

COOLING RATE DEPENDENT UNDERFILL FRACTURE TOUGHNESS

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Abstract

The failure of organic packages during thermal cycling is often associated with failure of the underfill by fracture. The fracture toughness of underfills measured by applying a mechanical load to the material at a constant temperature is used as a measure of the propensity of underfill fracture. However, this fracture toughness does not take into account temperature effects including cooling rates. To include temperature effects a fracture toughness based on thermally applied stresses is defined and a method to measure this thermally induced fracture toughness is described. Results on two commercial underfills are presented. Comparison of the conventional, mechanically induced fracture toughness and the new, thermally induced fracture toughness shows that underfill fracture toughness including thermal effects is significantly smaller than the conventional values. This indicates that the conventional method overestimates the stress required to introduce fracture in these materials.

Key words: underfill, thermal fracture toughness, viscoelastic stress

Introduction

The lifetime of organic packages during thermal cycling is determined by the level of stress that is experienced by the underfill particularly in the areas of greatest distance from neutral point (DNP) and by the level of stress that leads to the development of a crack in the underfill. These stresses are thermal stresses originating from the differences in the coefficients of thermal expansion (CTE) among the different materials in a package. To avoid failure, durability properties such as strength and fracture toughness of the underfill must be able to resist the applied thermal load.

To identify materials which will survive repeated thermal stress loading, fracture toughness measurements are conventionally carried out by applying a mechanical load at a fixed temperature (typically room temperature). This fracture toughness which we will call the

mechanically induced fracture toughness is used as one of the most critical criteria to select an optimum underfill material. However, package materials actually experience fracture that is induced by thermally induced stress rather than by mechanically induced stress. If there is a difference in the two fracture values, thermal package life time predictions will be improved by using the thermally induced fracture toughness.

In this paper we review a method to determine the thermally induced fracture toughness by inducing the growth of a crack during application of a thermal stress [1]. The thermally induced fracture toughness measured thus for two materials is compared to their counterpart mechanically induced fracture toughness.

Theoretical Analysis

Single edge notched tension (SENT) tests [2] have been popular, since it is easy to obtain the critical stress intensity factor (SIF) of a SENT specimen by measuring the critical load. However, in the thermally induced fracture test used here both ends of the test specimen are clamped hindering free rotation, and Tada's SIF equation [2] is not valid. To account for the effects of clamped ends the required weight functions for the SENT specimens need to be derived using FEM calculations [3, 4]. In a similar fashion, Park and Feger [1] calculated the thermal SIF for thermal SENT specimen geometries using a commercial FEM software.

Tang et al. [5] characterized the thermal fatigue behavior of an epoxy using the elastic modulus of the epoxy to calculate the strain energy release rate over a range of approximately 100°C. However, it is more accurate to use the relaxation modulus of the epoxy which may vary by an order of magnitude from the elastic modulus especially at temperatures close to the glass transition temperature range (T_g). The calculation of accurate thermal stress values reached during cooling requires accurate specimen moduli. In this study, the master curve of the underfill relaxation modulus vs. temperature was determined for two underfill materials and the thermal shift factors were taken into account for the transient temperature effects in order to calculate the cooling rate-dependent viscoelastic thermal stress in the underfill.

The stress intensity factor for the single edge notched test specimen in Figure 1 is expressed as

$$K_I = \sigma \sqrt{\pi a} F(a/W), \quad (1)$$

where $F(a/W)$ is the weight function, σ is the applied stress, and a is the crack length [2].

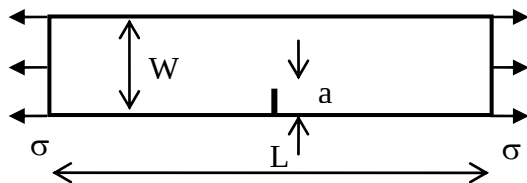


Figure 1. Single edge notched tension specimen

The weight function for the thermal SENT in Figure 2 may be defined similarly as

$$F_T \left(\frac{a}{W} \right) = - \frac{K_{TI}}{\sqrt{\pi a \Delta T \Delta \alpha E}}, \quad (2)$$

where K_{TI} is the mode-I thermal SIF, ΔT is the temperature change, $\Delta \alpha$ is the CTE mismatch

between a sample and a rigid fixture, and E is the modulus of the specimen. Although a uniform stress σ is applied on the specimen in Eqn. (1), non-uniform stresses arise when both ends of a SENT specimen are fixed requiring derivation of a weight function for the thermal SENT specimens used here. The weight function for SENT specimens with clamped ends is more complex because rotation of the ends is constrained.

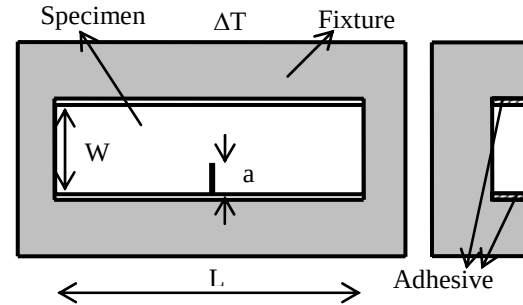


Figure 2. Thermal SENT specimen

John et al. [3] used finite element methods to derive an analytical weight function for fixed ratios of sample length to width. With this in mind the weight function for CTE mismatched boundary conditions was determined by FEM calculation for a length-to-width ratio of 6.64 used for the thermal SENT specimens of this study. Quarter point elements were used to create a singularity at the crack tip. The weight function in Eqn. (2) was derived and fitted using a cubic polynomial for simplicity:

$$F_T \left(\frac{a}{W} \right) = C_3 \left(\frac{a}{W} \right)^3 + C_2 \left(\frac{a}{W} \right)^2 + C_1 \left(\frac{a}{W} \right) + C_0 \quad (3)$$

where C_0 , C_1 , C_2 and C_3 were determined as -3.999, 2.882, 0.4039 and 1.121, respectively. The weight functions for the two cases were compared and their convergence for $a/W \ll 1$ was shown [1].

The thermal stress intensity factor can then be expressed as

$$K_{TI} = -\sigma_T \sqrt{\pi a} F_T(a/W), \quad (4)$$

where the thermal stress is

$$\sigma_T = -\Delta T \Delta \alpha E, \quad (5)$$

where $\Delta \alpha = \alpha_s - \alpha_f$, and α_s and α_f the CTE's of the specimen and the fixture, respectively. The effect of the fixture modulus in this study is negligibly small because Invar is by far stiffer than underfills. It should be noted that Eqn. (5) is valid for elastic materials because it is

expressed as a time-independent modulus. For viscoelastic materials which have time-dependent moduli, the thermal stress is expressed as

$$\sigma_T(t) = -\int_0^t E(t'-\tau') \frac{\partial(\varepsilon_T)}{\partial \tau} d\tau, \quad (6)$$

with the thermal strain

$$\varepsilon_T(t) = \Delta\alpha [T(t)] \Delta T(t), \quad (7)$$

and $E(t)$ is the relaxation modulus and the reduced time [6] is

$$t' - \tau' = \int_{\tau}^t \frac{1}{a[T(\xi)]} d\xi, \quad (8)$$

where $a[T(t)]$ is the thermal shift factor.

The convolution integral involving the thermal shift factors for the thermal stress in Eqn. (6) was calculated using a numerical integration scheme. [7] The validity of the numerical integration scheme was checked.

The thermally induced fracture toughness K_{TIC} can then be experimentally determined by measuring the critical temperature T_C at which a crack grows during transient thermal loading. Measurement of a critical temperature replaces measurement of a critical load carried out in the case of mechanically-induced fracture tests. The corresponding critical time t_C must be used for calculating the critical thermal stress as in

$$K_{TIC} = -\sigma_T(T_C, t_C) \sqrt{\pi a F_T(a/W)}. \quad (9)$$

This equation is capable of taking cooling rate effects into account since the convolution integral contains the time and temperature terms of Eqn. (6).

Experimental Results

The relaxation modulus and CTE of two commercial underfills were measured. Both are required to calculate the thermal stress in Eqn. (6). Thermal SENT tests and conventional mechanical fracture toughness tests were conducted.

A dynamic mechanical analyzer (DMA) was used to determine the relaxation modulus over a temperature range of approximately 200°C. The glass transition temperatures of underfill A (UF-A) and B (UF-B) were found to be 150 and 100°C, respectively. The master curves were constructed in Figure 3 and fitted using a Prony series. The Prony series fits were

then used to compute the viscoelastic thermal stress. As the reference temperatures for both underfills their respective glass transition temperatures were chosen even though this choice is somewhat arbitrary.

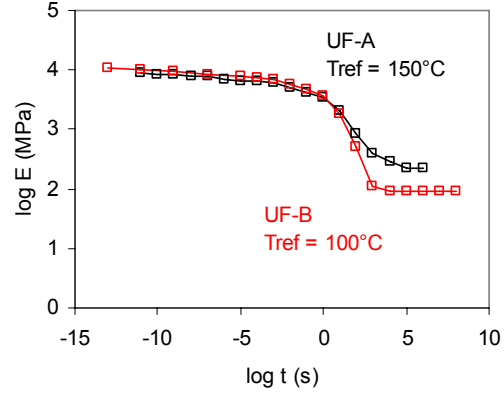


Figure 3. Master curve Prony series fits

The thermal shift factors in Figure 4 were fit using a fifth order polynomial [1].

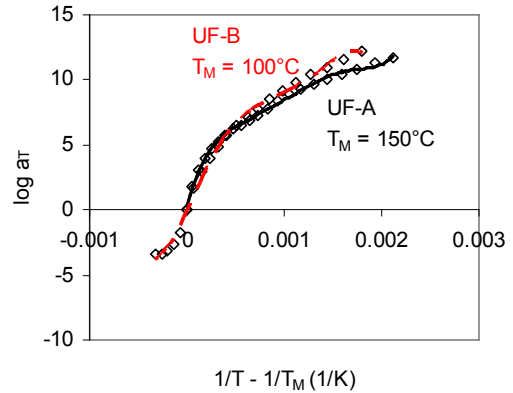


Figure 4. Thermal shift factors for $T_M = 150^\circ\text{C}$ (UF-A) and 100°C (UF-B)

Neither the WLF nor the Arrhenius expression was used because the master curves contain the relaxation modulus above and below the glass transition [6]. The WLF equation can represent the thermal shift at temperatures above the glass transition and the Arrhenius equation below it.

The coefficient of thermal expansion was measured with a thermo-mechanical analyzer (TMA) at a slow heating rate of 2°C/min. UF-A exhibited a constant CTE of 30.06 ppm/K from 23°C to 150°C while UF-B showed a nonlinear thermal expansion behavior in Figure 5. Note that the glass transition temperature of UF-B is approximately 100°C.

The reference temperature was taken to be the same as the glass transition temperature.

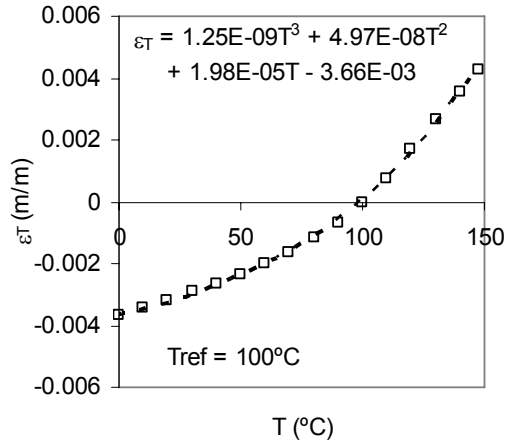


Figure 5. Thermal expansion of UF-B

The mechanically induced fracture toughness, K_{IC} , of the underfill in Figure 6 was determined using single edge notched bending (SENB) specimens using Tada's weight function [2] in the fracture toughness equation:

$$K_{IC} = P_C \frac{f(a/W)}{B\sqrt{W}}, \quad (10)$$

where P_C is the critical load at which the crack grows or the maximum load for this case. The specimen was nominally 4.3 mm wide, 60 mm long and 4 mm thick in the out-of-plane direction. UF-A with a room temperature yield strength of 64 MPa satisfies the ASTM standard requirement for specimen dimensions:

$$a, B, (W - a) > 2.5 \left(\frac{K_I}{\sigma_{YS}} \right)^2, \quad (11)$$

The crack length of the UF-B specimen was slightly smaller than this requirement, but this seemed not to affect its fracture toughness at crack length ratios from 50 to 73%. This may be attributed to the relatively small maximum strain or brittleness of UF-B. The maximum strain of the stress-strain curves was measured to be smaller than even 1.5% at room temperature and 5% at 75°C. Both maximum strain values are by far smaller than those for structural metals for which the dimension requirement of this ASTM standard were developed.

The fracture toughness of UF-B increases with decreasing temperature while that of UF-A remains quite independent of temperature over a wide range of temperature. The fracture toughness of UF-B drops at

temperature above its glass transition temperature.

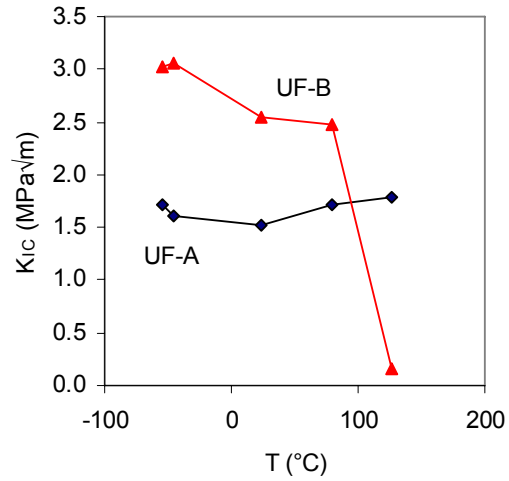
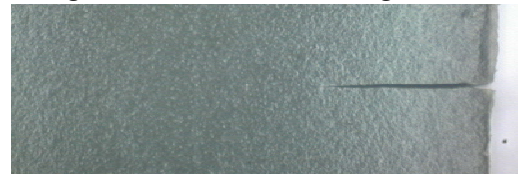


Figure 6. Mechanical fracture toughness

The thermally induced fracture toughness, K_{TIC} , of the underfills was determined using thermal SENT specimens [1]. The fixtures were made of Invar (CTE of 2 ppm/K) and stainless steel (CTE of 16 ppm/K) in order to create different thermal mismatch stresses. Various initial crack length ratios, from 20 to 70%, were created in the underfill samples to induce thermal fracture throughout a wide range of temperatures.

Both sample and fixture were placed in an environmental chamber and heated gradually to 150°C. After thermal equilibration at 150°C, the specimen was attached at both ends to the fixture using a resin similar to the underfill. After full cure the specimen was cooled at a constant rate of 1 or 4°C/min. Crack growth was monitored during cool down. Critical temperature T_C and time t_C at which the specimen fractured into two pieces were recorded. The initial crack length of each specimen was measured by microscope after the test.

Cracks shortly before and after crack growth of the two underfills are shown in Figure 7. Both crack faces (Fig. 7(a)) were completely closed before cooling started. The difference in gaps ahead and behind the initial crack tip indicated that viscoelastic deformation occurred during the transient thermal loading.



(a) UF-A at 61°C shortly before critical fracture



(b) UF-A at 58°C after critical fracture



(c) UF-B at -46°C after critical fracture

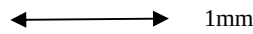


Figure 7. Crack tip deformation pattern

Discussion

Critical temperature and critical time of the two underfills were measured over a wide range of crack length ratios and are plotted in Figure 8. T_C decreases with decreasing initial crack lengths because samples with longer cracks experience greater thermal stress intensity factors. A large difference in the critical temperatures of the two underfills was observed. UF-B exhibited significantly lower critical temperatures than UF-A. This implies qualitatively that UF-B can sustain larger thermal stresses than UF-A. Because of its simplicity, we propose to use the critical temperature measurement method as a technique for pre-screening materials. However, the critical temperature is not a material property because it is a parameter which combines the modulus, CTE and thermal fracture toughness.

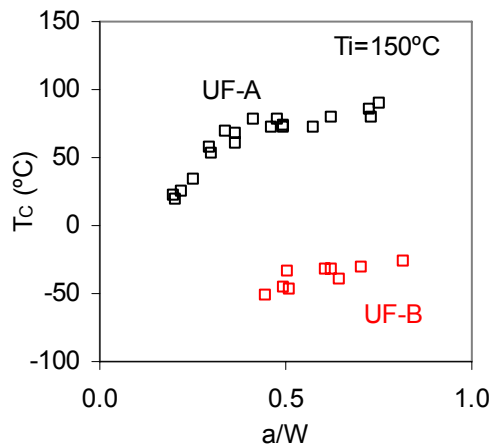
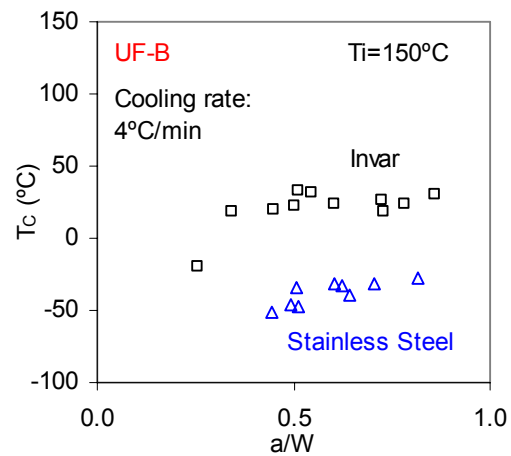


Figure 8. Critical temperature measurements

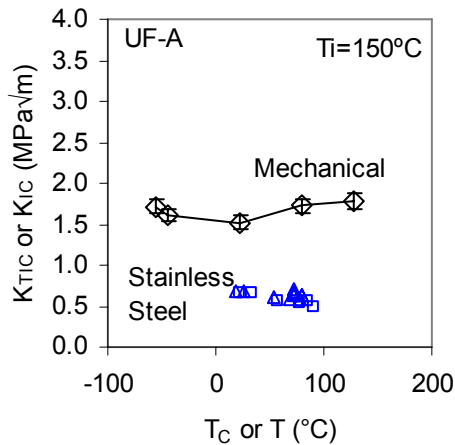
The effect of the CTE mismatch on the T_C of UF-B was determined using Invar and stainless steel fixtures (Fig. 9). The CTE of Invar is significantly smaller than that of stainless steel such that the CTE mismatch between the Invar fixture and the sample resulted in a higher T_C . This indicates that the critical temperature range can be controlled through the choice of the fixture material in order to match it with an operating temperature or other temperature of interest. For instance, stainless steel fixtures may be more appropriate than Invar for accelerated thermal cycles that require subambient temperatures.

Figure 9. CTE mismatch effect on T_C

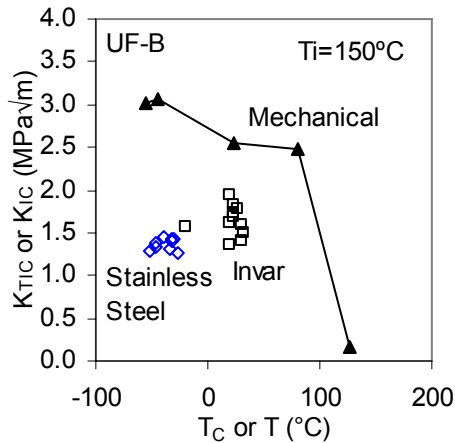
The thermal fracture toughness of the two underfills in Figure 10 was calculated using Eqn. 9. Both underfills consistently exhibited large differences between the mechanical and the thermal fracture toughness. In case of UF-A, the thermal fracture toughness is approximately 35% of the mechanical fracture toughness at high temperatures. The thermal fracture toughness of UF-B is nearly 50% of the mechanical one at subambient temperature. The mechanical fracture toughness increases upon cooling from room temperature while the thermal fracture toughness decreases for the same temperature range. According to these results the underfill reliability for at least these two underfills in electronic packages is highly overestimated when the mechanical fracture toughness is used.

Both the mechanical and thermal fracture toughness of UF-B are consistently greater than those of UF-A. In addition, the ratio between the K_{TIC} 's of the two underfills seems

to be same as that of the K_{IC} 's. This implies that it is possible to correlate K_{IC} to K_{TIC} .



(a) UF-A



(b) UF-B

Figure 10. Thermal fracture toughness

Conclusions

We have shown that the values of fracture toughness obtained for filled organic materials such as underfills depend on the method by which the stress is applied to the fracture samples. To obtain values that are most advantageously used in the prediction of underfill reliability we propose using thermal transient loading which simulates the stresses experienced during reliability testing and in the field. Fracture toughness values that are obtained with conventional mechanical loading were found to be significantly higher than those obtained by thermal loading. This indicates that conventionally obtained fracture toughness values will overestimate the reliability of

materials in applications which are experiencing thermal stresses.

We showed further that the critical temperature described above can be used as a parameter to rank underfills in terms of their reliability in high performance packaging applications.

Finally the results for the two investigated underfills are consistent over a wide range of temperatures and crack length to width ratios.

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