

Effect of annealing on the microstructure and mechanical properties of 1060Al/T2Cu composite plates after cold rolling

Haoran Guo

Huihui Nie

niehuihui0618@163.com

hongsheng chen

Jingyi Li

Wentao Fan

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Abstract: The effects of annealing at 150 °C and 400 °C on the interface structure and mechanical properties of 1060Al/T2Cu composite plates after cold rolling are investigated. After cold rolling, the pre-existing interface fractures along the rolling direction (RD), and its thickness decreases to 2.3 μm. On the Cu side, a new P{011}<122> texture forms; the intensities of Goss{110}<011> and Brass{011}<211> textures increase, whereas those of Copper{112}<111> and S{123}<634> textures decrease, indicating the transformation of Copper and S textures toward Goss and Brass textures. After annealing at 150 °C, the interface exhibits a three-layer structure with a thickness of 2.7 μm and remains fractured along RD. From the Cu side to the Al side, Cu₉Al₄ and CuAl are identified, whereas no stable compound phase forms in the interface adjacent to the Al side. Recovery occurs in both Cu and Al, and the textures of the constituent layers remain essentially unchanged. The UTS, YS, and EL of the composite plates are 194 MPa, 162 MPa, and 18%, respectively, all higher than those in the rolled condition, demonstrating a simultaneous improvement in strength and ductility. After annealing at 400 °C for 1 h, the interface thickness reaches 7.5 μm, and the phases from the Cu side to the Al side are Cu₉Al₄, CuAl, and CuAl₂. The initially fractured interface evolves into a structure containing fine pores. After high-temperature annealing for 2 h, the interface thickness increases to 8.7 μm; the pores disappear and a fully continuous three-layer structure is formed without a change in phase constitution. Only a very weak P texture remains on the Cu side, whereas P and Brass textures form in Al. On the Cu side, low-angle grain boundaries merge and transform into high-angle grain boundaries, accompanied by a marked reduction in dislocation density, indicating that static recrystallization is dominant in Cu. In contrast, high-temperature recovery occurs on the Al side, producing a microstructure dominated by recovered subgrains. The UTS and YS decrease, whereas EL increases, reaching 169 MPa, 92 MPa, and 40%, respectively.

Keywords: Cu/Al composite plates; cold rolling; annealing; interface; texture

Introduction

Cu exhibits high electrical and thermal conductivity together with good ductility and is therefore widely used in aerospace, electronics, construction, and electric-power applications. However, its high cost and relatively high density limit broader application. Al offers the advantages of low cost, low density, and good corrosion resistance. Cu/Al composite plates combine the complementary advantages of Cu and Al and thereby provide favorable overall properties [1-2]. At present, cold roll bonding is the most widely used route for manufacturing Cu/Al composite plates because of its simple operation, high production efficiency, and low cost [3-5].

The microstructure and properties of Cu/Al composite plates produced by cold rolling are commonly tailored by annealing [6-7]. Variations in annealing temperature (low-temperature annealing at 150-250 °C, intermediate-temperature annealing at 300-350 °C, and high-temperature annealing above 350 °C) and holding time can alter the grain size, compositional homogeneity, and texture components [8], thereby regulating the mechanical properties. Previous studies show that the interface thickness increases with increasing annealing temperature and holding time [9], and that several IMCs, including Cu_9Al_4 , CuAl_2 , and CuAl , form at the interface [10]. Cheng et al. [11] reported that the interface thickness follows a power-law relationship with annealing time but an exponential relationship with annealing temperature, indicating that temperature has a much stronger effect on interface growth than time. The effect of a 50 °C increase in annealing temperature on the thickness of the interfacial IMC layer is approximately equivalent to extending the annealing time by a factor of 2-3 [12]. The annealing temperature primarily governs changes in the phase type within the interface, whereas the annealing time mainly controls thickness uniformity.

In summary, the evolution of the bonding interface and microstructure of Cu/Al composite plates during intermediate-temperature annealing has been relatively well established, whereas considerably fewer studies have addressed low- and high-temperature conditions. Therefore, the precipitation and evolution of IMC phases at the interface, together with the evolution of Cu and Al grain size and orientation during low- and high-temperature annealing, are investigated here to clarify the mechanisms by which microstructural evolution affects the mechanical properties.

1 Experimental materials and methods

The starting material consisted of 3 mm thick 1060Al/T2Cu composite plates, with Cu and Al layer thicknesses of 1 mm and 2 mm, respectively. The chemical compositions are listed in Table 1. The 3 mm thick Cu/Al composite plates were subjected to two-pass cold rolling on a BKDΦ130 laboratory rolling mill to a final thickness of 1.5 mm, corresponding to a total thickness reduction of 50%. After cold rolling, the Cu and Al layer thicknesses were 0.54 mm and 0.96 mm, respectively. The macroscopic appearances before and after rolling are shown in Fig. 1. Rolling direction (RD), transverse direction (TD), and normal direction (ND) denote the longitudinal, transverse, and through-thickness directions, respectively. The composite plates after cold rolling were annealed in a KSL-1200X muffle furnace at 150 °C for 2 h (denoted A150/2h) and at 400 °C for 1 h and 2 h (denoted A400/1h and A400/2h), followed by air cooling. After grinding and mechanical vibratory polishing, the microstructures on the ND-RD plane were characterized using optical microscopy

(OM), scanning electron microscopy (SEM, JSM-IT700HR), and electron backscatter diffraction (EBSD). For EBSD specimen preparation, Ar⁺ ion polishing was performed at 6.5 keV for 25 min. EBSD measurements were conducted at an accelerating voltage of 20 keV with a step size of 0.15 μm , and the data were processed using CHANNEL 5. Tensile tests along RD were conducted at room temperature on a universal testing machine at a displacement rate of 0.5 mm/min to determine UTS, YS, and EL. The fracture surfaces were examined after tensile testing. Vickers microhardness was measured using an FM-700 microhardness tester under a load of 2 N and a dwell time of 10 s.

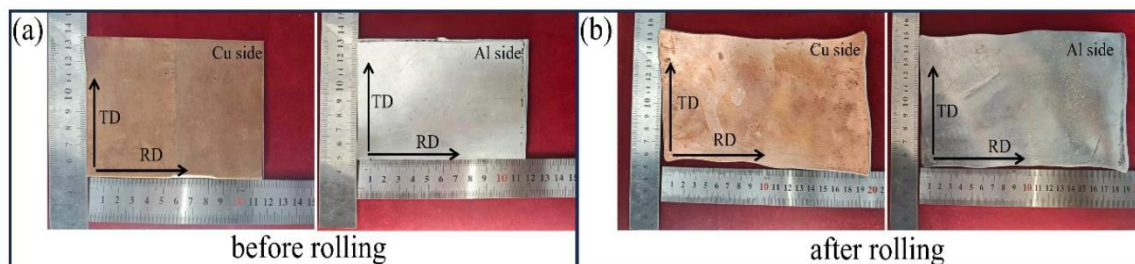


Fig. 1 Macroscopic morphology of Cu/Al composite plates

Table 1 Chemical compositions of 1060Al and T2Cu

Material		Chemical composition (wt.%)						
1060Al	Al	Cu	Si	Fe	Mn	Mg	Zn	Ti
	Bal.	0.05	0.25	0.35	0.03	0.03	0.1	0.03
T2Cu	Cu	P	Ni	Si	Fe	Zn	S	Ag
	Bal.	0.07	0.02	0.04	0.08	0.08	0.09	0.1

2 Results and discussion

2.1 Interface microstructure

2.1.1 Interface morphology and composition

Figure 2 presents SEM images of the Cu/Al interface together with EDS point and line scan results, and the evolution of the interface is illustrated in Fig. 3. Before rolling, the interface is flat and contains three continuously distributed interface layers. After cold rolling, the interface fractures along RD, and its thickness decreases to 2.3 μm under the rolling load. Based on the chemical compositions measured at the points in Fig. 2 and the equilibrium Cu-Al binary phase diagram [13-14], the interface adjacent to the Al side before rolling is identified as CuAl. Although rolling promotes interdiffusion between the Cu and Al layers, the phase constitution of the interface remains unchanged. In A150/2h, the interface remains discontinuous, whereas the pores between the interface and the matrix become smaller. The interface thickness increases slightly to 2.7 μm because, at the low annealing temperature, most Cu and Al atoms cannot overcome the diffusion-energy barrier and the interfacial reaction remains weak. After A150/2h, the interface is divided into three layers. Adjacent to the Cu matrix, the Al and Cu atomic fractions are 33.42 at.% and 66.58 at.%, respectively, suggesting that the unstable phase in the specimen before rolling transforms into Cu₉Al₄. In the middle layer, the Al and Cu atomic fractions are 48.88 at.% and 51.12 at.%, respectively, corresponding to CuAl. No stable compound forms in the layer adjacent to the Al side.

After A400/1h, the interface thickens along both RD and ND to 7.5 μm , and the initially fractured morphology evolves into a structure containing fine pores. This indicates that the high temperature sharply increases the atomic diffusion rate and promotes more extensive mutual diffusion of Cu and Al. The phase adjacent to the Cu side is Cu_9Al_4 and the middle phase is CuAl . These two layers only thicken and retain the same phases as in A150/2h, whereas the layer adjacent to the Al side transforms into CuAl_2 . Thus, the phase sequence across the interface from Cu to Al is Cu_9Al_4 , CuAl , and CuAl_2 . With increasing annealing time, the interface thickness in A400/2h increases to 8.7 μm , while the pores disappear and a fully continuous three-layer interface forms. The phase constitution remains the same as in A400/1h. However, Fig. 2k shows that the CuAl_2 layer adjacent to Al is the thickest, indicating the highest growth rate.

During Cu-Al interdiffusion, the interface thickness is governed by the atomic diffusion rate, which is related to the diffusion coefficient. The diffusion coefficient D depends on the annealing temperature T and the activation energy for diffusion Q , and can be described by the Arrhenius equation [15-16]:

$$D = D_0 \exp\left(-\frac{Q}{RT}\right) \quad (1)$$

where D_0 is the diffusion constant in m^2/s , Q is the activation energy for diffusion in J/mol , T is the absolute temperature in K , and R is the molar gas constant with a value of $8.14 \text{ J/(K}\cdot\text{mol)}$. Both D_0 and Q are intrinsic material parameters and are independent of temperature. Therefore, increasing the annealing temperature increases the diffusion rate of atoms at the interface and accelerates interface growth. Accordingly, the interface in A400/2h is substantially thicker than that in A150/2h. In general, increasing the annealing time also increases the interface thickness [17], explaining why the interface in A400/2h is slightly thicker than that in A400/1h. However, excessively high annealing temperatures or excessively long annealing times can cause a rapid increase in interface thickness and a pronounced deterioration in material properties [18].

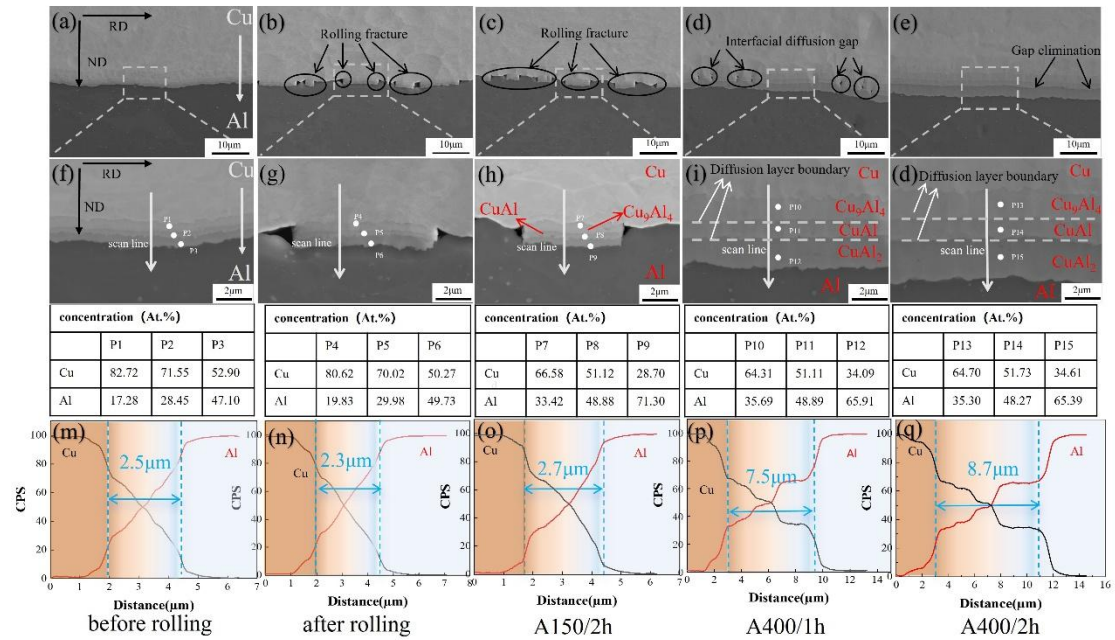


Fig. 2 SEM images and EDS point and line scan results of Cu/Al composite plates

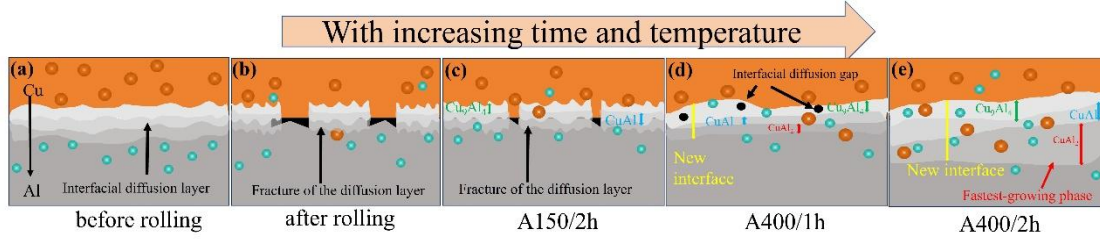


Fig. 3 Schematic illustration of the interface evolution mechanism in Cu/Al composite plates

2.1.2 EBSD characterization and texture evolution

Figures 4 and 5 show the EBSD results for the composite plates before and after rolling and after annealing, including recrystallization fraction (RX) maps, grain-boundary (GB) maps with grain-size information, kernel average misorientation (KAM) maps, and inverse pole figure (IPF) maps. Figures 6a, b, d, and e summarize the recrystallization fractions and the fractions of high-angle grain boundaries (HAGBs) and low-angle grain boundaries (LAGBs), while Fig. 6g schematically illustrates the microstructural evolution mechanism during high-temperature annealing. The geometrically necessary dislocation (GND) density was also calculated to assist in clarifying the recrystallization mechanisms in Al and Cu [19-20], as shown in Figs. 6c and f. The GND density was calculated as [21-22]:

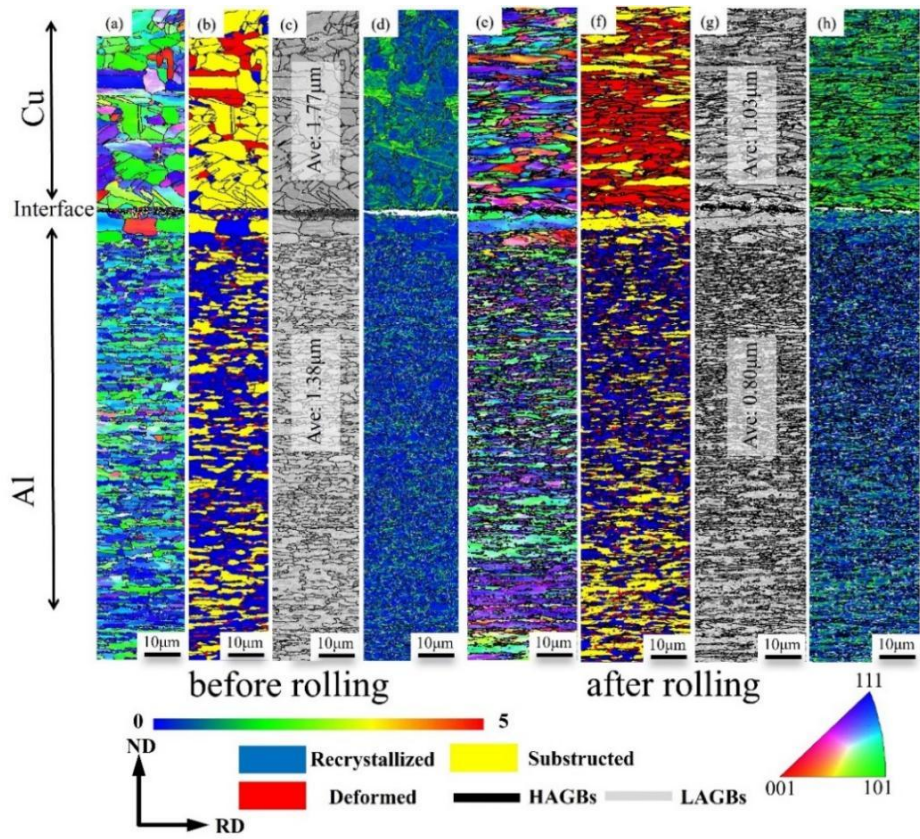
$$\rho_{GND} = \frac{2KAM}{\mu b} \quad (2)$$

where ρ_{GND} is the GND density, μ is the measurement step size, and b is the Burgers vector of the face-centered cubic structure. The b values for Al and Cu are 0.286 nm and 0.256 nm, respectively.

Before rolling, the average grain size of Cu is 1.77 μm , with substructured grains accounting for 79%, whereas the average grain size of Al is 1.38 μm and most Al grains are elongated along RD. Both constituent layers are dominated by HAGBs and exhibit relatively low dislocation densities. After cold rolling, the grains in both layers become further elongated along RD and are markedly refined. The average grain sizes decrease to 1.03 μm in Cu and 0.8 μm in Al. The fraction of LAGBs in Cu increases to 65%, and the GND density increases from $16.1 \times 10^{15} \text{ m}^{-2}$ before rolling to $39.6 \times 10^{15} \text{ m}^{-2}$ after rolling. In Al, the GND density increases from $9.79 \times 10^{15} \text{ m}^{-2}$ to $18.1 \times 10^{15} \text{ m}^{-2}$. The substantially larger increase in GND density in Cu indicates that the Cu layer accommodates more deformation during rolling. The fraction of deformed grains in Cu reaches 61%, whereas the Al microstructure is dominated by recrystallized and substructured grains, accounting for 53% and 36%, respectively. This difference arises because Cu has a low stacking-fault energy (SFE), which restricts cross-slip and promotes dislocation pile-up [23], whereas the higher SFE of Al facilitates recovery through dislocation cross-slip and climb, leading to subgrain formation and recrystallization [24]. In addition, the fraction of recrystallized Al grains increases and the grains become finer toward the interface, whereas regions farther from the interface contain a larger fraction of substructured grains and coarser grains. This spatial variation is associated with the shear-strain gradient near the interface. The marked difference in deformation resistance between the interface and Al produces a strain gradient that decreases from the interface into the Al layer [23]. The high strain near the interface provides sufficient driving force for recrystallization; dislocation consumption and fine-grain formation during recrystallization consequently produce a high recrystallized fraction and a low dislocation density near the interface.

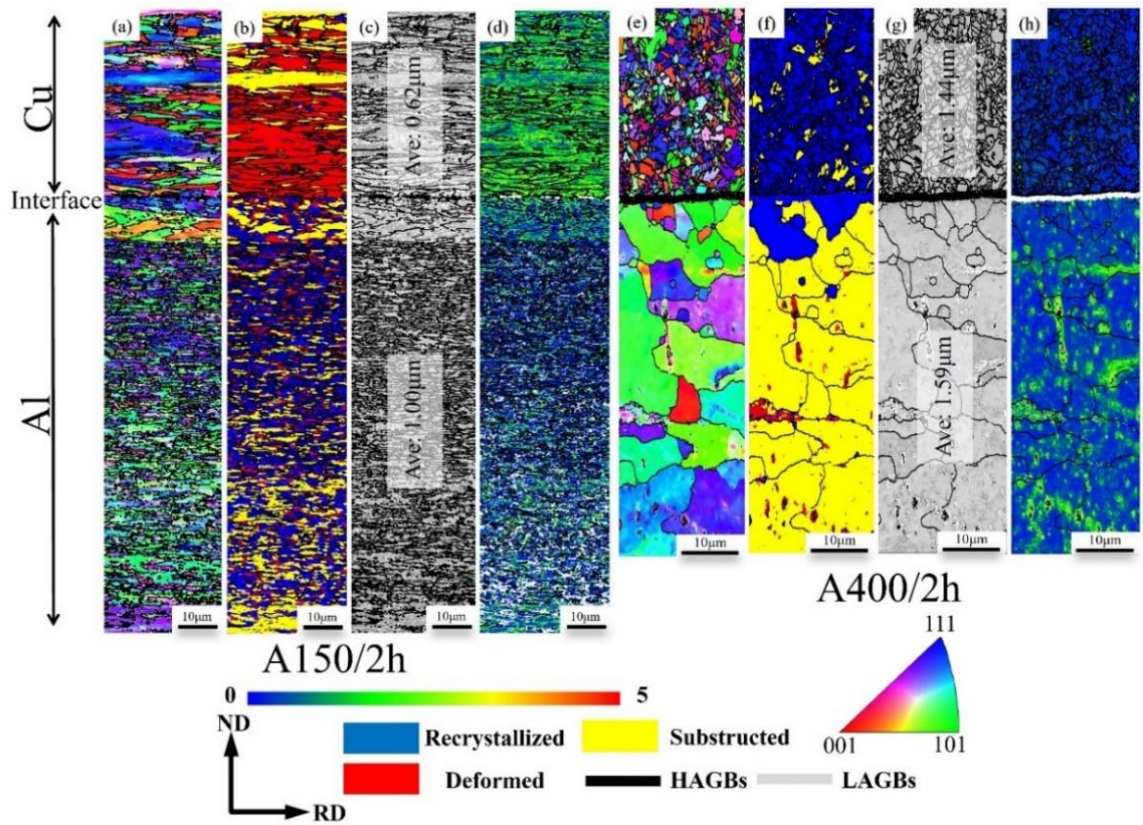
In A150/2h, the Cu side retains clear characteristics of the deformation structure introduced by cold rolling. The grain size decreases to 0.62 μm , and the fraction of recrystallized grains increases slightly, while the fraction of deformed grains remains high. Meanwhile, the fraction of LAGBs increases and the GND density decreases slightly. These results indicate that static recovery and dislocation rearrangement dominate on the Cu side, while the stored energy introduced by rolling is not fully released and static recrystallization remains incomplete. A similar trend is observed in Al, although slight grain growth occurs. Therefore, recovery dominates in both Cu and Al in A150/2h.

In A400/2h, the elongated deformed Cu grains transform into equiaxed recrystallized grains, the recrystallized fraction increases to 88%, and the grain size increases from 1.03 μm after rolling to 1.44 μm . The GND density in Cu decreases sharply from $39.6 \times 10^{15} \text{ m}^{-2}$ after rolling to $5.76 \times 10^{15} \text{ m}^{-2}$. LAGBs merge into HAGBs, and the HAGB fraction increases markedly to 96%, indicating that static recrystallization dominates in Cu during high-temperature annealing. In Al, the GND density decreases significantly from $18.1 \times 10^{15} \text{ m}^{-2}$ after rolling to $2.18 \times 10^{15} \text{ m}^{-2}$, whereas the LAGB fraction increases from 20% to 75%. Deformed grains are nearly eliminated; however, the microstructure does not evolve through a simple continued increase in recrystallized grains. Instead, a large number of subgrains appear and the grain size increases from 0.8 μm to 1.59 μm . This behavior indicates high-temperature recovery in Al. After extensive dislocation annihilation, the remaining dislocations rearrange into LAGBs, producing a microstructure dominated by recovered subgrains [25]. The different microstructural evolution of Cu and Al can be attributed to two factors. First, the high SFE of Al facilitates dislocation cross-slip and climb, allowing stored energy to be released more readily through recovery and subgrain growth during high-temperature annealing [26]. Second, the relatively thick IMC layer in A400/2h may restrict grain-boundary migration and the full growth of recrystallized grains, thereby stabilizing the LAGBs and subgrain structures produced by dislocation rearrangement. As the IMC layer becomes thicker, the constraints on grain-boundary migration and the interfacial pinning effect become stronger, suppressing recrystallization; this effect is more pronounced on the Al side [27]. Consequently, after A400/2h, static recrystallization occurs in Cu, whereas high-temperature recovery dominates in Al.



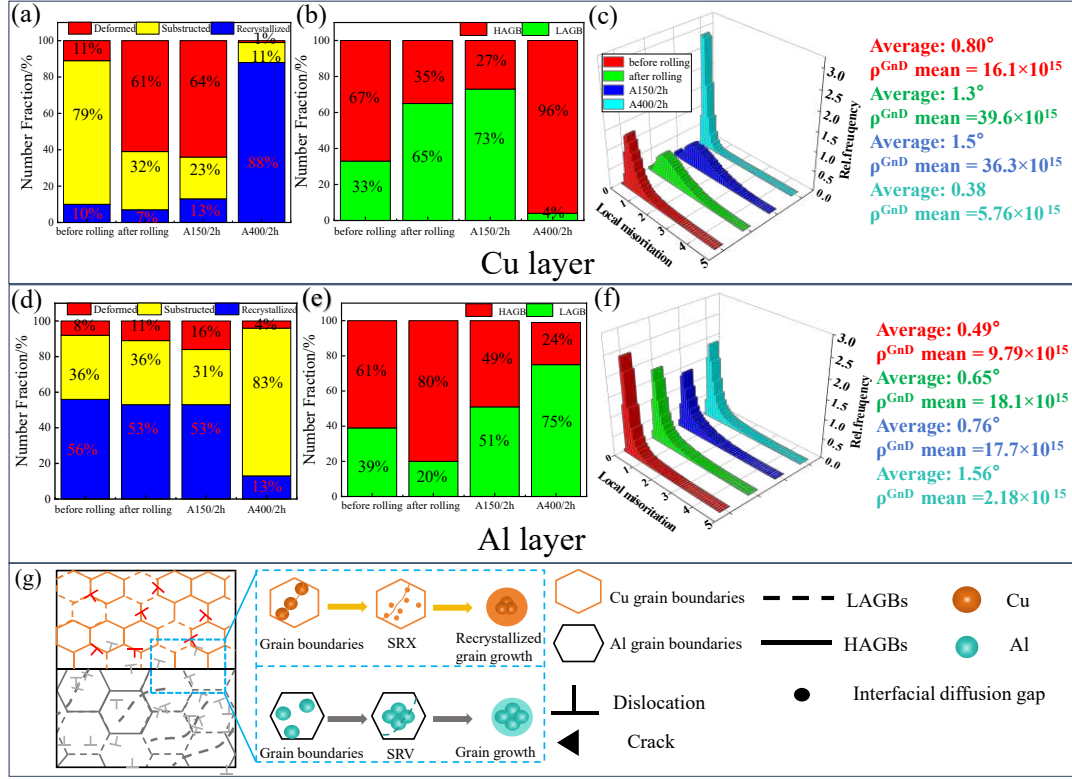
(a, f) IPF maps; (b, g) RX maps; (c, h) GB maps; (d, i) KAM maps

Fig. 4 EBSD maps before and after rolling



(a, f) IPF maps; (b, g) RX maps; (c, h) GB maps; (d, i) KAM maps

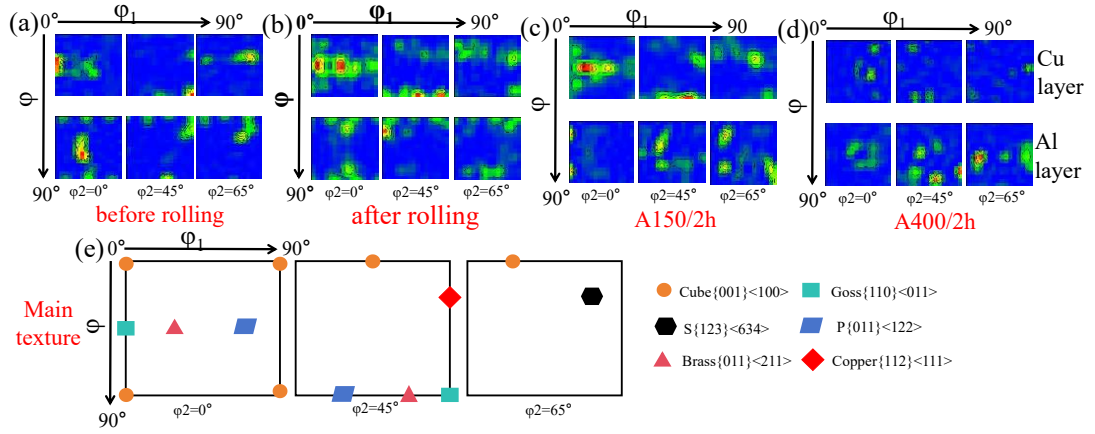
Fig. 5 EBSD maps of A150/2h and A400/2h



(a, d) recrystallization statistics; (b, e) HAGB/LAGB statistics; (c, f) GND maps; (g) schematic of the recrystallization mechanism during high-temperature annealing

Fig. 6 Recrystallization and HAGB/LAGB statistics, GND maps, and schematic of the high-temperature annealing mechanisms of Cu and Al in Cu/Al composite plates

Figure 7 shows the orientation distribution functions (ODFs) of the composite plates. Before rolling, the main texture components in Cu are the Goss $\{110\}<011>$ recrystallization texture and the S $\{123\}<634>$, Copper $\{112\}<111>$, and Brass $\{011\}<211>$ plane-strain textures. In Al, the main texture components are the Brass $\{011\}<211>$ and S $\{123\}<634>$ plane-strain textures and the Cube $\{001\}<100>$ recrystallization texture. After cold rolling, a P $\{011\}<122>$ texture forms on the Cu side. The Goss and Brass textures become stronger, whereas the Copper and S textures become weaker, indicating a transformation of the Copper and S textures toward the Goss and Brass textures. This evolution mainly arises because grains with the Copper orientation undergo lattice rotation around the $<110>$ crystal axis during large deformation [28-29]. The texture on the Al side also changes, leaving only a very weak Cube texture. This occurs because deformation of the Al layer is constrained by the Cu layer, disrupting the plane-strain state; the Brass and S textures present before rolling are therefore destroyed or weakened, and only a weak initial Cube texture remains [30]. After A150/2h, the texture on the Cu side remains essentially unchanged from that after rolling. The Cube texture remains in Al, with a slightly increased intensity. After A400/2h, the S, Copper, Goss, and Brass deformation textures in Cu disappear, and only a very weak P recrystallization texture remains. This is because almost all Cu grains are recrystallized after A400/2h [29], and the deformation structure is completely replaced by the recrystallized structure, resulting in the disappearance of the deformation textures. The texture in Al changes substantially, with the formation of a P recrystallization texture and a Brass plane-strain texture.



(a-d) ODFs of the Cu and Al sides of the Cu/Al composite plates; (e) distribution of ideal texture components for FCC metals

Fig. 7 ODFs of the Cu and Al sides and the distribution of ideal texture components for FCC metals

2.2 Mechanical properties and fracture surfaces

Figure 8 shows the engineering stress-strain curves of the tensile specimens along RD together with the UTS, YS, EL, and Vickers hardness. Rolling increases the hardness of both the Cu and Al layers. After rolling, the Cu hardness is approximately 120 HV, representing an increase of 71.4% relative to that before rolling, whereas the Al hardness is 43 HV, corresponding to an increase of approximately 33.3%. According to the rule of mixtures under the iso-strain assumption, the tensile strength of laminated composite plates can be approximated by the thickness-fraction-weighted sum of the strengths of the constituent layers [31-32]. Because the hardness of Cu is approximately 2.8 times that of Al, and because Cu has a lower SFE, weaker dynamic recovery during cold rolling, and substantially stronger work hardening than Al, the Cu layer is the primary load-bearing phase for strength in these composite plates, whereas the Al layer contributes less to the overall strength. The strength evolution of the composite plates is therefore analyzed below in terms of the evolution of the Cu microstructure and hardness. After rolling, the UTS, YS, and EL are 175 MPa, 142 MPa, and 15%, respectively. Relative to the condition before rolling, the UTS remains essentially unchanged, the YS increases by 56%, and the EL decreases. The pronounced increase in YS is mainly attributed to strong work hardening of the Cu layer, whereas the Al layer undergoes substantial dynamic recovery because of its high SFE and therefore exhibits weaker work hardening and makes only a minor contribution to the increase in YS [31].

After A150/2h, the UTS, YS, and EL are 194 MPa, 162 MPa, and 18%, respectively. Compared with the rolled condition, the UTS increases by 10%, while the YS and EL also increase slightly, demonstrating a simultaneous improvement in strength and ductility. The increase in strength results from the combined effects of enhanced interface bonding and residual-stress relaxation. First, large pore defects remain between the interface and the matrix after cold rolling, which facilitates premature delamination along the interface during tension and promotes stress concentration. After A150/2h, the pore size decreases and interface bonding is strengthened [33-34]. Second, the deformation incompatibility between Cu and Al during cold rolling produces substantial residual stresses and strain gradients in the composite plates. Low-temperature annealing relaxes the residual stresses and homogenizes the strain distribution, thereby reducing premature yielding and necking caused by local stress concentrations [33]. After A400/2h, the UTS, YS, and EL are 169 MPa, 92 MPa, and 40%, respectively, and the EL increases markedly. The Cu and Al hardness values

decrease to 73 HV and 32 HV, respectively. The substantial strength decrease of the composite plates is dominated by static-recrystallization softening of the Cu layer. After annealing at 400 °C, the recrystallized-grain fraction in Cu increases to 88%, extensive dislocation annihilation occurs, and the hardness decreases from 120 HV to 73 HV, a reduction of 39%. Consequently, the Cu layer, which is the primary load-bearing phase for strength, is strongly softened. Although high-temperature recovery also occurs in Al and its hardness decreases from 43 HV to 32 HV, Al is intrinsically a secondary contributor to strength and therefore has a limited influence on the overall strength decrease. The increase in EL is attributed to the recovery of ductility through Cu recrystallization and Al recovery, together with the improved interface bonding state.

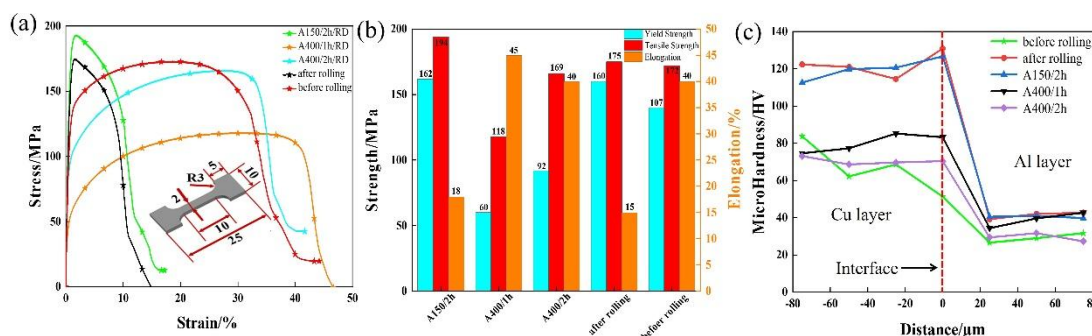


Fig. 8 Engineering stress-strain curves, UTS, YS, EL, and Vickers hardness of Cu/Al composite plates

As shown in Fig. 9, before rolling the fracture surface on the Al side exhibits pronounced necking, with a necking ratio of 86.7%. The surface consists of cleavage facets and numerous dimples. The dimples are relatively large but sparse, and the area ratio of cleavage facets to dimples is approximately 1:1, indicating clear mixed-fracture characteristics [34]. Pronounced necking is also observed on the Cu side, with a necking ratio of 88.2%. The fracture surface is composed almost entirely of continuous cleavage facets with only a few dimples, indicating brittle fracture. The interface initially delaminates from the bonding surface on the Al side and remains mostly attached to the Cu side after delamination, indicating a higher bonding strength with Cu. After cold rolling, the Al-side fracture surface exhibits quasi-cleavage characteristics, with a necking ratio of 76.8%. No clear ductile-fracture features are observed, indicating a transition in Al from mixed fracture to brittle fracture [34]. The Cu side remains dominated by large cleavage facets with sparse dimples, indicating that the fracture mode remains brittle. As shown in Fig. 9b4, the necking ratio is 76.2%. Delamination still initiates between the interface and the Al side, as before, and most of the interface remains attached to the Cu side after delamination, again indicating stronger bonding with Cu.

For A150/2h, the Al-side fracture surface consists of cleavage facets and some dimples. Compared with the condition before rolling, the dimples are deeper and denser but slightly smaller. The necking ratio on the Al side is 85.2%, indicating that Al accommodates more deformation, possibly because annealing releases part of the stress introduced by cold rolling. The Cu side remains dominated by a continuous cleavage facet containing a small number of pore defects, and the necking ratio is 80.6%, also indicating improved deformability. Two delamination features are observed at the interface: separation occurs from both the Al and Cu sides. This suggests that bonding between the interface and Al has improved, which is favorable for plastic deformation. However, a larger fraction of the interface remains attached to the Cu side, indicating that the bonding strength on the Cu side remains higher. After A400/2h annealing, the Al-side fracture surface is composed mainly of quasi-cleavage facets and exhibits more pronounced necking, with a

necking ratio of 91%. In addition, a large amount of the interfacial layer remains attached to the Al side, indicating increased bonding strength, which contributes to the higher EL. The Cu-side fracture surface is barely observable, indicating extremely pronounced necking. The deformability of Cu is effectively improved after annealing at 400 °C, possibly owing to the onset of recrystallization. Parts of the interface also remain attached to Cu, indicating that the bonding strength on the Cu side remains high. The interface still begins to delaminate from the Al side, indicating that bonding at the Cu-side boundary remains stronger [35]. This observation is consistent with the diffusion of Cu toward Al and the greater interface thickness.

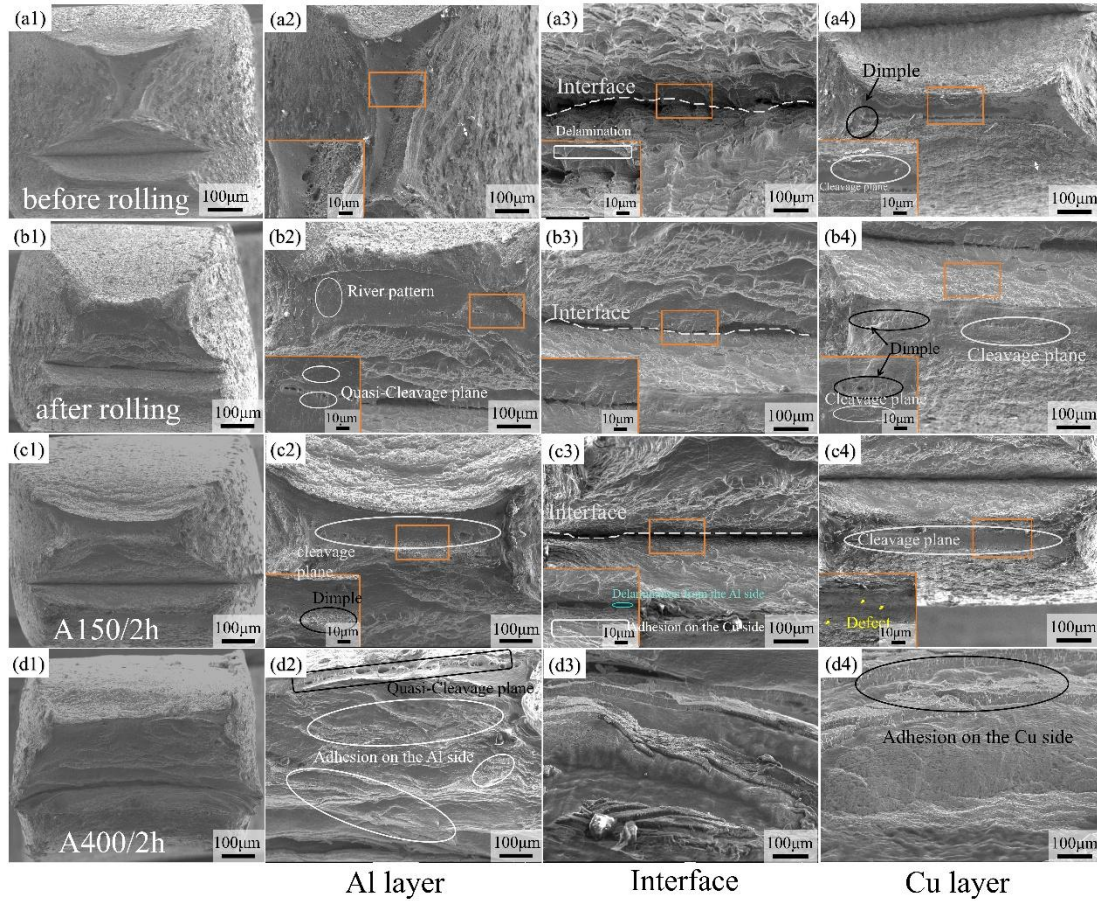


Fig. 9 Tensile fracture surface of Cu/Al composite plates

3 Conclusions

The microstructural evolution of the constituent layers and the interface in Cu/Al composite plates in the rolled and annealed conditions, together with the associated mechanisms governing the mechanical properties, is investigated. The main conclusions are as follows:

1) Cold rolling disrupts the initially continuous interface structure, whereas annealing promotes restoration of the interface and formation of IMCs. After cold rolling, the interface fractures along RD and its thickness decreases to 2.3 µm. After annealing at 150 °C, the interface remains fractured and only Cu₉Al₄ and CuAl are formed. Annealing at 400 °C markedly promotes Cu-Al diffusion, producing a continuous three-layer interface composed of Cu₉Al₄, CuAl, and CuAl₂, with the interface thickness increasing to 8.7 µm.

2) Cu and Al exhibit different microstructural evolution mechanisms during annealing because of their intrinsic material characteristics and the constraints imposed by the interface. During cold rolling, a higher dislocation accumulation and deformation stored energy develop on the Cu side, whereas stronger recovery occurs in Al because of its high SFE. Annealing at 150 °C mainly promotes recovery and dislocation rearrangement in both Cu and Al. After annealing at 400 °C, pronounced static recrystallization occurs on the Cu side, producing a microstructure dominated by recrystallized grains, whereas the Al side is constrained by the interfacial IMC layer and is dominated by recovered subgrains.

3) Annealing changes the texture evolution of the composite plates. Cold rolling promotes the transformation of the Copper and S textures toward the Goss and Brass textures on the Cu side, while weakening the initial textures on the Al side. Low-temperature annealing largely preserves the existing texture characteristics. After high-temperature annealing, the deformation textures on the Cu side disappear and only a weak P recrystallization texture remains, whereas P recrystallization and Brass textures form on the Al side.

4) After rolling, the UTS and YS of the composite plates increase, with UTS, YS, and EL values of 175 MPa, 142 MPa, and 15%, respectively, while the Cu and Al hardness values are 120 HV and 43 HV. Low-temperature annealing produces a simultaneous improvement in strength and ductility, whereas high-temperature annealing mainly improves ductility. Annealing at 150 °C promotes the healing of interfacial pores, strengthens interface bonding, relaxes residual stresses, and homogenizes the internal stress field. Consequently, the UTS, YS, and EL reach 194 MPa, 162 MPa, and 18%, respectively, yielding optimized overall mechanical properties. After annealing at 400 °C, static recrystallization strongly softens the Cu layer, which is the primary load-bearing phase for strength, and its hardness decreases from 120 HV to 73 HV. This softening is the principal cause of the reductions in UTS and YS to 169 MPa and 92 MPa, respectively, whereas recovery of the Al layer makes only a minor contribution to the strength decrease. Meanwhile, Cu recrystallization and Al recovery restore ductility, increasing EL to 40% and producing good plasticity.

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