Reliability analysis of solder joints due to creep and fatigue in microelectronic packaging using microindentation technique

D. Chicot a, K. Tilkina a, K. Jankowski b, A. Wymysłowski b,*

a University Lille Nord of France, USTL, LML, CNRS, UMR 8107, F-59650 Villeneuve d’Ascq, France
b Wrocław University of Technology, Wyb. Wyspiańskiego 27, 50-370 Wrocław, Poland

**Abstract**

Solder joints in microelectronics are used for electrical signals transmission, heat conduction and structural support. One of the key problems referring to solders in microelectronics is reliability due to typical failure modes as creep and fatigue. The above paper focuses on the experimental measurements and corresponding analysis with the microindentation tests of the SAC 405 solder alloy due to creep and fatigue. The creep, resulting from the application of a constant load during a long time, is represented by an original law between the indenter displacement and time. The fatigue due to repeated loading–unloading cycles is characterized by the law of Manson–Coffin which is adapted for connecting the plastic indentation strain to the number of cycles.

© 2013 Elsevier Ltd. All rights reserved.

1. Introduction

In microelectronic packaging, the solder joints constituted of tin–lead alloy until these last decades have played a major role in the reliability of the components. But due to the restriction of the use of certain hazardous substances such as lead in electrical and electronic equipment (RoHS regulation) [1], new alloys for solder joints have been developed. Among them, the SAC 405 alloy (S–SnAg4Cu0.5) takes place as a lead-free solder to replace the common tin–lead eutectic alloy (63/37) [2]. Nevertheless, the SAC 405 alloy tends to be more brittle due to a high stiffness and excessive solder interfacial reactions. As mentioned by Chin et al. [2], this leads to higher occurrences of solder joints failure during the dwell-time at a constant load is representative of a creep phenomenon by indentation and the measurement of compressive stress (mechanical loading), temperature cycles (thermal loading), the current flow (electrical loading) or the humidity [3].

On the other hand, the development during these two last decades of the instrumented indentation equipment allows for a large variety of mechanical testing conditions by indentation. In fact, using load–displacement cycles of the indenter together with different loading and unloading rates allows for analysis of the visco-elastic-plastic behavior of materials. Moreover, the maximum indentation load can be maintained during a given time period where the indenter continues to penetrate into the material for which the magnitude of penetration depends on the mechanical behavior of the tested material. Such an indenter displacement observed during the dwell-time at a constant load is representative of a creep phenomenon by indentation and the measurement of power-law creep parameters by instrumented indentation methods, which has been extensively discussed by Su et al. [4]. Besides, successive cycles can be performed using the same maximum load with increasing loading mode to assess the plastic characterization beneath the indenter, thus the phase transformation under indentation stress or the failure of coatings as mentioned by Xu et al. [5] when studying the indentation fatigue behavior of polycrystalline copper. Consequently, instrumented indentation seems to be a suitable technique for separately or jointly realizing coupled indentation conditions, i.e. creep and/or fatigue for example. This is one of the main advantages of such mechanical testing thus allowing to analyze the mechanical behavior of tested material under multiple mechanical solicitations [6,7]. In this work, a methodology based on the use of the microindentation technique is proposed to study separately creep and fatigue phenomena on the SAC 405 alloy. The objective to model coupled solicitations by indentation is currently under investigation and will be
presented in the near future. To limit the number of testing conditions and to compare with the Failure and Reliability Investigation System [3], the indentation testing conditions for creep and fatigue analyses have been chosen according to the FRIS testing conditions.

2. Experimental analysis

Microindentation experiments were carried out with a microhardness CSM 2-107 Tester equipped with a Vickers diamond indenter. The load range of the instrument is from 0.1 to 20 N. The load resolution is 100 μN and the depth resolution is 0.3 nm, these values being provided by the CSM Instruments Group. In this work, microindentation creep tests have been performed at a constant loading rate of 100 N/min with holding loads ranging from 1 to 20 N to investigate the indentation creep behavior over 1800 s of dwell-time. The microindentation fatigue tests have been performed during 100 cycles at constant maximum holding loads ranging between 1 and 20 N with a null dwell-time to avoid creep phenomenon in this test. The loading and unloading rates have been chosen in order to obtain 1, 2.5, 5, 10 and 15 cycles per minute. Fig. 1 shows an example of load–displacement curves for creep analysis (Fig. 1a) and for fatigue analysis (Fig. 1b) resulting from the application of 15 N and 1 cycle per minute. As it is visible in Fig. 1a, the indenter continues to penetrate into the material during the plateau observed at the maximum load. The total length of this plateau reflects the creep sensibility of the material but its analysis deals with the study of the indenter displacement as a function of time. In Fig. 1b, the maximum indenter displacement after each cycle was decreasing, which thus exhibits the hardening phenomenon. For information, it can be noted that the total displacement reached after 100 cycles under the constant applied load of 15 N, close to 85 μm, is quite similar to the total displacement obtained after 1800 s of dwell-time under the same maximum applied load. The total displacement after creep and cyclic loading is quite similar around 20 μm.

3. Creep

Creep is a physical phenomenon associated with irreversible deformation of the material resulting from a constant loading applied during a sufficient time-period. For a purely visco-elastic material, the deflection which is observed during the creep-period is null after the complete unloading. Conversely, for a visco-plastic material the residual indentation depth is time-dependent. Usually, creep manifests in time with a part of elastic deformation, which results in the diffusion and movements of atoms, and another part of plastic deformation where dislocations motion is initiated inside the material. In this case when the material exhibits a “visco-elastoplastic” behavior, different stages of deformation can be observed, i.e., the instantaneous elasticity (the material undergoes the deformation when the stress is applied), the delayed elasticity and viscous flow (the stage where deformation increases linearly for a given time and then rapidly leads to material failure). During the dwell-time at the maximum load, the indenter usually continues its displacement for which the magnitude depends on the applied load, the holding time and the temperature [8]. Fig. 2 shows an example of the indenter displacement as a function of time and resulting from the application of 15 N. The indentation data are taken from the plateau data observed at the maximum applied load of 15 N as given in Fig. 1a.

The indentation creep data shown in this figure present an indentation curve similar to that of an ordinary creep curve. The first stage of the curve records a rapid increase of deformation with time, with a decreasing rate, followed by a steady-state region where indentation sizes increase linearly with time. To analyze the variation of the indenter displacement as a time-function rep-
resented in Fig. 2, two approaches can be generally applied, one consists in the use of descriptive laws of creep [9–11] and the other one is based on combining basic rheological elements [12–14]. As mentioned by Lucas and Oliver [15] and Goodall and Clyne [16], a variety of analytical relations for determining the exponent which is used to characterize the steady state have been developed, including constant-displacement (or load-relaxation) testing, constant-loading-rate testing, constant-strain-rate testing and constant-load testing. The relationship between the indentation strain rate and the yield stress at constant temperature is represented by a classical power law where the exponent is material dependent [17]. In addition by using the relation of Tabor [18] in which the yield stress is linked to the hardness by a factor 3, it is possible to directly connect the indentation strain rate to the hardness. By combining such relations and the relation between the hardness and the indenter displacement, Mahmudi and Rezaee-Bazzaz [9] suggest to represent the time dependence of the indentation displacement, \( h(t) \), by the following relation:

\[
h(t) = (K_1 \cdot (2m' + 1) \cdot F_{m'} \cdot t + K_2)^{1/(2m'+1)}
\]  

(1)

where \( F \) is the indentation load, \( K_1 \) and \( K_2 \) are constants material dependent and \( m' \) the exponent representative for the steady state. Note that this relation has been further simplified by the same authors as follows:

\[
h(t) = h_0 + K_0t^{1/(2m'+1)}
\]  

(2)

where \( h_0 \) is the displacement at the onset of creep and \( K_0 \) and \( m' \) are determined by fitting the experimental data. Note that \( m' \) must be different to \( m \).

As mentioned by Mahmudi et al. [19], the stress exponent values have been frequently used to identify the mechanisms controlling the deformation process. However, the authors recognize that such an identification of the diffusion process does not provide sufficient information to draw insightful comparisons on creep mechanisms of materials without additional information on activation energies. At best, creep stress exponents can only help to narrow the field among many theoretical possibilities [20]. In fact, deformation of polycrystalline materials at temperatures above 0.5 \( T_m \) can take place by different deformation mechanisms, associated with different stress exponent values. It is reported that if \( m' \) in Eq. (1) is close to 1, then the creep is associated to a phenomenon of diffusion [21] and when \( m' \) is close to 2, creep results in a sliding at the grain boundaries [22]. For values ranging between 4 and 6, Uzun et al. [23] attributed the creep to dislocation climb. Mahmudi et al. [19], noted that in their work, the relatively high exponent values ranging between 8 and 10 imply that the operative creep mechanism is due to a dislocation creep. This result was confirmed by Torre et al. [24] whom suggest for values higher than six that the mechanism of creep is due to the movement of dislocations. However, it is noticeable that for such high values of \( m' \), the different mechanisms of creep cannot be totally dissociated. However, it is admitted that, when \( m' = 12 \), the main creep mechanism is linked to the movement of dislocations. More recently, Choi et al. [10] suggested a formula for an empirical fitting of a displacement \( h \) and time \( t \) as follows:

\[
h(t) = h_0 + Bt^n + Ct
\]  

(3)

where \( n, B \) and \( C \) are fitting parameters.

We can note that Eq. (3) is in fact Eq. (2) where an additional term proportional to time has been added. Ma et al. [8] highlighted that time \( t \) in these relations must be corrected by taking into account time \( t_c \) when the creep process is really started at the indentation load, i.e., \( t = t_{creep} - t_c \), as it has been performed in Fig. 2. More recently, Jones and Grasley [11] proposed the description of the indentation displacement variation as a function of time as follows:

\[
h(t) = 1.21 \sqrt{F \cdot J(t) \cdot \cot \theta}
\]  

(4)

where \( \theta = 68^\circ \) is the half-angle of the referred conical indenter and \( J(t) \) the following complex time-function:

\[
J(t) = \frac{1}{E_0} + \frac{1}{E_1} \left( \frac{1 - e^{-\beta t_0}}{1 - e^{-t}} \right)
\]  

(5)

where \( E_0, E_1 \) and \( \beta \) are fitting parameters. The normalized time \( t_0 \) is chosen to match the length of the load dwell time.

Jones and Grasley [11] indicate that the presence of the term \((1 - e^{-t})\) in the denominator of the time-function \( J(t) \) serves to normalize the entire term from \( 1/E_0 \) at \( t = 0 \) to \( (1/E_0 + 1/E_1) \) at \( t = t_0 \), contrary to 0 and \( 1/E_1 \), respectively, as mentioned by the authors in their paper [11]. Compared to the others analytical models [9,10], the form of this creep compliance presents the advantage that the fit parameters possess physical meaning. In fact, when \( t_0 \) is the total length of the dwell-time of the test in question, the constitutive parameter \( E_0 \) controls the initial (instantaneous) elasto-plastic penetration into the material, \( E_1 \) controls the magnitude of the time-dependent penetration into the material over time \( t_0 \), and \( \beta \) controls the shape of the relaxation/retardation function. Nevertheless, it is critical to note that the two elastic moduli involved in Eq. (5) are not connected to the elastic modulus of the material. For the authors, this is probably due to the fact that the initial penetration almost certainly includes plastic deformation in addition to elastic deformation.

Besides, the creep behavior can be described by rheological models associating dashpots and springs in series, in parallel or combined. Indeed, based on the model of Maxwell–Voigt, Radok [25] and Lee and Radok [26] who studied the problem of viscoelasticity using a correspondence principle to describe the analogy between a mechanical system (the model) and visco-elastic behavior (to be modeled). In such a model, the elasticity constants from contact equations are replaced by time dependent operators [12], Fischer-Cripps [13] collected the three main types of assembly. These models were originally defined by Maxwell, Voigt and Maxwell–Voigt for which the main difference between them are the number of fitting parameters leading to the improvement of the data fit. However, Chicot and Mercier [14] noted that there is no similarity from the reduced elastic modulus associated to the springs and the actual elastic modulus of the material. For this reason, the authors proposed an alternative approach that overcomes the loading history of the material by deleting the spring in series in the model of Maxwell–Voigt. It is noticeable that all the equations expressing the indenter displacement as a function of the creep-time and resulting from rheological models present a general relationship for which the form is very similar to that of the model of Jones and Grasley [11]. That is why we focused this study on the application of the models of Mahmudi and Rezaee-Bazzaz [9], Choi et al. [10] and Jones and Grasley [11].

Thus different models have been applied to model the creep indentation data variation. The deviations between the modeling curve and the experimental points were minimized by the least-square method in order to find the best set of parameters. As a main conclusion, the model of Jones and Grasley [11] leads to fitting parameters independent on the indentation load as it was expected but, as already mentioned by the authors, the elastic moduli of 3600 GPa and 25 GPa for \( E_0 \) and \( E_1 \), respectively, greatly differ from the elastic modulus of the tested material equal to 52 GPa [27]. Note that \( \beta \) is a constant close to 0.5 and \( t_0 \), which has been determined by fitting, is equal to 1700 s close to the length of the load hold period, here 1800 s. Application of Eqs. (2) and (3) also allows adequately to represent the indentation variation of the indentation displacement variation as a function of time as follows:
data. As a main result, the two exponents, $m'$ and $n$, are equal to 1.5 and 0.3, respectively. Note that there is no possible correlation between $m'$ and $n$ due to the introduction of the term $B t$ in Eq. (3). Parameter $K_0$ is found load-dependent. Note that the simplified Eq. (2) is a rough approximation of Eq. (1). However by applying Eq. (1), we remarked that the experimental values of the term $K_1$, $(2m'+1)^{m'}$ is equal to a certain extent to the values of $K_2$. On the other hand, it is clear that the multiplication of fitting parameters improves the quality of the representation of the experimental data. However, since the future objective is to study the coupled creep and fatigue behaviors by indentation, we suggest limiting the number of fitting parameters. That is why we propose to represent the time dependence of the indenter displacement by the following simple relation which presents the advantage to have only one fitting parameter:

$$h(t) = K_3 \cdot (1 + t)^{1/2m' - 1} = h_0 \cdot (1 + t)^{1/2m' - 1}$$

where the constant $K_3$ must be equal to $h_0$ when $t = 0$. Consequently, Eq. (6) only involves one fitting parameter, i.e. the exponent $m'$ which must differ from $m'$ and $m$.

Fig. 3 shows that Eq. (6) is able to adequately represent the creep behavior of the SAC 405 alloy. Note that Eq. (6) is valid after an initial range of 23 s. In this figure, we notice that deviations of the ideal curve (black line) from the experimental data can be observed at longer times. In fact, the exponent $m'$ has not a constant value but it varies as a function of the indentation load, as it is visible in Fig. 4.

However, the values of the exponent are close enough to consider a mean value of 13.5 ± 1.2. Note that in the work of Mahmudi and Rezaee-Bazzaz [9], this exponent is also independent on the applied load. According to the value of the exponent $m'$, it seems to indicate that the main creep mechanism is the dislocation motion as mentioned by Torre et al. [24].

In the past, Mahmudi et al. [28] have studied creep behavior of the lead-free Sn–5%Sb solder alloy by applying long time Vickers indentation testing under a constant load of 15 N at elevated temperatures in the range of 298–405 K. Based on the steady-state power law creep relationship, the stress exponents were determined in the range of ~3 and ~14 depending on the wrought and cast conditions of the material, respectively. The authors suggested that the dominant creep mechanism in the wrought condition is grain boundary diffusion whereas it is a dislocation creep mechanism over the whole temperature range investigated for the cast condition. Geranmayeh and Mahmudi [29] found that this result is in good agreement with those reported for the same material in conventional creep testing at room temperature. In supplementary work, Geranmayeh and Mahmudi [20] added that a high value of the stress exponent implies that the operative creep mechanism is a dislocation creep, which in fact is independent of a grain size. In addition by analyzing Berkovich depth-sensing indentation tests on Sn–3.5Ag–0.75Cu solder alloy with different loading rates, Ma and Fusahito [30] found that the resulting indentation load-depth curves are rate dependent and present varying creep penetration depths during the same dwell time. The authors thus confirmed that the derived value of the stress exponent is consistent with the results obtained from conventional uniaxial tensile and compression experiments of a bulk solder alloy.

4. Fatigue

On the other hand by studying fatigue by indentation, Xu et al. [31] showed that there are some similarities in the behavior of the indentation fatigue depth propagation and conventional fatigue crack propagation. They concluded that the indenter displacement is greatly influenced by overloading and underloading during the fatigue indentation test. In addition, based on microstructural observations of indentation cross-sections obtained after indentation fatigue on polycrystalline copper, Xu et al. [5] showed that the damage mechanism of indentation fatigue results in the nucleation, formation and propagation of microcracks around a flat indentation [5]. According to the authors, this could be confirmed by a presented SEM cross-section of a copper sample which was subjected to a cyclic maximum loading decrease for a high-low loading sequence of two load blocks (Fig. 5) [5].

Xu et al. [5] advanced that the nucleation and accumulation of cavities are responsible to the development of cracks which is the main damage mechanism during fatigue by indentation. For that, Fig. 5a shows some cyclic lines for which the radii decrease with increasing distance from the indenter. Fig. 5b shows the local deformation near the indentation where many cavities are distributed along the cyclic lines and even where some cracks in higher density cyclic lines are located near the indenter edge. Fig. 5c gives the SEM images in the zone below the indentation corresponding to Fig. 5a. In this figure, there are more cavities and cracks close to the free end of indentation. Fig. 5d show the SEM images of pile-up around the sample corresponding to Fig. 5a. From these observations, the authors concluded that the main deformation mechanism under the indenter during the indentation fatigue is probably due to the grain boundaries of heavily deformed grains,
The indentation. (c) Higher magnification below the indentation. (d) Higher magnification SEM image of the pile-up [5].

The damage was expected to be the result from the grain boundaries of heavily deformed grains by the cyclic plastic deformation. As a main conclusion, they showed that the behavior of indentation depth propagation under cyclic fatigue loading conditions was similar to that of fatigue crack propagation, although there was no crack in indented materials in advance. In fact, the reason why fatigue loading sinks the indenter further into the material than static loading can be explained using the analogy of crack propagation. A crack will not propagate under a static insufficient stress intensity factor, but will propagate under fatigue loading for the same maximum stress intensity factor. Suppose that the stress concentration is not at a crack tip, but under an indenter. For static loading, plastic deformation at the stress concentration can reduce the effect by shielding. This stops further plastic deformation or propagation of the plastic zone. Such static loading is the usual hardness test. However if the loading involves periodic unloading or fatigue loading, instead of static loading, the unloading part of the cycle may allow the dislocations to rearrange themselves to reduce the internal stresses. This means that during the next loading, the renewed stress concentration can cause more dislocations to propagate the plastic zone again. Indeed, after a certain number of cyclic indentations, the emission and retraction of dislocations to propagate the plastic zone again. Indeed, after a certain number of cycles, the indenter displacement reached at the end of each indentation cycle, i.e., \( h_{\text{ind}} = (h_{i+1} - h_i)/h_0 \). In this work, one hundred fatigue indentation cycles have been performed with maximum loads located between 1 and 20 N with loading and unloading rates chosen between 2 and 40 N/min for obtaining a frequency between 1 and 15 cycles per minute. As a main result, the values of \( e_{\text{ind}}/C_0 \) and \( c \) are found to be independent both of the loading rate and of the indentation load. For the SAC 405 alloy, \( e_{\text{ind}}/C_0 \) is equal to 5.5% ± 0.3 and \( c \) to 1.11 ± 0.03. Fig. 6 shows the variation of the plastic indentation strain as a function of the number of cycles in bi-logarithmic coordinates.

As it is visible in Fig. 6, the relation of Manson–Coffin is able to adequately represent the fatigue indentation data obtained on the
SAC 405 alloy. Moreover, the fatigue ductility exponent of $-1.1$ found in this work differs from classical values ranging between $-1.5$ and $-1.6$ as indicated by Morrow [34]. However, Alush et al. [35] found a value of $-1.2$ when studying the mechanical fatigue behavior of commercially pure polycrystalline copper heat treated at 643 K for 1 h and in addition authors noted that the exponent can vary in a greater extent, i.e. between $-0.5$ and $-2$ for unalloyed steels. The authors attributed such a variation to the influence of precipitates which can significantly change the obstacles system during dislocation motion.

About the use of instrumented indentation to study fatigue behavior, it is important to note that only few works are devoted to the determination of a fatigue criterion by indentation. Basaran et al. [36] when studying experimental damage mechanics of microelectronic solder joints under fatigue loading used increment loading in nanoindentation for quantifying the fatigue damage by means of the elastic modulus variation. Consequently, they defined an elastic modulus degradation which is considered by the authors as a physically detectable quantity of material degradation. Another approach was employed by Ye et al. [37] who used instrumented indentation to study local mechanical properties of welded joints subjected to low-cycle fatigue loadings. The authors calculated: elastic modulus, yield stress and stress exponent from nanoindentation loading curve performed after a given number of fatigue cycles resulting from classical mechanical tests. As a main result, a strain amplitude-dependent variation in the local mechanical properties in the joint regions was obtained. An empirical criterion for judging the fracture locations of a weldment based on the yield stress mismatch ratio was also introduced to assess the integrity of the present welded joints subjected to low-cycle fatigue loadings. Another work by Xu et al. [5] presents the influence of overloading and underloading on the instrumented indentation results. It was shown that an increase in the maximum load can accelerate indentation depth propagation, while a decrease in the maximum load can retard indentation depth propagation. Further experiments showed that a sudden increase in maximum load after achieving a steady state followed by cycling at normal loading conditions can also delay indentation depth propagation, while a sudden drop in maximum load had a contrary effect. Those experimental phenomena implied that there were some similarities in the behavior of indentation fatigue depth propagation and conventional fatigue crack propagation. This result is encouraging to study creep and fatigue behavior of materials by means of instrumented indentation.

5. Conclusion

Due to the numerous possibilities of loading conditions, the instrumented indentation test seems to be appropriate in order to study the creep behavior (time-dependent indenter displacement under a constant applied load) and the fatigue behavior (cyclic indenter loading) of massive materials. For modeling these behaviors, different models can be used to characterize separately the mechanical properties under creep or fatigue tests. For creep analysis, the displacement is expressed as a function of time where different parameters can be used to represent the creep sensibility. The main parameter is time or stress-exponent which can be in a first approach connected to the creep mode, i.e. phenomenon of diffusion, sliding at grain boundaries, dislocation climb or movement of dislocations according to its value which is usually located between 1 and 15. For fatigue analysis, the indenter displacement can be expressed as a function of time, number of cycles or load. It was shown that the relation between a relative indenter displacements calculated after each loading cycle can be expressed as a function of the number of cycles following a relationship similarly to that of a Manson–Coffin law. As a main conclusion, we propose the use of two simple models: one for representing the creep behavior by the stress exponent $m = (3.5)$ and the other one for the fatigue behavior by using the Manson–Coffin law allowing the determination of the fatigue ductility coefficient $c_{\text{will}} = (5.5\%)$ and the fatigue ductility exponent $c = (1.11)$ for the SAC 405 alloy.

References