Solder electromigration behavior in Cu/electroless Ni–P plating/Sn–Cu based joint system at low current densities

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Abstract

Electromigration (EM) in solder joints has great influence on their reliability. Nevertheless, few reports have been published on the EM in solder joints with Ni–P barrier layers at lower current densities less than 10 kA/cm^2. In the present study, EM in Cu/Ni–P/Sn–0.7Cu/Ni–P/Cu joints was investigated at 150 °C with current densities of 5.0 and 7.5 kA/cm^2. The breakdown mode was open failure of the solder joint on the cathode. It was found that Ni in the Ni–P barrier layer diffused toward the anode, resulting in a thicker P-rich layer, which caused the cracks and the delamination of the P-rich layer. Additionally, the diffusion of Sn detached the solder from the Ni_3SnP intermetallic compound on the cathode.

Key words

electromigration, solder, low current density, power module, Ni–P plating, Sn–Cu

I. Introduction

Electromigration in the semiconductor field has been studied extensively for decades to understand and control this phenomenon on the metallization. Electromigration is the movement of metal atoms in the direction of strong electron flow, which is termed “electron wind”, resulting from momentum transfer between conducting electrons and diffusing metal atoms. As a device’s temperature or current density increases, its mean time to failure caused by electromigration decreases, this relationship is expressed well by Black’s equation [1]–[7]. For example, the current density of fine pitch solder bumps between the chip and substrate are a major concern in the semiconductor field.

It is well known that current densities of 10 kA/cm^2 may cause serious electromigration. Although power modules with insulated–gate bipolar transistors (IGBTs) and diodes do not require joints as small as those of the flip chip, the power modules for hybrid electric vehicles usually operate above 200 A. In addition, the service temperature of engine compartments is very high. These conditions raise serious concerns about electromigration failure in the power modules for hybrid electric vehicles. The current density of power devices containing solder in double-sided cooling power modules varies from 0.24 kA/cm^2 to 0.4 kA/cm^2, and the influence of solder electromigration is believed to be minimal in the current operating environment [8]–[12].

Environmentally friendly vehicles must achieve good fuel-economy and high–power density simultaneously. To meet these demands, vehicles require power module with low energy consumption, high heat dissipation, and high-thermostability. The current density in the solder joints will increase for high-thermostability SiC power devices and miniaturized devices [2], [3].

Power modules can be miniaturized by integrating 6-in-1 and 2-in-1 circuit compositions, which operate a three-phase alternating current motor, into one package [15], [16]. Kadoguchi et al. reported that an emitter electrode in the upper arm IGBT is usually jointed to a collector electrode in the lower arm IGBT for 2-in-1 power modules. This joint is smaller than that in power devices, so its current density is greater and solder electromigration occurs more easily [16]. Thus, it is necessary to investigate the mechanism of solder electromigration and to enhance the electromigration resistance of power modules.

Several studies have reported that joints with under bump metallurgy, such as Ni and Ni/Ti, could enhance the electromigration reliability of Sn–Cu, Sn–Ag, and Sn–Pb solders at a high current. However, there are few studies on solder electromigration at current density lower than 10 kA/cm^2, which is a realistic value [17]–[28]. Thus, in the present paper we investigated electromigration in a joint of electroless Ni–P plating/Sn–0.7Cu based joint system at current densities lower than 10 kA/cm^2 at 150 °C. We identified the diffusion behavior of each element caused by electromigration mainly by comparing the solder joint interfaces on the anodes and cathodes.

II. Experimental procedure

Figure 1. (a) and (b) show an overview of the sample and a schematic cross-section of the cathode respectively. The electrodes were made from oxygen-free copper (C1020) and had a length of 1 mm and a cross section of 0.4 × 0.4 mm. The soldering pad for the Cu electrode was finished by electroless Ni–P/Au plating. The thickness of the Ni–P plating was 7–13 µm. Two electrodes were jointed with a 0.5-mm-diameter solder ball. The solder composition was M725 (Sn–0.7Cu–Ni–P, wt.% ; Senjyu Metal Industry Co., Ltd.). Figure 2. shows a scanning electron microscope (SEM) image of the cross-section of a joint interface. The intermetallic compound (IMC) at the interface was ~5.0 µm thick. Energy dispersive X-ray spectroscopy (EDX) showed that the IMC consisted of 32.8 at.% Cu–20.8 at.% Ni–46.4 at.% Sn. Thus, the IMC composition can be estimated as $(Cu,Ni)_6Sn_5$. A P-rich layer appeared and was ~1.3 µm thick, and this layer consists of mixture of Ni_3P and Ni [29]. Figure 1. (b) confirmed the sample composition. We applied

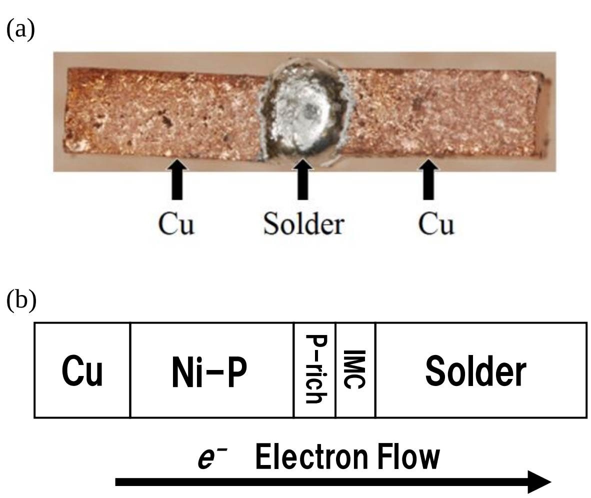

Figure 1. EM test sample
(a) Overview (b) Schematic of cathode side

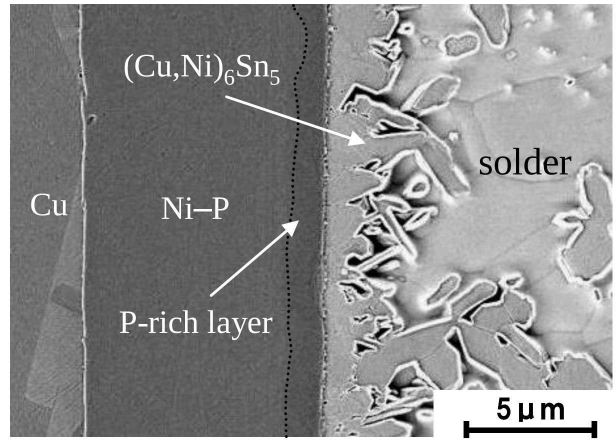

Figure 2. Cross-section SEM image of the solder joint interface after reflow

currents of 8.0 and 12 A to the solder joint to generate current densities of 5.0 and 7.5 kA/cm^2, respectively. The test sample was placed on a hot plate, which maintained the joint temperature at 150 °C when a specified constant direct current passed through the joint. The temperature profile of the test sample was determined by thermocouple and by measuring the resistance change in the junction line between Cu/solder/Cu joints. The change in resistance caused by electromigration was measured by monitoring the voltage. We defined the failure criterion as a 20% increase in resistance from the initial value. The cross-section of the failed sample was polished and studied using SEM. The electromigration failure mode was investigated by analyzing the elemental composition with EDX. To check the electromigration phenomenon over time, a joint was studied after applying 5.0 kA/cm^2 at 150 °C for 250, 500, 750, and 2500 h.

III. Results and discussion

A. *Electromigration stress test*

Figure 3. shows the change in resistance over time for current densities of 7.5 and 5.0 kA/cm^2 applied at 150 °C. The failure times were 1060 and 2320 h, respectively. Figure 4. (a) and (b) show the failure samples after these tests. In both failed samples, the solder detached from the Cu electrode on the cathode side. Samples maintained at 150 °C without applied current did not fail at the joint interface. Thus, we conclude that in both samples with applied current, the failure mode was caused by electromigration. Figure 5. shows a cross-section SEM images of the failed sample stressed at 150 °C and 5.0 kA/cm^2. This sample was detached at the solder interface of the cathode side as shown in Figure 5. (a). Figure 5. (b) and (c) show enlarged images of the detached area. The $(Cu,Ni)_6Sn_5$ IMC which remained after reflow, disappeared. The Ni–P plating transformed into a Ni_3SnP

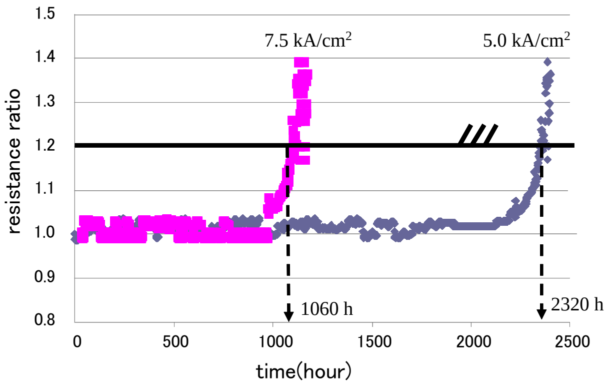

Figure 3. Change of resistance ratio under electromigration stress

IMC and P-rich layer. Delamination occurred between the P-rich layer and the Cu electrode. The Ni_3SnP IMC and P-rich layer were cracked. The Cu electrode formed Cu_3Sn and Cu_6Sn_5 IMCs, as shown in Figure 5. (b). The crack in the Ni_3SnP IMC and P-rich layer existed on the Cu–Sn IMCs at Cu electrode, and Sn was supplied from this crack to the Cu electrode. At this time, the diffusion direction of Sn was opposite the electron flow. Thus, it is believed that Sn is transferred from the cathode to the anode by thermal diffusion. Figure 5. (d) shows an enlarged image of the anode side, revealing that Ni–P plating and $(Cu,Ni)_6Sn_5$ IMC remained. The P-rich layer on the anode side was ~1.5 µm thick. It grew thicker after reflow but the rate on the anode side was slower than on the cathode one.

Thus, electromigration enhanced the diffusion of Ni in the Ni–P plating on the cathode toward the solder.

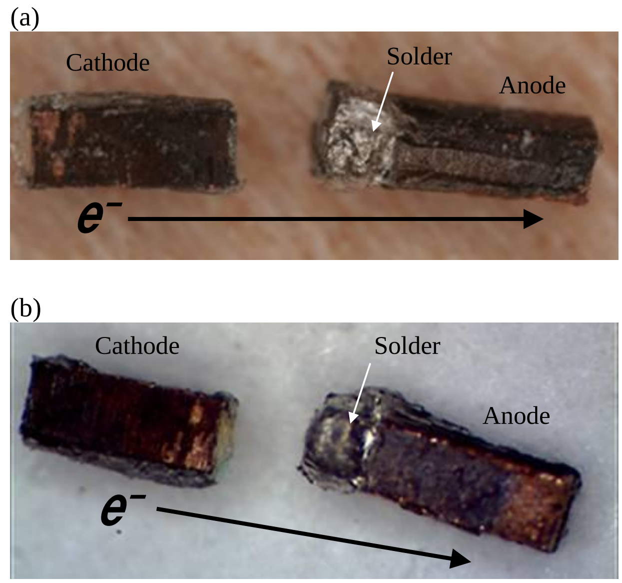

Figure 4. Overview of failure samples
(a) 5.0 kA/cm^2, 2320 h (b) 7.5 kA/cm^2, 1060 h

B. Diffusion behavior of Ni from the Ni-P plating toward the solder with electromigration stress

By diffusion of the Ni from Ni–P plating toward the solder, the Ni–P plating reformed the P-rich layer.

To confirm that this behavior was influenced by electromigration, we analyzed sample cross-sections after applying a stress of 150 °C and 5.0 kA/cm^2 for 250, 500, 750, and 2500 h. No samples exceeded the current breakdown criterion of 20% resistance increase from the initial value. Figure 6. shows cross-section SEM images after applying a

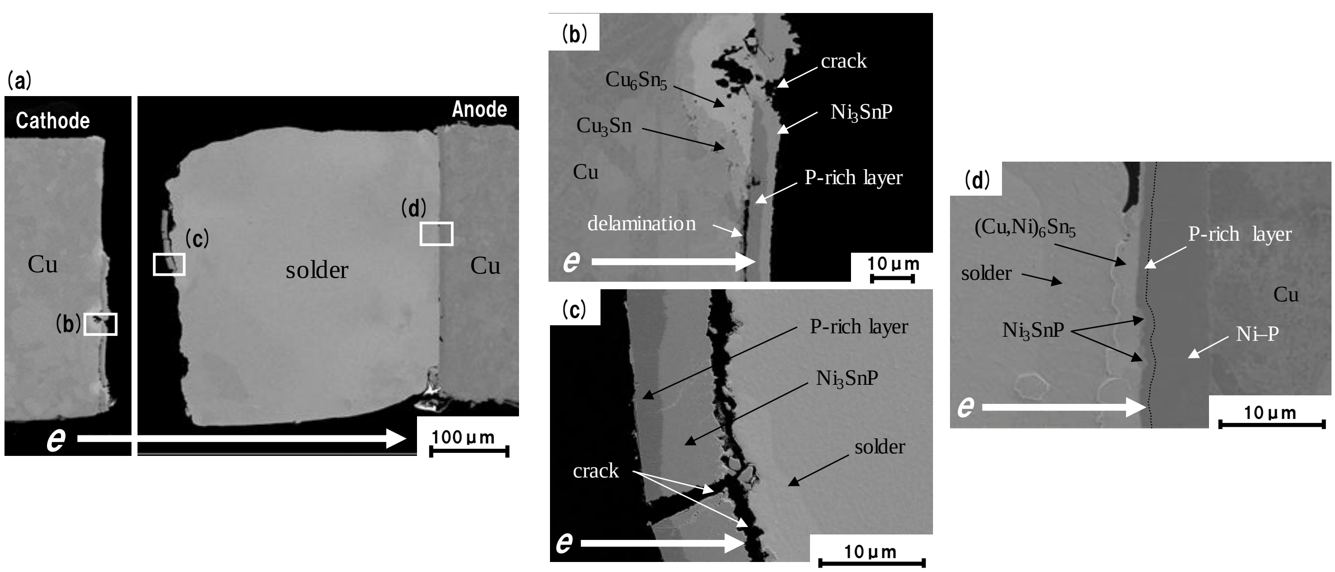

Figure 5. (a) Cross-section SEM image of the solder joint after 2320 h at 150 °C and 5.0 kA/cm^2
(b), (c) enlarged images of cathode side (d) enlarged images of anode side

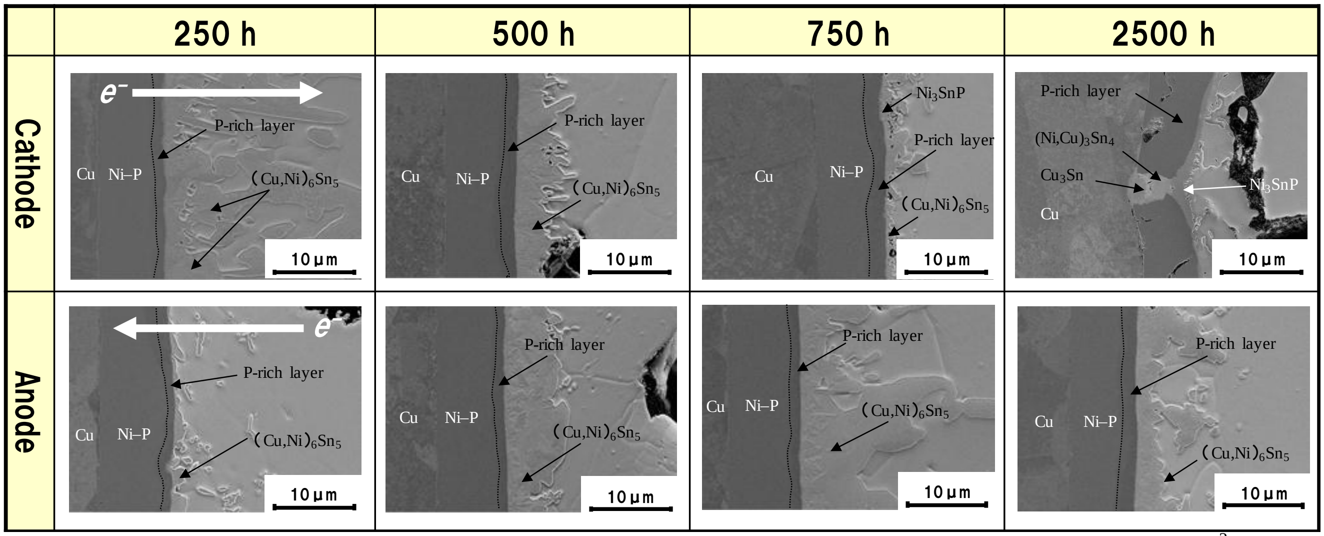

Figure 6. Cross-section SEM images of the solder joint on the cathode and anode at 150 °C and 5.0 kA/cm^2

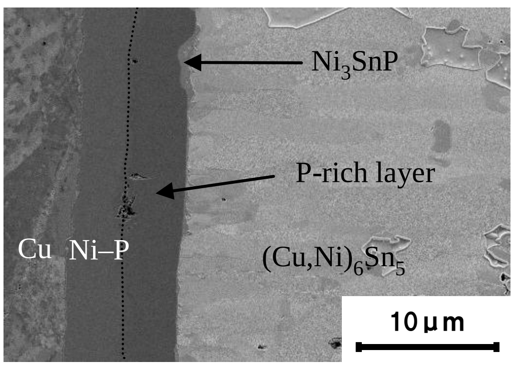

Figure 7. Cross-section SEM image of the solder joint after 2500 h at 150 °C with no applied current

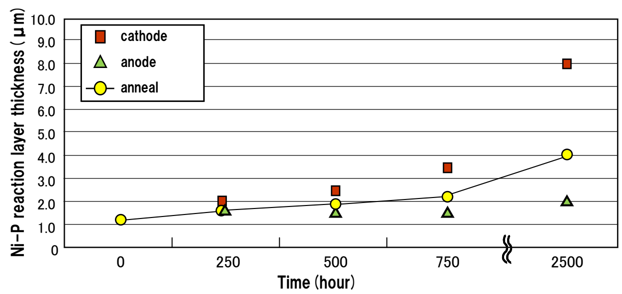

Figure 8. The change of Ni–P reaction layer thickness at 150 °C and 5.0 kA/cm^2

stress of 150 °C and 5.0 kA/cm^2. The (Cu,Ni)$_6$Sn$_5$ IMC on the cathode side was ~20 μm thick after 250 h, approximately four times thicker than original value. After 500 h and 750 h, the (Cu,Ni)$_6$Sn$_5$ IMC became thinner. After 2500 h, the (Cu,Ni)$_6$Sn$_5$ IMC disappeared from the joint interface, and a Ni$_3$SnP IMC formed there instead. The P-rich layer on the cathode side grew with time, and all the Ni–P plating changed into the P-rich layer (~8 μm thick), forming cracks. The Ni–P plating layer (9.5 wt.% P) was amorphous, and compressive stress was applied in this layer. The transformation from amorphous (Ni–P) to crystal (Ni$_3$P) led to volume shrinkage, which caused cracks and delamination [30],[31]. Two types of IMC appeared inside the cracks of the P-rich layer on the cathode side after 2500 h. It is believed that one IMC on the Ni$_3$SnP IMC side to be (Ni,Cu)$_3$Sn$_4$ because its composition from elemental EDX analysis was 34.1 at.% Ni–10.5 at.% Cu–55.5 at.% Sn. It is believed that the other IMC was Cu$_3$Sn because its composition was 75.2 at.% Cu–24.8 at.% Sn. The diffusion direction of Sn in the Ni$_3$SnP IMC was opposite the electron flow. Thus, it is believed that the Sn was transferred from the Ni$_3$SnP IMC to cracks in the P-rich layer by thermal diffusion, and reacted with the Cu electrode. As a result, Cu$_3$Sn and (Ni,Cu)$_3$Sn$_4$ IMCs were formed on the Cu electrode. The P-rich layer on the anode side was ~2.0 μm thick after 2500 h. It grew thicker after reflow, but the rate on the anode side was lower than on the cathode side.

Figure 7. shows a cross-section SEM image of a joint annealed at 150 °C for 2500 h with no applied current. A Ni–P plating layer remained, and the interface of the Cu electrode was undetached, although the (Cu,Ni)$_6$Sn$_5$ IMC and P-rich layer grew thicker than original value. The P-rich layer was ~4.0 μm thick. Electromigration almost doubled the growth rate of the P-rich layer on the cathode side and halved that on the anode side after 2500 h.

Figure 8. shows the change in thickness of the Ni–P plating over time at 150 °C with a current density of 5.0 kA/cm^2 and without applied current. The reaction rate of Ni–P on the cathode side at 150 °C with applied current was faster than that without applied current up to 2500 h. Thus, it is believed

that electromigration accelerated the diffusion of Ni from the Ni–P plating on the cathode toward the solder. The P-rich layer on the anode side grew slowly with applied current than without. Thus, the electromigration suppressed the diffusion of Ni in the Ni–P plating on the anode toward the solder.

C. *Breakdown mechanism of the cathode side with electromigration stress*

Figure 9. shows an approximate breakdown mechanism for the cathode side, produced by analyzing the electromigration failure mode and test samples with electromigration stress.

(1) Electromigration diffused Ni in the Ni–P plating and accelerated the formation of a P-rich layer. When the P-rich layer reached a Cu electrode, this layer delaminated from the interface and cracks formed in this layer.

(2) Sn diffused from the Ni_3SnP IMC to the cracks in the P-rich layer and formed Cu_3Sn and $(Ni,Cu)_3Sn_4$ IMCs on the Cu electrode side.

(3) Thermal diffusion provided Sn from the solder side and formed Cu_3Sn and Cu_6Sn_5 IMCs on the Cu electrode interface.

(4) Sn on the cathode was diffused by electromigration and broke down between the Ni_3SnP IMC and the solder.

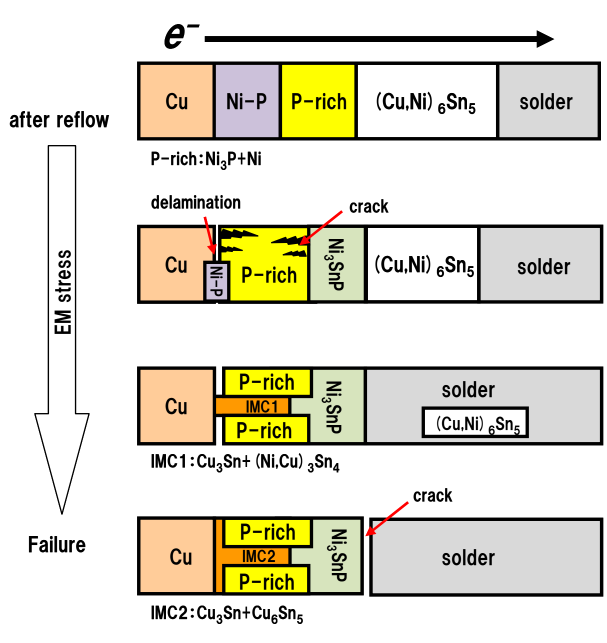

Figure 9. Schematic of electromigration failure mechanism on the cathode

IV. Conclusions

We investigated electromigration in the joint between Ni–P plating and Sn–Cu solder at 150 °C and current densities lower than 10 kA/cm^2.

(1) The times to failure for the electromigration tests were 1060 and 2320 h at current densities of 7.5 and 5.0 kA/cm^2, respectively. Electromigration breakdown occurred even at these current densities. The breakdown surface was the interface between the Ni_3SnP IMC and the solder on the cathode side.

(2) The P-rich layer on the cathode side grew more quickly than that on the anode side, and it grew faster with applied current than without. The diffusion direction of Ni was the same as the electron flow. Thus, electromigration accelerated Ni diffusion on the cathode side.

(3) The P-rich layer on the anode side grew slower than that on the cathode side, and it grew more slowly with applied current than without. The diffusion direction of Ni toward the solder was opposite the electron flow. Thus, electromigration suppressed Ni diffusion on the anode side.

When the P-rich layer reached the Cu electrode, cracks formed. Sn diffused from the Ni_3SnP IMC to these cracks in the P-rich layer and formed Cu_3Sn and Cu_6Sn_5 IMCs with the Cu electrode. The diffusion of Sn detached the solder from the Ni_3SnP IMC on the cathode.

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