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# The effects of Ni addition on microstructure evolution and mechanical properties of solder joints undergoing solid-liquid electromigration

Hongyu Qiu<sup>a</sup>, Xiaowu Hu<sup>a,\*</sup>, Xiongxin Jiang<sup>a</sup>, Qinglin Li<sup>b</sup>

<sup>a</sup> School of Mechanical & Electrical Engineering, Nanchang University, Nanchang 330031, China <sup>b</sup> State Key Laboratory of Advanced Processing and Recycling of Nonferrous Metals, Lanzhou University of Technology, Lanzhou 730050, China

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# 1. Introduction

With the trend of miniaturization and multi-functionalization in electronic products in recent years, the feature sizes of solder joints are decreasing and current density crossing the interconnects of electronic packages are increased dramatically [1,2]. Recently, electromigration in the semiconductor field has been studied extensively to understand its mechanism [3,4]. Some studies reported that the additives such as Ni and Ni/Ti could enhance the electromigration reliability of Sn-Cu, Sn-Ag and Sn-Pb solder joints in electromigration process [5]. The occurrence of Cu-Ni cross-solder interaction affected the interfacial reaction during reflowing and solid-solid electromigration [6]. Chen et al. has reported that microtensile toughness of the solder/Ni interface was better than that of solder/Cu after solid-solid electromigration [7]. In addition, the introduction of Ni in solder joints prevented the preferred migration of metal atoms/ions in solder joints during the electromigration process [8]. However, because the diffusion rate of atoms in liquid is several orders of magnitude higher than that in solid, the interfacial reaction and failure mechanism caused by solid-liquid electromigration are quite different from those caused by solid-solid electromigration [6].

In this study, the solid-liquid electromigration soldering process was investigated to reveal the microstructure evolution, grain orientation and tensile strength of Cu-Ni-Sn intermetallic

\* Corresponding author. E-mail address: huxiaowu@ncu.edu.cn (X. Hu).

## ABSTRACT

The effect of solid-liquid electromigration on Cu-xNi/Sn-3.0Ag-0.5Cu/Cu-xNi (x = 0, 1.5 and 5 wt%) joints bonded at 260 °C with electric density of  $2.89 \times 10^2$  A/cm<sup>2</sup> was investigated. The results showed that intermetallic compounds (IMCs) performed a non-interfacial growth in the Cu-5Ni/SAC305 system. The electron backscatter diffraction (EBSD) analysis revealed that the orientation of the Cu<sub>6</sub>Sn<sub>5</sub> [1 0 0] direction parallelled to the direction of electric current, however, the addition of Ni could induce that the growth of (Cu,Ni)<sub>6</sub>Sn<sub>5</sub> grain exhibited isotropy. Additionally, the additional Ni resulted in significant reduction in grain size and enhancement in the tensile strength of joints.

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compounds (IMCs) between the SAC305 solder and Ni-contained Cu substrate.

### 2. Experimental procedures

The sandwich alloy/SAC305/alloy interconnection structures were utilized in the solid-liquid electromigration experiment in this study. The Cu-*x*Ni (x = 0, 1.5 and 5 wt%) alloys were cut into  $3 \times 3 \times 15$  mm<sup>3</sup> plates, and those plates were polished and cleaned, the SAC305 foil was used as interlayer with the dimensions of  $3 \times 3 \times 0.5$  mm<sup>3</sup>. The bonded samples were subjected to a direct current stress of  $2.89 \times 10^2$  A/cm<sup>2</sup> in a purified N<sub>2</sub> atmosphere (99.999%), and then the experimental facility was placed in an oven and heated to 260 °C for 2, 5 and 10 h, respectively.

The samples were embedded in epoxyresin, ground and metallographically polished. The cross-section of joints were characterized by a scanning electron microscope (SEM). Electron backscatter diffraction (EBSD) detector was used to examine the grain orientation of formed in IMC joints. In the EBSD analysis, ND (normal direction) and TD (lateral direction) coincided with the directions perpendicular to the cross-section and parallel to the electric current direction, respectively.

## 3. Results and discussions

Fig. 1a1–a3 shows the cross-section SEM microstructure of Cu/SAC305/Cu joints bonded at 260 °C for various hours with electric current, which clearly indicates the asymmetric growth





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Fig. 1. The cross-section SEM images of Cu-xNi/SAC305/Cu-xNi (x = 0, 1.5 and 5 wt%) joints under condition of 260 °C, 2.89 × 10<sup>2</sup> A/cm<sup>2</sup> for 2, 5 and 10 h, (a1–a3) Cu/SAC305; (b1–b3) Cu-1.5Ni/SAC305; (c1–c3) Cu-5Ni/SAC305; (d–g) EDS analysis results.

pattern of interfacial IMCs (Cu<sub>6</sub>Sn<sub>5</sub> and Cu<sub>3</sub>Sn) during the solidliquid electromigration process. It was obviously seen that the total IMC thickness was much higher at the anode than that at the cathode under electric current stressing. The unbalanced growth of IMC at cathode and anode was due to the Cu atoms migrating into the liquid solder volume driven from the cathode to the anode under electron wind force, resulting in the consumption of Cu atoms at the cathode and accumulation of Cu atoms at the anode [9]. The electrons formed a strong flow field when the electric current passed, and the atoms were transferred by the electron wind [10].

Fig. 1b1–b3 presents the cross-sectional microstructures of Cu-1.5Ni/SAC305/Cu-1.5Ni joints, which also clearly indicates the asymmetric growth pattern of IMC during the solid-liquid electromigration process. Furthermore, the significant polarity of IMC was also observed, Cu-Ni-Sn IMCs at the anode became thicker than those at the cathode. Regarding the asymmetric growth pattern of Cu-Ni-Sn IMCs, the reason should be explained by the theory mentioned above that Cu and Ni atoms migrating into the liquid solder volume traveled from the cathode to the anode under the electromigration.

Fig. 1c1–c3 displays the cross-sectional microstructures of Cu-5Ni/SAC305/Cu-5Ni joints. It was clearly shown that the IMCs performed a non-interfacial growth in the Cu-5Ni/SAC305 system with the application of electric current. As seen in Fig. 1c1–c2, after

a solid-liquid electromigration bonding process for 2 and 5 h, a very thin layer of (Cu,Ni)<sub>3</sub>Sn formed at the Cu-5Ni/SAC305 interface. Minor thick and dispersed particles of (Cu,Ni)<sub>6</sub>Sn<sub>5</sub> were detected in the residual solder. With the bonding time increased to 10 h, as shown in Fig. 1c3, the (Cu,Ni)<sub>6</sub>Sn<sub>5</sub> grains completely performed a non-interfacial growth. In detail, the (Cu,Ni)<sub>6</sub>Sn<sub>5</sub> grains were dispersed throughout the solder joint and formed numerous isolated "islands" within the solder. Obviously, Cu and Ni atoms migrated into the liquid solder under the effect of the electronic wind. However, the diffusion rate of Ni atoms in solder was slower than that of Cu atoms, which caused the majority of Ni atoms to be difficult to diffuse toward the anode interface in the Cu-5Ni/ SAC305 system. Meanwhile, electromigration resulted in enormous energy in molten solder, leading to a dynamically disequilibrium state [11,12]. After power supply cutting off, the supersaturated solder was returned to the static equilibrium state due to the cooling of bonding joints. The precipitated Ni element was uniformly distributed within solder matrix and provided plentiful nucleation sites for (Cu,Ni)<sub>6</sub>Sn<sub>5</sub>, which resulted in the formation of isolated IMCs "islands". Thus, under the driven of metal atom supersaturation, the IMCs non-interfacially grew in the Cu-5Ni/SAC305 couples with electric current.

Fig. 2a–f shows the results of the corresponding EBSD analysis of the cross-sectional surfaces of Cu-xNi/SAC305/Cu-xNi (x = 0,



**Fig. 2.** EBSD analysis of cross sectional grain structures of solder joints bonding for 10 h, planar inverse polar figures: (a) Cu/SAC305, (b) Cu-1.5Ni/SAC305, (c) Cu-5Ni/SAC305; (d–f) the corresponding polar figure; (g) the average grain size of IMCs and (h) the lattice structure of Cu<sub>6</sub>Sn<sub>5</sub>.

1.5 and 5 wt%) joints. As seen in Fig. 2a-c, the corresponding planar inverse polar figures in the transverse direction (TD) were employed to investigate the grain orientation. The lattice structure of Cu<sub>6</sub>Sn<sub>5</sub> belonging to the hexagonal system is shown in Fig. 2h. Some Ni atoms replaced part of positions of the Cu atoms, thereby formed  $(Cu,Ni)_6Sn_5$  with the same lattice structure. The (100) polar figures of the Cu<sub>6</sub>Sn<sub>5</sub> or (Cu,Ni)<sub>6</sub>Sn<sub>5</sub> phases were shown in Fig. 2d–f. As illustrated in Fig. 2a and g, the color of Cu<sub>6</sub>Sn<sub>5</sub> grains with the average size of 5.14 µm were mostly blue or blue-violet after bonding with electric current at 260 °C for 10 h in Cu/SAC305 system, indicating that the [100] crystal orientation tended to be parallel with the TD direction of the sample frame. In order to demonstrate the orientation evolution tendency more clearly, the corresponding  $(1 \ 0 \ 0)$  polar figures of the Cu<sub>6</sub>Sn<sub>5</sub> grains were shown in Fig. 2d. The red regions were principally located near the end of the TD axis, which also meant that the orientation of the [1 0 0] direction of Cu<sub>6</sub>Sn<sub>5</sub> parallelled to the electric current direction. As observed from Fig. 2b and g, the color of (Cu,Ni)<sub>6</sub>Sn<sub>5</sub> grains with the average size of 5.3716 µm was polychrome in Cu-1.5Ni/SAC305 system, and the corresponding (100) polar figures of the  $(Cu,Ni)_6Sn_5$  (Fig. 2e) shows that the red regions were distributed around the TD axis and were asymmetrical, meaning

that the orientation of  $(Cu,Ni)_6Sn_5$  [1 0 0] direction was at an angle to the current direction. As seen in Fig. 2c and g, the average grain size of  $(Cu,Ni)_6Sn_5$  grains was remarkably reduced to 3.85 µm. Meanwhile, color of  $(Cu,Ni)_6Sn_5$  grains were complicated in Cu-5Ni/SAC305 system (Fig. 2c), and the red regions were randomly distributed around the center (Fig. 2f), which indicated that  $(Cu,Ni)_6Sn_5$  grain growth exhibited isotropy.

Fig. 3 shows the tensile strengths of different joints bonded at 260 °C for 10 h. With the increment of Ni particle, the tensile strength of the joints increased gradually, as shown in Fig. 3a. The tensile strength of Cu/SAC305 system with asymmetrical growth pattern of IMC was the lowest (10 MPa), and Cu-5Ni/SAC305 system with non-interfacial growth pattern of IMC exhibited the highest tensile strength (21.48 MPa). Fig. 3b and c shows that the surface fractures tended to propagate within the IMCs and the cleavage plane was also observed, which indicated that brittle fracture occurred in Cu/SAC305 and Cu-1.5Ni/SAC305 samples. Interestingly, the fracture behaviour changed from brittle fracture to mixed fracture for Cu-5Ni/SAC305 sample since the fracture surface consisted of solder and (Cu,Ni)<sub>6</sub>Sn<sub>5</sub> particles and the dimpled zone appeared, as shown in Fig. 3d. Summarily, the improvement of tensile strength of Cu-5Ni/SAC305 joint could be



Fig. 3. (a) Tensile strength of joints bonding under the electric density of  $2.89 \times 10^2$  A/cm<sup>2</sup> at 260 °C for 10 h, and the fracture surfaces of: (b) Cu/SAC305, (c) Cu-1.5Ni/SAC305, (d) Cu-5Ni/SAC305; (e) brittle fracture mode; (f) mixed fracture mode.

ascribed by two aspects. On the one hand, the refinement and homogenization of IMC resulted in the improvement of the mechanical properties. On the other hand, the non-interfacial growth of IMC led to the thinning of interfacial IMCs layer and the  $(Cu,Ni)_6Sn_5$  particles dispersed throughout the solder could hinder the crack propagation [1].

## 4. Conclusion

In summary, the addition of Ni to the Cu substrate was an effective method to improve reliability of joints bonded using solid-liquid electromigration process. The IMCs performed a non-interfacial growth in the Cu-5Ni system, and the growth of  $(Cu,Ni)_6Sn_5$  exhibited isotropy compare to the orientation of the Cu<sub>6</sub>Sn<sub>5</sub> [1 0 0] direction paralleling to the direction of electric current in Cu/SAC305 system. In addition, the tensile strength were significantly enhanced due to the grains refinement and the non-interface grown IMC with  $(Cu,Ni)_6Sn_5$  particles distributed in solder, which all were induced by the increment of Ni addition.

#### **Declaration of Competing Interest**

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.

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