

# Characterization of Viscoelastic Behaviour of a Molding Compound with Application to Delamination Analysis in IC Packages

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## Abstract

Over the glass transition temperature, molding compounds strongly exhibit viscoelastic behavior which causes their Young's moduli to be not only temperature-dependent but also time-dependent. In the present study, the stress relaxation test is used for the characterization of the viscoelasticity of an epoxy molding compound. The viscoelastic properties of the epoxy molding compound are determined in terms of Prony coefficients, relaxation time and time-temperature shift factors. Furthermore, the effect of viscoelasticity on the delamination in IC packaging is investigated.

## 1. Introduction

Realistic modeling and analysis of the mechanical performance and reliability of electronic packages requires sophisticated constitutive models for the many complex, non-traditional engineering materials that constitute these intricate devices. Plastic encapsulation is employed for a significant percentage of IC packages due to its cost effectiveness. The proper modeling of the behavior of such plastic materials is becoming important in reliability studies of these IC packages.

In the analysis of plastic-encapsulated IC packages, the viscoelastic property of materials such as epoxy molding compound or underfill has usually been neglected and the epoxy molding compound is often modeled as an elastic material. However, over the glass transition temperature molding compound strongly exhibits viscoelastic behavior which causes its Young's modulus to be not only temperature-dependent but also time-dependent. Bimaterial interfacial delamination between molding compound and copper during solder reflow is expected to be influenced by the viscoelasticity of the molding compound.

The temperature dependence of the modulus, strength and thermal expansion coefficient of epoxy molding compounds has been studied extensively with many investigators reporting the temperature dependence of the dynamic modulus instead of the static modulus. For determining the viscoelastic properties of a polymer, several different experimental methods have been used, including stress-relaxation test, dynamic mechanical analysis test and creep-compliance test. Several studies describing the thermo-viscoelastic response of epoxy molding compounds have appeared in the literature. Julian [1], Kenner [2] and Brian et al. [3] characterized the relaxation modulus in uniaxial tension for the relaxation modulus. Yeung [4], Driel [5] and He [6] investigate the viscoelastic behavior of molding compound and underfill materials using dynamic mechanical analysis (DMA).

In the present study, the stress relaxation test is used for the characterization of the viscoelasticity property of an epoxy molding compound. The viscoelastic properties of epoxy molding compound are obtained in the form of the Prony coefficients, relaxation time and time-temperature shift factors. Energy release rates  $G$  based on J integral are used for the calculation of the effect of viscoelasticity on the delamination in a PQFP during lead-free solder reflow with thermal load.

## 2. Experimental Procedure of Material Characterization of Viscoelastic Molding Compound

In this study, the stress relaxation test is used to determine the relaxation modulus of a material as a function of time  $t$ . In its idealized form, the test involves inducing a known strain instantaneously in a specimen and subsequently holding the strain constant. The stress is measured indirectly from the load and decreases over time. For a uniaxial stress-relaxation test, the relaxation modulus can be defined as

$$E(t) = \sigma(t) / \varepsilon_0 \quad (1)$$

where  $\sigma(t)$  is the stress as a function of time and  $\varepsilon_0$  is the constant strain applied during the relaxation test (see Fig. 1).

In this study, the specimens are 125mm in overall length and had a rectangular cross-section with nominal dimensions of 12.5mm by 1.5mm in the test section. The stress relaxation experiments are conducted with an Instron Microtester 5848 equipped with a thermal oven and the initial span length between the upper high temperature pneumatic clip and the lower one is 100mm (see Fig. 2). A constant strain 0.25% is applied suddenly (in one second) to a specimen (see Fig. 3) and the resulting load is measured as a function of time (see Fig. 4) over 30 minutes. The time-dependent stress is calculated from the load and the relaxation modulus  $E(t)$  is determined as the time-dependent stress divided by the applied strain (see Fig. 5) based on equation (1).

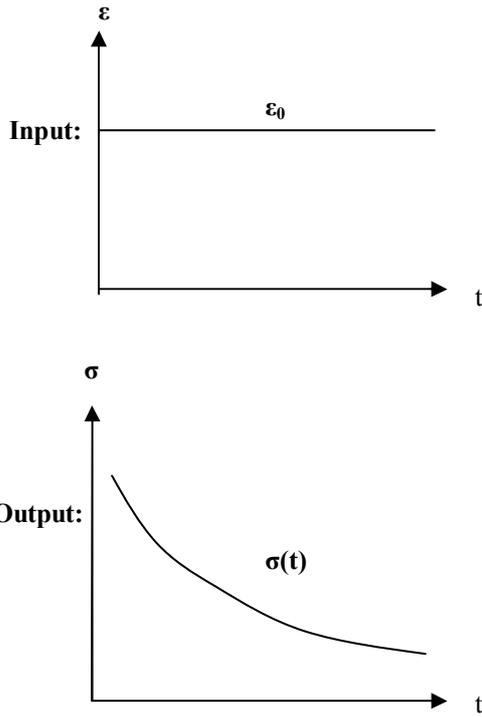


Fig. 1 Strain input and stress output during a stress relaxation test.

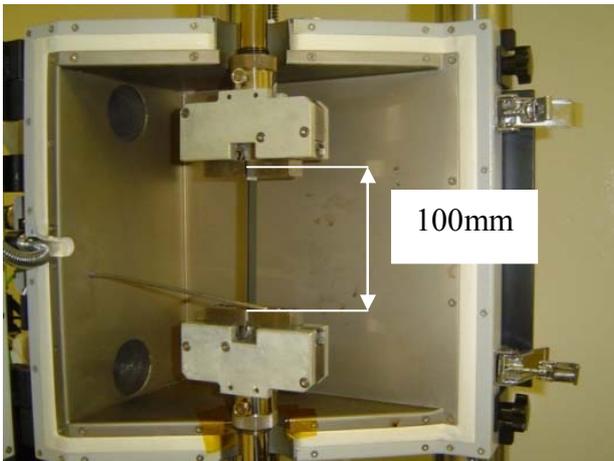


Fig. 2 Stress relaxation test setup.

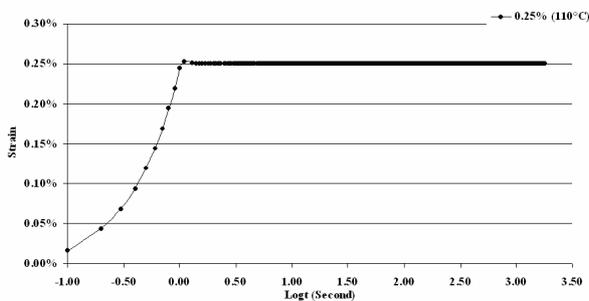


Fig. 3 Constant strain added for stress relaxation testing at 110°C.

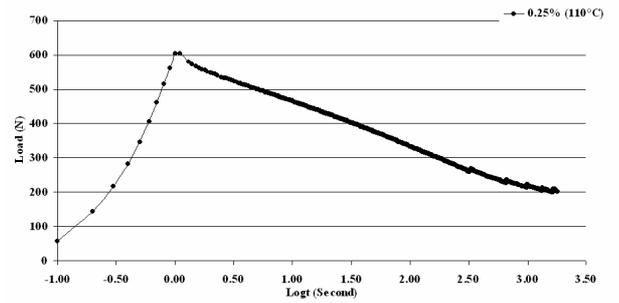


Fig. 4 Load curve obtained at 0.25% strain as a function of time at 110°C.

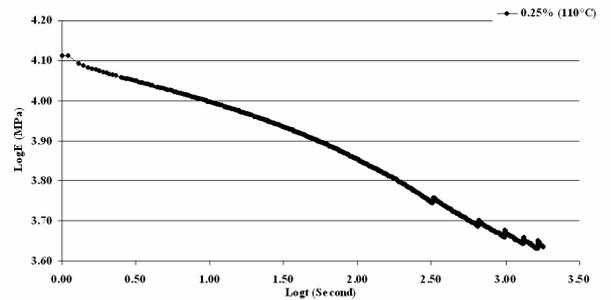


Fig. 5 Uniaxial stress relaxation modulus obtained at 0.25% strain as a function of time at 110°C.

### 3. Uniaxial Relaxation Tests

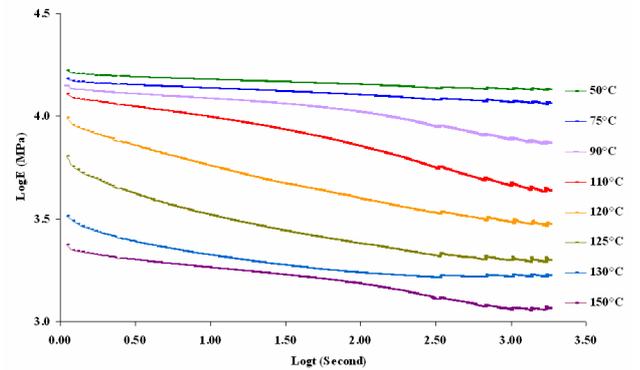


Fig. 6 Uniaxial stress relaxation modulus obtained at 0.25% strain as a function of time for the temperature range 50°C to 150°C.

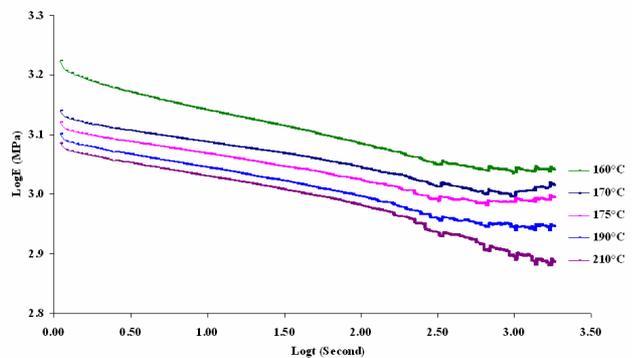


Fig. 7 Uniaxial stress relaxation modulus obtained at 0.25% strain as a function of time for the temperature range 160°C to 210°C.

The uniaxial stress relaxation modulus for an epoxy molding compound has been determined at 0.25% strain for the temperature range 50°C to 220°C. The selection of 0.25% strain is a good choice according to the study done by Julian [1] and Brian [3]. Typical results obtained at 0.25% strain for around 30 minutes are shown in Fig. 6 and Fig. 7.

From Fig. 6, it can be seen that the time-dependent behavior is most pronounced in the vicinity of temperature 110°C due to the reason that 110°C is the glass transition temperature of the epoxy molding compound investigated. In this transition region, the relaxation modulus has its strongest dependence on time and also is very susceptible to slight thermal variations over which the material can undergo drastic changes in its mechanical behavior. Comparing Fig. 6 with Fig. 7, it seems that the temperature effect is pronounced near the glass transition temperature (110°C) and less pronounced at low (from 50°C to 70°C) or high temperatures (from 160°C to 210°C).

The viscoelastic behavior of a molding compound can be analyzed using the time-temperature superposition principle (TTS), which is based on the empirical observation that the viscoelastic behavior of polymer materials at a reference temperature  $T_0$  can be related to the behavior at another temperature  $T$  by changing the experimental time or frequency scale [7-9]:

$$E(t, T_0) = E(\alpha_T t, T) \quad (2)$$

where  $E$  is the relaxation modulus and  $t$  is the time. The physical meaning of equation (2) is clear: the relaxation modulus determined at temperature  $T$  and time  $\alpha_T t$  is equivalent to the relaxation modulus determined at reference temperature  $T_0$  and time  $t$ . Generally speaking, from the polymers with the glass transition temperature  $T_0$ , the shift factor  $\log \alpha_T$  was found to approximately follow the Williams-Landel-Ferry (WLF) equation [7]:

$$\log \alpha_T = -\frac{C_1(T - T_0)}{C_2 + (T - T_0)} \quad (3)$$

where  $C_1$  and  $C_2$  are constants,  $T_0$  is the reference temperature which is often taken as the glass transition temperature  $T_g$ . In this equation,  $\log \alpha_T$  is positive when  $T < T_0$  and negative when  $T > T_0$ . As well as the WLF equation, the Arrhenius equation is also a popular method for fitting the shift factor:

$$\log \alpha_T = A + \frac{B}{C_2 + T_{ABS}} \quad (4)$$

where  $A$  and  $B$  are constants,  $T_{ABS}$  is the absolute temperature in degrees Kelvin.

At low temperatures, the relaxation process is slow and requires a longer time for experimental observation; at high temperatures, it is fast and a shorter time is required for experimental measurement. Therefore, the time-temperature superposition principle (TTS) assumes that changing the experimental temperature has the same effect as changing the relaxation time. Using the TTS experimental results, it is

possible to construct a master curve at an arbitrary reference temperature which extends the relaxation modulus beyond the range of laboratory scale in either the time or the frequency domain. This is important for accurately modeling the package stress because it allows one to consider the time-dependence of the mechanical response of the polymer.

Based on the stress relaxation data in Fig. 6 and Fig. 7, one master curve can be formed by shifting the curves in Fig. 6 and Fig. 7 referenced to the stress relaxation curve at the reference temperature 110°C (see Fig. 8). The lower temperature relaxation modulus curves are shifted to the left along the time axis and the higher temperature relaxation modulus curves are shifted to the right along the time axis which can extend the time to a longer time range. For the master curve generated, the horizontal shift factors are functions only of temperatures and they can be fitted using the Williams-Landel-Ferry (WLF) equation. Because of this, the shift factors versus temperature are plotted for 110°C reference temperature (see Fig. 9). Some experimental errors are to be expected in obtaining these shift factors.

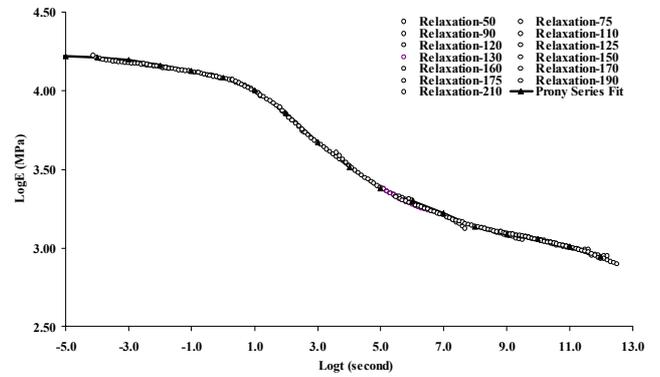


Fig. 8 Stress relaxation master curve obtained at 0.25% strain at  $T_0 = 110^\circ\text{C}$ .

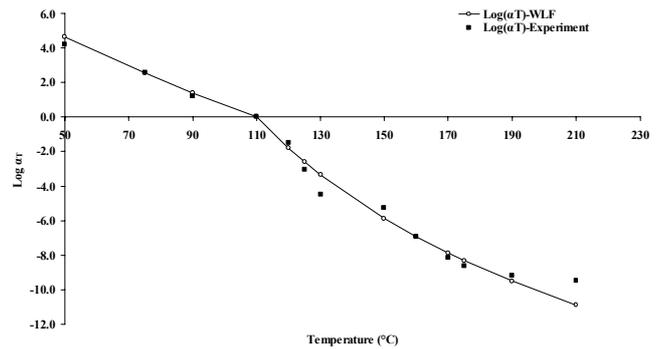


Fig. 9  $\text{Log} \alpha(t)$  versus temperature for  $T_0 = 110^\circ\text{C}$ .

In Fig. 9,  $\text{Log} \alpha_T$  is plotted as a function of temperature for a reference temperature of  $T_0 = 110^\circ\text{C}$ . Within the experimental temperature range, the WLF equation given in equation (3) can be used to fit the shift factor data rather well, as shown in Fig. 9. The fitting parameters are

$$C_1 = 30 \text{ and } C_2 = 500 \text{ when } T < T_0 = 110^\circ\text{C}$$

$$C_1 = 25 \text{ and } C_2 = 130 \text{ when } T > T_0 = 110^\circ\text{C}$$

These results deviate from the typical “universal” values of  $C_1 = 17.44$  and  $C_2 = 51.6$  [7] when  $T_g$  is selected as the reference temperature. In fact,  $C_1$  and  $C_2$  do vary from polymer to polymer and large deviations from the “universal” values have been observed in many polymer systems. In addition, values of  $C_1$  and  $C_2$  depend strongly on the choice of reference temperature  $T_0$ : a small variation in  $T_0$  can cause large deviations from the “universal” values.

Table I Comparison of parameters of WLF equation.

	Material	$T_0$ (°C)	$C_1$	$C_2$
Yeung T.S. [4]	EMC	175	27.423	500
Brian D.H. [3]	EMC	100	16.77	65
He Yi [6]	Non-flow Underfill	100	37.99	200.27
This study	EMC	110	30	500
			25	130

From Fig. 9, it is observed that the shift factor undergoes a noticeable change in slope as it passes through the glass-transition temperature. Because of this, the shift factor was fitted separately both above and below the glass-transition temperature. The reason for the changes of the shift factor through the glass-transition temperature may be due to the change of material property in the transition region. At lower temperatures, the material exhibit “glass-type behavior” which is characterized by a relatively high elastic relaxation modulus. At higher temperatures, the material undergoes “rubber-type behavior” where the elastic modulus is greatly reduced with respect to the modulus at lower temperature. The drastic changes of the relaxation modulus from the lower temperature to higher temperature may cause the noticeable change of the shift factor in the transition region.

This phenomenon was also found in research work done by Yeung and Yuen [4], Julian [1] and Brian [3]. Yeung and Yuen [4] used the constant parameters  $C_1$  and  $C_2$  for calculation of the shift factor in the whole temperature range which caused large errors in some shift data (see Fig. 10). From Fig. 9 and Fig. 10, it can be seen that fitting separately both above and below the glass-transition temperature can improve the accuracy of shift factor fitting. In contrast, Julian [1] chose the Arrhenius equation as an approximation of the horizontal shift factor and the shift factor was fitted separately both above and below the glass-transition temperature (see Fig. 11). Brian [3] chose Arrhenius equation for the fitting of the shift factors below the glass-transition temperature and WLF equation for the fitting of the shift factors above the glass-transition temperature (see Fig. 11).

From Fig. 11, it can be seen that the shift factors obtained in the present study develop in a trend similar to that obtained by Julian [1] for the section above the glass-transition temperature of 110°C. There is some difference which may be

attributed to the fact that Brian [3] chose 100°C as a reference temperature (see Table I) while 110°C is the reference temperature used in the present study. It should be noted that the polymeric molding compounds studied by both Julian [1] and Brian [3] is different from the epoxy molding compound studied here.

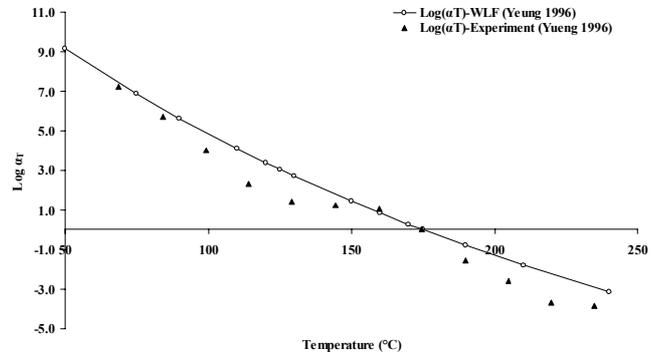


Fig. 10  $\text{Log}a(t)$  versus temperature for  $T_0 = 175^\circ\text{C}$  by Yeung and Yuen [4].

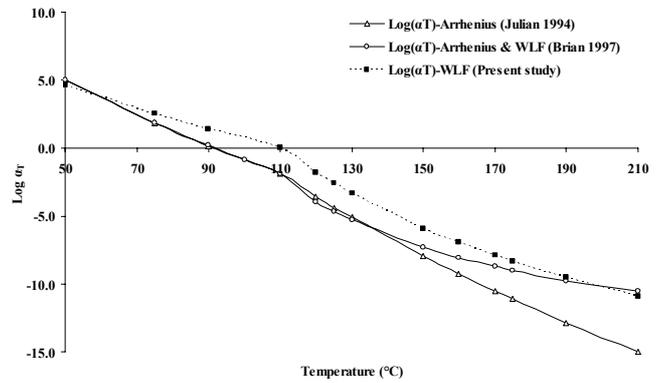


Fig. 11  $\text{Log}a(t)$  versus temperature for  $T_0 = 110^\circ\text{C}$  by Julian [1] and  $T_0 = 100^\circ\text{C}$  by Brian [3].

A Prony series was used to fit the master relaxation curve and this series is of the form

$$E(t) = E_\infty + \sum_{i=1}^N E_i e^{-\frac{t}{\tau_i}} \quad (5)$$

where  $E_\infty$  represents the equilibrium modulus after the molding compound has fully relaxed;  $E_i$  and  $\tau_i$  are the Prony coefficient and relaxation time for each element, respectively.

The Prony coefficients for the master curve are given in Table II. The fitting was effected by using the collocation method. The Prony representations of the master curve combined with the temperature dependence of the horizontal shift factor is required for finite element calculations based on linear representation of the time-dependent constitutive behavior. This master curve and the Prony series fit are shown in Fig. 8 which clearly shows that the Prony series provides an excellent fit to the stress relaxation master curve.

Table II Prony coefficients for 0.25% strain master curve.

i	$\tau_i$	$E_i$ (Mpa)
1	1.00E-04	200
2	1.00E-03	800
3	1.00E-02	1500
4	1.00E-01	1000
5	1.00E+00	1100
6	1.00E+01	2700
7	1.00E+02	2900
8	1.00E+03	2500
9	1.00E+04	900
10	1.00E+05	950
11	1.00E+07	600
12	1.00E+08	120
13	1.00E+09	180
14	1.00E+11	200
15	3.00E+12	250
	$E_\infty$	700

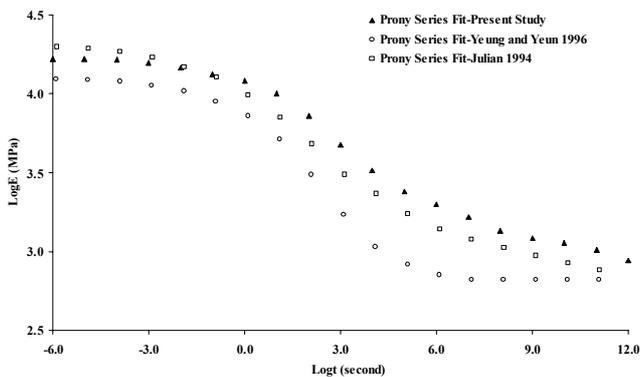


Fig. 12 Comparison of viscoelastic properties obtained by Yeung/Yuen (1996), Julian (1994) and present study (2006) at  $T_0 = 110^\circ\text{C}$ .

Based on the Prony series and the shift factors, the relaxation modulus of three kinds of molding compound at temperature  $T = 110^\circ\text{C}$  have been plotted in Fig. 12. It can be seen that the viscoelastic properties of the epoxy molding compound (Cookson Semiconductor Packaging Materials) in this study are close to the properties of the polymeric molding compound investigated by Julian [1]. On the other hand, the epoxy molding compound studied by Yueng and Yeun [4] is softer than the epoxy molding compound in this study. However, all these kinds of molding compound strongly exhibit viscoelastic behavior which causes its Young's modulus to be not only temperature-dependent but also time-dependent.

#### 4. Effect of Viscoelasticity on Delamination

Williams/Marshall [10] put forth the idea that fracture mechanics for viscoelastic materials could be treated with an approach that is similar to the traditional fracture mechanics developed for elastic materials by simply replacing the time-

independent values of modulus with the equivalent viscoelastic relaxation modulus. Thus, the J integral has been used for the analysis of the delamination between leadframe pad and molding compound in IC packaging. The aim of this study is to check the effect of viscoelastic material property on delamination analysis.

In this study, the entire thermal history of a plastic IC package is simulated during a lead-free solder reflow process lasting about 8 minutes. Two-dimensional finite element models were constructed for a 160-leaded PQFP the dimensions of which are shown in Fig. 13.

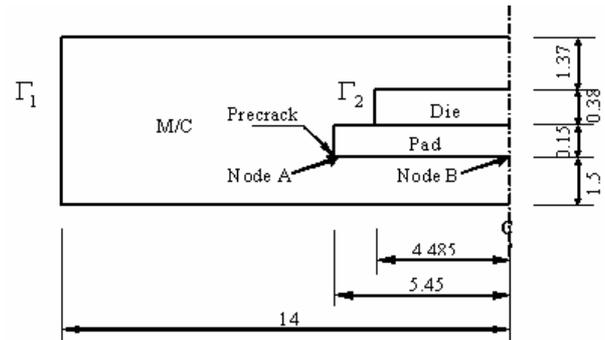


Fig. 13(a) Cross-section of PQFP studied.

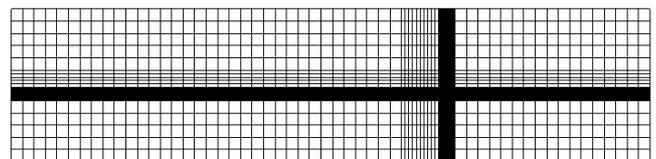


Fig. 13(b) Finite element mesh of left half of package ( $a = 1.5\text{mm}$ ).

A state of zero thermal stress in the package will be reached when the package is at a uniform temperature equal to the curing temperature of  $175^\circ\text{C}$ . During a lead-free solder reflow process, the thermal strain energy release rate at the crack tip will become almost zero at around 250s and 390s. All the data presented in this study are taken from the values in the critical range of 250s-390s for lead-free solder reflow. The material properties, boundary and loading conditions are same as those used in reference [11] except epoxy molding compound whose properties have been measured in this study and used for this simulation.

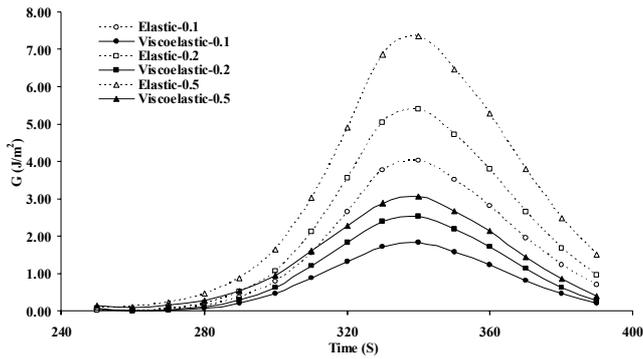


Fig. 14 Comparison of  $G(t)$  during lead-free solder reflow process ( $a = 0.1, 0.2$  and  $0.5\text{mm}$ ).

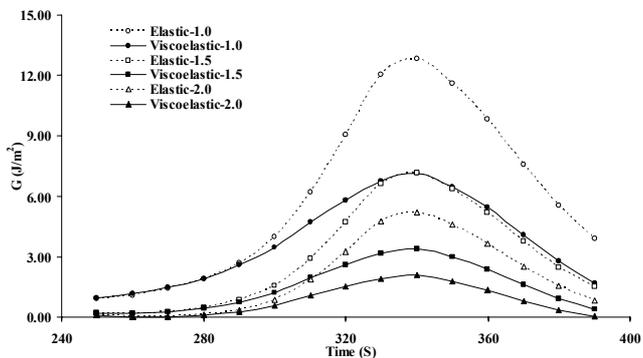


Fig. 15 Comparison of  $G(t)$  during lead-free solder reflow process ( $a = 1.0, 1.5$  and  $2.0\text{mm}$ ).

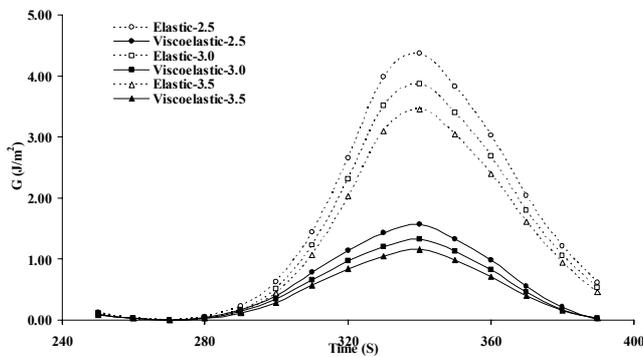


Fig. 16 Comparison of  $G(t)$  during lead-free solder reflow process ( $a = 2.5, 3.0$  and  $3.5\text{mm}$ ).

Figs. 14 and 15 compare energy release rate results obtained using the viscoelastic property of molding compound and using linear elastic property of the molding compound. It can be seen that energy release rates using the viscoelastic property of molding compound are very different from the solutions using linear elastic property of molding compound (more than 200%), especially near the maximum point of the solder reflow temperature profile where the temperature effect on molding compound is strong. Fig. 16 shows that the maximum energy release rate is significantly lower (more than 300%) by considering the viscoelastic behavior of the molding

compound than by considering linear elastic behavior of the molding compound. Thus the maximum energy release rate would be quite different during lead-free solder reflow process if the molding compound is considered to be viscoelastic rather than linear elastic. However, this does not imply that predictions of delamination assuming linear elastic behaviour of molding compounds are wrong. This is because the energy release rate  $G$  is only half the story. The fracture toughness needs also to be obtained as a function of the viscoelasticity of the mold compound. Only then can prediction of delamination be attempted. Further studies will be conducted on the effect of viscoelasticity on fracture toughness. From Figs. 14 and 16, it can be seen that the  $G(t)$  difference between using viscoelastic property and linear elastic property is smaller for interface cracks of medium length.

## 5. Conclusions

Above the glass transition temperature, molding compounds strongly exhibit viscoelastic behavior which is temperature-dependent and time-dependent. The viscoelastic properties of an epoxy molding compound (the Prony coefficients, relaxation time and time-temperature shift factors) are measured by performing stress relaxation tests using an Instron Microtester. The properties of epoxy molding compound measured compared well with the properties of molding compound obtained by other researchers. Furthermore, it has been shown that the energy release rate at a crack in a package computed for a viscoelastic molding compound is significantly lower than that for a linear elastic molding compound. However, it is difficult to tell how this might affect the prediction of the delamination temperature unless the effect of viscoelasticity on the interface fracture toughness is also known.

## Acknowledgments

The author would like to acknowledge the support of the National University of Singapore in providing a research scholarship under which this research has been conducted.

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