Influence of External Interface Normal Stress on the Growth of Cu–Sn IMC During Aging

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Abstract

A U-shape clamp was designed to apply stress perpendicular to the interface of Cu/Sn/Cu solder joints, and its influence on the growth behavior of Cu–Sn intermetallic compound (IMC) during thermal aging at 150 °C was investigated. The results show that compared with the sample at general stress-free state, the growth rate of IMC under compression is faster, while that under tension is slower. Moreover, the interface between IMC and Sn is smoother under compressive stress, and the corresponding IMC grains are smaller and more uniform than that under tensile stress. According to the growth kinetic analysis, the growth of IMC under general, compressive and tensile states is all controlled by the combination of grain boundary diffusion and volume diffusion with a similar growth exponent ($n \approx 0.4$). However, external stress can affect the Ostwald ripening process of grain growth, causing a change of grain size and grain boundary density in the IMC layer. As a result, the IMC growth behavior at the interface of the solder joint will be affected by the applied external normal stress.

Keywords Cu–Sn IMC · Growth behavior · External stress effect · Isothermal aging

1 Introduction

Traditional 2D-packaging has reached its limitation after decades of development. New technologies such as 3D-packaging are trying to achieve higher packaging density by integrating more components in the Z-axis [1]. But the transformation of packing technology also raises a problem that solder joints within 3D-packaging structure will work in a different and surely more complex circumstance compared with those within 2D structure [2]. In 2D-packaging, solder joints generally serve under planar stress, which parallels to the PCB (printed circuit board) as well as to the Cu–Sn

reaction interface in most circumstances. This planar shear stress can cause the failure of solder joints failure commonly [3, 4]. However, solder joints in 3D structure would also work under stress perpendicular to the reaction interface which is introduced by extreme temperature gradient [5]. Furthermore, with the miniaturization of electronic devices, the volume ratio of intermetallic compound (IMC) to each solder joints rises quickly [6]. IMC would play a more prominent role in the reliability of microelectronics.

Due to the significance of the IMC and the complicated service circumstances, the effect of external stress on the growth behavior of Cu–Sn IMC is essential for the better understanding on the reliability of solder joints in 3D-packaging. Among the literature, Vianco et al. [7] first reported the influence of external stress on the IMC growth during aging. Ngoh et al. [8] designed a C-ring to induce in-plane stress on the solder/substrate interface, and an increase in IMC growth rate was found when compressive stress was applied. Similar results were also revealed by Hu et al. [9]. However, all the works mentioned above are mainly focused on the stress that parallels to the reaction interface. In this manuscript, we applied external normal stress using a U-shape clamp to the interface of Cu/Sn/Cu solder joint, and the influence of normal-plane stress on the IMC growth behavior was first investigated and discussed.

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2 Experimental

Pure bulk Cu (99.99%) was cut into small bars with dimensions of 2 mm × 2 mm × 10 mm, and pure Sn ingot (99.999%) was cut into small flakes in size of 2 mm × 2 mm × 1 mm by spark cutting. After cutting, the Cu bars and Sn flakes were immersed in ethanol for 5 min for ultrasonic cleaning, then ground and polished to obtain a smooth and oxide-free surface. Surface-polished copper bars and tin flakes were then aligned and assembled as Cu/Sn/Cu sandwich structure and reflowed in a four-section reflow oven. The reflow peak temperature was set to be 250 °C, and the melting time of Sn was controlled within 5 s to ensure the formation of a thin layer of initial IMC.

A U-shape clamp made of 65Mn spring steel plate was employed as stress source in this experiment. The configuration/size of the clamp and the sample are presented in Fig. 1a. To introduce compressive stress, the length of the sample was machined larger than the width of the U-shape clamp. After the sample was mounted within the clamp, the U-shape clamp would squeeze the sample due to its deformation (Fig. 1b). If the length of the sample was shorter than the width of the fixture, the sample would bear tensile stress in the same way.

To estimate the stress in the sample, one can treat the clamp as a combination of two cantilever beam and suppose the curve part in the clamp does not deform after the sample was fixed on. Moreover, compared with the length of the clamp arm, the deformation value of clamp is rather small. Therefore, the applied stress would strictly parallel to the long axis of the sample, namely being perpendicular to the Cu/Sn/Cu interfaces. Then the load at the tip of one

cantilever beams can be expressed by the deflection equation [10]:

$$y_{\max} = \frac{Pl^3}{3EI}, \quad (1)$$

where $y_{\max}$ is the deflection of the cantilever beam after the sample was fixed on, P is the load on the tip of the cantilever beam, l is the length of the cantilever beam, E is the elastic modulus of 65Mn steel, and I is the cross-sectional moment of inertia of the cantilever beam which can be calculated by [10]:

$$I = \frac{bh^3}{12}. \quad (2)$$

In Eq. (2), b is the width of the clamp and h is the thickness of the clamp. Then, the stress in the sample can be calculated by [10]:

$$\sigma = \frac{P}{A_c}, \quad (3)$$

where A_c is the area of the cross section of the clamp. All the values of the variables mentioned above are summarized in Table 1.

The calculated result in Table 1 shows that the absolute value of the stress in the sample is about 4.16 MPa according to Eq. (3). This stress value seems small compared with others' work on the influence of external in-plane stress on Cu–Sn IMC [8]. But the yield strength of pure tin at 150 °C is 7.6 MPa [11]. During aging treatment, the pure tin would creep and cause the tension sample break apart. To ensure the results as obvious as possible, the value must be kept at a high level but with a limitation to keep the experiment feasible. At the same time, for controlling the experimental conditions more comparable, the length of the sample was carefully designed to keep the absolute stress value of compression as the same as that of the tension state.

After fixed into the clamp, three groups of samples (compression group, tension group and general stress-free group) were aged at 150 °C for 100 h, 250 h, 500 h, 1000 h, respectively (Fig. 1c). After aging, the samples were mounted in

Fig. 1 Photographs of **a** the configuration of the spring steel clamp and as-reflowed general sample, **b** clamps with compression and tension samples, **c** schematic illustration of the experiment process

Table 1 Value of variations used in estimating the stress in the sample and results obtained from each equation

Variables	Values	Results
$y_{\max}$	2.5 mm	$P \approx 16.6 \text{ N}$
l	4 mm	
E	$1.97 \times 10^5 \text{ MPa}$	
b	5 mm	$I = 0.72 \text{ mm}^4$
h	1.2 mm	
A_c	4 mm^2	$\sigma \approx 4.16 \text{ MPa}$

epoxy resin and gone through standard metallographic processes to expose the reaction interface. At last, excessive Sn was dissolved in a 20% HNO_3 solution to uncover the IMC grains. The reaction interface and the IMC grains were observed by an FEI Quanta 600 scanning electron microscope (SEM). The thickness of IMC was calculated by a traditional approach which is extensively described [12]. The size of IMC grain and its distribution was measured and analyzed by an image analysis software.

3 Results and Discussion

3.1 Interfacial Morphology Evolution of Solder Joints

The cross-sectional and top view images of the as-reflowed sample are shown in Fig. 2. It can be seen that fine IMC grains are closely arranged at the reactive Cu/Sn interface. The interfaces between Cu/IMC and IMC/Sn are clear, and no pores or inclusions are observed (Fig. 2a). Cu and Sn are metallurgically bonded by the formation of IMC, which was identified as Cu_6Sn_5 . Besides, due to the short reflow time, the IMC thickness is rather thin (about 1 μm in Fig. 2a) and the size of IMC grains is uniform with a radius of approximately 1 μm (Fig. 2b). In the literature about the interfacial reaction between Cu and Sn, for the convenience of calculation and establishment of theoretical models, some diffusion couples were designed by diffusion bonding or electroplating to avoid the initial IMC formation [13, 14], so that the IMC thickness (x) would be taken to grow from 0 mm to a certain value. However, by such sample treatment method, thin oxide layer was observed between two reaction components which would act as a diffusion barrier between them [15]. Moreover, in such method each diffusion couple would go through an uncertain nucleation incubation period which

would be counted into the total aging time in subsequent kinetic analysis [16]. According to Dybkov's modeling [17], if two reaction components were perfectly bonded (metallurgically bonded without any IMC formation), at the beginning stage of aging, the growth behavior of a single IMC layer would obey a linear law rather than a parabolic one. Therefore, there is no need to keep IMC thickness to 0 before the aging experiment if we only focus on the long-time aging behavior. In this study, the samples were prepared by a reflow process, but the melting time of Sn was controlled to just 5 s to keep the generated IMC as thin as possible. In this way, Cu and Sn were metallurgically bonded before aging treatment, which avoids the aforementioned formation of oxide layer. Moreover, the thin initial IMC should not have a significant effect in the later solid-state reaction.

Before any further statistical analysis, cross-sectional morphology of each sample gone through different aging time was well examined and is compared in Fig. 3. It can be found that interfacial morphology of three groups of samples varies from each other greatly among compressive state (Fig. 3a), stress-free state (Fig. 3b) and tensile state (Fig. 3c). All the samples have a straight Cu/IMC interface, while the undulation of their Sn/IMC interfaces differs with the applied stress state. During the aging experiment, for the sample under stress-free condition after 100-h aging (Fig. 3b1), the Sn/IMC interface is basically straight with some protuberances (marked by downward arrows in Fig. 3) arranged on the IMC layer. With the increase in aging time (Fig. 3b2, b3), the IMC layer became thicker and the protuberance also grew larger. When aging time reaches 1000 h (Fig. 3b4), the Sn/IMC interface exhibits a zigzag morphology. However, for samples under compressive state, only a few protuberances can be found after 100-h aging (Fig. 3a1). The IMC layer got thicker with the increased aging time, while the growth of the protuberance in Fig. 3a2, a3 is not as obvious as that

Fig. 2 SEM images of **a** the cross section of as-reflowed general sample and **b** the top view of IMC phases after etching away the excessive tin

Fig. 3 Interfacial morphology evolution of samples under **a** compression, **b** stress free and **c** tension after (1) 100-h, (2) 250-h, (3) 500-h and (4) 1000-h aging

in stress-free samples (Fig. 3b2, b3). Finally, in the 1000-h aging sample (Fig. 3a4), the Sn/IMC interface is smoother compared with that in stress-free sample (Fig. 3b4) and the thickness of IMC layer is more uniform. In contrary, more protuberances can be found in samples under tensile state aged for 100 h (Fig. 3c1). The growth of the protuberance is more obvious during the aging treatment, but the growth of total IMC layer seems to be delayed (Fig. 3c2, c3). At last, Sn/IMC interface in sample aged for 1000 h is winging and rough as shown in Fig. 3c4. Overall, for three groups of samples, the total IMC thickness increases with the aging time, and the variation of interface morphology also becomes dramatic. It is necessary to mention that the IMC layer is continuous during aging without any break, since the deformation in the clamp is too small (compared with the arm length of the clamp) to generate effective bending force within sample. Therefore, the interfacial protuberance morphology in Fig. 3 reflects the real growth behavior of IMC during aging under different stress states.

3.2 Growth Rate and Kinetic Analysis of IMC

IMC thicknesses (x) of three groups of samples as a function of aging time (t) are plotted in Fig. 4a, and the average values of the IMC thickness of each sample are listed in Table 2. As illustrated in Fig. 4a, compared with sample under stress-free state, IMC in samples under compressive state exhibits higher growth rate. In contrast to compression, tension has an inhibition effect to the growth of IMC. In general, the relationship between IMC thickness (x) and aging time (t) can be described as a power function [18]:

$$x = k \left(\frac{t}{t_0} \right)^n, \quad (4)$$

where x is the thickness of IMC, k is the proportional coefficient, t is the aging time, and t_0 is the unit time, which is 1 h in this study. n is the indicator of growth mechanism, for interfacial reaction controlled by volume diffusion, $n = 0.5$; when n is close to 0.3, the interfacial reaction should be

Fig. 4 **a** IMC thickness as a function of aging time and **b** $\log x$ versus $\log t$, where x is the IMC thickness and t is the aging time

Table 2 Average IMC thickness of samples under different stress conditions for different aging time

Aging time (h)	Average IMC thickness (μm)		
	Compression	Stress free	Tension
100	2.70	2.74	2.40
250	4.75	4.19	3.43
500	5.70	5.53	4.86
1000	7.51	6.92	5.56

mainly controlled by grain boundary diffusion [19]. Equation (4) can be rearranged as:

$$\log x = \log k + n \log \left(\frac{t}{t_0} \right). \quad (5)$$

Thereby, n could be determined by the slope of $\log x$ - $\log t$ plot as drafted in Fig. 4b. As calculated and marked in Fig. 4b, $n=0.43$ for IMC growth under compressive condition, $n=0.41$ for stress-free condition and 0.38 for tensile condition. The growth exponents of all three conditions fall closely around 0.4, which manifests that the growth mechanism of IMC under different stress conditions is similar. They are all controlled by grain boundary diffusion and volume diffusion simultaneously. However, the IMC growth rate under compressive stress is higher than that under tensile stress.

3.3 Top View and Grain Size Distribution of IMC

Since the aging treatment was conducted at the same temperature, the vacancy concentration inside every IMC grain

should be similar [20], which means the volume diffusion contributed growth is also similar for all three groups of samples. Therefore, the different growth rate under different stress states should mainly be introduced by grain boundary diffusion. Therefore, the average size of IMC grains which is an estimator of grain boundary density should be an important indicator of the diffusion reaction mechanism. But IMC grain boundary is hard to be disclosed by selective etch technology, and only a few grains can be analyzed by electron back-scattered diffraction (EBSD). In this work, excessive Sn was removed by etching method as mentioned above. The top view images of the IMC grains in samples under three stress conditions after 1000-h aging are shown in Fig. 5. The top surface of compressive sample (Fig. 5a, d) is almost flat, and only a few large grains outgrow the surface, but for the tensile sample, the top surface (Fig. 5c, f) is almost full of bulky protruding grains. Stress-free sample has similar surface feature (Fig. 5b, e) but with less enlarged grains. As shown in Fig. 2a, each sandwiched solder joint has two interfaces. Figure 5a–c presents the interfacial IMCs at one side, while Fig. 5d–f shows those at the other side. It can be seen that there are no obvious differences in the IMC morphology between two opposite interfaces. Actually, since the as-reflowed Cu/Sn/Cu sample and the clamp are geometrically symmetric, it is reasonable to expect that both interfaces have identical features. In the meantime, these features and the evolution of the morphology are in well coincidence with the morphological features of the cross-sectional interface presented in Fig. 3.

In order to understand the different IMC growth rate of samples under different stress states, the average radius of IMC grains after 1000-h aging was calculated. The grain size distribution was presented as histograms, and

Fig. 5 Top view of the IMC grains under **a** compression, **b** stress free and **c** tension after 1000-h aging, **d–f** are the opposites interface of **a–c**, respectively

the results were fitted by normal distribution function in Fig. 6 and Table 3. It is concluded that the average IMC grain size of the compressive sample is about $2.74\text{ }\mu\text{m}$, while that of stress-free sample and tensile sample are about $3.37\text{ }\mu\text{m}$ and $3.91\text{ }\mu\text{m}$, respectively. Larger grain size means lower grain boundary density. As discussed before (Fig. 4b), the growth of IMC under these three stress conditions is jointly controlled by grain boundary diffusion and volume diffusion. So, the samples with higher grain boundary density (compressive samples) should have higher diffusion flux and therefore higher growth rate. The normal distribution fitting results (see Fig. 6 and Table 3) show that the IMC grain size of compressive sample has smaller expectations ($x_c = 2.42$), variance ($w = 1.69$), average radius ($r_{\text{aver}} = 2.74\text{ }\mu\text{m}$) and maximum radius ($r_{\text{max}} = 12.22\text{ }\mu\text{m}$) compared with stress-free sample ($x_c = 2.82$, $w = 1.83$, $r_{\text{aver}} = 3.37\text{ }\mu\text{m}$, $r_{\text{max}} = 13.42\text{ }\mu\text{m}$), which means that the average grain size of the compressive sample is smaller than the stress-free sample and its grain size distribution is more uniform. However, the tensile sample has opposite values ($x_c = 3.29$, $w = 1.89$, $r_{\text{aver}} = 3.91\text{ }\mu\text{m}$, $r_{\text{max}} = 14.04\text{ }\mu\text{m}$), and the corresponding IMC grain size is bigger. This is also consistent with the morphological differences presented in Fig. 5 and can be well understood by comparing the distribution fitting curves of three stress conditions as shown in Fig. 6d.

3.4 Growth Model of IMC Under Different Stress States

Based on the results discussed above, the growth behavior of IMC under different stress states can be described and is illustrated in Fig. 7. According to Ostwald ripening mechanism [21], during the initial stage of aging experiment, each IMC grain has a small and substantially uniform grain size (Fig. 7b1). As the aging process goes on, some grains with crystallographic priority would grow into advantage grains with larger grain size compared with nearby grains due to energy fluctuations (Fig. 7b2). Because the entire system tends to reduce its total surface energy, advantage grains are easier to grow and the growth of other grains would be inhibited and even be swallowed by advantage grains in the subsequent aging process. Finally, the grain size of advantage grains would surpass other grains excessively as shown in Fig. 7b3, and the entire surface morphology would appear like some super-large grains scattered on the background of small grains. The above description is the growth behavior of IMC grains in the general stress-free sample in this experiment. The morphological characteristics of IMC grains in the stress-free sample are shown in Fig. 3b, Fig. 5b, e, which are in good agreement with the mechanism described in Fig. 7b.

The growth of IMC is a volume expansion process which needs to overcome volumetric work to maintain the growth

Fig. 6 Histogram of grain size distribution with normal distribution fitting in the samples under **a** compression, **b** stress free and **c** tension after 1000-h aging, **d** the comparison of the normal distribution curves

Table 3 Statistical parameters of IMC grain size under different stress states after aging for 1000 h, where r_{max} represents maximum radius, r_{aver} represents average IMC grain size, x_c and w are the expectations and variance of the normal distribution fitting results, respectively

Parameters	Compression	Stress free	Tension
r_{max} (μm)	12.22	13.42	14.04
r_{aver} (μm)	2.74	3.37	3.91
x_c	2.42	2.82	3.29
w	1.69	1.83	1.89

of IMC grains [22]. One of the mechanisms of the occurrence of tin whiskers in solder joints is explained by volume expansion following oxide layer crack introduced by IMC growth [23]. Under the compressive condition, IMC grains need to overcome more volumetric work compared

with stress-free condition. Initially, the advantage grains have a higher growth rate than other grains, which means that they will expand more volume per unit time. That is to say, they will overcome more volumetric work to maintain their growth priority. Consequently, under the effect of compressive stress, the growth of advantage grains will be suppressed (Fig. 7a2), while other grains will get growth space. Finally, compared with the case without stress, the Ostwald ripening process is delayed in compressive state, and the average IMC grain size is reduced, and its size is evenly distributed around a certain value. Since the difference of grain size is reduced, the surface morphology of the entire IMC will become flat (Fig. 7a3). On the contrary, under the action of tensile stress, less volumetric work is required for IMC grains to grow. For the growth of advantage grains, the volume work to be overcome is reduced, and its rapid growth

Fig. 7 Illustration of different growth behaviors of IMC grains under stress states of **a** compression, **b** stress free, **c** tension during thermal aging

is further expanded. Hence the Ostwald ripening process is accelerated, and the overall grain size becomes larger and the size distribution is more dispersed. Therefore, the surface of the IMC has a waving feature with large undulations (Fig. 7c3). In addition, the increase in grain size represents the decrease in grain boundary density within the IMC layer. Since the growth of IMC is controlled by grain boundary diffusion and bulk diffusion together (Fig. 4b), the compressive sample has a smaller grain size and a larger grain boundary density; hence, the total grain boundary diffusion flux per unit time increases, resulting in a higher growth rate of IMC layer. On the contrary, the Ostwald ripening process becomes faster under tensile stress; the average IMC grain size increases; the grain boundary density decreases; and the growth rate is also reduced. Therefore, the final IMC layer of tensile state is thinner than that of stress-free and compressive states.

4 Conclusion

To investigate the effect of external normal stress on the IMC growth during thermal aging, a U-shape clamp was designed to apply different stress states on the interface of Cu/Sn/Cu solder joint. It is found that for all compressive,

stress-free and tensile stress states, the IMC growth is all controlled by grain boundary diffusion and volume diffusion simultaneously with a growth exponents (n) of about 0.4. However, compared with the general stress-free state, compression will enhance the growth rate of IMC and result in a smoother Sn/IMC interface. Besides, the IMC grain in compressive sample is more uniform and a smaller grain size. It is proposed that the Ostwald ripening process of IMC grain is delayed by compressive stress, resulting in a smaller grain size (average 2.74 μm after 1000-h aging) and higher grain boundary density to accelerate the IMC growth. On the contrary, tensile stress can accelerate the Ostwald ripening process, which leads to a larger grain size (average 3.91 μm after 1000-h aging) and a lower grain boundary density compared to the general stress-free state (the average IMC size is 3.91 μm after 1000-h aging). The Sn/IMC interface is rougher due to the existence of advantage grain, and the IMC growth rate is lower compared to that in general stress-free sample.

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References

- [1] R. Terrill, G.L. Beene, IEEE Aerosp. Appl. Conf. Proc. **2**, 347 (1996)
- [2] K.N. Tu, Y. Liu, Mater. Sci. Eng. R **136**, 1 (2019)
- [3] J.H.L. Pang, D.Y.R. Chong, T.H. Low, IEEE Trans. Compon. Packag. Technol. **24**, 705 (2001)
- [4] C. Noritake, P. Limaye, M. Gonzalez, B. Vandevelde, in *7th International Conference on Thermal, Mechanical and Multiphysics Simulation and Experiments in Micro-Electronics and Micro-Systems* (2006)
- [5] K.N. Tu, Microelectron. Reliab. **51**, 517 (2011)
- [6] S. Annuar, R. Mahmoodian, M. Hamdi, K.N. Tu, Sci. Technol. Adv. Mater. **18**, 693 (2017)
- [7] P.T. Vianco, J.J. Stephens, J.A. Rejent, IEEE Trans. Compon. Packag. Manuf. Technol. Part A **20**, 478 (1997)
- [8] S.L. Ngoh, W. Zhou, J.H.L. Pang, J. Electron. Mater. **37**, 1843 (2008)
- [9] T.C. Hu, F.C. Hsu, A.W. Huang, M.T. Lin, J. Electron. Mater. **41**, 3309 (2012)
- [10] D.W.A. Rees, *Mechanics of Solids and Structures*, 1st edn. (World Scientific Publishing Company, Singapore, 2000)
- [11] V. Tzimouli, H. Tsandali, ASM Metals Handbook, Vol 2 Properties and Selection: Nonferrous Alloys and Special-Purpose Materials, 10th ed. (1998)
- [12] Y. Yuan, D. Li, Y. Guan, H.J. Seifert, N. Moelans, J. Alloys Compd. **686**, 794 (2016)
- [13] R. Zhang, Y. Tian, C. Hang, B. Liu, C. Wang, Mater. Lett. **110**, 137 (2013)
- [14] M. Onishi, H. Fujibuchi, Trans. Jpn. Inst. Met. **16**, 539 (1975)
- [15] T. Takenaka, S. Kano, M. Kajihara, N. Kurokawa, K. Sakamoto, Mater. Sci. Eng. A **396**, 115 (2005)
- [16] V.A. Baheti, S. Kashyap, P. Kumar, K. Chattopadhyay, A. Paul, J. Alloys Compd. **727**, 832 (2017)
- [17] V.I. Dybkov, Reaction Diffusion and Solid State Chemical Kinetics (2002)
- [18] Y.C. Cheng, Y.T. Wang, F.C. Hsu, F.C. Lu, C.L. Wu, M.T. Lin, J. Electron. Mater. **44**, 604 (2015)
- [19] Y.L. Corcoran, A.H. King, N. de Lanerolle, B. Kim, J. Electron. Mater. **19**, 1177 (1990)
- [20] W.K. Liao, C.M. Chen, M.T. Lin, C.H. Wang, Scr. Mater. **65**, 691 (2011)
- [21] J. Almer, *Growth and Coarsening, Ostwald Ripening in Material Processing* (Springer, Berlin, 2002)
- [22] N. Jadhav, E.J. Buchovecky, L. Reinbold, S. Kumar, A.F. Bower, E. Chason, IEEE Trans. Electron. Packag. Manuf. **33**, 183 (2010)
- [23] K.S. Kumar, L. Reinbold, A.F. Bower, E. Chason, J. Mater. Res. **23**, 2916 (2008)