

Early-Stage Interfacial Intermetallic Growth and Local Micromechanical Response of Sn-3.0Ag-0.5Cu Joints on Immersion-Tin-Finished Cu during Thermal Cycling

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(Received: 5 September 2026 / Revised: 15 October 2026 / Accepted: 21 October 2026 / Published online: 24 October 2026)

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ABSTRACT

Sn-3.0Ag-0.5Cu (SAC305) solder joints were reflowed on immersion-tin (ImSn)-finished Cu pads. The joints were then cycled between 0 and 100 °C for 0, 100, 200 and 300 cycles. The aim was to characterize the earliest stage of interfacial and matrix evolution before macroscopic damage develops. Optical microscopy, scanning electron microscopy, energy-dispersive X-ray spectroscopy (EDS) and depth-sensing nanoindentation were used to characterize the samples. The mean thickness of the interfacial intermetallic compound (IMC) layer increased from 3.76 µm in the as-reflowed state to 3.98, 4.05 and 4.37 µm after 100, 200 and 300 cycles, respectively. This is a net increase of 16.2%. Over this early interval a linear dependence on cycle number described the data better ($R^2 = 0.95$) than a parabolic, diffusion-controlled law ($R^2 = 0.85$). The growth rate was about 1.9 nm per cycle. EDS of the interfacial band was 48.8 at.% Cu and 51.2 at.% Sn. This composition is consistent with Cu₆Sn₅ as the dominant reaction product. After 300 cycles, the hardness of the solder matrix decreased by 12.0%, from 0.221 to 0.195 GPa. The hardness of the solder adjacent to the IMC decreased by 7.2%, from 0.224 to 0.208 GPa. The near-interface solder remained harder than the matrix in both states. Both reductions agree within 1% with those predicted from the increase in maximum penetration depth. The reduced modulus scattered more than the difference between conditions. This scatter is consistent with β-Sn elastic anisotropy and indentation pile-up ($h_f/h_{max} \geq 0.95$). The results show that mild cycling thickens the Cu-Sn interfacial

layer and softens the adjacent solder at the same time. Consequently, the local strength mismatch across the joint increases well before any crack is detectable.

Keywords: Electronic packaging; immersion tin; intermetallic compound; nanoindentation; solder joint reliability

INTRODUCTION

Sn-Ag-Cu (SAC) alloys are the standard Pb-free interconnect materials in electronic assemblies. Sn-3.0Ag-0.5Cu (SAC305) is the most widely used composition. Each solder joint must provide electrical, thermal and mechanical continuity at the same time [1]. SAC305 melts between about 217 and 220 °C. Its service temperature therefore corresponds to a homologous temperature of about 0.6. As a result the mechanical response of the joint depends strongly on time, temperature and strain rate. The microstructure is also inherently unstable in service [1]. In addition typical joints contain only one or a few grains of body-centered tetragonal β -Sn. Along the c -axis, the coefficient of thermal expansion of β -Sn is about two times the value along the a -axis. Its Young's modulus along the c -axis is about three times the value along the a -axis [2,3,4]. Nominally identical joints can therefore show markedly different local elastic, thermal and plastic responses.

The surface finish of the printed circuit board controls the reaction path between the molten solder and the underlying Cu. An ImSn finish does not act as a diffusion barrier. The thin Sn deposit dissolves into the Sn-rich melt during reflow. The joint interface is then governed by the Cu-Sn reaction. Scalloped Cu_6Sn_5 forms next to the solder. A thinner Cu_3Sn layer may form next to Cu when the thermal exposure is sufficient [5,6,7,8]. Cu and Sn also react measurably at room temperature [9]. Part of an ImSn deposit may therefore convert to Cu_6Sn_5 during storage before assembly. Surface finish influences both IMC morphology and joint strength. Muhd Amlı et al. [10] reported that ImSn produced smaller and more numerous primary Cu_6Sn_5 crystals than immersion silver. ImSn also gave higher joint strength. Similar finish-dependent differences exist between organic solderability preservative and electroless nickel/immersion gold finishes [11]. Recent work continues to use the surface finish to control interfacial IMC growth during aging [12].

Two processes act on the joint during temperature cycling. First, time spent at the upper temperature promotes Cu-Sn interdiffusion. The interfacial layer thickens as a result [13,14,15]. Second, the Cu pad, laminate, attached components and anisotropic β -Sn grains undergo different amounts of thermal expansion. This thermal-expansion mismatch imposes cyclic viscoplastic strain. Over longer or more severe cycling the strain drives recovery, subgrain formation, recrystallization and Ag_3Sn coarsening. Intergranular cracks eventually form in the high-strain regions of the joint [3,4,16,17,18,19,20]. Anisotropy alone can generate local damage in free-standing Sn-rich specimens [21]. No external constraint is needed. These mechanisms were established by electron backscatter diffraction (EBSD) including in-situ observations [22]. They were also interpreted with crystal-plasticity modeling [4,23].

Thermal exposure also changes the local mechanical properties of the solder. Isothermal aging reduces the shear strength and apparent stiffness of SAC305 joints [14],

[24]. Aging-induced microstructural changes have also been correlated with nanoindentation response [25]. Fahim et al. cycled SAC305 ball-grid-array joints between -40 and 125 °C [26]. Both the modulus and the hardness of the solder decreased significantly within 250 cycles. The Cu pad was unaffected. Nanoindentation suits such studies because it probes micrometer-scale volumes. However, the measured response depends on β -Sn orientation. It also depends on the distance to hard intermetallic phases [27,28,29].

Most detailed studies addressed the long-term aging, thousands of cycles, or harsh profiles such as -40 to 125 °C. Many other studies also use Ni-based or organic finishes [16,30,31,32]. Much less attention has been paid to mild cycling between 0 and 100 °C. It is not yet clear whether such a profile changes both the Cu-Sn interface and the local mechanical response of SAC305 on ImSn-finished Cu within the first few hundred cycles. Previous work on SAC305/ImSn joints under thermal cycling focused on the directional growth of the interfacial IMC [33]. The early stage is important. Interfacial and matrix evolution may be detectable well before a crack or a failure signal appears. This early state is the reference against which later degradation must be judged.

The present study addresses this gap. The interfacial IMC thickness of SAC305/ImSn/Cu joints was measured after 0 , 100 , 200 and 300 cycles between 0 and 100 °C. The elemental composition of the interfacial band was determined by EDS. The local hardness and reduced modulus of the solder matrix and of the solder adjacent to the IMC were compared in the as-reflowed and 300 cycle states. The working hypothesis is that early cycling produces measurable interfacial growth and softens the solder. This combination would increase the mechanical mismatch across the joint. Throughout the paper, phase assignments and mechanistic interpretations are restricted to what the available evidence can support.

MATERIALS AND METHOD

Commercial test boards with ImSn-finished Cu pads were supplied by RedRing Solder (M) Sdn. Bhd. (Figure 1(a)). Commercial SAC305 solder paste was printed manually onto the pads. A 6-mil (≈ 152 μm) stainless-steel stencil was used (Figure 1(b)). The paste was reflowed in a convection reflow oven (Madell Technology Corporation) at a set temperature of 225 °C for 8 s. This temperature is slightly above the melting range of the alloy [1]. The assemblies were cooled in still air to room temperature. They were then sectioned into coupons with a diamond cut-off wheel.

The coupons were cycled in air between 0 and 100 °C. These limits correspond to the nominal extremes of Test Condition J of JESD22-A104 [34]. The ramp rate was 10 °C min^{-1} . The dwell time was 60 min at each temperature extreme. One complete cycle therefore lasted about 140 min. Specimens were withdrawn after 100 , 200 and 300 cycles. Uncycled joints served as the as-reflowed reference.

The specimens were cold-mounted in epoxy. They were ground successively on 800 , 1000 , 1200 and 2400 grit SiC papers. They were then polished with 6 , 3 and 1 μm diamond suspensions. Selected cross sections were etched in 2 vol.% HNO_3 in methanol to reveal the solder microstructure. The solder/Cu interface was examined by optical microscopy and by field-emission scanning electron microscopy (ZEISS GeminiSEM 500). EDS spectra were acquired from rectangular areas within the interfacial band and

in the adjacent solder. The spectra were quantified in weight percent by the standardless routine of the acquisition software. The values were then converted to atomic percent. The average IMC thickness, $\bar{x}$, was determined from cross-sectional micrographs as:

$$\bar{x} = \frac{A_{IMC}}{L} \quad (1)$$

where A_{IMC} is the measured area of the interfacial layer and L is the corresponding projected interface length. Seven fields were measured for each cycle count and averaged. Eq. (1) integrates over the scallop profile. Therefore, $\bar{x}$ is an area-equivalent thickness rather than a local scallop height.

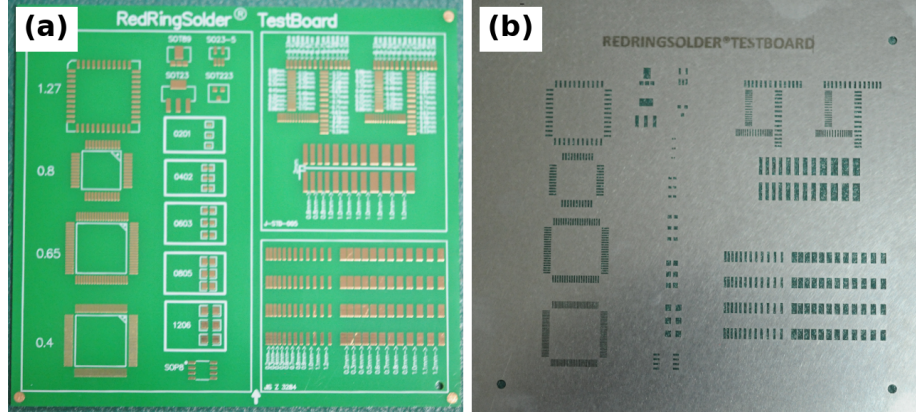


Figure 1. (a) ImSn-finished test board and (b) 6-mil stencil used to print the SAC305 solder paste

Depth-sensing nanoindentation was performed at room temperature on polished cross sections of the as-reflowed and 300-cycle specimens. A diamond indenter of Berkovich geometry was used. Each test was load-controlled to a peak load P_{max} of 10 mN. The loading and unloading rates were set to 0.5 mN min^{-1} . The dwell at peak load was 30 s. Indents were placed in two regions. The first region was the solder matrix away from the interface. The second region was the solder immediately adjacent to the IMC layer, hereafter called the near-IMC solder. Hardness H and reduced modulus E_r were obtained by the Oliver-Pharr method [35,36] as follows:

$$H = \frac{P_{max}}{A_c} \quad (2)$$

$$E_r = \frac{\sqrt{\pi}}{2\beta} \cdot \frac{S}{\sqrt{A_c}} \quad (3)$$

$$h_c = h_{max} - \varepsilon \frac{P_{max}}{S}. \quad (4)$$

Here, S is the initial unloading stiffness and A_c is the projected contact area at the contact depth h_c . The term β is a geometry correction factor and $\varepsilon = 0.75$ for a Berkovich tip. The specimen modulus E_s was estimated from:

$$\frac{1}{E_r} = \frac{1 - \nu_s^2}{E_s} + \frac{1 - \nu_i^2}{E_i} \quad (5)$$

where $E_i = 1141$ GPa and $\nu_i = 0.07$ for diamond. A Poisson's ratio ν_s of 0.36 was assumed for SAC305. Five indents were programmed per specimen. An indent was retained only if it could be traced clearly to a specimen condition and an indentation region. Its unloading curve also had to give a physically meaningful fit. One 300-cycle matrix indent failed this test. Its fitted E_r of 1126 GPa exceeded the modulus of diamond. Its fit error was also more than two orders of magnitude larger than those of the accepted indents.

RESULTS AND DISCUSSION

Figure 2 shows optical cross sections of the interface after 0, 100, 200 and 300 cycles. In the as-reflowed state (Figure 2(a)), a continuous and non-planar reaction band lines the entire solder/Cu boundary. Its shape is consistent with a scalloped Cu-Sn reaction front rather than a residual Sn coating. No separate ImSn layer can be distinguished. This agrees with the dissolution of the deposit into the melt during reflow [5]. The band remains continuous after 100, 200 and 300 cycles (Figs. 2b to 2d). It also becomes progressively more distinct with cycling. Occasional dark features appear along the interface, most clearly in Figure 2(b). Their origin cannot be resolved at this magnification. They may be polishing relief, particle pull-out, or voids. These features were excluded from the thickness measurements.

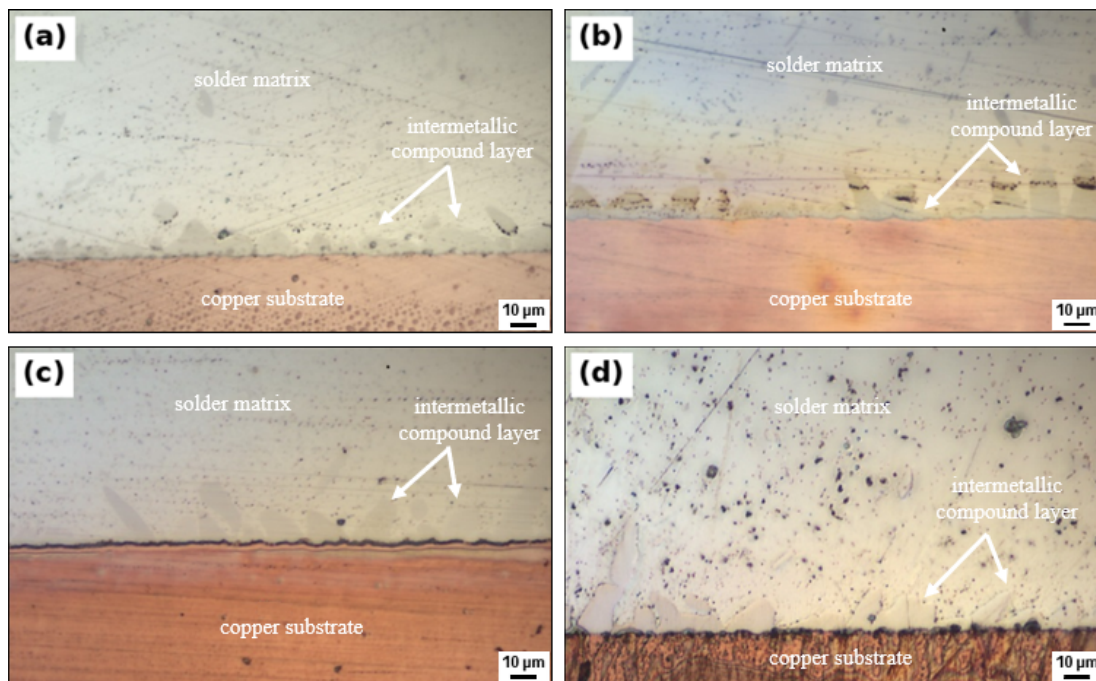


Figure 2. Optical cross sections of the SAC305/ImSn/Cu interface after (a) 0, (b) 100, (c) 200 and (d) 300 thermal cycles between 0 and 100 °C

Figure 3 shows the etched solder microstructure. In the as-reflowed state (Figure 3(a)), the matrix consists of coarse β -Sn regions. These regions are separated by darker-etching eutectic networks. Faceted Cu_6Sn_5 scallops protrude from the interface into the

solder. After 200 cycles (Figure 3(b)), the etched contrast is finer and more uniform. After 300 cycles (Figure 3(c)), the contrast also becomes banded and directional. These observations show a change in the scale and heterogeneity of the etched features. They do not prove β -Sn grain refinement or recrystallization. Optical etch contrast in Sn-rich solder also depends on crystal orientation, etching response and the distribution of intermetallic particles. Such assignments require orientation mapping [16,19].

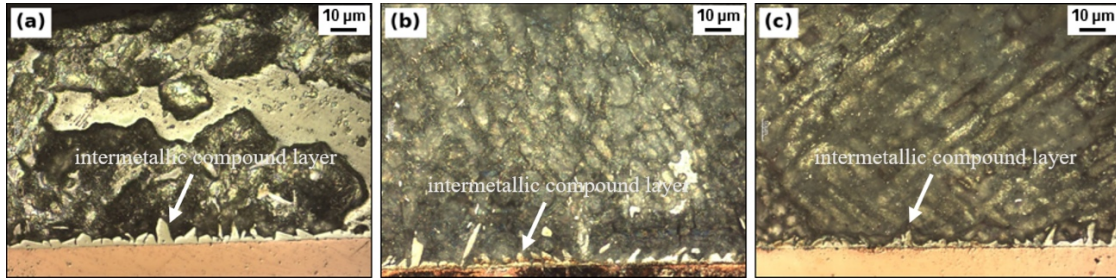


Figure 3. Etched cross-sectional microstructures of SAC305/ImSn/Cu joints after (a) 0, (b) 200 and (c) 300 thermal cycles

Figure 4 shows a representative electron image of the interface and the corresponding EDS spectra. Table 1 lists the quantified compositions. The area placed across the interfacial band (Spectrum 9) gave 33.8 wt.% Cu and 66.2 wt.% Sn. These values correspond to 48.8 at.% Cu and 51.2 at.% Sn. This composition is close to that of Cu_6Sn_5 (54.5 at.% Cu). It is far below that of Cu_3Sn (75.0 at.% Cu). The area in the solder just above the band (Spectrum 8) contained 24.2 at.% Cu. This value is well above the Cu content of the SAC305 matrix. The analyzed volume there therefore includes Cu_6Sn_5 scallops or particles embedded in β -Sn.

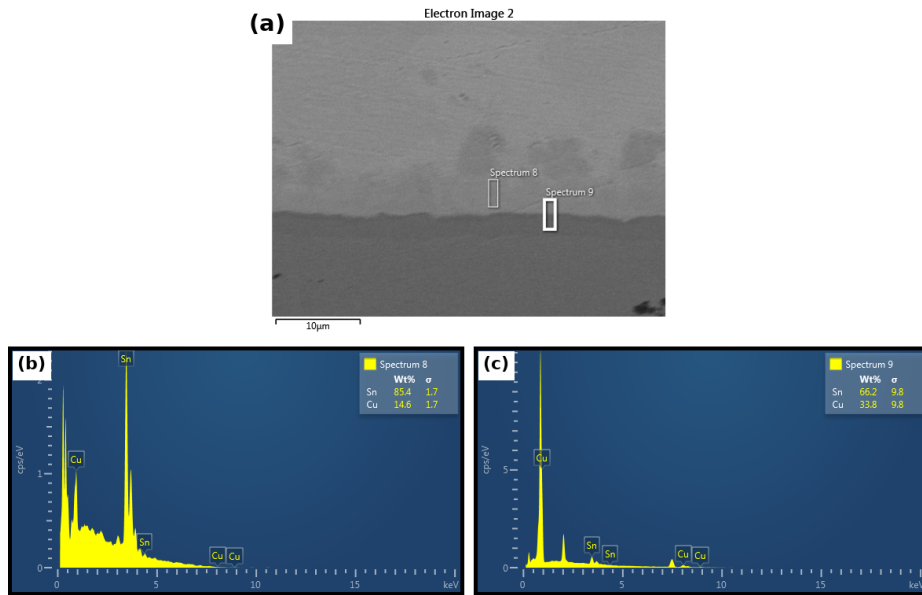


Figure 4. (a) Electron image of the SAC305/Cu interface showing the EDS analysis areas. (b) EDS spectrum from the solder adjacent to the interfacial band (Spectrum 8). (c) EDS spectrum from the interfacial band (Spectrum 9).

The composition of the interfacial band is consistent with Cu_6Sn_5 as the dominant reaction product. This phase is expected at a SAC305/Cu interface [5,6]. No composition approached 75 at.% Cu. Hence, no Cu_3Sn layer thick enough to be analyzed separately was present. A thin Cu_3Sn sublayer next to the pad may still exist. Such a layer can form during reflow or early aging [14,31]. It cannot be excluded at the present spatial resolution. The interfacial product is therefore designated as a Cu-Sn IMC consistent with Cu_6Sn_5 .

Table 1. EDS Compositions of the interfacial region

Location	Cu (wt.%)	Sn (wt.%)	σ (wt.%)	Cu (at.%)	Sn (at.%)
Spectrum 8 (solder adjacent to band)	14.6	85.4	1.7	24.2	75.8
Spectrum 9 (interfacial band)	33.8	66.2	9.8	48.8	51.2

Table 2 lists the mean IMC thickness at each cycle count. In the as-reflowed state, the mean thickness was 3.76 μm . This value is substantial for such a brief reflow. The set temperature was only a few degrees above the liquidus. For comparison, Hamasha et al. reported about 2 μm for SAC305 on Cu after reflow at 245 °C for 45 s [14]. Several factors may contribute. Cu-Sn compounds form even at room temperature [9]. Part of the ImSn deposit may therefore have converted to Cu_6Sn_5 by solid-state reaction before assembly. This pre-existing compound would become part of the interface after reflow. In addition, the area-based measure of Eq. (1) includes the scallop profile. This profile is pronounced in the as-reflowed state (Figure 3(a)) [7,8].

Table 2. Mean interfacial IMC thickness as a function of thermal-cycle count

Thermal cycles	Approx. elapsed time (h)	Mean thickness (μm)	Increase (μm)	Increase (%)
0	0	3.76	0	0
100	233	3.98	0.22	5.9
200	467	4.05	0.29	7.7
300	700	4.37	0.61	16.2

Each value is the mean of seven measurement fields evaluated with Eq. (1). The elapsed time is based on a cycle period of about 140 min.

The thickness then increased to 3.98, 4.05 and 4.37 μm after 100, 200 and 300 cycles. The total increase of 0.61 μm is 16.2% of the initial thickness. IMC growth in Sn-rich solder/Cu couples is commonly described by:

$$x - x_0 = kt^n \quad (6)$$

where x is the thickness after exposure time t and x_0 is the as-reflowed thickness. The term k is a thermally activated rate constant and n is a time exponent. A value of $n = 0.5$ indicates volume-diffusion control. A value of $n = 1$ indicates interface-reaction control [5,15]. Here, the cycle number N was used as a proxy for time. A linear fit gave $x = 3.755 + 1.90 \times 10^{-3}N \mu\text{m}$, with $R^2 = 0.95$. A fit against $\sqrt{N}$ gave $R^2 = 0.85$ (Figure 5). A free power-law fit to the three increments gave $n \approx 0.87$. With only three points, however, this exponent is poorly constrained. An effective parabolic coefficient was also calculated for comparison with isothermal data. Using the total elapsed time of 700 h, $k_{\text{eff}} = (x^2 - x_0^2)/t = 2.0 \times 10^{-18} \text{ m}^2 \text{ s}^{-1}$.

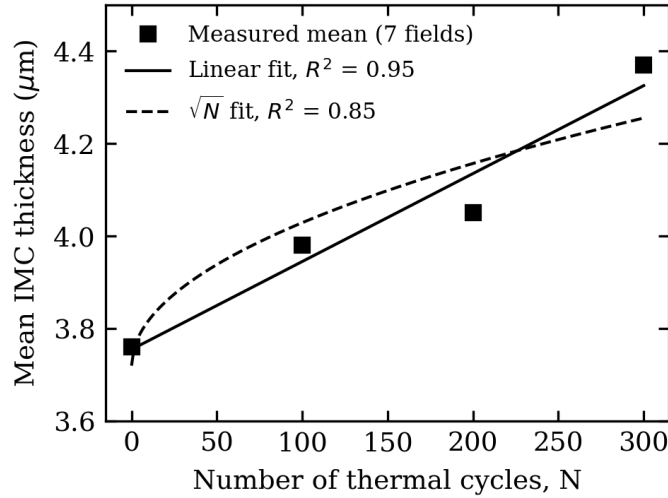


Figure 5. Mean interfacial IMC thickness versus number of thermal cycles. Linear and parabolic ($\sqrt{N}$) least-squares fits are shown

The better linear fit should be interpreted with caution. It rests on four means without dispersion data. Nevertheless, a near-linear early stage is physically plausible. Scalloped Cu_6Sn_5 grows by fast diffusion along the channels between scallops. This growth is coupled with ripening [7]. During solid-state exposure, the layer also changes from a scalloped to a more planar shape. This change alters the effective kinetics. A single parabolic law is therefore not expected to hold from the as-reflowed state onwards [7], [15]. In addition, exposure during cycling is intermittent. Growth occurs mainly during the upper dwell and the upper part of each ramp. Li et al. showed that IMC growth and fracture in SnAgCu/Cu joints depend on the thermal path and not only on elapsed time [13]. The coefficient k_{eff} is therefore a lower bound for the growth rate at 100°C .

Figure 6 shows representative load-displacement curves. Table 3 lists the individual retained indents and Figure 7 plots them. All curves showed a large increase in displacement during the dwell at peak load. The increase was 143 nm for the as-reflowed matrix indent. It was 148 nm for the 300-cycle near-IMC indent. These values are about 10.6% of the maximum depth. They reflect the strong indentation creep of Sn-rich solder at room temperature. Elastic recovery on unloading was small. The ratio of final to maximum depth, h_f/h_{max} , was 0.95 and 0.98 for the two curves in Figure 6.

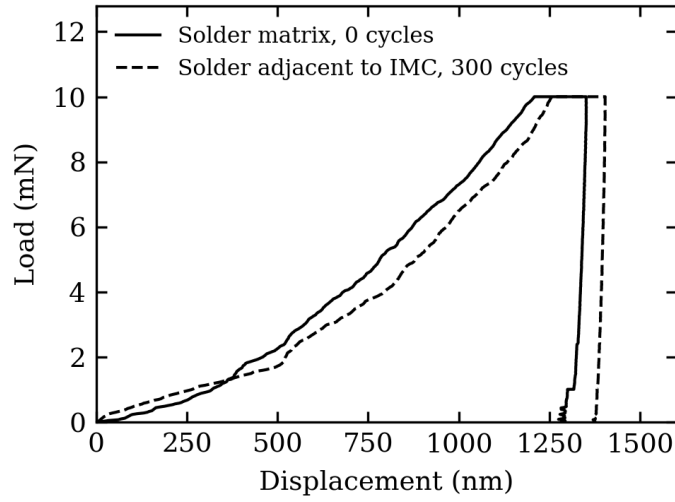


Figure 6. Load-displacement curves at a peak load of 10 mN. The curves are for an indent in the solder matrix of the as-reflowed joint and an indent in the solder adjacent to the IMC layer after 300 cycles.

Table 3. Individual nanoindentation results at a peak load of 10 mN

Region / condition	$h_{\max}$ (nm)	h_c (nm)	H (GPa)	E_r (GPa)	E_s (GPa)
Matrix, 0 cycles	1351.8	1334.6	0.221	89.0	84.0
Matrix, 300 cycles (1)	1429.3	1419.2	0.196	55.5	50.8
Matrix, 300 cycles (2)	1442.3	1431.3	0.193	81.4	76.2
Near-IMC, 0 cycles (1)	1317.2	1305.1	0.231	81.0	75.9
Near-IMC, 0 cycles (2)	1357.5	1349.0	0.217	111.4	107.4
Near-IMC, 300 cycles (1)	1367.5	1365.4	0.212	58.8	53.9
Near-IMC, 300 cycles (2)	1403.8	1392.7	0.204	82.8	77.7

h_c is the contact depth reported by the instrument. E_s was calculated from E_r using Eq. (5) with $\nu_s = 0.36$. Values are for individual indents.

In the solder matrix, H decreased by 12.0%. It fell from 0.221 GPa in the as-reflowed state to a mean of 0.195 GPa after 300 cycles. Over the same interval, the mean maximum depth increased by 6.2%, from 1351.8 to 1435.8 nm. For a self-similar indenter at constant load, H is proportional to h^{-2} . The depth increase therefore predicts a hardness reduction of 11.4%. This prediction is within 1% of the measured value. The reduction is thus not an artefact of unloading-curve fitting.

This softening agrees in direction with the degradation reported for aged and cycled SAC305. Aging reduces the shear strength and stiffness of the joint [14,24]. Cycling between -40 and 125 °C reduces the nanoindentation hardness and modulus within 250 cycles [26]. The present data show that a similar softening is already detectable under the much milder profile of 0 to 100 °C.

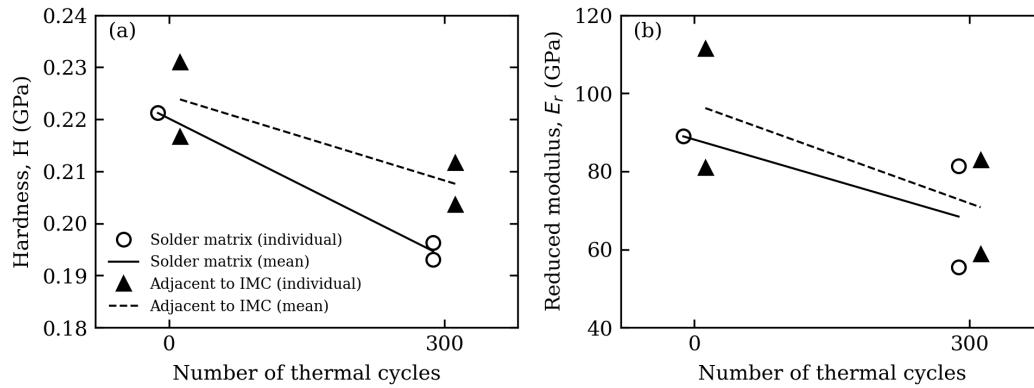


Figure 7. (a) Hardness and (b) reduced modulus of individual indents in the solder matrix and in the solder adjacent to the IMC, before cycling and after 300 cycles

This softening agrees in direction with the degradation reported for aged and cycled SAC305. Aging reduces the shear strength and stiffness of the joint [14,24]. Cycling between -40 and 125 °C reduces the nanoindentation hardness and modulus within 250 cycles [26]. The present data show that a similar softening is already detectable under the much milder profile of 0 to 100 °C.

Three mechanisms may explain the softening. They are not mutually exclusive. First, Ag_3Sn particles provide much of the dispersion strengthening of SAC305. These particles coarsen during thermal exposure and cyclic strain accelerates the coarsening. Tian et al. quantified such strain-enhanced coarsening in SAC305 micro-joints [18]. Coarsening increases the interparticle spacing. It therefore reduces the Orowan contribution to the flow stress [16,17]. Second, 100 °C corresponds to a homologous temperature of about 0.76. At this temperature, the dislocation substructure retained from solidification and cooling can recover rapidly. Third, cyclic strain can form subgrains. In highly strained regions, it can also form recrystallized grains [16,17,19]. The optical evidence in Figure 3 is consistent with microstructural change. However, it cannot separate these mechanisms.

The cyclic strain in these coupons is also expected to be modest. No component was mounted on the coupons. The global expansion mismatch that dominates package-level thermomechanical fatigue is therefore largely absent. The strain comes only from local sources. These sources include β -Sn anisotropy, the Sn/Cu mismatch and the Cu/laminate mismatch transmitted through the pad [20,21]. The observed softening is therefore likely dominated by thermally driven aging processes. Locally strain-enhanced coarsening probably makes a secondary contribution.

The near-IMC solder followed a similar trend. Its mean H decreased by 7.2%, from 0.224 to 0.208 GPa. Its mean maximum depth increased by 3.6%. This depth change predicts a hardness reduction of 6.8%. The prediction is again within 1% of the measured value. The near-IMC solder was harder than the matrix in both states. The difference was 1.2% before cycling and 6.6% after 300 cycles. The near-IMC solder therefore softened less than the matrix.

Two effects are likely responsible. First, the indentation is not confined to the solder. For a Berkovich indent with a contact depth of about 1.35 μm , the impression edge

is about 10 μm long. The plastic zone extends even further. An indent placed near a 4 μm thick IMC layer is therefore constrained by it. Cu_6Sn_5 is much stiffer and harder than solder. Its modulus is about 110 to 125 GPa and its hardness is several GPa [28,29,37]. Cu is also much stiffer than SAC305. The highest single E_r value (111.4 GPa, near-IMC, as-reflowed) most likely reflects this elastic contribution from the substrate side. Second, the solder close to the interface contains more Cu_6Sn_5 particles and scallop protrusions. This is shown by the 24.2 at.% Cu in Spectrum 8 and by the protrusions in Figure 3(a). This composite-like region is stronger than the bulk matrix.

The growing hardness contrast has a direct mechanical consequence. The bulk solder softens, while the solder next to the thickening and brittle IMC layer keeps more of its strength. Cyclic deformation will then tend to concentrate in the softer matrix just beyond the stiff interfacial zone. Recrystallization and crack paths are commonly observed in this region in thermally cycled SAC305 joints [4,16]. Prolonged aging or cycling eventually shifts fracture toward the interfacial Cu_6Sn_5 [24,31,32]. The present results suggest that the conditions for this localization begin to develop within the first few hundred mild cycles.

The reduced modulus behaved differently. The condition means decreased from 89.0 to 68.4 GPa in the matrix. They decreased from 96.2 to 70.8 GPa in the near-IMC solder. However, the spread within a single condition was as large as the difference between conditions. The two valid 300-cycle matrix indents gave 55.5 and 81.4 GPa. The two as-reflowed near-IMC indents gave 81.0 and 111.4 GPa. The E_r data therefore do not resolve an effect of cycling on stiffness.

The E_r values of 55 to 111 GPa are also higher than the Young's modulus usually quoted for bulk SAC305, which is about 40 to 55 GPa [1]. Three factors explain this behaviour. First, the unloading curves indicate substantial pile-up. The ratio h_f/h_{max} exceeded 0.95. Pile-up becomes significant in the Oliver-Pharr analysis when this ratio exceeds about 0.7 and work hardening is limited. The contact area is then underestimated and both H and E_r are overestimated [36,38]. Second, β -Sn shows a three-fold elastic anisotropy. Indents in grains of different orientation can therefore return very different moduli [4,27]. Without orientation data, the modulus of each indent cannot be normalized. Third, the strong creep during the 30 s dwell (Figure 6) can continue into the start of unloading. This creep distorts the stiffness S , especially in soft and rate-sensitive solders [27,36]. For these reasons, the E_r values are treated as comparative indicators only. Hardness is less affected. The same pile-up bias applies to all indents and the depth-based check confirmed the relative change.

CONCLUSIONS

The early response of SAC305 joints on ImSn-finished Cu to temperature cycling between 0 and 100 $^{\circ}\text{C}$ was characterized by optical microscopy, SEM-EDS and nanoindentation. A continuous Cu-Sn IMC layer formed at the interface during reflow. EDS gave 48.8 at.% Cu for this layer. This value is consistent with Cu_6Sn_5 as the dominant phase. No Cu_3Sn layer thick enough to be analyzed was detected. The IMC layer thickened from 3.76 to 4.37 μm over 300 cycles. This is an increase of 16.2%. Over this early interval, a linear dependence on cycle number described the growth better than a parabolic law. This behavior is consistent with scallop-controlled and intermittent early

growth. After 300 cycles, the hardness of the solder matrix decreased by about 12%. The hardness of the solder adjacent to the IMC decreased by about 7%. Both changes agree with the measured increase in penetration depth. They also agree in direction with the softening reported for aged and more severely cycled SAC305. The solder adjacent to the IMC remained harder than the matrix. The hardness contrast increased from about 1% to about 7%. This indicates a developing strength gradient that favors strain localization near the interface. The reduced modulus was dominated by pile-up, β -Sn anisotropy and creep effects. It did not resolve a cycling effect. The micromechanical trends rest on one or two indents per condition. They are therefore indicative. Nevertheless, the combined evidence shows that interfacial and matrix evolution begins within the first few hundred mild cycles. These results provide a reference state for assessing later degradation. Future work should include intermediate cycle counts, larger indent populations and orientation-resolved characterization.

ACKNOWLEDGMENTS

The authors wish to express their sincere gratitude to Universiti Kebangsaan Malaysia (UKM) (grant code: GUP-2024-055). The authors would like to thank RedRing Solder (M) Sdn. Bhd. for supplying the test boards. The authors also thank the Microelectronic Packaging and Materials (MIPAC) Laboratory for the use of their metallography facilities and extend their appreciation to MIPAC Science Officer, Mr. Saiful Bahari Bakarudin, for conducting the nanoindentation testing.

AUTHOR CONTRIBUTIONS

Maria Abu Bakar: conceptualization, methodology, formal analysis, investigation, data collection visualization, writing (original draft). Azman Jalar, Mohamad Riduwan Ramli and Norliza Ismail: data interpretation and writing (review and editing).

CONFLICT OF INTEREST

The authors declare no competing interests.

DATA AVAILABILITY

The data supporting the findings of this study are available from the corresponding author upon reasonable request.

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