

Effects of friction time and rotational speed on aluminum-copper friction weld quality

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ABSTRACT

Rotary friction welding is a promising solid-state joining technique for dissimilar aluminium–copper joints. This study investigated the effects of friction time (30–60 s) and rotational speed (1000–2000 rpm) on the microstructure and mechanical properties of AA6061–Cu joints. The welded joints were characterized by optical microscopy, SEM–EDX, micro-Vickers hardness, and tensile testing. Increasing the friction time from 30 to 60 s increased the ultimate tensile strength from 19.15 MPa to 28.44 MPa and increased the average IMC thickness from 43.68 μm to 136.78 μm . At a constant friction time of 60 s, the rotational speed of 1300 rpm produced the highest tensile strength of 63.90 MPa, the highest average hardness of 74.52 HV, and an average IMC thickness of 204.61 μm . These results indicate that appropriate control of friction time and rotational speed improves interfacial bonding and mechanical performance through controlled IMC evolution and thermo-mechanical interaction.

Keywords: aluminium, copper, friction time, rotary friction welding, rotational speed.

INTRODUCTION

Welding technology is one of the most important joining processes in the manufacturing, transportation, and electrical industries. These industries require materials that are strong, lightweight, ductile, corrosion-resistant, and cost-effective, particularly for automotive and aerospace applications [1]. In many cases, dissimilar materials are joined to combine their desirable properties. One of the most widely used dissimilar material combinations in the electronics, electrical, and electric mobility sectors is aluminium and copper, which play a crucial role in the development of integrated, optimized, and lightweight electrical conductors [2, 3].

These two materials are widely used as electrical conductors because of their low electrical resistivity, which is $2.65 \times 10^{-8} \Omega \cdot \text{m}$ for aluminium and $1.68 \times 10^{-8} \Omega \cdot \text{m}$ for copper at 20 °C [4]. Although copper exhibits higher electrical

conductivity, its application is often limited by its higher cost and lower availability in the market. In contrast, aluminium offers several advantages, including a lower melting point, excellent corrosion resistance, lower cost, and abundant availability [2–6]. Among aluminium alloys, AA6061, which is primarily composed of aluminium, magnesium, and silicon, is considered one of the most promising materials for industrial manufacturing applications [1]. However, joining dissimilar materials remains challenging because of differences in their thermophysical properties. Conventional fusion welding methods are generally unsuitable for such material combinations, as they tend to promote the formation of hard and brittle intermetallic compounds (IMCs) at the interface [10]. Consequently, rotary friction welding (RFW) has emerged as a more effective alternative, capable of producing strong joints despite significant differences in physical and chemical characteristics, including electrochemical

potential, stiffness, and thermal expansion coefficients, between the materials [9–11].

Friction welding can produce joints with improved structural integrity, thereby reducing the risk of premature failure [14]. This process is a solid-state joining technique in which heat is generated through interfacial friction between metals combined with the application of axial pressure [15]. Compared with conventional fusion welding, friction welding offers several advantages, including the absence of flux requirements, minimal spatter formation, and a significantly lower likelihood of weld defects [16]. However, achieving high joint quality remains challenging because of differences in the chemical composition as well as the mechanical and physical properties of the dissimilar metals being joined [17, 18]. The key parameters influencing joint performance include friction pressure, friction time, rotational speed, and material surface condition [19]. Aluminium and copper are considered a relatively mature and well-understood dissimilar metal pair; therefore, the optimization of RFW parameters is less complex than that required for more advanced materials. In addition, friction welding of aluminium and copper can produce high-quality joints with greater cost efficiency [3, 20, 21].

Several previous studies have investigated the effects of welding parameters on aluminium–copper friction welding. Chapke et al. [22] reported that a friction time of 4 seconds was sufficient to improve joint strength in AA6063–Cu welds. Similarly, Pavethan et al. [23] and Mattie, et al. [24] found that a friction pressure of 60–80 MPa, friction time of 4–5 s, and rotational speed of around 1500 rpm produced high joint strength, increased hardness, and a more homogeneous microstructure. According to Huber et al. [25], maintaining a proper balance between friction pressure, friction time, and rotational speed is essential to prevent the excessive formation of IMC layers, which can reduce mechanical properties. Overall, previous research indicates that increasing friction time and rotational speed enhances heat input, atomic diffusion, and IMC thickness. However, excessive heat can cause grain growth and brittleness in the joint. The balance of friction time and rotational speed parameters in the friction welding of AA6061 and pure copper needs to be further investigated to improve the material's strength and hardness. Recent work by Milašinović et al. (2024) demonstrated excellent joint efficiency using short friction durations

(~4 s) combined with thermography, electrical conductivity measurements, and process monitoring. Their study achieved 100% joint efficiency, 61% ductility, and electrical conductivity exceeding 92% of the theoretical value. However, the study primarily focused on process optimization under short welding cycles. In contrast, the present study investigates the thermo-diffusion mechanisms occurring during prolonged friction durations (30–60 s), where IMC evolution, thermo-mechanical equilibrium, and fracture behavior become dominant [26].

Although several studies have investigated Al–Cu rotary friction welding, most have focused on short friction durations (2–6 s) and conventional parameter optimization. The thermo-diffusion behavior under prolonged friction durations and its influence on IMC growth stability, fracture behavior, and thermo-mechanical equilibrium remain insufficiently understood. Therefore, this study investigates the individual and sequential effects of friction time (30–60 s) and rotational speed (1000–2000 rpm) on diffusion-assisted bonding, IMC evolution, hardness distribution, and tensile performance of AA6061–Cu rotary friction welded joints. Although Al–Cu rotary friction welding has been extensively investigated, previous studies have mainly focused on process optimization under short friction durations (2–6 s). The present work does not aim to propose a new welding process; instead, it provides a mechanistic understanding of diffusion-controlled interfacial evolution during prolonged friction durations (30–60 s). Particular emphasis is placed on the relationship among diffusion kinetics, IMC growth, thermo-mechanical equilibrium, TMAZ localization, and fracture behavior.

MATERIALS AND METHODS

Materials

The materials used in this study were AA6061-T6 aluminium alloy (or the actual temper condition) and commercially pure copper cylinders. According to the manufacturer's specification, the AA6061 alloy has a typical ultimate tensile strength of approximately 310 MPa, yield strength of approximately 276 MPa, and Vickers hardness of approximately 107 HV. The temper condition was maintained throughout the machining process prior to welding. Magnesium and

silicon serve as the primary alloying elements in AA6061. The chemical compositions of both materials are presented in Table 1 and Table 2. The AA6061 and pure copper cylinders were machined to a diameter of 13 mm and a length of 72 mm, as illustrated in Figure 1(a).

Figure 2 presents a schematic representation of the rotary friction welding process performed using a lathe-based CL-3000 Series machine (XYZ Machine Tools Ltd., Birmingham, United Kingdom), equipped with a 3 HP motor, a maximum rotational speed of 3000 rpm, and a 6-inch chuck. During welding, the pure copper cylinder rotates, while the AA6061 aluminium cylinder remains stationary. The process begins with an initial friction phase that generates heat, followed by the forging phase, which produces a solid-state joint. The welding process employed in this study was continuous-drive friction welding (CDFW). In this process, the rotational speed was continuously supplied by the lathe motor and maintained constant during the friction stage until the predetermined friction time was reached, after which the upsetting pressure was applied to complete the solid-state joint. Unlike Inertia Friction Welding (IFW), no flywheel was used as the source of rotational energy. This study aims to determine the optimal friction time and rotational speed parameters for dissimilar rotary friction welding, serving as a continuation of previous research that identified an optimal friction pressure of 1.96 MPa [27]. The detailed welding parameters, including variations in friction time and rotational speed for the Al–Cu joints, are summarized in Table 3.

Figure 2 presents the detailed actual flow process of Al–Cu rotary friction welding used in this study. The process began with the preparation and machining of AA6061 aluminium and pure copper rods into cylindrical specimens

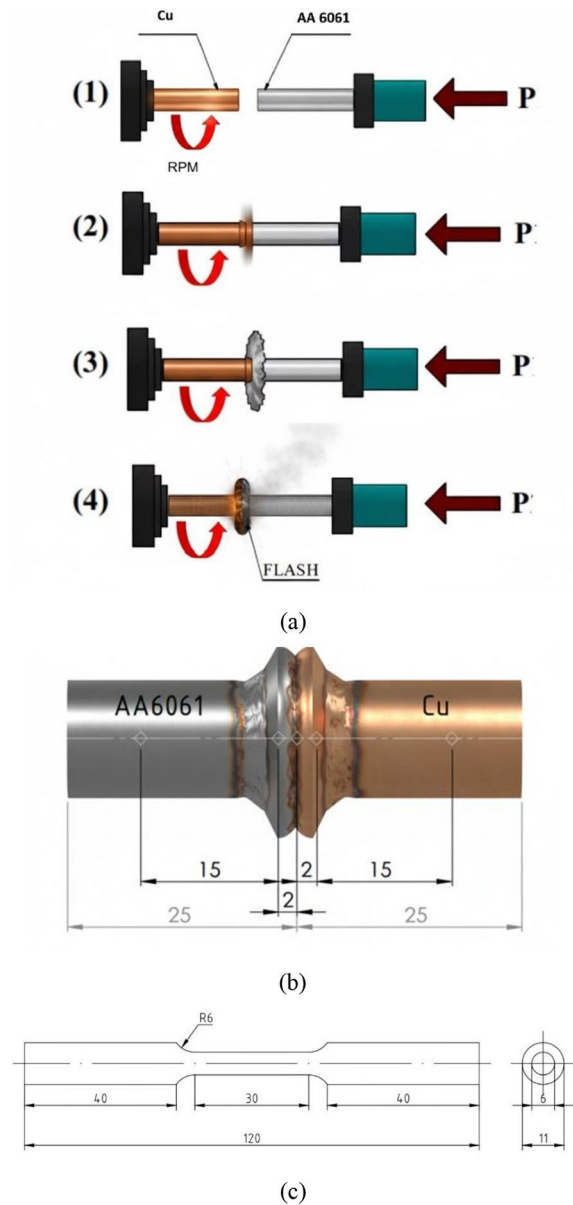


Figure 1. Experimental procedures and specimen geometries used in AA6061–Cu rotary friction welding:
(a) welding process schematic,
(b) hardness test positions, and
(c) tensile specimen dimensions (unit: mm)

Table 1. Chemical composition of aluminium 6061 and pure copper [27]

Material	%Al	%Cr	%Cu	%Fe	%Mg	%Mn	%Si	%Ti	%Zn
AA6061	97.40	0.125	0.206	0.0257	1.070	0.079	0.69	0.15	0.014
Cu	-	-	100	-	-	-	-	-	-

Table 2. Initial mechanical properties of the base materials

Material	Initial condition	Vickers hardness (HV0.3)	Typical UTS (MPa)
AA6061	Commercial rod	96 ± 3	260–310
Cu	Commercial rod	58 ± 2	210–240

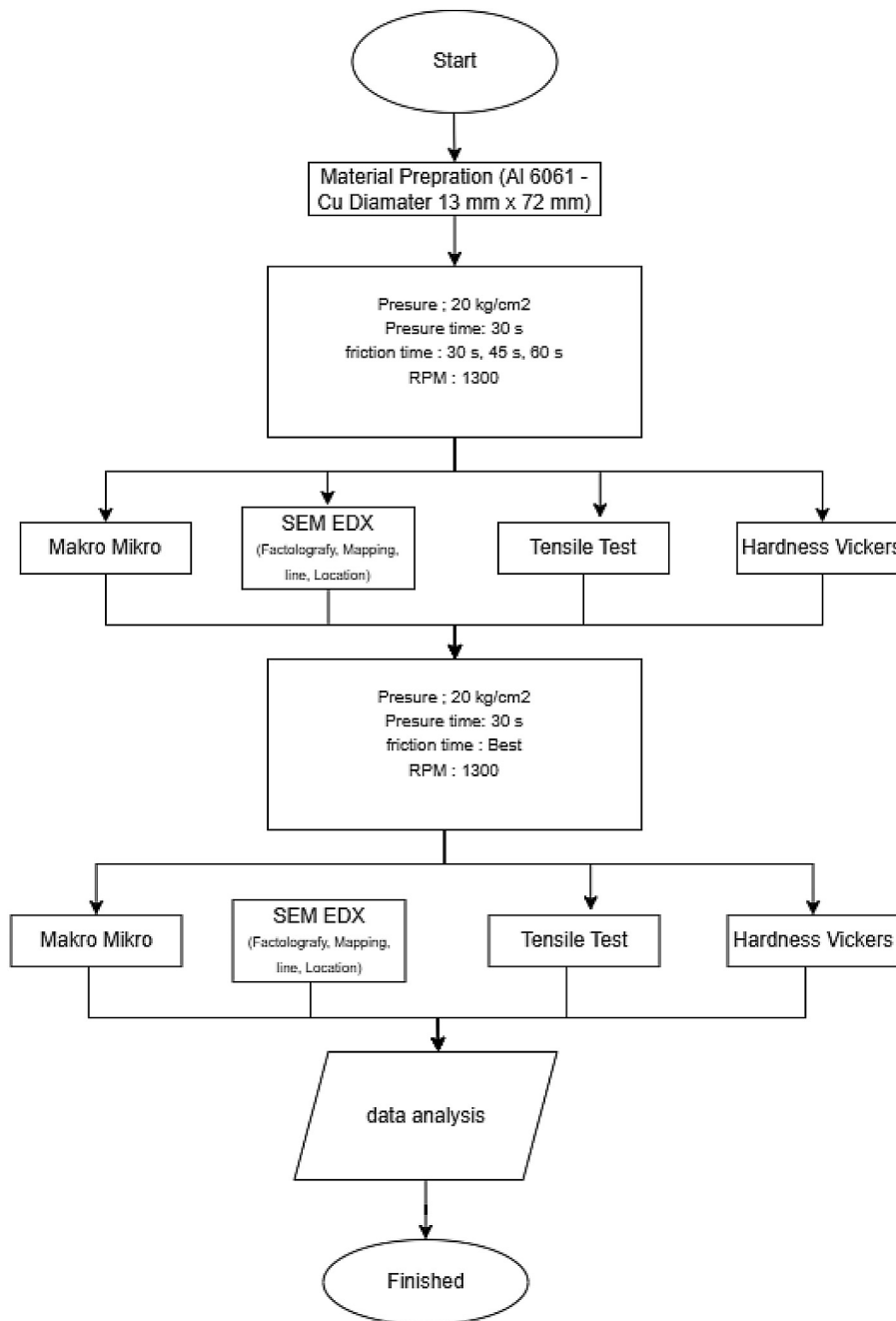


Figure 2. Experimental workflow

with dimensions of 13 mm diameter and 72 mm length. The specimen surfaces were ground and cleaned using acetone to remove oxide layers and contaminants prior to welding. The copper specimen was mounted on the rotating chuck, while the AA6061 specimen was fixed on the stationary side. Rotary friction welding was then performed under a constant friction pressure of 1.96 MPa and upsetting time of 30 s. The friction time was varied between 30–60 s and the rotational speed between 1000–2000 rpm. During the friction stage, interfacial heat was generated through

severe plastic deformation and sliding friction, promoting thermo-mechanical mixing and diffusion bonding. Subsequently, the upsetting stage consolidated the joint and formed the flash region. After welding, the specimens were sectioned and characterized using optical microscopy, SEM–EDX, micro-Vickers hardness testing, and tensile testing to evaluate the CWZ, TMAZ, HAZ, IMC formation, and mechanical performance of the joints.

The present study employed a sequential one-factor-at-a-time (OFAT) experimental

Table 3. Process parameters and variations in rotary friction welding

Specimens	Friction pressure (kg/cm ²)	Pressure time (s)	Rotational speed (rpm)	Friction time (s)
1	20	30	1300	30
2	20	30	1300	45
3	20	30	1300	60
4	20	30	1000	60
5	20	30	1300	60
6	20	30	2000	60

design rather than a full factorial design. Friction time was first varied while maintaining a constant rotational speed of 1300 rpm. Subsequently, the optimum friction time was fixed (60 s), and rotational speed was varied. Therefore, the experimental design was intended to isolate the individual influence of each process parameter rather than evaluate their statistical interaction. The identical parameters used for Specimen 3 and Specimen 5 were intentionally applied as experimental replication to evaluate repeatability and consistency of the thermo-mechanical response under identical welding conditions. Replication allows verification of IMC growth stability and mechanical performance reliability, ensuring that the observed trends are not caused by random experimental variation.

The axial pressure (1.96 MPa) and upsetting time (30 s) were kept constant based on previous optimization results in order to isolate the thermo-diffusion effects of friction time and rotational speed on interfacial bonding, IMC formation, and mechanical performance of the welded joints.

Weld joint characterizations

The physical and mechanical properties of the aluminium–copper friction welds were evaluated through several tests. The physical characteristics were analyzed using macro- and microstructural observations of the weld cross-section, while the mechanical properties were assessed based on microhardness distribution and tensile strength. Macro- and microstructural examinations were performed using a Boeco BM-180/I/PL optical microscope with magnifications ranging from 40X to 1000X. The specimens were prepared in accordance with ASTM E407-99, Standard Practice for Microetching Metals and Alloys, through grinding, polishing, and etching processes. Keller’s reagent (5 mL HNO₃, 2 mL

HF, 3 mL HCl, and 190 mL H₂O) was used for the AA6061 side, while HNO₃ was used for the pure copper side. The optical microscope, operating with visible light, was employed to observe weld morphology and grain boundaries.

Grain size measurements were performed using ImageJ software based on optical micrographs obtained from the weld interface region. The average grain size was determined from representative microstructural areas to evaluate the effect of friction time and rotational speed on grain evolution. Additionally, the microstructure and fracture surfaces of the Al–Cu friction welds were analyzed using scanning electron microscopy (SEM) and energy-dispersive x-ray spectroscopy (EDX). SEM provides high-resolution imaging for detailed examination of microstructural features, crack propagation, and intermetallic phase formation at a finer scale [28]. EDX, commonly coupled with SEM, was used to identify alloying elements and intermetallic phases in specific microstructural regions. All SEM–EDX analyses were performed using an FEI INSPECT F50 system.

The micro-Vickers hardness test was conducted using an EW-105 Series microhardness tester manufactured by Everone, Shanghai, China. The test followed the ASTM E92 standard, applying a 300 gf load with a dwell time of 10 seconds. Measurements were taken evenly across three key regions of the friction weld: the CWZ, thermomechanically affected zone (TMAZ), and HAZ. The measurement locations are illustrated in Figure 1 (c). Tensile strength testing of the aluminium–copper friction welds was performed according to ASTM E8/E8M – Standard Test Methods for Tension Testing of Metallic Materials, as shown in Figure 1 (d). The tests were carried out using a Tinius Olsen TO-5000 Series universal testing machine with a maximum load capacity of 5 kN and a crosshead speed range of 0.001–1000 mm/min.

RESULTS AND DISCUSSION

Analysis of friction time

Macro and microstructure

Figure 3 shows the thickest TMAZ was observed at 30 seconds, followed by a gradual reduction in width at 45 and 60 seconds. The reduction in TMAZ width observed at longer friction times (45–60 s) suggests a gradual transition from deformation-dominated bonding toward diffusion-assisted metallurgical bonding. This interpretation is supported by the formation of a thicker and more continuous IMC layer, SEM–EDX compositional distribution, and the improvement in hardness and tensile strength at extended friction durations. Extended friction duration increases cumulative heat input, promoting dynamic recrystallization and stabilization of interfacial diffusion within the central weld zone (CWZ). Under near steady-state thermo-mechanical conditions, plastic deformation becomes more localized at the interface rather than spreading across the TMAZ. Therefore, microstructural evolution shifts from plastic flow expansion toward diffusion-controlled IMC growth. Similar behavior has been reported in extended-duration friction welding where diffusion kinetics dominate interfacial bonding mechanisms [29]. The narrowing of the TMAZ indicates that plastic deformation becomes increasingly localized at the interface as diffusion-controlled bonding becomes dominant. Under extended friction duration, thermo-mechanical equilibrium stabilizes the CWZ, reducing the extent of deformation spreading into the surrounding TMAZ region. Quantitative measurements of the TMAZ width were performed using

Image software based on optical micrographs. The average TMAZ width decreased from approximately 1450 μm at 30 s to 830 μm at 60 s, indicating increased localization of plastic deformation near the weld interface. This reduction in TMAZ width corresponded with higher CWZ hardness and improved tensile strength, confirming that diffusion-controlled bonding became more dominant at longer friction durations. All specimens exhibited flash formation around the joint, caused by uneven temperature distribution under axial pressure and rotation [30, 31]. The flash appeared narrower at 60 seconds. Due to the difference in thermal conductivity, more flash formed on the aluminium side, which softens more easily than copper.

The microstructure of the base metals (BM) for aluminium and copper is shown in Figures 4(a) and 4(b). The aluminium consists of two main phases: the primary $\alpha\text{-Al}$ phase and the secondary $\beta\text{-Mg}_2\text{Si}$ phase, while the copper exhibits coarse equiaxed grains with random orientation. Figure 5 presents the microstructure of the aluminium–copper friction weld joints at different friction times. The 30-second specimen (Figure 5(a)) shows a weak bond with unbonded areas due to insufficient heat and pressure, and the IMC layer is barely visible [32]. At 45 seconds (Figure 5(b)), a thin IMC layer begins to form along with small cracks, indicating the initial stage of metallurgical bonding. The 60-second specimen (Figure 5(c)) exhibits a defect-free interface with a thicker and more uniform IMC layer, suggesting enhanced atomic diffusion and improved metallurgical bonding. However, excessive friction duration can lead to overheating and unstable material flow, potentially generating microdefects such as voids and cracks [33].

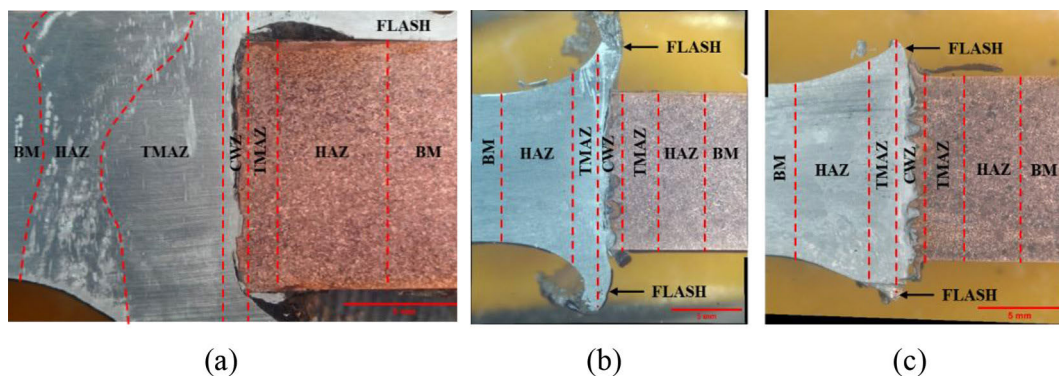


Figure 3. Macrostructure of aluminium–copper friction weld joints at friction times of (a) 30 seconds, (b) 45 seconds, and (c) 60 seconds

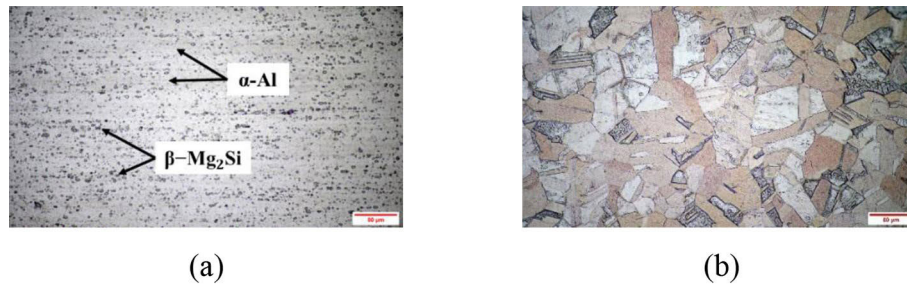


Figure 4. Microstructure of the base metals: (a) AA6061 aluminium and (b) pure copper

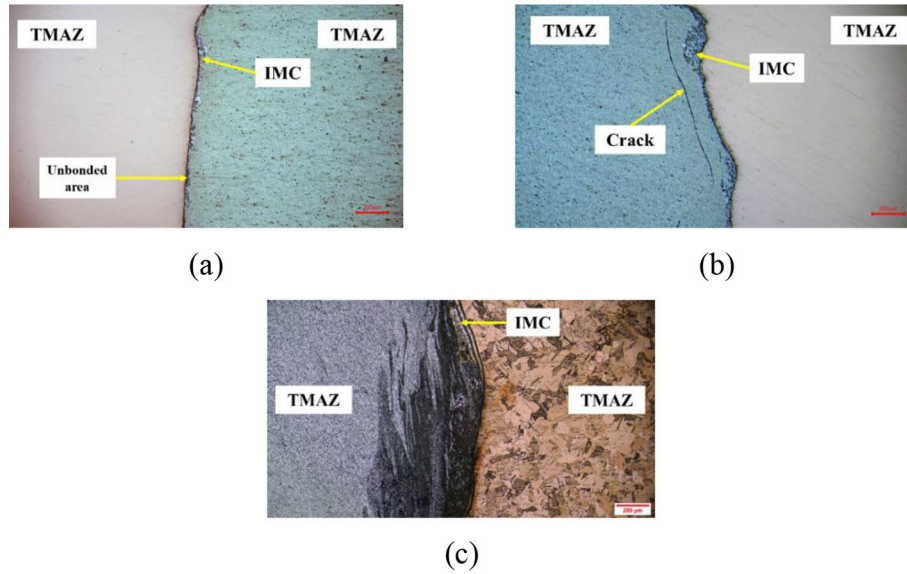


Figure 5. Microstructure of aluminium–copper friction weld joints at different friction times: (a) 30 seconds, (b) 45 seconds, and (c) 60 seconds

The diffusion coefficient (D), representing the rate of atomic diffusion, was calculated using Equation (1) and is presented in Table 4. The exponential dependence of the diffusion coefficient on temperature confirms that IMC growth follows a thermally activated Arrhenius-type mechanism. The results indicate that increasing the friction time and temperature leads to higher D values, accelerating atomic diffusion and promoting the formation of a thicker intermetallic compound (IMC) layer. This explains why the specimen with a 60-second friction time exhibited the thickest IMC layer, as observed in the microstructural analysis.

$$D = D_0 \exp(-Q/RT) \quad (1)$$

Based on the diffusion rates presented in Table 4, the IMC layer thickness of each specimen was measured using Image software. The results in Table 5 show that the IMC thickness, from greatest to smallest, corresponds to friction times of 60, 45,

and 30 seconds. Increasing the friction time enhances heat input and plastic deformation, which accelerates atomic diffusion and IMC growth [20, 33, 35]. However, excessively long friction times may lead to the formation of a brittle IMC layer [26].

Figure 6 shows the effect of friction time on grain size in AA6061–Cu friction weld joints. Based on the measurement results, the average grain size increased from $4.26 \pm x \mu\text{m}$ at 30 s to $7.44 \pm y \mu\text{m}$ at 45 s, and subsequently decreased to $5.71 \pm z \mu\text{m}$ at 60 s, indicating the occurrence of grain coarsening followed by dynamic recrystallization under prolonged friction time. The increase in grain size at a friction time of 45 seconds indicates that heat accumulation during the welding process has driven grain growth due to increased grain boundary mobility. Under these conditions, the heat generated is high enough to enlarge the grain size, but has not yet reached a stable condition that supports optimal recrystallization. Conversely, at a friction time of 60 seconds, there is

Table 4. Interdiffusion parameters of Al and Cu at different friction times

Condition	Diffusion constant ($\times 10^{-4} \text{ m}^2\text{s}^{-1}$) [34]	Activation energy (kJ/mol) [34]	Specimens	Temperature (K)	Diffusion coefficient (m^2s^{-1})
Al in Cu (D_{Al}^{Cu})	0.131	185.2	1	498	4.910×10^{-25}
			2	499	5.371×10^{-25}
			3	509	1.291×10^{-24}
Cu in Al (D_{Cu}^{Al})	0.654	1.41	1	498	4.652×10^{-24}
			2	499	4.655×10^{-24}
			3	509	4.686×10^{-24}

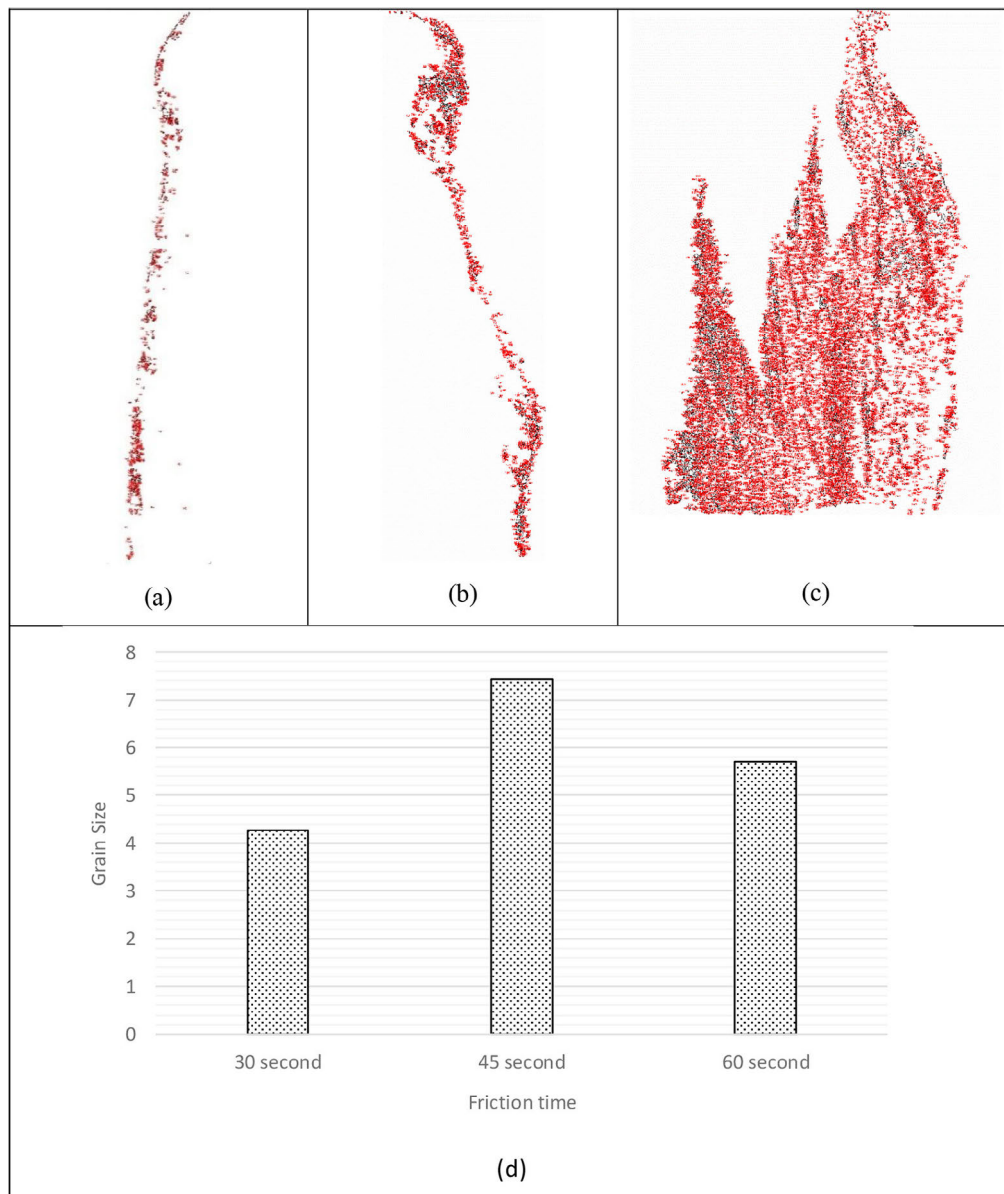


Figure 6. Grain size analysis of aluminium–copper friction weld joints at different friction times: (a) 30 seconds, (b) 45 seconds, (c) 60 seconds, (d) Grain size distribution

a decrease in grain size compared to 45 seconds. This phenomenon indicates that longer plastic deformation and more intensive atomic diffusion

have triggered the dynamic recrystallization process, resulting in the formation of new, finer, and more uniform grains around the joint interface.

The improved tensile strength of the 60 s specimen is associated with a more uniform metallurgical interface and a more continuous IMC layer, indicating improved interfacial bonding. The present results do not imply that increasing IMC thickness alone enhances joint strength, since excessive IMC growth is generally detrimental because of its brittle nature.

The relatively high standard deviation observed at the friction time of 45 s indicates that the joint was in a transitional thermo-mechanical condition between deformation-dominated bonding and diffusion-assisted metallurgical bonding. Under this intermediate condition, heat distribution and plastic deformation were not yet fully stabilized, resulting in local variations in IMC growth along the interface. To reduce experimental uncertainty, IMC thickness measurements were conducted at five different locations using ImageJ software, and average values with standard deviations were calculated. In addition, experimental replication under identical welding parameters was performed to evaluate process reproducibility and thermo-mechanical consistency.

The Table 6 ANOVA results show that the p-value (0.011) is lower than the significance level of 0.05, indicating that friction time has a statistically significant effect on IMC thickness. Therefore, the variation in IMC thickness among 30 s, 45 s, and 60 s conditions is not caused by random measurement error but is influenced by thermo-diffusion behavior during the welding process. The relatively large standard deviation

observed at 45 s represents a transition phase from mechanical interlocking to diffusion-dominated metallurgical bonding, resulting in non-uniform IMC growth.

The IMC layer thickness was measured using ImageJ software at five different locations along the weld interface in order to reduce local measurement bias. Statistical analysis including mean and standard deviation was performed to evaluate measurement consistency. The relatively large variation observed at a friction time of 45 s indicates a transition condition from mechanical interlocking to diffusion-controlled metallurgical bonding. At this stage, thermo-mechanical equilibrium has not yet been fully achieved, resulting in local variations in heat input and plastic deformation. One-way ANOVA statistical testing (95% confidence level) was applied to evaluate the significance of differences in IMC thickness between friction time variations. Error bars representing standard deviation were included to illustrate measurement dispersion and improve data reliability.

IMC thickness values presented in Table 5 represent the average of five measurements taken at different interface locations to reduce local measurement bias. The formation of intermetallic compounds (IMCs) in the CWZ was analyzed using SEM-EDX to identify the elemental composition and phase types at the joint interface [36]. Each specimen exhibited three distinct IMC phases, as illustrated in Figures 7–9. At a friction time of 30 seconds, the CWZ was not

Table 5. IMC layer thickness of specimens at different friction times

Length measurement	Specimens (μm)		
	30 seconds	45 seconds	60 seconds
1	44.55	180.408	167.825
2	18.361	87.336	153.045
3	73.362	67.249	132.176
4	25.343	82.974	135.651
5	25.775	239.303	95.208
Average	43.678	131.454	136.781

Table 6. One-way ANOVA results for IMC thickness at different friction times

Source of variation	SS	df	MS	F	p-value
Between groups	29269.53	2	14634.77	6.72	0.011
Within groups	26132.34	12	2177.69		
Total	55401.87	14			

fully developed, showing a thin IMC layer (Figure 7(a)), while specimens welded for 45 and 60 seconds displayed thicker and more uniform IMC layers (Figures 8(a)–9(a)). Based on SEM–EDX compositional analysis and comparison with the Al–Cu binary phase diagram reported in the literature, the intermetallic regions in the CWZ likely correspond to $\theta(\text{Al}_2\text{Cu})$, $\eta(\text{AlCu})$, and $\zeta(\text{Al}_3\text{Cu}_4)$ phases. It should be noted that the phase identification in this study was based primarily on SEM–EDX compositional analysis and literature comparison. Direct crystallographic validation using XRD or TEM–SAED was not performed; therefore, the reported IMC

phases should be interpreted as probable phase formations rather than definitive crystallographic identification. The formation of Cu-rich intermetallic phases is associated with asymmetric diffusion behavior, where Cu exhibits higher diffusivity in Al compared to Al diffusion in Cu. This concentration gradient promotes localized enrichment of Cu atoms, stabilizing θ -phase formation under extended thermo-mechanical interaction. Point A contained Cu concentrations of 18.51%, 25.04%, and 33.82%, corresponding to the hypereutectic $\alpha(\text{Al}) + \theta(\text{Al}_2\text{Cu})$ phase with an atomic ratio of $n(\text{Al}):n(\text{Cu}) \approx 2:1$, consistent with the θ -phase characteristics [37–40]. Point B

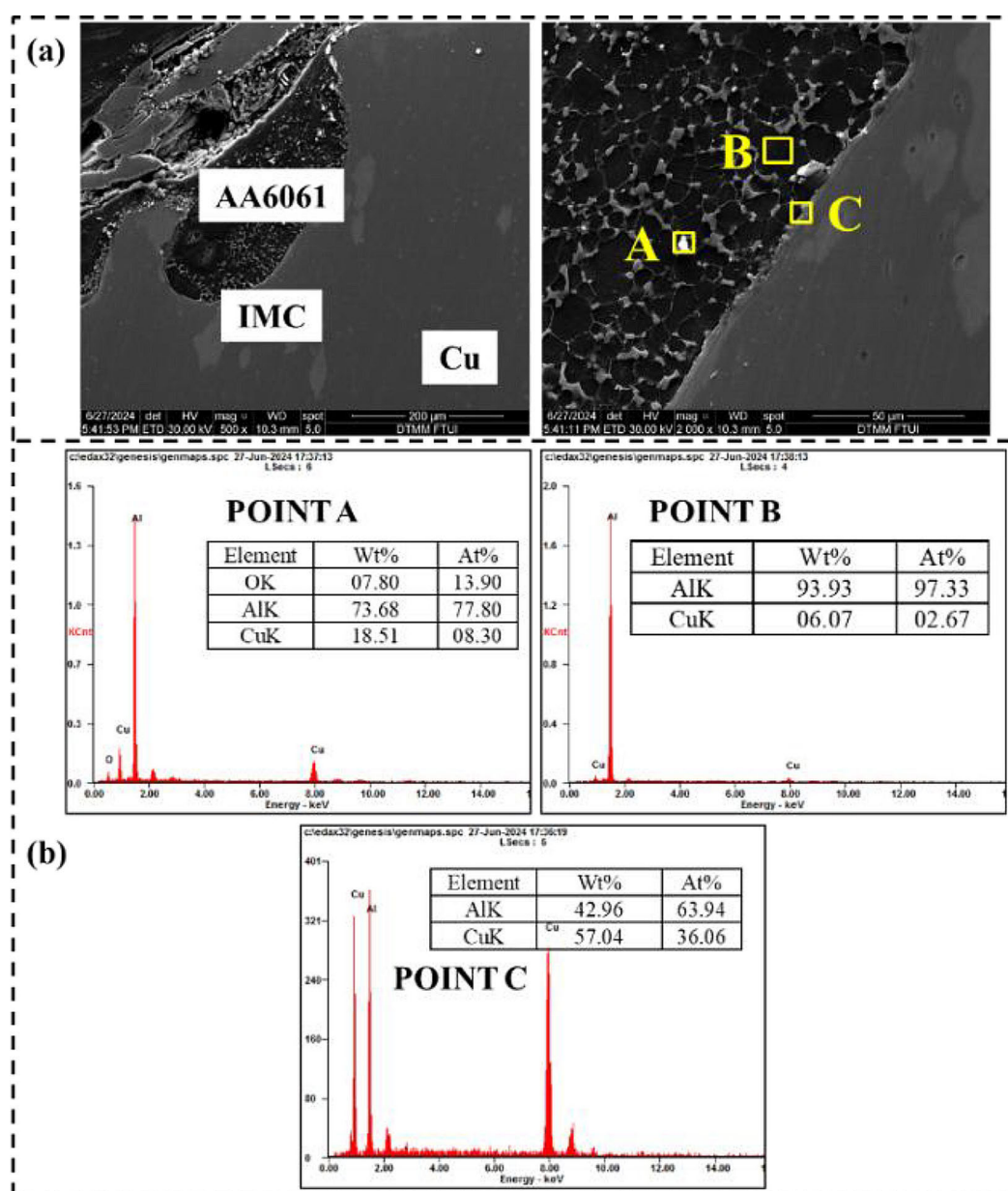


Figure 7. (a) SEM image and (b) EDX mapping of the aluminium–copper friction weld interface at a friction time of 30 seconds

contained 6.07%, 6.29%, and 6.04% Cu, indicating a hypoeutectic $\alpha(\text{Al})$ -dominated phase [37–39]. Meanwhile, point C showed Cu contents of 57.04%, 78.47%, and 61.44%, representing $\eta_1 (\text{AlCu} (\text{ht})) + \eta_2 (\text{AlCu} (\text{rt}))$ phases in the 30 s and 60 s specimens, and $\zeta_1 (\text{Al}_3\text{Cu}_4 (\text{ht})) + \zeta_2 (\text{Al}_3\text{Cu}_4 (\text{rt}))$ phases in the 45 s specimen [41, 32, 35, 38, 39].

Vickers microhardness

Figure 10 shows the Vickers microhardness distribution of the aluminium–copper friction

welds at different friction times. The average hardness values for friction times of 30, 45, and 60 seconds were 57.02 HV, 57.76 HV, and 57.66 HV, respectively. Increasing the friction time enhances frictional heat and atomic diffusion between the metals, leading to a thicker and more continuous intermetallic compound (IMC) layer [20]. The highest hardness values were observed in the TMAZ due to plastic deformation, with the copper side exhibiting greater hardness because of its superior thermal resistance. The TMAZ hardness values of 55.1 HV, 55.6 HV, and 58.1 HV correspond to

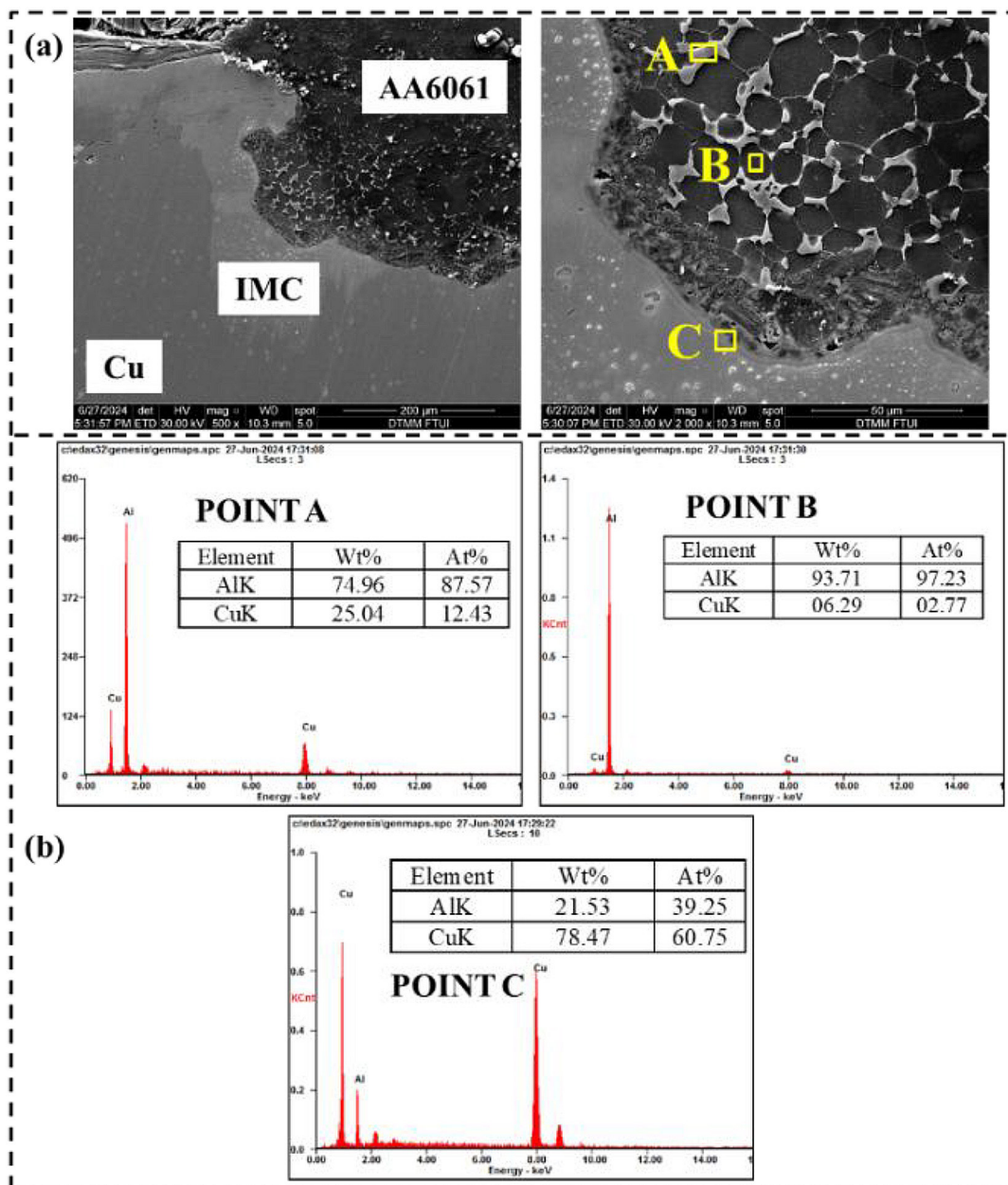


Figure 8. (a) SEM image and (b) EDX mapping of the aluminium–copper friction weld interface at a friction time of 45 seconds

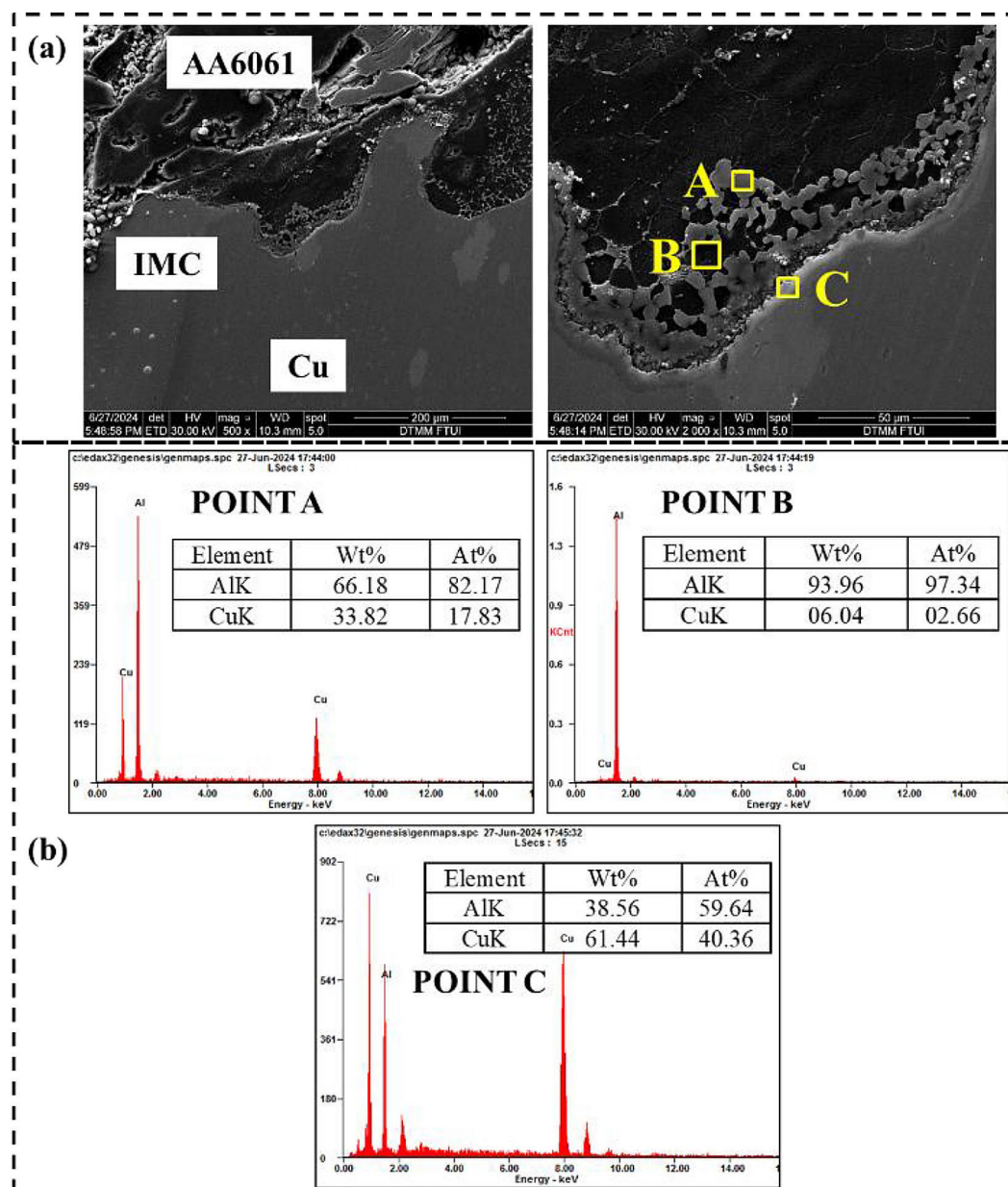


Figure 9. (a) SEM image and (b) EDX mapping of the aluminium–copper friction weld interface at a friction time of 60 seconds

the increased IMC thickness and atomic diffusion [20, 26]. Conversely, the region adjacent to the TMAZ exhibited only minor variations in hardness. Therefore, the present results do not provide sufficient evidence to identify a distinct HAZ based solely on hardness measurements – approximately 57.3 HV on the aluminium side and 54.5 HV on the copper side – caused by thermal softening and dissolution of Mg_2Si precipitates [42]. Therefore, a friction time of 60 seconds provides the most balanced heat input and material mixing, resulting in optimal hardness across the weld zones [43].

Tensile strength

Figure 11 presents the stress–strain curves of aluminium–copper friction welds at different friction times, while Figure 12 compares their ultimate tensile strength and yield strength. The tensile strengths for friction times of 30, 45, and 60 seconds were 19.15 MPa, 26.85 MPa, and 28.44 MPa, respectively. The significant reduction compared to the base metals [28, 44, 22] is attributed to the presence of the HAZ and the formation of intermetallic compounds (IMCs) [22]. At 30 and 45 seconds, the IMC layer was incomplete and exhibited defects such as cracks

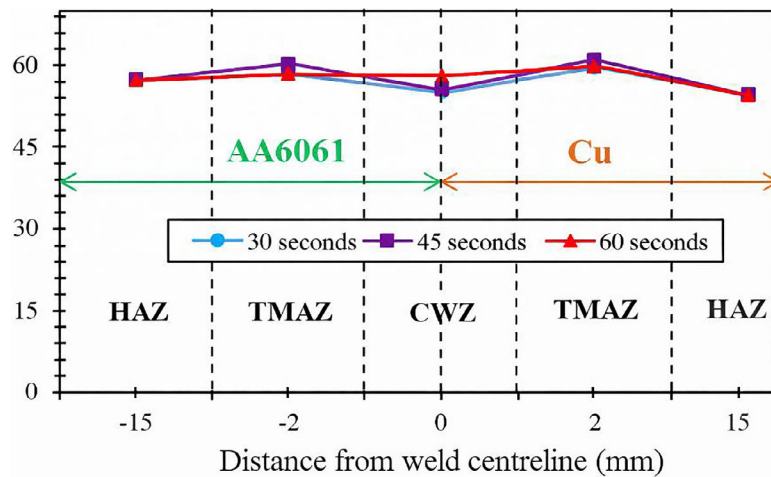


Figure 10. Vickers microhardness distribution of aluminium–copper friction welds at different friction times

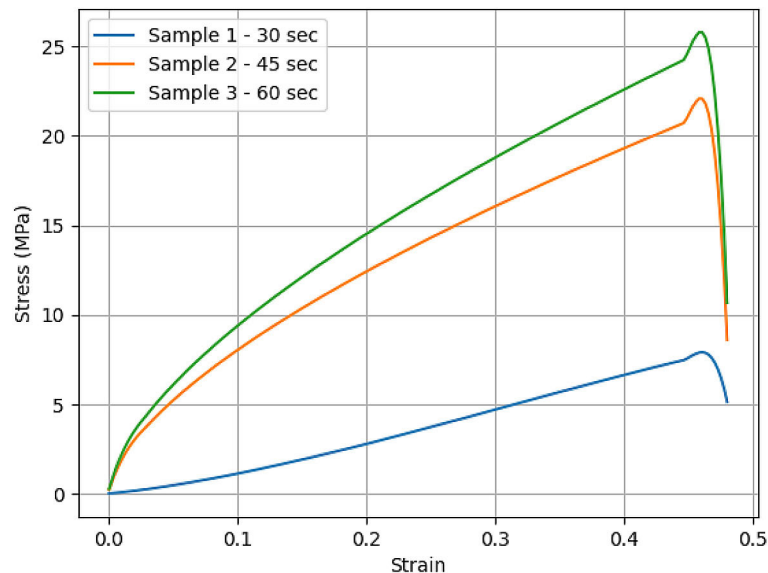


Figure 11. Stress–strain curves of aluminium–copper friction welds at different friction times

and unbonded regions, leading to lower tensile strength. In contrast, the 60-second specimen displayed a more homogeneous and defect-free IMC layer, resulting in the highest tensile strength. A longer friction time allows for a gradual temperature rise toward thermal equilibrium, improving weld integrity [33]. Therefore, a friction time of 60 seconds is recommended as the optimal condition for aluminium–copper friction welding.

The stress–strain curves presented in this study were used to compare the tensile response of the welded joints under different welding parameters. The discussion focuses on ultimate tensile strength, yield strength, and fracture behavior. The initial slope of the stress–strain curves was not used to determine the intrinsic

Young's modulus of the welded joints; therefore, no comparison of elastic modulus is made in this study. Minor fluctuations observed in the stress–strain curves are associated with localized brittle fracture events within the IMC layer. The presence of $\theta(\text{Al}_2\text{Cu})$ phase reduces plastic strain continuity, causing temporary stress relaxation during crack propagation.

Figures 13–15 illustrate the fracture morphology of the tensile-tested aluminium–copper friction welds at different friction times. All specimens fractured within the CWZ and predominantly exhibited brittle fracture features characterized by cleavage patterns [45], [46]. The 30 s and 45 s specimens displayed both sticking and faceted surfaces, whereas the 60 s

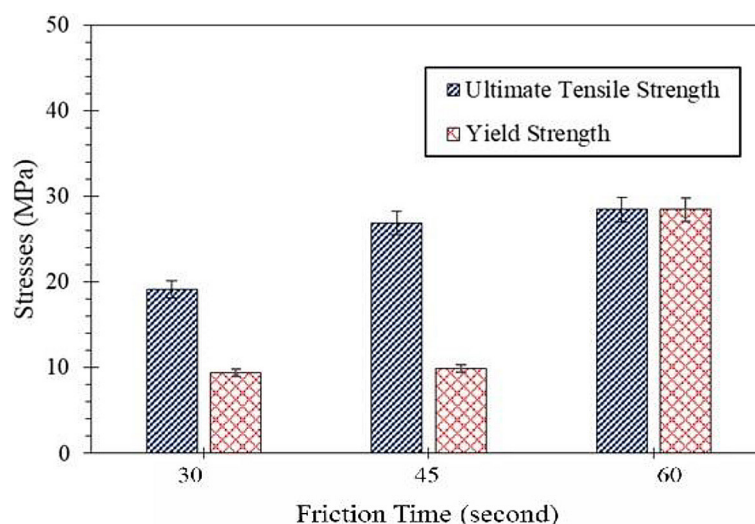


Figure 12. Tensile strength of aluminium–copper friction welds at different friction times

specimen showed only a faceted surface—indicating a more complete metallurgical bond resulting from enhanced material diffusion at longer friction durations [47]. EDX analysis of the 30 s and 45 s specimens (Figures 12–13) revealed a ductile $\alpha(\text{Al})$ phase at point A, a brittle Cu-rich ($>53\%$) intermetallic $\theta(\text{Al}_2\text{Cu})$ phase at point B, and a mixed $\alpha(\text{Al})+\theta(\text{Al}_2\text{Cu})$ phase with transitional ductile–brittle behavior at point C. Meanwhile, the 60 s specimen (Figure 16) exhibited only the $\alpha(\text{Al})+\theta(\text{Al}_2\text{Cu})$ phase, confirming optimal diffusion and the formation of a well-bonded but not excessively thick IMC layer. Overall, the balanced phase distribution and strong metallurgical bonding observed at a friction time of 60

seconds indicate that this parameter provides the most optimal welding condition. SEM observations suggest that crack initiation predominantly occurred within the IMC layer located in the CWZ, particularly in regions enriched with brittle $\theta(\text{Al}_2\text{Cu})$ and $\eta(\text{AlCu})$ phases. Although the 60 s friction time improved metallurgical bonding through the formation of a more continuous IMC layer, the fracture behavior remained dominated by brittle cleavage mechanisms because of the limited plastic deformability of the intermetallic phases. Crack propagation tended to follow the IMC interface due to local stress concentration and mechanical incompatibility between Al, Cu, and the intermetallic compounds.

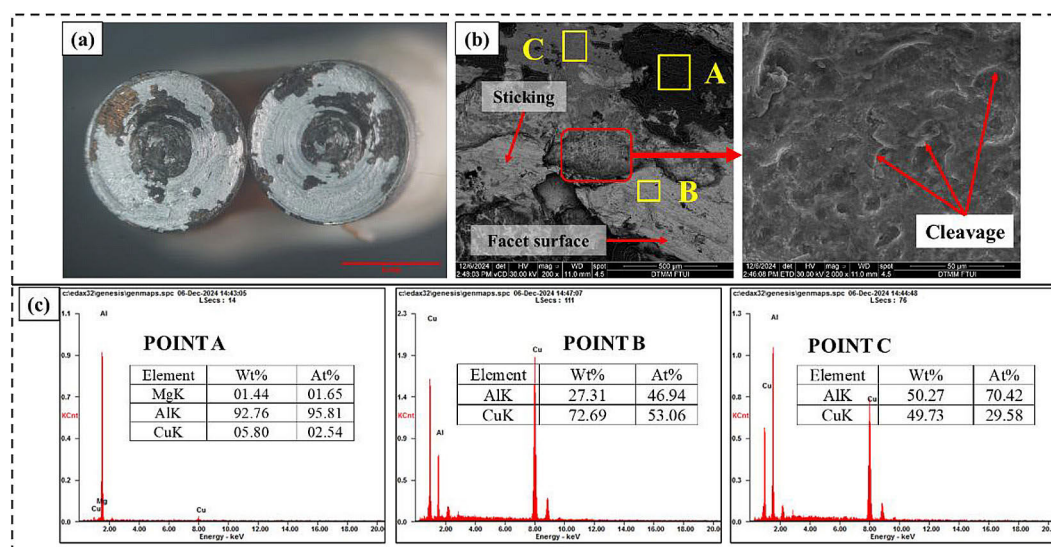


Figure 13. (a) Macrostructure of the specimen welded at a friction time of 30 seconds after the tensile test, (b) SEM image of the fracture surface, and (c) EDX analysis of the fracture surface

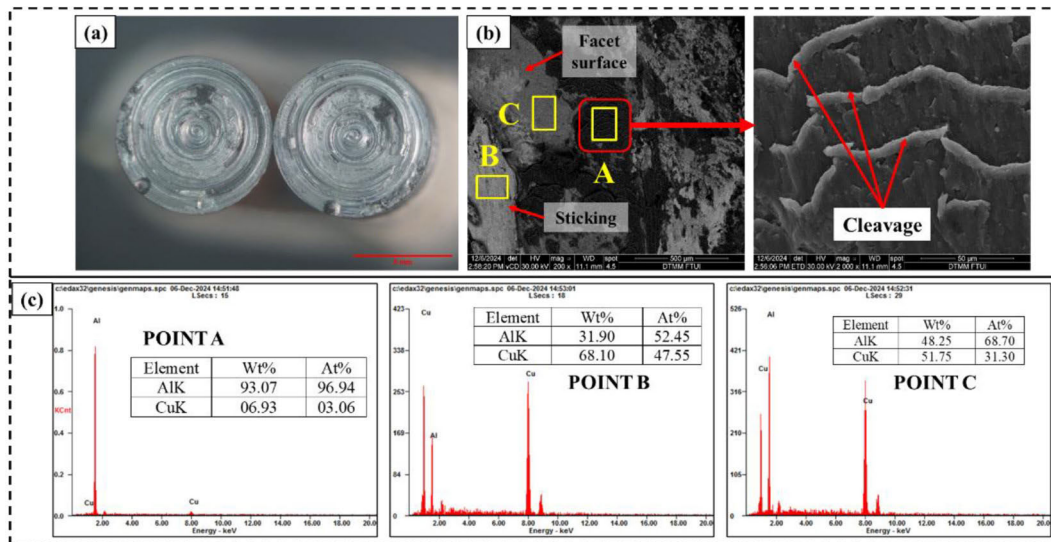


Figure 14. (a) Macrostructure of the specimen welded at a friction time of 45 seconds after the tensile test, (b) SEM image of the fracture surface, and (c) EDX analysis of the fracture surface

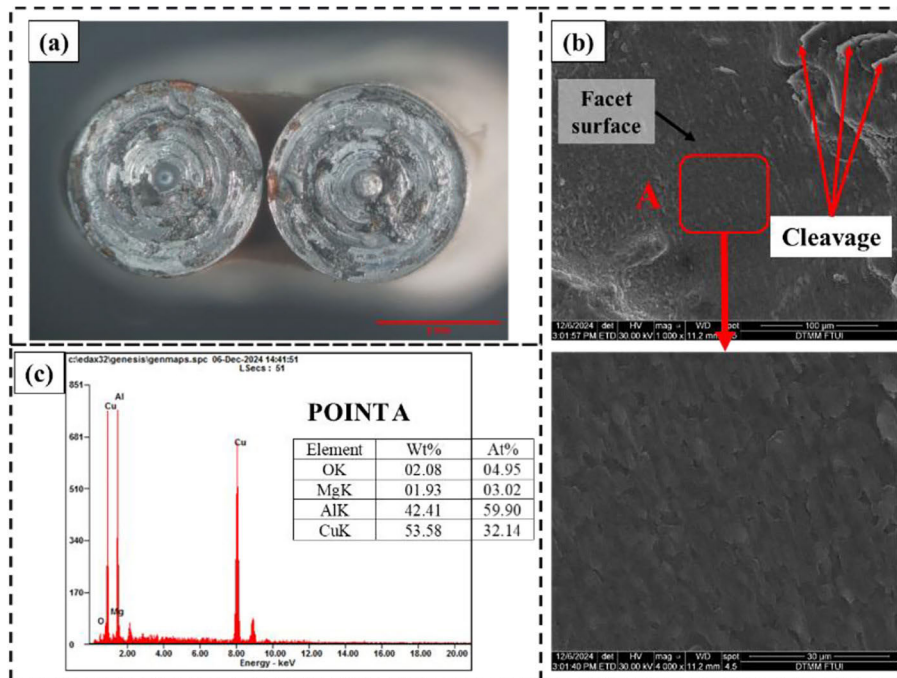


Figure 15. (a) Macrostructure of the specimen welded at a friction time of 60 seconds after the tensile test, (b) SEM image of the fracture surface, and (c) EDX analysis of the fracture surface

The results obtained from friction time variation indicate that thermo-mechanical interaction and diffusion-assisted bonding strongly influence IMC growth and joint integrity. However, friction time alone does not fully determine the weld behavior because rotational speed also governs heat generation rate, plastic flow intensity, and thermal stability during rotary friction welding. Therefore, the following section further discusses the coupled effect of rotational speed on

microstructural evolution, diffusion kinetics, and mechanical performance of AA6061–Cu joints.

Analysis of rotational speed

Macro and microstructure

Compared with friction time, rotational speed more directly controls instantaneous heat generation and frictional energy input, thereby affecting

diffusion kinetics and IMC growth behavior differently. Consequently, the combined interaction between friction time and rotational speed determines the thermo-mechanical equilibrium and final weld quality. Figure 16 shows the macrostructure of aluminium–copper friction welds at different rotational speeds. As illustrated, increasing the rotational speed broadens the TMAZ and narrows the HAZ. Higher rotational speeds are generally associated with increased frictional heat generation and greater plastic deformation during rotary friction welding, as suggested by the observed flash formation, IMC growth, and microstructural evolution [48]. It should be noted that direct temperature measurement and thermo-mechanical simulation were not conducted in this study. Therefore, the thermal behavior discussion is based primarily on indirect experimental observations and previously reported thermo-mechanical friction welding theory. However, the rapid temperature rise restricts heat dissipation, leading to a narrower HAZ [49]. All specimens exhibited flash formation around the joint due to uneven heat and pressure distribution during welding [30, 31]. The flash becomes more pronounced at higher rotational speeds, as excessive heat drives more material out of the weld zone [50, 49].

Figure 17 illustrates the effect of rotational speed on the microstructure. As the rotational speed increases, grain coarsening becomes more pronounced due to rapid grain growth and increased precipitation caused by higher heat input and slower cooling rates [49]. All specimens still exhibit unbonded areas resulting from uneven heat distribution and axial pressure [32], however, the presence of intermetallic compound (IMC) layers in each specimen confirms

the formation of strong metallurgical bonding. In general, the formation of IMCs and microdefects is a typical phenomenon in friction welding, driven by the combined effects of frictional heat and plastic deformation. Because aluminium has a lower melting point and is softer than copper, it undergoes plastic deformation earlier in the CWZ, leading to greater material flow and more flash formation along the joint edges [20].

According to Mattie et al. [24], the rotational speed does not directly affect the overall welding process but has a significant influence on the degree of plastic deformation. Greater deformation enhances the atomic diffusion rate, thereby promoting the formation of intermetallic compounds (IMCs). As shown in Table 7, a rotational speed of 1300 rpm yields the highest diffusion coefficient (D), indicating the most intensive atomic diffusion under this condition.

Figure 18 shows the effect of rotational speed on the grain size of AA6061–Cu friction welded joints. The average grain sizes obtained were 6.291 μm at 1000 rpm, 9.544 μm at 1300 rpm, and 7.707 μm at 2000 rpm. The increase in grain size to a maximum value at 1300 rpm indicates that the rotational speed produces sufficient heat input to significantly promote grain growth. Higher temperatures increase atomic diffusivity and grain boundary mobility, resulting in more intensive grain enlargement compared to the 1000 rpm condition. Although the largest grain size was found at 1300 rpm, this condition actually produced the best mechanical properties, including the highest tensile strength and the most optimal hardness distribution. This indicates that joint performance is not only influenced by grain size, but also by the formation of a more

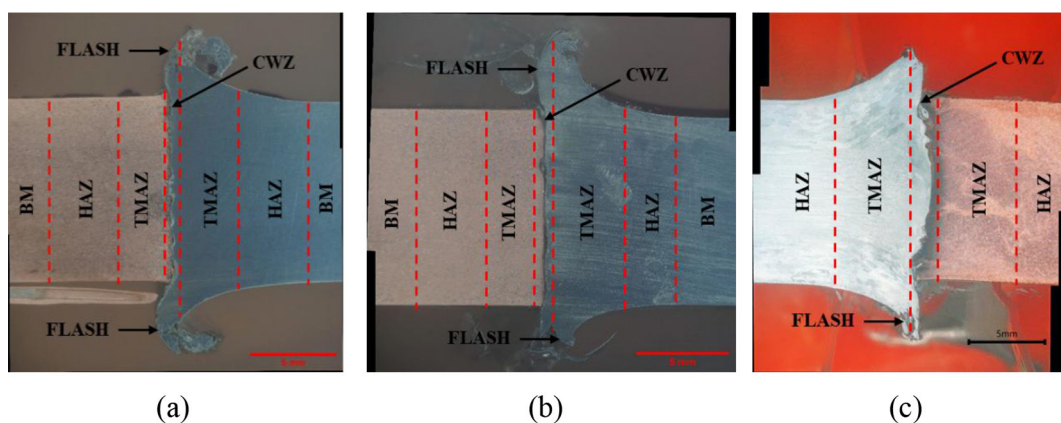


Figure 16. Macrostructure of aluminium–copper friction weld joints at rotational speeds of (a) 1000 rpm, (b) 1300 rpm, and (c) 2000 rpm

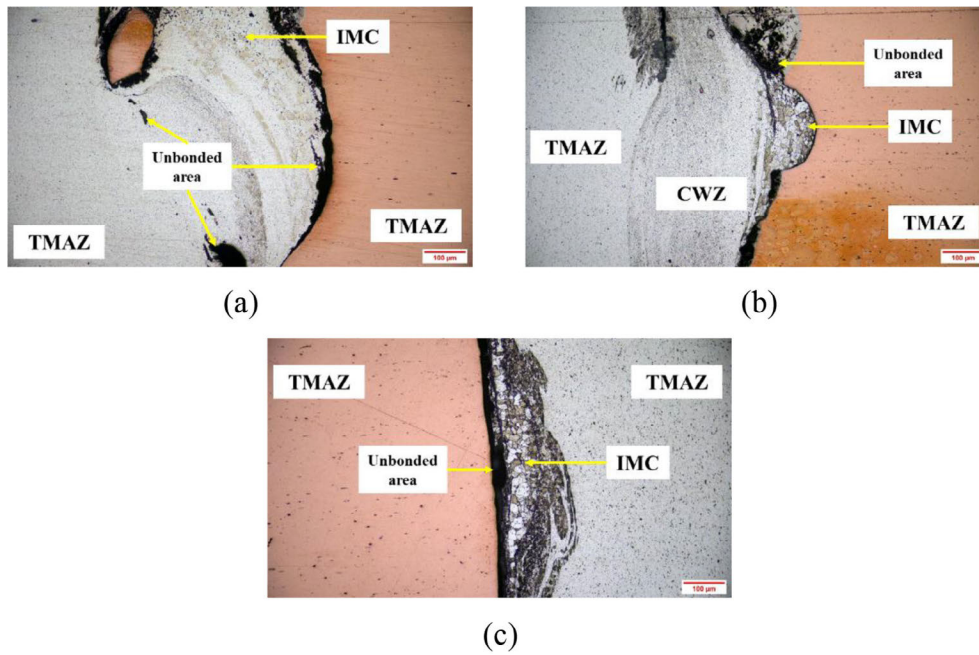


Figure 17. Microstructure of aluminium–copper friction welds at different rotational speeds: (a) 1000 rpm, (b) 1300 rpm, and (c) 2000 rpm

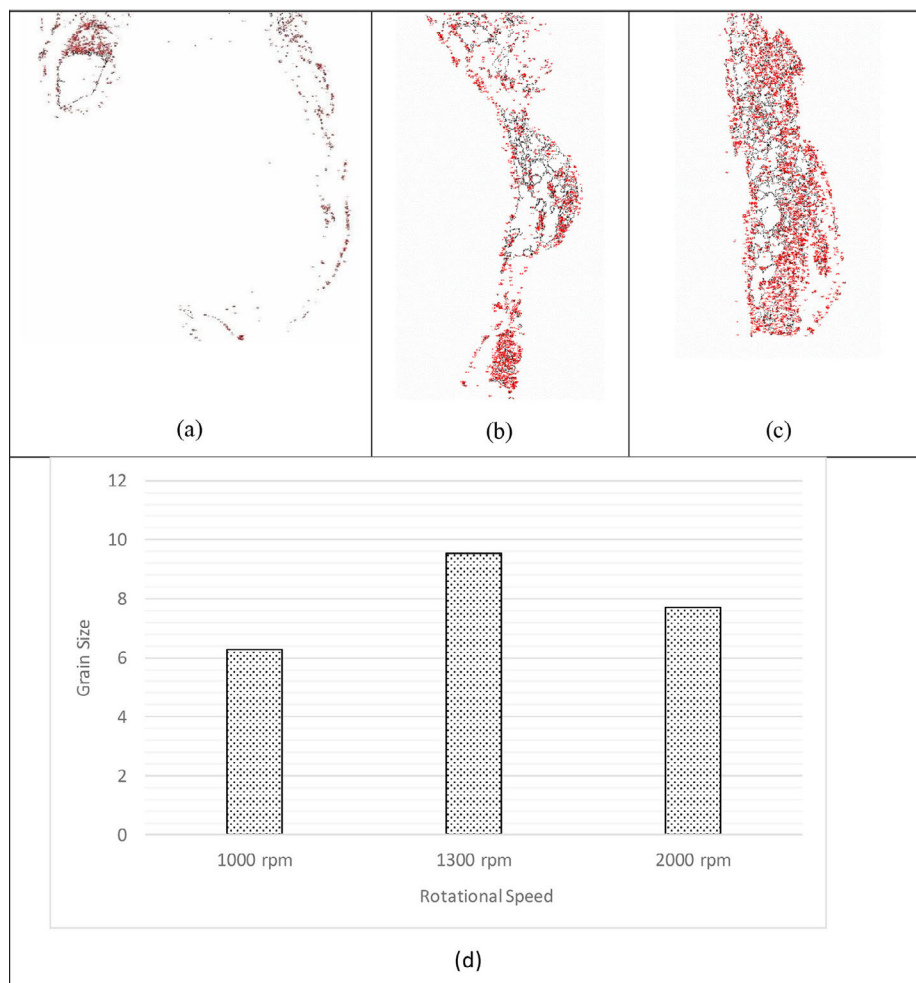


Figure 18. Analysis grain size of aluminium–copper friction welds at different rotational speeds: (a) 1000 rpm, (b) 1300 rpm, (c) 2000 rpm, and (d) Grain size distribution

uniform and stable intermetallic layer (IMC). At 1300 rpm, an average IMC thickness of 204.605 μm was obtained, indicating optimal diffusion of Al and Cu atoms. When the rotational speed was increased to 2000 rpm, the grain size decreased to 7.707 μm . This decrease is thought to be caused by more extreme thermomechanical conditions resulting in greater plastic deformation and the emergence of local recrystallization that inhibits further grain growth. However, excessive heat also encourages the formation of more brittle intermetallic phases, resulting in a decrease in tensile strength compared to the 1300 rpm condition.

Table 8 presents the IMC thickness sequence according to rotational speed, namely 1300, 2000, and 1000 rpm. The greatest thickness at 1300 rpm results from sufficient heat input to promote grain growth without causing significant IMC remelting [49, 51]. Based on the macro- and microstructural observations, a rotational speed of 1300 rpm is considered optimal for improving the mechanical properties of the joint.

The present experimental results do not indicate a single critical IMC thickness that universally governs joint strength. Although the specimen produced at 60 s friction time exhibited an average IMC thickness of approximately 137 μm , the replicated specimen welded under the same nominal condition showed an average IMC thickness of approximately 205 μm together with considerably higher tensile strength. These observations suggest that joint performance is influenced not only by IMC thickness but also by IMC continuity, local phase distribution, diffusion uniformity, and thermo-mechanical stability during welding. Therefore, IMC thickness alone should not be considered the sole controlling factor for joint strength. Thin IMC layers were associated with incomplete interfacial bonding, while excessively thick IMCs promoted brittle fracture due to the increased presence of $\theta(\text{Al}_2\text{Cu})$ and $\eta(\text{AlCu})$

phases within the CWZ. Therefore, optimum mechanical properties were achieved when IMC growth remained sufficiently continuous without becoming excessively brittle.

For the rotational speed variations (Figures 19–21), all three specimens exhibited three types of intermetallic compounds (IMCs). Point A contained 31.66%, 27.35%, and 44.48% Cu, corresponding to the hypereutectic phase $\alpha(\text{Al}) + \theta(\text{Al}_2\text{Cu})$, with an atomic ratio of $n(\text{Al}):n(\text{Cu}) \approx 2:1$ [37–40]. Point B showed Cu contents of 6.7%, 4.27%, and 1.72%, representing the hypoeutectic $\alpha(\text{Al})$ phase [37–39]. Meanwhile, Point C contained 65.82%, 15.21%, and 68.99% Cu, indicating the presence of $\theta(\text{Al}_2\text{Cu}) + \eta_2(\text{Al-Cu}(\text{rt}))$ phases at 1000 and 2000 rpm, and $\alpha(\text{Al}) + \theta(\text{Al}_2\text{Cu})$ at 1300 rpm [41, 32, 35, 38, 39].

Micro Vickers hardness

Figure 22 illustrates the micro-Vickers hardness distribution of the aluminium–copper friction weld joints at different rotational speeds. The average hardness values at 1000, 1300, and 2000 rpm were 64.92 HV, 74.52 HV, and 70.72 HV, respectively. Hardness variation is influenced by recrystallization, atomic diffusion, IMC formation, and phase transformations [49]. Although the microstructure at 1300 rpm appears coarser than at 1000 rpm, it exhibits the highest hardness due to the thicker IMC layer [41]. However, excessively thick IMCs tend to be brittle and can reduce joint toughness. The increased hardness observed in the CWZ is likely associated with intermetallic compound formation, localized plastic deformation, and possible dynamic recrystallization behavior during rotary friction welding [52], while lower hardness values in the TMAZ and HAZ were caused by rapid grain growth and thermal softening [50, 42, 53]. Additionally,

Table 7. Interdiffusion parameters of Al and Cu at different rotational speeds

Condition	Diffusion constant ($\times 10^{-4} \text{ m}^2\text{s}^{-1}$) [34]	Activation energy (kJ/mol) [34]	Specimens	Temperature (K)	Diffusion coefficient (m^2s^{-1})
Al in Cu ($D_{\text{Al}}^{\text{Cu}}$)	0.131	185.2	1	590	5.251×10^{-22}
			2	612	4.906×10^{-25}
			3	528	6.236×10^{-24}
Cu in Al ($D_{\text{Cu}}^{\text{Al}}$)	0.654	1.41	1	590	4.957×10^{-24}
			2	612	2.041×10^{-21}
			3	528	4.743×10^{-24}

Table 8. IMC thickness of specimens at different rotational speeds

Length measurement	Specimens (μm)		
	1000 rpm	1300 rpm	2000 rpm
1	149.464	193.807	93.392
2	156.119	155.762	78.419
3	129.956	204.425	182.379
4	119.212	246.018	192.073
5	130.419	223.011	203.524
Average	137.034	204.605	149.957

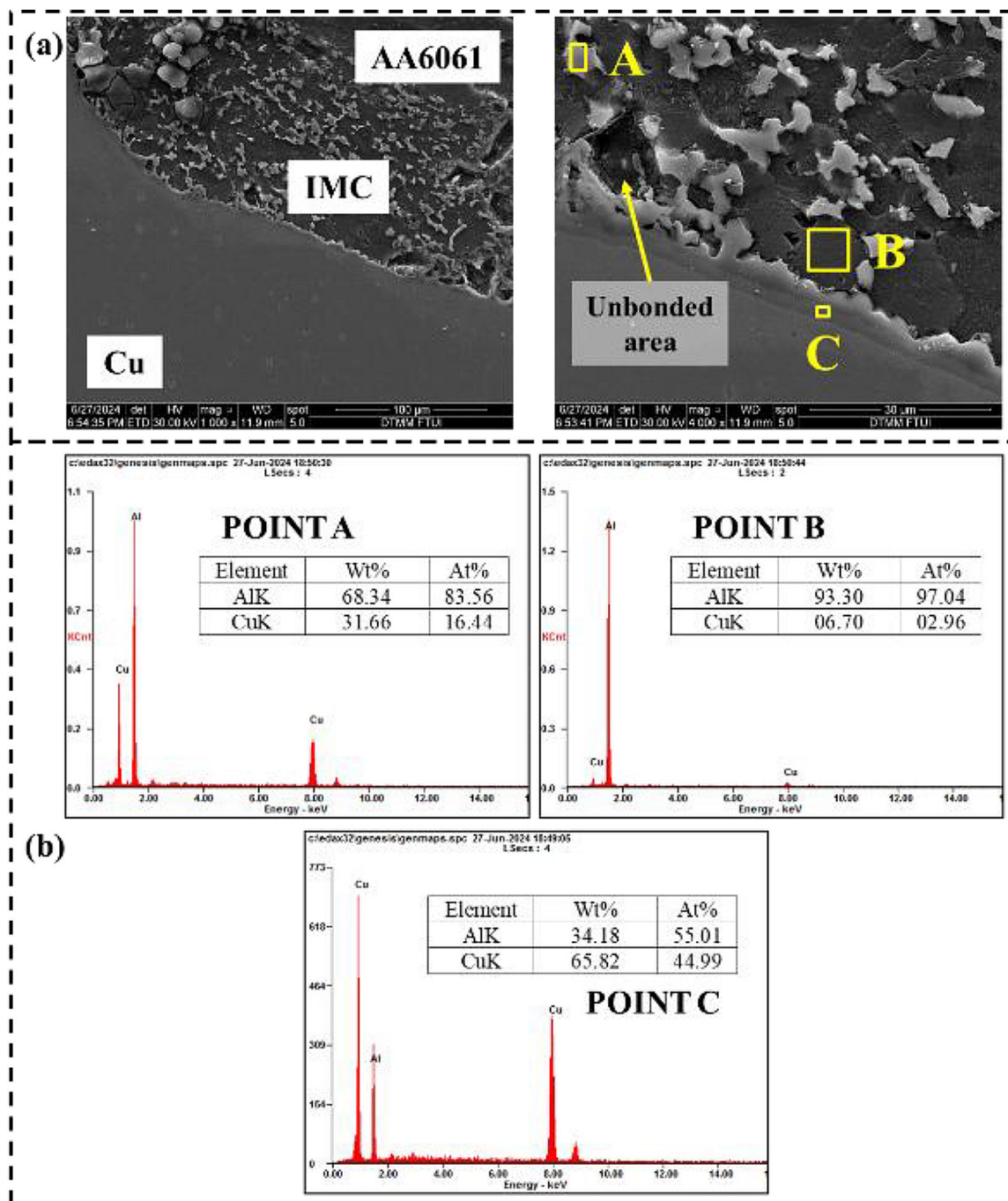


Figure 19. (a) SEM image and (b) EDX mapping of the aluminium–copper friction weld joint at a rotational speed of 1000 rpm

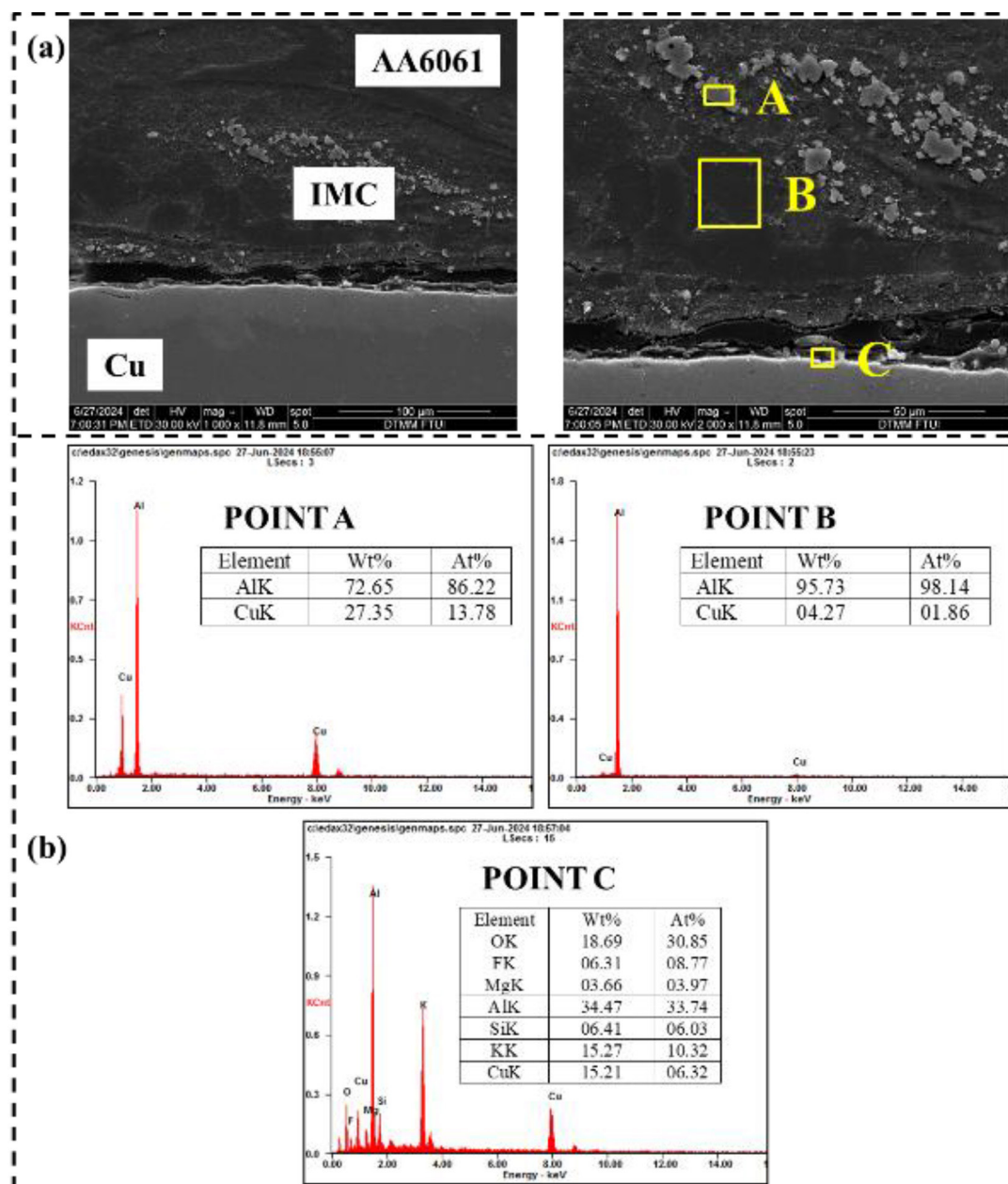


Figure 20. (a) SEM image and (b) EDX mapping of the aluminium–copper friction weld joint at a rotational speed of 1300 rpm

plastic deformation in the CWZ prevents grain growth, further increasing hardness [49].

It should be noted that quantitative grain size measurement and phase fraction analysis were not performed in this study. Therefore, the interpretation of recrystallization behavior and phase-related strengthening mechanisms is based primarily on qualitative microstructural observations and hardness trends. The CWZ hardness values at 1000, 1300, and 2000 rpm were 82.7 HV, 113.5 HV, and 110.3 HV, respectively. Meanwhile, the aluminium-side TMAZ recorded 67.2 HV, 86.9 HV, and 71.8 HV, and the copper-side TMAZ

measured 62.9 HV, 60.4 HV, and 59.7 HV. Overall, a rotational speed of 1300 rpm provided the most stable thermo-mechanical equilibrium and material mixing, resulting in the optimal hardness across all weld zones.

Tensile strength

Figure 23 presents the stress–strain curves for different rotational speeds, while Figure 24 compares the ultimate tensile strength (UTS) and yield strength of the joints. The tensile strengths at 1000, 1300, and 2000 rpm were 2.95 MPa,

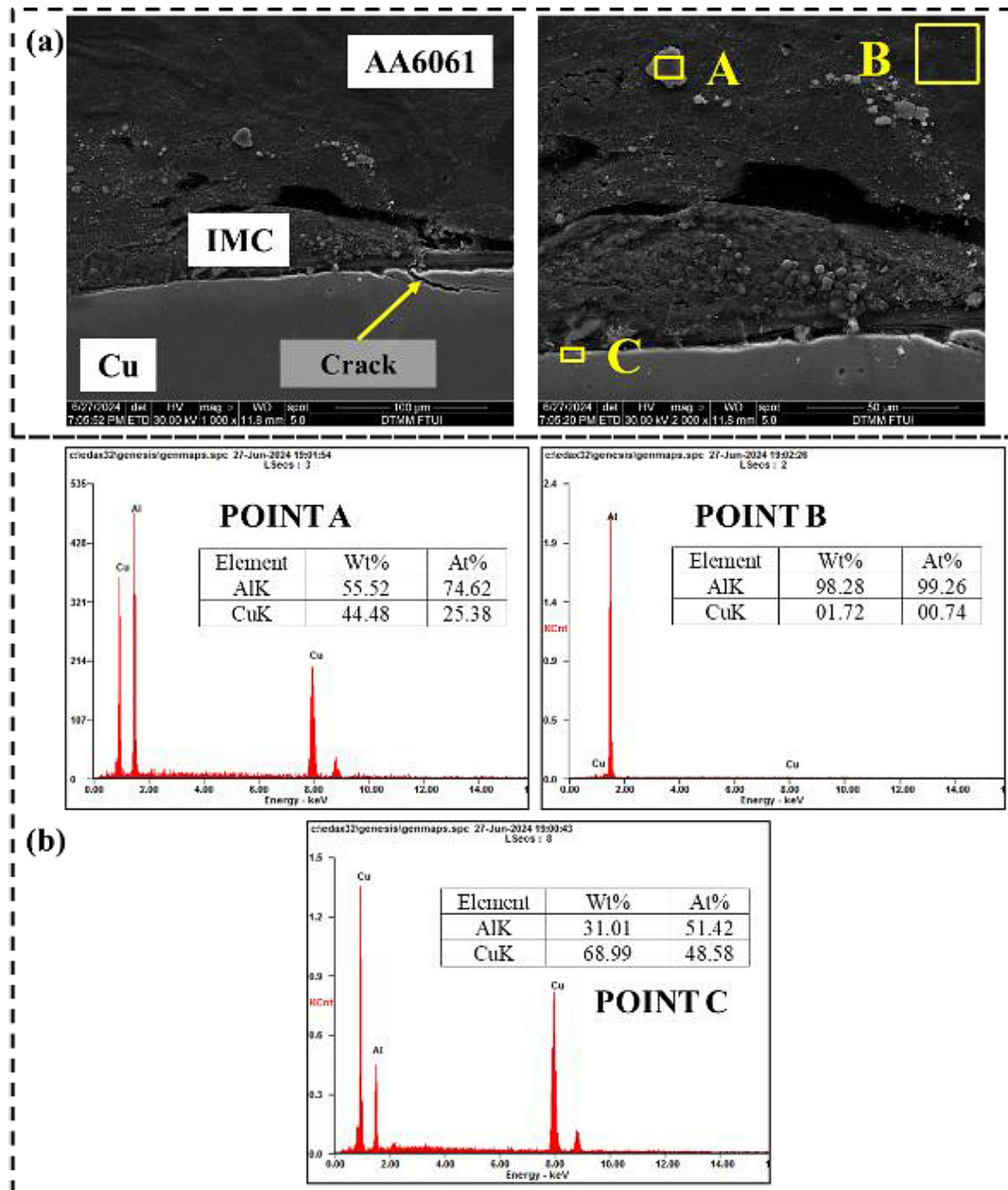


Figure 21. (a) SEM image and (b) EDX mapping of the aluminium–copper friction weld joint at a rotational speed of 2000 rpm

63.90 MPa, and 21.14 MPa, respectively. The significant increase in tensile strength at 1300 rpm compared to friction-time variations is attributed to the synergistic stabilization of plastic flow and diffusion-controlled IMC growth under optimized thermo-mechanical conditions. The reduction in tensile strength compared to the base metals [22, 28, 44] was primarily due to the presence of the heat-affected zone (HAZ) and the formation of brittle intermetallic compounds (IMC) [22]. Increasing the rotational speed increased the heat generation rate and enhanced the transfer of aluminium onto the copper surface

[26]. According to Li et al. [54], high heat flux and temperature rise occur at the beginning of the process but decrease as rotational speed declines. Consequently, excessive grain growth and residual stresses develop, reducing the tensile strength of the welds [49, 51].

The initial negative stress region observed in the raw tensile curve originated from load cell offset during the early stage of data acquisition and does not represent actual mechanical behavior of the welded joint. The curve was subsequently corrected through stress-axis normalization for accurate UTS interpretation.

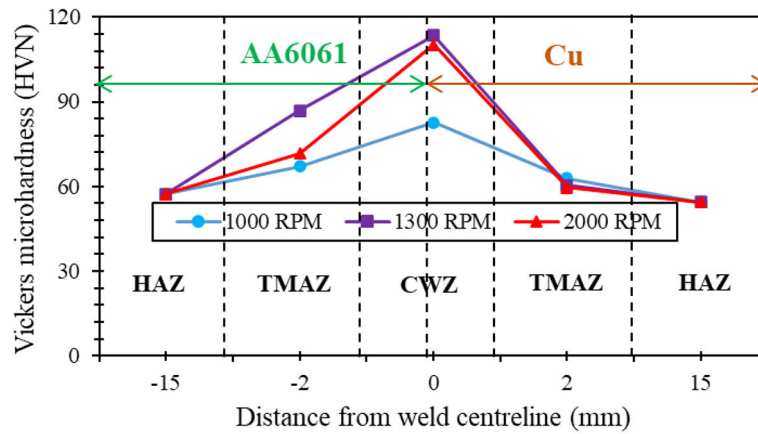


Figure 22. Micro-Vickers hardness distribution of aluminium–copper friction weld joints at different rotational speeds

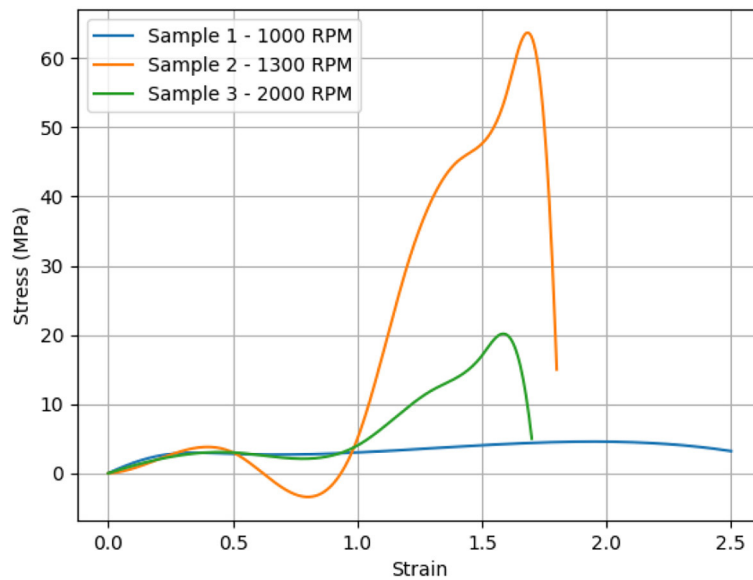


Figure 23. Stress–strain curves of aluminium–copper friction welds at various rotational speeds

The superior mechanical performance observed at 1300 rpm is associated with a balanced thermo-mechanical condition that promotes stable plastic flow and controlled IMC growth. At lower rotational speed (1000 rpm), insufficient heat generation limited atomic diffusion and resulted in incomplete metallurgical bonding. Conversely, excessive heat generation at 2000 rpm accelerated grain coarsening and brittle IMC formation, reducing joint strength. Therefore, 1300 rpm provided the optimum condition for achieving a homogeneous interfacial microstructure and improved mechanical integrity.

Figures 25–27 illustrate the fracture morphology of aluminium–copper friction welds under different rotational speeds. All specimens fractured within the CWZ, exhibiting a mixed

ductile–brittle mode at 1000 and 2000 rpm – characterized by the presence of dimples and cleavage – while the 1300 rpm specimen predominantly showed a cleavage pattern [45, 46]. The 1000 rpm specimen also displayed sticking features, indicating strong metallurgical bonding. EDX analysis (Figures 24(c)–26(c)) revealed ductile $\alpha(\text{Al})$ at point A. At point B, specimens welded at 1000 and 1300 rpm contained brittle intermetallic ζ_1 phases, whereas the 2000 rpm specimen showed $\theta + \eta_2$ phases, which are also brittle. Point C consisted of a mixture of $\alpha(\text{Al})$ and $\theta(\text{Al}_2\text{Cu})$ phases, exhibiting transitional mechanical characteristics between ductility and brittleness. Overall, the 1300 rpm rotational speed produced the most balanced microstructure – optimizing hardness, tensile strength, and crack resistance by minimizing

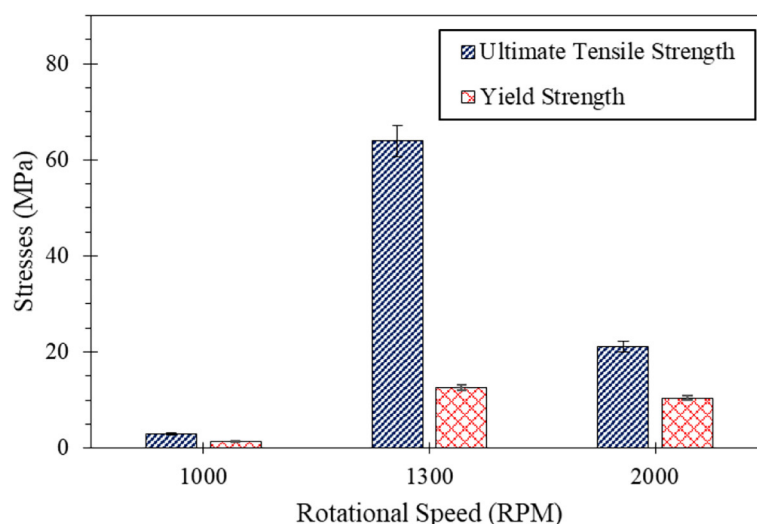


Figure 24. Tensile strength of aluminium–copper friction welds at different rotational speeds

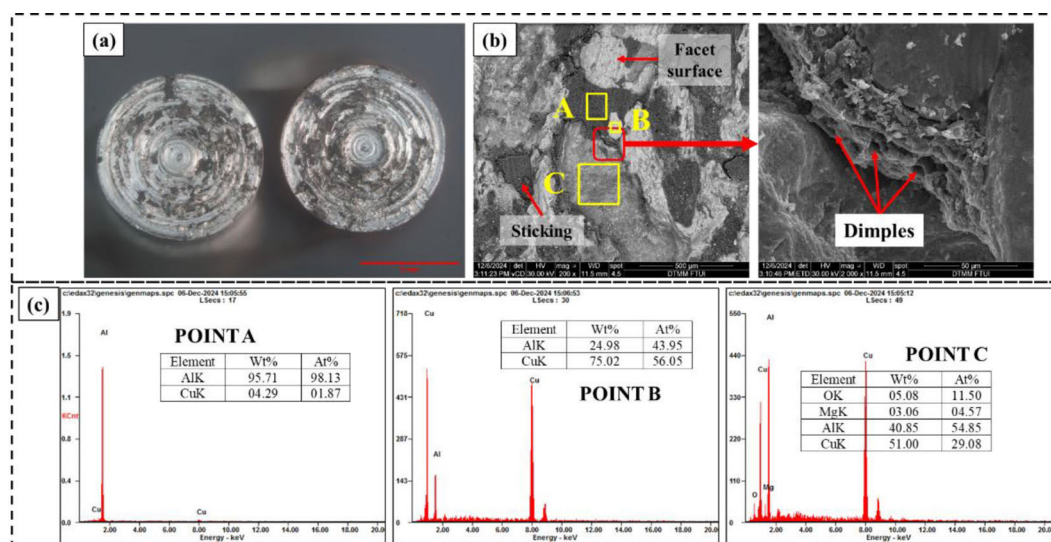


Figure 25. (a) Macrostructure of the specimen welded at a rotational speed of 1000 rpm after the tensile test, (b) SEM images of the fracture surface, and (c) EDX analysis of the fracture surface

excessive intermetallic phase formation. The oscillations observed in the stress–strain curves are attributed to intermittent microcrack initiation and propagation along brittle IMC layers within the CWZ. Due to the limited ductility of intermetallic compounds, localized fracture events cause temporary stress drops followed by load redistribution. Similar discontinuous deformation behavior has been reported in dissimilar metal joints containing brittle intermetallic interfaces.

Fracture location analysis revealed that most tensile specimens fractured near the CWZ interface where the IMC layer was concentrated. Specimens exhibiting limited elongation generally fractured directly within the brittle IMC region, whereas specimens with improved ductility showed partial

plastic deformation extending toward the aluminium side due to its lower hardness and higher ductility compared with Cu and the intermetallic layer.

The fracture morphology analysis in this study was primarily qualitative and focused on identifying dominant fracture characteristics such as cleavage facets, dimples, sticking regions, and crack propagation paths. Quantitative fracture mechanics parameters, including fracture toughness, crack density distribution, and localized strain analysis, were not evaluated in the present work. Therefore, the fracture behavior interpretation was mainly correlated with tensile strength results, IMC distribution, and interfacial microstructural observations. The experimental results suggest that joint strength

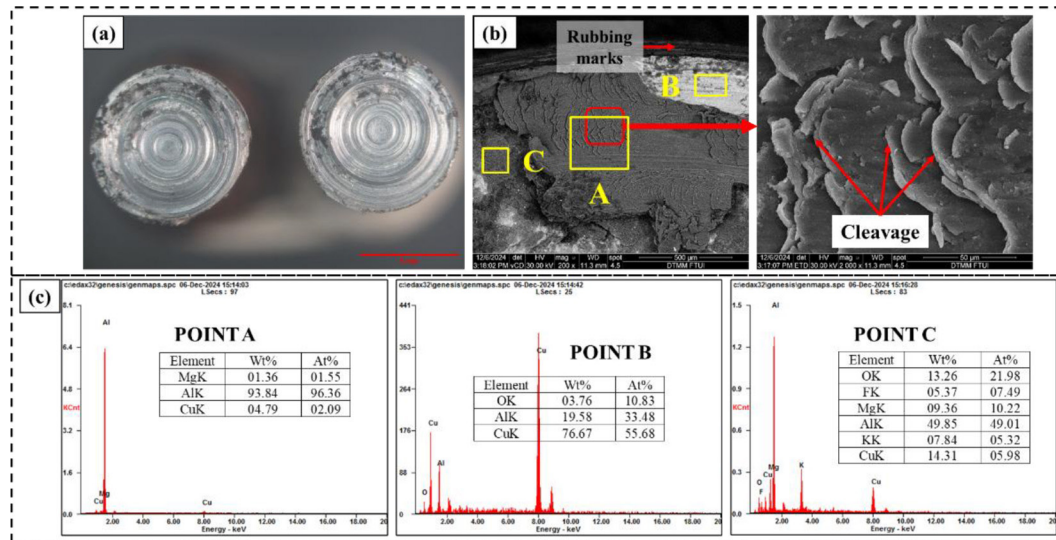


Figure 26. (a) Macrostructure of the specimen welded at a rotational speed of 1300 rpm after the tensile test, (b) SEM images of the fracture surface, and (c) EDX analysis of the fracture surface

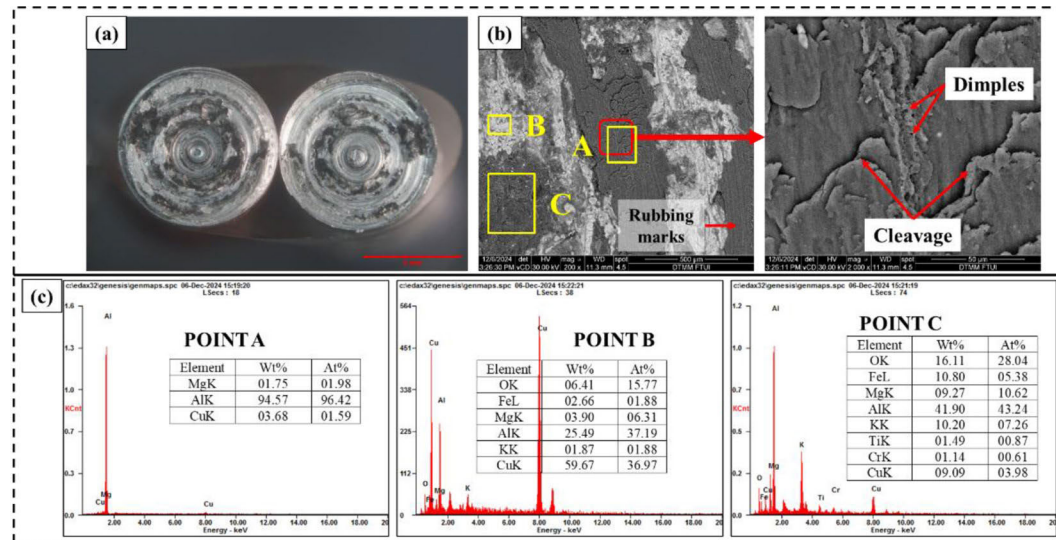


Figure 27. (a) Macrostructure of the specimen welded at a rotational speed of 2000 rpm after the tensile test, (b) SEM images of the fracture surface, and (c) EDX analysis of the fracture surface

tended to improve when the IMC thickness was within the approximate range of 130–140 μm under the present welding conditions. This observation may indicate the existence of a favorable IMC thickness range that balances metallurgical bonding and brittleness; however, further validation through additional datasets and fracture mechanics analysis is required. The proposed IMC thickness range should be interpreted as an experimental trend specific to the present welding parameters rather than as a universal critical value for Al–Cu rotary friction welding systems. Although the present results provide important insight into thermo-mechanical behavior and

IMC evolution during Al–Cu rotary friction welding, several microstructural aspects still require further investigation at the grain scale. It should be noted that grain-scale characterization techniques such as EBSD or TEM were not employed in the present study. Therefore, the proposed deformation synergy and recrystallization behavior were interpreted primarily from qualitative microstructural observations, hardness evolution, and thermo-mechanical response during welding. Future work employing grain orientation mapping and strain localization analysis is recommended to further validate deformation mechanisms within the CWZ.

CONCLUSIONS

This study investigated the effects of friction time and rotational speed on diffusion behavior, intermetallic compound (IMC) formation, and mechanical performance of AA6061–Cu rotary friction-welded joints. The results show that joint performance is governed primarily by diffusion-controlled interfacial reactions rather than heat input alone. Increasing friction time enhanced atomic diffusion and IMC formation, improving metallurgical bonding; however, excessive IMC growth promoted brittleness and crack initiation. An intermediate rotational speed of 1300 rpm provided the most stable thermo-mechanical condition, resulting in uniform IMC growth, the highest tensile strength, and a homogeneous hardness distribution. Within the experimental conditions investigated, the combination of 60 s friction time and 1300 rpm generally exhibited the most favorable thermo-mechanical characteristics. However, because noticeable variation was observed between replicated specimens, additional replicated experiments are required before this condition can be considered statistically optimal.

It should be noted that direct experimental validation of diffusion activation energy, such as EPMA line scanning or concentration-gradient analysis, was not performed in this study. Therefore, the diffusion analysis is intended primarily to support qualitative interpretation of thermo-diffusion behavior during rotary friction welding. The noticeable variation observed between replicated specimens indicates that the current experimental setup still exhibits process variability. Future work should include a larger number of replicated experiments together with statistical analysis to improve confidence in the determination of optimum welding parameters. One limitation of the present work is that the interface temperature was measured using an infrared thermometer on the external weld surface. Consequently, the recorded temperatures do not exactly represent the actual temperature inside the bonding interface. Future studies employing infrared thermal imaging, embedded thermocouples, or thermo-mechanical finite element simulations are recommended to improve the quantitative accuracy of diffusion kinetics analysis.

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