Evaluation of Fatigue Crack Propagation of Sn–5.0Sb/Cu Joint Using Inelastic Strain Energy Density

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In this study, evaluation of fatigue crack properties in the Sn–5.0Sb/Cu joint was performed by using the inelastic strain energy density W_{in} around the crack tip which was calculated by FEM. W_{in-c} given by the multiplication of W_{in} obtained from an arbitrary-sized square region surrounding the crack tip by the side length of the square region – did not depend on the size of the square region and the element size. The fatigue crack propagated in the solder layer and the power law of Paris law type between its fatigue crack propagation rate and ΔW_{in-c} held. However, the power exponent in the fatigue crack propagation law differed depending on the regions of ΔW_{in-c} . The power exponent became about 1 in the low ΔW_{in-c} region and very large in the high ΔW_{in-c} region. In the low ΔW_{in-c} region, fatigue fracture propagated along the high angle grain boundaries formed ahead of the crack by the continuous dynamic recrystallization. On the other hand, the fracture transformed to static fracture mode in the high ΔW_{in-c} region and the cleavage fracture was observed. The large power exponent in the high ΔW_{in-c} region was attributed to the cleavage fracture. [doi:10.2320/matertrans.MH201813]

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1. Introduction

Power semiconductor modules using wide-band gap semiconductors, such as SiC are expected to operate at higher temperatures than in conventional conditions. Along with an increase in the operating temperature, high thermostability of materials that compose a power semiconductor module is now an important issue. Especially, the realization of high thermostability of die attach materials used for joining semiconductors, insulated substrates, and heat dissipating structure members is a matter of urgent attention. In the context, Sn–Sb based alloys are now drawing attention as a candidate for die-attach material. Die-attach materials are subjected to cyclic thermal stress ascribable to the differences in thermal expansion coefficients of members of the power module, which induces a fatigue crack to propagate, reducing heat dissipation characteristics and failure of the module is caused in the end. Therefore, evaluation of fatigue crack propagation rate in die attach materials is critical to guarantee reliability of power modules.

Since die-attach materials are used in the environment beyond small scale yielding, ideal for the evaluation of fatigue crack propagation rate is to use the cyclic J -integral ΔJ .¹⁾ Although fatigue crack propagation in solder alloys and Ag nanoparticles for the die-attach have been evaluated by ΔJ ,^{2–4)} the method of calculating ΔJ by the finite element method (FEM) is accompanied by difficulties and there is a problem to consider creep deformation into ΔJ . With these taken into consideration, evaluation by a scalar parameter as an alternate is desired. As a method without using a fracture mechanics parameter, there have been efforts to calculate the inelastic strain energy density range ΔW_{in} around the crack tip by the FEM and the result has been used as a parameter to evaluate the fatigue crack propagation.^{5,6)} As it is easy to

calculate ΔW_{in} by FEM, ΔW_{in} can be expected as an alternate to ΔJ . However, since unlike the path-independent J -integral, ΔW_{in} is dependent on the element size of FEM model and the distance from the crack tip, the dependence of ΔW_{in} needs to be considered. Because of the foregoing, the author *et al.* have investigated a method to reduce the dependence of ΔW_{in} on the element size and the distance from the crack tip.⁷⁾

In this study, fatigue crack propagation characteristic of Sn–5.0 mass% Sb that is candidate materials for die attach of power modules was evaluated by the method using inelastic strain energy density.

2. Experimental Procedure

2.1 Specimen

Sn–5.0 mass% Sb (hereafter mass% is omitted) was used in this study. A flat plate scarf joint with 300 μm of solder layer width as shown in Fig. 1 was employed for specimen of fatigue crack propagation testing. The joining angle was 45 degrees from the stress axis to produce the mixed mode loading of mode I and mode II. A notch was created on one side at the end of the joint. The notch with the length of 500 μm was introduced by a precision wire saw (K. D. Unipress). The joining temperature was +50 K above the melting temperature and soldering was done in the atmosphere with the use of RA flux. Then, the specimen surface was polished with SiC grinding papers and diamond particles. Finally, the surface was mirror-finished by polishing with colloidal silica suspension. Figure 2 shows the initial microstructure of the joint. The joint consists of few

Fig. 1 Specimen geometry for fatigue crack propagation test.

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Fig. 2 Low magnification optical micrograph around the notch tip (left side) and high magnification optical micrograph in front of the notch tip (right side) of Sn–5.0Sb/Cu joint.

Fig. 3 Fatigue crack propagation testing machine.

crystalline grains of β -Sn solid solution with Cu_6Sn_5 dispersoids. Although it is considered that Sb forms a solid solution in β -Sn and Sb exceeding the solubility limit forms SnSb intermetallic compounds, the compounds were not observed under the optical microscope.

2.2 Fatigue crack propagation test

Fatigue crack propagation tests were performed in pulsating displacement-controlled tensile-tensile mode. The control waveform was symmetric triangle wave, the test temperatures were 348 K and 398 K, and the total displacement range was $5\mu\text{m}$ – $11\mu\text{m}$, and the displacement rate was $6\mu\text{m s}^{-1}$. Figure 3 shows the testing system used in this study. The test machine shown in Fig. 3 is a micro-load fatigue test machine (Saginomiya Seisakusho: LMH207-20) equipped with a piezo actuator which was controlled by the displacements measured by the capacitance sensor fixed below the specimen. The test temperatures were controlled by ceramic heater installed inside the specimen fixing jigs. Fatigue crack lengths were measured from video images recorded by a microscope (Leica: Z16 APO) installed on the upper part of the test machine.

2.3 Evaluation of fatigue crack propagation rate

In fracture mechanics, the cyclic J -integral ΔJ shown in eq. (1) is used as the driving force for fatigue crack propagation in the condition beyond small scale yielding.¹⁾

$$\Delta J = \int_{\Gamma} \left(\Psi(\Delta \varepsilon_{ij}) dy - \Delta T_i \frac{\partial \Delta u_i}{\partial x} ds \right) \quad (1)$$

where Γ is the integral path encircling a crack tip in a counter-clockwise direction, Ψ is the quantity relating to strain energy density, ΔT_i is the vector of surface force on the path, Δu_i is the displacement vector along Γ , and ds is the increment in length along Γ . In regard to solder alloys and sintered Ag nanoparticles, the studies on fatigue crack propagation by using ΔJ have been published.^{2–4)} However, the method of calculating ΔJ for other than standardized fracture mechanics test specimens has not been established yet. Furthermore, C^* integral is necessary for the time-dependent crack propagation analysis in fracture mechanics.^{8–10)} Therefore, the use of fracture mechanics parameters involves difficulty in the analysis of fatigue crack propagation in real power semiconductor modules. Given the analysis of thermal fatigue crack propagation in power semiconductor module structures, a scalar parameter that is an alternative to ΔJ is necessary. For an alternate scalar parameter, there are attempts seen in which the inelastic strain energy density range ΔW_{in} around the fatigue crack tip is calculated by FEM to evaluate the fatigue crack propagation rate.^{5–7)} Although ΔW_{in} depends on the element size in the FEM model and the distance from the crack tip, it can take not only plastic deformation but also creep deformation into consideration. In this study, ΔW_{in} was employed to evaluate fatigue crack propagation. The FEM model shown in Fig. 4 was used for the calculation of ΔW_{in} and the FEM analysis was performed for each predetermined length of the fatigue crack. The element type was a quadrilateral 8 nodes plane strain element. A solver of ANSYS 17.2 was used. As for material properties, Cu was regarded as an elastic body and Sn–5.0Sb as an elasto-plastic body. For the plastic hardening law, the non-linear kinematic hardening law (Chaboche model) shown in eq. (2) was used.

$$\dot{\chi}_i = \frac{2}{3} \sum_{i=1}^n A_1 \dot{\varepsilon}_p - A_2 \chi_i \dot{p} \quad (2)$$

where $\dot{\chi}_i$ is the back stress rate, $\dot{\varepsilon}_p$ is the plastic strain rate, χ_i is the back stress, $\dot{p}$ is the cumulative plastic strain rate, and A_1 and A_2 are the material constants as determined by inverse analysis based on the experimentally obtained load–displacement hysteresis loop. One example of material constants for the Chaboche model is presented in Table 1. A square area of $60\mu\text{m} \times 60\mu\text{m}$ around a crack tip was divided into squares with sides of $10\mu\text{m}$ and ΔW_{in} was calculated as an average

Fig. 4 FEM model for calculation of ΔW_{in} .

Table 1 One example of material constants for the Chaboche model.

Test temperature / K	σ_y / MPa	A_1	A_2
348	14	24000	1400
398	10	16000	1000

of element solutions of this area. The equation for calculating ΔW_{in} is shown in eq. (3).

$$\Delta W_{in} = \frac{\sum \Delta W_{in}^{element} \times V^{element}}{\sum V^{element}} \quad (3)$$

where $\Delta W_{in}^{element}$ is the element solution of the amount of change in inelastic strain energy density per cycle, and $V^{element}$ is the element volume.^{5,7)}

2.4 Microstructural observation

Microstructures were observed with an optical microscope (Carl Zeiss: Axio Imager A1m). Furthermore, crystallographic orientation analysis was performed, using the technique of Electron Back Scatter Diffraction (EBSD, TSL: OIM). Specimens were prepared by mechanical polishing using SiC polishing paper and diamond paste, followed by further polishing using a colloidal silica suspension to remove the layers that were damaged in the mechanical polishing process.

3. Results and Discussion

3.1 Effect of element size and distance from crack tip on inelastic strain energy density

Since ΔW_{in} is dependent on the element size in the FEM model and the distance from the crack tip, these types of dependence need to be clarified. The HRR singularity field is formed surrounding the crack tip and the stress and the strain in the singularity field can be described by eq. (4) and eq. (5).^{11,12)}

$$\sigma_{ij} = k_1 \left(\frac{J}{r} \right)^{\frac{1}{n+1}} \cdot \tilde{\sigma}_{ij}(\theta) \quad (4)$$

$$\varepsilon_{ij} = k_2 \left(\frac{J}{r} \right)^{\frac{n}{n+1}} \cdot \tilde{\varepsilon}_{ij}(\theta) \quad (5)$$

where J is the J -integral, r is the distance from a crack tip, n is the strain hardening exponent, and others are constants. In the HRR singularity field, stress and strain are proportional to $(1/r)^{1/(n+1)}$ and $(1/r)^{n/(n+1)}$, respectively. Since the strain energy density is the integral of stress with strain, the strain energy density W in the HRR singularity field is inversely proportional to r irrespective of the strain hardening exponent of the material. When a sufficiently large plastic deformation is loaded, like in this study, strain energy density W becomes mostly equivalent to inelastic strain energy density W_{in} , on the basis of which it can be predicted that an average of W_{in} in the square area surrounding the crack tip is inversely proportional to the length of a side of the square area L_{area} . In order to confirm the prediction, the dependence of W_{in} on L_{area} was investigated by creating a FEM model of the flat

Fig. 5 Illustration showing HRR singularity and W_{in} acquisition area around crack tip in FEM analysis.

Fig. 6 Relationship between W_{in} and L_{area} for each $L_{element}$ (elasto-plasticity analysis).

plate specimen with a single crack to be subjected to mode I loading. The dependence of W_{in} on the element size $L_{element}$ was also investigated. Figure 5 is an illustration of element meshing surrounding the crack tip and the acquisition range of W_{in} . In accordance with eq. (3), W_{in} is determined as an average of element solutions in the square area. For the plastic constitutive equation, the Chaboche model was used and the constant for Sn-5.0Sb obtained from the tensile testing at the strain rate of 10^{-5} s^{-1} at 398 K was used ($\sigma_y = 4 \text{ MPa}$, $A_1 = 8000$, $A_2 = 1700$).

Figure 6 shows the relationship between W_{in} and L_{area} for each $L_{element}$. The gradient of a straight line became -1 . As predicted above, W_{in} was the power of -1 to L_{area} , i.e. becomes inversely proportional. As for this inversely proportional relationship, similar results had been obtained in sintered Ag nanoparticles^{7,13)} and the inverse relationship can be established regardless of materials. Furthermore, W_{in} calculated as an average in the square area has negligible effects of element size in the square area. Since W_{in} and L_{area} are inversely proportional to each other, the proportional constant of a straight line is determined from W_{in} calculated in an arbitrary size of square area as in eq. (6).

$$W_{in-c} = W_{in} \cdot L_{area} \quad (6)$$

where W_{in-c} is the proportional constant. As this W_{in-c} is not dependent on L_{area} , the use of W_{in-c} allows the evaluation of fatigue crack propagation that is not dependent on the element size and the size of the square area. However, it

should be considered that there is a case in which creep deformation becomes dominant in the real operating condition of power semiconductor modulus. The stress and strain surrounding the crack tip under creep dominance is described by C^* -integral in eq. (7) instead of J -integral.⁸⁻¹⁰⁾

$$C^* = \int_{\Gamma} \left(W^* dy - \sigma_{ij} n_j \frac{\partial u_i}{\partial x} ds \right)$$

$$W^* = \int_0^{\dot{\epsilon}_{kl}} \sigma_{ij} d\dot{\epsilon}_{ij} \quad (7)$$

where W^* is the stress work ratio, $\dot{u}_i$ is the displacement rate vector, and $\dot{\epsilon}_{kl}$ is the creep strain rate. It is known that even under creep deformation, stress and strain in the steady state creep area sufficiently close to a crack tip exhibit HRR singularities shown in eq. (8) and eq. (9).¹⁴⁾

$$\sigma_{ij} = \left(\frac{C^*}{A I_n r} \right)^{\frac{1}{n+1}} \cdot \tilde{\sigma}_{ij}(\theta) \quad (8)$$

$$\epsilon_{ij} = \left(\frac{C^*}{A I_n r} \right)^{\frac{n}{n+1}} \cdot \tilde{\epsilon}_{ij}(\theta) \quad (9)$$

where n is the creep stress exponent, and others are constants. Although the exponential terms in eq. (8) and eq. (9) become the functions of stress exponents unlike in elasto-plasticity, W_{in} is expected to become the power of -1 to r even under creep deformation. Therefore, the dependence of L_{area} and $L_{element}$ on W_{in} under creep deformation was also investigated. Solder was regarded as an elasto-creep body and for the creep constitutive equation, the Norton's law was used as shown in eq. (10).

$$\dot{\epsilon}_{ss} = A_3 \sigma^n \quad (10)$$

where $\dot{\epsilon}_{ss}$ is steady-state strain rate, A_3 is the material constant, and n is the stress exponent. The material constant of Sn-5.0Sb was obtained from the creep testing at 398 K ($A_3 = 2.69 \times 10^{-12}$, $n = 6.56$, the unit of σ and ϵ_{ss} are MPa and s^{-1}).

The relationship between W_{in} and L_{area} is shown in Fig. 7. As in elasto-plastic analysis, W_{in} has almost no effects of element size and becomes inversely proportional to L_{area} . Therefore, even in the condition in which creep deformation

becomes dominant, fatigue crack propagation can be evaluated by using W_{in-c} calculated by eq. (6).

In this study, fatigue crack propagation rate of Sn-5.0Sb was evaluated by Paris type fatigue crack propagation law using ΔW_{in-c} shown in eq. (11).

$$\frac{da}{dN} = C_1 \Delta W_{in-c}^{C_2} \quad (11)$$

where C_1 and C_2 are the material constants. When sufficiently large plastic deformation is loaded like in the study, $W_{in} \simeq W$. However, the total strain energy density W needs to be used to obtain the inverse relationship as describe above in small load conditions.

3.2 Evaluation for fatigue crack propagation rate of Sn-5.0Sb

3.2.1 Fatigue crack propagation rate

The relationship between fatigue crack length and the number of cycles is shown in Fig. 8. It is clear that with an increase in the total displacement range, W_{in} around the crack tip becomes large, causing the fatigue crack propagation rate to increase. The fatigue crack propagation rate da/dN was calculated from the fatigue crack propagation curve and the results organized by using ΔW_{in-c} are shown in Fig. 9. There is variability among the test results, but a power law that should be shown in eq. (11) holds between the fatigue crack propagation rate and ΔW_{in-c} . The power exponent of the fatigue crack propagation law differs depending on the levels of ΔW_{in-c} and beyond the point exceeding $\Delta W_{in-c} = 100 \text{ N m}^{-1}$. The power exponent became about 1 in the low ΔW_{in-c} region and more than 3 in the high ΔW_{in-c} region. It has been reported that when fatigue crack propagation rate of solder alloys is evaluated by ΔJ , the power exponent of the fatigue crack propagation law becomes about 1.²⁾ In addition, according to a research report on heat resistance alloys, the power exponent of the fatigue crack propagation law in the time-dependent type fatigue becomes 1.¹⁵⁾ From these research reports, the exponent of fatigue crack propagation law in the time-dependent type fatigue becomes 1 regardless of the kind of material. As the test temperature of this study is more than 60% of the melting temperature of Sn-5.0Sb,

Fig. 7 Relationship between W_{in} and L_{area} in for each $L_{element}$ (elasto-creep analysis).

Fig. 8 Relationship between fatigue crack length and number of cycles for each testing condition.

Fig. 9 Fatigue crack propagation rate as a function of ΔW_{in-c} .

the time-dependent type fatigue is dominant for the test conditions in this study. It is thought to be that the power exponent in the low ΔW_{in-c} region showed the value of the time-dependent type fatigue.

On the other hand, a large power exponent was obtained in the high ΔW_{in-c} region and its value is very large as a ductile solder alloy. In the high ΔW_{in-c} region, failure mode transcends the region of continuous crack propagation caused by fatigue and seems to become crack propagation caused by static fracture mode. And there are no effects of temperatures on the fatigue crack propagation rate.

3.2.2 Mechanism on fatigue crack propagation

Shown in Fig. 10 are bright-field images of fatigue crack propagation paths in the tests in the low and high regions of ΔW_{in-c} . The fatigue crack propagated in the solder bulk in any ΔW_{in-c} region. In the low ΔW_{in-c} region, the fatigue crack deflected microscopically. In the high ΔW_{in-c} region, the crack propagated linearly. The fracture mechanism varies according to the regions of ΔW_{in-c} , which relates to the change in the exponent in the fatigue crack propagation law.

Fig. 10 Optical micrographs showing crack propagation path in low ΔW_{in-c} and high ΔW_{in-c} region.

Fig. 11 Polarized light images, orientation mapping and grain boundary structure of around fatigue crack in low ΔW_{in-c} region.

Figure 11 shows polarized light image around a fatigue crack in the low ΔW_{in-c} region, and the crystalline orientation mapping and the grain boundary structure image by EBSD. In the grain boundary structure image, the red lines are subgrain boundaries with the misorientation less than 15 degrees and the blue lines are high angle grain boundaries with the misorientation more than 15 degrees. In the low ΔW_{in-c} region, fine grains exist around the fatigue crack, and most of them were high angle grain boundaries. It has been reported that β -Sn, which is the matrix of Sn alloy has microstructural changes called “continuous dynamic recrystallization” during high temperature cyclic deformation.^{16–18)} The continuous dynamic recrystallization is a phenomenon in which subgrains are formed in arranging dislocations through dynamic recovery and the continuous deformation increases the dislocation density, which makes the subgrains continuously rotate causing recrystallization to progress.¹⁹⁾ In the condition of the low ΔW_{in-c} region in this study, the continuous dynamic recrystallization occurred. Figure 12 shows the relationship between fatigue crack length and the number of cycles in the test of low ΔW_{in-c} region. A staircase pattern was observed as shown in Fig. 12 and it is confirmed from the video records that in the phenomenon, crack propagation and non-propagation were repeated. The result shows that during the period of forming subgrains around the crack tip, propagation of the crack ceases and when a high angle grain boundary is formed by the rotating grains, the fatigue crack propagates along the high angle grain boundaries. In other words, it can be considered that the continuous dynamic recrystallization and subsequent grain boundary fracture in the low ΔW_{in-c} region are the fatigue crack propagation mechanism.

Figure 13 shows polarized light image around the fatigue crack in the high ΔW_{in-c} region, crystalline orientation mapping and grain boundary structure. In the images, the continuous dynamic recrystallization observed in the low

Fig. 12 Relationship between fatigue crack length and number of cycles in low ΔW_{in-c} condition.

Fig. 13 Polarized light images, orientation mapping and grain boundary structure around fatigue crack in high ΔW_{in-c} region.

ΔW_{in-c} region was not seen in the high ΔW_{in-c} region, but the cleavage fracture in grain interior was observed. Under a large strain energy condition, the fatigue crack propagation does not seem to be the continuum crack propagation mechanism by fatigue, but seems to become crack propagation by the static fracture mode mechanism. In this study, from $\Delta W_{in-c} = 100 \text{ N m}^{-1}$ onwards, the fracture mechanism seems to transit to the static fracture mode mechanism. The crack propagated in a brittle manner which was prompted by the cleavage fracture before the occurrence of the continuous dynamic recrystallization. The cleavage fracture contributes to increase the power exponent beyond the point exceeding $\Delta W_{in-c} = 100 \text{ N m}^{-1}$.

4. Conclusions

- (1) It was found that W_{in-c} obtained by the multiplication of the average of element solutions of the inelastic strain energy density in an arbitrary-sized square region

surrounding the fatigue crack tip by the side length of the square region is not dependent on the size of the square region and the element size. This can be explained by the theory of HRR singularity and can also be true either in elasto-plastic analysis or in elastic creep analysis.

- (2) In the conditions of this study, the fatigue crack of the Sn-5.0Sb/Cu joint propagated in the solder layer and the power law of Paris law type holds between the fatigue crack propagation rate and ΔW_{in-c} .
- (3) The power exponent in the fatigue crack propagation law became about 1 in the low ΔW_{in-c} region and it became very large in the region over $\Delta W_{in-c} = 100 \text{ N m}^{-1}$.
- (4) In the low ΔW_{in-c} region, the fatigue crack propagated along high angle grain boundaries formed by the continuous dynamic recrystallization. On the other hand, in the high ΔW_{in-c} region, the cleavage fracture occurred due to the static fracture mode.
- (5) In the high ΔW_{in-c} region, the power exponent in the fatigue crack propagation law showed the larger value than that in the low ΔW_{in-c} region because of the occurrence of the cleavage fracture caused by the static fracture.

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