MICROMATERIALS AND NANOMATERIALS

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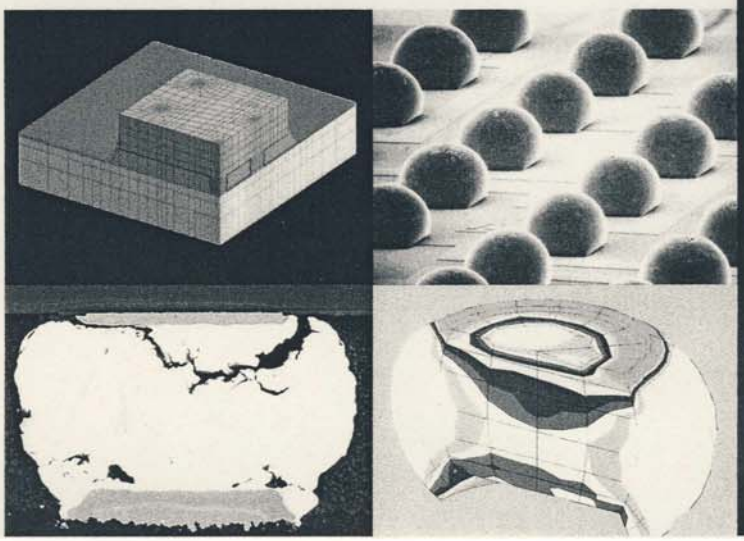
materials mechanics and packaging

dedicated to the memory of Andreas Schubert



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IMPRESSUM

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Dr. Andreas Schubert

1956 – 2003

I wish to thank all the colleagues and friends who have contributed to this booklet. It shows the recognition and respect you showed to Andreas.

He was an exceptional and affectionate man and I am grateful for having had the opportunity to live and work with him for such a long time. I miss him badly.

Ilona Soremski

A Severe Loss

In Memoriam Dr.-Ing. Andreas Schubert

Andreas Schubert was one of the internationally best-recognised materials scientists in the field of materials mechanics for electronic packaging from Germany. He suddenly, and unfathomably for all of us, died on the night of January 27th, 2003.

The Fraunhofer Institute IZM was the centre of his professional life. A professorship at the Technical University Berlin was due to be awarded to him, most of all for his remarkable achievements in the area of packaging techniques for microsystem technologies.

After his course of studies in materials science at the Mining Academy Freiberg, he finished his postgraduate studies there with a doctoral degree. For six years, he worked hard specialising on fracture mechanics and micromechanics at the Institute for Mechanics at the Academy of Sciences, Chemnitz, after which he assumed work at the Fraunhofer Insitute IZM in 1993, right when it was founded. This year, which sees the 10th anniversary of IZM – His institute, would also have been a year in which he could have looked back at 10 years of continuous achievement and success. The numerous excellent presentations, given at important international conferences are examples of these, just as much as the large number of papers published in journals with a wide international renown.

At the 50th ECTC, in Las Vegas, the probably internationally best-recognised conference series on electronic packaging he was awarded the prize of 'Outstanding Paper'. Above that, IEEE honoured him by elevating him to IEEE Senior Member. A variety of other awards and honours, though, were granted him for best papers at quite a range of scientific conferences.

Dr. Schubert was managing editor of the journal of Microsystem Technologies, as well technical chair of a number of international conferences on advanced packaging materials. As deputy of the department head of the IZM's department of Mechanical Reliability and Micro Materials as well as in his capacity of scientific director of the Micro Materials Center Berlin, his share in the scientific achievements and development of core competencies in thermomechanical reliability of the IZM was tremendous.

Having supervised many research projects, and also having coordinated the IZM research programme Reliability and Life-time Evaluation, and, more over, having been a partner for colleagues as well as customers, here and abroad, who showed commitment more than anything else. Therefore, Dr. Andreas Schubert's part in the growth of IZM just cannot be underestimated.

Building on his achievements, we will carry forward his work as we know he would have done. Thus honouring his professional life, will do our best to keep his memory alive.

Prof. Dr. Dr.-Ing. E. h. H. Reichl
Director of Fraunhofer IZM

Prof. Dr. rer. nat. habil B. Michel
Head of Micro Materials Center Berlin at IZM

Characterization and Modeling of Moisture Behavior of Electronic Packaging

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Abstract

This paper presents an overview of recent advances in characterization and modeling of moisture behavior of electronic packaging. Moisture diffusion modeling is discussed first, followed by the characterization of moisture related properties including diffusivity and hygroswelling properties. The emphasis was given on the characterization of interface toughness with moisture and temperature effects in nonlinear conditions. Then the micromechanics approach is introduced to study the vapor pressure evolution, void growth instability, and constitutive modeling.

1. Introduction

Ever since the discovery of the "popcorn" failure of plastic-encapsulated IC packages in the 1980s, much effort has been devoted to understand the failure mechanism. Fukuzawa et al. [1] first studied popcorn problems and assumed that the pad-encapsulant had fully delaminated and proceeded to determine the size of the pad and the thickness of the encapsulant in order to avoid package cracking under the combined action of vapor pressure and thermal loading. Later, Tay and Lin [2 – 5] conducted a series of work on the moisture diffusion and heat transfer in plastic IC packages, and studied the dynamics of moisture diffusion, hygrothermal stresses and delamination using interface fracture mechanics approach. Galloway et al. [6] contributed to the moisture diffusion modeling and characterization for various kinds of plactic materials. Wong et al. [7] followed Galloway's approach to propose an alternative variable for moisture diffusion modeling. Tee et al. [8 – 11] developed a fully integrated modeling approach to investigate the moisture behavior. Dudek et al. [35, 36] presented parametric studies on moisture diffusion and popcorn cracking.

Most of previous studies assumed that the delamination exists before reflow, and considered vapor pressure as traction loading subjected to the delaminated interfaces. It is still not clear how delamination initiates with the moisture and temperature effect. Fan and Zhang [12 – 16] recently introduced micromechanics approach to study the moisture behaviors and delamination initiation of electronic packaging. In this paper, the recent advances in characterization and modeling of moisture behavior of electronic packaging are presented. Moisture diffusion modeling is discussed first, followed by the characterization of moisture related properties including diffusivity and hygroswelling properties. The emphasis was given on the characterization of interface toughness with moisture and temperature effects in nonlinear conditions. Then the micromechanics approach is introduced to study

the vapor pressure evolution, void growth instability, and constitutive modeling.

2. Moisture Diffusion Modeling

Transient moisture diffusion follows

$$\frac{\partial^2 C^1}{\partial x^2}+\frac{\partial^2 C^2}{\partial y^2}+\frac{\partial^2 C^3}{\partial z^2}=\frac{1}{\alpha_D}\frac{\partial C}{\partial t} \qquad (1)$$

where C is the local concentration, x, y, z are coordinates, α_D is the moisture diffusivity, and t is the time. An interfacial concentration discontinuity will occur whereever two materials having different saturated concentrations are joined, as shown in Fig. 1. Therefore, a new variable that could enforce field continuity is required for modeling of moisture diffusion in a multi-material system.

Galloway et al. [6] defines a new variable $\varphi = C/S$ which is continuous along the bimaterial interface. S is the solubility defined by $S = C_{sat}/p_{ext}$, where p_{ext} is the ambient vapor pressure under the given humid conditions. Wong et al. [7] introduced an alternative variable w, the so-called 'wetness' defined by $w = C/C_{sat}$, which is also continuous along the interface.

It is pointed out by Fan et al. [15] that the two approaches are equivalent. When φ is used, we have

$$\varphi_1 = \varphi_2 \ , \quad \alpha_{D1}S_1\frac{\partial \varphi_1}{\partial n}=\alpha_{D2}S_2\frac{\partial \varphi_2}{\partial n} \qquad (2)$$

When w is used

$$w_1 = w_2 \ , \quad \alpha_{D1}C_{sat1}\frac{\partial w_1}{\partial n}=\alpha_{D2}C_{sat2}\frac{\partial w_2}{\partial n} \qquad (3)$$

where the subscripts 1 and 2 represent different materials, respectively. Tay and Lin [4] performed finite element modeling on moisture diffusion coupled with heat transfer in plastic IC packages. Tee et al. [8] conducted a 3-D finite element moisture diffusion modeling for PBGA package.

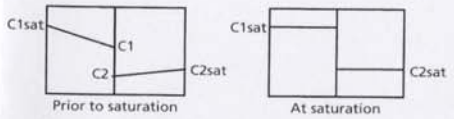

Fig. 1. Moisture concentration across bi-material interface

3. Characterization of Moisture Related Material Properties

3. 1. Diffusivity & Saturation Concentration Measurement

The Arrhenius equation can be used to describe the dependecy of diffusivity and saturation concentration on temperature as following, respectively

$$D = D_o e^{\frac{Q}{RT}} \ , \quad \psi = \frac{C_{sat}}{\rho_g} = \psi_0 \exp(Q_\psi / RT) \qquad (4)$$

Moisture absorption tests can be performed at different temperature levels to obtain the material constants. Galloway et al. [6] described the test procedures and provided data for various kinds of materials such as die attach, BT, mold compound and solder mask. Tee et al. [10] tested different kinds of underfill materials and used the test data for finite element modeling.

3.2. Hygroswelling Characterization

Material swells when moisture is absorbed. The change in dimension and weight can be related by the coefficient of moisture expansion β for the strain versus moisture concentration as following:

$$\varepsilon_h = \beta C \qquad (5)$$

where ε_h is the hygro strain, and C is the moisture concentration. The general hygroswelling characterization technique was developed in ref. [10]. In addition to hygroswelling, vapor pressure-induced expansion (swelling) is another factor to cause the stresses, which is discussed in refs. [9, 11]. Tay and Lin [3] studied the dynamics of moisture difussion, hygrothermal stresses and delamination in plastic IC packages. Tee et al. [9] developed an integrated approach to consider thermal

expansion, hygroswelling, as well as the vapor pressure induced expansion for the delamination in QFN packages.

3. 3. Interfacial Fracture Toughness Measurement

The development of interfacial fracture toughness tests is essential for the designing of efficient interfaces. Interface fracture toughness tests can be divided into two categories: fracture toughness tests using simplified specimen and in situ testing of microelectronic devices and their packages. Only if the plastic zone is small can interfacial fracture be characterized by interfacial toughness and the mode-mixity alone. Otherwise there will be crack growth resistance and the ideal interfacial toughness test would deliver the parameters that enable the actual fracture process to be characterized either by the fracture process zone model (EPZ) or dislocation free zone model (DFZ) [17]. The test should be capable of delivering the results for a range of known mode-mixities. Perhaps the most versatile test specimen is the four-point-bend mixed mode (see Fig. 2) which is either bilayer or a film sandwiched between two layers and its various verities such as the asymmetric double cantilever beam [18, 19].

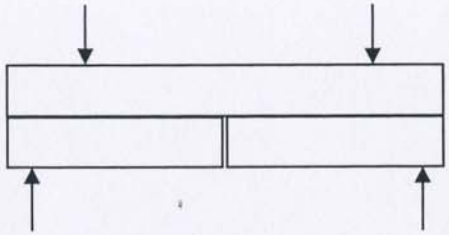

Fig. 2 A 4-point-bend mixed-mode specimen

The advantage of having a constant crack tip moment is that, if the plastic zone is small, propagation is steady state. The energy release rate which is independent of crack length, can be obtained from simple beam theory and along with mode-mixity has been given by Hutchinson and Suo [18]. If a thin film is sandwiched between two substrates, the residual stress does not affect the energy release rate because it is not released during the delamination. Charalambides et al. [20, 21] discussed some of the complexities of this test. Yao et al. [22] investigated the underfill/substrate interfacial toughness, in which a rigorous procedure to prepare the test specimen is discussed, and another configuration of four-point bend specimen is proposed.

Gurumurthy et al. [19] has used the asymmetric double cantilever beam to study a model interface between a model epoxy underfill and a polyimide coated dilycidyl ether of bisphenol-A (DGEBA). By varying the height ratio of the DGEBA to the underfill a range of mode-mixities was obtained.

Handling very thin films and introducing interfacial cracks can be a problem. Bagchi and his co-workers [23, 24] have devised a method of using the residual stresses in deposited thin metal film to cause the delamination. Often the residual stresses resulting from the deposition of a thin film will be insufficient to cause the delamination, but a chromium film can be deposited on top of the metal film that has high tensile residual stresses. This deposition can be performed at room temperature by electron beam evaporation and does not react with the underlying film. The thickness of chromium layer can be increased until the metal film delaminates.

With in situ interfacial toughness tests the mode-mixity cannot be easily controlled but they represent the actual interface of the microelectronic package. For measuring the interfacial toughness between the chip passivation and the underfill of flip chips, IBM has developed a test based on the compact tension specimen [25]. A precrack was introduced into the underfill. The purpose of the test was to study the effect of contamination by flux on the interfacial toughness and the effect of subjecting the chip to the pressure cooker accelerated test. Hence it was not important to control the mode-mixity and rigid arms were attached to the chip which do not allow the mode-mixity to be modeled. However, in principle it would be possible to use arms that were more flexible to control the mode-mixity.

Indentation methods of measuring the fracture toughness of thin films have long been used. Drory and Hutchinson [26] described a method of measuring the interfacial toughness of a brittle film on a ductile substrate with a conical brale indentor.

A ductile film on an elastic substrate may buckle if it delaminates under the compressive stress induced by the indentation. The possible delamination modes have been suggested by De Boer and Gerberich [27] for the micro-wedge indentation of ductile films that have been patterned into thin lines. G. Q. Zhang et al. [28] also studied the buckling driven interface delamination between a thin metal layer and a ceramic substrate.

4. Modeling of Vapor Pressure

It is important to understand how vapor pressure is evolved during temperature rise. One of critical issues in developing a vapor pressure model is to find out the moisture density in *voids*, denoted as ρ [12, 13, 14],

$$\rho = \frac{\mathrm{d}m}{\mathrm{d}V_f} \tag{6}$$

where dm is the mass of moisture per unit volume of free spaces in material, denoted by dV_f. The moisture concentration C is defined as

$$C = \frac{\mathrm{d}m}{\mathrm{d}V} \tag{7}$$

where dV is the element volume of the porous material, which contains free spaces dV_f. It should be noted that, because of the inhomogeneous character of a porous material, the element should be established over a (finite) representative volume, RVE [12].

Introducing the void volume fraction f according to

$$f = \frac{dV_f}{dV} \quad 0 < f \le 1 \tag{8}$$

the following relation between ρ and C can be obtained:

$$\rho = \frac{\mathrm{d}m}{\mathrm{d}V_f} = \frac{\mathrm{d}m}{\mathrm{d}V}\frac{\mathrm{d}V}{\mathrm{d}V_f} = C / f \tag{9}$$

The following condition is used to determine the moisture state in voids at preconditioning of temperature T_0,

$$\begin{cases} \rho \le \rho_g(T_0) & \text{for vapor phase at } T_0 \\ \rho > \rho_g(T_0) & \text{for mixed liquid/vapor phase at } T_0 \end{cases} \tag{10}$$

where ρ_g is the saturated vapor density, which can be obtained from the steam table as function of temperature.

When the moisture is at mixed liquid/vapor phase, it is necessary to know at which temperature the moisture can be *fully* vaporized. This temperature is called the *phase transition temperature*, denoted by T_1, which can be determined by

$$\rho(T_1) = \rho_g(T_1) \tag{11}$$

Now the vapor pressure in voids can be determined by the moisture state analyzed above. When the moisture is in the mixed liquid/vapor phase, the vapor pressure maintains the saturated vapor pressure p_g as function of temperature (from steam table), i.e.,

$$p(T) = p_g(T) \text{ for mixed/liquid vapor phase} \tag{12}$$

When the moisture is in single vapor phase, the ideal gas law can be followed to calculate the vapor pressure as following:

$$p\mathrm{d}V_f = \mathrm{d}mRT \text{ or } p = \rho RT,\ pf = CRT \tag{13}$$

where R is the universal gas constant (= 8.314 J /mol).

The two vapor-phase states (p,f,T,C) and (p_r,f_r,T_r,C_r) are then related by

$$\frac{p}{p_r} = \frac{Tf_rC}{T_rfC_r} \text{ for single vapor phase from } T_r \text{ to } T \tag{14}$$

Three distinct cases for the vapor pressure evolution have been identified [12], and are shown in Fig. 3.

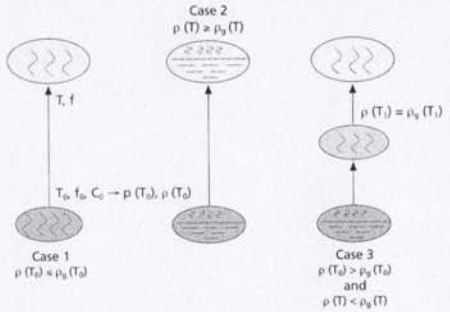

Fig. 3. Three distinct cases for vapor pressure evolution from the preconditioning temperature T_0 to the current temperature T.

The vapor pressure solution can be summarized as follows:

Case 1: when $C_0/f_0 \le \rho_g(T_0)$,

$$p(T) = \frac{C_0 p_g(T_0)}{\rho_g(T_0) f}\frac{T}{T_0}[1-3\alpha(T-T_0)] \qquad (15)$$

Case 2: when $\frac{C_0}{f}[1-3\alpha(T-T_0)] \ge \rho_g(T)$

$$p(T) = p_g(T) \qquad (16)$$

Case 3: when $C_0/f_0 > \rho_g(T_0)$,

and $\frac{C_0}{f}[1-3\alpha(T-T_0)] < \rho_g(T)$

$$p(T) = p_g(T_1)\frac{T}{T_1}\frac{f(T_1)}{f}\frac{1-3\alpha(T-T_0)}{1-3\alpha(T_1-T_0)} \qquad (17)$$

where T_1 is determined by equation (11).
The above model includes an unknown f, and the current void volume fraction. Obviously, the vapor pressure is dependent on the void deformation, and should be solved together with the governing equations of deformation.

5. Modeling of Void Growth Instability

The interface delamination is considered as the consequences of micro-voids growth, nucleation and coalescence. A micromechanics analysis of a single void is useful to reveal the fundamental failure mechanisms associated with the initiation of interfacial delamination. The use of single void model is also helpful to investigate the role of vapor pressure on void behavior.
As mentioned before, the vapor pressure model should be used together with the equations of deformation and constitutive relations to investigate material behavior during soldering. For the purpose of analysis a spherical volume of material containing a (concentric) spherically shaped micro-void is considered [13]. The material is considered incompressible. The inner radial surface is subjected to internal vapor pressure, induced by the moisture inside. A radial stress σ^T is applied to outer radius to represent the thermal stress as function of temperature rise. The neo-Hookean model is introduced to describe the deformation behavior of rubber-like material. The stored energy function can be written as follows:

$$W = \frac{\mu}{2}(\lambda_1^2+\lambda_2^2+\lambda_3^2-3),\ \lambda_1\lambda_2\lambda_3 = 1 \qquad (18)$$

where μ is the shear modulus and λ_i are the principal stretches. It can be seen that the shear modulus is the only material property introduced in this stress-strain relation. Guo and Cheng [29] established the equilibrium solution of a spherically symmetric cell in current configuration, which can be expressed explicitly in terms of the initial and current void volume fractions f_0 and f through,

$$\frac{\sigma^T(T)+p(f_0,f,C_0,T,T_0)}{\mu} = 2\left(\frac{1-f}{1-f_0}\right)^{1/3}+\frac{1}{2}\left(\frac{1-f}{1-f_0}\right)^{4/3}-2\left(\frac{f_0}{f}\frac{1-f}{1-f_0}\right)^{1/3}-\frac{1}{2}\left(\frac{f_0}{f}\frac{1-f}{1-f_0}\right)^{4/3} \qquad (19)$$

Equation (19) displays a nonlinear and non-monotonic relation between the applied stress (the sum of thermal stress and vapor pressure) and the void volume fraction f. This applied stress

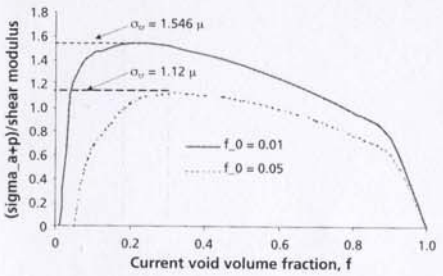

Fig. 4: The sum of thermal stress and vapor pressure applied to a cavity in a finite neo-Hookean rubber-like matrix versus the evolution of the void volume fraction f (f_0=0.01 and 0.05).

(relative to the shear modulus) versus the evolution of the void volume fraction f is shown in Fig. 4 [13]. The unstable void growth takes place when the peak value (of thermal stress + vapor pressure) is reached.
It is noted that using the neo-Hookean model according to equation (19), the critical stress is of the order of (1.1 ~ 1.6 times) the shear modulus μ. Assuming that $\mu = \sim E/3$, where at reflow temperature 220 °C, with a typical Young's modulus E = ~500 MPa, the critical stress is of order of magnitude 165 – 260 MPa. The saturated vapor pressure of 2.32 MPa at 220 °C which is very small compared to the critical stress (165 – 260 MPa) obtained for the rubber model.
From this observation it can be understood that the void unstable-growth is unlikely observed in bulk material. Although the moisture exists and is evaporated anywhere in polymer materials when the entire package is exposed to the reflow temperature at 220 °C, the rupture of bulk material prior to interface delamination can hardly happen. This means that above model applies to the void behaviors in bulk. In the following section, the model will be extended to investigate the void behaviors at interfaces.
The void behavior at interface is different from that within the bulk. The void growth at the interface is not only controlled by the total stress but also the interface strength. Fig. 5 sketches the void behavior at the interface, in which three stages are involved [13]. At the beginning, the void at the interface has an initial void volume fraction f_0, in which a certain amount of moisture is condensed into liquid. With the increase of temperature, thermal stress and vapor pressure are developed and subjected to the void. The void will reach the equilibrium at the void volume fraction f_1. Stage 1 shows no difference with behaviors of voids in bulk and the stress-level is much less than the critical stress at which the void will 'burst'. However, due to the fact that the interfacial strength is weaken by the moisture intake at high temperature, the void will continue to grow, as shown in stage 2 of Fig. 5. The new

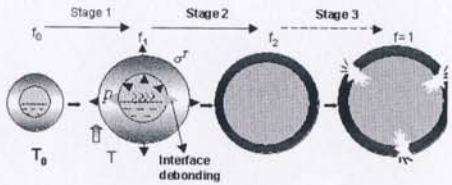

Fig. 5. Schematic description of the void behaviors at interface. Three stages are involved.

equilibrium will be reached at the void volume fraction f_2. At this new equilibrium position, the problem can be treated as the equilibrium for a void with the initial volume void fraction $(f_0+f_2-f_1)$. If the applied stress reaches the critical stress with $(f_0+f_2-f_1)$, the void growth becomes unstable (3rd stage). Otherwise, the void growth will stop here and no further delamination occurs. The void behavior at stage 2 is related to the interface properties. A general relationship between the void-growth and the moisture contents and the temperature may be postulated as following form:

$$\dot{f} = kCe^{-\frac{Q}{RT}}\dot{T} \qquad (20)$$

where C is the moisture concentration and T the temperature, R is the universal gas constant. K and Q are interface constants to be determined. It can be seen that the rate of void-growth is

propertional to moisture concentration C and temperature T. Some materials exhibit excellent resistance to moisture absorption with low C, while other materials show interface strengths being quite sensitive to moisture absorption with high k. There is no direct correlation between the amount of moisture absorbed and the void-growth at interface since different materials have different k. Some failures may occur for materials with minor moisture absorption (low C), but the material has very high k. Some other materials do not fail, even with major moisture absorption (very high C), due to the excellent resistance of interface strength at high moisture concentration and high temperature (very low k). However, for same material the correlation between the delamination and the moisture absorption is direct and obvious. Therefore, equation (20) is a generic form to describe the void behavior on interface.

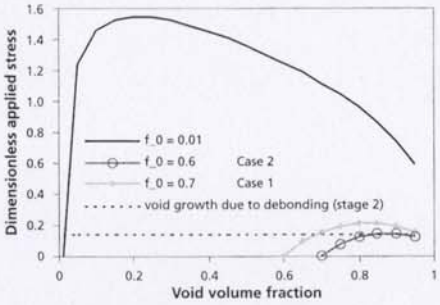

Fig. 6. void growth at interfaces

The exact determination of the material properties such as k is challenging. Instead of determining equation (20), in the following, equation (20) will be applied to see how much void growth in stage 2 will lead to the void unstable void growth. Let's assume that the initial void volume fraction is 0.01. At the beginning the void will deform along the solid line shown in Fig. 6. Then void growth enters second stage, in which the stress-level does not change but void grows following equation (20). This stage is shown in Fig. 6 as dotted line, which will intersect with another equilibrium curve. The void will not grow further if stress-level is still below the critical stress. Otherwise, like case 2 shown in Fig. 6, the delamination will take place.

6. Micromechanics-based Modeling

To link the results of the single void behavior to descriptions of material behavior in a macroscopic scale is a critical issue. Homogenization processes can be applied for this purpose. There are several theories to establish the relationships between the microscopic and macroscopic variables [30, 31] for porous material, in which the void volume fraction f is treated as a *field* variable: a damage parameter to represent the local material behavior. $f = 1$ at a particular (continuum) point implies that delamination takes place at this 'point'. The evolution equation is required for the void volume fraction f. For homogeneous material in bulk, the growth rate can be written as [13]

$$\dot{f} = \dot{f}_{\text{growth}} + \dot{f}_{nucleation} \tag{21}$$

$$\dot{f}_{\text{growth}} = (1-f)\dot{E}_{kk} \tag{22}$$

$$\dot{f}_{nucleation} = A\dot{\sigma}_e + B\dot{\Sigma}_m \tag{23}$$

At interface, the impact of interface on void growth should be included [13]

$$\dot{f} = \dot{f}_{\text{growth}} + \dot{f}_{nucleation} + \dot{f}_{debonding} \tag{24}$$

$$\dot{f}_{\text{growth}} = (1-f)\dot{E}_{kk} \tag{25}$$

$$\dot{f}_{nucleation} = A\dot{\sigma}_e + B\dot{\Sigma}_m \tag{26}$$

$$\dot{f}_{debonding} = kCe^{-\frac{\Omega}{kT}}\dot{T} \tag{27}$$

Gurson [30] assumes that the matrix material follows the classical elastic-plastic flow rule with Von-Mises yielding criterion. He established the relationship between the macroscopic and microscopic variables by using the averaging method over a cell containing a single void. Finally the macroscopic plastic potential that represents the yielding condition has the form

$$\Phi = \left(\frac{\Sigma_e}{\sigma_e}\right)^2 + 2f\cosh\left(\frac{3\Sigma_m}{2\sigma_e}\right) - (1+f^2) = 0 \tag{28}$$

in which Σ_e denotes Mises equivalent macroscopic stress, Σ_m the mean macroscopic stress, σ_e the current matrix flow strength of matrix and f the current void volume fraction. Equation (28) shows the effect of mean stress and void volume fraction on the material's yielding. Tvergaard [32] improved the model predictions for periodic arrays of cylindrical and spherical voids by introducing two factors q_1 and q_2, as following

$$\Phi = \left(\frac{\Sigma_e}{\sigma_e}\right)^2 + 2q_1 f\cosh\left(\frac{3q_2\Sigma_m}{2\sigma_e}\right) - (1+q_1 f^2) = 0 \tag{29}$$

The original Gurson-Tvergaard model requires each finite element to be modeled as porous material anywhere over a given structure. This usually gives rise to the difficulties in numerical implementation. Also, the model is very sensitive to the element size. The concept of *cell model* was first introduced by Xia and Shih [33] to tackle this problem. In the cell model, only a material layer of characteristic thickness D is modeled by Gurson-Tvergaard relation (see Fig. 7). Beyond this region, the conventional material without voids is applied. Therefore, it can be assumed that voids are present only in the material layer from the very beginning. This model has advantage that each cell behaves as a basic material unit containing a void and can be considered as a representative volume element pertaining to the specific material considered. The discrete, three-dimensional nature of a cell enables it to capture the important features from the crack formation to the propagation of a macro-crack. Since the Gurson-Tvergaard model is not able to capture the coalescence phase, the cell model will use a linear traction-separation law [32] to supersede the Gurson model when the void volume fraction reaches a critical value.

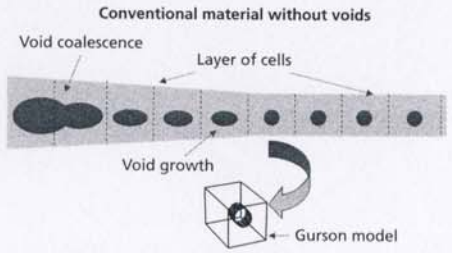

Fig. 7. Schematic of the cell model that is used to model a special material layer only such as an interface layer.

The cell model has recently been extended to model the interface delamination in plastic IC packages [34]. The vapor pressure effect is investigated by the comparison between the baked and unbaked packages. The vapor pressure model by equation (15) for case 1 is used. However, the initial vapor pressure is assumed to be as high as 2.6 times of yielding strength. According to equation (15), the initial pressure is very small and can never exceed $5.27e^{-2}$ MPa.

One of concerns in using the Gurson-Tvergaard model is its validity for the polymer materials. Thermoset materials behave like the rubber-like at high temperatures. Thermoplastic materials behave more like a viscous fluid or visco-elasto-plastic. Nevertheless, this mechanism-based approach provides insights into the failure of plastic packages arising from thermal and vapor pressure effects in the initiation of micro-voids, void growth, and the coalescence of voids. A specific micro-mechanics model for porous polymer materials with moisture effect is another subject for future study.

7. Discussions

Moisture absorption has been long recognized as the root cause for the delamination failure at reflow. However, one of important conclusions from the vapor pressure modeling is that the maximum vapor pressure at reflow is not always proportional to the moisture absorption. The

vapor pressure maintains the saturated value (e.g. = 2.32 MPa at 220 °C) no matter how much moisture is absorbed, as long as the moisture is not fully vaporized. In fact, the interface delamination not only depends on the vapor pressure, but also on the interface strength as well. When the vapor pressure maintains its saturated value, the interface strength becomes a key factor for the delamination.

The correlation between the interface strength and moisture absorption is complicated. Some materials exhibit excellent resistance to the moisture absorption, while some materials' interface strength is very sensitive to the moisture absorption. Therefore, there is no direct correlation between the amount of moisture absorbed and the failures when different materials are selected. Some failures may occur for the materials with less moisture absorption, since the adhesion of these materials is weakened significantly with moisture absorption. While others do not fail, even with significant moisture intake due to the excellent resistance of interface strength against the moisture and temperature. For the same material, however, the correlation between the delamination and the moisture absorption is direct and obvious.

It seems not appropriate to use the moisture absorption as criterion to evaluate the material's performance at reflow. The interface adhesion after moisture absorption at high temperature becomes one of most important indicators to identify the failures. The characterization and definition of the interface strength at high temperature with moisture absorption is, however, somehow ambiguous, when different methodologies and measurement techniques are applied. Nevertheless, it is noted that the interface strength at high temperature is a *comprehensive* property. Thermal stress and vapor pressure come with the temperature rise automatically and hence are the 'built-in' stresses. In other words, the interface strength measured at high temperature after preconditioning includes the effects of thermal stress and vapor pressure to a certain degree already (thermal stresses in test specimen is different from those developed in actual packages).

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