Interfacial microstructure evolution and thermal reliability of copper/nickel joints formed by ultrasonic bonding*

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Ultrasonic bonding can be of use in developing electronic packaging technology at lower temperatures to afford more reliable bonded joints. We bonded Cu and Ni sheets together and used high-temperature testing to evaluate the thermal reliability of the joints at 673 K. Furthermore, we observed the microstructure of the joint interface and evaluated its effect on the fracture load of the Cu-Ni interface. The fracture load of the Cu-Ni interface of the joints heated at 673 K temporarily increased at 300 h and then decreased after 1000 h, and the heated joints showed good thermal reliability. Obvious changes were detected in the microstructure of the interface in the heated joint. The bonded region had obviously expanded vertical to the direction of the ultrasonic bonding, and the Cu side of the joint was composed almost entirely of coarse grains because almost all the Cu grains grew during the high-temperature test. However, a finely recrystallized microstructure and a diffusion layer had formed, especially at the Cu-Ni bonding interface. We discussed the effects of the changes in the microstructure of the Cu-Ni interface and the bonding material on the fracture load of the Cu-Ni joint. The diffusion and recrystallization region was not always stronger than the Cu base metal, suggesting that expanding the bonded region is important to improve the fracture load of the Cu-Ni interface. The ratio of the unbonded region to the bonded one at the Cu-Ni interface decreased from 72 to 54 % after the high-temperature test, indicating that diffusion bonding had occurred during the high-temperature test performed at 673 K. Therefore, the fracture load of the interface between the bonded Cu and Ni was improved at 673 K.

Key Words: Ultrasonic bonding, Copper/nickel bonding, Interfacial microstructure, Thermal reliability, Heat treatment

1. Introduction

Ultrasonic bonding is a method of solid-state bonding that can be used to form strong joints at low temperature in a short amount of time. It has been widely used in electronic packaging processes such as the Au and Al wire bonding process and applied to various bonding materials employed in electronic products. There are numerous studies on ultrasonic bonding of Al and Cu wires or ribbons and of Au and Cu balls¹⁻⁸⁾. Cu has recently attracted much attention as a potential ultrasonic bonding material for application in electronic packaging. Cu is cheaper than Au. In addition, Cu has better electrical and mechanical properties than Au has. From these characteristics, Cu is expected to be applied not only in wire bonding but also in bonding large-area terminals requiring high current capacity and thermal reliability in power modules^{2), 8), 11)}. On the other hand, Ni is one of the primary metallizing materials used in electronic packaging. Therefore, ultrasonic bonding of Cu and Ni is expected to be applied to various products. Furthermore, intermetallic compounds are not formed in Cu/Ni alloys, because the binary Cu-Ni system is completely soluble in the solid state (see in Fig. 1). Thus, Cu-Ni

joints formed by ultrasonic bonding are expected to exhibit long-term reliability at high temperatures. However, there are only a few reports on the ultrasonic bonding of such system¹¹⁾, and the interfacial microstructure and thermal reliability of Cu-Ni joints are not well understood. We previously reported the initial fracture load of Cu -Ni joints formed by ultrasonic bonding and the thermal reliability of the joints at 473 K, and demonstrated that Cu and Ni sheets could be joined by selecting the optimum bonding conditions and that the resulting bonded joints had good thermal reliability¹²⁻¹⁴⁾. However, we did not evaluate the thermal reliability of Cu-Ni joints, which is required for application in electronic products such as power modules.

In this work, we bonded Cu and Ni sheets together by ultrasonic vibration and used high-temperature testing to evaluate the thermal reliability of the joint at 673 K. Furthermore, we observed the microstructure of the joint and evaluated its effect on the fracture load of Cu-Ni joint.

Fig. 1 Binary phase diagram of Cu and Ni¹⁰⁾. The dotted line is spinodal line.

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Fig. 2 Schematic diagram illustrating the basic principle of ultrasonic bonding used in current work.

Fig. 3 Bonding profile for Cu and Ni.

2. Experimental procedure

2.1 Ultrasonic bonding

We used an ultrasonic bonding machine to bond together Cu and Ni sheets whose thickness, width, and length were 0.50, 10, and 25 mm, respectively. The surfaces of the Cu and Ni sheets were cleaned in alcohol before bonding. The sheets were bonded by pressing them together and vibrating them with a 2.0×2.0 mm bonding tool. The frequency of the vibration was 30 kHz. The schematic of the basic principle of ultrasonic bonding and the profile of ultrasonic bonding are shown in Figs. 2 and 3, respectively.

2.2 Thermal reliability tests

High-temperature tests were performed using an oven, at 673 K in air, and the thermal reliability was evaluated from the tensile shear load of the joints treated at high temperature. The tensile testing speed was 10 mm/min.

2.3 Observation of structure in bonded region of joints

The interface of the initial bonded joint and that of the joint subjected to high-temperature heating were analyzed by scanning electron microscopy (SEM), electron probe micro-analysis (EPMA), and electron backscatter diffraction (EBSD) analysis. Cross-sections of the joints were cut either parallel or vertical to the direction of ultrasonic vibration and the surfaces of the cross sections were polished using an Ar-ion beam. From the parallel cross-section, we observed deformations in the bonded materials and measured the length of the diffusion path between the bonded

Fig. 4 Tensile shear load of joints bonded for 0.80 s at maximum amplitude of $21.5 \mu\text{m}$ under trigger load of 30 N and maximum load of 120 N, and heated at 673 K.

Fig. 5 Cross-sectional SEM image of breakage point after tensile shear test. Joint was heated at 673 K for 300 h.

Cu and Ni. From the vertical cross-section, we estimated the amount of unbonded area at interface between the Cu and Ni to evaluate the fracture load of the Cu-Ni interface.

3. Result and discussion

3.1 Thermal reliability of the joint tested at 673 K

Figure 4 shows the results of the tensile shear test for bonded joints heated at 673 K for various length of time. The joints were bonded for 0.80 s at a maximum amplitude of $21.5 \mu\text{m}$ under a trigger load of 30 N and a maximum load of 120 N^{12-14} . The strength of the initial joint (*i.e.*, the bonded joint before the high-temperature test) was 386 N. The Cu side of the joint near the interface and the Cu-Ni bonding interface of the joint had started to fracture. From observations of the fractured surfaces, Cu was detected in the fractured surface of Ni side of the joint¹²⁻¹³. The bonded joint showed good thermal reliability at 673 K, even after 1000 h. The joint strength temporarily improved to 523 N at 300 h and then decreased to 349 N after 1000 h. The Cu base metal fractured during the tensile shear tests performed after 100, 300, 500, 800, and 1000 h. Figure 5 shows the cross-sectional image of a fracture point in the joint heated at 673 K for 300 h, indicating that the crack had progressed toward the Cu base metal from the edge of the bonding interface during the tensile shear tests. Therefore, the strength of the joints heated at 673 K was higher than that of the initial joints.

3.2 Microstructures of joint interfaces

Figures 6 and 7 show the SEM images, results of elemental analysis, and image quality (IQ) mapping images analyzed using

Fig. 6 Cross-sectional SEM images and line scans for initial bonded Cu/Ni interface (a) and (b) and for bonded Cu/Ni interface after high-temperature test was performed at 673 K for 1000 h (c) and (d), respectively¹⁴.

EBSD at the Cu-Ni bonding interface. We observed and analyzed the joint interface before and after the high-temperature test. We found that pressing and vibration of the bonding tool generated plastic deformation in the Cu base metal and that fine grains had formed at the bonding interface of the initial bonded joint; further there were no obvious diffusion of Cu and Ni atoms at bonding interface because the electron-probe diameter of the EPMA was almost 1 μm. The microstructure of the interface of the bonded Cu/Ni joint heated at 673 K for 1000 h was much different from that of the initial bonded Cu/Ni joint. Almost all the Cu grains grew during the high-temperature test, and the Cu side of the joint was composed almost entirely of coarse grains. However, vertically extending finely recrystallized grains had formed at the bonded interface, especially at the interface of the Cu-Ni. The line scans suggested, an approximately 15-μm-thick layer of Ni had diffused into the Cu side of the joint. Such a diffusion layer was also found in the joint heated at 673 K for 100 h. A finely recrystallized interfacial structure and diffusion layer had formed at 673 K during the high-temperature test.

The thermal interdiffusion distance was estimated using the following equation:

$$x = (2Dt)^{1/2}, \quad (1)$$

where x represents the diffusion distance, t represents the diffusion time, and D is calculated from the Arrhenius equation:

$$D = D_0 \exp\left(-\frac{Q}{kT}\right), \quad (2)$$

where D_0 represents the pre-exponential factor, Q represents the activation energy, k is the Boltzmann constant, and T represents the diffusion temperature. The interdiffusion distance was calculated by substituting $D_0 = 0.16 \text{ cm}^2/\text{s}$, and $Q = 1.8 \text{ eV}$ for the ranges 573–773 K¹⁵) into Eq. (2). T and t are assumed to be = 673 K and $3.6 \times 10^6 \text{ s}$, respectively. Although the estimated diffusion distance was 1.95 μm, the observed diffusion distance was almost one order of magnitude greater than the estimated diffusion distance. Some studies have previously reported unique diffusion and fine recrystallization at the interfaces of metal joints^{16, 17}). In those reports, diffusion-induced recrystallization (DIR) is explained as a phenomenon that involves both recrystallization and the diffusion of solute atoms along moving grain boundaries to form new grains for various solute concentrations. The solute atoms penetrate the grain boundary network at rates much greater than the estimated volume diffusion because of DIR^{16, 17}). In this work, fine grains formed at the bonding interface of the initial bonded joint so that Ni could quickly diffuse into the Cu side of the joint along the grain boundaries during high-temperature test. This finding suggests that the DIR region could form on the Cu side of the joint and that the Ni atoms that had penetrated the grain boundary network

Fig. 7 IQ map images of (a) initial bonded Cu/Ni interface and (b) bonded Cu/Ni interface heated at 673 K for 1000 h¹⁴⁾.

Fig. 8 Tensile shear load for initial joint and joint heated at 673 K for 100 h and bonded for 0.30 s at maximum amplitude of 10.7 μm under trigger load of 10 N and maximum load of 70 N.

could prevent the grain boundaries vertical to the bonded interface from moving during the high-temperature test. Thus, vertically extending finely recrystallized grains could form at the bonded interface, especially at the interface of the Cu-Ni.

3.3 Effects of changes in microstructure of bonded interface on fracture load of Cu-Ni interface

From the obtained results, we found that the fracture load of the Cu-Ni interface heated at 673 K was higher than that of the initial Cu-Ni interface and that an interface composed of a finely recrystallized structure and diffusion layer formed at 673 K during the high-temperature test. We evaluated the effects of the structural change at the bonding interface and the material on the fracture load of Cu-Ni interface by observing the fractured surface of the joint interface before and after the high-temperature test. However, we could not obtain specimens which fractured near the interface after the high-temperature test because the joints that were bonded for 0.80 s at a maximum amplitude of 21.5 μm under a trigger load of 30 N and a maximum load of 120 N and heated at 673 K had fractured at the Cu base metal. Therefore, we prepared joints that were bonded for 0.30 s at a maximum amplitude of 10.7 μm under a trigger load of 10 N and a maximum load of 70 N. These joints also fractured near the interface similarly to the previous joints. However, the fracture load of the new initial joints was 70 N. We performed a high-temperature test at 673 K for 100 h on the new joints. Figure 8 shows the results of the high-temperature test. The joint strength increased to 180 N for the joint heated at 673 K for 100 h,

Fig. 9 Optical and SEM images of initial fracture on Ni side of joint (a) and (b) and of fracture on Ni side of joint heated at 673 K for 100 h (c) and (d).

Fig. 10 Cross-sectional SEM images of (a) initial joint and (b) joint after high-temperature test was performed at 673 K for 100 h. Joints were bonded for 0.30 s at maximum amplitude of 10.7 μm under trigger load of 10 N and maximum load of 70 N.

Fig. 11 Ratio of unbonded region as observed from cross-section of joint vertical to direction of ultrasonic bonding for initial joint and joint after high-temperature test was performed at 673 K for 100 h.

and all the new heated joints fractured near the interface. Therefore, we observed the fractured surfaces of the new joints by preparing lower-strength joints. Figure 9 shows the fractured surfaces that formed during the tensile test of the initial joint and the joint heated at 673 K for 100 h. An elliptic bonded region whose major axis was vertical to the direction of the ultrasonic bonding had formed (Figs. 9 (a) and (c)). Furthermore, Figs. 9 (b) and (d) show that alternating bonded and unbonded regions had formed in the direction of the ultrasonic bonding. Ni atoms were detected at point A in the unbonded region, and Cu atoms were detected at point B in bonded region as shown in Figs. 9 (a) and (b). This result indicated that either the Cu side of the joint near the interface or the Cu-Ni bonding interface was the starting point of the fracture in the joints. The joints fractured, as shown in Figs.

9 (c) and (d), after the high-temperature test performed at 673 K for 100 h. Although a similar bonded region formed in the heated joint, the region had obviously expanded vertical to the direction of the ultrasonic bonding, as shown in Figs. 9 (c) and (d). This result indicated that diffusion bonding had occurred between Cu and Ni during the high-temperature test. The diffusion bonding between the Cu and Ni was one reason for the improved interfacial fracture load. From the results of the elemental analysis shown in Fig. 9 (d), Ni and Cu atoms were detected at points C and E, respectively. However, both Ni and Cu atoms were also detected at point D in a bonded region. The ratio of Ni to Cu atoms in the bonded region was 66:34. These results indicated that not only the diffused region but also the Cu side of the joint near the Cu-Ni interface or the bonding interface could be the starting point of the fracture. Therefore, the joint strength of the diffused region in the joint interface was not always greater than the strength of the Cu base metal, suggesting that either the diffused region had no negative effect on the fracture load or that the effect of formation of the diffused region on the fracture load of the Cu-Ni interface may be negligible. We suppose that expansion of the bonded region is important for improving the fracture load of Cu-Ni interface.

Figures 10 and 11 show stronger evidence for the occurrence of diffusion bonding between Cu and Ni during the high-temperature test. We evaluated the ratio of unbonded region by observing a cross-section vertical to the direction of the ultrasonic bonding in the initial joint and in the joint heated at 673 K for 100 h. In typical diffusion bonding between Cu and Ni, the voids that form as the unbonded region at the Cu-Ni interface will disappear during diffusion bonding. Therefore, we estimated the ratio of the area of the unbonded region to evaluate the strength of the Cu-Ni interface. We observed the microstructure of the Cu-Ni interface, as shown in Fig. 10, from cross sections of the whole Cu-Ni interface polished using Ar-ion beam. Figure 11 shows the ratio of the unbonded region in the initial joint and in the joint heated at 673 K for 100 h. The ratio of the unbonded region decreased from 72 % to 54 % after the joint was heated. This result indicated that diffusion bonding occurred between Cu and Ni during the high-temperature test performed at 673 K for 100 h. Therefore, the fracture load between the bonded Cu and Ni improved during the high-temperature test performed at 673 K. As a result, the improving of the fracture load and annealing of the Cu base metal could occur, the joint strength temporarily improved, and the Cu base metal fractured during tensile shear test after the joint was heated at 673 K, as mentioned in section 3.1. We suppose that the decrease in the strength of the joint heated at 673 K between 300 and 1000 h, is partly due to the oxidation of the Cu base metal. The Cu base metal clearly thinned

because Cu oxide layers spontaneously grew on and peeled off in the joint heated at 673 K between 300 and 1000 h, which may be why the Cu base metal weakened and the joint strength decreased.

4. Conclusions

We bonded Cu and Ni sheets together using ultrasonic vibration and used high-temperature testing to evaluate the thermal reliability of the joints at 673 K. Furthermore, we observed the microstructure of the joints and evaluated the effect of the microstructure of the joints on the fracture load of the Cu/Ni interface. The obtained results are summarized as follows:

- 1) The bonded joints heated at 673 K for 1000 h shows good thermal reliability. The strength of the joint temporarily improved to 523 N at 300 h and then decreased to 349 N at 1000 h. The Cu base metal in the heated joint fractured during the tensile shear test.
- 2) The microstructure of the Cu-Ni interface in the heated joint was obviously different from that of the Cu-Ni interface in the initial joint. Almost all the Cu grains grew, and a unique interface composed of a finely recrystallized structure and diffusion layer formed at 673 K during the high-temperature test.
- 3) The improved fracture load of the Cu-Ni interface was due to the diffusion bonding, and annealing of the Cu base metal could occur. The strength of the heated joint temporarily increased, and the Cu base metal fractured during the tensile shear test of the joint heated at 673 K. The Cu base metal thinned because Cu oxide layers spontaneously grew on and peeled off, which may be why the Cu base metal weakened and the joint strength decreased.

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