Research on Cu/stainless steel composite thin strips by two-pass cold roll-bonding with intermediate annealing

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Research Article

Keywords: Cu/SS composite thin strips, Intermediate annealing, Deformation behaviors, Finite element simulation, Bonding mechanisms

Posted Date: August 28th, 2023

DOI: https://doi.org/10.21203/rs.3.rs-3284263/v1

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Abstract

Roll-bonding is feasible to fabricate T2 copper (Cu)/ Stainless steel (SS) composite thin trips, which have great potential in micromanufacturing, robotics, aerospace, and other applications. However, the effective bonding of Cu and SS could be hindered by limited diffusion of the elements and uncoordinated deformation of the metal matrices. In this study, Cu/SS composite strips with 0.24mm thickness were prepared by the two-pass cold rolling process with intermediate annealing at 400 ~ 1000°C. The influences of the intermediate annealing process on the tensile and peeling strength were investigated. Finite element simulation and microstructure evaluation were carried out to analyze the deformation behaviors and bonding mechanisms of the strips. The results indicate that the deformation coordination in the second-pass rolling and the bonding strength were improved by appropriate intermediate annealing processes. The difference in the deformation resistance between Cu and SS became the lowest by intermediate annealing at 1000°C, while the deformation of Cu and SS was severely uncoordinated by annealing at 600°C. The peel strength and elongation of the strips annealed at 1000°C were 11.65 ± 0.7N/mm and 34.8 ± 1.3% after the second-pass rolling, which were 79.23% and 6.64 times higher than the strips manufactured without intermediate annealing, respectively. In this work, Cu/SS thin strips with high bond strength and ductility were successfully manufactured by appropriate intermediate annealing process, and the bonding mechanisms were systematically discussed.

1 Introduction

Laminated bimetal composite strips can give full play to the advantages of the constituent materials [1, 2]. Copper (Cu) /stainless steel (SS) composite strips have excellent thermal and electrical conductivity of Cu, as well as high strength and corrosion resistance of SS [3, 4]. Cu/SS thin strips has great potential in micromanufacturing, robotics, aerospace, and other applications. Therefore, reliable and efficient manufacturing of Cu/SS composite strips is urgently needed.

Welding techniques are widely used to fabricate Cu/SS composite strips. The thermal properties of Cu and SS are significantly different, which hinders the joining of the materials in conventional welding processes (e.g., friction welding, ultrasonic welding, and laser bonding) [5]. The large differences in the thermal properties could result in hot cracks, abnormal heat flow, and high residual stresses in the welds [6]. Explosive welding is used to fabricate composite plates by the quick and large deformation owing to high pressure and high temperature transmitted to the collision points [7–9]. Zhang et al. [10] studied the interfacial microstructure of Cu/steel composite produced by explosive bonding. The transition layers between the copper and steel had thicknesses varying from 50µm in vortex regions to 5µm in solid-solid bonding regions. In another study by Bina et al. [11], copper/SS304L composites with high strength and ductility were fabricated via explosive welding. The wavy interfaces were observed in the composites, and the elemental diffusion at the interfaces was enhanced by heat treatment.

However, the energy released by explosives could easily melt or break the metals. Thus, explosive welding is unsuitable for manufacturing Cu/SS thin strips [12]. Although diffusion bonding can be used to fabricate composite strips, the productivity is usually limited. The production of thin composite strips by rolling is one of the most efficient way for manufacturing composite thin strips [13]. It has been reported that roll bonding with adequate heat treatment is an efficient method to produce Cu/SS composite strips [14].
In roll bonding processes of Cu/SS, the interfacial bonding state and deformation behavior of the component materials need to be controlled to improve the bonding strength of the composite. Yang et al. [15] reported a diffusion-rolling procedure of Cu/Fe multilayered composites. Cu and Fe were bonded together by diffusion bonding and then cold-rolled to enhanced the bonding strength. Khalid A et al. [16] investigated the influence of annealing temperature on the mechanical properties of Cu-steel-Cu rolled composite sheet. The study shows that as the annealing temperature increased from 0°C to 700°C, the stamping formability improved, and the bonding strength decreased. The compressive stress at the interface was conducive to the bonding of the composite sheet. Large differences in the deformation behaviors of the component materials could result in insufficient bonding strength. Xiao et al. [17] prepared Ti/Al clad plates by applying different rolling temperatures. The results show that increasing the temperature of the Ti layer benefits the deformation coordination of Ti and Al. Wang et al. [18] prepared Cu-Al bimetallic composite plates with a thickness of 2mm by a horizontal continuous composite casting-rolling process. For cold rolling, the first-pass reduction should exceed 60% to avoid delamination cracking. When the rolling temperature rises to 200–300°C, the first-pass reduction can be decreased to 55% due to better deformation coordination of the component materials.

So far, there have been few researches on the rolling of Cu and SS composite strips. The effective bonding of Cu and SS could be hampered by limited diffusion of the elements and uncoordinated deformation of the metal matrices. In this work, Cu/SS composite strip with a thickness of 0.24mm was fabricated by a two-pass cold rolling method with intermediate annealing. Appropriate intermediate annealing was beneficial to the bonding strength. The effects of the processing parameters on deformation coordination, interface status, and bonding strength were systematically investigated.

2 Experimental and finite element simulation

2.1 Experimental method

The chemical compositions of the SS and Cu strips are listed in Table 1. The original thickness of the SS strips was 0.2 mm, while the Cu strip was 0.3 mm. Before the roll bonding processes, the SS strips were annealed at 1050°C for 4h, and the T2 copper strips were annealed at 260°C for 1h.

<table>
<thead>
<tr>
<th>materials</th>
<th>C</th>
<th>Cr</th>
<th>Ni</th>
<th>S</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>O</th>
<th>Fe</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>SUS304</td>
<td>0.08</td>
<td>17.06</td>
<td>8.05</td>
<td>0.02</td>
<td>0.46</td>
<td>1.32</td>
<td>0.03</td>
<td>—</td>
<td>Bal.</td>
<td>—</td>
</tr>
<tr>
<td>T2</td>
<td>—</td>
<td>—</td>
<td>≤ 0.005</td>
<td>≤ 0.005</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>≤ 0.06</td>
<td>—</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

As shown in Fig. 1. The strips were grinded by a T-shaped wire brush before assembling the billets. After the grinding processes, the hardening layer was formed on the surfaces to be bonded. The hardening layer on the SS was distributed in a long strip shape along the grinding direction. The billets were bound with aluminum wire after the surface treatment. The rolling experiment was carried out on a twin-roll mill. The Cu/SS composite strips were rolled and bonded by two-pass cold rolling processes with intermediate annealing at different temperatures (400–1000°C). The parameters of the roll-bonding processes are listed in Table 2. The
reduction rate of the first-pass and second-pass cold rolling were 40% and 20%, respectively. The diameter of the rollers was 160mm and the rolling speed was 53.6mm/s. The temperatures of intermediate annealing were selected as 400°C, 600°C, 800°C and 1000°C and the electrical resistance furnace was filled with argon as a protective atmosphere. In this paper, the specimens manufactured using those annealing temperatures were named as “1P”, “1P-400”, “1P-600”, “1P-800”, “1P-1000”, “2P”, “2P-400”, “2P-600”, “2P-800”, and “2P-1000”, where “1P” and “2P” means the specimens manufactured by first-pass and second-pass rolling.

<table>
<thead>
<tr>
<th>Intermediate annealing temperature(°C)</th>
<th>Roll speed (mm/s)</th>
<th>Roll diameter (mm)</th>
<th>First-pass rolling</th>
<th>Second-pass rolling</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>Reduction rate</td>
<td>After intermediate annealing</td>
</tr>
<tr>
<td>No annealing</td>
<td>53.6</td>
<td>160</td>
<td>40%</td>
<td>1P</td>
</tr>
<tr>
<td>400</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>1P-400</td>
</tr>
<tr>
<td>600</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>1P-600</td>
</tr>
<tr>
<td>800</td>
<td></td>
<td></td>
<td></td>
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</tr>
<tr>
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<td></td>
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<tr>
<td>1000</td>
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</tr>
</tbody>
</table>

Tensile and peeling tests were carried out to evaluate the overall and bonding strengths of the SS/Cu composite strips. The dimensions of the specimens are shown in Fig. 2. The tensile direction was the rolling direction (RD). According to the instructions of ASTM-E8M-15[19] and ASTM-D903-93[20] standards, the speeds of tensile and peeling tests were selected as 0.3 mm/min and 25 mm/min. The average peel strengths are calculated from Eq. (1)[1, 21, 22].

\[
\text{Average peel strength} = \frac{\text{average load}}{\text{bond width}} \, (N/mm) \quad (1)
\]

Three specimens at each manufacturing condition were tested in order to obtain the average values. The cross-sectional Vickers microhardness within the surface layer was measured by an HVT-1000 micro-Vickers hardness tester for a dwell time of 10 s, the Cu was employed a 50 g load and the 304 SS was employed a 300 g load.

The interface and microstructure of the composite strips were observed by optical microscopy (OM) and scanning electron microscopy (SEM). The element distribution at the bonding surfaces and the interface was analyzed by energy dispersive spectrometry (EDS). For the microstructural observation, the copper was etched using 10 g FeCl₂, 30 ml HCl and 120 ml H₂O for 10 s; the stainless steel was etched using 25 ml HCl, 50 ml 10% CrO₂ aqueous solution.

### 2.2 Numerical simulation method

Finite element (FE) simulation was carried out to calculate the deformation behaviors of SS and Cu in the second-pass rolling. In order to prevent excessive complications and to reduce the calculation time, the roll
process was conducted in 2D Dynamic/Explicit due to its considered to be a plane strain problem [23] and the sheet length was reduced to 6 mm (the complete rolling process was guaranteed without affecting the calculation results). The meshing diagram of the finite element model is shown in Fig. 3. The coefficient of friction between the roll and the Cu layer, the roll and the SS layer were set to 0.15,0.3, respectively [24, 25]. The interface of Cu and SS was set to bonded state. The geometric models were meshed with CPE4R meshes, the Cu layer contained 1800 elements and the SS layer contained 1400 elements. After the first-pass rolling, the thickness of SS and Cu was 0.17mm and 0.13mm, respectively, and the intermediate annealing treatment was carried out under different conditions. The properties of the materials were set based on the experimental data in Table 3.

<table>
<thead>
<tr>
<th>Material</th>
<th>Dimensions(mm)</th>
<th>Sample</th>
<th>Hardness (HV)</th>
<th>Yield strength (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>SS</td>
<td>0.13x6</td>
<td>1P</td>
<td>324.69</td>
<td>1161.21</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1P-400</td>
<td>316.97</td>
<td>1083.49</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1P-600</td>
<td>312.81</td>
<td>1053.18</td>
</tr>
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<td></td>
<td></td>
<td>1P-800</td>
<td>140.85</td>
<td>631.58</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1P-1000</td>
<td>132.32</td>
<td>566.76</td>
</tr>
<tr>
<td>Cu</td>
<td>0.17x6</td>
<td>1P</td>
<td>86.95</td>
<td>323.48</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1P-400</td>
<td>66.91</td>
<td>222.28</td>
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<td></td>
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<td>1P-600</td>
<td>46.34</td>
<td>183.88</td>
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<tr>
<td></td>
<td></td>
<td>1P-800</td>
<td>41.49</td>
<td>161.50</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1P-1000</td>
<td>37.14</td>
<td>158.87</td>
</tr>
</tbody>
</table>

3 Results

3.1 Analysis of the thickness ratios and rolling force

In addition, the unit width rolling force of Cu/SS composite strip simulation results under various conditions was extracted, as shown in Fig. 5. It can be deduced that by increasing the intermediate annealing temperature, the required rolling force for the second-pass rolling decreases.

3.2 Evolution of the Cu/SS interface

Figure 6 shows the SEM images of the Cu/SS interface. As shown in Fig. 6(a) 1P, the discontinuous hardening layer was observed at the composite interface formed after rolling. The broken hardening layer of SS destroyed the surface of the Cu matrix under the action of the rolling force, and the fresh Cu matrix was
bonded with the fresh SS matrix by the first-pass rolling, forming a mechanical bonding at the interface. There were numerous pores at the interface (marked in blue dotted circles in Fig. 6(a)), indicating the metallurgical bonding area of the SS and Cu strip at the interface was relatively small. Hence, the bonding of Cu/SS strips by first-pass cold rolling was weak.

Intermediate annealing has a dramatic effect on eliminating the pores at the interface. The effect was more pronounced at higher annealing temperatures. As shown in Fig. 6(d) and (e), the unbonded pores are completely eliminated by intermediate annealing at 800 and 1000 °C and improved the interface bonding status.

As shown in Fig. 6(f-j), the hardening layer of SS at the composite interface is further broken and squeezed into the Cu matrix after the second-pass rolling. Hence, the mechanical bonding of the interface was strengthened. In the second-pass rolling process, when the deformation of the composite strip matrix was uncoordinated seriously, and it led to cracks between the hardening layer and the steel matrix, which was detrimental to the bonding strength. As shown in Fig. 6(f), (g) and (h), cracks can be found between the hardening layer forming mechanical bonding and the SS matrix near the interface of specimens 2P, 2P-400 and 2P-600. 2P-600 specimens have a higher number of cracks compared with 2P and 2P-400. However, there are no obvious cracks of SS in the 2P-800 and 2P-1000 specimens (Fig. 6(i)(j)).

EDS line scanning was carried out on the interface and the results are shown in Fig. 7. As shown in Fig. 7(a) 1P, the diffusion layer depth of 1P is 1 µm. As shown in Fig. 7(b), the thickness of the diffusion layer slightly increases as the annealing temperature increases. The diffusion layer after intermediate annealing at 1000 °C was 1.2 µm. This was due to the gradual occurrence of thermal excitation of atoms and the diffusion of atoms was promoted at the interface as the temperature increased [3, 26, 27]. The inter-solubility between Fe and Cu elements is relatively low and no obvious intermetallic compound was found at the interface. The thickness of the diffusion layer was less than 3µm under different process parameters. Therefore, the influence of the element diffusion layer on the bonding strength in the intermediate annealing and the second-pass rolling processes were basically identical [28].

### 3.3 Peel strength and interface bonding mechanism

There were more pores at the bonding interface after first-pass rolling and the bonding strength of the composite strips was small. As shown in Fig. 8, both the intermediate annealing and the second-pass rolling processes have a significant influence on the peel strength. On the one hand, 1P-800 and 1P-1000 specimens had higher peel strength than 1P-400 and 1P-600, while 1P-600 specimens had the low peel strength (2.7 ± 0.46N/mm); on the other hand, the peeling strength increased after the second-pass rolling. At the intermediate annealing temperature of 600°C, the residual stress was eliminated, resulting in a weakened mechanical bonding. Hence, 1P-600 showed the lowest peeling strength. The most significant improvement of the peel strength by the second-pass rolling was achieved at intermediate annealing of 1000°C. The highest peel strength was achieved in 2P-1000 specimens (11.65 ± 0.70N/mm), which is 5.30N/mm higher than 1P-1000 specimens. The detailed discussion is given in Section 4.2.
The post-peeling morphology of Cu/SS was observed. EDS elemental analysis was performed on the Cu element of the peeled SS surface. As shown in Fig. 9(a) 1P, the EDS scan shows that the content of Cu bonded on the SS side is 5.7 at.% after the first-pass rolling. The Cu left on the peeled surface of the steel side was the area that formed mechanical bonding and metallurgical bonding.

The Cu content on the SS side of different conditions is shown in Fig. 9(b). The content of Cu bonded on the SS side increased slightly after intermediate annealing at 400°C (10.1 at.%) and 600°C (7.0 at.%), while there were dramatic increases of the Cu content at the peeled surfaces of 1P-800 (21.7 at.%) and 1P-1000 (30.9 at.%). The bonding strength of 1P-800 and 1P-1000 composite strips was higher. As shown in Fig. 9(c) 1P-800, when the intermediate annealing temperature was 800 °C, flakes of Cu appeared on the peeled surface of the SS side. As shown in Fig. 9(d) 1P-1000, the fresh SS metal exposed on the surface was almost completely covered by Cu, and the adhered Cu surface contained more dimples and small Cu ridges [29].

After the second-pass rolling process, the bonding strength of the composite strips were further improved than that after intermediate annealing, but the bonding strength was almost unchanged when the intermediate annealing temperature was 600°C, which will be discussed in Section 4.2. As shown in Fig. 9(e) 2P-1000, after the second-pass rolling, the number and size of Cu ridges increased and the Cu content reached 39%. The bonding strength was the highest.

### 3.4 Microstructure evolution

As shown in Fig. 10(a) 1P, the Cu and SS grains were deformed grains elongated along RD by the first-pass rolling. After intermediate annealing at 400°C, the Cu in 1P-400 specimens recrystallized and the grain size increased to around 20µm, forming a typical recrystallization microstructure with annealing twins (Fig. 10(b) 1P-400). The grain size of Cu further increased to around 40µm in 1P-600 specimens, while there was no significant change of the SS deformed grains and they were fiber grains (Fig. 10(c) 1P-600). By intermediate annealing at 800°C, the SS recrystallized to equiaxed recrystallized grains, while the Cu grains increased significantly compared with 600°C, resulting in a single-layer grain structure (Fig. 10(d) 1P-800). The SS fully recrystallized when the annealing temperature was 1000°C. The grain size of SS was around 40µm in 1P-1000 specimens (Fig. 10(e) 1P-1000). As shown in Fig. 10(f) 2P-1000, the grains of both SS and Cu were elongated along RD.

### 3.5 Tensile properties and fracture mechanism

As shown in Fig. 11. The tensile properties of the Cu/SS strips, including ultimate tensile strength (UTS) and elongation, were significantly influenced by both intermediate annealing and the second-pass rolling. As shown in Fig. 11(a), the UTS of the Cu/SS composite strip after intermediate annealing slightly decreases with the increase of annealing temperature, but the elongation significantly increases. The maximum elongation of the 1P-1000 sample was 78 ± 0.9%, and the UTS was 405.2 ± 3.59 MPa. After intermediate annealing at 1000 °C, the recrystallization structure formed by intermediate annealing weakens work hardening, and higher elongation was obtained. As shown in Fig. 11(b), the UTS of the composite strip decreases slightly as the annealing temperature increases. After the second-pass rolling, the elongation of the 2P-1000 specimens decreased to 34.8 ± 1.3% and the UTS increased to 535.7 ± 6.3 MPa.
The fracture morphology of the composite strips after intermediate annealing and the second-pass rolling was observed. Interfacial delamination can be found at the fracture zones. There were obvious differences between the Cu and the SS when the composite strip was fractured. The fracture morphology of the Cu side of the composite strip was basically the same, but the difference in the SS was great. The recrystallization temperature of Cu is low, and the grains appear equiaxed at a low intermediate annealing temperature. The recrystallization temperature of steel is high, and it is still in the recovery stage at a low temperature. According to the fracture morphology of the Cu side in Fig. 12(a) 1P-600, when the intermediate annealing was 600°C, it can be found that the Cu fractures upward from the interface, a whole tearing edge and a tearing ridge were formed at the edge. During the fracture process, great plastic deformation was produced, the section shrinks and the fracture shape was close to the tip of the cone. It indicated that pure shear ductile fracture occurs in the Cu layer of the composite strip. The grain on the SS presented a fibrous state, which was mainly a brittle fracture. As shown in Fig. 12(b) 1P-1000, when the intermediate annealing was at 1000°C, the grain on the steel side began to recrystallize, and the fracture dimples increased, mainly manifested as a ductile fracture. As shown in Fig. 12(c) (d), After the second-pass rolling, the composite interface was more closely combined and the delamination degree of the composite thin strip was reduced.

It is found that the tensile strength and elongation at the break of the composite strip are mainly related to the side structure of the steel by analyzing the tensile properties and tensile fracture of the composite strip after intermediate annealing and the second-pass rolling.

4 Discussion

4.1 Effect of the intermediate annealing on the composite interface

By the first-pass rolling process, the hardening layer of SS was broken and Cu was squeezed into the cracks, forming a bonding interface with residual compressive stress. Intermediate annealing has different impacts on the mechanical bonding and metallurgical bonding of the interface.

4.1.1 Effect of the intermediate annealing on the mechanical bonding

The first-pass rolling led to the mechanical bonding of Cu and the broken hardening layer of SS. The mechanical bonding introduces residual stress to the interface and promotes the initial bonding of the interface. However, the residual compressive stress was effectively eliminated by the intermediate annealing, and the mechanical bonding of the interface was relaxed. Hence, the intermediate annealing is detrimental to the mechanical bonding effect. At lower annealing temperatures (400°C and 600°C), the mechanical bonding introduced by the first-pass rolling was effectively eliminated, while the metallurgical bonding was not significantly pronounced [30]. Thus, as shown in Fig. 13, the peel strength of the 1P-400 and 1P-600 specimens was lower than 1P-800 and 1P-1000. The 1P-600 specimens had the lowest peel strength since the residual stress was sufficiently eliminated. This is in agreement with the previous work reported by Al-Ghamdi and Hussain [16, 31].
4.1.2 Effect of the intermediate annealing on the metallurgical bonding

During the cold rolling and annealing processes, elements in the Cu and SS matrices diffused at the interface, leading to the metallurgical bonding of the strips. A number of pre-bonding points formed by mechanical bonding have been produced by the first-pass rolling process. The atoms at the vicinities of the pre-bonding points have been activated by the intermediate annealing treatment, and new bonding points were produced by the mutual diffusion at the interface. Thus, as shown in Fig. 6, the unbonded areas (pores) at the interface of the 1P specimens are effectively eliminated by the intermediate annealing.

According to classical kinetics, diffusion is typically modeled using Fick's first law, which emphasizes the flux of diffusing material is proportional to the diffusion coefficient \(D\). The relationship between diffusion coefficient \(D\) and temperature \(T\) is in accordance with the Arrhenius equation [32] [33]:

\[
D = D_0 \exp \left( \frac{-Q}{RT} \right)
\]

where \(D_0\) is the diffusivity constant, \(Q\) is the activation energy for diffusion, \(R\) is the gas constant and \(T\) is the absolute temperature. According to the relationship above, an increase in temperature can lead to an exponential increase in the diffusion coefficient, which makes the degree of atomic interdiffusion more intense.

The diffusion is more pronounced at higher annealing temperatures. Hence, the peel strength and the Cu content on the peeled surfaces was increased after intermediate annealing. With the increase of intermediate annealing temperature, the diffusion area at the composite interface increased. After the first-pass rolling, The Cu content on the surface of the SS strip was 5.7%, mainly distributed in the crack of the fractured hardening layer. After intermediate annealing at 1000°C, the Cu content on the surface of the SS strip was 39%.

Intermediate annealing temperature has a weak effect on the thickness of the diffusion layer. Cu-Fe exists in the form of limited solid solubility of Fe in Cu (\(\varepsilon\) phase) and Cu in Fe (\(\alpha\) phase), and no intermetallic phases was produced [34].

In summary, the bonding strength of the Cu/SS strips was influenced by the mechanical bonding and metallurgical bonding effect. The residual compressive stress at the interface produced by the first-pass rolling was eliminated by the annealing processes. As a result, the mechanical bonding was weakened as the annealing temperature increased. Meanwhile, the diffusion phenomenon at the interface was enhanced by increasing the annealing temperature, and the metallurgical bonding was enhanced consequently. As shown in Fig. 13, the bonding strength of 1P-600 reduced, since the mechanical bonding was relaxed and the metallurgical bonding was not significantly pronounced. The highest bonding strength was achieved in the 1P-1000 specimens, where the metallurgical bonding had a significant effect.
4.2 Effect of intermediate annealing on deformation coordination

4.2.1 Evolution of the matrices’ microstructure and properties

The microstructure and consequent mechanical properties of the Cu and SS strips were significantly influenced by the annealing process, and annealing treatment can eliminate the strain hardening caused by rolling and improve the microstructure. As shown in Fig. 14, the grains in the SS strips are elongated by the first-pass rolling, and the micro-Vickers hardness is relatively high. This is because the strain and dislocation density imposed by rolling are large [35–37]. The recovery and recrystallization of deformed grains in the SS strips were not completed by annealing at 400 and 600°C. Unlike the SS materials, the recrystallization temperature of Cu is significantly lower. Hence, the Cu materials were completely recrystallized by the intermediate annealing processes. The heat-treated Cu had equiaxed grains, whose size increased with the annealing temperature. The deformed grains in the SS strips fully recrystallized after intermediate annealing at 800 and 1000°C. For these reasons, the difference in the micro-Vickers hardness of the Cu and SS was dramatically smaller after annealing at the higher temperatures (800 and 1000°C). After annealing at 600°C, the difference in the hardness was the largest, indicating the deformation of the two materials became the most discordant.

4.2.2 Deformation behavior of composite strip

The cross shear zone is the area between the two junctions of the front and rear sliding zones of the component metal contact surface up-roll and down-roll (neutral layer with a friction shear stress of 0). According to the film theory, the presence of the cross shear zones is conducive to the strong initial bonding of dissimilar materials [38]. As shown in Fig. 15, the hardening layers on SS were fractured under the action of shear force during the first-pass rolling, exposing fresh metal and forming initial mechanical bonding. The radial rolling force in this area almost reaches the maximum, and the metal atoms form metallurgical bonding under the extrusion [39].

After the first-pass rolling, an initial bonding was formed at the interface of the composite strip. During the second-pass rolling, due to the different deformation resistance of Cu and SS, there was still frictional shear force. The frictional shear force causes cracks between the hardening layer and the matrix, damaging the original mechanical bonding and metallurgical bonding, which was not conducive to further improvement of the bonding strength. The cloud diagram of friction Shear stress under rolling condition 2P is shown in Fig. 16 (a). It can be seen that there were areas with zero friction Shear stress in the contact area between Cu/SS and roll, and the two areas with zero friction stress are not in the same vertical plane. As shown in Fig. 16(b), the size of the cross shear zone is 38.8µm. At different intermediate annealing temperatures, the deformation resistance between the two metals was different, and the size of the cross shear zone at the interface was different. As shown in Fig. 16(c), after intermediate annealing at 600 °C, the difference in deformation resistance between Cu and SS is the largest and the length of the friction shear zone is the largest.

4.3 Bonding mechanism of Cu/SS composite strips
The deformation and bonding mechanism of this process is summarized, as shown in Fig. 17.

During the first-pass rolling, the deformation of the two metals was not coordination and a strong shear friction was generated at the interface. The hardening layer of SS was broken and cracks were formed on the surface. The fresh Cu matrix was exposed to bond the fresh SS matrix under the action of the rolling force, forming a mechanical bonding at the interface [40]. After the first-pass rolling, there were residual compressive stress and numerous pores at the interface. The bonding mechanism was mainly mechanical bonding and a weak metallurgical bonding. Hence, the bonding strength of the composite strips were weak. The intermediate annealing eliminated the residual compressive stress and thus weakened the mechanical bonding. The pores were eliminated and the element diffusion was promoted by intermediate annealing, resulting in the increase of metallurgical bonding area and enhancement of metallurgical bonding. Intermediate annealing 800°C and 1000°C composite strips had higher peel strength than 400°C and 600°C. The microstructure and consequent mechanical properties of the Cu and SS strips were significantly influenced by the intermediate annealing, and the deformation coordination of Cu and SS materials were changed in the second-pass rolling process. The thickness of the composite strip was reduced and the bonding strength increased after the second-pass rolling. However, when the deformation resistance difference between the two metals was large, a large cross-shear zone will be formed at the interface of the composite strip, which had negative influence on further improving the bonding strength of the composite composite strip. When the intermediate annealing was 600°C, the deformation resistance of Cu and SS was great difference, and the bonding strength was almost unchanged after the second-pass rolling. In summary, appropriate intermediate annealing effectively improved the bonding strength by improving the interface bonding state after the first-pass rolling and coordinating metals deformation in the second-pass rolling.

5 Conclusions

In this work, Cu/SS composite thin strips were manufactured by a two-pass cold rolling technique with intermediate annealing processes. Intermediate annealing enables the effective control of the interfacial status and deformation behavior during subsequent rolling. The conclusions are as follows:

(1) Appropriate intermediate annealing processes between the two passes could effectively improve the bonding strength and ductility of the Cu/SS thin strips. By intermediate annealing at 1000°C, the peel strength and elongation of the strips after second-pass rolling were (11.65 ± 0.70N/mm) and (34.8 ± 1.3%), which are 79.23% and 6.64 times higher than the strips manufactured without annealing, respectively.

(2) The interfacial status after the first-pass rolling was effectively controlled by the intermediate annealing. The interfacial bonding strength was resulted from the competition of the mechanical locking and metallurgical bonding. The intermediate annealing processes led to the relaxation of the mechanical bonding. At the lower intermediate annealing temperatures (400°C and 600°C), the mechanical bonding was significantly eliminated, while the metallurgical bonding was more pronounced at the higher intermediate annealing temperatures (800°C and 1000°C).

(3) The microstructure and consequent deformation behaviors of the Cu and SS matrices in the second-pass rolling process were controlled by the intermediate annealing. At the lower intermediate annealing temperature
(400°C and 600°C), Cu has completed recrystallization and the microstructure of SS was still deformed grains. At this point, there was a significant difference in deformation resistance between Cu and SS strips, resulting in large cross shear zone and low bonding strength. By annealing at 1000°C, the difference in the deformation resistance was the lowest, which was beneficial to the bonding strength.

**Declarations**

**Funding**

This work is financially supported by National Natural Science Foundation of China (project No. U22A20188, No. 52105391, No. 51974196); Shanxi Provincial Fundamental Research Program (No. 20210302124321); Natural Science Foundation of Shanxi Province (No. 20210302124426).

**Authors’ contributions**


**Ethics approval**

Not applicable.

**Consent to participate**

Not applicable.

**Consent for publication**

Not applicable.

**Competing Interests**

The authors have no relevant financial or non-financial interests to disclose.

**Declaration of interests**

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

The authors declare the following financial interests/personal relationships which may be considered as potential competing interests:

**References**


    utm_source=xmol&utm_medium=affiliate&utm_content=meta&utm_campaign=DDCN_1_GL01_metadata.
    Accessed 21 Jul 2023


Figures

Figure 1

Rolling process flow chart of Cu and SS composite strips.

Figure 2

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Figure 4

(a) The experimental and simulated values of 2P; (b) The Cu-SS thickness ratio of composite strips.
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<table>
<thead>
<tr>
<th>Intermediate annealing temperature</th>
<th>After intermediate annealing</th>
<th>After the second-pass rolling</th>
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**Figure 6**

SEM images of the Cu/SS interfaces produced by different conditions: (a) 1P; (b) 1P-400; (c) 1P-600; (d) 1P-800; (e) 1P-1000; (f) 2P; (g) 2P-400; (h) 2P-600; (i) 2P-800; (j) 2P-1000.
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Roll-bonding mechanisms

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