the M800Al joint along the stringer direction are compared with the previously reported results of 6061Al/Cu bottom joints with different thicknesses, as shown in Fig. 18. The increase in strain rate in the NZ caused by the eccentric pin is conducive to grain refinement and promotes intermixing, which in turn triggers recrystallization mechanisms, reduces the thickness of IMCs, and diminishes the possibility of internal angles or Cu hooks becoming fracture initiation sites, thereby improving the bonding strength of the interface. This improvement is attributed to the favorable metallurgical bonding and the enhanced dual-mechanical interlocking effect, namely the longer, more vortex-like interlocking interface at the Al/Cu

boundary and the fine bonding between the Cu hooks and the TMAZ. The results in Ref. [9, 22, 39–46] indicate that the tensile strength is decreased with the increase in sheet thickness, and the values obtained in the present study are comparable to the reference data, suggesting that the observed dual-mechanical interlocking enhances the mechanical properties of the joint.

4 Conclusions

1) The eccentric structure promotes material migration in both horizontal and vertical directions, and the eccentric pin with three cutting faces shows a stronger enhancement effect on vertical material movement. The skin boss design reduces Al/Cu mixing, as well as the design of the height of the Cu hook and the distribution of Cu in NZ.

2) The small eccentric progressive pin exerts a significant stirring effect on the local material at the corner. It scrapes the Cu hook on the RS, driving the hook tip deep into the pin threads and facilitating Cu fragmentation. The additional three cutting faces further enhance the shearing and fragmentation in the corner region and on the Cu hook. The eccentric pin with three cutting faces promotes the transformation from LAGBs to HAGBs, tends to realize DRX via DDRX and GDRX, and refines grains in the SZ. The average grain size at the center of the NZ interface is reduced from 2.56 μm to approximately 1.06 μm , while the average grain size inside the Cu hook on the RS is decreased from 4.13 μm to about 1.26 μm .

3) The cyclic action of the M-type tool enhances plastic flow in the NZ, leading to the formation of multilayered IMCs at the interface with a thickness reduced to 0.81–1.15 μm . Meanwhile, the vortex-like interlocking structure formed at the interface is beneficial for enhancing in mechanical properties. The ultimate tensile strength of the M800Al joint along the stringer direction reaches 176 MPa, showing an increment of 23.9% compared to that of the S800Al joint.

4) The vortex characteristics of material flow in the SPSZ and the evolution of the Cu hook are mainly influenced by three factors. Firstly, theoretical inference suggests that under the experimental parameters, material on the RS undergoes backward migration; secondly, the asymmetric small shoulder periodically changes the flow direction of part of the material; thirdly, the periodic change in the gap between the pin and the corner die scrapes off the mixture adhered to the pin tip. The eccentric pin with three cutting surfaces changes the reaction force direction more frequently, promoting the coordinated deformation of Al/Cu.

5) The strengthening mechanism of high-quality T-lap joints obtained by eccentric progressive pins is as follows. Firstly, the stirring range of the NZ is expanded, which intensifies plastic deformation, reduces the thickness at the tip of the Cu hook, and improves the bonding quality between the Cu hook and the matrix. Secondly, the vortex-like mechanical interlocking at the interface is enhanced, and the IMCs are thinned, thereby improving the bonding strength of the joint. Thirdly, the grains at the interface are refined, the difference

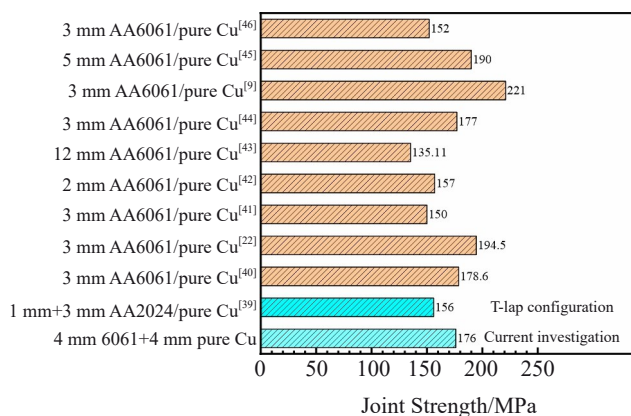


Fig.18 Comparison of joint strength of different Al/Cu FSW joints

in grain size across the interface is reduced, the distribution of Cu fragments in the NZ is optimized, and the hardness of the NZ is increased.

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采用预置焊丝和偏心阶梯焊具实现高质量 Al/Cu T-搭接FSW 接头

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摘 要: 通过搅拌摩擦焊 (FSW) 制备的 Al/Cu T-搭接接头中, 焊核区存在的钩状结构是裂纹的萌生源, 严重影响接头性能。采用偏心阶梯焊具对 6061 铝合金与纯铜进行 T-搭接 FSW, 通过在内角预置两种成分焊丝, 试验了 3 种焊具对 T-搭接接头的适用性, 以期改善钩状结构与基体的结合质量, 从而获得高质量的接头。优选两种焊具, 观察其对 Al/Cu 交界面处晶粒形态及金属间化合物 (IMCs) 形貌影响, 并评估了接头沿壁板和筋板方向的力学性能。结果表明, 在减小界面 IMC 层厚度的同时, 小针偏心的焊具减薄了 Cu 钩端部厚度, 扩大了焊核区体积, 且 Cu 钩外围与基体达到了冶金结合, 提高了焊接质量。小针偏心加三切面的 M800Al 焊具, 其周期性剪切作用促使形成涡流状互锁界面, 且从 Cu 钩内刮擦的更多碎片混入整个焊核区, 细化晶粒的同时有助于不连续动态再结晶和几何动态再结晶。采用偏心焊具制备的 M800Al 接头沿筋板方向的抗拉强度达 176 MPa, 与无偏心焊具制备的接头强度相比增加了 23.9%, 这为异种金属 T-搭接接头性能改善提供了一种可行的解决方案。

关键词: Al/Cu 异种合金; FSW; T-搭接接头; 预置焊丝; 偏心阶梯焊具; 显微组织; 性能

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