THERMOMECHANICAL EFFECTS IN ELECTRONIC PACKAGES

by

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ABSTRACT

Many electronic components are composite structures consisting of silicon, metal and polymer, e.g., IC plastic packages and high density interconnects. Due to property mismatch between these materials, high thermal and shrinkage stresses develop during manufacturing and operation, leading occasionally to mechanical failure. This can be cracking in any of the materials of the component or (and) cracking along any of the interfaces. To evaluate the strength of the component, a general analysis scheme is implemented: 1) The residual stresses in a model geometry, representing the component, are calculated by a viscoelastic analysis. 2) The energy release rates for propagation of crack-like defects in the model, either existing or assumed, are calculated by Finite Element Analysis and the modified crack closure method. 3) The results are expressed in dimensionless form so that they are generally applicable. 4) A fracture criterion is introduced by comparing the calculated energy release rate to its critical value. The critical energy release rate for propagation of a crack in any of the materials of the component is measured by fracture toughness experiments and for propagation of a crack along any of the interfaces by the Island Blister Test (IBT). A number of test structures have been constructed to verify the predictions of the present method. The experimental results have shown good agreement with the analysis.

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I dedicate this work to my late father Konstantinos.
1. INTRODUCTION

1.1 Background

The problem of reliability of structures has been one of the most important issues that engineers have had to face. To address it several approaches have been devised. The oldest is trial and error; the actual structure is built and appropriate intuitive modifications are introduced until it performs its intended function without failure. Although time and resource consuming, this approach is still in use in areas where scientific knowledge is not sufficiently developed.

A newer method consists of the development of a general understanding of the mechanical behavior of structures, which allows for the a priori selection of all the important design parameters for their safe and effective performance. This is quick and efficient and is the preferred method in applications for which the appropriate amount of knowledge exists or can be acquired.

The first step in the evaluation of the mechanical performance of a structure is to calculate the most severe stress state which develops during its construction and operation, using appropriate stress analysis techniques. Once the stresses and strains have been calculated, they can be introduced in one of the many failure criteria to predict whether or not failure will occur in the actual structure [1]. However, this is not possible in cases where stress singularities are present since failure would be predicted at almost zero load.

More recently, an alternative has been developed to avoid problems related to such stress singularities. This is fracture mechanics. It has been
observed that flaws can be detected in most materials used in any application and even if flaws cannot be detected, the mechanical strength of a material behaves as if a characteristic flaw existed in it. Also, often it is necessary to introduce flaws in the components of a structure so they can perform their intended function: holes are drilled in plates to bolt them together or notches are introduced for mechanical interlocking etc.

In 1921 Griffith [2] was the first to observe that the strength of glass fibers increased as the diameter of the fibers decreased, approaching the theoretical strength. He concluded that the fiber behaved as if there existed a crack-like flaw of a characteristic size in the material. Performing a linear elastic analysis based on energy balance, he derived an expression for the strength of the fiber. The concept of the elastic energy release rate played a key role in his analysis; this is the potential energy released by the loaded solid per unit of crack extension. He identified the critical value of the energy release rate, at which the crack rapidly propagated, as twice the surface energy, both normalized to the unit advance of the crack.

A few decades later, Irwin [3], [4], found that the presence of plastic deformation ahead of the crack tip decisively influences the critical value of the energy release rate. For example, it is higher under plane stress conditions than under plane strain, due to the different amounts of plasticity present. Thus, the critical energy release rate is that obtained under plane strain condition, since only then it is independent of the thickness. The work of Irwin and others led to modern Fracture Mechanics, one part of which is Linear Elastic Fracture Mechanics (LEFM) [5], [6]. According to LEFM, if a crack is present in a loaded linear elastic solid, with limited yielding ahead of the crack tip (small scale yielding conditions), the crack will propagate when the
energy release rate obtained from an elastic analysis is greater than the critical energy release rate of the material, the latter having been measured in standardized tests.

In the present research, Fracture Mechanics is employed to solve model problems associated with the mechanical performance of electronic components, such as high density interconnects and plastic encapsulated chips. The general methodology is as follows: (i) A crack-like flaw is assumed to exist in the structure. This flaw can either be one which is introduced in the structure for functional reasons or it can be an inherent one. (ii) The energy release rate associated with extension of the crack, under the given loading and geometry, is calculated by the Finite Element Method (FEM), [7]-[9]. (iii) Then the calculated energy release rate is compared to the critical value obtained by fracture toughness tests or adhesion measurements on the actual materials. (iv) Failure is predicted when the calculated energy release rate exceeds the critical energy release rate of the material.

1.2 High Density Interconnects

A complicated electronic assembly such as a computer consists of many smaller substructures (memory chips, the Central Processing Unit (CPU), power supply, etc.) which perform various digital and analog operations. All these parts must be connected by some means. Currently the connections are made by printed circuit cards and boards, which are substrates with metal wires deposited on their surfaces. The various electronic parts are attached at specific locations on the card or board, and the entire assembly is connected and ready to function.
In recent years, using new materials and new technology, the size of electronic components has decreased significantly so, more of them can be put in a smaller space. At the same time, however, the interconnections have become more complicated and they must be accommodated in smaller spaces. This need has created High Density Interconnects (HDIs) [10]: a number of printed circuit boards laminated one upon another. Connections between layers are made by metal filled holes between them at appropriate positions; these are called vias.

Usually a HDI is built on a silicon or ceramic wafer substrate. On its surface, the first level of copper wires is deposited; they are less than 25 μm wide and less than 10 μm thick [10]. Then, a polymer film is applied by spin casting, covering the wires. Polyimides are most often used in this application because of their good thermal stability, wear resistance, mechanical strength and low dielectric constant; the latter permits faster propagation of electrical signals in the wires. After the coating has been deposited, circular or square holes are etched in it at appropriate positions and these vias are filled with copper. Next, the second level of copper wires is placed on top of the polymer film, then it is covered by a second polymer film and the cycle is repeated until the desired number of connections have been made. The total thickness of all the layers is less than 20 μm.

Each polymer layer is cured at 250 - 400 °C, so, during manufacturing, high thermal stresses develop. An accurate determination of them is crucial to any analysis trying to predict the strength of such structures. This, also requires accurate knowledge of the mechanical properties of the materials. Since most polymers exhibit viscoelastic behavior over the temperature range
mentioned, in Chapter 2, a method is presented to obtain their viscoelastic properties in a form useful for FEM analysis.

Two types of failures are considered in this work; i) loss of adhesion between the film and the substrate, and ii) through the thickness cracking of the film. As mentioned earlier, the parameter which represents the resistance of a material against cracking is its critical value of the energy release rate, measured by fracture toughness experiments [6]. Similarly, for an interface, it is the critical value of the energy release rate of that interface (also called the specific fracture energy) and a test to measure the latter is the Island Blister Test (IBT) [11]. An improvement in the analysis of the IBT is presented in Chapter 3, where the experimental values obtained from it are converted to interface specific fracture energy by a large deflection elastic-plastic analysis.

The problem of the loss of adhesion between the polymer film and the substrate is the topic of Chapter 4. This is caused by the residual stress which develops in the film from the differential cooling shrinkage between them. The most severe case is an interfacial crack from the edge of the film, where a stress singularity exists, even without a crack [12]. A numerical method to calculate the energy release rate for the propagation of such a crack is presented in Chapter 4. If the release rate is greater than the specific fracture energy of the interface, a loss of adhesion is probable.

In Chapter 5, through the thickness cracking of the polymer film when the via holes are introduced is examined. To be conservative, we analyze a three dimensional crack with a length equal to the characteristic dimension of the via. The problem of the interaction of such cracks also is addressed. The solutions of these problems provide general guidelines for the design of high
density interconnects and for the selection of materials to use in them. They also lead to the design of test structures for quality control purposes.

1.3 Plastic Encapsulated Chips.

In an Integrated Circuit (IC) [13], the various components are manufactured on a single piece of silicon, the chip or die, by carefully doping, depositing or growing in it, at precise locations and amounts, various other materials. These give to the structure the necessary properties to perform its intended function. Using modern fabrication techniques, very small chips can perform very complicated tasks. The chip, is bonded to a metal frame, the lead frame, with an adhesive (epoxy or polyimide) or with solder. The lead frame serves a number of purposes: (i) It provides structural support to the chip. (ii) It is a kind of conveyor belt, which transports the chip through various assembly steps. (iii) It provides the chip with electrical and thermal connections to the board. (iv) It serves as a support frame to the molten plastic in the encapsulation step.

After the chip has been attached to the lead frame, the electrical connections between the chip and the lead frame are made with fine gold wires and the entire assembly is encapsulated in plastic, such as epoxy. Transfer molding is the process used for this purpose. The plastic protects the chip from adverse environmental influences, such as moisture, radiation, heat etc., and from external mechanical influences, such as bending during handling, impact etc. Plastic is preferred over other types of encapsulation, e.g., metal or ceramic hermetic packages, due to lower cost, and easier fabrication process.
Failures in plastic encapsulated IC packages can occur during manufacturing and operation. These include cracking of the plastic, cracking of the chip, loss of adhesion between the chip and the plastic, and loss of adhesion between the lead frame and the plastic. In Chapter 6 we offer a general analysis of cracking in the polymer, when no delaminations are present.
2. VISCOELASTIC PROPERTIES OF POLYMERS FOR FINITE ELEMENT METHOD ANALYSIS

2.1 Introduction

High thermal and shrinkage stresses develop in electronic packages during manufacturing and operation due to the thermal and shrinkage mismatch between their components which are usually made of silicon, metals and polymers. To make estimates of the stability of such packages, the stresses developed during the curing of the polymer must be calculated.

Finite Element Analysis [7-9], is the most effective method of stress analysis used for electronics packages [14, 15]. To produce useful results, the properties of the materials must be known with reasonable accuracy. In particular, the viscoelastic properties of the polymer films are required, because the films exhibit the highest thermal mismatch. Polyimides are most commonly used for such applications [16] because of their low dielectric constants, high tear resistance, high tensile strength and high temperature resistance. The structure, properties and residual stresses generated during curing and thermal cycling of certain polyimides and benzocyclobutene based polymers have been studied extensively [17-19].

Usually, the residual stresses during curing on silicon substrates are found experimentally from curvature measurements [20]. These have been done for the polymers of interest in the present work: biphenyldianhydride - phenyldiamine (BPDA-PDA) [21], hexafluorodianhydride - aminophenoxybiphenylin (HFDA-APBP) [22], and divinylsiloxane - bisbenzocyclobutene (DVS-BCB) [19]. The structures of the three are shown in Figure 2-1. BPDA-PDA is marketed by Du Pont, HFDA-APBP by Amoco and
DVS-BCB by Dow Chemical under the trade names Pyralin-LX 2611, Ultradel 4212 and Cyclotene, respectively. Despite shrinkage at the curing temperature due to solvent expulsion and polymerization, it has been found that in those polymers, the residual stresses at room temperature are caused by the thermal mismatch between the substrate and the polymer. At the curing temperature, any stresses which arise are relaxed. (It should be mentioned however, that in certain other polymer films shrinkage stresses do contribute to the final residual stress state [21, 23].) Since the polymers under consideration exhibit viscoelastic behavior, their viscoelastic properties must be obtained in a form useful for the FEM analysis.

2.2 Viscoelastic Analysis

The constitutive equation of a polymer film is assumed to be of the following linear viscoelastic form [24]:

\[
\sigma(x,t) = \int_{0}^{t} 2G(\tau - \tau') e(x,\lambda)d\lambda + \int_{0}^{t} K(\tau - \tau') \phi(x,\lambda)d\lambda \tag{2-1}
\]

where \(\sigma\) is the stress at location \(x\) and 'real time' \(t\), \(G\) is the relaxation shear modulus which depends on the material time \(\tau\), \(K\) is the bulk modulus which also depends on the material time \(\tau\), \(e\) is the deviatoric strain rate, \(\phi\) is the volumetric strain rate, \(I\) is the unit tensor, \(\tau\) is the "material time" corresponding to the real time \(t\), and \(\tau'\) is the "material time" corresponding to the real time \(\lambda\).

The "material time", \(\tau\), is related to the "real time", \(t\), by the expression:

\[
\tau(t) = \int_{0}^{t} \frac{ds}{A(T(s))} \tag{2-2}
\]
where $A(T(s))$ is the shift factor at temperature $T$.

The distinction between "real time", $t$, and "material time", $\tau$, is necessary for viscoelastic materials since the response of the material to an imposed strain depends upon the temperature of the experiment: the higher the temperature, the less time it takes for the material to stress relax a given amount after the imposition of a step strain. The time scaling factor between any temperature and the reference temperature (which can be chosen arbitrarily) is the shift factor, $A(T(s))$ above. Obviously, if $T=T_{\text{ref}}$ then $A(T_{\text{ref}})=1$. The shift factor for a given material is found from experiments, by the method of time-temperature superposition [24].

In polymers, the bulk modulus is a much less sensitive function of time than the relaxation shear modulus. Furthermore, in most cases, the contribution of the volumetric strain rate i.e., the second term of Equation (2-1), to the stress state is small compared to that of the deviatoric strain rate i.e., the first term of Equation (2-1). Therefore, to simplify the analysis, we assumed that the bulk modulus, $K$, does not vary with time.

In thin films, it is convenient to obtain the relaxation shear modulus, $G$, from the Young's modulus, $E$, and the Poisson's ratio, $\nu$, by the expression:

$$G = \frac{E}{2(1+\nu)}$$  \hspace{1cm} (2-3)

The variation of the Young's modulus with material time, after a step strain is applied to the material at time zero, is given by the following expression:

$$E(\tau) = E_r + \sum_{j=1}^{n} E_j \exp\left(-\frac{\tau}{\tau_j}\right)$$  \hspace{1cm} (2-4)
where $E_r$ is the relaxed Young's modulus, and $E_j$ is the contribution to the Young's modulus, $E(\tau)$, of the relaxation mechanisms represented by the relaxation time, $\tau_j$. The Poisson's ratio, $v$, also is a function of time, but it is not easy to measure in thin films, and in Equation (2-3), we assume it to be constant. The error introduced in the analysis by this approximation is estimated to be less than 7%. (Varying the Poisson's ratio between 0.30 and 0.50 in Equation (2-3), produces a change of less than 7% above or below the value obtained with an average Poisson's ratio of 0.40). The material parameters of Equation (2-4) are found by dynamic measurements.

When a rectangular strip of a linear viscoelastic material is subjected to a uniaxial cyclic strain, the stress will be out of phase with the applied strain. Stress and strain can be represented as rotating vectors in the complex plane. The stress can be separated into two components, one in phase with the applied strain and the other leading the strain by a ninety degree angle. The ratio of the magnitude of the in-phase stress to the magnitude of strain is related to the recoverable part of the energy of one cycle and is called the storage modulus. The ratio of the magnitude of the out of phase stress to the magnitude of the strain is related to the energy dissipated in one cycle and is called the loss modulus. These two quantities are usually given by the Debye equation:

$$E^* = E_r + \sum_{j=1}^{n} \frac{E_j \omega^2 \tau_j^2}{1 + \omega^2 \tau_j^2} + i \sum_{j=1}^{n} \frac{E_j \omega \tau_j}{1 + \omega^2 \tau_j^2}$$  \hspace{1cm} (2-5)$$

where $\omega$ is the cyclic frequency in (rads/sec). The real part of the above expression is the storage modulus and the imaginary part the loss modulus. According to linear viscoelasticity, the material constants in the above equation are the same as those in Equation (2-4). Therefore, instead of
relaxation experiments one can use dynamic analysis to obtain the constants of Equation (2-4).

There are two common functional forms for the dependence of the shift factor on temperature. One is an Arrhenius expression and the other is the WLF equation [24]. The former is valid at temperatures near a secondary transition and the latter at temperatures above the glass transition temperature; neither is valid for the entire use temperature range of a polymer.

The shift factor $A(T)$ can be obtained over the entire use temperature range, by the method of time-temperature superposition. Briefly, this consists of the following steps: (i) Modulus data at discrete points of a wide temperature range are collected within a small frequency range. (ii) All data are plotted on a modulus versus frequency plot. (iii) A reference temperature is chosen. (vi) All data above the reference temperature are shifted horizontally and sequentially, in a decreasing temperature order, toward the higher frequency values so that a smooth curve is produced. (v) The opposite procedure is repeated for the data below the reference temperature. In this manner, data over a wide temperature but limited frequency range can be converted to data over a wide frequency range at the reference temperature, and vice-versa. The shift factor at a given temperature is the horizontal shift of the data of that temperature during this procedure. (For a general discussion of the viscoelastic properties of polymers the reader is referred to [24].)
2.3. Method

A method to obtain the material constants of Equation (2-4) and of (2-5) is as follows:

(i) The variation of the storage modulus of the polymer with temperature and frequency is obtained from one sample for each polymer by Dynamic Mechanical Analysis (DMA). Films 20 mm long, 10 mm wide and 20 µm thick for BPDA-PDA, 15 µm thick for HFDA-APBP and 25 µm thick for DVS-BCB were analyzed in a Seiko 200 DMA system. Figures 2-2, 2-3 and 2-4 show the results for the three polymers.

(ii) A time-temperature shift is performed on the DMA data to yield the master curve: i.e. the variation of the storage modulus with frequency. See Figures 2-5, 2-6 and 2-7. The shift factor as a function of temperature is given in Figures 2-8, 2-9 and 2-10.

(iii) Each master curve is fitted by the Debye equation with a selected number of terms. Each term represents a relaxation constant and a relaxation time; from these is calculated the relaxed modulus needed for the FEM analysis. In this work nine terms of the Debye equation were chosen and the results of the fit for the three polymers are shown in Table 2-2. A number of points equal to the number of terms in the Debye equation were chosen at equally spaced intervals along the storage modulus axis of the master curve. The frequencies at each of these values of the storage modulus were obtained from the plot. The relaxation times were assumed to be inversely proportional to the cyclic frequencies. A system of nine linear equations resulted from Equation (2-5) and they were solved for the nine unknown
values of $E_j$. Due to the form of this system, all nine $E_j$ are the same for each of the polymers. Therefore, only one value needs to be reported in Table 2-2.

(iv) In this system since temperature is the main variable, the dependence of the shift factor on temperature should be known. In ABAQUS [25], which is the FEM code employed in this work, the shift factor is assumed to be given by the WLF equation, however as mentioned, WLF is valid only for a small temperature range above the glass transition. Clearly, the actual dependence of the shift factors on temperature, shown graphically in Figures 2-8, 2-9 and 2-10 for the three polymers of interest, needs to be used. Therefore, a user subroutine was constructed to replace the default WLF equation in ABAQUS.

The above method has been coded into a FORTRAN program which is presented in Appendix A.

2.4. Results

To assess the validity of the method and of the approximations involved, the residual stresses during cooling from their curing temperatures of BPDA-PDA, HFDA-APBP and DVS-BCB films attached to silicon substrates were calculated. Since the stress state in the films is simple biaxial tension, rectangular strips of the polymer films were considered for the FEM analysis. The thickness of the strip was 20 $\mu$m for BPDA-PDA, 15 $\mu$m for HFDA-APBP and 25 $\mu$m for DVS-BCB. The length of the strip was (arbitrarily) taken to be 20 $\mu$m, and each was constrained on one pair of opposing sides. Also, they were constrained in the thickness direction to produce the desired biaxial stress state, i.e., they were under plane strain conditions. Rigid body motions were excluded. The coefficients of thermal expansion (CTE) of the strips were taken to be those of the polymer films minus the CTE of the silicon substrate. All
material properties used for the analyses are shown in Tables 2-1, and 2-2. Due to the simplicity of the model and of the stress state, a single, eight node isoparametric quadrilateral under plane strain was sufficient to model the structure.

For the viscoelastic analysis, the thermal loading of BPDA-PDA was uniform cooling from 400 °C to 25 °C, and three cooling times of 6 sec, 40 min and 24 hours were considered. The results of the analyses are shown in Figure 2-11. In the same figure, the results of the elastic analysis with a temperature dependent but time independent modulus have also been included for comparison. At ambient, the calculated residual stress in BPDA-PDA is 4.22 MPa for a cooling time of 40 min, which agrees well with the reported experimental value of 4 MPa obtained from curvature measurements [21].

For the viscoelastic analysis of HFDA-APBP, the thermal loading was uniform cooling from 310 °C to 25 °C, and three cooling times of 6 sec, 30 min and 24 hours were considered. The results of the analyses are shown in Figure 2-12. The calculated stress of 34 MPa at ambient for a cooling time of 30 min is within close range of the experimentally obtained value of 37 MPa [22]. The same kind of elastic analysis is included for comparison.

The thermal loading of DVS-BCB was uniform cooling from 250 °C to 25 °C for cooling times of 6 sec, 60 min and 24 hours. The results are shown in Figure 2-13. The calculated stress of 36.5 MPa for the cooling time of 60 min is in good agreement with the experimentally obtained value of 38 MPa [19]. An elastic analysis has also been included in this case.
2.5 Conclusions

A method to obtain the viscoelastic properties of polymer films in a form useful for Finite Element Method (FEM) analysis has been developed. The method consists of the following steps: (i) The variation of the storage modulus with temperature and frequency is obtained from a Dynamic Mechanical Analysis (DMA). (ii) A time-temperature shift is performed on the DMA data to yield the master curve and the variation of the shift factor with temperature. (iii) The Debye equation with a given number of terms is fitted to the master curve. This yields a number of relaxation constants and times which can be used in the FEM analysis performed by ABAQUS. (iv) A user subroutine replaces the default WLF equation in ABAQUS to perform the time-temperature shift.

The method has been used to calculate the variation of stress with temperature, for a number of cooling times, from the curing temperatures, in three polymer films used in electronic packages: BPDA-PDA, HFDA-APBP and DVS-BCB. The calculated stresses agree well with available experimental data. For comparison, the linear elastic analysis overestimates the residual stress in all these polymers.

The material properties obtained from the present method can be useful in Finite Element Analyses to find the stresses in complex structures associated with electronic packages. Such structures can be copper lines embedded in films, copper vias, free edges, etc. Although the method was developed to be used with the FEM code ABAQUS, it is general enough to be adaptable to other FEM codes.
<table>
<thead>
<tr>
<th>Material</th>
<th>Young’s Modulus (GPa)</th>
<th>Poisson’s ratio</th>
<th>CTE (E-6/°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>BPDA-PDA</td>
<td>8.300</td>
<td>0.40</td>
<td>4.2</td>
</tr>
<tr>
<td>HFDA-APBP</td>
<td>2.700</td>
<td>0.40</td>
<td>50</td>
</tr>
<tr>
<td>DVS-BCB</td>
<td>2.600</td>
<td>0.35</td>
<td>52</td>
</tr>
<tr>
<td>Silicon</td>
<td></td>
<td></td>
<td>2.3</td>
</tr>
</tbody>
</table>

Table 2-1. Elastic material properties.
<table>
<thead>
<tr>
<th>j</th>
<th>$E_j=0.911 \text{ GPa}$</th>
<th>$E_j=0.281 \text{ GPa}$</th>
<th>$E_j=0.169 \text{ GPa}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\tau_1$ (sec)</td>
<td>$0.39131E+13$</td>
<td>$0.57919E+15$</td>
<td>$0.64273E+12$</td>
</tr>
<tr>
<td>$\tau_2$ (sec)</td>
<td>$0.24270E+9$</td>
<td>$0.35690E+14$</td>
<td>$0.12409E+7$</td>
</tr>
<tr>
<td>$\tau_3$ (sec)</td>
<td>$0.11131E+6$</td>
<td>$0.33572E+12$</td>
<td>$30.744$</td>
</tr>
<tr>
<td>$\tau_4$ (sec)</td>
<td>$50.264$</td>
<td>$0.10844E+8$</td>
<td>$0.10689E-1$</td>
</tr>
<tr>
<td>$\tau_5$ (sec)</td>
<td>$0.5009E-1$</td>
<td>$0.72281E-5$</td>
<td>$0.65855E-4$</td>
</tr>
<tr>
<td>$\tau_6$ (sec)</td>
<td>$0.16951E-3$</td>
<td>$0.29728E-10$</td>
<td>$0.10812E-5$</td>
</tr>
<tr>
<td>$\tau_7$ (sec)</td>
<td>$0.35556E-6$</td>
<td>$0.59706E-13$</td>
<td>$0.10049E-7$</td>
</tr>
<tr>
<td>$\tau_8$ (sec)</td>
<td>$0.14631E-9$</td>
<td>$0.43016E-15$</td>
<td>$0.47346E-10$</td>
</tr>
<tr>
<td>$\tau_9$ (sec)</td>
<td>$0.31183E-13$</td>
<td>$0.16020E-17$</td>
<td>$0.41944E-14$</td>
</tr>
</tbody>
</table>

Table 2-2. Viscoelastic relaxation constants and times of the three polymers.
Figure 2-1. Chemical structure of (a) BPDA-PDA, (b) HFDA-APBP, and (c) DVS-BCB, repeat units.
Figure 2-2. Dependence of storage modulus of BPDA-PDA on temperature for 0.5 Hz (lower curve), 5 Hz (middle curve) and 20 Hz (upper curve), from DMA analysis.
Figure 2-3. Dependence of storage modulus of HFDA-APBP on temperature for 0.5 Hz (lower curve), 5 Hz (middle curve) and 20 Hz (upper curve), from DMA analysis.
Figure 2-4. Dependence of the storage modulus of DVS-BCB on temperature for 0.5 Hz (lower curve), 5 Hz (middle curve) and 20 Hz (upper curve), from DMA analysis.
Figure 2-5. Dependence of storage modulus of BPDA-PDA on logarithm of frequency at a reference temperature of 220 °C, as obtained from time-temperature superposition of DMA data.
Figure 2-6. Dependence of storage modulus of HFDA-APBP on logarithm of frequency at a reference temperature of 220 °C, as obtained from time-temperature superposition of DMA data.
Figure 2-7. Dependence of storage modulus of DVS-BCB on logarithm of frequency at a reference temperature of 200 °C, as obtained from time-temperature superposition of DMA data.
Figure 2-8. Dependence of logarithm of the shift factor of BPDA-PDA on temperature, for a reference temperature of 220 °C.
Figure 2-9. Dependence of logarithm of the shift factor of HFDA-APBP on temperature, for a reference temperature of 220 °C.
Figure 2-10. Dependence of logarithm of the shift factor of DVS-BCB on temperature, for a reference temperature of 200 °C.
Figure 2-11. Residual stresses in BPDA-PDA as function of temperature, assuming elastic behavior of the polyimide (open diamonds), viscoelastic behavior of the polyimide and cooling times of 6 seconds (circles), 40 minutes (squares) and 24 hours (triangles).
Figure 2-12. Residual stresses in HFDA-APBP as function of temperature, assuming elastic behavior of the polyimide (open diamonds), viscoelastic behavior of the polyimide and cooling times of 6 seconds (circles), 30 minutes (squares) and 24 hours (triangles).
Figure 2-13. Residual stresses in DVS-BCB as function of temperature, assuming elastic behavior of the polymer (open diamonds), viscoelastic behavior of the polymer and cooling times of 6 seconds (circles), 60 minutes (squares) and 24 hours (triangles).
3. ELASTIC-PLASTIC FINITE ELEMENT ANALYSIS OF THE ISLAND BLISTER TEST

3.1. Introduction

The mechanical integrity of many electronic components is determined by the strengths of the interfaces between dissimilar materials, thus, knowledge of the strength is important to the design and evaluation of such structures. One parameter which measures the strength of an interface is the critical value of the static fracture energy release rate or specific fracture energy ($\gamma_a$). It is defined as the energy required to propagate a preexisting crack at the interface over a unit area. For analysis, it is assumed to be the resistive work done by the tractions a short distance ahead of the crack (Figure 3-1a), when slowly released to propagate the crack by that distance (Figure 3-1b), divided by the new crack surface area. To avoid any confusion, by crack surface area we mean the area of an imaginary surface between the two actual crack surfaces, i.e., it is half the total new surface which is created by the propagation of the crack. This work is:

$$\gamma_a = \frac{\partial W_a}{\partial A} = \frac{1}{\Delta A} \int_0^{\Delta A} \int_0^\tau_i \text{du}_i \text{ ds} \quad \text{when } \Delta A \to 0 \quad (3-1)$$

$$i = 1, 2, 3$$

where $W_a$ is the fracture energy, $A$ is the crack surface area, $\tau_i$'s are the tractions on the prospective crack surface and $u_i$'s are the displacements of the crack surface. Note that if $\gamma_a$ is interpreted as surface energy, it includes both surfaces created by the propagation of the crack. In expression (3-1) the summation convention over repeated indices has been used. If the materials
of the system are linearly elastic under small displacements, then Equation (3-1) becomes:

\[
\frac{\partial W_a}{\partial A} = \frac{1}{2} \Delta A \int_0^{\Delta A} \tau_i u_i \, ds \quad \text{when } \Delta A \to 0 \quad (3-2)
\]

\[i = 1, 2, 3\]

The specific fracture energy, \(\gamma_a\), is assumed to be the sum of three components which are not always independent. The normal mode (Mode I or peeling mode), \(\gamma_I\), is produced by the work of the tractions normal to the interface. The shear mode (Mode II or shearing mode), \(\gamma_{II}\), is produced by the work of the shear tractions parallel to the interface and parallel to the direction of crack propagation. The antiplane mode (Mode III or tearing mode), \(\gamma_{III}\), is produced by the work of the shear tractions parallel to the interface and perpendicular to the direction of crack propagation. If a coordinate system is set up at a position on the crack front such that the 1 axis is perpendicular to the crack surface, the 2 axis is parallel to the crack surface and to the direction of crack propagation, and the tractions and displacements of Equation (3-1) are referred to it, then the three terms of the equation correspond to the three modes.

The external force does work on the system; this is converted to fracture energy, is stored as elastic energy, and is absorbed by plastic deformation. Thus:

\[
\frac{\partial W_{ex}}{\partial A} + \frac{\partial W_a}{\partial A} + \frac{\partial W_{el}}{\partial A} + \frac{\partial W_p}{\partial A} = 0 \quad (3-3)
\]
where $W_{ex}$ is the work from the external force, $W_a$ is the fracture energy or fracture surface energy, $W_{el}$ is the stored elastic energy, $W_p$ is the plastic work, and $A$ is the crack surface area.

Numerous methods to measure the specific fracture energy of interfaces have been proposed [26], and a few are widely used. Among these are the scratch test [27], the pull test [28], the peel test [29], and the blister test [30] and all have variations; each has certain advantages and disadvantages. In the scratch test, a blunt stylus is dragged across the surface of a coated substrate. The load applied to the stylus is increased until the coating delaminates. The load at which this occurs is a measure of the adhesion, but determining when this occurs and how to interpret the results are problematical [26, 31].

In the pull test a stud adhered to the film which is bonded to the substrate is pulled until failure occurs at the film-substrate interface. The test is sensitive to misalignment and to the thickness of the adhesive layer. The geometry produces a complicated stress state which is difficult to control and the adhesive used to attach the stud must be stronger than the interface and have no adverse interaction with it [26, 32].

All of the many different peel tests pull an adhering film from a substrate. If the film is strong and perfectly elastic, the test is easy to perform and to analyze. If the film is strong and ductile, extensive plasticity arises and the analysis becomes more complicated. If the film is weak or brittle, it may break before the interface fails; then the peel test is not possible [33-35].

The blister test was developed to avoid some of the problems of the peel tests. In it, a circular hole is formed through the substrate. Pressure is
applied to the exposed back surface of the film bridging the hole until interfacial failure occurs around the bonded periphery. In this test however, it is difficult to control the pressure and weak films may rupture before any interfacial failure occurs.

A variation is the island blister test (IBT) [11]. In it, a small substrate island at the center of the blister remains bonded to the film (Figure 3-2). Pressure is applied to the unbonded film annulus until failure occurs around the periphery of the island. By altering the ratio of the island diameter to the outer diameter of the annulus, interfacial failure can be produced without rupture in most films. Further, the plastic work factor is smaller than most other peel tests because such a small volume of film is involved.

In the IBT, the pressure at which peeling at the island occurs is measured. To convert this to interfacial specific fracture energy, a stress analysis is needed. Allen and Senturia formulated one [11], using the following assumptions: i) the film remains elastic during deformation, ii) the film stress does not vary from the initial residual stress and iii) the film behaves as an ideal membrane with no bending stiffness. The solution identified the total debonding energy as a function of the size of the island and the pressure on the film. Later, Lai and Dillard [36], improved the analysis by eliminating the assumptions about bending stiffness and residual stress. In fact, they showed that high stresses can develop in the film near the island but they did no plasticity analysis nor did they try to discriminate among the three modes of crack propagation.

In this work, we present Finite Element Method modeling of the Island Blister Test which addresses the plasticity in the film and separates the
normal and shear components of the specific fracture energy. Some results of
the application of the model to the interface between Ultradel 4212, a
polyimide (Figure 2-1b), and a chromium coated silicon substrate, done by
Sikorski [37, 38], are given.

3.2. Finite Element Method Modeling

Using axial symmetry, the island blister geometry shown in Figure 3-2
was modeled by one row of one thousand axisymmetric eight-noded
quadrilaterals. This choice was justified by the fact that the average ratio of
film thickness to blister radius was 1 to 1000. Also, as shown later, this type of
mesh produces results which agree well with experimental deflection
measurements. The substrate was assumed to be perfectly rigid and all nodes
over the island were rigidly pinned to it. This is justified by the large
difference between the moduli of the polymer films and the metals or silicon.
The nodes along the axis of symmetry in the middle and the nodes on the
right edge of the film were pinned in the radial direction. The bottom node of
the right edge was also pinned in the meridian direction to prevent
separation of the film from the substrate after the application of pressure. The
large deflection elastic-plastic analysis was performed by ABAQUS [25].

To perform a valid analysis, the behavior of the film must be known
with reasonable accuracy. The plastic part of the behavior of Ultradel 4212 was
obtained from tensile experiments at different strain rates, as shown in Figure
3-3. The incremental theory of plasticity and the Von Mises yield criterion
were assumed to describe the yielding behavior of the polymer under a
generalized stress state. The strain rate dependence of the plastic behavior of
the polymer, i.e., its viscoplastic behavior, was described by the following equation, available in ABAQUS:

\[ \dot{\varepsilon}_{pl} = D \left( \frac{\sigma_{pl}}{\sigma_o} - 1 \right)^b \]  \hspace{1cm} (3-4)

where \( \dot{\varepsilon}_{pl} \) is the equivalent plastic strain rate, \( \sigma_{pl} \) is the equivalent plastic stress, \( \sigma_o \) is the static yield stress and D and b are material constants. In the case under consideration here, D=5.63 and b=2.20, from Figure 3-3. To a first approximation, the viscoelastic relaxation of the film on the time scale of the experiment was taken into account by using for the elastic part of its behavior the 5-min relaxation modulus (2.3 GPa), at room temperature, instead of its glassy modulus (2.7 GPa) [39], [40]. It should be noted however, that this approximation does not constitute a true viscoelastic analysis of the debonding processes, which was neglected. The Poisson’s ratio was taken as 0.40.

A variant of the modified crack closure method was used to obtain the different modes of the specific fracture energy. In the original method [41], the specific fracture energy associated with the propagation of a crack in a linear elastic system, given by Equation (3-2), was calculated from the nodal forces and displacements of four noded quadrilaterals. The method has been extended to two dimensional and three dimensional higher order elements by a number of authors [42 - 44]. In the case of eight noded quadrilaterals under plane stress or plane strain conditions, the Mode I and Mode II components of the specific fracture energy of an interfacial crack between a linear elastic material and a rigid substrate are given by the following formulas referring to Figure 3-4:
\[ \gamma_I = \frac{1}{2 \Delta e t} ( F_{z,i} u_{z,i+2} + F_{z,i+1} u_{z,i+3} ) \]  

\[ \gamma_{II} = \frac{1}{2 \Delta e t} ( F_{x,i} u_{x,i+2} + F_{x,i+1} u_{x,i+3} ) \]  

In the above formulas, \( \gamma_I \) and \( \gamma_{II} \) are the Mode I and Mode II components of the specific fracture energy, respectively; \( F_{z,i} \) and \( F_{z,i+1} \) are the nodal forces of nodes \( i \) and \( i+1 \) along the \( z \) axis, respectively; \( u_{z,i+2} \) and \( u_{z,i+3} \) are the displacement of nodes \( i+2 \) and \( i+3 \) along the \( z \) axis; \( F_{x,i} \) and \( F_{x,i+1} \) are the nodal forces of nodes \( i \) and \( i+1 \) along the \( x \) axis; \( u_{x,i+2} \) and \( u_{x,i+3} \) are the displacement of nodes \( i+2 \) and \( i+3 \) along the \( x \) axis, respectively; \( \Delta e \) is the length of each of the two adjacent elements and \( t \) is their thickness, which is usually taken as one. The modified crack closure method is also very accurate for relatively coarse meshes. This derives from the fact that the Finite Element Method can calculate nodal forces and nodal displacements much more accurately than stresses and strains [7-9].

Equations (3-5) and (3-6) are not directly applicable to the case here because instead of plane stress or plane strain conditions we have axisymmetric geometry. In FEM, the nodal forces for axisymmetric geometries are the total forces on a complete circle, with the center as the axis of symmetry and the radius as the distance of the node from the axis. A second reason is that the film behavior is, in general, non-linear and so is the relation between nodal forces and displacements. To account for these differences, the following method was used to calculate the Mode I and Mode II components of the interface specific fracture energy, given in general by Equation (3-1). The first step of the FEM model loads the film with the
residual stresses. The next step loads the film with the critical pressure in ten increments. The reaction forces of nodes $i$ and $i+1$ of element $e_j$, which were initially pinned to the substrate were recorded at the critical pressure value. These two nodes were unpinned from the substrate and the reaction forces on them were relaxed equally in each of ten increments. The displacement of each node in each of these increments was recorded. The relations are:

$$\gamma_1 = \frac{1}{\pi \left( (r_i + \Delta e)^2 - r_i^2 \right)} \sum_{j=1}^{10} \left( F_{z,i}^j \Delta u_{z,i}^j + F_{z,i+1}^j \Delta u_{z,i+1}^j \right) \quad (3-7)$$

and

$$\gamma_{11} = \frac{1}{\pi \left( (r_i + \Delta e)^2 - r_i^2 \right)} \sum_{j=1}^{10} \left( F_{r,i}^j \Delta u_{r,i}^j + F_{r,i+1}^j \Delta u_{r,i+1}^j \right) \quad (3-8)$$

where the superscript $j$ refers to the increment number, and $r_i$ is the distance of the crack tip from the center of the island.

To verify that the mesh was adequate for our purposes, convergence studies were performed with three meshes of 500, 1000 and 1500 elements. The results showed that the mesh with 1000 elements was efficient and economic for the analysis. The measured deflection of a 11 $\mu$m thick film with an inner radius of 1 mm and outer radius of 13 mm was compared to the deflection calculated from the FEM analysis. The results in Figure 3-5 show good agreement with the experimental data.

Appendix B, contains a listing of an ABAQUS input file for the elasto-plastic analysis of the Island Blister Test, as described in this section.
3.3 Results and Discussion

Based on the Island Blister Test model of this work, Sikorski [37, 38], analyzed his experimental results of an Ultradel 4212 film attached to a chromium coated silicon wafer. Figure 3-6 shows that for the system under consideration, stresses greater than the yield point are developed near the island during debonding. This agrees with predictions made by Liechti et al. based on nonlinear Von Karman plate theory [45]. The maximum Von Mises stress in the film does not change significantly as the crack propagates inward on the island (Table 3-1a) but, it is greater for thinner films (Table 3-1b). (Note that in Table 3-1b, the 50 μm point is not experimental. It was generated by FEM analysis and a trial and error procedure; a number of different critical pressures, p_c's, were assumed to be applied to a 50 μm thick film, and the corresponding γ_t's were calculated from FEM analyses, as described in the previous section. The maximum Von Mises stress corresponding to the 50 μm point in Table 3-1b was obtained from the FEM analysis for the p_c which gave a γ_t of 60 J/m². As will be shown next, this is the fracture criterion for the system). This suggests that above a critical thickness, no plastic work is done and the system remains in the elastic regime. All the films which were tested were below this critical thickness so plasticity was important and had to be considered for accurate calculation of the specific fracture energy [35], [45, 46].

To investigate the effect of film thickness on the components of the specific fracture energy and the specific plastic work, i.e., the plastic work dissipated per unit new crack surface area, three specimens of thicknesses 5.6, 11, and 19 μm were tested. The results are shown in Table 3-2 and in Figure 3-7. It is seen that the Mode I component of the specific fracture energy (γ_t) is

51
independent of thickness and inner radius except for small values of the latter. The apparent increase in $\gamma_I$ for smaller inner radii is an artifact of testing, as discussed further in the next paragraph. The Mode II component of the specific fracture energy ($\gamma_{II}$) increases with increasing thickness. The peel angle decreases with increasing thickness, as shown in Figure 3-8. There is no apparent trend in the dependence of the specific plastic work on film thickness.

Because reproducibility is an important requirement of any test, it was investigated by fabricating and testing a series of samples which were nominally the same. Data from samples from five different wafers (11.0 $\mu$m Ultradel 4212 on Cr) are compared in Figure 3-9. The data are divided into two sets. Set A was the first one to be tested. Notice that for this set, the value of $\gamma_I$ increases as the test progresses and $r_i$ decreases, although the agreement from sample to sample is good. Set B was built at the same time as set A but was tested following more rigorous standardization of the testing technique. For this set, the rate of pressure increase was controlled at 11.5 Pa/sec (0.10 psi/min) and crack propagation was defined as the crack front moving 1 $\mu$m over a period of 15 seconds. Using these criteria, much more consistent data were obtained. To ensure that the improvement was due to the testing technique and was not sample related, one of the samples of set A (3B3) was tested again using the standardized procedure. The resulting point fell well within set B. We conclude that this standardization of testing procedure provides more consistent data and removes the dependency of $\gamma_I$ on $r_i$.

The results of this work show that the Mode I component of the specific fracture energy ($\gamma_I$) is essentially independent of thickness and inner radius, i.e. independent of geometry. However, the Mode II component ($\gamma_{II}$)
depends on the film thickness as does the total specific fracture energy \( (\gamma_a) \), which is the sum of the two components. Thus, it is apparent that interfacial crack propagation occurs in this system when the Mode I component reaches a critical value. The value of the Mode II component and of the total specific fracture energy is dependent upon the geometry, the material properties and the loading at the moment when the Mode I component reaches its critical value. Therefore, unlike the Mode I component, they are not material constants and they cannot uniquely characterize the strength of the interface. Essentially, this result is the same as that reached by a number of authors in the past [47-49]; that fracture is determined by both, the total specific fracture energy and the mode mixity.

3.4 Conclusions

A large deflection elastic-plastic finite element modeling of the island blister test, is presented. The model can separate the Mode I and Mode II components of the specific fracture energy and the plastic work. Based on this model, Sikorski [37, 38], calculated the Mode I and Mode II components of the specific fracture energy and the specific plastic work in a polyimide-on-metal system (Ultradel 4212 on Cr). Several observations for this system are:

1. Accounting for plasticity eliminates the thickness dependence of the Mode I component of the specific fracture energy, \( \gamma_I \).

2. Crack propagation occurs when \( \gamma_I \) reaches a critical value, about 60 J/m\(^2\). This is the fracture criterion for this system.
3. For small values of the inner radius \((r_i)\), \(\gamma_I\) increases with decreasing \(r_i\). This is an artifact of testing, since these data were generated before the testing procedure was standardized.

4. Reproducibility of the test is enhanced by controlling the rate of pressurization (11.5 Pa/sec) and by precise definition of what constitutes crack-front motion.
Table 3-1. (a) Maximum Von Mises stress as function of inner radius for a 11 μm thick film. (b) Maximum Von Mises stress as function of film thickness for an inner radius of 0.39 mm. Note that the 50 μm point is not experimental but was generated, for comparison, by assuming that $\gamma_l=60$ J/m$^2$ and determining the associated $p_c$ from FEM analysis, by trial and error (see text for more details).
<table>
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<tr>
<th>Film Thickness (μm)</th>
<th>Inner Radius (mm)</th>
<th>$\gamma_I$ (J/m²)</th>
<th>$\gamma_{II}$ (J/m²)</th>
<th>$\gamma_a$ (J/m²)</th>
<th>Specific Plastic Work (J/m²)</th>
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<td>24</td>
<td>85</td>
<td>150</td>
</tr>
</tbody>
</table>

Table 3-2. Components of specific fracture energy and specific plastic work for three Ultradel 4212 films on Cr.
Figure 3-1. Crack front position before (a) and after (b) crack propagation.
a) TOP VIEW

![Diagram of top view showing Island and Polymer Film]

b) CROSS SECTION A-A

![Diagram of cross section showing Si Substrate and Film]

Figure 3-2. Island Blister Test (IBT) geometry.
Figure 3-3. Tensile constitutive behavior of Ultradel 4212, as function of strain rate.
Figure 3-4. Schematic of two crack tip elements in their undeformed state (a), and their deformed state before (b) and after (c) crack propagation.
Figure 3-5. Deflection of half of the island blister membrane as obtained by experimental measurements (squares) and the FEM solution (solid line) [37].
Figure 3-6. Von Mises stress in the film as function of radial position [37].
Figure 3-7. Mode I specific fracture energy as function of inner radius, for three films of different thicknesses [37].
Figure 3-8. Peel angle as function of film thickness [37]. Note that the 50 μm (solid) point is not experimental but was generated, for comparison, by assuming that $\gamma_f = 60$ J/m$^2$ and determining the associated $p_c$ from FEM analysis, by trial and error (see text for more details).
Figure 3-9. Mode I specific fracture energy as function of inner radius, for five different films having the same thickness [37].
4. AN EDGE CRACK BETWEEN A FILM AND A RIGID SUBSTRATE

4.1 Introduction

The first to find the form of the crack tip singularity, for the problem of an interface crack between two dissimilar materials (Figure 4-1a), was Williams [50]. His analysis is based on asymptotic expansion of a stress function, such that it will satisfy the local boundary conditions near the crack tip. Recently, Rice [51], has provided a more complete solution of this problem, based on Muskhelishvilli's [52] complex variable method for the solution of plane linear elastic problems. Later, Cherepanov [53], England [54], Rice and Sih [55], and Suo and Hutchinson [56], provided solutions for specific problems. Williams' elastic solution has the following problem; it predicts the existence of an oscillating stress singularity at the crack tip. The oscillating stress singularity leads to oscillating crack opening displacements near the crack tip, which, in turn, lead to interpenetration of the crack faces. This is an unacceptable result physically, and it has been recognized as such by all of the above authors.

Another problem of the linear elastic solution is that Irwin's crack closure integral [4], which is the basis for the calculation of the energy release rate of cracks in homogeneous materials, cannot be applied in the case of interfacial cracks. The reason is that both stresses and crack opening displacements oscillate in such a way that their product does not converge. To overcome this, Salganik and Malyshov [57], used only the non-oscillating part of the solution to calculate the energy release rate. The same assumption was used by Hutchinson et al. [58], to calculate the energy release rate in various interfacial problems. Alternatively, one can use the J-integral for such
calculations, again provided that the oscillating parts of the elastic solution are removed.

To avoid such an unacceptable solution Comninou [59], and other authors, postulated that a contact zone between the crack faces must exist near the crack tip. To prove this, she attempted to solve the problem of a crack between two dissimilar materials, subjected to a remote uniform tensile stress. Using the technique of distribution of dislocations, and assuming that the distribution of edge dislocations is non-singular, she expressed the problem in terms of two integral equations. Based on these, she found a non-oscillating shear singularity and a non-oscillating compressive singularity, at the crack tip. She also provided a numerical solution of the integral equations.

The treatment of the problem by Comninou, has several flaws. First, the assumption that the distribution of edge dislocations is non-singular leads to a degenerate system of two integral equations. In it, instead of two unknown functions, i.e., the distributions of edge and shear dislocations which satisfy the boundary conditions of the problem, the author obtained an unknown function, the distribution of edge dislocations, and an unknown constant, the length of the contact zone. This system fails in the limit, when the elastic constants of the two materials approach each other. Also, the numerical solution was solved in an indirect way and has obvious convergence problems, i.e., it is in the limit of numerical accuracy, no complete solution could be obtained, and a small change in Dundur’s β parameter leads to a very large change in the size of the contact zone. Further, the predicted compressive singularity is clearly counterintuitive and contradicts the elastic solution of the same problem by Rice and Sih (which
also cannot be accurate since it contains the oscillating singularity). Thus the conclusion can be drawn that the small displacement theory of elasticity cannot properly handle the form of the stress field in the vicinity of the crack tip. Strong supporting evidence is found in the work of Knowles and Sternberg [60]. Using the large displacement theory of elasticity, they showed that for incompressible materials under plane stress, no oscillating singularity exists, and no contact between the crack faces is predicted. Since the small displacement theory is inadequate, an alternative approach was sought.

4.2 The edge crack

One way to overcome the difficulty associated with the elastic solutions is to use a numerical method to directly calculate the energy release rate, without having to calculate the exact form of the stress and strain fields in the vicinity of the crack tip. The Finite Element Method and the modified crack closure method are the techniques which are used here, to obtain the energy release rate of an interface crack between a film and a rigid substrate (Figure 4-1b). Plane strain conditions are assumed.

It is known that for a large crack, the fracture energy release rate is:

\[ G = \frac{\sigma_0^2 t (1-v^2)}{2 E} \]  

(4-1)

where \( \sigma_0 \) is the biaxial residual stress, \( t \) is the thickness of the film, \( E \) is its modulus, and \( v \) is its Poisson's ratio. The validity of Equation (4-1) can be proven, if we refer to Figure 4-2. A crack advance of distance \( dx \) is equivalent to the crack remaining stationary while a strip of material of width \( dx \) is removed from the right edge of the model, and added to the left edge. The energy released in the model, by the removal of the strip at the right edge
(divided by dx) is given by (4-1), keeping in mind that no energy is released in the thickness direction, because of the plane strain conditions. The addition of the strip to the left edge does not add any energy to the system, because for a long crack, the left edge is stress free in the x direction. Equation (4-1) can be used to check the accuracy of the finite element calculations. (The same result has also been obtained by Suo and Hutchinson [56], from a more general treatment.)

An example of an FEM mesh used to model the situation is shown in Figure 4-3 and the mesh near the crack tip, after deformation, is shown in Figure 4-4. The results of the analysis are expressed in a dimensionless form. The form is suggested by Equation (4-1):

\[ \frac{GE}{\sigma_0^2 t (1-v^2)} \]  \hspace{1cm} (4-2)

Figure 4-5, shows the dependence of this quantity on the crack length normalized by the thickness. The results of the FEM calculations are in excellent agreement with Equation (4-1), for large cracks. i.e., \( L/t > 1 \) and mesh convergence studies showed the ratio of Mode I to Mode II components increases with increasing mesh refinement. Also, for a given mesh size, decreasing the Poisson’s ratio increases the ratio. Recently, more detailed calculations by Shaffer [61] suggest that the Mode II contribution on the total energy release rate shown in Figure 4-5, is negligible, i.e., the loading at the crack tip is pure Mode I.

4.3 Application

The results of the previous section can be used to calculate the critical thickness of a polymer film attached to a rigid substrate, above which loss of
adhesion is likely to occur. This critical thickness is calculated from Equation (4-1), by substituting $\gamma_l$ for $G$, and solving for the thickness:

$$t_{crit} = \frac{2 \gamma_l E}{\sigma_0^2 (1-\nu^2)} \quad (4-3)$$

where $\gamma_l$ is the Mode I component of the specific fracture energy. This is measured in the Island Blister Test, as described in the previous chapter. The critical thickness will be a function of temperature, since all the quantities on the right hand side of Equation (4-3) are temperature sensitive. (In using Equation (4-3), the assumption was made that the loading at the crack tip is purely Mode I [61].) An example is given in Figure 4-6, where the critical thickness of Ultradel 4212 on chromium has been plotted as function of temperature. The residual stress as a function of temperature has been taken from Chapter 2, and its specific fracture energy on chromium, from Chapter 3. Because of a lack of data, the specific fracture energy was assumed to be independent of temperature, hence the plot in Figure 4-6 should not be interpreted literally.

Evidence for the validity of the analysis can be found in the work of Shaffer [61]. An 83 $\mu$m thick Ultradel 4212 film on Cr did not debond at a temperature of -196 °C, as expected from the analysis which predicts a critical thickness of 103 $\mu$m at this temperature (see Figure 4-6). Also, applying the edge delamination test [61, 62] for measurement of adhesion, he found the adhesion of Cycloprene on an aluminum substrate to be about 4.66 J/m$^2$, at room temperature. From Equation (4-3), we find the critical thickness of this system to be 19 $\mu$m, where the Cycloprene modulus, residual stress and Poisson's ratio at room temperature, were taken from Chapter 2. His experiments [61] have shown that delaminations are indeed produced in
CycloTene films thicker than 19 μm, on aluminum, but not in films thinner than this value.

4.4 Conclusions

Unlike the case of a crack in a homogeneous material, the analysis of a crack at the interface between two dissimilar materials has encountered many problems. Currently, the validity of the linear elastic analyses remain questionable. We attempt to overcome these difficulties by using a Finite Element Analysis and the modified crack closure method. This avoids the necessity of calculating stress and displacement fields in the vicinity of the crack tip. The results show good agreement with analytical solutions, in the limiting case of long cracks. The problem of Mode mixity is addressed in detail by Shaffer [61], who concluded that the Mode II component is negligible. The analysis identifies the critical thickness above which loss of adhesion between a compliant material and a rigid substrate will occur, using a simple functional parameter. The latter was used to calculate the critical thickness, as a function of temperature, for Ultradel 4212 film attached to a chromium coated rigid substrate and for CycloTene film attached to an aluminum coated silicon substrate. The experimental results of Shaffer [61], show reasonable agreement with the present analysis.
Figure 4-1. a) An interface crack between two dissimilar materials and b) an interface crack between a material and a rigid substrate.
Figure 4-2. Schematic of crack propagation, as removal and addition of material.
Figure 4-3. Example of FEM mesh used for the calculation of the energy release rate of an interface crack.
Figure 4-4. Detail of deformed FEM mesh in the vicinity of the crack tip.
Figure 4-5. Energy release rate of an interface crack between a film and a rigid substrate, as function of crack length normalized by the thickness.
Figure 4-6. Critical thickness for loss of adhesion, of an Ultradel 4212 film on Chromium, as function of temperature.
5. THE THROUGH-THICKNESS CRACKS

5.1 Introduction

The problem of through-thickness cracking of a film attached to a substrate has been addressed by a number of authors. Gecit [63] analyzed the problem of a single crack oriented perpendicular to the film plane, with length smaller than or equal to the thickness of the film (Figure 5-1a). Assuming plane strain conditions and using a semi-analytical procedure, he calculated the stress intensity factors for a number of combinations of film and substrate properties. Lu and Erdogan [64, 65], expanded Gecit's method to include more material combinations and to find solutions to a greater number of problems. Using a dislocation formulation, Civilek [66], developed a method for the case of one or more cracks in a finite strip of a single material. Beuth [67], summarized the results of previous work and presented the complete solution procedures over the full range of elastic mismatches between film and substrate. Since these cannot yield exact analytical solutions, approximate formulas and information on their applicability were given for relevant practical problems. These authors provide stress intensity factors for partially through-thickness cracks propagating toward the substrate. As a side result, they also give the average stress intensity factor for cracks channeling though the film but they cannot provide the full solution to the problem, since its three dimensional nature renders such plane analyses invalid.

Nakamura and Kamath [68], used a 3-D FEM analysis and the J-integral method, to calculate the stress intensity factor along the edge of a finite through-thickness crack in a film attached to a rigid substrate (a channeling crack) (see Figure 5-1b). They showed the average stress intensity factor along the edge of the crack is independent of its length, if it is more than twice the
film thickness. Also, they found the stress intensity factor is a function of distance from the substrate: it increases as the free surface is approached. When film-substrate adhesion around the vicinity of the crack is lost, the average stress intensity factor of the crack increases. Their work presented no experimental results to verify their predictions; it is very intensive computationally, requiring between 5000 and 7000 eight-noded 3-D brick elements.

Instead of the J-integral approach here we use the FEM and the modified crack closure method to calculate the stress intensity factors for an array of colinear through-thickness cracks, in a film perfectly attached to a rigid substrate. (The single crack is the special case when the cracks are far apart.) Also, changes in the shape of the three dimensional crack front before continuous crack propagation is addressed by an approximate method. Experiments on thin film structures containing fabricated cracks, are used to check the validity of the analysis.

5.2 FEM Analysis of a system of colinear cracks

The array of colinear cracks in a film attached to a rigid substrate is analyzed first. A top view of two of the cracks is shown in Figure 5-2 and one crack is in Figure 5-3. In these figures, L is the crack length, t is the film thickness and b is half the distance between the tips of the cracks. The ratio of length to film thickness was chosen as two, because for any greater ratio the stress intensity along the crack front is independent of it [68]. The periodic boundary conditions and the symmetry of the model make appropriate one fourth of the three dimensional geometry, Figure 5-4; a typical FEM mesh, containing 912 twenty-noded brick elements, is also shown. The bottom
nodes are pinned in all three directions. All side nodes, except those of the crack surfaces, are pinned in the 1 and 2 directions. The loading is an equibiaxial residual stress and Figure 5-5 shows the displaced mesh.

Since the conditions in a three dimensional geometry are neither plane stress nor strain, the appropriate parameter to describe the stress intensity along the crack front is the energy release rate. For generality, the following dimensionless parameter is used:

\[
\frac{G_I E}{\sigma_0^2 t (1-v^2)}
\]

(5-1)

where, \(G_I\) is the Mode I energy release rate, \(E\) is the Young's modulus of the film, \(v\) is the Poisson's ratio of the film, \(t\) is the film thickness, and \(\sigma_0\) is the biaxial residual stress. The dependence of this parameter on the ratio of half-distance between cracks to film thickness, \((b/t)\), and on the ratio of the depth of the crack front to the film thickness, \((z/t)\), is shown in Figure 5-6. It is apparent that the interaction becomes significant for distances between crack tips less than one film thickness; this amplifies the energy release rate along the crack front of both cracks to levels higher than in a single crack. The dependence of the amplification factor on distance between crack tips (normalized by the thickness) has been expressed in a closed functional form by curve fitting. The resulting expression is:

\[
\frac{G_I(b/t)}{G_I(oo)} = f(b/t) = 1 + 0.0515 \left(\frac{b}{t}\right)^{-1.058}
\]

(5-2)

This is shown in Figure 5-7; the expression has an error of less than 5% for any value of the ratio \(b/t\).
To check the accuracy of the FEM results, we compared them to existing analyses, for the case of a single crack. Nakamura and Kamath [68], expressed their results in terms of the stress intensity factor, $K_I$, instead of the energy release rate, assuming plane strain conditions. We have expressed our results in the same form, in Figure 5-8. For plane strain conditions, the relation between $G_I$ and $K_I$ is:

$$G_I = \frac{K_I^2 (1-v^2)}{E} \quad (5-3)$$

and for plane stress conditions:

$$G_I = \frac{K_I^2}{E} \quad (5-4)$$

In both cases, the appropriate dimensionless parameter is:

$$\frac{K_I}{\sigma_0 \sqrt{l}} \quad (5-5)$$

The maximum of the curve of Figure 5-8 is 1.29 for plane strain, and differs by less than 1% from the value of 1.30 given by the authors [68]. Also, the average of 1.03 is identical to that of the authors [68] and differs by less than 1% from the value of 1.04, given by Beuth [67].

5.3 The curved three-dimensional crack front

The analysis of the previous section was done for a straight crack front, perpendicular to the plane of the film. In the literature, it is assumed that such will start propagating when its average energy release rate exceeds the critical energy release rate of the material but no experimental verifications have been presented. Further, the possibility of the crack front assuming a
different shape of lower energy release rate, thus arresting further crack propagation, is not mentioned nor analyzed. We address this issue next.

Thermodynamics teaches that a system under constant temperature and pressure will assume a state which minimizes its free energy. The free energy of a cracked elastic system is the sum of its potential energy and the crack surface energy. Applying this to the three dimensional crack front, we expect that a change in its shape should occur, when the energy release rate at any location along the crack front exceeds the critical energy release rate of the material. The new stable shape should be such that its energy release rate, at any position on it, is lower than the critical energy release rate of the material. If such a shape does not exist, catastrophic crack propagation will occur.

There exist no analytical solutions for the energy release rate of a three dimensional crack front as function of its shape. Even in the simplest case of a straight front, the analysis is numerical. The analysis of the problem is very tedious and very intensive computationally, because of the infinite number of possible shapes that the crack front can assume, and the extreme distortion of most FEM meshes which will model them. Because of this, we do not solve the problem exactly. Instead, we use a simple method to obtain an approximate estimate of the minimum energy release rate crack front profile, and a value for its energy release rate.

The release rate is in units of energy per unit area: an energy flux. We assume the crack front is subjected to two energy fluxes: a horizontal one, $G_h$, which drives it forward, and a vertical one, $G_v$, which drives it toward the substrate. By vertical we mean the direction perpendicular to the plane of the film and by horizontal the direction parallel to the plane of the film.
Referring to Figure 5-9, the total energy release rate of a small part of the crack front is related to $G_h$ and $G_v$ by the following equation:

$$G \, ds \, dl = G_h \, dz \, dl_x + G_v \, dx \, dl_z$$  \hspace{1cm} (5-6)

or

$$G = G_h \, \sin^2 \theta + G_v \, \cos^2 \theta$$  \hspace{1cm} (5-7)

where:

$$\tan \theta = \frac{dz}{dx}$$  \hspace{1cm} (5-8)

Both, $G_h$ and $G_v$ depend on the position along the crack front and the crack front shape. The two quantities can be found rigorously from numerical analysis, but as mentioned, this is tedious and time consuming. To simplify, we assume they are given by the following equations:

$$G_h = G_{sf} \, \sin \theta$$  \hspace{1cm} (5-9)

and

$$G_v = G_{pc} \, \cos \theta$$  \hspace{1cm} (5-10)

where $G_{sf}$ is the release rate of a straight vertical front (see Figure 5-6) and $G_{pc}$ is the release rate of a partial through-thickness crack, which can be found in Gecit [63]. The two release rates have been plotted, as functions of $z/t$, in Figure 5-11. They provide the dependence of $G_h$ and $G_v$ on $z/t$, while the trigonometric functions provide the dependence on crack front shape. (This particular choice of trigonometric functions is justified by the fact that for a vertical front ($\theta=90$), $G_h$ should be equal to $G_{sf}$, and for a horizontal front
\( (\theta=0), \ G_v \) should be equal to \( G_{pc} \). The experimental results of the next section will give us a measure of the validity of Equations (5-9) and (5-10).

Substituting Equations (5-9) and (5-10) into Equation (5-7), we get:

\[
G = G_{sf} \sin^3 \theta + G_{pc} \cos^3 \theta \tag{5-11}
\]

The extremum for \( G \), with respect to the angle, exists when its first derivative with respect to the angle becomes zero. Differentiating (5-11) we obtain:

\[
\frac{\partial G}{\partial \theta} = 3 \sin \theta \cos \theta \left( G_{sf} \sin \theta - G_{pc} \cos \theta \right) \tag{5-12}
\]

which is zero when:

\[
G_{sf} \sin \theta - G_{pc} \cos \theta = 0 \tag{5-13}
\]

for any angle greater than zero and smaller than ninety degrees. The extremum is a minimum since the second derivative:

\[
\frac{\partial^2 G}{\partial \theta^2} = 3 \left( \cos^2 \theta - \sin^2 \theta \right) \left( G_{sf} \sin \theta - G_{pc} \cos \theta \right) + 3 \sin \theta \cos \theta \left( G_{sf} \cos \theta + G_{pc} \sin \theta \right) \tag{5-14}
\]

is always positive.

Equation (5-13) can be re-written as:

\[
\frac{G_{pc}}{G_{sf}} = \tan \theta = \frac{dz}{dx} \tag{5-15}
\]

Equation (5-15) can be solved numerically to obtain the minimum energy release rate crack front shape, which is shown in Figure 5-10. The energy release rate of that crack front, as a function of distance from the
substrate, is shown in Figure 5-11. In the Figure, $G_{pc}$ and $G_{sf}$ have also been plotted, for comparison.

### 5.4 Application to high density interconnects

The results show that for a crack with a vertical front, the maximum energy release rate is located at a position slightly below the film surface, and is:

$$G_{max}^{vf} = \frac{1.66 \sigma_0^2 t (1-\nu^2)}{E} \quad (5-16)$$

For the minimum $G$ crack front profile, the maximum energy release rate is located at $z/t=0.6$, and is:

$$G_{max}^{cf} = \frac{0.99 \sigma_0^2 t (1-\nu^2)}{E} \quad (5-17)$$

As already discussed, the initially vertical crack front will change in profile when its maximum energy release rate is greater than the critical energy release rate of the material ($G_{crit}$). The curved front will propagate catastrophically when its maximum energy release rate is greater than $G_{crit}$. These events occur at the following two critical thicknesses, obtained from Equations (5-16) and (5-17), by equating the right hand side to $G_{crit}$:

$$t_{cf}^{vf} = \frac{G_{crit} E}{1.66 \sigma_0^2 (1-\nu^2)} \quad (5-18)$$

and

$$t_{cf}^{cf} = \frac{G_{crit} E}{0.99 \sigma_0^2 (1-\nu^2)} \quad (5-19)$$
The critical thicknesses are functions of the temperature, since the residual stress $\sigma_0$, the critical energy release rate $G_{\text{crit}}$, the Young's modulus $E$, and the Poisson's ratio $\nu$, are functions of temperature. The critical thicknesses of three polymer films used in HDI's, DuPont's BPDA-PDA, Amoco's Ultradel 4212, and Dow's Cyclotene, are shown in Figures 5-12 and 5-13. To obtain these thicknesses, we used the residual stresses obtained in Chapter 2, and the fracture toughness values in Table 5-1. (The experimental procedure for the fracture toughness measurements is described in Appendix C.)

To check the validity of the analysis and of the assumptions of this work, a number of experiments were performed. In these, slits representing cracks from 50 to 1000 $\mu$m long, were etched in Cyclotene films spun-cast on chromium coated silicon wafers. According to Figures 5-12 and 5-13, at room temperature (20°C), a crack front profile change should occur above a thickness of 26 $\mu$m. Catastrophic crack propagation should occur above a thickness of 43 $\mu$m.

The results of the experiments are tabulated in Table 5-2. It was observed that the crack front profile change occurred in all films thicker than 25 $\mu$m, while no change was observed in samples below that thickness. Figure 5-14a shows an optical micrograph of such a feature. The length of the feature is about equal to the film thickness, as predicted by the analysis (see Figure 5-7). Figure 5-14b, is an SEM micrograph of a similar feature, where the curved nature of the crack front, of the same form as that of Figure 5-8, is evident. Also, Table 5-2 shows that catastrophic failures were observed in films thicker than 35 $\mu$m. Figure 5-15a, b shows the microscopic appearance of a failed sample. It is observed that an interconnecting network of cracks, resembling a
jigsaw puzzle, exists. The trajectories of the cracks always go through slits or other etched features.

5.5 Conclusions

A system of colinear through-thickness cracks, in a polymer film bonded to a rigid substrate, was analyzed. The analysis showed that the interaction between cracks results in amplification of the energy release rate to levels higher than those of a single crack. The amplification factor decreases rapidly with increasing distance between cracks. The case of a single crack is reached, when the distance between crack tips is greater than one film thickness.

The energy release rate of a single crack, with a vertical straight front, is a function its distance from the substrate. Its maximum is observed slightly below the free surface. A change in the crack front profile occurs when the maximum energy release rate exceeds the critical energy release rate of the material. Then, the crack front attains a curved shape of minimum energy release rate. When the maximum energy release rate of that profile exceeds the critical energy release rate of the material, catastrophic crack propagation follows.

A number of experiments were performed to check the validity of the analysis. In these, slits were etched in Cyclotene films of various thicknesses, attached to chromium coated silicon substrates. Above a threshold thickness value the slits grew to form a stable curved crack front profile. If this thickness was greater than a second threshold value, (which is greater than the first) catastrophic crack propagation took place. The two thickness threshold limits
agree well with those predicted by the analysis; The results suggest the analysis is valid.
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<td>3.388</td>
<td>2.237</td>
</tr>
<tr>
<td>DVS-BCB</td>
<td></td>
<td>0.24</td>
<td></td>
</tr>
</tbody>
</table>

Table 5-1. Fracture toughnesses (MPa m$^{1/2}$) of the three polymers, at three temperatures and at a strain rate of 0.1/min. All polymers were about 10 μm thick.

<table>
<thead>
<tr>
<th>Thickness</th>
<th>15 μm</th>
<th>25 μm</th>
<th>30 μm</th>
<th>40 μm</th>
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<tbody>
<tr>
<td>Shape change</td>
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<td>No</td>
<td>Yes</td>
<td></td>
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<tr>
<td>Fracture</td>
<td>No</td>
<td>No</td>
<td>No</td>
<td>Yes</td>
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</tbody>
</table>

Table 5-2. Failure modes of DVS-BCB films of various thicknesses, in which slits had been etched. The first row shows the thickness of the films, the second whether crack front shape change occurred, and the third, whether catastrophic crack propagation occurred.
Figure 5-1. a) A partial through-thickness crack and b) A through-thickness crack.
Figure 5-2. Top view of two colinear through-thickness cracks.
Figure 5-3. A single through-thickness crack.
Figure 5-4. Undeformed FEM mesh of one fourth of the model geometry.
Figure 5-5. Deformed FEM mesh of one fourth of the model geometry.
Figure 5-6. Energy release rate along the crack front, as function of distance from the substrate and half distance between cracks, both normalized by the thickness.
Figure 5-7. Dependence of the energy release rate amplification factor, on the half distance between cracks normalized by the thickness.
Figure 5-8. Dependence of the stress intensity factor on the distance from the substrate normalized by the thickness.
Figure 5-9. Schematic of the relation of the horizontal and vertical components to the total energy release rate, at a position on the crack front.
Figure 5-10. Minimum energy release rate crack front profile.
Figure 5-11. Energy release rate of the minimum energy crack front profile, as function of distance from the substrate normalized by the thickness.
Figure 5-12. Vertical crack front critical thicknesses, of BPDA-PDA, HFDA-APBP, and DVS-BCB.
Figure 5-13. Curved crack front critical thicknesses, of BPDA-PDA, HFDA-APBP, and DVS-BCB.
Figure 5-14. a) Optical micrograph of a slit extension and b) SEM micrograph of a slit extension which shows a curved crack front shape, in Cyclotene films 28 μm thick at room temperature.
Figure 5-15. Optical micrographs from a failed sample. a) cracks emanating from slits and b) cracks emanating from a square feature, in Cyclotene films 39 µm thick at room temperature.
6. CRACKING IN INTEGRATED CIRCUIT PLASTIC PACKAGES

6.1. Introduction

The encapsulation of IC plastic packages creates stresses due to the thermal mismatch between the plastic encapsulant, the silicon chip and the lead frame metal. These may lead to a number of failures such as cracked packages, cracked chips, delaminations at the various interfaces and metal deformation. Environmental factors, such as the presence of moisture, have also been shown to promote failures.

A number of works have addressed this area. Dale and Oldfield [69], provided a concise description of the stress mechanisms which are encountered in plastic IC packages during manufacturing. They identified the encapsulation process and the die attachment as the two most damaging fabrication steps. Based on Timoshenko’s analysis of laminated systems [70], they calculated the strain which is developed in each of the components of the assembly. By using interferometry, X-ray diffraction and strain gauges they showed their analytical calculations predicted accurately the strain in the middle of the chip. Using photoelasticity, Kotake and Takasu [71], measured the stresses developed in silicon chips after die attachment and encapsulation. They found the stresses changed significantly between those two steps. After die-attachment the chip was under tension near the surface away from the leadframe and under compression on the opposite side. After encapsulation, the entire chip was under compression. Kesslel et al. [72], using a test chip with diffused resistor strain gauges measured the residual strains after die-attachment. A number of die-attachment/leadframe combinations were tested. Their analysis showed that vertical die-cracking was the result of
bending due to the thermal mismatch between die and leadframe, which was aggravated by the presence of defects in the die-attaching material. A stress reduction in the chip due to the relaxation of thick adhesive layers was also noted. With temperature-compensated piezoresistive strain gauges attached to the chip, Usell and Smiley [73], measured the average strain in the chip during encapsulation. Two encapsulants were considered and a simple beam model with temperature dependent properties was used to calculate the average strain in the chip, with good results. An FEM model of the package was not so successful, because the code did not include temperature dependent properties. A gap between the chip and the encapsulant was observed and this was attributed to thermal stresses and to the swelling stresses caused by moisture absorption.

Based on measurements of the deformation of aluminum lines in the chip, Thomas [74], concluded that the stress distribution at the corners of the chip was not symmetrical. A number of encapsulants were tested but no improvement in metal deformation was observed. Thermal fatigue of the package produced microcracks in the plastic which were recognized as major contributors to metal deformation. Recognizing the importance of accurate determination of the stresses in the package, Groothuis et al. [75], used a two dimensional FEM model to calculate the locations of stress concentration for various combinations of geometries and material properties. Strain gauge measurements showed excellent correlation between the analysis and experimental results. Further, failures were observed at the locations of maximum stress concentration calculated by the analysis. Proceeding one step further and using simultaneous stress and heat transfer FEM analysis of a 2-D model, Miyake et al [76], calculated the stress and temperature distributions in
a plastic package. Steady-state cooling and heating, corresponding to temperature cycling of the package, and transient state heating, corresponding to solder dipping of the device, were considered. Strain gauge and temperature measurements using the $V_f$ temperature dependence of a diode, showed good agreement between analysis and experiment. Edwards et al. [77], built a test structure to investigate the fracture mechanisms and the stress distribution in plastic packages. A number of combinations of molding compound, leadframe material and chip coatings were tested and a relation between the measured level of compressive stress and the amount of damage observed in the package was found. They concluded that the test structure was a good way to evaluate different material combinations for packaging applications.

Nishimura et al. [78], investigated the effect of the lead frame material on encapsulant cracking. Two crack assisting mechanisms were identified. The first is delamination between the chip pad and the encapsulant and this occurs with low CTE leadframe materials. The second is delamination or sliding of the die-bonding layer which takes place with high CTE leadframe materials. They concluded that package cracking can be minimized by using a lead frame material with a CTE between those of silicon and the encapsulant, by using a moderately compliant layer, or by using a low modulus encapsulant. Recognizing the need for a more quantitative approach to package failure, Nishimura at al., [79], used fracture mechanics to estimate the fatigue life of IC plastic packages. Following experimental observations, they assumed that a delamination exists between the chip pad and the encapsulant and that an initial crack is present in the molding compound. The fatigue crack propagation in the molding material was assumed to follow the Paris
equation [80], and it was obtained from bending fatigue of sharp-notched specimens. FEM analysis was used to calculate the range of the stress intensity factor of the crack in the actual package. Their analysis showed good agreement with experiments and it provides a method to analyze strength problems in packages. The limitation is that the mode of failure and the initial crack size are determined by experimental observations, after cracking has already occurred. No predictive capability was presented. Van Vroonhoven [81], also employed fracture mechanics to analyze the effect of delaminations on the probability of cracking. A generic geometry, a wedge between two dissimilar elastic materials, was used to model the interface between the chip and the molding compound. The stress intensity at the corner of the wedge is characterized by a stress-singularity parameter which depends on the elastic properties of the materials and on the presence or absence of adhesion. (The stress intensity at the corner increases when adhesion is lost). It was also shown that for uniform temperature distribution the stress singularity does not depend on thermal influences. The approach does not provide a criterion for delamination or for cracking but it explains qualitatively why the loss of adhesion may cause cracking in the plastic.

The existence of different failure mechanisms in IC packages, the range of package sizes and of the properties of the encapsulants, make their general treatment very difficult. This is evident in all of the above work where failure mechanisms are experimentally identified and then quantitative results are presented only for particular cases. In our work, based on a model geometry, we present a method to predict cracking in plastic IC packages under monotonic thermal loading, in the absence of any delaminations.
6.2. **Numerical - Analytical**

The problem can be generalized by considering a generic square chip. A chip to package ratio of 1:2 was chosen (Figure 6-1). The solution is general enough to provide guidelines for the design of chips of more complicated geometries. The loading is the residual stress developed by polymerization shrinkage and the thermal mismatch between the polymer and the chip. Plane strain conditions are assumed.

Due to symmetry, if a crack develops in the polymer, it will extend along the direction defined by the diagonal of the package. Thus, the analysis considers only one eighth of the assembly shown in Figure 6-1. The stress analysis was performed by the Finite Element Method, using the commercial code ABAQUS. The mesh was prepared with PATRAN and it is shown in Figure 6-2. The modified crack closure method [41] was used to calculate the energy release rate of the crack.

A number of cases were analyzed. These involved different crack lengths, and different polymer and chip moduli and Poisson’s ratios. From the results, it became apparent that it was possible to generalize them by defining a dimensionless parameter, the M factor, when plane strain conditions obtain:

\[
\frac{(1-v) G E_p}{(1+v) (1-1.8v) \sigma_0^2 L f(L/D)} = M
\]  

(6-1)

In the expression, G is the energy release rate for propagation of a crack of length L, L/D is the ratio of crack length to chip diagonal, f(L/D) is a function of L/D, presented in Table 6-1, E_p is the modulus of the polymer, v is the Poisson’s ratio of the polymer, and \( \sigma_0 \) is an equivalent residual stress. This
is defined as the stress developed under plane stress conditions in a uniaxially constrained polymer strip with a coefficient of thermal expansion (CTE) equal to the difference between the CTE of the polymer and the CTE of the chip, when the strip is subjected to the same processing and thermal history as that of the actual package. (The ends of the strip are held fixed). All the material variables in equation (1) may vary with temperature.

Equation (6-1) can also be rewritten in terms of the stress intensity factor $K$, of the crack instead of its energy release rate, $G$. In plane strain, these two quantities are related by the equation:

$$ G = \frac{K^2 (1-v^2)}{E} \quad (6-2) $$

so Equation (6-1) becomes:

$$ \frac{(1-v) K}{\sqrt{(1-1.8v)}} \frac{\sigma_0}{\sqrt{f(L/D)}} = \sqrt{M} \quad (6-3) $$

The dependence of $M$ on $E_p/E_c$ and on $v$, is shown in Figure 6-3 which can be used to predict whether or not propagation of an existing crack of known length will occur. To do this, we need to calculate:

$$ \frac{(1-v) G_c E_p}{(1+v)(1-1.8v) \sigma_0^2 L f(L/D)} = M_c \quad (6-4) $$

where $G_c$ is the critical value of the energy release rate of the polymer at the temperature considered. For a known ratio of $E_p/E_c$ and a known value of the polymer's Poisson's ratio, if the value of $M_c$ is below that of the appropriate Poisson's ratio's curve, crack propagation will occur. If $M_c$ is above the relevant curve, the crack remains static.
As seen in Figure 6-3, the M factor is a weak function both of \( \frac{E_p}{E_c} \) and \( v \). For small cracks, it is also essentially independent of the crack length to chip diagonal ratio \( (L/D) \). If \( \frac{E_p}{E_c} \) ranges from 0.01 to 0.12, and the Poisson’s ratio of the polymer is between 0.25 and 0.40, \( M \) can be approximated by:

\[
\frac{(1-v) G E_p}{(1+v) (1-1.8v) \sigma_0^2 L} = 2.8 \tag{6-5}
\]

Thus, if \( M < 2.8 \) the crack grows; if \( M > 2.8 \), the crack remains static.

The analysis assumes a crack exists in the polymer at the corner of the chip but usually such a crack is absent so the analysis is not directly applicable. This can be remedied by invoking Griffith’s concept of an inherent flaw [2]; we assume the Griffith’s flaw extends from the corner of the chip into the polymer, along the direction of the diagonal. The size of the inherent flaw can be determined from tensile and fracture toughness experiments on the encapsulant material, assuming it is brittle. If the tensile strength of the material at a given temperature is \( \sigma_u \) and its critical energy release rate is \( G_c \) then its inherent flaw size is:

\[
L_{cr} = \frac{2 G_c E_p}{(1-v^2) \pi \sigma_u^2} \tag{6-6}
\]

By substituting this crack length into Equation (6-5) and solving for \( G \) we get:

\[
G = \frac{5.6 (1-1.8v) G_c \sigma_0^2}{(1-v)^2 \pi \sigma_u^2} \tag{6-7}
\]

Fracture will occur at the residual stress level at which \( G \) becomes equal to \( G_c \) so equating the right hand side of Equation (6-7) to \( G_c \) and solving for \( \sigma_0 \) we obtain:
\[
\sigma_0 = \frac{0.74 (1-\nu) \sigma_u}{1 - 1.8v}
\]  
(6-8)

Failure will occur at the temperature at which the above relation is satisfied. To estimate that temperature, we need to know the equivalent residual stress, the tensile strength and the Poisson’s ratio of the polymer, all as functions of temperature. It is known that the tensile strength of a brittle material is a stochastic quantity. Thus, one should expect that the fracture temperatures of the packages will also follow some stochastic distribution. The distribution of tensile strengths and fracture temperatures are not those implied by Equation (6-8) however. The reason for this is the fact that a chip-package will break when a crack appears in any one of its four corners so the probability that it will fracture at a temperature corresponding to a tensile strength below its average value is 15/16, and above it is 1/16. The fracture temperatures of the packages will shift toward the lower values of the tensile strength. This must be kept in mind when evaluating the experimental results.

In the analysis, the corner of the chip has been assumed to be perfectly sharp; this produces the most severe stress concentration at the corner. If the corner is rounded, the concentration decreases as the radius of curvature increases. The limiting case is a circular chip (Figure 6-4). The solution of this problem provides an upper limit for the strength of the assembly and gives an estimate of the sensitivity of Equation (6-8) to the radius of curvature.

The elastic solution of the problem of a hollow cylinder under plane strain conditions, filled with a perfectly rigid material and subjected to a residual strain \(\varepsilon_0\), can be obtained from the solution of a hollow cylinder under internal pressure and plane strain conditions, found in Saada [82]. The
two problems are equivalent if the internal pressure produces a radial

displacement of \((1+\nu)\varepsilon_o r_i\) at the internal radius \(r_i\). This pressure is:

\[
\rho = E \varepsilon_o \frac{(1-\lambda^2)}{[1+\lambda^2(1-2\nu)]} \tag{6-9}
\]

where \(\lambda\) is the ratio of internal to external radius:

\[
\lambda = \frac{r_i}{r_e} \tag{6-10}
\]

The hoop stress \(\sigma_{\theta}\) at the internal radius is:

\[
\sigma_{\theta} = E \varepsilon_o \frac{(1+\lambda^2)}{[1+\lambda^2(1-2\nu)]} \tag{6-11}
\]

When \(\lambda=0.5\) then:

\[
\sigma_{\theta} = \sigma_o \frac{1.25}{1.25 - 0.5\nu} \tag{6-12}
\]

where

\[
\sigma_o = E \varepsilon_o \tag{6-13}
\]

If an inherent flaw exists at the internal radius of the cylinder then the

critical residual stress \(\sigma_o\) is that for which \(\sigma_{\theta} = \sigma_u\). Thus, the condition of

fracture in this geometry is:

\[
\sigma_o = (1 - 0.4\nu) \sigma_u \tag{6-14}
\]

This is the upper limit for the strength of an encapsulated chip-package with

the geometry considered here. The relatively small difference between the

values given by Equations (6-8) and (6-14), for typical values of Poisson’s ratios

of polymers, indicates that for a square chip the effect of a small radius at its

corner is negligible.
For comparison, the package fracture analysis has also been repeated assuming plane stress conditions, even though this is not appropriate for an actual chip-package, because the adhesion between the chip and the polymer laterally constrains the polymer and produces plane strain conditions near the chip corner. The M parameter for plane stress is:

$$\frac{G E_p}{(1-1.127v) \sigma_0^2 L f(L/D)} = M$$  \hspace{1cm} (6-15)

where all the quantities in the above expression have been defined previously. The dependence of M on the other system parameters is shown in Figure 6-5. Again, M is nearly constant and equal to 2.8 in the region of interest. Following a derivation similar to that for plane strain conditions, we obtain the following expression for the strength of the package, which should be compared to equation (6-8):

$$\sigma_0 = \frac{0.74 \sigma_u}{\sqrt{(1-1.127v)}}$$  \hspace{1cm} (6-16)

(The effect of a corner radius on the strength of the package is expected to be small, for the same reasons as described for plane strain).

6.3. Experimental

To assess the value of the analysis, square glass chips were encapsulated in a commercially available encapsulant, E-565 from Emerson & Cumming. This is a single component epoxy system cured at 123 °C for 2 hours. A viscoelastic analysis of it was done [40], between 130 °C and -140 °C at three different frequencies: 0.5, 5 and 20 Hz. The variation of the storage modulus with temperature and frequency is shown in Figure 6-6 and the modulus versus frequency at a reference temperature of 20 °C is shown in Figure 6-7.
The shift factor versus temperature, is presented in Figure 6-8. The relaxation constants and times are shown in Table 6-2. The CTE of this polymer was measured to be 130 ppm/°C above 40 °C, and 55 ppm/°C below -10°C; in the range between these two temperatures it was assumed to vary linearly. The modulus of the glass chip is 70 GPa, its Poisson's ratio is 0.25 and its coefficient of thermal expansion is 2.3 ppm/°C in the temperature range of interest.

Based on the above mechanical and thermal properties, a viscoelastic analysis was performed in ABAQUS [25] to obtain the residual stress \( \sigma_0 \) as function of temperature; the results are shown in Figure 6-9. Above room temperature the residual stress is small but it increases rapidly below room temperature attaining high values below -100 °C. To estimate the failure temperature for this assembly we need to know the tensile strength of the polymer at these low temperatures. A number of tensile experiments were performed on polymer strips 7 cm long, 6 mm wide and 0.7 mm thick, at -70 °C which is the low temperature limit of our environmental chamber, an INSTRON model 3111. The values are shown in Table 6-3; only specimens which broke at locations away from the grips, are reported. (The fracture stress at lower temperatures is not expected to vary more than a few percent from these values [83]). It is seen that the lowest stress is 23.6 MPa and the highest 62.6 MPa, with an average of about 42 MPa. According to Equation (6-8), the residual stress at fracture should vary between 18 and 46 MPa with an average of 31 MPa. From Figure 6-9 we can obtain the fracture temperatures: -104 °C to -217 °C, with an average of about -157 °C.

Two sets of experiments were performed. In the first, eight glass chips 1.27 cm square and 3 mm thick were encapsulated in the polymer (Figure 6-10a). To avoid any bending effects, all specimens were polished on both sides
to ensure that the chip was in the center of the thickness of the samples. They were then cooled from room temperature in a small chamber, until they broke (Figure 6-10b). The failure temperatures were recorded and are shown in Table 4. They range from -98 °C to -174 °C which agrees well with the range predicted by the analysis. Also, the average fracture temperature of -142 °C is higher than that predicted by the analysis, -157 °C, for the reason already mentioned.

The second set of experiments used glass chips 2.54 cm square and 3mm thick. The failure temperatures are shown in Table 6-4. Again, the range (-108 °C to -160 °C) and average (-135°C) are close to those of the smaller chips; as expected from the analysis, the size of the chip-package does not significantly affect the fracture process.

We recognize that the model used is a very simple one, much simpler than any real IC geometry and construction. We believe, however, that the tools presented in this analysis are generally valid, and could be applied successfully to such real packages. The tools are: 1) the viscoelastic method to predict residual thermal stresses. 2) Use of the FEM to analyze stress concentrations, the locations of which usually are obvious. 3) Use of the modified crack closure method to calculate the energy release rate of a crack in such locations, and expressing the results in a dimensionless factor or parameter. 4) Invoking the inherent flaw size concept from Griffith, if the materials involved are brittle. In essence, this comprises a virtual crack, specific to the material.

Examination of Figure 6-10 shows that the glass chips in the model also break. This occurs because once the polymer at the chip corner starts to crack,
the chip there goes into tension along its diagonal and fracture of it soon occurs. We know that in real packages, chip corner breakage often is seen, but with the crack perpendicular to the diagonal rather than parallel to it. This is due to chip bending as has been shown by Kessel et al. [72]. Bending effects are not included in the present model.

6.4 Effect of a compliant interlayer on package cracking

The effect of a compliant interlayer between the chip and the polymer, on the energy release rate of a crack in the polymer at the corner of the chip, was investigated numerically by the Finite Element method. The thickness of the compliant layer was $1/100$ of the chip semi-diagonal, i.e., $t/D = 1/100$. The geometry of the model in this case is shown in Figure 6-11. Due to symmetry, only one eighth of the model needs to be considered, i.e., region oab of Figure 6-11. The FEM mesh used for this reason is shown in Figure 6-2. Similar to the case without a compliant interlayer, the loading of the assembly is a residual stress developed from the thermal mismatch between the materials of the assembly.

Figure 6-12 shows the variation of the ratio of the energy release rate of a crack when a compliant layer is present to that of the same crack when there exists no compliant layer in the assembly ($G_r/G_p$), as a function of the crack length divided by the chip semi-diagonal ($L/D$). The values in that Figure have been obtained assuming that the ratio of the modulus of the rubber to that of the polymer is 0.01, i.e., $E_r/E_p = 0.01$. Also the Poisson’s ratio of the rubber has been taken as 0.49 and that of the polymer as 0.30. It is evident that for small cracks, the presence of the compliant layer increases the driving force for crack propagation. For larger cracks, i.e, $L/D > 0.2$, the rubber layer
has the opposite effect on the crack driving force. The explanation for this behavior is that for small cracks, the rubber layer can produce higher crack opening displacements near the chip corner, because the polymer can displace more easily along the side of the chip, compared to when no rubber is present. Also, near the corner, the almost incompressible behavior of the rubber, due to the triaxial stress state, does not allow it to have any stress relieving effect. For larger cracks, the previous two factors do not exist and in addition, due to the biaxial stress state, the rubber relieves the residual stresses, thus reducing the crack driving force.

Figure 6-13 shows the variation of Gr/Gp, as function of Er/Ep, for two different Poisson’s ratios of the rubber, i.e., 0.40 and 0.49, for a small crack of length L/D = 0.02. The Poisson’s ratio of the polymer has been kept constant at 0.30. It is observed that for most practical cases, (Poisson’s ratio of the rubber greater than 0.49; ratio of rubber to polymer modulus of less than 0.05) the rubber has an adverse effect on the driving force for small cracks. Since inherent flaws can be considered small cracks (for instance a typical inherent flaw of 50 μm and a typical chip of a few millimeters give a ratio of L/D less than 0.05, which is considered a small crack according to Figure 3-12) it can be concluded that a rubber interlayer will have an adverse effect on the strength of the package.

6.5 Conclusions

A model geometry of a square chip encapsulated in a polymer has been used to investigate cracking in plastic IC packages. It was assumed that no adhesive failures were present in the package prior to cracking. The loading of the package was thermal shrinkage stresses from uniform cooling of the
assembly after curing of the polymer. The energy release rate for propagation of a crack of variable length, from the corner of the chip into the polymer was calculated by the Finite Element Method. The results were expressed in a dimensionless parameter. By assuming that an inherent flaw exists at the corner of the chip in the polymer, we were able to obtain an expression for the strength of the package, when no detectable cracks are present. The expression, which obtains at fracture, relates the residual stress in the polymer to its tensile strength. Experiments performed with glass chips and a commercial epoxy encapsulant showed reasonable agreement with the predictions of the analysis.

Finally, from an FEM analysis of a model package it was shown, that a rubber interlayer between the chip and the polymer increases the driving force for crack propagation of small cracks, from the corner of the chip into the polymer. Since, inherent flaws are small cracks, it is concluded that the rubber interlayer will decrease the strength of the package against cracking.
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</tr>
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Table 6-1. Dependence of $f(L/D)$ on $L/D$ (see text for details).
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Table 6-2. Relaxation constants and times of encapsulant E-565.
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</tr>
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<td>62.2</td>
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Table 6-3. Tensile strength of E-565 at -70 °C.
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<th>$T_{fracture}$ ($^\circ$C)</th>
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<tr>
<td>-168</td>
<td>-160</td>
</tr>
<tr>
<td>-174</td>
<td>**</td>
</tr>
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</table>

Table 6-4. Fracture temperatures in degrees Celsius, of a number of smaller and larger packages. The asterisks signify that two specimens had not broken at -175 $^\circ$C.
Figure 6-1. Model geometry.
Figure 6-2. FEM mesh of one eighth of the model.
Figure 6-3. Dependence of the dimensionless energy release rate under plane strain on the ratio of polymer to chip Young’s modulus and the polymer Poisson’s ratio.
Figure 6-4. Geometry of a circular chip.
Figure 6-5. Dependence of the dimensionless energy release rate under plane stress on the ratio of polymer to chip Young’s modulus and the polymer Poisson’s ratio.
Figure 6-6. DMA results for encapsulant E-565.
Figure 6-7. Dependence of storage modulus of E-565 on frequency at a reference temperature of 20 °C.
Figure 6-8. Dependence of the logarithm of the shift factor of E-565 on temperature.
Figure 6-9. Dependence of equivalent residual stress of E-565 on temperature.
Figure 6-10. Photographs of unbroken (a) and broken (b) chips (1.27 cm square) encapsulated in E-565.
Figure 6-11. Model geometry with the presence of a rubber interlayer between chip and polymer.
Figure 6-12. Ratio of energy release rates with and without rubber, as function of crack length to chip semi-diagonal ratio.
Figure 6-13. Ratio of energy release rates with and without rubber, as function of rubber to polymer Young’s modulus ratio.
Electronic packages are composite structures made with a combination of materials, such as polymers, metals, and ceramics. Often they are subject to thermal loading. Due to the property mismatch between their constituent materials, high thermal and processing stresses develop, often leading to fracture. A general methodology, based on fracture mechanics, has been introduced in this work to analyze thermomechanical effects in electronic packages. In particular, the strength of packages against cracking is of primary concern. The methodology is as follows:

(i) To apply fracture mechanics, a crack has to exist in the structure. If such cracks exist, we proceed to the next step. If not, an assumed crack is introduced in the structure, in the place of another stress concentrating feature which already exists, e.g., a hole. If such features do not exist, a Griffith flaw, an inherent crack, is assumed to exist at the location of maximum stress concentration in the structure.

(ii) The energy release rate for propagation of the crack, under a variety of thermal loadings, is calculated by the Finite Element Method, and the modified crack closure method. When possible, the results are expressed in a dimensionless form so they can be generally applicable. For the FEM to produce accurate results, the time-dependent and time-independent material mechanical properties, such as viscoelasticity, plasticity, visco-plasticity must be known. The time-independent properties are obtained from tensile experiments at different temperatures. A technique, based on dynamic measurements at different temperatures and frequencies and the time
temperature superposition principle, has been used to obtain the viscoelastic properties of the polymers in a form useful for FEM analysis.

(iii) Then, the calculated energy release rates, from step two, are compared to the critical energy release rates for propagation of the cracks, which are obtained by independent experimental measurements on simpler structures. The critical energy release rate for propagation of a crack in a homogeneous material is obtained from fracture toughness experiments. The critical energy release rate for propagation of a crack at the interface between two materials, is obtained from the Island Blister Test. A large displacements elastic-plastic analysis of this test has been developed, to convert the experimental pressure measurements to interface critical energy release rate values.

(vi) If the calculated energy release rate is greater than its critical value, failure is postulated.

The method has been used to evaluate the strength against cracking and loss of adhesion, of a polymer film bonded to a rigid substrate. Both phenomena are relevant to High Density Interconnects. The results were expressed in dimensionless forms. From the calculations a critical thickness emerges, above which failure should occur. The results indicate that only very brittle polymers may show polymer cracking in usual interconnect applications. This is due to the relatively low energy release rates, associated with cracks in films attached to rigid substrates, and the small thicknesses of HDI's. Three different polymer films, ВРДА-РДА, HFDA-АРБР, and DVS-BCB, attached to a metal coated silicon substrate, were considered as examples. The results of the experiments showed good agreement with the analysis.
Unlike the case of film cracking, a loss of adhesion or delamination between the film and the substrate is more probable. The reason is the much lower critical energy release rates associated with interfacial cracks. In most polymers, these are orders of magnitude lower than encountered with the propagation of cracks in the homogeneous material. The critical energy release rate for crack propagation in HFDA-APBP is about 3000 J/m², while for a crack at the interface between HFDA-APBP and chromium it is about 60 J/m².

The method has also been applied to the case of a square chip encapsulated in polymer. In this case, the inherent flaw size concept was invoked. A critical relation between the residual stress in the polymer and its tensile strength was predicted to hold at fracture. If the residual stress and the tensile strength are known as functions of temperature, the temperature of fracture can be calculated. Tests with square glass chips encapsulated in an epoxy showed reasonable agreement.

The results of this work indicate that for an accurate prediction of the fracture process, the time dependent and time independent properties of the materials must be known, in a form useful for numerical stress analysis. For polymers in particular, their properties are complicated functions of time, temperature and stress levels. In most commercial FEM codes, there exist no appropriate models for such properties. In ABAQUS for example, non-linear viscoelastic behavior is not included in the material properties library, so, one could not model time-dependent behavior which is also dependent upon the stress level, without having to create his own user subroutine with the correct material model. Such work will be very useful for many applications involving large deformations, but it will require expertise in finite element
methods, mechanics, and computer programming, and long development and validation time.

Another issue, not addressed in this work, is fatigue fracture of electronic components. This involves two phases before failure: Crack initiation and slow crack propagation. Initiation is one of the problems least understood in fracture mechanics but perhaps, the problem can be reduced to that of crack propagation, by assuming the existence of an inherent flaw or virtual crack in the same manner introduced in this work. Then, the method illustrated in this work could be used to obtain the dependence of the energy release rate of the crack on the crack length, in the model structure. The dependence of the crack length on time, can be found independently, from simpler fatigue crack propagation experiments, using one of the existing crack propagation laws.

Finally, another issue which warrants further work is that of the effect of processing on the mechanical properties of the materials. This can be done in two ways; one is to perform a parametric study where the processing of a material or a structure is varied and then its mechanical properties are measured. The correlation between processing and properties can be obtained empirically. A second approach would attempt to model the microscopic processes which are responsible for the macroscopic properties of the system, perhaps using molecular modeling, statistical mechanics and other computer simulation techniques. This approach will require more basic research and may yield results only in the long term.
APPENDIX A

Program Name: tshift.f

Description: This program accepts as input the values of the storage modulus at three different frequencies and at a number of temperature intervals spanning the temperature range of interest, as obtained from a Dynamic Mechanical Analysis (DMA) of a sample. The program performs the time temperature shift (see Ferry's book "Viscoelastic properties of polymers" for an explanation of the time temperature shift) and produces as output the following four (4) files:

(i) "outs.dat": This output file contains either error messages, in case that the execution of the FORTRAN program has stopped for some reason, or the results of the execution.

(ii) "outx.dat": This output file contains two columns of numbers. The first column contains the temperatures and the second the corresponding values of the logarithms of the shift factor. These raw data are suitable for plotting the logarithm of the shift factor as function of temperature by using some other software.

(iii) "outmx.dat": This output file contains two columns of numbers. The first column contains values of the logarithms of the frequency and the second the corresponding values of the storage modulus. These raw data are suitable for plotting the master curve, i.e. the variation of the storage modulus with the logarithm of frequency.
(iv) "outaba.dat" : This output file contains the viscoelastic properties of the analyzed sample in a form suitable for direct input in the ABAQUS input file. This is actually the only necessary file for the ABAQUS analysis.

**Input File : "inshift.dat"**

This file should contain all input data necessary for the execution of the FORTRAN program "tshift.f". The following information should be provided, line by line, in the format explained below:

Line 1 : NFREQ (I10), 1b, NTEMP (I10), 1b, Tref (G10.4), 1b, NCON (I10)

Line 2: W1 (G10.4)

Line 3: W2 (G10.4)

Line 4: W3 (G10.4)

Line 5: T1 (G10.4), 1b, E'w1 (G10.4), 1b, E'w2 (G10.4), 1b, E'w3 (G10.4)

Line 6: ..............................................................

Line n+4: Tn (G10.4), 1b, E'w1 (G10.4)......

where:

NFREQ = number of frequencies that data are input. Always set this to 3.

NTEMP = number of temperature points that data are input.
Tref= Reference temperature at which output data will be produced. It can be any temperature, chosen by the user, in the temperature range where DMA data have been obtained.

NCON= number of relaxation constants in the PRONY series expansion. This number has to be greater or equal to 2. The larger this number the better the accuracy of the FEM analysis is going to be (10 terms were taken in the example).

W1= The minimum frequency of the data (in Hz).

W2= The intermediate frequency of the data (in Hz).

W3= The maximum frequency of the data (in Hz).

T1= The lowest temperature that data are input.

E'w1= Storage modulus at the lowest frequency and at the corresponding temperature on the same line (T1 for Line4).

E'w2= Storage modulus at the intermediate frequency and at the corresponding temperature on the same line.

E'w3= Storage modulus at the highest frequency and at the corresponding temperature on the same line.

Tn= The highest temperature of input data.

1b= 1 blank space between entries.
The symbols in the parentheses next to an entry are the FORTRAN FORMAT of that entry. Commas (,) should not be entered in actual input file. They are used above simply to separate entries.

There are a few important things to remember when the input file is created: (i) The temperatures should be given in ascending values as we move down the input file. ii) The storage modulus at a given frequency decreases as the temperature increases. This means that the storage moduli are in descending values as we move down the input file. (iii) For two consecutive temperatures, the storage modulus at the lower temperature and the lowest frequency should be smaller than the storage modulus at the higher temperature and the highest frequency. This limitation is necessary in order to assure that the time-temperature shift can be performed. Therefore, the DMA analysis should provide data at relatively small temperature intervals e.g. less than 7 degrees Celsius (°C). The listing of the FORTRAN program tshift.f follows:

```
C********************************************************************************
C     PROGRAM NAME: tshift.f       *
C     by                                *
C     George Margaritis               *
C     M.I.T.                           *
C     Dept. of Mat. Science & Engineering    *
C     Cambridge, MA 02139     *
C12345********************************************************************************
C$S=0
C$T=0
C$P=0
C     Program that performs time-temperature superposition of
C     viscoelastic data and calculates the shift factor as
C     function of temperature from the storage modulus,
C     the storage modulus as function of Logf and creates the part
C     of ABAQUS input file which contains the viscoelastic material
C     properties.
C     This program works only if
C     (i) for fixed freq. E'(T1) > E'(T2) for T1 < T2
```
(ii) for two successive temperatures, 
\[ E'(\text{Freq min, Tsmaller}) < E'(\text{Freq max, Tlarger}) \]

Data should be checked by user!

\begin{verbatim}
DIMENSION EXSTOR(200,10), SH(200), STORF(1000,2), W(10), SHR(200)
& ECON(100), RELTIM(100)
OPEN (UNIT=5, FILE='inshift.dat')
OPEN (UNIT=6, FILE='outshift.dat')
OPEN (UNIT=7, FILE='outx.dat')
OPEN (UNIT=8, FILE='outmx.dat')
OPEN (UNIT=9, FILE='outaba.dat')
OPEN (UNIT=10, FILE='outsf.dat')
OPEN (UNIT=11, FILE='outmc.dat')
OPEN (UNIT=12, FILE='outdma.dat')
\end{verbatim}

NFREQ= # of different freq. measured, NTEMP=# of T slices

TREF= Reference Temperature at which the results are printed

READ (5,10) NFREQ, NTEMP, TREF, NCON

10 FORMAT (I10:1X,I10,G10.4,1X,I10)

W(I)=Matrix with the frequencies at which data have been obtained

Frequencies should be given in ascending order e.g. \( W(1)=W_{\text{min}} \)

READ (5,15) (W(I),I=1,1,NFREQ)

15 FORMAT (G10.4)

EXSTOR(I,J)=Storage Modulus of Ith T slice & Jth frq. value

READ (5,20) ((EXSTOR(I,J),J=1,NFREQ+1), I=1,NTEMP)

Be sure to set the number in front of G10.4.. to NFREQ+1

20 FORMAT (4(G10.4,1X))

Checks if data fulfill requirements above.

\begin{verbatim}
IF (TREF.LT.EXSTOR(1,1).OR.TREF.GT.EXSTOR(NTEMP,1)) THEN
WRITE (6,21)
21 FORMAT ('Reference Temperature out of Range. Execution Stops.')
GO TO 300
END IF
\end{verbatim}

DO 28 I1=1,NTEMP-1

\begin{verbatim}
IF (EXSTOR(I1,2).GT.EXSTOR(I1+1,NFREQ+1)) THEN
WRITE (6,22) I1,I1+1
22 FORMAT ('T-slice=','I3,4X,I3,/',
& 'CONDITION (II) VIOLATED. NOT ENOUGH T-SLICES.','/',
& 'EXECUTION STOPS. CHECK DATA AGAIN!')
GO TO 300
END IF
\end{verbatim}

DO 25 I2=2,NFREQ+1

\begin{verbatim}
IF (EXSTOR(I1,I2).LE.EXSTOR(I1+1,I2).OR.
& EXSTOR(I1,1).GE.EXSTOR(I1+1,1)) THEN
WRITE (6,23) I1,I1+1
23 FORMAT ('T-slice=','I3,4X,I3,4X,/',
& 'DATA NOT IN PROPER ORDER. PROGRAM STOPS.','/',
& 'CHECK DATA FOR ASCENDING T VALUES AND',
\end{verbatim}
& ' DESCENDING E VALUES')
GO TO 300
END IF
25 CONTINUE
DO 27 I3=3,NFREQ+1
IF (EXSTOR(I1,I3-1).GE.EXSTOR(I1,I3)) THEN
WRITE (6,26) I1
26 FORMAT ('T-slice='I3,3X,/, & 'FOR A T-SLICE Es ARE NOT IN ORDER. CHECK DATA')
GO TO 300
END IF
27 CONTINUE
28 CONTINUE
C Writes DMA data in file outdma.dat for Mac
DO 35 I1=1,NFREQ+1
WRITE (12,29)
29 FORMAT ('**')
DO 32 I2=1,NTEMP
WRITE (12,30) EXSTOR(I2,I1)
30 FORMAT (G10.4)
32 CONTINUE
35 CONTINUE
C Part of program which calculates shift factor of T.
SH(I1)=0.
40 DO 70 I1=1,NTEMP-1
EB=EXSTOR(I1,2)
DO 50 I2=1,NFREQ-1
EL=EXSTOR(I1+1,I2+1)
EG=EXSTOR(I1+1,I2+2)
IF (EL.LE.EB.AND.EG.GT.EB) GO TO 60
IF (EB.GT.EXSTOR(I1+1,NFREQ+1)) GO TO 300
50 CONTINUE
60 A=10
RA=LOG(W(I2+1)/W(I2))/LOG(A)
SH1=(EB-EL)/(EG-EL)*RA
SH2=-(LOG(W(I2)/W(I1))/LOG(A)+SH1)
SH(I1+1)=SH(I1)+SH2
70 CONTINUE
C Part of program which calculates Log aT vs. T starting
C at the reference temperature.
DO 72 I1=1,NTEMP-1
TL=EXSTOR(I1,1)
TH=EXSTOR(I1+1,1)
IF (TREF.LT.TL.OR.TREF.GT.TH) GO TO 72
SL=SH(I1)
SHI=SH(I1+1)
SHRE=SHI-(TH-TREF)/(TH-TL)*(SHI-SL)

72 CONTINUE
    DO 74 I1=1,NTEMP
    SHR(I1)=SH(I1)-SHRE

74 CONTINUE
C   Part of the program which prints the shift factor vs. Temp.
    WRITE (6,80) TREF
80 FORMAT (5X,'Tref=',G15.5)
    WRITE (6,90)
90 FORMAT (10X,'TEMP',15X,'Log aT')
    DO 120 I1=1,NTEMP
    WRITE (6,100) EXSTOR(I1,1), SHR(I1)
100 FORMAT (5X,G15.5,3X,G15.5)
C   Prints LogaT vs. T for xmath and Mac
    WRITE (7,110) EXSTOR(I1,1), SHR(I1)
110 FORMAT (G15.5,5X,G15.5)
    WRITE (10,115) EXSTOR(I1,1)
115 FORMAT (G15.5)
120 CONTINUE
    WRITE (10,123)
123 FORMAT (**)
    DO 128 I1=1,NTEMP
    WRITE (10,125) SHR(I1)
125 FORMAT (G15.5)
128 CONTINUE
C   Part of the main program which gives the Master Curve
    WRITE (6,130)
130 FORMAT (5X,'Stor. Modulus',10X,'Log Freq')
    DO 170 I1=1,NTEMP
    DO 160 I2=1,NFREQ
      I3=(I1-1)*NFREQ+I2
      STORF(I3,1)=EXSTOR(I1,I2+1)
      A=10
      STORF(I3,2)=LOG(W(I2))/LOG(A)+SHR(I1)
C   Prints E' vs. logf
    WRITE (6,140) STORF(I3,1),STORF(I3,2)
140 FORMAT (5X,G15.5,5X,3X,G15.5)
C   Prints E' vs. logf for xmath and Mac
    WRITE (8,150) STORF(I3,2),STORF(I3,1)
150 FORMAT (G15.5,4X,G15.5)
    WRITE (11,155) STORF(I3,2)
155 FORMAT (G15.5)
160 CONTINUE
170 CONTINUE
    WRITE (11,171)
171 FORMAT (**)

147
DO 174 I1=1,NTEMP
DO 173 I2=1,NFREQ
I3=(I1-1)*NFREQ+I2
WRITE (11,172) STORF(I3,1)

172 FORMAT (G15.5)
173 CONTINUE
174 CONTINUE
C    Creates part of ABAQUS input file which contains the
C    viscoelastic properties
WRITE (9,175)
175 FORMAT (**VISCOELASTIC, TIME=PRONY")
   EG=EXSTOR(I1,3)
   ER=EXSTOR(NTEMP,3)
   EK=(EG-ER)/NCON
   ECON(I1)=ER+0.5*EK
DO 180 I1=2,NCON
   ECON(I1)=ECON(I1-1)+EK
180 CONTINUE
DO 250 I1=1,NCON
DO 230 I2=1,NTEMP-1
   IF (ECON(I1).LE.EXSTOR(I2,3).AND.ECON(I1).GT.EXSTOR(I2+1,3);
&    GO TO 200
GO TO 230
200 FL=LOG(W(2))/LOG(A)+SHR(I2)
   FH=LOG(W(2))/LOG(A)+SHR(I2+1)
   EL=EXSTOR(I2,3)
   EH=EXSTOR(I2+1,3)
   FREQL=FL+(ECON(I1)-EL)/(EH-EL)*(FH-FL)+0.79818
   RELTIM(I1)=10**(-FREQL)
230 CONTINUE
   EK1=EK/EG
   WRITE (9,240) EK1,RELTIM(I1)
240 FORMAT (G15.5,1X,", 0.,',1X,G15.5)
250 CONTINUE
   WRITE (9,260) NTEMP
260 FORMAT (**TRS, DEFINITION=USER'/
&  *USER SUBROUTINE'/
&'C12345*/'
&' C    This Sub is entered in the ABAQUS input file'/
&' C    instead of the default WLF equation, for'/
&' C    time-temperature shift in linear'/
&' C    viscoelastic analysis'/
&' SUBROUTINE UTRS(SHIFT, TEMP, DTEMP, TIME, DTIME,/
&'   &PREDEF, DPRED, STATEV, CMNAME, COORDS)/'
&' IMPLICIT REAL*8 (A-H,O-Z)/'
&' CHARACTER*8 CMNAME'/
&' DIMENSION SHIFT(2), PREDEF(1), DPRED(1), STATEV(1)'
&' &, COORDS(1), SFDAT(‘,I10,’), T(2), SF(2)'
&' &C Make sure that the first line of input file'
&' &C has the # of data points. Also make sure the'
&' &C Temperature range in input file includes'
&' &C that given in ABAQUS input file.'
&' &C NDAT=# of data points, SFDAT(J,1)=Temp'
&' &C , SFDAT(J,2)=Log aT'
    &' &C************ DATA ***************')
WRITE (9,263) NTEMP
263 FORMAT (6X,’NDAT=,’I10)
C Prints LogaT vs. T for ABAQUS input file
    DO 285 I1=1,NTEMP
        WRITE (9,265) I1,EXSTOR(I1,1)
265 FORMAT (6X,’SFDAT(‘,I3,’)=‘,G15.5)
        WRITE (9,270) I1,SHR(I1)
270 FORMAT (6X,’SFDAT(‘,I3,’)=‘,G15.5)
285 CONTINUE
        WRITE (9,290)
290 FORMAT( ’C******* END OF DATA ***************’)
    &’ T(1)=TEMP-DTEMP’
    &’ T(2)=TEMP’
    &’ DO 40 I1=1,2’
    &’ DO 30 I2=1,NDAT-1’
    &’ TL=SFDAT(I2,1)’
    &’ TH=SFDAT(I2+1,1)’
    &’ AL=SFDAT(I2,2)’
    &’ AH=SFDAT(I2+1,2)’
    &’ IF (TL.GT.T(I1).OR.TH.LT.T(I1)) GO TO 30’
    &’ SF(I1)=AL+(AH-AL)*(T(I1)-TL)/(TH-TL)’
    &’ SHIFT(I1)=10**(SF(I1))’
    &’ GO TO 40’
    &’ 30 CONTINUE’
    &’ 40 CONTINUE’
    &’ RETURN’
    &’ END’
300 STOP
END

An example of an input file follows:

3  40  20.  10
.5
5.
20.
| -139.4 | 4.537 | 4.690 | 4.845 |
| -136. | 4.321 | 4.485 | 4.648 |
| -125.2 | 4.109 | 4.213 | 4.331 |
| -120.3 | 3.989 | 4.118 | 4.282 |
| -115.5 | 3.913 | 4.064 | 4.206 |
| -110.7 | 3.864 | 3.982 | 4.122 |
| -100.7 | 3.752 | 3.812 | 3.983 |
| -95.6 | 3.620 | 3.770 | 3.878 |
| -90.4 | 3.549 | 3.703 | 3.797 |
| -85.4 | 3.534 | 3.631 | 3.725 |
| -80.3 | 3.445 | 3.560 | 3.662 |
| -75.4 | 3.385 | 3.506 | 3.603 |
| -70.3 | 3.374 | 3.469 | 3.546 |
| -65.2 | 3.315 | 3.407 | 3.484 |
| -60.0 | 3.300 | 3.343 | 3.411 |
| -54.7 | 3.168 | 3.292 | 3.348 |
| -49.5 | 3.125 | 3.208 | 3.276 |
| -44.0 | 3.077 | 3.137 | 3.219 |
| -38.5 | 2.997 | 3.078 | 3.150 |
| -33.2 | 2.908 | 3.018 | 3.089 |
| -27.8 | 2.790 | 2.926 | 3.016 |
| -23.2 | 2.753 | 2.850 | 2.931 |
| -17.4 | 2.618 | 2.770 | 2.840 |
| -12.2 | 2.545 | 2.659 | 2.734 |
| -6.9 | 2.416 | 2.561 | 2.636 |
| -1.7 | 2.228 | 2.410 | 2.503 |
| 3.5 | 2.018 | 2.211 | 2.344 |
| 8.8 | 1.732 | 1.963 | 2.114 |
| 14.4 | 1.356 | 1.631 | 1.802 |
| 20.0 | 0.8467 | 1.184 | 1.408 |
| 25.2 | 0.3880 | 0.7170 | 0.9677 |
| 30.9 | 0.1507 | 0.3495 | 0.5505 |
| 36.3 | 0.06353 | 0.1579 | 0.2975 |
| 41.8 | 0.03835 | 0.07210 | 0.1321 |
| 47.2 | 0.03074 | 0.04541 | 0.07046 |
| 52.5 | 0.02667 | 0.03459 | 0.04668 |
| 57.8 | 0.02466 | 0.02964 | 0.03643 |
| 63. | 0.02289 | 0.02716 | 0.03037 |
| 68.4 | 0.02282 | 0.02528 | 0.02815 |
| 73.7 | 0.02192 | 0.02443 | 0.02634 |

Finally, the following is an example of an ABAQUS input file, which can give the residual stress in a uniaxially constrained strip of material, as function of
temperature. This is relevant to the analysis of Chapter 6. (To obtain the residual stress in a biaxially constrained strip, relevant to the analysis of Chapter 2, we need only to change the type of elements from plane stress to plane strain, i.e. from CPS4 to CPE4).

*HEADING
VISCOELASTIC STRESS RELAXATION / TEMPERATURE
*PREPRINT, HISTORY=NO, MODEL=NO, ECHO=NO
*NODE, NSET=BEAM
1,0,0
2,0,10
3,10,0
4,10,10
5,20,0
6,20,10
7,30,0
8,30,10
*ELEMENT, TYPE=CPS4, ELSET=ALL
1,1,3,4,2
2,3,5,6,4
3,5,7,8,6
*BOUNDARY
1,1,2
2,1
7,1
8,1
*SOLID SECTION, ELSET=ALL, MATERIAL=POLYMER
*MATERIAL, NAME=POLYMER
*ELASTIC
4845, 0.3
*EXPANSION
50.E-6, -120
50.E-6, -20
120.E-6, 20
120.E-6, 70
*VISCOELASTIC, TIME=PRONY
.99479E-01 , 0., 23.560
.99479E-01 , 0., .62219
.99479E-01 , 0., .30236E-01
.99479E-01 , 0., .85673E-03
.99479E-01 , 0., .78290E-05
.99479E-01 , 0., .12082E-07
.99479E-01 , 0., .14323E-11
*TRSF, DEFINITION=USER
*USER SUBROUTINE
C12345*
C This Sub is entered in the ABAQUS input file
C instead of the default WLF equation, for
C time-temperature shift in linear
C viscoelastic analysis
SUBROUTINE UTRS(SHIFT, TEMP, DTEMP, TIME, DTIME,
&PREDEF, DPRED, STATEV, CMNAME, COORDS)
IMPLICIT REAL*8 (A-H,O-Z)
CHARACTER*8 CMNAME
DIMENSION SHIFT(2), PREDEF(1), DPRED(1), STATEV(1)
& COORDS(1), SFDAT( 40,2), T(2), SF(2)
C Make sure that the first line of input file
C has the # of data points. Also make sure the
C Temperature range in input file includes
C that given in ABAQUS input file.
C NDAT=# of data points, SFDAT(J,1)=Temp
C , SFDAT(J,2)=Log aT
C************ DATA **************

| NDAT | 40 |
| SFDAT( 1,1) | -139.40 |
| SFDAT( 1,2) | 23.667 |
| SFDAT( 2,1) | -136.00 |
| SFDAT( 2,2) | 22.475 |
| SFDAT( 3,1) | -125.20 |
| SFDAT( 3,2) | 20.924 |
| SFDAT( 4,1) | -120.30 |
| SFDAT( 4,2) | 19.994 |
| SFDAT( 5,1) | -115.50 |
| SFDAT( 5,2) | 19.491 |
| SFDAT( 6,1) | -110.70 |
| SFDAT( 6,2) | 19.076 |
| SFDAT( 7,1) | -100.70 |
| SFDAT( 7,2) | 17.892 |
| SFDAT( 8,1) | -95.600 |
| SFDAT( 8,2) | 17.012 |
| SFDAT( 9,1) | -90.400 |
| SFDAT( 9,2) | 16.551 |
| SFDAT(10,1) | -85.400 |
| SFDAT(10,2) | 16.397 |
| SFDAT(11,1) | -80.300 |
| SFDAT(11,2) | 15.623 |
| SFDAT(12,1) | -75.400 |
| SFDAT(12,2) | 15.127  |
| SFDAT(13,1) | -70.300 |
| SFDAT(13,2) | 15.011  |
| SFDAT(14,1) | -65.200 |
| SFDAT(14,2) | 14.370  |
| SFDAT(15,1) | -60.000 |
| SFDAT(15,2) | 14.021  |
| SFDAT(16,1) | -54.700 |
| SFDAT(16,2) | 12.935  |
| SFDAT(17,1) | -49.500 |
| SFDAT(17,2) | 12.417  |
| SFDAT(18,1) | -44.000 |
| SFDAT(18,2) | 11.617  |
| SFDAT(19,1) | -38.500 |
| SFDAT(19,2) | 10.629  |
| SFDAT(20,1) | -33.200 |
| SFDAT(20,2) | 9.8202  |
| SFDAT(21,1) | -27.800 |
| SFDAT(21,2) | 8.9526  |
| SFDAT(22,1) | -23.200 |
| SFDAT(22,2) | 8.5711  |
| SFDAT(23,1) | -17.400 |
| SFDAT(23,2) | 7.6830  |
| SFDAT(24,1) | -12.200 |
| SFDAT(24,2) | 7.0426  |
| SFDAT(25,1) | -6.9000 |
| SFDAT(25,2) | 6.1530  |
| SFDAT(26,1) | -1.7000 |
| SFDAT(26,2) | 5.1141  |
| SFDAT(27,1) | 3.5000  |
| SFDAT(27,2) | 4.0372  |
| SFDAT(28,1) | 8.8000  |
| SFDAT(28,2) | 2.8179  |
| SFDAT(29,1) | 14.400  |
| SFDAT(29,2) | 1.4623  |
| SFDAT(30,1) | 20.000  |
| SFDAT(30,2) | .00000E+00 |
| SFDAT(31,1) | 25.200  |
| SFDAT(31,2) | -1.3115 |
| SFDAT(32,1) | 30.900  |
| SFDAT(32,2) | -2.4268 |
| SFDAT(33,1) | 36.300  |
| SFDAT(33,2) | -3.3505 |
| SFDAT(34,1) | 41.800  |
| SFDAT(34,2) | -4.0966 |
SFDAT(35,1) = 47.200
SFDAT(35,2) = -4.6153
SFDAT(36,1) = 52.500
SFDAT(36,2) = -5.1292
SFDAT(37,1) = 57.800
SFDAT(37,2) = -5.5328
SFDAT(38,1) = 63.000
SFDAT(38,2) = -5.9473
SFDAT(39,1) = 68.400
SFDAT(39,2) = -5.9758
SFDAT(40,1) = 73.700
SFDAT(40,2) = -6.3344

C********** END OF DATA **************

T(1) = TEMP - DTEMP
T(2) = TEMP
DO 40 I1 = 1, 2
DO 30 I2 = 1, NDAT - 1
TL = SFDAT(I2, 1)
TH = SFDAT(I2 + 1, 1)
AL = SFDAT(I2, 2)
AH = SFDAT(I2 + 1, 2)
IF (TL .GT. T(I1) .OR. TH .LT. T(I1)) GO TO 30
SF(I1) = AL + (AH - AL) * (T(I1) - TL) / (TH - TL)
SHIFT(I1) = 10**SF(I1)
GO TO 40
30 CONTINUE
40 CONTINUE
RETURN
END

*INITIAL CONDITIONS, TYPE=TEMPERATURE
BEAM, 70
*AMPLITUDE, NAME=COOL, TIME=A, VALUE=A
0, 70, 1800., -120.
*RESTART, WRITE, FREQUENCY=150
*STEP, INC=150, CYCLE=10
*VISC0, PTOL=.05, CETOL=1E-4
30., 1800.
*TEMPERATURE, AMPLITUDE=COOL
BEAM, -120.
*EL PRINT, POSITION=CENTROIDAL
S11
TEMP
*END STEP
APPENDIX B

This Appendix contains an example of an ABAQUS input file, for the Island Blister Test elastoplastic analysis.

*HEADING
IBT - ELASTIC/PLASTIC ANALYSIS
*PREPRINT, HISTORY=NO, MODEL=NO, ECHO=NO
*NODE
1, 0, 0
59, 0.390, 0
2002, 0, 0.0055
2060, 0.390, 0.0055
4003, 0, 0.0110
4061, 0.390, 0.0110
2001, 13.000, 0
4002, 13.000, 0.0055
6003, 13.000, 0.0110
*NGEN, NSET=ALL
1, 59
59, 2001
2002, 2060, 2
2060, 4002, 2
4003, 4061
4061, 6003
*NSET, NSET=LSUP, GEN
1, 59
*NSET, NSET=LSUP1, GEN
1, 57
*NSET, NSET=LEFT
1, 2002, 4003, LSUP
*NSET, NSET=RIGHT
2001, 4002, 6003
*NSET, NSET=LEFT1
1, 2002, 4003, LSUP1
*ELEMENT, TYPE=CAX8
1, 1, 3, 4005, 4003, 2, 2004, 4004, 2002
30, 59, 61, 4063, 4061, 60, 2062, 4062, 2060
*ELGEN, ELSET=SUP
1, 29, 2
*ELGEN, ELSET=MEMBRANE
30, 971, 2
*ELSET, ELSET=ALL
SUP, MEMBRANE
*ELSET, ELSET=NONE
  11
*ELSET, ELSET=SCOTT1, GEN
  1, 100, 1
*ELSET, ELSET=SCOTT2, GEN
  100, 1000, 25
*ELSET, ELSET=SCOTT
  SCOTT1, SCOTT2
*NSET, NSET=ONE
  59, 58
*SOLID SECTION, ELSET=ALL, MATERIAL=POLYMER
*MATERIAL, NAME=POLYMER
*ELASTIC, TYPE=ISO
  2300., 0.40
*PLASTIC
  50., 0.
  59.0, 0.0033
  68., 0.0133
  78., 0.0283
  87.02, 0.0483
  94.1, 0.0783
  101.3, 0.1283
  104.8, 0.1783
  122.1, 0.2783
  149.6, 0.3493
*RATE DEPENDENT
  5.63, 2.20
*AMPLITUDE, NAME=LOAD, TIME=A, VALUE=R
  1.0, 0, 1.5, 0.75, 7.5, 1
*BOUNDARY
  LEFT, 1, 2
  RIGHT, 1, 2
*INITIAL CONDITIONS, TYPE=STRESS
  ALL, 25.0, 0, 25.0, 0
*STEP, INC=1, CYCLE=5
*STATIC, PTOL=1E-1
  1., 1.
*EL PRINT, POSITION=CENTROIDAL, ELSET=NONE
*S NODE PRINT, NSET=ONE
U
*ENERGY PRINT
ALLSE
*END STEP
*STEP, NLGEOM, INC=100, CYCLE=15, ROTTOL=0.78, SUBMAX
**STATIC, PTOL=1E-1, DIRECT=NO STOP**
0.1, 0.5
**DLOAD, AMPLITUDE=LOAD**
MEMBRANE, P1, 0.0113
**EL PRINT, POSITION=CENTROIDAL, ELSET=NONE, FREQ=100**
S
**NODE PRINT, FREQ=100, NSET=ONE**
U
RF
**NODE PRINT, FREQ=100, NSET=LSUP1**
U
**ENERGY PRINT**
ALLSE
**END STEP**
**STEP, NLGEOM, INC=100, CYCLE=15, ROTTOL=0.78, SUBMAX**
**STATIC, PTOL=1E-1, DIRECT=NO STOP**
1.0, 6.0
**DLOAD, AMPLITUDE=LOAD**
MEMBRANE, P1, 0.0113
**EL PRINT, POSITION=CENTROIDAL, ELSET=NONE, FREQ=100**
S
**NODE PRINT, FREQ=100, NSET=ONE**
U
RF
**NODE PRINT, FREQ=100, NSET=LSUP1**
U
**ENERGY PRINT**
ALLSE
**END STEP**
**STEP, NLGEOM, INC=100, CYCLE=15, ROTTOL=0.78, SUBMAX**
**RESTART, WRITE, FREQUENCY=100**
**STATIC, PTOL=1E-1, DIRECT=NO STOP**
0.10, 1.
**BOUNDARY, OP=NEW**
LEFT1, 1, 2
RIGHT, 1, 2
**EL PRINT, POSITION=CENTROIDAL, ELSET=SCOTT, FREQ=100**
MISES
**NODE PRINT, NSET=ONE, FREQ=1**
U
RF
**ENERGY PRINT**
ALLSE
**END STEP**
APPENDIX C

Single edge notched strips of the polymer are used to measure its fracture toughness, according to the following procedure: (i) A silicon wafer is coated with the pre-polymer solution, by spinning. (ii) The polymer is cured, according to the manufacturers specifications. The thickness of the cured polymer is in the range of 5 to 20 μm, measured by profilometry. (iii) Polymer strips, 6 mm wide, with small notches, 1 mm long, are patterned into the polymer by photolithographic techniques. (vi) The strips are removed from the wafer by dipping it into an HF solution, and then are washed with water. (v) The strips are glued onto paper frames, as shown in Figure C-1. (vi) The paper frame is placed between the grips of the testing machine and then is cut at the locations indicated in Figure C-1. (vii) The test is performed.

For single edge notched strips, the fracture toughness can be calculated by the following formula [84]:

\[ K_I = \sigma \sqrt{\pi L \ f(L/B)} \]  \hspace{1cm} (C-1)

where, \( \sigma \) is the stress at which the strip brakes, and:

\[ f(L/B) = 0.265 \ (1-L/B)^4 + \frac{0.857 + 0.265 \ L/B}{(1-L/B)^{3/2}} \]  \hspace{1cm} (C-2)
Figure C-1. Sample mounted on a paper frame for fracture testing.
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BIOGRAPHICAL NOTE

Georgios Margaritis

The author is a native of Greece, as his long name ending in an s, implies. He was born in Athens in 1962, in a city with a glorious past and blue skies, smog forbidden, of course. He finished high-school in the city of Iraklion, on the island of Crete, where people live the words of Pericles “Happiness exists in Freedom and Freedom in Bravery”. He studied Civil Engineering at the University of Thessaloniki, from where he received his degree in 1987. Driven by the thirst for knowledge and new experiences, he came to the other “land of the free and home of the brave”, the United States. Actually, he received his MS in Civil Engineering in 1990 from the University of Illinois at Chicago, which is more like the home of the Cubs (or maybe the Bulls). Lured by the siren of the East Coast, Boston, he came to continue his graduate studies at the Massachusetts Institute of Technology. The professional sports teams in Boston might not win as often as those in Chicago, but, at least, the schools are better. He is currently finishing his Doctorate in Materials Science and Engineering, with major in Polymers. He has four publications, and is looking forward to a bright future.