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A New Creep Model for SnAgCu Lead-Free Composite Solders: Incorporating Back Stress

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Abstract

The paper presents improved constitutive models for SnAgCu solder. In the present study, the constitutive behavior for creep performance of 95.8Sn-3.5Ag-0.7Cu lead-free solder was investigated. The secondary creep stage was focused on. It is shown that the stress exponent n can be well-defined into two stress regimes: low stress and high stress. A new constitutive model, which considered back stress, is proposed to describe the creep behavior of SnAgCu solder. In this model, back stress, being a function of applied shear stress in the low stress regime and particle size, volume fraction, coarsening of IMC particles in high stress regime, is introduced to construct the relationship between the creep strain rate and shear stress. The creep mechanism in these two stress regimes was studied in detail. In low stress regime, dislocations pass through the matrix by climbing over IMC particles. While in high stress one, dislocations are glide-controlled. According to the different creep mechanisms in both stress regimes, the back stress was calculated respectively and then incorporated into Arrhenius power-law creep model. It is demonstrated that the predicted strain rate-shear stress behavior employing the modified creep constitutive model considering back stress is consistent well with the experimental results.

1. Introduction

In recent years, due to the increasing environmental and health concerns over the toxicity of lead and strict legislation to ban the use of lead-based solders, many lead-free solders are developed. Among these solders, Sn-Ag-Cu alloys are one of the leading alternatives to replace the widely used Sn-Pb solders [1,2]. Microelectronic solder joints are typical exposed to the thermomechanical cycling conditions with high homologous temperatures [3], and creep is the mainly deformation mechanism of solder joints during service. Creep is the process of plastic deformation over time, under a constant load and high temperature [4]. Therefore, to increase solder joint reliability requires a fundamental understanding of the creep deformation behavior of solders.

All creep strain-time curves include three stages: primary, steady-state and tertiary creep. In general, attention was focused on the steady-state creep for it accounts for most of the creep rupture life of solders [5]. There are many creep models existing to describe the steady-state creep. Based on the simple Arrhenius power-law creep model, Shi et al. [5], Wen et al. [6], Xiao et al. [7] have developed models to predict the second stage creep behavior. However, these models are often found to break down at high stress regime. As a result, the hyperbolic creep model [8,9,10] is used to avoid the breaking down of the power law. However, the hyperbolic model is only an empirical equation and fails to

explain the physical mechanism during creep. Both Arrhenius and hyperbolic models ignore the microstructure change during the creep deformation. In fact, microstructure such as dislocation structures and crystallographic texture in materials strongly affect inelastic deformation [11]. This leads to the conclusion that a model of creep deformation behavior must consider not only the macroscopic deformation behavior but also the dislocation structures.

Furthermore, descriptions of elevated temperature creep are often based on the idea that internal stress or back stress (threshold stress) which oppose deformation process. In precipitation hardened or dispersion metals (such as SnAgCu solders), the second phases (Ag_3Sn and Cu_3Sn_6) are the obstacles against dislocation motion during creep. These obstacles are the main source of back stress. Hence, in order to better describe the creep mechanism, it is important to incorporate the back stress into the existing Arrhenius based creep model.

In this study, a large number of creep tests were carried out at various stress levels from 4-18Mpa, over a wide temperature range from 25 °C -75°C on miniature lap-shear Sn-Ag-Cu solder joints. These tests are to determine the creep behavior of the solder. Based on the experimental creep data, two stress regimes are identified and the back stress in the respective stress regimes is calculated and introduced into the modified creep model. The modified model was verified against the experimental results.

2. Experimental Procedures

In this study, 95.8Sn-3.5Ag-0.7Cu (in wt. %) with particle size of 25-45 μm was used as the solder alloy. Details of the processing of the solder sample were described elsewhere [12]. For creep tests, a micro-lap-shear solder joint sample was specially designed. The detail of fabrication of test samples, creep tests and microstructural characterization can be found in ref. [13].

3. Experimental results and discussion

3.1 Creep test

For creep testing, all the samples conducted displayed a typical creep curve which included three stages mentioned before. After termination of a short primary creep stage, the steady state stage can be observed in this solder under all test conditions. Strain increases rapidly with time in the tertiary stage.

The dependence of steady-state creep rate $\dot{\epsilon}$ on applied stress (τ) and temperature (T) can be characterized using a power-law equation of the form [14]

$$\dot{\epsilon} = A \left(\frac{\tau}{G} \right)^n \exp \left(-\frac{Q}{kT} \right) \quad (1)$$

where A is material constant, n is stress exponent and Q is creep activation energy; G is the temperature dependent intrinsic shear modulus and k is the molar gas constant. The term $(\frac{\tau}{G})^n$ describes the stress influence on creep strain

rate. The term $(\exp(-\frac{Q}{kT}))$ describes the temperature effect.

The shear modulus G was computed from the following equation [13]:

$$G = 16302 - 40.5(T - 273) \quad (2)$$

Figure 1 shows a logarithmic plot of the creep strain rate ($\dot{\epsilon}$) vs modulus compensated shear stress (τ/G). The slope of the fitted line is the stress exponent n. Figure 1 shows that the stress exponent of the solder joint under testing temperatures of 25°C, 50°C, 75°C can be well defined into two stress regimes: low stress and high stress.

Three trends can be found from Figure 1. First, stress exponent n is larger in high stress regime than in low stress regime at any given temperature; second, in both low and high stress regimes, stress exponent n decreases with increasing temperature; third, the transition point from low stress regime to high stress one decreases with increasing temperature too. The above observations can be explained as follows.

As we know, creep is determined by both the applied stress and temperature. As temperature increases, the dislocations acquire more energy to move and hence creep strain rate becomes less dependence on stress. While n stands for the stress sensitivity to the strain rate, the decreasing dependence implies that n decreases with the increasing temperature. In other words, creep strain rate dependence on temperature becomes more when temperature increases.

Fig. 1. Stress exponents of SnAgCu solder joints at 25°C, 50°C, 75°C

The activation energies (Q) of creep at two stress regimes were determined from the creep data using the equation as follows

$$Q = -k \left[\frac{\partial \ln \dot{\epsilon}}{\partial (1/T)} \right]_r \quad (3)$$

Activation energy of creep can be described as the energy barrier that is need for an atom to move from a high energy location to a lower energy location. For SnAgCu samples, the Q values were determined to be in the range from 21.8 to 28.9 kJ/mol. The Q values in our study are lower than the results presented by other investigators for Sn-Ag-Cu solder [1, 5, 15, 30]. However, direct comparison of the absolute Q values with the existing literatures is not meaningful since the Q values are affected by many factors, solder composition, applied stress, temperature, sample size, *et al* [13].

3.2 Microstructural Characterization

Figure 2 shows a typical SEM micrograph of the solder. The microstructure of SnAgCu alloys consists of proeutectic β matrix (which is nearly pure Sn) interspersed with various amounts of eutectic microconstituent comprising a dispersion of intermetallics Ag_3Sn and Cu_6Sn_5 . The size distribution and spacing of the intermetallic phase was studied using SEM micrograph. Figure 2 shows the fine and homogeneous dispersion of intermetallics in the matrix. Figure 3 shows the distribution of second phase size. The average particle size is about $0.564 \mu m$ and the volume fraction of intermetallic particles is about 9.5%. They are calculated using Agamennone *et.al*'s method [16].

Fig. 2. Typical SEM micrographs showing the microstructure of SnAgCu solder

4. New improved constitutive creep model

4.1 Incorporating back stress into power law model

As can be seen in Figure 2, there are second phase particles in the SnAgCu solder. These particles can act as obstacles for dislocation motion. The obstacles create a back stress that resists the motion of dislocations [15], and hence the back stress should be incorporated into equation (1). According to Henshall *et al*'s work [17], the third term in equation 1

$(\exp(-\frac{Q}{kT}))$ can be rewritten as D in equation (4) and (5) so

that the activation energy for deformation is constant at high temperature and decreases linearly starting at a transition

temperature T_t . The new creep model can thus be written as follows.

$$\dot{\varepsilon} = A \left(\frac{\tau - \sigma_b}{G} \right)^{n_0} D \quad (4)$$

$$D = \exp \left\{ - \frac{Q^*}{kT_t} \left[\ln \left(\frac{T_t}{T} \right) + 1 \right] \right\}; T < T_t$$

$$D = \exp \left(- \frac{Q^*}{kT} \right); T \geq T_t \quad (5)$$

where σ_b is the back stress, n_0 is the stress exponent after incorporating back stress, Q^* is the activation energy at high temperature

In our study, when temperature rises from 25°C to 50°C, the value of n reduced rapidly in both low and high stress regime. Hence, 50°C can be considered to be the transition temperature. It is consistent well with Kerr's work [15]. According to our results in 3.1 which can be compared with El-Bahay *et al's* activation energy [31], the activation energy at high temperature can be assumed as 30 kJ/mol.

Fig. 3. Particle distribution of intermetallics particles

4.2 Creep mechanism analysis and back stress calculation

From the creep shear strain rate vs shear stress curve, we have two-tiered behaviour comprised of a low-stress regime (LSR) and high stress regime (HSR). The observed change in the stress exponent between low stress regime and high stress regime can be linked to a change in the creep mechanism. Thus, the creation of the back stress in the two regimes should be different.

In SnAgCu solders, there are two phases in microstructure: soft matrix Sn and hard Intermetallic (IMC) particles. Hence, it is easy for dislocations to pass through matrix. During creep, when dislocations pass through matrix between particles, they will be pinned or arrested at IMC particles because of attractive character of interface between matrix and particles [18]. Only a few dislocations can pass through the matrix by climbing over particles to take part in the deformation. With τ/G increases, there will be more dislocations get stuck at IMC particles. When τ/G reaches threshold value, the dislocations between particle-matrix get enough energy to

detach itself from the interface by gliding over the particles. These dislocations can lead to deformation. The dislocation motion during creep process can be seen in Figure 4.

In other word, in the low stress regime, creep mechanism is dislocation climb [19,20]. The back stress is provided by the creation of new dislocation segments [15]. As result, the back stress has a linear relationship with the applied stress. And back stress in this stress regime is insensitive to the particle size and interparticle spacing [19,21,22]. We assumed that the solder was a well-annealed alloy, i.e. back stress $\sigma_b = 0$ at $t=0$, and the subsequent build-up of σ_b with creep strain will oppose the applied stress, as given by

$$\sigma_b = a\tau \quad (6)$$

$$\sigma_e = \tau - \sigma_b = (1-a)\tau \quad (7)$$

where a is a constant, σ_e is the effective stress.

Hence, determining the values of a will enable us to obtain the value of the back stress in the low stress regime.

Fig. 4. Dislocation motion during (a) low stress regime and (b) high stress regime

Equation (4) can be rewritten by combining equation (7)

$$\dot{\epsilon} = A \left[\frac{(1-a)\tau}{G} \right]^{n_0} D \quad (8)$$

From Figure 1, it is observed that in low stress regime, the range of n is from 4.8-3.2 without considering the back stress. Incorporating the back stress into the model, the stress effect on creep strain rate will be reduced. According to Equation (8), when n_0 is assumed as 3, the relationship between $\dot{\epsilon}^{1/n_0}$ and τ is linear, and the slope is $(AD)^{1/3} G^{-1} (1-a)$. We need to know the values of A and D to determine the values of a . A and D can be calculated in the high stress regime.

In the high stress regime, dislocation is glide-controlled [15]. The particle is much harder than the matrix, hence the dislocation cannot cut or shear the particles. Orowan bowing of dislocations is the most likely mechanism [20] as schematically shown in Figure 5 [23]. As the stress applied, the dislocation line is bowing around two particles. An additional stress is required to overcome the energy increase arising from elongation of the dislocation line. This stress is referred as Orowan stress, and back stress can be derived from Orowan stress. Hence, back stress has a close relationship with IMC particle size and interparticle spacing.

Fig. 5. Orowan bowing mechanism

In the high stress regime, the back stress from particles for precipitation hardened materials can be derived by using the approach of Orowan stress (σ_{Or}) [24]

$$\sigma_{Or} = \frac{0.84MGb}{\lambda - 2r} \quad (9)$$

where M is the Taylor factor, r is the mean radius of particles, λ is the interparticle spacing and b is Burger's vector.

The interparticle spacing λ was determined using the following relation [15], which assumes an ordered arrangement of particles.

$$\lambda = d \left(\frac{\pi}{6f} \right)^{1/2} \quad (10)$$

where d is the average intermetallic particle diameter and f is the volume fraction of Ag_3Sn and Cu_6Sn_5 particles.

The back stress is expressed as [15, 18, 22, 24, 25]

$$\sigma_b = \sigma_{Or} \sqrt{1 - k_r^2} \quad (11)$$

where k_r is the intensity of the attraction between the particles and dislocations, usually between 0-1.

In the above parameters, r and f can be determined from the typical SEM micrographs in this study, $G = 16302 - 40.5(T - 273)$ [13], $d = 2r$, $k_r = 0.87$ [24], $M = 3$ [24, 26], $b = 0.317 \text{ nm}$ [15].

The creep deformation can be considered as a kind of aging process. In SnAg-based solders (eg., Sn-Ag and Sn-Ag-Cu), the presence of Ag_3Sn and Cu_6Sn_5 particles in the microstructure can undergo significant in situ strain-enhanced coarsening during aging, resulting in continuous evolution of the joint creep behaviour during service [22]. The Lifshitz-Slyozov-Wagner (LSW) model provides a first-order prediction of the normal homogeneous coarsening [27, 28], and it has been employed by many investigators [4, 20, 24, 28, 29] for the description of the microstructural coarsening during creep. The LSW model is given as

$$r^3 - r_0^3 = At \quad (12)$$

where A is a constant; t is the time; r_0 is the original particle radius.

During isothermal aging creep, the above coarsening model was modified by Dutta *et al* [22] as follows

$$r^3 - r_0^3 = K \frac{C_{eq}}{T} D_{sol} t \quad (13)$$

where K is constant; C_{eq} (in mole fraction) and D_{sol} (in m^2/s) are equilibrium solubility and diffusivity of the solute (Ag) in Sn. $K_{SnAgCu} = 5.23 \times 10^{-5} \text{ mK}$, $C_{eq} \approx 2.94 \times 10^{-1} e^{-26576.5/RT}$

and $D_{sol} = 1.8 \times 10^{-5} e^{-77000/RT}$ [22]. Here only the effect of Ag is considered as the solubility limits of Cu in Sn ($\sim 0.03\%$ wt.%) are much lower than that of Ag in Sn ($\sim 0.1 \text{ wt.}\%$) [32].

Using the above equations, the back stress in both low stress regime and high stress regime can be calculated. The parameters in the modified model can be summarized in Table 1.

Table 1. Summary of Global Parameters in the modified model

A /s	G MPa	Q^* kJ/mol
5.61E+11	16302-40.5(T-273)	30

In low stress regime, a values at different testing temperatures are determined and present in Table 2. It was observed that as the testing temperature arises, the values of a increase correspondingly. This can be attributed to an increased amount of dislocations climbing over the intermetallic particles and passing through the matrix, as the temperature increases. The increase in newly created

dislocation segments will in turn lead to an increase in the back stress.

Table 2. Results of b value at different temperatures

Temperature K	a
298	0.57
323	0.74
348	0.82

In high stress regime, the back stress at the beginning of creep is calculated in Table 3. A trend of decreasing back stress with increasing testing temperature was observed. This observation can be attributed to the creep mechanism being dislocation glide-controlled in high stress regime. As temperature increases, the temperature effect on creep becomes more prominent. It will result in more energy for dislocation motion. Thus, additional stress required to overcome the energy increase arising from elongation of the dislocation line may be reduced. This additional stress is proportional to back stress.

Table 3. Back stress at different temperatures in high stress regime

Temperature K	σ_b MPa
298	7.95
323	7.42
348	6.90

4.3 Verification the modified creep model with experimental results

In order to verify the modified model, a comparison between the modified model and the experimental data is made and presented in Figure 6. The results of the model agree very well with the experimental data, indicating the feasibility of the model to predict creep behavior.

(a)

(b)

(c)

Fig. 6. Measured shear-stress versus strain rate behaviour compared to modified creep model predictions (a) 25°C, (b) 50°C, (c) 75°C

5. Conclusions

(1) The mechanism of creep during low stress regime and high stress regime was analyzed. In low stress regime, dislocations pass through the matrix by climbing over the second phase particles. The back stress was proportional to applied shear stress. However, in high stress regime, dislocations are glide-controlled. Orowan bowing of dislocations is the most likely mechanism. The back stress has a close relationship with precipitate particle size and volume fraction. The coarsening of intermetallic particles was also taken into account.

(2) The back stress in both stress regimes was calculated and incorporated into the simple Arrhenius power-law creep model. All data in all stress regimes can be rationalized by a power-law with a stress exponent of 3. According to our results in 3.1, the activation energy at high temperature will be taken as 30kJ/mol.

(3) The prediction results employing the modified creep model agree very well with the experimental data. Hence, a creep strain rate incorporating a back stress is capable of describing the stress and temperature dependence of the stationary or minimum creep strain rate.

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