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Microstructural Control of the Interface Layer for Strength Enhancement of Dissimilar Al/Cu Joints via Ni Addition during TIG Arc Brazing

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Abstract: Dissimilar metal joining between Al and Cu is effective for reducing the weight and cost of electrical components. In this study, dissimilar lap joining of pure Al to pure Cu with an Al-Ni filler material was conducted using tungsten inert gas (TIG) arc brazing, and the effect of Ni on the joint strength associated with the microstructure of the intermetallic compound (IMC) layer at the dissimilar interface was examined. The addition of Ni effectively increased the interfacial strength of the joints. Regardless of the addition of Ni, the joints fractured in the thick Al₂Cu layer formed at the Al/Cu interface. However, the Ni addition reduced the thickness of the IMC layer and led to the formation of Al₇Cu₄Ni particles in the weakest Al₂Cu layer. Both the thickness reduction and reinforcing Al₇Cu₄Ni particle formation are thought to contribute to the increase in joint strength of the Al/Cu dissimilar interface.

Keywords: intermetallic compound; dissimilar metal joining; aluminum alloy; copper

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1. Introduction

A multimaterial structure of dissimilar metals is required for performance improvement in commercial products. In particular, the combination of Al and Cu could reduce the weight and cost of electrical components [1,2]. The formation of an ideal structure in practical applications may require the dissimilar joining of Al alloy to Cu. Mechanical joining [3] has potential as a solution for the dissimilar joining, but it generally results in an increase in weight of the products, which may be unaccepted in practical use. Therefore, metallurgical joining is often required as a suitable technique without an increase in weight of the products. However, a sound metallurgical joint is a great challenge to achieve because of the formation of an intermetallic compound (IMC) layer [4–6] at the Al/Cu dissimilar interface through mutual diffusion during the joining process. The IMCs exhibit low ductility [5,7], resulting in a significant decrease in the interfacial strength of the Al/Cu joint.

To achieve satisfactory dissimilar metal joints that exhibit good mechanical and electrical properties, reduction in the IMC layer thickness at the dissimilar interface is generally attempted because a thin IMC layer can alleviate the degradation of the joints [2,5,8–11]. Pan et al. [11] conducted heat treatment of Al/Cu dissimilar friction welds and showed that a sufficient interfacial strength of the Al/Cu dissimilar joint was obtained with an IMC layer thinner than 1 µm, which led to failure in the heat-affected zone. Xue et al. [10] examined the effect of heat treatment on Al/Cu dissimilar friction stir welds and reported that the interfacial strength of dissimilar joints gradually decreased with the growth of a brittle IMC layer. Therefore, many attempts using low heat-input welding have been made to reduce the thickness of the IMC layer. For instance, the application of brazing

between liquid Al and solid Cu, such as arc brazing [12–15] and laser brazing [16–18], and solid-state joining, such as friction welding [19,20], friction stir welding [10,21–24], and ultrasonic spot welding [25–27], successfully suppressed the growth of the IMC layer, thereby improving the mechanical properties of the dissimilar joints.

The addition of alloying elements also has the potential to improve the mechanical properties of dissimilar joints. The microstructure of the IMC layer, such as the phase, grain size, thickness, and morphology, is strongly affected by alloying elements in the metals [28–33]. Xia et al. [28] showed that $\text{Cu}_{15}\text{Si}_4$ was formed at the interface between vacuum-brazed Al-Si alloy and Cu. Zhou et al. [32] reported that $\text{Al}_{4.2}\text{Cu}_{3.2}\text{Zn}_{0.7}$ was formed in a laser weld between Al and brass. These studies show that the addition of alloying elements is an effective method for changing the joint properties at the dissimilar interface. However, the effect of alloying elements on joint properties associated with the microstructure of the IMC layer at the Al/Cu dissimilar joint interface has not been systematically investigated.

Based on these findings, our research group attempted to examine the effect of alloying elements on the strength of the Al/Cu dissimilar joint obtained via laser brazing [34]. This study showed that the addition of Ni to Al effectively improved the strength of the $\sim 5\ \mu\text{m}$ thick multi-IMC structured layer due to the formation of a (Cu,Ni)Al layer at the weak interface between the Al_2Cu and Al_4Cu_9 layers. However, the fracture of Al/Cu dissimilar joints generally occurs at various locations and is governed by the thickness of the multi-IMC structured layer [8,10,11,35,36]. These results indicate that the effect of alloying elements on the strength of Al/Cu dissimilar joints should depend on the joining method. However, the effect of Ni on the joint strength obtained with various joining methods has not yet been clarified.

In this study, Al/Cu dissimilar lap joining with an Al-Ni filler material was performed by tungsten inert gas (TIG) arc brazing that exhibited a relatively low cooling rate and a long reaction time. The effect of Ni on the interfacial strength at the Al/Cu dissimilar joints was examined, and we attempted to clarify the microstructure of the IMC layers governing joint strength.

2. Materials and Methods

2.1. Materials

Oxygen-free Cu 1020 and pure Al 1050 were used as the base materials, and their chemical compositions are shown in Tables 1 and 2, respectively. The length, width, and thickness of the sheets were 100, 50, and 2 mm, respectively. To examine the effect of alloying elements on the joint strength and microstructure of the IMC layer, the Al-Ni filler material was prepared using a conventional casting method, and Al-X filler materials (X = Mn, Si, Cr, Zn, or Sn) were also used for comparison. The content of each alloying element in Al was adjusted from 0 to 7 at.%. The cast Al alloys were subjected to hot rolling to reduce solidification segregation and finished by cold rolling. The Al-X filler materials were finally prepared as bars with length, width, and thickness of 100, 4, and 2 mm, respectively.

Table 1. Chemical composition of C1020 used in this study.

Element	O	Cu
wt.%	0.0001	99.99

Table 2. Chemical composition of A1050 used in this study.

Element	Si	Fe	Cu	Ti	V	Al
wt.%	0.10	0.24	0.01	0.01	0.02	99.59

2.2. Joining

The Al/Cu dissimilar lap joining with Al-X filler materials was performed via TIG arc brazing. A schematic of this process is presented in Figure 1. The surfaces of the joined materials were polished with 400-grit emery paper to remove the oxide layer, and thereafter covered by a flux (FT-170, TOYO METAL, Osaka, Japan) to improve the weldability of the dissimilar metals. The pure Al sheet was placed on the Cu sheet to achieve a lapped length of 20 mm, and the Al-X filler material was placed adjacent to the pure Al sheet, as shown in Figure 1. To increase the weldability, these materials were heated up to 200 °C with a hot plate under the Cu sheet during the joining process. The joining parameters were as follows: Joining current of 140 A, arc voltage of 11 V, arc length of 2 mm, and joining speed of 200 mm/min. Ar shielding gas with a 20 L/min flow rate was employed for arc brazing.

Figure 1. Schematic of dissimilar tungsten inert gas (TIG) arc brazing.

2.3. Mechanical Testing

The interfacial strength of the Al/Cu dissimilar joints was evaluated by tensile shear strength testing. The configuration of the tensile specimen is presented in Figure 2. The specimens for this test were prepared as strips perpendicular to the joining direction. To fix the tested area in the tensile specimens at 20 mm², two slits were cut into the Al fusion zone and Cu sheet using an electrical discharge machine, as shown in Figure 2b. The tensile shear strength testing was performed on a screw-driven testing machine at a crosshead speed of 0.3 mm/s at room temperature. Spacers having the same thickness as the joined sheets were employed, as shown in Figure 2a, for pure shearing during the tensile testing. The hardness and effective elastic modulus of the IMC formed at the dissimilar interface were examined by a nanoindentation system (Triboindenter, Hysitron Inc., Minneapolis, MN, USA). The measurement was performed using a Berkovich indenter (Triboindenter, Hysitron Inc., Minneapolis, MN, USA) under a 2000-μN load. The relationship between the reduced modulus and the elastic modulus of the materials is given by Equation (1) [37].

$$\frac{1}{E} = \frac{1 - \nu_s^2}{E_s} - \frac{1 - \nu_i^2}{E_i} \quad (1)$$

where E is the reduced modulus; ν_s and E_s are Poisson's ratio and elastic modulus of the specimen, respectively; and ν_i and E_i are Poisson's ratio and elastic modulus of the indenter, respectively. The values of ν_i and E_i of the Berkovich indenter in this study were 0.07 and 1140 GPa, respectively. The effective elastic modulus, E^* , which is used for comparison of the mechanical properties of the IMCs, is defined by Equation (2).

$$E^* = \frac{E_s}{1 - \nu_s^2} \quad (2)$$

Figure 2. (a) Configuration and (b) location of the slits in the specimens used for tensile shear testing.

E^* was calculated from Equation (2) using the reduced modulus and mechanical properties of the indenter.

2.4. Microstructural Characterization

After the tensile shear strength test, the fracture surfaces were examined via scanning electron microscopy (SEM) to determine the fracture mode at the Al/Cu dissimilar joint interface. The chemical compositions of the fracture locations containing IMCs were examined using an electron probe microanalyzer (EPMA, JEOL, Akishima, Japan). The Al/Cu dissimilar joint interface, including the reaction layer, was examined via optical microscopy (OM), SEM, EPMA, and the electron backscatter diffraction (EBSD) technique. The samples for EBSD analysis were prepared by vibratory polishing in a 50 mL H_2O + 50 mL Struers OP-A solution for 10.8 ks (3 h).

3. Results

3.1. Microstructural Characterization

The relationship between the content of Ni in the Al filler material and the tensile shear strength is shown in Figure 3, and the effect of the other alloying elements is also presented for comparison. Regardless of the addition of alloying elements to the Al filler material, all tensile specimens fractured at the interface between the Al fusion zone and the Cu sheet. The tensile shear strength of the joint produced without alloying elements was approximately 35 MPa. The addition of Ni, Mn, Si, Cr, and Zn increased the interfacial strength of the Al/Cu dissimilar joint at appropriate amounts of the alloying elements. Among the alloying elements that exhibited a positive effect, the addition of Ni at a concentration higher than 1 at.% exhibited the most effective improvement in joint strength, which exceeded 45 MPa. On the contrary, the addition of Sn to the Al filler material tended to decrease joint strength, resulting in a strength reduction to 12 MPa in the worst case. This study attempted to elucidate the reason for the increase in interfacial strength of the Al/Cu dissimilar joint by the addition of Ni.

Figure 3. Effect of alloying elements in the Al filler material on the strength of the Al/Cu dissimilar joints.

3.2. Characteristics of the Reaction Layer

The backscattered electron (BSE) images of the Al/Cu dissimilar joint interface without alloying elements are presented in Figure 4. The reaction layer formed at the dissimilar interface consisted of a eutectic structure in the Al fusion zone and an IMC layer adjacent to the Cu, as shown in Figure 4a. Based on the Al-Cu binary phase diagram, the eutectic structure was predicted as Al + Al₂Cu [38]. The high magnification images shown in Figure 4b,c indicate that the IMC layer consisted of several IMC phases. The chemical compositions of the IMC phases produced without alloying elements are shown in Table 3, and the analysis locations are indicated in Figure 4c. Based on the results of the chemical composition analysis, the thick IMC layer below the eutectic structure was Al₂Cu, and the thin IMC layers near the Cu, in order of appearance from the Al side, were AlCu, Al₃Cu₄, and Al₄Cu₉. The chemical composition of the thinnest IMC layer adjacent to Cu could not be analyzed by EPMA in this study, but the Al-Cu binary phase diagram [38] implies that the thinnest layer was AlCu₃ or AlCu₄.

Figure 4. (a) Backscattered electron (BSE) image of the interfacial region with a reaction layer, (b,c) highly magnified BSE images of the intermetallic compound (IMC) layer located above the Cu in the Al/Cu dissimilar joint without alloying elements.

Table 3. Chemical compositions of the IMC phases formed without alloying elements indicated in Figure 4.

Location	Content (at.%)		Phase
	Al	Cu	
A	67.1	32.9	Al ₂ Cu
B	52.7	47.3	AlCu
C	44.1	55.9	Al ₃ Cu ₄
D	33.5	66.5	Al ₄ Cu ₉

The BSE images of the Al/Cu dissimilar joint interface produced with the addition of 2.3 at.% Ni are presented in Figure 5, and the results of chemical composition analysis of IMC phases found in Figure 5c are shown in Table 4. Similar to the joint interface produced without alloying elements, the dissimilar interface had a typical reaction layer corresponding to the joining between liquid Al and solid Cu [4,5,12–18], i.e., a eutectic structure formed in the Al fusion zone and an IMC layer adjacent to the Cu. The IMC layer mainly comprised a thick Al_2Cu layer, as shown in Figure 5b and Table 4, as was found in the joint produced without alloying elements (Figure 4). Meanwhile, the addition of Ni produced particles in the Al_2Cu layer with a higher Ni content than the other IMC phases, as shown in Figure 5c. The phase map including the particles detected using the EBSD technique is given in Figure 6. $\text{Al}_7\text{Cu}_4\text{Ni}$ particles (corresponding to the green region) were formed in Al_2Cu (corresponding to the red region). Thin IMC layers were observed above the Cu, as shown in Figure 5c. Two thin IMC layers were formed in the joint produced with the addition of Ni. The chemical composition analysis showed that the thin IMC layer adjacent to Cu was Al_4Cu_9 , and the middle layer was Al_3Cu_4 or $(\text{Ni,Cu})\text{Al}$, as shown in Table 4.

Figure 5. (a) BSE image of the interfacial region with a reaction layer, (b,c) highly magnified BSE image of an IMC layer located beside Cu in the Al/Cu dissimilar joint with the addition of 2.3 at.% Ni to the Al filler material.

Table 4. Chemical compositions of the IMC phases formed with 2.3 at.% Ni indicated in Figure 5.

Location	Content (at.%)			Phase
	Al	Cu	Ni	
E	65.2	33.7	1.1	Al_2Cu
F	62.0	35.1	2.9	-
G	44.6	54.7	0.6	Al_3Cu_4 or $(\text{Cu,Ni})\text{Al}$
H	33.6	66.1	0.3	Al_4Cu_9

Figure 6. Electron backscatter diffraction (EBSD) phase map at the Al_2Cu layer with Ni-containing particles formed at the joint by the addition of 2.3 at.% Ni.

3.3. Fracture Location within the Dissimilar Joint

The BSE images of the fracture surfaces of the tensile specimens produced without alloying elements and with 2.3 at.% Ni are presented in Figure 7. The results of a chemical composition analysis of the fracture surfaces are listed in Table 5, and the identified phases, which were comprehensively judged using the chemical compositions and the microstructures on the cross-sections analyzed by EBSD, were also presented. Regardless of the Ni addition, all the fractured specimens exhibited a flat surface without any dimple patterns, indicating a brittle fracture. In the joints without alloying elements, BSE images of both the Al- and Cu-side fractured surfaces displayed negligible contrast differences arising from the chemical composition, which the chemical composition analysis revealed to be Al_2Cu . These results showed that during the tensile shear test, the brittle fracture occurred in the Al_2Cu layer formed at the Al/Cu dissimilar joint interface. However, additional particles containing Ni, i.e., $\text{Al}_7\text{Cu}_4\text{Ni}$ particles, were observed on the fracture surfaces of the joint produced with the Al-Ni filler material. Based on the cross-sectional observation (Figure 5) and the chemical composition analysis (Table 4), it was suggested that the joint failed in the Al_2Cu layer with $\text{Al}_7\text{Cu}_4\text{Ni}$ particles.

Figure 7. BSE images of the fracture surfaces of the Al/Cu dissimilar joints without alloying elements and with the addition of 2.3 at.% Ni.

Table 5. Chemical compositions of the fracture surfaces indicated in Figure 7.

Location	Content (at.%)			Phase
	Al	Cu	Ni	
I	68.9	31.1	-	Al_2Cu
J	69.4	30.6	-	Al_2Cu
K	69.8	29.9	0.3	Al_2Cu
L	60.2	36.2	3.6	$\text{Al}_7\text{Cu}_4\text{Ni}$
M	67.7	31.9	0.4	Al_2Cu
N	55.3	41.6	3.1	$\text{Al}_7\text{Cu}_4\text{Ni}$

3.4. Hardness of Fractured IMCs

The hardness and effective elastic modulus of the fractured IMCs, i.e., Al_2Cu and $\text{Al}_7\text{Cu}_4\text{Ni}$, obtained by nanoindentation measurements, are shown in Figure 8. The Al_2Cu

at the dissimilar joint produced without alloying elements exhibited hardness of 8.8 GPa and elastic modulus of 142 GPa. These mechanical properties of Al_2Cu were hardly affected by the addition of Ni. Meanwhile, the hardness and elastic modulus of $\text{Al}_7\text{Cu}_4\text{Ni}$ were 10.8 GPa and 198 GPa, respectively, which were higher than those of Al_2Cu . Based on the identification of the fracture location (Figure 7 and Table 5), this result indicates that hard and highly elastic $\text{Al}_7\text{Cu}_4\text{Ni}$ particles were present on the fracture path during the strength testing.

Figure 8. Hardness and effective elastic modulus of fractured IMCs in the Al/Cu dissimilar joints.

4. Discussion

4.1. Formation of the Reaction Layer

During the reaction between liquid Al and solid Cu, a reaction layer consisting of an Al + Al_2Cu eutectic structure and a thick Al_2Cu layer was typically formed at the dissimilar interface [4,5,12–18]. Rapid dissolution of Cu into liquid Al resulted in a hypereutectic composition, leading to the formation of an Al_2Cu layer followed by a eutectic structure during the solidification process [4,5]. Further reaction results in the formation of several thin IMC phases, such as AlCu , Al_3Cu_4 , Al_2Cu_3 , and Al_4Cu_9 [5], by interdiffusion between Al_2Cu and solid Cu. In this study, the relatively large heat input by TIG arc brazing and the accompanying heating using a hot plate should result in a high joining temperature for a long reaction time, which causes the formation of a thick Al_2Cu layer with several IMCs at the Al/Cu dissimilar joint interface, as shown in Figure 4.

According to the Al-Cu-Ni ternary phase diagram [39], it is expected that the addition of Ni resulted in the formation of $\text{Al}_7\text{Cu}_4\text{Ni}$ and $(\text{Ni,Cu})\text{Al}$. $(\text{Ni,Cu})\text{Al}$ is generally a stable phase having large negative free energy over a wide range of chemical compositions and temperatures. Furuya et al. [34] reported that a $(\text{Ni,Cu})\text{Al}$ layer was observed between the Al_2Cu and Al_4Cu_9 layers at the dissimilar interface during laser brazing (which has lower heat input than TIG arc brazing) between Al-Ni alloy and pure Cu. These results suggest that a $(\text{Ni,Cu})\text{Al}$ layer is inevitably formed by the addition of Ni at the Al/Cu dissimilar interface, implying that the IMC phase of location G in Figure 5c was $(\text{Ni,Cu})\text{Al}$. This would explain why the addition of Ni to the Al filler material promotes the formation of $\text{Al}_7\text{Cu}_4\text{Ni}$ particles and a $(\text{Ni,Cu})\text{Al}$ layer, instead of several Al-Cu binary IMCs, as shown in Figure 5c.

4.2. Thickness of the IMC Layer

The thickness of the IMC layer at the dissimilar joint interface was reduced by the addition of Ni, as shown in Figures 4b and 5b. A decrease in the thickness of the IMC layer formed at a dissimilar interface increases joint strength [2,5,8–11]. Xue et al. [10] conducted heat treatment of Al/Cu friction stir welds and demonstrated that thick IMC layers exhibited low strengths at a dissimilar joint interface. Lee et al. [9] also reported that the growth of an IMC layer decreased the joint strength of Al/Cu dissimilar friction

stir welds. The thickness reduction of the IMC layer by the addition of Ni might be a microstructural factor that improves the interfacial strength of the Al/Cu dissimilar joint.

4.3. Dispersion of the IMC Particles

Several studies have reported crack propagation in the Al_2Cu layer formed at the Al/Cu dissimilar joint [11–13]. Pan et al. [11] examined the fracture location of the Al/Cu dissimilar joint with various IMC thicknesses and reported that a fracture of the Al/Cu dissimilar joint propagated in Al_2Cu when a relatively thick IMC layer formed at the interface. Feng et al. [12] reported that the Al/Cu dissimilar joint produced by metal inert gas arc brazing using cold metal transfer technology mainly failed in a thick Al_2Cu layer, even though several IMCs were observed at the dissimilar interface. These studies imply that the failure in the Al_2Cu layer can be dominant in the Al/Cu dissimilar joint with a thick IMC layer, which is consistent with the result of this study, as shown in Figure 7.

The strength improvement of brittle materials by reinforcing particles is caused by the mechanisms of crack bowing, crack deflection, and crack bridging [40–43]. Serbena et al. [41] conducted heat treatment of photo-thermo-refractive glass and showed that crack deflection was caused by residual stress arising from the elastic mismatch between the matrix and fine NaF particles, increasing the fracture strength and toughness. They [42] also reported that ceramic particles of $\text{Li}_2\text{Si}_2\text{O}_5$ effectively increase the flexural strength of lithium disilicate glass and concluded that crack bowing is the main strengthening mechanism responsible for the higher elastic modulus and toughness of the particles. Khan [43] et al. examined the mechanical properties of Ca- α -SiAlON with a dispersion of hard SiC particles and showed that crack deflection and crack bridging improved the toughness of the material. In this study, the effective and dominant mechanisms of strengthening by particles are unclear, but there is a high possibility that the hard and highly elastic $\text{Al}_7\text{Cu}_4\text{Ni}$ particles acted as reinforcing particles in the brittle Al_2Cu layer.

4.4. Strength Improvement Mechanism with Addition of Ni

The effect of the addition of Ni on the microstructure of the IMC layer at the Al/Cu dissimilar joint interface is summarized in Figure 9. The IMC layer at the dissimilar interface produced without alloying elements comprised a thick Al_2Cu layer and several thin IMC layers in order from the Al side to the Cu side, as shown in Figure 4 and Table 3. The dissimilar joint failed in the Al_2Cu layer during strength testing, as presented in Figure 7 and Table 5. The IMC layer produced with Ni also consisted of a thick Al_2Cu layer and thin IMC layers, but the thick Al_2Cu layer included $\text{Al}_7\text{Cu}_4\text{Ni}$ particles (Figure 5). The thickness of the IMC layer was clearly reduced by the addition of Ni into the Al filler material. The fracture of the dissimilar joint also occurred in the Al_2Cu layer, but the hard and highly elastic $\text{Al}_7\text{Cu}_4\text{Ni}$ particles were present on the fracture path that acted as reinforcing particles in the brittle Al_2Cu layer, as shown in Figure 7. The thickness reduction and dispersion strengthening caused by the addition of Ni could be a reason for the apparent improvement in the strength of the Al/Cu dissimilar joint by the addition of Ni.

Figure 9. Schematic of the Al/Cu dissimilar interface and fracture location of the joint produced without alloying elements, and with the addition of Ni to Al.

Our research group previously showed that the strength of the Al/Cu dissimilar joint obtained via laser brazing was improved by the addition of Ni, resulting from the formation of a (Cu,Ni)Al layer at the weak interface between the Al₂Cu and Al₄Cu₉ layers [34], which is a different mechanism obtained in this study. The definitive difference in the Al/Cu dissimilar joints produced by arc and laser brazing is the fracture location within the IMC layers formed at the joint interface. The fracture location of the Al/Cu dissimilar joints is generally governed by the thickness of the IMC layer [8,10,11,35,36]. Wallach et al. [35] reported that the failure mechanism of the Al/Cu joint transformed from intergranular between different IMC layers into transgranular in IMC when the thickness of IMC layer is greater than 25 µm. Pan et al. [11] also reported that, in the case of a thin IMC layer, the fracture occurred at the interface between the Al₂Cu and Al₄Cu₉ layers, while in the case of a thick IMC layer, the crack propagated inside the Al₂Cu layer. In our research, the Al₂Cu layer with over 30 µm thickness (produced via TIG arc brazing) tended to fail in the Al₂Cu layer, whereas an IMC layer thinner than about 5 µm (produced via laser brazing) mostly failed at the interface between IMC layers [34]. The difference in fracture location observed between the arc- and laser-brazed Al/Cu dissimilar joints indicates a different mechanism for strength improvement. However, the reason why the fracture location within the IMC layer depends on the thickness remains unclear.

5. Conclusions

The effect of the addition of Ni to the Al filler material on the strength associated with the microstructure of the IMC layer at the Al/Cu dissimilar joint interface was examined. The addition of Ni effectively increased the interfacial strength of the dissimilar joint. The IMC layer at the dissimilar joint interface was composed mainly of a thick Al₂Cu layer, resulting in brittle fracture of the joint, regardless of the addition of Ni to Al. However, the addition of Ni reduced the thickness of the Al₂Cu layer at the dissimilar interface and yielded the formation of Al₇Cu₄Ni particles in the fractured Al₂Cu layer. This study showed that the thickness reduction and reinforcing particles formation by the addition of Ni to Al increased the interfacial strength of the Al/Cu dissimilar joint.

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