Characterization and Prediction of Fracture within Solder Joints and Circuit Boards

by

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Abstract

Double cantilever beam (DCB) specimens with distinct intermetallic microstructures and different geometries were fractured under different mode ratios of loading, $\psi$, to obtain critical strain energy release rate, $J_c$. The strain energy release rate at crack initiation, $J_{ci}$, increased with phase angle, $\psi$, but remained unaffected by the joint geometry. However, the steady-state energy release rate, $J_{cs}$, increased with the solder layer thickness. Also, both the $J_{ci}$ and $J_{cs}$ decreased with the thickness of the intermetallic compound layer.

Next, mode I and mixed-mode fracture tests were performed on discrete ($l=2$ mm and $l=5$ mm) solder joints arranged in a linear array between two copper bars to evaluate the $J = J_{ci} (\psi)$ failure criteria using finite element analysis. Failure loads of both the discrete joints and the joints in commercial electronic assemblies were predicted reasonably well using the $J_{ci}$ from the continuous DCBs. In addition, the mode-I fracture of the discrete joints was simulated with a cohesive zone model which predicted reasonably well not only the fracture loads but also the overall load-displacement behavior of the specimen. Additionally, the $J_{ci}$ calculated from FEA
were verified estimated from measured crack opening displacements in both the continuous and discrete joints.

Finally, the pad-crater fracture mode of solder joints was characterized in terms of the $J_{ci}$ measured at various mode ratios, $\psi$. Specimens were prepared from lead-free chip scale package-PCB assemblies and fractured at low and high loading rates in various bending configurations to generate a range of mode ratios. The specimens tested at low loading rates all failed by pad cratering, while the ones tested at higher loading rates fractured in the brittle intermetallic layer of the solder. The $J_{ci}$ of pad cratering increased with the phase angle, $\psi$, but was independent of surface finish and reflow profile. The generality of the $J = J_{ci}(\psi)$ failure criterion to predict pad cratering fracture was then demonstrated by predicting the fracture loads of single lap-shear specimens made from the same assemblies.
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I would like to dedicate this thesis to my wife Sridevi and my father P Subba Raju.
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Chapter 1

1 Introduction

1.1 Motivation

Solder joints act as electrical and mechanical interconnects between integrated circuit (IC) chips and printed circuit boards (PCBs) in electronic devices. Failure of these devices is often due to cracking in the solder or at the solder-pad interface under various thermal and mechanical loading conditions during assembly, testing, or service.

The majority of the research in this area has dealt with thermal fatigue. However solder joint failure under drop impact and vibration loading has also been of great interest, and the mechanical strength of solder joints has become an important parameter for the reliable performance of surface mount electronic devices. This is especially true in higher density array packages, larger printed circuit boards, and electronic devices for aerospace and automotive applications, where mechanical loads can be a significant cause of failure during service and assembly [1-3]. However, relatively little attention has been paid to the development of methods of predicting the fracture load of solder joints under mechanical loads, applied either directly on components or induced by the bending or twisting of printed circuit boards (PCBs).

Most existing experimental methods to evaluate the strength of solder joints under mechanical loads [2-7] are primarily qualitative and do not provide fundamental mechanical properties such as the strain critical energy release rate, $J_c$, that can be used to predict the failure load of joints in other configurations or loads. Although some attempts have been made to measure fundamental properties (fracture and strength properties) that govern solder joint failure [8-12], these studies are incomplete and have been purely experimental with no failure load predictions. Also, these studies did not address some key issues such as the R-curve behavior of solder joints and the relationship between solder microstructure and the measured fracture
parameters. Furthermore, most of these earlier studies have been focused on SnPb solder which is currently being replaced by lead-free solders in almost all the electronic devices.

This conversion to lead-free electronics has also brought changes to other materials such as PCBs. The epoxy-based PCB laminates that are compatible with higher lead-free reflow temperatures can be more brittle than earlier materials, raising new reliability issues. Among these is the increased propensity for PCB surface epoxy cracking beneath the copper pads of solder joints, also known as pad cratering [13]. This mode of failure has been observed widely in both low and high strain rate mechanical loading conditions, such as the quasi-static bending of PCBs and board-level drop tests [14,15]. At present, there are no widely accepted standards for fracture testing of PCBs and pad-crater cracking is generally assessed using qualitative tests [16,17] that have limitations similar to the earlier mentioned qualitative tests.

Fracture-based criteria such as the critical energy release rate, $J_c$, as a function of the mode ratio of loading, $\psi$, have been used widely to predict failure in adhesive joints [18-20]. In this case, fracture is predicted when the applied strain energy release rate at the particular mode ratio of loading (calculated from the applied loads, specimen dimensions and mechanical properties) equals the critical value for fracture at that mode ratio. A similar approach was adopted in this thesis to treat the solder joint failure: both cracking within the solder and the pad-crater fracture. The fracture behavior of the joints was investigated and properties were measured for each failure mode, and then these properties were used in finite element analysis to predict the failure loads.

1.2 Objectives

The main objectives of the thesis were:

1. To understand how different processing, geometric, and loading factors (mode ratio, $\psi$) influence the fracture of solder joints and to measure the critical energy release rate of fracture as a function of the mode ratio, $J_c(\psi)$ or $G_c(\psi)$.

2. To carry out mode-I and mixed-mode fracture tests on model solder joints which mimic real solder joints in electronic packages, and evaluate different failure criteria such as the critical strain energy release rate as a function of the mode ratio, $J_{ci}(\psi)$, and cohesive zone
models with finite element (FE) models. To test commercial microelectronic packages, and predict the fracture of solder joints in these packages using the same method.

3. To characterize the pad-crater fracture of commercial lead-free assemblies in terms of the critical strain energy release rate, $J_{ci}$, measured at various mode ratios, $\psi$. To demonstrate the predictive capability of this fracture criterion by measuring the fracture loads of test specimens made from these same PCB-CTBGA assemblies and comparing with the strength predictions made using the FEA and the measured $J_{ci}(\psi)$ fracture envelope.

1.3 Thesis Outline

Chapter 2 and 3 describe work done as part of the first objective; i.e., to study the fracture behavior of solder joints. Chapter 2 presents the double cantilever beam (DCB) specimen preparation and fracture testing methods established for measuring the fracture toughness of solder joint systems. R-curves were measured from the DCBs made with copper and lead-free solder (96.5Sn3Ag0.5Cu, SAC305) under a range of standard industrial processing conditions, and the results are presented. These experiments examined the relationship between the crack length, the critical fracture energy, the solder joint microstructure, the mode ratio of loading, and the crack path and fracture surface morphology. Such R-curve data for solder joints has never been published before. Also, measuring this data at different mode ratios reflecting the range of application of solder joints in actual components was another unique contribution. Finally, this chapter also presents the effect of the local geometry of the leading edge of the solder layer on crack initiation.

A major portion of Chapter 2 was published as a research article in Materials Science and Engineering A and was presented at an international conference.

Solder joints in electronic devices range in thickness from approximately 500 μm to less than 100 μm, and are used to join various substrates having different bending stiffnesses. These

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1 Nadimpalli SPV, Spelt JK. R-curve behaviour of Cu-Sn3.0Ag0.5Cu solder joints: Effect of mode ratio and microstructure. Mater Sci Eng A 2010; 527: 724-34.

factors can, in principle, affect the stress state at a crack tip and alter the effective fracture toughness of the joint. Hence, the experimental procedures established in Chapter 2 were used in Chapter 3 to examine the effect of solder joint thickness and substrate stiffness on the R-curve behavior of SAC305 lead-free solder joints under different mixed-mode loading conditions. The crack paths and fracture surfaces of the specimens were explained using an elastic-plastic finite element model. Further, some preliminary tests were performed to get a qualitative understanding of the effect of loading rate on fracture properties. This chapter was submitted for publication as an article in *Engineering Fracture Mechanics*\(^3\). The observations of Chapters 2 and 3 are relevant to the prediction of fracture in both small joints such as BGAs in microelectronic applications and relatively large solder joints such as those used to connect power electronics modules.

Chapter 4 examines two different solder joint fracture criteria that would permit the prediction of solder joint failure for a wide range of joint geometries and types of load. The first part of this chapter presents fracture experiments on Cu-SAC305 DCB specimens under mode-I loading to measure both \(G_{ci}\) and cohesive zone model (CZM) parameters. Some of the \(G_{ci}\) data presented here was from Chapter 2 and 3. In the second part of Chapter 4, mode-I fracture tests were performed on discrete 2 mm and 5 mm solder joints arranged in a linear array between two copper bars, in order to understand the fracture behavior of individual solder joints and to evaluate the \(G_{ci}\) and CZM failure criteria using linear elastic finite element analysis. This chapter was published as a research paper in *Engineering Fracture Mechanics*\(^4\). This is probably the first paper which has presented and validated the predictive capability of solder joint fracture criterion for joint strength prediction. Some of the contents of this chapter were also presented at an international conference\(^5\).

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\(^3\) Nadimpalli SPV, Spelt JK. Effect of geometry on the fracture behaviour of lead-free solder joints. Article submitted to Engng Fract Mech on April 2010.


\(^5\) Nadimpalli SPV, Spelt JK. Predicting the strength of solder joint using cohesive zone modeling. SMTA international conference on soldering and reliability, Toronto, May 19-22, 2009.
Chapter 5 verifies the validity of critical strain energy release rate criteria for a range of mixed-mode loading conditions. The first part of the chapter presents mixed-mode fracture experiments on Cu DCB specimens joined with a continuous SAC305 solder layer in order to measure the fracture parameters corresponding to crack initiation; i.e. the critical strain energy release rate as a function of the mode ratio of loading, $G_{ci}(\psi)$ and $J_{ci}(\psi)$, where $\psi$ is the phase angle of loading. In the second part of Chapter 5, fracture tests were performed on discrete 2 mm and 5 mm solder joints arranged in a linear array between two copper bars to evaluate the proposed failure criterion, $J_{ci}(\psi)$, using elastic-plastic finite element analysis. Finally, the $J_{ci}$ values calculated using FEA were validated by comparing them with the $J$-integral values estimated from the measured critical opening displacements near the location of crack initiation. The contents of this chapter were published as a research article in *Engineering Fracture Mechanics*. The same procedure was also extended to predict the fracture of solder joints in a commercial plastic ball grid array (PBGA) package-PCB assembly. This work was presented at an international conference.

Chapter 6 presents the pad-crater fracture prediction methodology. The pad crater failure was characterized in terms of the critical strain energy release rate, $J_{ci}$, measured at various mode ratios, $\psi$. Fracture specimens were prepared from a commercial PCB rated for lead-free assembly and assembled with “chip array thin core ball grid array” (CTBGA) packages. The specimens were fractured at low and high loading rates in various bending configurations to generate a range of mode ratios. $J_{ci}$ for pad cratering was calculated from the measured fracture strength and specimen deformation using a linear elastic finite element analysis (FEA). The predictive capability of the approach was then demonstrated by measuring the fracture loads of single lap-shear specimens made from these same PCB-CTBGA assemblies and comparing with

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7 Nadimpalli SPV, Spelt JK. A geometry and size independent failure criterion for fracture prediction in lead-free solder joints. SMTA international conference on soldering and reliability, Toronto, May 18-20, 2010.
the strength predictions made using the FEA and the $J_{ci}(\psi)$ failure criterion. This work was submitted for publication as a research paper in *Microelectronic Reliability*\(^8\).

Finally, the Chapter 7 presents the conclusions and recommended future work.

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\(^8\) Nadimpalli SPV, Spelt JK. Prediction of pad cratering fracture at the copper pad-printed circuit board interface. Submitted for publication in Microelectronics Reliability 2010.
1.4 References


Chapter 2

2 R-curve Behavior of Cu-Sn3.0Ag0.5Cu Solder Joints: Effect of Mode Ratio and Microstructure

2.1 Introduction

Microelectronic packages experience various thermal and mechanical loading conditions during their assembly, testing and service phases. Although most research has focused on solder joint failure due to thermal fatigue, the mechanical strength of solder joints is also an important performance parameter in many devices. This is especially true in higher density array packages, larger printed circuit boards, and electronic devices for aerospace and automotive applications, where mechanical loads can be a significant cause of failure during service and assembly [1-3]. In general, a high solder joint strength is considered as one indicator of reliability over the service life and strength tests such as ball shear and board bending are used widely in the industry [1-3].

Microelectronic package failures are often due to cracking in the solder or at the solder/pad interface under mixed-mode loading conditions; i.e. when the solder is subject to loads that create a combination of a tensile, opening-mode strain energy release rate, $G_I$, as well as a shear mode value, $G_{II}$. The mode ratio of loading is commonly defined by the phase angle, $\psi = \arctan\left(\sqrt{G_{II} / G_I}\right)$; i.e. $\psi = 0^\circ$ is pure mode I while $\psi = 90^\circ$ is pure mode II. Early work on solder joint fracture [4-7] focused mainly on understanding the mode I fracture behavior of SnPb solder joints. Subsequent studies [8 - 13] provided an understanding of solder fracture under various mixed-mode conditions. For example, Choi and co-workers [10, 11] carried out theoretical and experimental fracture studies on brass/solder/brass sandwich specimens with 0.5 and 2 mm solder layer thicknesses. They observed that fracture toughness increased with solder layer thickness and that the fracture toughness of the 0.5 mm layer did not change significantly for the range of mode ratios considered ($-25^\circ$ to $5^\circ$). Siow and Manoharan [8, 13] measured
mode I-III fracture energies of SnPb and SnAg solder joints with bare copper and nickel-plated substrates. They found that the latter specimens were tougher than the bare copper ones, suggesting that intermetallics play a significant role in fracture, and that the SnPb joints were tougher than the SnAg joints in mode I. Mixed mode I-III loading had a lower fracture toughness compared with the mode I values.

In all of these studies [5-13], the authors used fatigue pre-cracking to create an initial sharp crack tip for subsequent fracture toughness testing, and with the exception of [5], made a measurement at only one crack length. Although this procedure creates a uniform starting condition for a single measurement, it neglects the possibility that toughness may depend on crack propagation length due to damage zone development at the crack tip; i.e. R-curve behavior as illustrated in Fig. 2.1. Logsdon et al. [5] did measure the R-curve of SnPb solder in mode I, but this was for a bulk specimen comprised only of solder rather than a joint. The fracture toughness of solder in a joint is quite different from that in the bulk because of the constraint within a joint [6, 10].

![Diagram of R-curve and DCB specimen](image)

Fig. 2.1 Schematic of the R-curve and DCB specimen. The overall length of the specimen was 160 mm. All dimensions in mm.
The main objective of the present chapter was to examine the crack growth toughening that may give rise to R-curve behavior in solder joints. This can be relevant in predicting the failure of both small joints such as BGAs in microelectronic applications and relatively larger solder joints such as those used to connect power electronics modules, larger capacitors and heat sink attachments. In the latter cases, it is hypothesized that cracks can form at a critical initiation strain energy release rate, $G_{ci}$, then grow stably for at least several millimeters (until $a_{LS}$ in Fig.2.1 which is rising slope length) as the damage zone develops and the toughness increases, creating the rising part of the R-curve. After a certain amount of subcritical crack growth, the critical strain energy release rate reaches a steady-state maximum value, $G_{cs}$. However, in smaller solder joints, failure would be governed by crack initiation at $G_{ci}$ since the joint is too short for appreciable crack growth.

Depending on the temperature of soldering and subsequent aging, two different types of intermetallic compounds (IMCs) form near a solder-copper interface; namely, Cu$_6$Sn$_5$ and Cu$_3$Sn [14-16]. Solder joints manufactured under typical conditions contain mainly Cu$_6$Sn$_5$, while Cu$_3$Sn is generally formed during aging in service or under very high soldering temperatures. Although the formation of these IMCs is essential for a strong solder joint, a number of studies have shown that excess IMC thickness can reduce solder joint strength because of the brittle nature of these compounds [17, 18]. Hwa-Teng et al. [17] carried out tensile and shear strength tests on Sn3.5Ag solder butt joints formed between two copper wires of 1 mm diameter. They found that both the tensile and shear strengths of the joint decreased with the IMC layer thickness. They also noticed that the IMC layer thickness significantly influenced the fracture morphology. Shin et al. [18] also observed that the shear strength of Cu-Sn solder balls decreased with a thicker IMC layer. Pratt et al. [19] confirmed this weakening with increasing IMC layer thickness using a mode-I chevron notch Sn-Pb solder joint fracture specimen with copper substrates. The same group of authors also discovered that the fracture initiation strength was strongly dependent on the surface roughness [20]. A recent fracture study by Hayes et al. [21] on modified compact tension specimens with Sn4Ag0.5Cu and Sn0.7Cu solders on Ni–Au, Ni–Pd, and Cu substrates resulted in similar findings. The importance of the IMC thickness and composition on fracture strength implies that meaningful data can only be obtained from fracture test specimens manufactured under realistic time-temperature conditions.
Consequently, the present chapter measured the R-curves of joints made with copper and lead-free solder (96.5Sn3Ag0.5Cu, SAC305) under a range of standard industrial processing conditions. The experiments examined the relationship between the crack length, the critical fracture energy, the time above liquidus, the mode ratio of loading, and the crack path and fracture surface morphology.

2.2 Experimental Procedures

2.2.1 Specimen Preparation

Figure 2.1 depicts the double cantilever beam (DCB) specimen which consisted of two copper bars joined with a 400 µm thick layer of SAC 305 solder. The Cu bars (C110 alloy) were cut to the required dimensions and the bonding surfaces were polished for 5 min using an orbital sander fitted with an ultra fine silicon carbide/nylon mesh abrasive pad. To avoid edge rounding, eight bars were placed adjacent to each other and sanded simultaneously. This process produced a repeatable surface roughness of \( R_a = 0.95 \mu m \), measured using an optical profilometer. This was very close to the average \( R_a \) of 1 µm that was measured on an organic solderability preservative (OSP) finish on two different commercial PCBs.

After polishing, the Cu bars were rinsed thoroughly with water, wiped with cheese cloth, and then rinsed with acetone. The surface areas where soldering was not required were masked with Kapton tape, taking care not to contaminate the cleaned surfaces to be soldered.

The prepared copper bars were then placed on a hot plate covered with aluminum foil and maintained at 290°C with the bonding surfaces vertical (Fig. 2.2). The temperature of the copper bars was monitored continuously with thermocouples inserted in holes drilled just beneath the surfaces to be soldered. When the temperature of the bars reached 220-225°C, a flux-cored SAC 305 0.75 mm solder wire (Kester Inc., USA) was touched to the prepared vertical surfaces so that they became rapidly covered with a very thin layer of solder. The bars were then clamped together against two 400 µm steel wires to maintain the desired spacing (Fig. 2.2). This procedure minimized voiding caused by flux entrapment as excess solder and flux residues flowed out of the joint as the bars were brought together. The entire soldering process from the time of first solder application took approximately 15 to 20 s. It was followed by a further 30 to
210 s dwell period on the hot plate, depending on the desired time above the solder liquidus temperature of 220°C (TAL). The specimens were then placed transversely in a small wind tunnel and cooled with forced air at a cooling rate of 1.4-1.6°C/s, which is typical of microelectronics manufacturing.

Fig. 2.2 Schematic of specimen arrangement during soldering. Excess solder was melted in the initial 2-3 mm gap before clamping both the bars against the steel wires.

Specimens were prepared with similar peak temperature and cooling rates but three different TALs (Fig. 2.3). After cooling to room temperature, the specimens were machined to remove the excess solder and to obtain the final dimensions. To prevent the copper from smear over the interfaces and obstructing the observation of cracks, machining was performed with small cutting depths, employing a very sharp tool. This also reduced the heating of the specimen. The loading pin holes were drilled in the copper bars taking care to ensure that the axes of the loading pin holes lay in a single plane perpendicular to the length of specimen and the solder plane to avoid twisting of adherends during testing. Coolant was used to prevent
excessive heating of the specimen during these machining and drilling operations. In order to assess the dependence of the fracture initiation load on the local geometry of the end of the solder layer, some specimens (TAL240 profile) were prepared with three different end geometries as shown in Fig. 2.4. These were formed using either Kapton tape (smooth square edge), a band saw cut (rough square edge) or the steel spacing wire (round smooth edge).

![Time-temperature profiles of the three specimen types.](image)

**Fig. 2.3** The time-temperature profiles of the three specimen types. Time above liquidus (TAL) measured in seconds.
2.2.2 Fracture Testing

The DCB solder joints were tested under mode-I and mixed-mode conditions using the load jig shown in Fig. 2.5 [22]. This had the advantage that the mode ratio of loading could be selected simply by adjusting the location of the link pins so that the forces on the upper and lower arms could be varied independently. As the load jig is statically determinate, the specimen forces can be calculated from the equilibrium considerations as

\[ F_1 = F \left(1 - \frac{s_1}{s_3}\right) \]  

\[ F_2 = F_1 \frac{s_1}{s_2} \left(\frac{1}{1 + \frac{s_3}{s_4}}\right) \]

where, \(s_1, s_2, s_3,\) and \(s_4\) are the distances between pin centers (Fig. 2.5), and \(F\) is the force applied to the load jig. \(F_1\) and \(F_2\) are the forces on the upper and the lower adherends of the specimen, respectively. Note from Eq. (2) that a given load jig configuration (i.e. set of pin locations) results in a constant \(F_2/F_1\) ratio that is independent of specimen geometry, crack length, and the applied load \(F\). Thus a single DCB geometry could be used to obtain the three mode ratios that were used, corresponding to phase angles of 0° (mode I), 25° and 45°.
The visible edge of the solder layer was painted with a thin layer of diluted paper correction fluid to facilitate the identification of the crack tip. A microscope on a micrometer stage with a field diameter of 1.9 mm was used to measure the crack length. The load on the specimen was increased with a constant cross-head speed of 0.1 mm/min until a critical load was reached when crack extension occurred. Crack initiation from the end of the solder layer in an uncracked specimen was defined as occurring at the load causing crack extension of 50-100 μm. In very few cases at higher mode ratios ($\psi = 45^\circ$), microcracks were first seen approximately 500 to 600 μm from the end of solder layer. In these situations, a 50-100 μm macro crack would form at the end of the solder layer almost immediately after the appearance of the microcracks, and hence the same definition of crack initiation was used in all cases. Since the load frame was stopped when crack growth occurred, the fracture was stable and a single specimen could be used to measure approximately 30 or more crack growth sequences, each providing a measure of $G_c$ at progressively longer crack lengths. A minimum of four DCB specimens were tested for each combination of TAL, end condition and mode ratio.

Fig. 2.5 Schematic of the DCB specimen mounted in the mixed-mode load jig [22].
2.2.3 $G_c$ Calculations

The fracture tests provided many measurements of the critical load corresponding to the onset of crack growth at increasing crack lengths, which were used to calculate the critical strain energy release rate of the solder joint system using the beam-on-elastic-foundation model of [23]. These equations were derived under the assumption that the adherends deform elastically. The expression for the energy release rate is given by,

$$G_c = \frac{12a^2}{E(h-t)^3} \left[ f_1^2 \Phi_I^2 + \frac{3}{4} f_2^2 \Phi_{II}^2 \right]$$

(3)

where $f_1$ and $f_2$ are the mode I and mode II critical loads per unit width of specimen, $\Phi_I$ and $\Phi_{II}$ are defined by

$$\Phi_I = 1 + 0.667 \frac{h}{a} \left[ 1 - \frac{t}{h} \right]^{3/4} \left[ 1 + \frac{t}{h} \left( \frac{2E_a}{E_{a}} - 1 \right) \right]^{0.25}$$

$$\Phi_{II} = 1 + 0.206 \frac{h}{a} \left[ 1 - \frac{t}{h} \right]^{1/2} \left[ 1 + \frac{2tE_a}{G_s h} \right]^{1/2}$$

(4)

and $a$, $h$ and $t$ are as defined in Fig. 2.1. $G_s$ and $E_a$ are, respectively, the solder shear and tensile modulus, and $E$ is the tensile modulus of the copper bars. The symbol $\alpha$ is a calibration constant equal to 2.946 [23].

The mechanical properties of Cu (C110 alloy) and Sn3Ag0.5Cu used in the present study are listed in Table 2.1. The applied mode ratio (phase angle) of loading, $\psi$, in each test was calculated as [23]
\[
\psi = \arctan \left( \frac{\sqrt{3}}{2} \frac{f_2 \Phi_{ll}}{f_1 \Phi_i} \right) \tag{5}
\]

Table 2.1 Mechanical properties of copper and solder [17].

<table>
<thead>
<tr>
<th></th>
<th>Tensile modulus (GPa)</th>
<th>Poisson ratio</th>
<th>Shear modulus (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cu C110</td>
<td>124</td>
<td>0.35</td>
<td>45.9</td>
</tr>
<tr>
<td>Sn3Ag0.5Cu</td>
<td>51</td>
<td>0.4</td>
<td>18.2</td>
</tr>
</tbody>
</table>

2.2.4 Microstructural Analysis

Some of the DCB specimens were prepared for microstructural analysis by cross-sectioning across the specimen width and wet grinding using 400, 800 and 1200 grit papers. The specimens were then polished with 6 µm diamond paste followed by 2 min of polishing with 0.05 µm colloidal silica. The prepared samples were subsequently analyzed with optical and scanning electron microscopy (SEM). The results of this analysis are presented in the following section.

2.3 Results and Discussion

2.3.1 Microstructural Characterization

Figure 2.6 shows the optical micrographs of the solder in specimens prepared with the TAL60 and the TAL120 profiles. The light regions are the primary tin dendrites and the dark regions are the eutectic phase containing a mixture of Sn, Cu_6Sn_3 and Ag_3Sn particles. The size of the primary tin dendrites in both specimens was similar because the bulk solder microstructure was mainly governed by the solidification rate which was constant in these specimens. The
visible primary Cu₆Sn₅ intermetallics in the solder matrix are characteristic of solder joints prepared with copper substrates. Figure 2.7 is a lower magnification optical micrograph of a TAL60 specimen showing that the Sn dendrites were smaller near the copper interfaces, becoming larger towards the center where the solder solidifies last. The orientation of the dendrites changes across grain boundaries as seen in Fig. 2.7 where the large grain has mostly vertically-oriented dendrites and the smaller grain below it has dendrites oriented in the horizontal direction. On average, approximately one or two grains were observed across the solder layer thickness regardless of the time above liquidus (TAL). This is consistent with the number of grains observed in the microstructure of some BGAs [24, 25].

Figure 2.8 shows a high magnification SEM image of the copper-solder interface. Only Cu₆Sn₅ IMCs were observed in the interfaces in all the specimens. It can be seen from Fig. 2.9 that the Cu₆Sn₅ intermetallic layer became thicker and more continuous as the TAL increased. The pits near the solder and IMC interface in Fig. 2.9(a) and (b) are an artifact of the polishing process. The average thickness of these IMC layers was estimated using digital image analysis (ImageJ [26]), dividing the area of the layer by its length. This gave IMC layer thicknesses (average±standard deviation) of 3±0.6 (n=4), 4±0.8 (n=3) and 5±0.2 (n=5) μm for TAL60, TAL120 and TAL240, respectively. The quantities in the brackets represent number of measurements. These thickness values are similar to what has been observed in manufactured solder joints [27].
Fig. 2.6 Optical micrographs showing the microstructure of solder layer in (a) TAL60 specimen and (b) TAL120 specimen. 40 µm scale bars.

Fig. 2.7 Optical micrograph showing solder microstructure in a TAL60 specimen. Two grains can be seen in the picture (part of the grain boundary is marked by the white dashed line). 200 µm scale bar.
Fig. 2.8 Interface IMC microstructural details of TAL240 specimen. The IMC layer thickness is shown at different points. Average thickness was 5 µm.

Fig. 2.9 Intermetallic compound microstructure of a) TAL60 b) TAL120 and c) TAL240 specimens.
2.3.2  R-curve and Fracture Behavior

Figure 2.10 (a) shows a typical R-curve from a single mode I fracture experiment with the SAC305 solder/Cu joint system processed at TAL120. After the initiation of the crack at approximately $G_{ci} = 480 \, \text{J/m}^2$, the fracture toughness increased with crack length for about 20 mm of crack growth before reaching a steady-state with an average $G_{cs} = 1,800 \, \text{J/m}^2$. The rising part and the steady-state region of the R-curves were modeled as straight lines using the following procedure. The first 5 points beginning with $G_{ci}$ were fitted with a straight line (points $i=1-5$). This was then repeated for subsequent sets of 5 points ($i=2-6, 3-7$ etc.) until both the slope of the linear fit and the $R^2$ value decreased for 3 consecutive sets of 5 points indicating that the rising part was merging with the steady-state plateau and that the linear fit was becoming less accurate. The third consecutive point defining the onset of this departure from linearity (around 20 mm in Fig. 2.10 (a)) was then defined as the end of the rising part, and rest of the points were defined as the steady state region of the R-curve.

This procedure did not work in a small fraction of the experiments (4 out of 20) as illustrated in Fig. 2.10 (b), either due to excess scatter in the data points of the transition region at the end of the rising part, or because there were insufficient measurement points on the rising part. The fracture surfaces of these specimens did not appear to be distinctive in any way. In such cases, straight lines were fitted to the initial section of the rising part of the R-curve and to the end of the steady-state region. The intersection of these lines was defined as the end of the rising part. For example, the end of the rising part occurs at around 7 mm in Fig. 2.10 (b), and the steady-state region starts at approximately at 18 mm, yielding $G_{ci}$ and $G_{cs}$ values of $380 \, \text{J/m}^2$ and $1,500 \, \text{J/m}^2$ in this case.

The increase in $G_c$ with crack length was due to the development of a damage zone ahead of the macroscopic crack tip consisting of micro-cracked and plastically deformed solder, as well as the development of crack bridging behind the crack tip. The size of this damage zone increased as the crack propagated, causing an increase in the toughness due to the dissipation of additional energy. As will be discussed below, the amount of crack growth before the damage zone reached a steady-state size and $G_c = G_{cs}$ depended on the mode ratio of the loading, and the solder microstructure. The effect of specimen stiffness on the R-curve was not studied in the present experiments. Hutchinson and Suo [28] reasoned that the length of the rising part of the
Fig. 2.10 R-curves of SAC305/Cu joint specimens tested under mode I loading and processed under (a) TAL120 and (b) TAL240 respectively.
R-curve of a delaminating composite DCB should decrease as the specimen becomes more compliant. Nevertheless, a linear elastic finite element analysis of the present Cu/SAC305 DCBs showed that the compliance changed negligibly with increasing crack length, so this effect was probably insignificant.

Figures 2.11 and 2.12 show typical fracture surfaces, illustrating the three-dimensional nature of the cracking and the extensive bridging between the copper bars by solder ligaments that resulted from crack propagation along alternating interfaces. As will be discussed below, the crack path was always through the solder, although very close to the interfaces. These mechanisms are analogous to ones seen in the fracture of some metal-ceramic joints and adhesive joints. For example, Evans et al. [29, 30] observed alternate debonding and bridging in Al₂O₃-Au joints, and showed experimentally that the R-curve behavior was due to the bulk plastic deformation of the Au bond layer at the bridging locations. Toughened epoxy adhesives bonding aluminum substrates exhibit similar bridging by ligaments of epoxy behind the macro-crack and micro-cracking ahead of the macro-crack [31].

Fig. 2.11 Fracture surfaces of specimens tested at $\psi = 25^\circ$, showing bridging ligaments and alternating crack path.
Fig. 2.12  Edge view of fracture specimen showing bridging zone approximately 4 mm behind the macro-crack tip due to crack jumping from lower interface to upper interface. Solder layer was 400 µm thick. Specimen tested at $\psi = 0^\circ$. 
2.3.3 Effect of Initial Conditions on Initiation $G_c$

The value of $G_{ci}$ largely governs the strength of small joints such as BGAs where only a small amount of crack growth and damage zone toughening can occur. In order to understand the significance of the starting conditions on $G_{ci}$, mode I fracture experiments were carried out on TAL240 specimens prepared with different starting geometries at the end of the solder layer as shown in Fig. 2.4. Figure 2.13 shows that the crack initiation was surprisingly independent of the pre-crack geometry; although there was slightly more scatter as the starting geometry became smoother (i.e. from “tape” to “wire”). This lack of dependence of $G_{ci}$ on the starting geometry is an indication that the differences in the stress concentration created by such macroscopic geometric changes are insignificant from a practical perspective; i.e. any differences are masked by the inherent scatter in the $G_{ci}$ measurements.

![Bar Chart](image)

**Fig. 2.13** The mean initiation toughness data $G_{ci}$ of mode I fracture as a function of different pre-crack geometries (TAL240 specimens). The data were obtained from 5 specimens in each case. The error bars represent ±95% confidence intervals (t-test).
2.3.4 Effect of Mode Ratio on R-curve Behavior

Figure 2.14 shows that the initiation toughness, $G_{ci}$ of TAL120 specimens was largely independent of phase angle from mode I to $\psi = 25^\circ$, which is in accordance with the observations of Choi et al. [10] at phase angles from $-25^\circ$ to $5^\circ$. It increased approximately 35% from 520 J/m$^2$ at mode I to 700 J/m$^2$ at $\psi = 45^\circ$. A comparison of the R-curves of these same specimens in Fig. 2.15 and Table 2.2, reveals that a similar trend existed for $G_{cs}$ as well; i.e. no difference between the two smallest phase angles, and an increase at $\psi = 45^\circ$. Further, it can be noted from Table 2.2 and Fig. 2.15 (d) that the rising slope and the rising length, $a_{Ls}$, was independent of phase angle over the range $\psi = 0^\circ$ to $45^\circ$. These trends in $G_{ci}$, $G_{cs}$ and the rising part of the R-curves were significant at the 95% confidence level using the t-test.

These observations are consistent with the fracture behavior of joints and laminates, where $G_c$ typically increases with the phase angle due to the increase in the damage zone size with increasing amounts of mode II (e.g. [28, 31]). This should not be confused with the increase in the damage zone size due to crack growth in the rising part of R-curve.

Fig. 2.14 $G_{ci}$ of TAL120 specimens as a function of phase angle of loading. At least four repeat experiments shown per phase angle. The line passes through the mean values.
The differences between the R-curves in Fig. 2.15 were also evident in the fracture surfaces (Fig. 2.16), with higher $G_c$ values corresponding to more three-dimensional, rougher crack paths. For example, the uppermost mode I curve in Fig. 2.15 (a) (+ symbol) lay above the others and corresponded to a specimen that had a fracture surface lying mostly at the midplane of the solder layer (Fig. 2.16 (a)). In contrast, the other mode I specimens tended to have crack paths similar to that shown in Fig. 2.11; smooth fracture surfaces composed of regions close to one or other of the interfaces. It was typical that more planar crack propagation in the middle of the solder layer was rougher and caused a greater $G_c$ than did propagation near the interfaces (Fig. 2.16). Similarly, the uppermost curve in Fig. 2.15(b) (○ symbol) corresponded to a fracture surface having more frequent crack jumps between interfaces over the initial 15 mm compared to the other specimens, which resulted in a higher $G_c$ (Fig. 2.16(b)). However, with further crack propagation the curve starts merging with the others (beyond 30 mm), and the fracture surface became similar to the others.

At higher phase angles (i.e., 25° and 45°), crack initiation and a few millimeters of stable growth always occurred close to the upper interface, before the crack path began to shift erratically between the two interfaces. Fracture surfaces corresponding to the 45° phase angle were more planar than those at lower phase angles, probably because of the tendency of constrained cracks to move closer to the more highly strained adherend (i.e., upper adherend of DCBs in the present tests), thereby maximizing tensile stress normal to the crack plane.
Fig. 2.15 R-curves of TAL120 specimens tested under (a) mode I, (b) $\psi = 25^\circ$, and (c) $\psi = 45^\circ$. 4 specimens tested in each case. (d) Comparison of R curves obtained by averaging the data in (a), (b) and (c).
Fig. 2.16 Low magnification images of (a) smooth fracture surface formed due to crack growth near the interfaces and (b) rough fracture surface due to crack growth in the middle of solder layer. Specimens were tested in mode I loading.

Table 2.2 Summary of mode I and mixed-mode R-curve parameters for TAL120 specimens. Four specimens tested in each case; ± one standard deviation.

<table>
<thead>
<tr>
<th>$\psi$</th>
<th>$G_{ci}$ (J/m$^2$)</th>
<th>Rising slope</th>
<th>Rising length $a_{Ld}$ (mm)</th>
<th>$G_{cs}$ (J/m$^2$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$0^\circ$</td>
<td>523 ± 70</td>
<td>90 ± 38</td>
<td>16 ± 6</td>
<td>1801 ± 102</td>
</tr>
<tr>
<td>$25^\circ$</td>
<td>503 ± 50</td>
<td>75 ± 16</td>
<td>18 ± 4</td>
<td>1730 ± 66</td>
</tr>
<tr>
<td>$45^\circ$</td>
<td>700 ± 149</td>
<td>76 ± 16</td>
<td>18 ± 4</td>
<td>2293 ± 128</td>
</tr>
</tbody>
</table>
2.3.5 Effect of Time-Temperature Profile on Mixed-Mode Fracture Behavior

Figures 2.15(b), 2.17 and 2.18 show the R-curves for a phase angle of 25° obtained from specimens prepared with TAL120, TAL60 and TAL240, respectively. It is evident that the R-curves for TAL60 specimens (Fig. 2.17) had a clearer transition between the rising portion and the steady-state region. This is attributed to the absence of bridging regions caused by alternate interface debonding (Fig. 2.11) in TAL60 specimens. The presence of such crack face bridging behind the macro-crack tip effectively lengthens the damage zone that extends ahead of the macro-crack, because the solder ligaments continue to bear load behind the crack and hence contribute to a rising $G_c$. In few cases $G_c$ decreased with crack length after reaching the plateau; e.g. the ◊ curve in Fig. 2.17. This decrease was consistent with a corresponding change in the fracture surface as shown in Fig. 2.19. The highly three-dimensional crack front in region ‘A’ changed to a more planar crack front in region ‘B’ after approximately 30 mm of crack growth, resulting in less energy dissipation and a decreased $G_c$. The R-curves for TAL240 (Fig. 2.18) show that the rising part lengths were shorter, ending at crack lengths of about 8 mm compared to 17 mm for TAL60 and TAL120. As will be explained below, this can be attributed to the intermetallic microstructure at the copper-solder interface which favored planar crack as opposed to a rougher, three-dimensional fracture surface.

Table 2.3 and Fig. 2.20 show that $G_{ci}$ decreased from TAL60 to TAL120, but did not change any further (95% confidence). However, the $G_{cs}$ decreased continuously as TAL increased. The slight increase in $G_{ci}$ from TAL120 to TAL240 was statistically insignificant (95% confidence). This trend in the fracture behavior with TAL can be attributed to the differences in the thickness of the intermetallic layers (Fig. 2.9). TAL240 specimens had a thicker IMC layer than did the TAL60 and TAL120 specimens, and were weaker as a result. This is in accordance with the observations made in [19, 21] on the effect of IMC layer thickness on the mode I fracture initiation. The rising length and slope were the same at TAL60 and TAL120, but were smaller at TAL240, simply because $G_{cs}$ was smaller while $G_{ci}$ did not change appreciably (95% confidence level).
Fig. 2.17 Critical strain energy release rate as a function of crack length for 4 TAL60 specimens tested at $\psi = 25^\circ$.

Fig. 2.18 Critical strain energy release rate as a function of crack length data obtained from 4 TAL240 specimens tested at $\psi = 25^\circ$. 
Fig. 2.19 Comparison of fracture surface and corresponding critical strain energy release rate for the TAL60 specimen corresponding to the lowest curve of Fig. 2.17.
Fig. 2.20 (a) Initiation energy $G_{ci}$ and (b) steady-state energy $G_{cs}$ as a function of time above liquidus (TAL). Four repetitions at each TAL, specimens tested at $\psi = 25^\circ$. 
Table 2.3  Summary of mixed-mode ($\psi = 25^\circ$) fracture data (of Fig. 2.15 (b), 2.17 and 2.18) showing the effect of TAL on the R-curve. Four specimens tested at each TAL; ± one standard deviation.

<table>
<thead>
<tr>
<th>TAL (s)</th>
<th>$G_{ci}$ (J/m$^2$)</th>
<th>Rising part slope</th>
<th>Rising length $a_{Ls}$(mm)</th>
<th>$G_{cs}$ (J/m$^2$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>60</td>
<td>677 ± 78</td>
<td>77 ± 11</td>
<td>17 ± 4</td>
<td>2040 ± 124</td>
</tr>
<tr>
<td>120</td>
<td>503 ± 50</td>
<td>75 ± 16</td>
<td>18 ± 4</td>
<td>1730 ± 66</td>
</tr>
<tr>
<td>240</td>
<td>581 ± 56</td>
<td>56 ± 16</td>
<td>8 ± 2</td>
<td>1240 ± 134</td>
</tr>
</tbody>
</table>

2.3.6  Fracture Surfaces and Failure Mechanisms as a Function of TAL

Crack growth is affected by the mechanics of the loading and the properties of the solder joint. The crack will tend to a path that is normal to the first principal stress, maximizing the mode I component of the strain energy release rate since the solder is weakest in mode I (over a small range of low phase angles, Table 2.2). The location of this path is related to the mode ratio of the loading, being in the midplane of the solder layer in mode I and along the more highly-strained copper bar at higher mode ratios [28]. The crack path will also be governed by microstructural variations in the solder and interface properties that create regions of weakness. At lower phase angles (i.e. $\psi$ value less than about 25$^\circ$; Table 2.3), the crack path is only weakly affected by the presence of mode II and the cracks may jump from one interface to the other depending on local intermetallic layer variations. However, at higher phase angles, as explained above, the crack tended to extend along a path closer to one interface which resulted in a more planar crack growth (section 2.3.4).

Figure 2.21 shows the typical fracture surfaces of specimens tested at $\psi = 25^\circ$ and prepared with the three TAL profiles, each creating a different intermetallic profile at the
interface. It is noted that TAL60 specimen fracture surfaces had more pits and sharp changes in elevation resulting from crack jumps from one interface to the other. The TAL120 case displayed crack bridging that resulted from fracture occurring along both interfaces leaving a ligament between the two copper bars. As discussed with Fig. 2.15 (b), this appears to be the reason for the longer rising part of the R-curves in this case. Although similar ligaments were visible in TAL240, they were very few and the fracture surface was relatively smooth with negligible crack jumps. As discussed below, this behavior of ligament formation can be attributed to the greater strength of the interfacial intermetallics in TAL60 and TAL120 specimens compared with the TAL240 specimens. The increasing roughness of the fracture surfaces corresponded to the increasing $G_c$ as TAL decreased (Table 2.3).

Fig. 2.21 Typical fracture surfaces of specimens prepared with TAL240, TAL120 and TAL60, tested at $\psi = 25^\circ$. Scale shown is in mm.
Figure 2.22 shows scanning electron micrographs of the TAL60 and TAL240 fracture surfaces. It is seen that the TAL240 specimens failed in a completely brittle manner through the intermetallic layer which was approximately 5 μm thick (Fig. 2.9). The crack growth was perpendicular to the columnar IMCs creating a regular hexagonal pattern with smooth facets that can be observed in the central part of the TAL240 image (Fig. 2.22 (b)). These hexagonal features are typical of a Cu₆Sn₅ failure, which is quite brittle [32, 33] and grows as individual hexagonal rods as seen in Fig. 2.23. The EDX analysis confirmed that these hexagonal features were indeed Cu₆Sn₅ particles (note the Cu peak in Fig. 2.22). In contrast, the TAL60 specimens failed in a ductile manner through the solder layer but very close to IMC layer. The fracture surface had rounder dimples which are typical of ductile failure consisting of void formation and coalescence [17, 19, 21]. The walls of these round pits were formed by solder, which is again confirmed with an EDX analysis. The intermetallic layer, which had a thickness of around 3 μm for this case (Fig. 2.9), did not play a significant part in the failure. As a result the toughness of the joint was very high (Table 2.3). The TAL120 fracture surfaces displayed a mixture of both ductile and brittle features.

These results are consistent with similar observations in [17, 19, 21], and show the significance of interfacial IMC microstructure on the fracture behavior of solder joints as a function of TAL (i.e. correlation of microstructure in Figs. 2.9 and 2.22 with toughness values in Table 2.3 and Fig. 2.20). The TAL60 specimen had the greatest $G_{cs}$ because it failed predominantly in a ductile mode (Fig. 2.22 (a)). Similarly, the TAL120 specimens, which failed with a mixture of ductile and brittle mechanisms, had higher toughness, $G_{cs}$, compared to the TAL240 specimens that fractured mainly in a brittle manner through the Cu₆Sn₅.
Fig. 2.22  SEM and EDX analysis of fracture surfaces of (a) TAL60, ductile failure through SAC solder, and (b) TAL240, brittle Cu₆Sn₅ cleavage. Scale bar 15 μm.
2.4 Conclusions

The fracture behavior of a Cu/Sn3Ag0.5Cu solder joint system was studied under mode I and mixed-mode conditions using DCB specimens manufactured using standard industrial processing conditions and three time-above-liquidus (TAL) values. The microstructures of the specimens were found to be similar to those seen in commercial solder joints, and all joints exhibited a qualitatively similar R-curve behavior, but with different values of toughness.

The phase angle had little effect on the critical strain energy release rate, $G_c$, between $\psi=0^\circ$ (mode I) and $25^\circ$, but caused a 35% increase at $\psi=45^\circ$ in both the initiation value, $G_{ci}$, and the steady-state value $G_{cs}$. This behavior was analogous to that seen in adhesive joints.

The solder joint toughness decreased as the time-above-liquidus increased. This was attributed to the intermetallic compound (IMC) layer thickness at the joint interface, which increased from an average value of 3 μm at TAL 60 s to 5 μm at TAL 240 s. The thinner IMC layer produced a tougher, more ductile fracture, while the thicker IMC layer caused brittle
Cu₆Sn₅ cleavage at a lower critical strain energy release rate. These differences were reflected in the morphology of the fracture surfaces and in the degree of crack bridging behind the macro-crack tip. The other R-curve parameters, such as rising length and slope were not affected significantly by the phase angle, but were affected by TAL.

The initiation strain energy release rate, which governs the failure of small joints such as BGAs, was largely independent of the geometry of the solder fillet at the free end of the solder layer.

2.5 References

3 Effect of Geometry on the Fracture Behavior of Lead-free Solder Joints

3.1 Introduction

Microelectronic package failures are often due to cracking in the solder or at the solder/pad interface under various thermal and mechanical loading conditions during assembly, testing, or service. The majority of the research in this area has dealt with thermal fatigue. Solder joint failure under drop impact and vibration loading has also been of great interest. But relatively little attention has been paid to the development of methods of predicting the fracture load of solder joints under quasi-static mechanical loads, applied either directly on components or induced by the bending or twisting of printed circuit boards (PCBs). Most existing experimental methods to evaluate the strength of solder joints under mechanical loads are primarily qualitative and do not provide fundamental mechanical properties that can be used to predict the strength of joints in other configurations or loads.

In general, solder joints fracture under mixed-mode loading such that the loads create a combination of a tensile, opening-mode strain energy release rate, $G_I$, as well as a shear mode value, $G_{II}$. The mode ratio of loading is often defined by the phase angle, $\psi = \arctan\left(\sqrt{\frac{G_{II}}{G_I}}\right)$; i.e. $\psi = 0^\circ$ is pure mode-I while $\psi = 90^\circ$ is pure mode-II. Early work on solder joint fracture [1-4] focused mainly on understanding the mode I fracture behavior of SnPb solder joints. Subsequent studies [5 - 10] investigated the solder fracture under mixed-mode conditions. For example, Choi and co-workers [7, 8] carried out fracture studies on brass/solder/brass sandwich specimens with 0.5 and 2 mm solder layer thicknesses. They observed that fracture toughness increased with solder layer thickness and that the fracture toughness of the 0.5 mm layer did not change significantly for the range of mode ratios considered (-25° to 5°). Siow and Manoharan [5, 10] measured mode I-III fracture energies of SnPb and SnAg solder. They found that the SnPb joints
were tougher than the SnAg joints in mode I, and mixed-mode I-III loading decreased the fracture toughness compared with the mode I values.

Most of the above studies are focused on lead-based solder alloys. Also, in all these studies [2-10], the authors used fatigue pre-cracking to create an initial sharp crack tip for subsequent fracture toughness testing, and with the exception of [2], made a measurement at only one crack length, omitting observations of possible R-curve toughening. Mixed-mode fracture experiments on lead-free Sn3Ag0.5Cu (SAC305) solder have been done with double cantilever beam specimens [11], where it was observed that cracks initiated at a relatively low critical energy release rate, \( G_{ci} \), and the joints toughened with crack growth until a steady-state value \( (G_{cs}) \) was reached. This implies that relatively large solder joints, such as those in heat sink attachments or power electronic modules, will have an ultimate strength that is greater than that indicated by crack initiation. In contrast, the fracture of a smaller joint such as a ball-grid array (BGA) is governed mainly by the critical energy release rate at initiation, \( G_{ci} \), since subcritical crack growth is small in this case. It was noted that \( G_{ci} \) and \( G_{cs} \) increased with the size of the mode II component [11]. It was also observed that the local geometry of the end of the solder joint had only a relatively small effect on \( G_{ci} \). In subsequent studies [12, 13], the \( G_{ci} \) obtained from a continuous joint DCB specimen was used to predict the mode-I and mixed-mode fracture loads of discrete 2 mm and 5 mm long SAC305 lead-free solder joints arranged in a linear array between two copper bars. These two studies showed that fundamental solder joint fracture properties can be used to predict the strength of solder joints of various shapes and sizes.

Solder joints in electronic devices range in thickness from approximately 500 μm to less than 100 μm, and are used to join various substrates having different bending stiffnesses. Experience with epoxy adhesive joints has shown that these factors can, in principle, affect the stress state at a crack tip and alter the effective fracture toughness of the joint [14]. The main objective of the present work was to examine the effect of solder joint thickness and substrate stiffness on the R-curve behavior of SAC305 lead-free solder joints under different mixed-mode loading conditions. The crack paths and fracture surfaces were explained using an elastic-plastic finite element model. Further, some preliminary tests were performed to get a qualitative understanding of the effect of loading rate on fracture properties. The observations are relevant to the prediction of fracture in both small joints such as BGAs in microelectronic applications and relatively large solder joints such as those used to connect power electronics modules.
3.2 Experimental Procedures

3.2.1 Specimen preparation

Figure 3.1 depicts the double cantilever beam (DCB) specimen consisting of two copper bars (C110 alloy) joined with a thin layer of SAC 305 solder. Specimens were made with two bar thicknesses, \( h = 12.6 \text{ mm} \) and 19 mm, and solder layers of either \( t = 200 \mu\text{m} \) or 400 \( \mu\text{m} \).

The specimen preparation procedure was similar to that used in [11]. The Cu bars (C110 alloy) were cut to the required dimensions and the bonding surfaces were polished to achieve surface roughness values similar to that of an organic solderability preservative (OSP) finish on commercial printed circuit boards \( (R_a=0.95 \mu\text{m}) \). They were then rinsed thoroughly with tap water, dried with cheese cloth, and then rinsed with acetone. The surface areas where soldering was not required were masked with Kapton tape, taking care not to contaminate the cleaned surfaces to be soldered. Round steel wires maintained the required gap of 200 \( \mu\text{m} \) or 400 \( \mu\text{m} \) between the Cu bars.

Fig. 3.1 Schematic of the DCB specimen is shown along with different parameters. The width of the specimen was 12.6 mm and overall length was 160 mm.
The masked copper bars were placed on a hot plate covered with aluminum foil and maintained at 290°C with the bonding surfaces vertical (Fig. 3.2). When the temperature of the bars reached 220-225°C (measured using an embedded thermocouple in each specimen), a flux-cored SAC 305 0.75 mm solder wire (Kester Inc., USA) was touched to the prepared vertical surfaces so that they became rapidly covered with a thin layer of solder. The bars were then clamped together against the steel wires to maintain the desired solder thickness. The time above liquidus of 120 s, and a peak temperature of between 245°C and 250°C were maintained during this process. The specimens were then cooled at a rate of 1.4-1.6°C/s, which is typical of microelectronics manufacturing. The local geometry of the beginning of the solder layer was defined by the smooth, square shape of the edge of the Kapton tape (see Fig. 3.1). Earlier work showed that the local edge geometry had a relatively small effect on the quasi-static fracture load at crack initiation [11].

After cooling to room temperature, the excess solder was removed from the sides of the specimens by peripheral milling using an end mill, and the hand polishing with 400 grit SiC paper to facilitate the observation of cracks in the solder layer. Finally, the loading pin holes were drilled in the copper bars. A detailed microstructural study of similar DCB specimens in Chapter 2 confirmed that the intermetallics and bulk solder were similar to those typical of commercial SAC305 solder manufactured with OSP finish [11].

![Fig. 3.2 Schematic of specimen arrangement during soldering of a DCB and discrete joint specimens [11].](image-url)
3.2.2 Fracture Testing

The DCB specimens were tested under mode-I and mixed-mode conditions using the load jig shown in Fig. 3.3 [15]. The mode ratio of loading was selected by adjusting the location of the link pins to apply unequal forces on the upper and lower arms, which induces a combination of tensile and shear stresses on the solder layer. Thus a single DCB geometry could be used to obtain the three mode ratios that were used, corresponding to phase angles of $0^\circ$ (mode I or pure tensile crack opening stress), $25^\circ$ and $45^\circ$.

The visible edge of the solder layer was painted with a thin layer of diluted paper correction fluid to facilitate the identification of the crack tip. A microscope on a micrometer stage with a field of view of 1.9 mm was used to observe crack initiation from the start of the solder layer defined by the Kapton tape and propagation from subsequent cracks, as well as to measure the distance from the crack tip to the loading pins ($a$ in Fig. 3.1). The load on the specimen was increased with a constant cross-head speed of 0.1 mm/min until a critical load was reached when crack extension occurred. Crack initiation from the end of the solder layer in an uncracked specimen was defined as occurring at the load causing crack extension of approximately 100 μm. This was also the resolution used to define all the subsequent crack extensions; i.e. the critical fracture load from an existing crack was recorded after approximately 100 μm of growth. After each crack extension the crosshead was reversed to decrease the load by about 50%. Since the fracture was stable, a single specimen could be used to measure 30 or more crack growth sequences, each providing a measure of $G_c$ at progressively longer crack lengths. A total of 33 DCB specimens were used to study the various combinations of solder layer thickness, Cu bar thickness, and mode ratio.

To develop a preliminary understanding of the effect of loading rate on the solder fracture, three DCB specimens with $t=200$ μm were loaded in mode-I at a rate of 0.1 mm/min until a crack was initiated. The same specimens were subsequently loaded at a rate of 5 mm/min until the previously initiated crack started propagating. Since the crosshead was stopped immediately after crack initiation in the initial loading, the damage zone development and R-curve toughening was assumed to have an insignificant effect on the measured $G_c$ during the
second loading.

![Fig. 3.3 Schematic of the DCB specimen mounted in the mixed-mode load jig [14].](image)

### 3.2.3 \( G_c \) Calculations

The fracture test of a single DCB specimen provided one critical load measurement corresponding to crack initiation from the start of the solder layer and many subsequent crack extension events (~30) corresponding to the resumption of crack growth at increasing crack lengths. These critical load and their corresponding crack lengths (\(a\) in Fig. 3.1) were then used to calculate the critical strain energy release rate of the solder joint system using the beam-on-elastic-foundation model of [16]. These equations were derived under the assumption that the Cu bars deform elastically. The expression for the energy release rate, \( G_c \), is given by,
where $F_1$ and $F_2$ are the mode I and mode II critical loads per unit width of specimen and are derived from the applied force $f_1$ and $f_2$ on the specimen as explained in [16]. The constants $\Phi_I$ and $\Phi_{II}$ depend upon the geometry and mechanical properties of the DCB specimen [16] and are provided in the Appendix. The mechanical properties of Cu (C110 alloy) and Sn3Ag0.5Cu used in these calculations are listed in Table 3.1. The applied mode ratio (phase angle) of loading, $\psi$, in each test was calculated as [16]

$$\psi = \arctan\left[\frac{\sqrt{3}}{2} \frac{F_2 \Phi_{II}}{F_1 \Phi_I}\right]$$

(2)

<table>
<thead>
<tr>
<th>Tensile modulus (GPa)</th>
<th>Poisson ratio</th>
<th>Shear modulus (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cu C110</td>
<td>124</td>
<td>0.35</td>
</tr>
<tr>
<td>Sn3Ag0.5Cu</td>
<td>51</td>
<td>0.4</td>
</tr>
</tbody>
</table>

3.3 finite element model

The deformation of the Cu-SAC305-Cu DCB specimen was simulated using ANSYS 12® finite element software, and the strain distributions were used to explain the crack paths observed in the DCB specimens tested under mode-I and mixed-mode loading. Figure 3.4 shows the finite element mesh of the DCB specimen with $h=12.6$ mm and $t=400$ μm, along with the boundary conditions. The Cu bars and the solder layer were meshed with PLANE 183 2-D 8-
node structural elements ranging in size from 0.05 mm in the solder layer to 1.5 mm in the Cu bar far from the start of the solder layer. The elements in the Cu bars were modeled as plane stress while the solder elements were modeled as plane strain. The constitutive behavior of the Cu was defined as isotropic linear elastic and the solder as isotropic elastic-perfectly plastic with the yield stress of 40 MPa [17].

Fig. 3.4 Finite element mesh of the DCB specimen along with boundary conditions. The magnified portion shows the mesh details near the beginning of the solder layer where the crack initiated.
3.4 Results and Discussions

3.4.1 R-curve Behavior of SAC305 Solder

Figure 3.5 shows a typical R-curve of a Cu-SAC305 solder joint with a 200 μm solder layer thickness tested at ψ=25°. Two distinct regions were apparent: a rising part and a steady-state region, both of which were modeled as straight lines using the procedure given in [11]. In this case, the initiation critical strain energy release rate, $G_{ci}$, was 517 J/m$^2$ while the steady-state value, $G_{cs}$, was 1,660 J/m$^2$.

![R-curve graph](image)

Fig. 3.5 R-curve of Cu-SAC305-Cu joint system with $t=200$ μm, tested at ψ=25°.

As explained in [11], the increase in $G_c$ with crack length (rising part) was due to the development of a damage zone ahead of the crack tip consisting of micro-cracked and plastically deformed solder, as well as the development of crack bridging behind the crack tip. The size of this damage zone increased as the crack propagated, causing an increase in the toughness due to the dissipation of additional energy, until it reached a steady-state size leading to crack
propagation at constant $G_{cs}$. As will be discussed below, the amount of crack growth before the damage zone reached a steady-state size and $G_c = G_{cs}$ depended on the mode ratio of the loading and the solder layer thickness. In addition, this damage zone evolution also depended on the solder microstructure [11]. Figure 3.6 shows the fracture surface of the specimen of Fig. 3.5, illustrating the three-dimensional nature of the crack propagation along alternating interfaces which caused solder bridging between the copper bars (i.e. solder ligaments continued to join the bars in the wake of the crack).

![Fracture surface of SAC305 solder specimen of Fig. 3.5. $t=200 \, \mu m$, tested at $\psi=25^\circ$.](image)

Fig. 3.6 Fracture surface of SAC305 solder specimen of Fig. 3.5. $t=200 \, \mu m$, tested at $\psi=25^\circ$.

### 3.4.2 Effect of Substrate Stiffness on Crack Initiation

Figure 3.7 shows that the mean $G_{ci}$ of the solder joint ($t=400 \, \mu m$ loaded at $\psi=45^\circ$) tended to increase with the increasing thickness of the Cu bars; i.e. from a mean $G_{ci} = 702 \, J/m^2$ for 12.6 mm bars to $G_{ci} = 842 \, J/m^2$ for 19 mm bars. However, this change in $G_{ci}$ was statistically insignificant (t-test, 95% confidence level) given the relatively large scatter in the crack initiation measurements.
In principle, an increase in $G_{cl}$ with beam thickness could be attributed to the stress distribution in the initiation region. Thicker beams would tend to distribute stresses over longer distances in the solder layer, thereby increasing the size of the damage zone causing slightly more energy dissipation than would thinner beams. A similar effect of increasing fracture energy with increasing beam stiffness was observed by Mangalgiri et al. [18] in the mode-I fracture toughness of a structural adhesive, although in that case crack propagation occurred from existing cracks rather than from an uncracked condition as in the current $G_{cl}$ measurements.

Further evidence of the relative insensitivity of $G_{cl}$ to changes in the solder layer stress distribution was found in a previous study [13], where the fracture loads of discrete 2 mm and 5 mm long SAC305 solder joints were predicted using the $G_{cl}$ measured from continuous DCB specimens. In that case, $G_{cl}$ was largely independent of the differences in solder layer constraint due to the changing solder layer length.

Fig. 3.7 Effect of beam thickness, $h$, on the $G_{cl}$ of Cu-SAC305-Cu joints with 400 μm solder layer thickness tested at $\psi=45^\circ$. The number of specimens tested in each case is indicated above each column, and the error bars represent the ± 95% confidence intervals (t-distribution).
3.4.3 Effect of Solder Thickness on the R-curve at Different Mode Ratios

The variation of the initiation strain energy release rate, $G_{ci}$, with solder thickness was statistically insignificant over the range $t=200 - 400 \, \mu m$ for the three phase angles, as seen in Fig. 3.8. In contrast, the steady-state critical strain energy release rate, $G_{cs}$, did increase significantly with the solder thickness (Fig. 3.9; t-test, 95% confidence level).

This observation that solder thickness does not influence $G_{ci}$ but does affect $G_{cs}$, can be attributed to the differences in constraint levels imposed by the Cu bars on the damage zone development in the solder layer. Crack initiation at the start of the solder layer occurs close to a free surface and is thus close to a state of plane stress. Moreover, the damage zone at initiation will be relatively small, and is unlikely to span the entire solder layer thickness; therefore, it will not be constrained by the adjacent Cu bars. However, as loading increases after initiation, the damage zone evolves and grows, leading to toughening as seen in Fig. 3.5. Eventually the damage zone extends across the solder thickness and its further expansion becomes constrained by the bounding Cu bars. This can also be seen from Table 3.2, which shows the average slope and length of the rising part of the R-curves measured for both solder thicknesses at the three phase angles. The rate at which damage zone develops (i.e. slope) was similar for both layer thicknesses (t-test, 95% confidence level), but the rising part of the R-curve was significantly longer for the thicker solder layer since the damage zone was larger and could grow further than in the thinner solder layer.

These observations imply that the effect of solder layer thickness can be neglected when predicting the strength of relatively short solder joints using $G_{ci}$. However, longer joints may support sufficient subcritical crack growth to realize appreciable toughening which will increase with the solder layer thickness.

The effect of phase angle on $G_{ci}$ and $G_{cs}$ was similar for both solder layer thicknesses; i.e. both parameters increased significantly with phase angle from 25° to 45° (Figs. 3.8 and 3.9; t-test, 95% confidence level).
Fig. 3.8 Initiation strain energy release rate, $G_{ci}$, as a function of mode ratio, $\psi$, obtained from $t=200$ $\mu$m and $t=400$ $\mu$m Cu-SAC305-Cu joints. The number of specimens tested in each case is indicated above each column, and the error bars represent the ± 95% confidence intervals (t-distribution).

Fig. 3.9 Steady-state strain energy release rate, $G_{cs}$, as a function of phase angle, $\psi$, obtained from $t=200$ $\mu$m and $t=400$ $\mu$m Cu-SAC305-Cu joints. The number of specimens tested in each
case is indicated above each column, and the error bars represent the ± 95% confidence intervals (t-distribution).

Table 3.2  R-curve parameters (defined in Fig. 3.5) for joints with solder thickness $t=200 \, \mu m$ and $t=400 \, \mu m$ at different phase angles (mean±standard deviation, $N$ is given in Fig. 3.9).

<table>
<thead>
<tr>
<th>$t$ (μm)</th>
<th>$\psi=0^\circ$</th>
<th>$\psi=25^\circ$</th>
<th>$\psi=45^\circ$</th>
<th>$\psi=0^\circ$</th>
<th>$\psi=25^\circ$</th>
<th>$\psi=45^\circ$</th>
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<td>200</td>
<td>71</td>
<td>70±18</td>
<td>110±50</td>
<td>12</td>
<td>11±0.4</td>
<td>10±1.8</td>
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<tr>
<td>400</td>
<td>90±38</td>
<td>75±16</td>
<td>76±16</td>
<td>16±6</td>
<td>18±4</td>
<td>18±4</td>
</tr>
</tbody>
</table>

3.4.4 Effect of Phase Angle on the Crack Path

Figure 3.10 illustrates the mechanism of crack propagation observed in the mode-I tests. In this example, the crack initially propagated along a path within the solder but very near the lower interface, and then jumped toward the upper interface, possibly because of random changes in the local intermetallic microstructure. Consequently, the fracture surface was highly three-dimensional, with jumps occurring in both the dominant growth direction along the specimen and across its width. In some regions, this caused the solder layer to form a ligament bridging both interfaces in the crack wake (Figs. 3.10 and 3.11), leading to another R-curve toughening mechanism. Similar behavior was observed in the mixed-mode case at $\psi=25^\circ$ for both solder thickness values. These mechanisms are analogous to ones seen in the fracture of some metal-ceramic joints and toughened adhesive joints. For example, Evans et al. [19, 20] observed alternate debonding and bridging in Al$_2$O$_3$-Au joints, and Papini et al. [21] observed bridging by ligaments of epoxy behind the macro-crack and micro-cracking ahead of the macro-crack.
Fig. 3.10 Mode-I fracture of DCB specimen showing the details of crack growth and deformation of the 400 μm solder layer.

The crack path and fracture surfaces were slightly different for specimens tested at \( \psi=45^\circ \). Figure 3.12 shows that a typical crack initiated and grew for approximately 2 mm near the upper interface and then shifted toward the lower interface where it remained until the specimen broke completely. This produced a predominantly planar fracture surface in contrast to the highly three-dimensional crack path observed at phase angles below 25° (Fig. 3.11). Similar crack growth behavior was observed in \( t=200 \) μm solder joints tested at \( \psi=45^\circ \).
To confirm that these differences in the fracture surfaces were due to the change in the mode ratio and not because of microstructural variations, some specimens ($t=400 \ \mu\text{m}$, $h=12.6 \ \text{mm}$) were initially loaded under $\psi=45^\circ$ and then subsequently fractured under mode-I loading as in Fig. 3.13. It is evident that the initial portion of the crack path resembles Fig. 3.12 at $\psi=45^\circ$, and the remaining portion of fracture surface is similar to that shown in Fig. 3.11 for a mode-I loading. Therefore, the observed differences in the fracture surfaces were attributed mainly to the influence of the mode ratio of loading and its effect on the stress state at the crack tip.

Fig. 3.11 Fracture surfaces of a mode-I DCB specimen ($t=400 \ \mu\text{m}$, $h=12.6 \ \text{mm}$).

Cracks in layered media such as adhesive and solder joints often follow a path that is normal to the maximum principal stress; therefore, the average crack path will be in the middle of the solder layer under mode-I loading. At $\psi=45^\circ$, since the upper Cu bar is more highly strained in this case ($f_1>f_2$, Fig. 3.1), the path normal to the maximum principal stress should cause crack propagation along the upper interface [22]. However, Fig. 3.12 shows that the crack actually followed a path near the lower Cu bar, and therefore was not growing in accordance with the maximum principal stress criterion. Instead, the crack followed a path defined by the
von Mises (equivalent) strain, $\varepsilon_{eq}$, as shown by the FE model of Figs. 3.14a and 3.14c which depict the von Mises strain fields at crack initiation ($G = \dot{G}_c = 700 \text{ J/m}^2$) and at a slightly higher load at a crack length of 200 $\mu$m ($G = \dot{G}_c = 770 \text{ J/m}^2$). The increase in $G_c$ with crack length is due to the R-curve toughening shown in Fig. 3.5. At crack initiation ($G = \dot{G}_c = 700 \text{ J/m}^2$, Fig. 3.14a), the maximum von Mises strain occurred at the start of the solder layer near the upper interface where crack initiation was observed ($\varepsilon_{eq} = 0.24$, averaged over four elements). Figure 3.14a also depicts a second, slightly smaller local maximum von Mises strain at the lower interface approximately 2 mm ahead of the site of crack initiation. This secondary local maximum is due to the distribution of the shear strain component induced by mixed-mode loading which displays the same pattern of two local maxima (Fig. 3.14b). With subsequent crack growth at higher loads and applied $G$ ($G_c = 770 \text{ J/m}^2$), the von Mises strain at the lower interface ($\varepsilon_{eq} = 0.50$) had increased to the point where it exceeded that at the site of crack initiation at the upper interface ($\varepsilon_{eq} = 0.40$). With further loading, the von Mises strain at the lower interface continues to grow ever larger than that at the upper interface and at some point crack initiation and growth are predicted to occur near the lower interface. This correlated very well with the observation from Fig. 3.12 that the crack initiated and propagated very near the upper interface approximately 2 to 2.5 mm before shifting to the lower interface where it continued to grow until the specimen broke completely. As a result, the fracture surfaces of specimens tested under $\psi = 45^\circ$ were predominantly planar along the lower interface. It is noted that these explanations assume that the IMC microstructure is uniform and identical along both interfaces. Local variations in the IMC layer can change the toughness and crack path as illustrated in Figs. 3.6 and 3.13.

Figure 3.14d shows the von Mises plastic strain contours in the DCB specimen subjected to mode-I loading corresponding to crack initiation. Only a portion near the start of the solder layer is shown, because the strains were negligible elsewhere. As expected by the symmetry of the loading, the von Mises strains were symmetrically distributed near both the upper and lower interfaces. This symmetric distribution of the maximum von Mises strain is consistent with the observed crack propagation near both interfaces in Figs. 3.10 and 3.11.

The observation that the solder crack path seemed to be governed by the location of the maximum von Mises strain suggests that the solder failed in a ductile manner, rather than a brittle failure which would tend to be dominated by the crack tip opening stress. This was confirmed by the scanning electron micrographs of the fracture surface near the crack initiation.
location under mode-I loading (Fig. 3.15a). The failure surface has features of classic ductile fracture, such as the equi-axed dimples formed by void nucleation, growth and subsequent coalescence [20]. The intermetallic compounds present near the interfaces (Fig. 3.15b), and impurities such as gas bubbles from the solder flux can act as nucleation sites for such voids [23].

![Fracture surfaces of \( t = 400 \mu m \) DCB specimen tested at \( \psi = 45^\circ \) loading.](image1)

Fig. 3.12 Fracture surfaces of \( t = 400 \mu m \) DCB specimen tested at \( \psi = 45^\circ \) loading.

![Fracture surfaces of the specimen (\( t = 400 \mu m \)) which was tested initially under mixed-mode loading of \( \psi = 45^\circ \) and subsequently fractured under mode-I.](image2)

Fig. 3.13 Fracture surfaces of the specimen (\( t = 400 \mu m \)) which was tested initially under mixed-mode loading of \( \psi = 45^\circ \) and subsequently fractured under mode-I.
\( G_c = 700 \, \text{J/m}^2 \)
(a) von Mises strain contours

\( G_c = 700 \, \text{J/m}^2 \)
(b) shear strain contours

\( G_c = 770 \, \text{J/m}^2 \)
Fig. 3.14 Von Mises strain contours of SAC305 solder ($\psi=45^\circ$, $t=400$ µm, $h=12.6$ mm, $a=40$ mm as in Fig. 3.1) corresponding to (a) the crack initiation load ($G_{ci}=700$ J/m$^2$), (b) shear strain contours for the same loading at crack initiation, and (c) von Mises strain contours for subsequent crack propagation load ($G_c=770$ J/m$^2$, 200 µm crack) at. The maximum von Mises strain values indicated by arrows are the values averaged over four elements. (d) Von Mises strain contours corresponding to mode-I crack initiation load.

Fig. 3.15 (a) Fracture surface of SAC305 solder joint with $t=400$ µm at the initiation location in a mode-I specimen, showing features of ductile failure. (b) The Cu$_6$Sn$_5$ intermetallic layer.
(IMC) at the solder-copper interface provides potential nucleation sites for voids and dimples formed during ductile failure.

### 3.4.5 Effect of Loading Rate on Fracture

Figure 3.16 shows that the mode-I $G_{ci}$ of $t=200 \mu$m joints increased significantly with the loading rate; i.e., from an average $G_{ci} = 544 \text{ J/m}^2$ for 0.1 mm/min to $G_{ci} = 1,230 \text{ J/m}^2$ at 5 mm/min. This corresponded to strain rates of approximately $6 \times 10^{-5} \text{ s}^{-1}$ and $1 \times 10^{-3} \text{ s}^{-1}$, respectively. As mentioned earlier, each DCB specimen was used to measure two $G_{ci}$ values, one at the lower loading rate from the start of the solder layer, and a second at the higher rate from the short crack (< 1mm) created after the first loading. Therefore, a very small amount of R-curve toughening (Fig. 3.5) may have artificially raised the apparent $G_{ci}$ at 5 mm/min. Nevertheless, this effect would have been small, since the rising slope for most cases was less than 90 J/m$^2$/mm which means that the increase in $G_c$ would have been less than 90 J/m$^2$. Therefore, the results indicate that the toughness of the solder joint increased markedly with loading rate. This is consistent with the results reported in [24] where the shear force required to break flip chip solder bumps increased with the shear speed. A comprehensive understanding of the solder joint fracture at higher strain rates would be useful for failure prediction of portable electronic devices during drop impact. It is anticipated that the critical strain energy release rate methodology to predict the fracture load of solder joints under quasi-static loading [12, 13] can be extended to higher strain rate conditions.
Fig. 3.16 Effect of loading rate on the mode-I $G_{ci}$ of SAC305-Cu solder joints with $t=200$ μm.

3.5 Conclusions

The effect of substrate stiffness on the initiation strain energy release rate, $G_{ci}$, was negligible, which implies that the $G_{ci}$ measured from DCB specimens can be used to predict the fracture loads of joints of much smaller joints such as BGAs.

Similarly, the variation of the $G_{ci}$ with solder thickness was statistically insignificant over the range $t=200 - 400$ μm for the three phase angles $\psi=0^\circ$, 25°, and 45°, respectively. In contrast, the steady-state critical strain energy release rate, $G_{cs}$, did increase significantly with the solder thickness. These observations imply that the effect of solder layer thickness can be neglected when predicting the strength of relatively short solder joints using $G_{ci}$. However, longer joints may support sufficient subcritical crack growth to realize appreciable toughening which will increase with the solder layer thickness.

Both $G_{ci}$ and $G_{cs}$ were relatively constant between $\psi=0^\circ$ and 25°, but increased at $\psi=45^\circ$ due to the increased contribution of shear loading. Similarly, the crack path was found to be influenced by the mode ratio of loading and its effect on the stress state at the crack tip. The crack path was highly three-dimensional for phase angles below $\psi=25^\circ$ and predominantly
planar for higher phase angles $\psi=45^\circ$. The crack paths followed the contour of maximum von Mises strain rather than the maximum principal stress.

Consistent with earlier work on solder joint strength, some preliminary results indicated that the loading rate had a significant effect on $G_{ci}$.

### 3.6 References


[17] Jenq ST, Chang HH, Lai YS, Tsai TY, High strain rate compression behavior for Sn-37Pb eutectic alloys, lead-free Sn-1Ag-0.5Cu and Sn-3Ag-0.5Cu alloys. Microelectron Reliab 2009; 49: 310-17.


Appendix 3.A

The parameters $\Phi_I$ and $\Phi_{II}$ which are used in $G_c$ calculations are defined as,

$$
\Phi_I = 1 + 0.667 \frac{H}{a} [(1 - \frac{T}{H})^3 [1 + \frac{T}{H} (2E_{a}/E_a - 1)]]^{0.25}
$$

$$
\Phi_{II} = 1 + 0.206 \frac{H}{a} \sqrt{1 - \frac{T}{H} \left[1 + \frac{2TEa}{GaH}\right]}
$$

(A1),

where $E$, $E_a$ and $\nu$, $\nu_a$ are the Young’s modulus and Poisson’s ratio of the bars and solder, respectively (Fig. A1), and $G_a$ is the shear modulus of solder. The symbol $\alpha$ is a calibration constant equal to 2.946 [16]. The rest of the parameters are defined in Fig. A1.

Fig. A1 Definition of symmetric DCB specimen parameters used for $G_c$ calculations.
Chapter 4

4 Fracture Load Prediction of Lead-free Solder Joints

4.1 Introduction

Solder joint failure by crack propagation, either due to thermal loads or mechanical loads, is a significant reliability concern in electronic devices. The majority of the research in this area has dealt with thermal fatigue. Solder joint failure under drop impact and vibration loading has also been of great interest, but relatively little attention has been paid to the development of methods of predicting the fracture load of solder joints under quasi-static mechanical loads, applied either directly on components or induced by the bending or twisting of printed circuit boards (PCBs). Most existing experimental methods to evaluate the strength of solder joints under mechanical loads are primarily qualitative and do not provide fundamental mechanical properties that can be used to predict the strength of joints in other configurations or loads. Therefore, tests such as the ball shear [1, 2], ball pull [3], board level bending [4], and board level drop tests [5,6] are useful mainly for quality control and not for failure prediction.

Tan et al. [7] have proposed a force-based failure criterion by measuring the strength of individual 500 μm Sn-Pb solder balls subjected to different combinations of normal and shear loads and creating a force-based failure envelope. Using this criterion they were able to predict the failure of solder joints in a board level bending test. The results illustrated the different roles of loads normal and parallel to the plane of the solder joint; however, the limitations of this type of force-based prediction method are similar to those of the previously mentioned quality control tests; i.e. force-based criteria are applicable only to the particular type of joint for which the force envelope was developed (e.g. a 500 μm solder ball in [7]). Fracture-based criteria such as the critical energy release rate $G_c$ as a function of the mode ratio of loading [8, 9,10] and cohesive zone models [11, 12, 13] have been used widely to predict failure in adhesive joints.
Siow and Manoharan [14, 15] used modified compact tension specimens to compare the fracture toughness of Sn-Pb and Sn-Ag solder joints under mode-I and mixed-mode I-III loading with copper and nickel finishes. They noted that Sn-Pb solder joints were tougher than Sn-Ag solder joints irrespective of substrate finish. This was because fracture occurred mainly in the bulk solder and not in the brittle intermetallic layer. An interesting, counter-intuitive result was that the mode-I fracture toughness was greater than that for mixed mode I-III loading.

The quasi-static fracture behavior of Cu/Sn3Ag0.5Cu (SAC305) solder joints was investigated using a double cantilever beam (DCB) specimen configuration loaded in mode I and various mixed-mode I-II conditions in Chapter 2 and [16]. It was observed that cracks initiated at a low critical energy release rate, \( G_{ci} \), and the joints toughened with crack growth until a steady state value (\( G_{cs} \)) was reached (see Fig. 4.1). This implies that relatively large solder joints, such as those in heat sink attachments or power electronic modules, will have an ultimate strength that is much greater than that indicated by crack initiation. However, the fracture of a smaller joint such as a BGA is mainly governed by the critical energy release rate at initiation, \( G_{ci} \), since subcritical crack growth is small in this case. It was noted that \( G_{ci} \) and \( G_{cs} \) increased with the size of the mode II component. It was also observed that the local geometry of the end of the solder joint had only a relatively small effect on \( G_{ci} \).

The fracture of a thin layer joining two substrates has been modeled successfully in many applications using the cohesive zone approach in which the global structural effect of the micro-mechanisms of fracture at the crack tip are mimicked using a traction-separation law [17]. The latter is simply the relation between the force (traction) between the fracture surfaces as a function of the separation between the fracture surfaces (Fig. 4.2). Two important parameters governing a traction-separation law are \( \Gamma \) and \( \hat{\sigma} \) which represent the area under the traction-separation curve (energy dissipation) and the maximum traction stress, respectively. The governing law for a specific interface is often found by assuming a relation of a certain form and then matching a finite element simulation with experimental observations. For example, Mohammed and Liechti [18] assumed a bilinear traction-separation law to simulate the four-point bending of an aluminum-epoxy adhesive specimen by using initial guess values of \( \Gamma \) and \( \hat{\sigma} \). The parameters were then adjusted to provide the best match with experimentally measured displacements near the adhesive crack tip. Towashiraporn and Xie [19] used a similar approach with a SnPb solder-copper tensile specimen to obtain the traction-separation relation. This
relation was then used to model the displacement behavior of a PCB under impact loads; however, no attempt was made to predict solder joint failure.

The objective of the present chapter was to examine two different solder joint fracture criteria that would permit the prediction of solder joint strength for a wide range of joint geometries and types of load. The first part of the work involved fracture experiments on Cu-SAC305 DCB specimens under mode-I loading to measure both $G_{ci}$ and cohesive zone model (CZM) parameters. In the second part, fracture tests were performed on discrete solder joints arranged in a linear array between two copper bars, in order to understand the fracture behavior of individual solder joints and evaluate the $G_{ci}$ and CZM failure criteria using linear elastic finite element analysis.

![R-curve of SAC305/Cu joint specimens tested under mode I loading](image)

Fig. 4.1 R-curve of SAC305/Cu joint specimens tested under mode I loading [16].
4.2 Experimental Procedures

4.2.1 Specimen Preparation

Two different specimen configurations were used - a double cantilever beam (DCB) fracture specimen (Fig. 4.3a) and a model joint specimen (Fig. 4.3b) that mimicked aspects of an array-type commercial electronic package, such as ball grid array (BGA). The DCB specimen consisted of two copper bars (C110 alloy, 160x12.6x12.6 mm) joined with a continuous layer of 400 µm thick Sn3.0Ag0.5Cu (SAC 305) solder (Fig. 4.3a). The model specimen consisted of similar copper bars joined with discrete 400 µm thick solder joints of two different lengths; $l = 2$ mm and $l = 5$mm (Fig. 4.3b).

The fabrication process of both specimen configurations was very similar. Initially, the Cu bars (C110 alloy) were cut to the required dimensions and the bonding surfaces were polished for 5 min using an orbital sander fitted with an ultra fine silicon carbide/nylon mesh abrasive pad. To avoid edge rounding, eight bars were placed adjacent to each other and sanded simultaneously. This process produced a repeatable surface roughness very close to that of an organic solderability preservative (OSP) finish on commercial PCBs [16]. After polishing, the Cu bars were rinsed thoroughly with water and wiped with cheese cloth to remove debris, and then rinsed with acetone.

Fig. 4.2 The bilinear cohesive zone traction-separation law used in the FE analysis.
Kapton tape was used to mask the solder joint areas (Figs. 4.3 and 4.4). The round steel wires shown in Fig. 4.3 maintained the required gap of 0.4 mm between the Cu bars.

Fig. 4.3 Schematic of a) the DCB specimen and b) discrete solder joint specimen ($l=2$ mm and 5 mm). The width of the specimen was 12.6 mm and overall length was 160 mm. All dimensions in mm.
The masked copper bars were placed on a hot plate covered with aluminum foil and maintained at 290°C with the bonding surfaces vertical (Fig. 4.4). The temperature of the copper bars was monitored continuously with thermocouples inserted in holes drilled just beneath the surfaces to be soldered. When the temperature of the bars reached 220-225°C, a flux-cored SAC 305 0.75 mm solder wire (Kester Inc., USA) was touched to the prepared vertical surfaces so that they became rapidly covered with a thin layer of solder. The bars were then clamped together against 400 μm steel wires to maintain the desired solder thickness (Fig. 4.4). This procedure minimized voiding caused by flux entrapment as excess solder and flux residues flowed out of the joint as the bars were brought together. In the model joint specimens (Fig. 4.3b) the steel wires were placed at each solder joint to maintain the gap and to shield each joint from compressive loads when the preceding was loaded during the fracture tests. The local geometry of the end of solder layer in the DCB was defined by the smooth, square shape of the edge of the Kapton tape. Earlier work showed that the local edge geometry had a relatively small effect on the quasi-static fracture load [16].

Fig. 4.4 Schematic of specimen arrangement during soldering of a DCB and discrete joint specimens.
The entire soldering process from the time of first solder application took approximately 15 to 20 s. It was followed by a further 100 s dwell period on the hot plate to achieve a time above liquidus of 120 s, and a peak temperature of between 245°C and 250°C. The specimens were then placed transversely on supports in a small wind tunnel and cooled with forced air at a cooling rate of 1.4-1.6°C/s (Fig. 4.5), which is typical of microelectronics manufacturing. A detailed microstructural study of similar DCB specimens confirmed that the intermetallics and bulk solder were similar to those typical of commercial SAC305 solder. After cooling to room temperature, the sides of the specimens were machined to remove the excess solder and create a smooth surface to facilitate the observation of cracks in the solder layer. During this step, approximately 300 μm of the copper layer was removed in about six passes using a sharp fly cutter to prevent copper from smearing over the interfaces and obstructing the view of cracks. Finally, the loading pin holes and clip gauge mounting holes were drilled in the copper bars.

![Fig. 4.5 Time temperature profiles of the DCB and model joint specimens.](image-url)
4.2.2 Fracture Tests

The DCB and the discrete joint specimens were tested in mode-I under displacement control with a constant cross-head speed of 0.1 mm/min. The crack initiation and propagation were observed using a microscope with 1.9 mm diameter field of view, mounted on a micrometer stage. The visible edge of the solder layer was painted with a thin layer of diluted paper correction fluid to facilitate the identification of the crack initiation and propagation. Table 4.1 shows the number of specimens tested for each configuration. A clip gauge was used to measure the opening displacement of the arms near the loading pins in the DCB and discrete joint specimens (Figs. 4.3a and 4.3b). In some of the DCB specimens, the opening displacement of the end of the solder layer was measured directly (Fig. 4.3a), in order to determine the CZM parameters for the solder. The load corresponding to crack initiation in DCB specimens was identified using two different methods. Visual inspection could reveal a crack of 50-100 µm length. Crack initiation also corresponded to the onset of nonlinearity in the load-displacement response of the DCB. Crack initiation in the discrete solder joint specimen was identified mainly by the visual method.

Table 4.1 Number of specimens tested in each configuration.

<table>
<thead>
<tr>
<th>Specimen type</th>
<th>Configuration (Fig. 4.3)</th>
<th>Number of specimens tested</th>
</tr>
</thead>
<tbody>
<tr>
<td>DCB</td>
<td>(a = 40.5) mm</td>
<td>9</td>
</tr>
<tr>
<td>Discrete joint specimen ((l = 2) mm)</td>
<td>(a = 37) mm, (p = 37) mm</td>
<td>5</td>
</tr>
<tr>
<td></td>
<td>(a = 44) mm, (p = 30) mm</td>
<td>3</td>
</tr>
<tr>
<td>Discrete joint specimen ((l = 5) mm)</td>
<td>(a = 40.5) mm, (p = 35) mm</td>
<td>6</td>
</tr>
</tbody>
</table>
4.3 Finite Element Models

4.3.1 Cohesive Zone Modeling

Crack initiation was simulated using the cohesive zone model available in ANSYS 12® finite element software [20]. The finite element mesh of the continuous joint DCB specimen and the discrete joint specimen are shown with the boundary conditions in Figs. 4.6 and 4.7, respectively. The cohesive zone modeling procedure for these two specimens was identical. The Cu bars and solder layer were meshed with the PLANE 82 2-D 8-node structural elements ranging in size from 0.07 mm in the solder layer to 1.5 mm in the Cu bar far from the crack tip. The model results were found to be independent of the mesh size in this range. The elements in the Cu bars were modeled as plane stress while the solder elements were modeled as plane strain. The mechanical behavior of the elements was defined as isotropic linear elastic with the properties of Table 4.2.

The upper solder-Cu interface in both the continuous joint DCB (Fig. 4.6) and the discrete joint DCB (Fig. 4.7) was modeled using 2-D surface-to-surface contact elements [20]. The surface on the solder layer side was defined as a contact surface with the 3-node CONTA 172 elements, and the surface on the Cu bar side was defined as target surface with the TARGE169 elements [20]. One advantage of CONTA 172 elements is that the mesh size of the two materials bounding the interface need not match and can be relatively coarse (Fig. 4.6). The constitutive behavior of this interface was represented with the bilinear (triangular) traction-separation (or cohesive zone) law implemented in ANSYS (Fig. 4.2).
Fig. 4.6 Finite element mesh of the DCB specimen showing the overall view with boundary conditions and the detail near the solder layer. The top solder-Cu interface was modeled using surface-to-surface contact elements with a traction-separation constitutive behavior.

Figure 4.2 shows the relation between separation, $\delta$, and the normal stress, $\sigma$, across the interface. The behavior is initially elastic, but after attaining a peak stress $\sigma^*$ the interface softens gradually reaching zero strength at a critical separation displacement, $\delta_c$. At this critical displacement the two materials separate completely, corresponding to crack initiation or extension. The area under the curve, $\Gamma$, represents the fracture energy dissipated by yield, cavitation and micro-cracking in the damage zone near the crack tip. The procedure used to determine these parameters is explained in Section 4.4.3. The initial slope $K$, the interface
stiffness, is not believed to play a significant role in the failure load predictions [21]; hence, a default contact stiffness option in ANSYS was used which resulted in a value of $10^{14}$ N/m$^3$. This value was reasonable, because it was high enough to avoid any contribution from the model interface to the overall compliance of the system [21] and low enough to prevent numerical convergence issues. This means that the primary functionality of the cohesive zone law in the current study was to mimic the fracture processes near the crack tip. The finite element mesh of the discrete joint specimen was similar to that of the DCB (Fig. 4.7).

Fig. 4.7  The finite element mesh of the discrete joint specimen ($l=2$ mm) with the boundary conditions and the detail near a typical solder joint. The upper solder-Cu interface was modeled using contact elements that behaved according to the cohesive zone law.
The applied loads in the simulation were increased slowly to compensate for the numerical instabilities that can arise due to the debonding processes. The cohesive zone model in ANSYS 12 uses an artificial damping coefficient $\eta$ [20] to stabilize the solution. A value of 0.0001 was used in the present analysis.

Table 4.2  Mechanical properties used in the FE analysis [16].

<table>
<thead>
<tr>
<th>Material</th>
<th>Tensile modulus (GPa)</th>
<th>Poisson ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cu (110 alloy)</td>
<td>124</td>
<td>0.35</td>
</tr>
<tr>
<td>SAC305 (Sn3.0Ag0.5Cu)</td>
<td>51</td>
<td>0.4</td>
</tr>
</tbody>
</table>

4.3.2  $G_{ci}$ Calculation Using Finite Elements

In contrast to the CZM simulation, a crack was modeled explicitly in this case. It was located in the middle of the solder layer, as shown in Fig. 4.8. The region near the crack tip was meshed with singular elements (i.e. 6-noded triangular elements formed by collapsing 8-noded PLANE183 elements) available in ANSYS (triangles surrounding crack tip in Fig. 4.8). These elements can capture the $r^{1/2}$ singularity near the crack tip and are efficient for $G$ calculations. The length of the singular element was maintained at 0.05 mm. Everywhere else, the model was meshed with PLANE183 elements; the solder in plane strain and the copper bars in plane stress. The mesh was graded ensuring that the solder layer contained at least 8 elements in the thickness direction (Fig. 4.8). The solution method was a linear elastic static case with the material properties of Table 4.2.
The strain energy release rate, $G$, was calculated using

$$G = (1 - \nu^2) \frac{K_I^2}{E},$$

(1)

where $\nu$ and $E$ are the Poisson ratio and tensile modulus of the solder, respectively, and $K_I$ is the mode-I stress intensity factor obtained directly from the FE model for the given boundary conditions (i.e., loading and constraint). A hypothetical crack of length equal to 0.5 mm was considered in the failure load predictions of 2 mm and 5 mm joints.

Fig. 4.8 Mesh near the crack tip, denoted as A, in a typical finite element mesh used for $G_c$ calculation.

### 4.4 Results and Discussions

#### 4.4.1 Crack Initiation and $G_{ci}$ of DCB

Figure 4.9 shows a typical load-displacement response of a DCB specimen under mode-I loading (data points recorded every 2.2 N), and depicts the onset of nonlinearity used to infer crack initiation. The procedure to find the onset of nonlinearity was similar to the method given
in [22, 23]; i.e. a best-fit straight line from the origin to a load equal to 80% of the maximum load was used to represent the elastic loading curve. This best-fit straight line is extended beyond the elastic region with the same slope. Crack initiation was assumed to occur when the difference between the measured load and the predicted elastic load exceeded 3%, as indicated by a filled circle in Fig. 4.9. The $G_{ci}$ corresponding to the onset of nonlinearity had an average value of 380 J/m$^2$, which was approximately 20% less than the average $G_{ci}$ calculated using visual inspection of the solder layer (480 J/m$^2$) as shown in Fig. 4.10. Perhaps the most significant reason for this difference was that the 50-100 μm crack was visible only at the edge of the specimen where a state of plane stress existed. The fracture surfaces revealed, however, that the crack front was usually curved, with a leading edge in the central portion of the joint width, where a state of plane strain existed, that was approximately 1 mm ahead of the cracks along the edges of the specimen. Thus the load corresponding to the appearance of a crack on the edge may have been slightly higher than the load at which initiation occurred in the middle of the joint width. Furthermore, the visual method will also overestimate the load because of the slight $R$-curve toughening that will occur between the point of initiation and a crack length of 50-100 μm. Therefore, the onset of nonlinearity provided a more conservative estimate of the joint fracture strength.

![Fig. 4.9 Typical load-displacement response of a DCB specimen tested under mode I loading.](image-url)
Fig. 4.10 Mode-I critical strain energy release rate corresponding to crack initiation for SAC305-Cu DCB specimens as detected by visual inspection and the onset of nonlinearity (NL).

It can be observed from Fig. 4.9 that the load on the DCB continued to increase beyond the crack initiation stage until it reached a maximum and then declined with increasing displacement. This increase in load after initiation was due to the R-curve behavior, as discussed previously and reported in [16]. Once the crack-tip damage zone attained a steady-state condition (corresponding to the maximum load in Fig. 4.9), the crack advanced at a constant steady-state energy release rate, \( G_{cs} \) [16]. Although the main objective of the current chapter was crack initiation and the conditions leading to it, understanding the above relation between the load-displacement response and R-curve toughening was useful in comparing the failure of solder joints of different lengths, as will be seen below.

4.4.2 Fracture of Discrete Joint Specimens

The load-displacement response of a typical discrete joint specimen \((l=5 \text{ mm})\) is shown in Fig. 4.11. As the adherends were pulled apart under mode I loading (Fig. 4.3b) the load
increased linearly with displacement until crack initiation at 670 N. As in Fig. 4.9 for the DCB specimens, the load continued to increase with further crack growth to reach a maximum. Similar behavior was observed in the loading of the 2nd and 3rd joints as well, resulting in the other two peaks in Fig. 4.11. This was again attributed to the toughening caused by the growth of the damage zone after initiation as the crack grew about 1 to 2 mm before propagating unstably after the maximum load was reached. Figure 4.12 shows that toughening was too small to be seen after initiation for the discrete joint specimens with \( l = 2 \) mm. In this case, crack initiation was followed immediately by rapid propagation since the 2 mm solder joint length was too small to sustain a period of stable, R-curve toughening.

Figure 4.13 shows the crack initiation loads of \( l = 5 \) mm joints as a function of loading arm length \( (a \text{ in Fig. 4.3b}) \) in six specimens. As expected, the failure loads decreased as the length of the moment arm increased. A similar trend is observed in Fig. 4.14 which shows the failure loads of \( l = 2 \) mm joints for ten different specimens, as function of loading arm length. The data points of Fig. 4.14 are the crack initiation loads, equal to the maximum loads at break, shown in Fig. 4.12, but for all of the tested \( l = 2 \) mm joints. Figure 4.15a shows both fracture surfaces of a DCB specimen, illustrating the three-dimensional nature of the crack path which was very close to either interface, causing macroscopically smooth fracture surfaces. The frequent crack jumps between the two interfaces across the width and along the crack growth direction were caused by the mode-I loading and the intermetallic microstructure at the interfaces [16]. In some cases, cracks grew along both interfaces simultaneously leaving raised solder foils that bridged the crack faces. Fig. 4.15b illustrates that the fracture surfaces observed in \( l = 5 \) and \( l = 2 \) mm joints were similar to those of the continuous solder joint of Fig. 4.15a. However, because there was very little stable, subcritical R-curve crack growth in the \( l = 2 \) mm joints, these cracks tended to propagate very quickly along only one interface, producing a macroscopically smooth, planar surface.
Fig. 4.11  Applied force vs. the opening displacement of the loading pins for a discrete joint specimen with \( l = 5 \) mm under mode-I loading.

Fig. 4.12  Load versus loading pin opening displacement of a discrete joint specimen of \( l = 2 \) mm, from mode-I experiments.
Fig. 4.13 Measured crack initiation loads for six specimens with $l=5$ mm and predictions based on the CZM ($I=\dot{G}_{ci}=380$ J/m$^2$ and $\dot{\sigma} =120$ MPa) and $G_{ci}$ methods. The symbols represent experimental data from the different specimens. The loading arm length is distance $a$ in Fig. 4.3b.

Fig. 4.14 Measured crack initiation loads for eight specimens with $l=2$ mm and predictions based on the CZM ($I=\dot{G}_{ci}=380$ J/m$^2$ and $\dot{\sigma} =120$ MPa) and $G_{ci}$ methods. The symbols represent experimental data from different specimens. The loading arm length is distance $a$ in Fig. 4.3b.
Fig. 4.15 (a) Fracture surfaces of a mode-I DCB specimen. (b) Comparison of fracture surfaces for DCB, $l=5$ mm and $l=2$ mm joints, showing only one side of each specimen. Scale in mm.
4.4.3 Determining the CZM Traction Separation Law

Since the goal of the present CZM simulations was the modeling of crack initiation, R-curve toughening behavior was not modeled, and hence the damage zone energy represented by $\Gamma$ was equated to 380 J/m$^2$, the average mode-I $G_{ci}$ measured for the SAC305-Cu DCB specimens. The value of $\hat{\sigma}$ was determined using an iterative procedure to obtain the best match between model predictions and the experimental measurements of the load-displacement behavior at the end of solder joint where cracks initiate. The opening displacements at the loading pins, $u$, (Fig. 4.6) were prescribed in the finite element (FE) model over the range used in the experiments, and the unknown parameter $\hat{\sigma}$ was initially set to 50 MPa, a value close to the yield stress of the solder. Then the value of $\hat{\sigma}$ was incremented to 100, 120 and 150 MPa in subsequent simulations to produce four simulated load-displacement curves (Fig. 4.16). It is seen that the $\hat{\sigma}$ parameter had a significant effect on the load-displacement response as well as the peak load.

The experimentally measured opening displacement at the end of the solder layer (i.e. at the location of crack initiation) as well as the crack initiation load were compared with these FE model results. Figure 4.17 shows that the results corresponding to $\hat{\sigma}$ =120 MPa and $\Gamma$ =380 J/m$^2$ provided a good match with the experimental observations, both in terms of the deformation response and the crack initiation load. As mentioned previously, since initiation corresponded to the ultimate joint strength in short joints, only the crack initiation was modeled with the CZM, so the toughening beyond initiation did not match the measured trend.

The sensitivity of the failure load prediction to uncertainty in $\Gamma$, or equivalently $G_{ci}$, was estimated by performing additional simulations using $\hat{\sigma}$ =120 MPa and $\Gamma$ values of 380 J/m$^2$ and 480 J/m$^2$, which were the average $G_{ci}$ from the nonlinearity and visual crack detection methods, respectively. Figure 4.18 shows that a 20% decrease in $\Gamma$ (i.e. from 480 to 380 J/m$^2$) produced an 11 % decrease in the predicted crack initiation load. However, a similar 20% change in $\hat{\sigma}$ (i.e. from 120 to 150 MPa) resulted in only a 3% change in the predicted load (Fig. 4.16). Hence, the failure load was relatively more sensitive to $\Gamma$ than the cohesive stress $\hat{\sigma}$. 
Fig. 4.16 FE CZM predictions of the DCB load vs. opening displacement near the crack initiation point shown in Fig. 4.3a for four values of $\sigma$, all of them using $G_{ci} = 380\text{J/m}^2$. 

![Graph showing FE CZM predictions of the DCB load vs. opening displacement for different stress levels.]

Experimental data
CZM simulation 120 MPa

Crack initiation
Toughening

Opening displacement at the end of solder layer (mm)
Force (N)
Fig. 4.17 Measured and predicted load vs. opening displacement near the leading edge of the solder layer (crack initiation point, Fig. 4.3a) for the CZM simulation corresponding to $G_{ci} = 380$ J/m$^2$ and $\sigma^\ast = 120$ MPa.

![Graph showing load vs. opening displacement for CZM simulation](image)

Fig. 4.18 Comparison of the predicted load-displacement response near the leading edge of the solder layer in the DCB for simulations using $\sigma^\ast = 120$ MPa and two values of $G_{ci}$.

4.4.4 Failure Load Predictions for Discrete Solder Joints

4.4.4.1 Discrete Joint Specimens with $l=2$ mm

The loads corresponding to crack initiation in 5 mm and 2 mm joints were predicted using the $G_{ci}$ and CZM models explained in Sections 4.3.1 and 4.3.2. Figure 4.14 shows the failure loads of $l=2$ mm joints in ten different specimens as function of loading arm length. It can be seen that the predictions of both the failure models were very close, and agreed
reasonably well with experiments. Table 4.3 shows that the models predicted the mean experimental failure loads for the $l=2$ mm joints to within 11%.

Table 4.3 Comparison of mean crack initiation loads from experiments with the predictions of the CZM ($\Gamma=G_{ci}=380 \text{ J/m}^2$ and $\sigma =120 \text{ MPa}$) and $G_{ci}$ methods for $l=2$ mm discrete joints.

<table>
<thead>
<tr>
<th>Moment arm length $a$ in Fig. 4.3b (mm)</th>
<th>Experimental failure load (N) (mean ± standard deviation)</th>
<th>CZM prediction</th>
<th>$G_{ci}$ prediction</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Load (N)</td>
<td>% difference</td>
</tr>
<tr>
<td>37</td>
<td>694 ± 138</td>
<td>770</td>
<td>11</td>
</tr>
<tr>
<td>74</td>
<td>447 ± 46</td>
<td>439</td>
<td>-2</td>
</tr>
<tr>
<td>104</td>
<td>360 ± 80</td>
<td>324</td>
<td>-10</td>
</tr>
<tr>
<td>111</td>
<td>297 ± 40</td>
<td>301</td>
<td>2</td>
</tr>
</tbody>
</table>

Due to the fact that these 2 mm long joints did not experience any toughening (Fig. 4.12) and crack initiation marked their failure, the CZM could predict the load-displacement behavior of the specimen quite accurately as shown in Fig. 4.19. The deviations in the curves after the peak loads for each joint in the sequence were due to the compliance of the load train (grips, links and load cell) which was not included in the CZM model.
4.4.4.2 Discrete Joint Specimens with \( l=5 \) mm

The model predictions for \( l=5 \) mm joints agreed reasonably well with the experiments as seen in Fig. 4.13 and Table 4.4, although the predictions of both models were consistently smaller than the measured joint strengths, and were less accurate than the predictions for the 2 mm joints (Table 4.3). This was probably attributable to the small amount of R-curve toughening that can occur in the 5 mm joints, causing the onset of final fracture to coincide with a higher strain energy release rate than the assumed \( \Gamma=\Gamma_{ci}=380 \) J/m\(^2\). For example, if \( \Gamma=\Gamma_{ci}=480 \) J/m\(^2\) were used to account for the additional toughening prior to failure in the 5 mm joints, both the CZM and \( G_{ci} \) models would predict the joint strength to within 12%.

Table 4.4 Comparison of mean crack initiation loads from experiments with the predictions of the CZM (\( \Gamma=\Gamma_{ci}=380 \) J/m\(^2\) and \( \tilde{\sigma} =120 \) MPa) and \( G_{ci} \) methods for \( l=5 \) mm discrete joints.
<table>
<thead>
<tr>
<th>Moment arm length ( a ) in Fig. 4.3b (mm)</th>
<th>Experimental crack initiation load (N)</th>
<th>CZM prediction</th>
<th>( G_{ci} ) prediction</th>
</tr>
</thead>
<tbody>
<tr>
<td>40</td>
<td>712 ± 60</td>
<td>667</td>
<td>-7</td>
</tr>
<tr>
<td>75</td>
<td>491 ± 17</td>
<td>400</td>
<td>-19</td>
</tr>
<tr>
<td>110</td>
<td>342 ± 14</td>
<td>295</td>
<td>-14</td>
</tr>
</tbody>
</table>

An important practical implication of the observations in this section is that fracture of a joint whose length is less than 2 mm does not undergo appreciable toughening, and hence the \( G_{ci} \) from the DCB specimen is sufficient to predict the ultimate strength of such short joints. It is likely therefore, that fracture initiation data obtained from a relatively large solder joint may be useful in predicting the quasi-static fracture of a solder ball in a ball grid array.

### 4.4.5 Effect of Joint Spacing and Substrate Stiffness on the Failure Load of the First Joint

The discrete joint specimens \( (l=2 \text{ mm}) \) with a pitch \( p=30 \) or 37 mm were designed to test the failure loads of individual solder joints. However, as the joint spacing \( p \) becomes smaller or the stiffness of the substrate changes, crack initiation, growth, and the failure load of the first joint will be affected by load sharing among neighboring joints.

Two sets of CZM simulations, with \( l=2 \) mm joints, were performed to investigate this behavior. In the first, the pitch \( p \) was varied to determine its effect on the failure load (simulations 1, 2 and 3 in Table 4.5). The second set of simulations examined the effect of loading arm bending stiffness on the failure load of the solder joints (simulations 2 and 4 in Table 4.5). The bending stiffness, \( EI \), of the loading bars was varied by changing the tensile
modulus while keeping the moment of inertia $I$ constant. The CZM parameters were $\Gamma = G_{ci} = 380$ J/m$^2$ and $\sigma = 120$ MPa. The displacement $u$ of the adherends was increased monotonically as the individual solder joints continued to fail one by one. The distance from the loading pins to the first joint ($a$ in Fig. 4.3b) was equal to 40 mm, and eight joints were modeled, with different $p$ values as per Table 4.5.

Table 4.5  Parameters used in CZM simulations of pitch and stiffness effects for an array of discrete joints with $l=2$ mm.

<table>
<thead>
<tr>
<th>Simulation</th>
<th>Pitch of joint $p$ (mm)</th>
<th>Tensile modulus $E$ of adherends (GPa)</th>
<th>Predicted Failure load of first joint (N)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>3</td>
<td>124</td>
<td>674</td>
</tr>
<tr>
<td>2</td>
<td>6</td>
<td>124</td>
<td>636</td>
</tr>
<tr>
<td>3</td>
<td>9</td>
<td>124</td>
<td>633</td>
</tr>
<tr>
<td>4</td>
<td>6</td>
<td>51</td>
<td>466</td>
</tr>
</tbody>
</table>

The force-displacement curve was extracted from the node where the displacement $u$ was prescribed (Fig. 4.3b), and is shown in Fig. 4.20 for the first set of simulations (1, 2, 3 of Table 4.5). The distinct peaks labeled as J1, J2, and J3 of the ‘pitch=3mm’ curve represent the failure loads of first three solder joints in the specimen. The three peaks observed in the other two curves have the same meaning. From a practical standpoint, the failure of the first joint, J1, is most important since it would define the onset of component failure. It is evident from the Fig. 4.20 that the failure load J1 decreased as the spacing (pitch) increased from 3-6 mm, but then remained unchanged from 6-9 mm (Table 4.5). This reflects the increased load sharing among joints as pitch decreases.
Fig. 4.20 The predicted force-displacement response of an array of discrete joints showing the effect of joint pitch. The peaks J1, J2, and J3 represent the failure of the first, second and third solder joints with \( l = 2 \) mm.

Figure 4.21 shows the predicted force-displacement curve from the second set of simulations (2, 4 of Table 4.5). It can be seen that a reduction of the substrate (loading arm) bending stiffness (simulated by changing \( E \) from 124 GPa to 51 GPa) resulted in a significant change in the failure load; i.e., a reduction from 636 N to 466 N. This was a consequence of greater load concentration at J1 with the more flexible adherend. This implies that stiffer materials increase the strength of solder joint arrays subject to external mechanical loads. It is interesting that this may conflict with a desire to use more flexible substrates to reduce thermal strains in solder joints [24].
4.4.6 Comparison of $G_{ci}$ and CZM Failure Criteria

It is evident from the previous sections that both the failure criteria, $G_{ci}$ and CZM, are capable of predicting the failure loads reasonably well. However, each model has certain advantages and limitations relative to the other. The $G_{ci}$ method is straightforward and simple to use. The only input parameter required for this model is $G_{ci}$, which can be obtained from simple DCB fracture experiments. However, three or four trial-and-error simulations are required to predict the failure load that gives $G=G_{ci}$ in each case. Each simulation takes about one minute with a computer that has 3.4 GHz dual core processor and 3.5 GB RAM. In contrast, the CZM requires no explicit finite element crack representation and requires only one simulation (running time about 2 h on a similar computer), to predict the failure load. A unique advantage of the CZM over the $G_{ci}$ method is that the CZM can simulate the simultaneous failure (damage) of joints, and hence can simulate the complex load sharing behavior in array joints such as BGAs. Also, the CZM can model crack growth behavior in longer joints, which is not possible with the $G_{ci}$ method. However, the numerical modeling of the CZM is relatively more complex compared
to $G_{ci}$ method and requires the determination of the solder-substrate system traction-separation law.

Hence, both these models can be useful depending upon the requirement. In other words, if the aim is just to predict the failure of solder joint then $G_{ci}$ method alone is sufficient. However, if the complex mechanical behavior along with the failure is to be modeled then CZM is the better option.

4.5 Conclusions

Continuous and discrete SAC305 solder joints of different lengths were made between copper bars under standard surface mount (SMT) processing conditions, and then fractured under mode-I loading. The load-displacement behavior corresponding to crack initiation and the subsequent toughening before ultimate failure were recorded.

The load corresponding to crack initiation in the DCB specimen with a continuous solder joint was identified using two different methods: one based on visual inspection and the other based on the onset of nonlinearity in load-displacement behavior. Based on these values, strain energy release rates at crack initiation, $G_{ci}$, were calculated. It was found that the $G_{ci}$ based on the onset of nonlinearity was, 380 J/m$^2$, while that from the visual method was 480 J/m$^2$. The larger value corresponding to the appearance of a crack at the edge of the joint was attributed to the state of plane stress at that location compared with the state of plane strain existing over most of the crack front.

From the discrete joint experiments, it was observed that R-curve toughening could increase the joint ultimate strength beyond the crack initiation value at $G_{ci}$. The longer joints ($l=5$ mm) experienced some toughening beyond crack initiation similar to that seen with the continuous solder joint DCB. The smaller discrete joints ($l=2$ mm) did not show any toughening and failed as soon as a crack initiated at $G_{ci}$. The fracture of these discrete solder joints was simulated using finite elements with two different failure criteria: one in terms of the critical strain energy release rate at initiation, $G_{cis}$, and another based on a cohesive zone model at the crack tip (CZM). The parameters of the CZM for the case of mode-I loading were obtained from an iterative
procedure using the measured crack initiation load and load-displacement curve of the continuous solder joint DCB. Both models predicted the fracture loads reasonably well (i.e., to within 12% accuracy). In addition, the CZM was able to predict accurately the overall load-displacement behavior of the discrete joint specimens and predict the load sharing that occurred between neighboring solder joints as a function of joint pitch and adherend stiffness. This has application in the modeling of the strength of solder joint arrays such as those found in ball grid array packages. These observations imply that the $G_{ci}$ obtained from a continuous solder joint DCB can be used to predict the ultimate strength of short joints with lengths less than 2 mm, provided that the microstructure of the joints was similar, i.e., the joints should have the same thermal history and substrate finish.

### 4.6 References


Chapter 5

5 Mixed-mode Fracture Load Prediction in Lead-free Solder Joints

5.1 Introduction

A large amount of research has been directed towards developing models to predict solder fracture under creep and fatigue loading and, thus the long-term life of solder joints. Solder joint failures due to impact forces and vibrations have also received a great deal of attention; however, relatively few efforts have been made towards developing methods to predict fracture loads of solder under quasi-static loads, either applied directly to components or induced by the bending or twisting of printed circuit boards (PCBs).

Existing strength evaluation methods for solder joints under mechanical loads such as the ball shear [1, 2], ball pull [3], board level bending [4], and board level drop tests [5] are primarily qualitative and provide relative rankings of joint strength. They do not measure fundamental mechanical properties that can be used to predict the strength of joints in other configurations or loads. For example, Tan et al. [6] have proposed a failure criterion based on a critical force by measuring the strength of individual 500 μm Sn-Pb solder balls subjected to combinations of normal and shear loads (i.e. mixed-mode loading). They predicted the failure of solder joints in a board level bending test with this force-based criterion. The results illustrated the different roles of loads normal and parallel to the plane of the solder joint; however, the limitation of this type of force-based prediction method is similar to that of the previously mentioned quality control tests; i.e. they are applicable only to the particular type of joint, substrate finish and component type for which the force envelope was developed (e.g. a 500 μm BGA solder ball with ENIG substrate finish in [6]).

Fracture-based criteria such as the critical energy release rate as a function of the mode ratio of loading, \( \psi \), have been used widely to predict failure in adhesive joints [7-10]. In this
case, fracture is predicted when the applied strain energy release rate at the particular mode ratio of loading (calculated from the applied loads, specimen dimensions and mechanical properties) equals the critical value for fracture at that mode ratio. In the following, if the applied strain energy release rate is calculated assuming elastic solder behavior it is denoted $G$, whereas it will be referred to as $J$ if calculated using an elastic-plastic model. Then $G_c$ and $J_c$ are the respective critical values for solder crack growth, while $G_{ci}$ and $J_{ci}$ are the critical values for crack initiation.

Yamada [11, 12] characterized the mode-I fracture of a soldered double cantilever beam (DCB) specimen consisting of two beryllium copper bars joined by a 250 μm thick 60Sn40Pb solder layer. It was observed that the strain energy release rate for mode-I fracture could be calculated analytically using either a beam on an elastic-plastic foundation model ($J$) or a beam on an elastic foundation model ($G$) [12]. Furthermore, it was demonstrated that the $J$-integral was path independent for interfacial cracks where the path of integration could cross the boundary of two dissimilar materials [13].

The quasi-static fracture behavior of Cu/Sn3Ag0.5Cu (SAC305) solder joints has been investigated using a DCB specimen loaded in mode I and various mixed-mode I-II conditions [14]. It was observed that cracks initiated at a low critical strain energy release rate, $G_{ci}$, and then the joints toughened with crack growth until a steady state value ($G_{cs}$) was reached. This R-curve behavior implies that relatively large solder joints, such as those in heat sink attachments or power electronic modules, will have an ultimate strength that is much greater than that indicated by crack initiation. In contrast, the fracture of a smaller joint such as a BGA is mainly governed by the critical energy release rate at initiation, $G_{ci}$, since subcritical crack growth is small in this case. It was noted that $G_{ci}$ and $G_{cs}$ increased with the size of the mode II loading component for phase angles greater than 25°. It was also observed that the local geometry of the end of the solder joint had only a relatively small effect on $G_{ci}$ [14].

The present authors were able to predict the fracture loads of discrete 2 mm and 5 mm SAC305 lead-free solder joints, arranged in a linear array between two copper bars, subjected to mode-I loads (i.e. forces normal to the plane of solder joint) using $G_{ci}$ and cohesive zone model parameters that were obtained from a continuous joint DCB specimen [15]. However, most solder joints in service experience mixed-mode loading, being subject to combinations of tensile and shear stress.
The objective of the present chapter was to examine solder joint fracture criteria that would permit the prediction of solder joint strength for a wide range of joint geometries under mixed-mode loading conditions. The first part of the work involved mixed-mode fracture experiments on Cu DCB specimens joined with a continuous SAC305 solder layer in order to measure the fracture parameters corresponding to crack initiation; i.e. the critical strain energy release rate as a function of the mode ratio of loading, $G_{ci}(\psi)$ and $J_{ci}(\psi)$, where $\psi$ is the phase angle of loading, expressing the mode ratio as

$$\psi = \arctan\left(\frac{K_{II}}{K_I}\right),$$

where $K_{II}$ and $K_I$ are mode-II and mode-I stress intensity factors, respectively.

In the second part of the present work, fracture tests were performed on discrete 2 mm and 5 mm solder joints arranged in a linear array between two copper bars, in order to understand the mixed-mode fracture behavior of discrete solder joints and to evaluate the proposed failure criterion, $J_{ci}(\psi)$, using elastic-plastic finite element analysis (FEA). Finally, the $J_{ci}$ values calculated using FEA were validated by comparing them with the $J$-integral values estimated from the measured critical opening displacements near the location of crack initiation.

### 5.2 Experimental Procedures

#### 5.2.1 Specimen Preparation

A double cantilever beam (DCB) fracture specimen made with a continuous solder joint (Fig. 5.1a) and DCB specimens made with discrete joints (Fig. 5.1b) were used in the fracture experiments. The arms of the DCBs consisted of two copper bars (C110 alloy, 160x12.6x12.6 mm) joined with either a continuous layer of 400 μm thick Sn3.0Ag0.5Cu (SAC 305) solder or with discrete 400 μm thick solder joints of two different lengths; $l = 2$ mm and $l = 5$mm (Fig. 5.1 (b)).
The soldering process used for both specimen configurations was very similar, and was the same as that in [14]. After cutting the Cu bars to the required dimensions, the bonding surfaces were polished for 5 min using an orbital sander fitted with an ultra fine silicon carbide nylon mesh abrasive pad. This process produced a repeatable surface roughness ($R_a=0.95 \, \mu m$) very close to that of an organic solderability preservative (OSP) finish on commercial PCBs [14]. The polishing was followed by a thorough rinsing with water to remove debris, drying with cheese cloth, and rinsing with acetone. Kapton tape was used to mask the solder joint areas (Figs. 5.1 and 5.2). Steel wires were used to maintain the 0.4 mm solder layer thickness.
Fig. 5.1 Schematic of a) the continuous solder joint DCB specimen and b) the discrete solder joint DCB specimen (l=2 mm and 5 mm). The width of the specimen was 12.6 mm and the overall length was 160 mm. All dimensions in mm. Not to scale.

The masked copper bars were placed on a hot plate maintained at 290°C with the bonding surfaces vertical (Fig. 5.2). When the temperature of the bars reached 220-225°C, a flux-cored SAC305 0.75 mm solder wire (Kester Inc., USA) was touched to the prepared vertical surfaces so that they became rapidly covered with a thin layer of solder. The bars were then clamped together against the 400 μm steel wires, which caused the excess solder and flux residues to flow out of the joint. This procedure minimized voiding due to flux entrapment. Embedded thermocouples were used to control the processing conditions such that the time above liquidus was 120 s and the peak temperature was between 245°C and 250°C. The specimens were then cooled in a small wind tunnel with forced air at a cooling rate of 1.4-1.6°C/s, which is typical of microelectronics manufacturing. A detailed microstructural study of similar DCB specimens confirmed that the intermetallics and bulk solder were similar to those typical of commercial SAC305 solder [14]. After cooling to room temperature, the sides of the specimens were machined to remove the excess solder and create a smooth surface to facilitate the observation of
cracks in the solder layer. Finally, the loading pin holes and clip gauge mounting holes were drilled in the copper bars.

All the discrete joint specimens ($l=2$ mm and 5 mm) contained two solder joints as in Fig. 5.1b except three specimens which contained three discrete joints each ($l=2$ mm). These three specimens had the first joint 45 mm from loading pins and the joints were spaced 30 mm apart. In the discrete joint specimens (Fig. 5.1b), the steel wires were placed just ahead of each solder joint to maintain the solder thickness and to shield each joint from compressive loads as the preceding joint was loaded during the fracture tests. The local geometry of the beginning of each solder joint in both specimens was defined by the smooth, square shape of the edge of the Kapton tape. Earlier work showed that the local edge geometry had a relatively small effect on the quasi-static fracture load at crack initiation [14].

Fig. 5.2 Schematic of specimen arrangement during soldering of continuous and discrete joint DCB specimens.
5.2.2 Fracture Testing

The specimens were tested under mode I and various mixed-mode conditions using the load jig shown in Fig. 5.3 [16]. The load jig enabled the forces on the upper and lower arms of the specimen ($F_1$ and $F_2$) to be independently varied by adjusting the location of the link pins, resulting in a range of mode ratios for a single DCB geometry and actuator. As the load jig is statically determinate, the specimen forces can be calculated from the equilibrium considerations as

$$F_1 = F \left(1 - \frac{s_1}{s_3}\right) \quad (2)$$

$$F_2 = F \frac{s_1}{s_2} \frac{1}{\left(1 + \frac{s_2}{s_4}\right)} \quad (3)$$

where, $s_1$, $s_2$, $s_3$, and $s_4$ are the distances between pin centers (Fig. 5.3), and $F$ is the force applied to the load jig. Note from Eq. (2) that a given load jig configuration (i.e. set of pin locations) results in a constant $F_2/F_1$ ratio that is independent of specimen geometry, crack length, and the applied load $F$. The relation between forces $F_1$, $F_2$ and the loading phase angle, $\psi$, and the energy release rate was given in [16, 17] for a continuous layer DCB as:

$$\psi = \arctan \left[ \frac{\sqrt{3} \left( \frac{F_2}{F_1} + 1 \right)}{2 \left( \frac{F_2}{F_1} - 1 \right)} \right], \quad (4)$$

$$G_c = \frac{12}{E_a h} \left[ f_1^2 \Phi_I^2 + \frac{3}{4} f_2^2 \Phi_{II}^2 \right], \quad (5)$$

where $E_a$ and $h$ are the Young’s modulus and height of the Cu bars, respectively, and $a$ is the arm length as defined in Fig. 5.1. The forces $f_1$ and $f_2$ are mode I and mode II components, respectively, and they are derived from the applied forces $F_1$ and $F_2$ [17]; the constants $\Phi_I$ and $\Phi_{II}$ depend on the geometry and material properties of the specimen [17]. Consequently, for continuous solder layers and discrete joints of sufficient size, the loading phase angle is independent of the crack length, $a$ (Fig. 5.1). However, as will be explained below, the additional compliance of the DCB specimen with the $l=2$ mm solder joints caused the phase angle to vary appreciably with $a$, making Eq. (4) invalid in this case. Therefore, to have a consistent calculation procedure for the three types of DCB specimens used here (continuous,
All the tests were conducted under displacement controlled conditions. The force, $F$, was increased steadily by moving the cross-head (actuator) at a constant rate of 0.1 mm/min. The leading edge of the solder layer (i.e. the location of crack initiation) was monitored with a microscope on a micrometer stage with a field diameter of 1.9 mm. The side of the specimen was painted with a thin layer of diluted paper correction fluid to facilitate the identification of the crack tip. The force, $F$, corresponding to a crack extension of greater than 100 μm in the uncracked joint was defined as the initiation load; although in most DCB specimens, a crack longer than 200 μm formed almost instantaneously at initiation. The number of specimens tested in each specimen configuration and loading condition is shown in the Table 5.1.

The load jig of Fig. 5.3 produced stable crack extension, and many crack growth sequences could be measured with a single DCB after initiation to define the R-curve and the steady-state critical strain energy release rate [14]. However, the present experiments focused on crack initiation in the continuous and discrete joint DCB specimens, since R-curve toughening will be negligible in the small joints typical of most microelectronics applications.

It is noted that a constant cross-head speed applied to the specimens of Fig. 5.1 will produce a solder strain rate that decreases slightly as the distance from the loading pins increases. For example, a finite element analysis of the $l=2$ mm discrete joint DCB specimen loaded under mode-I, showed that a constant cross-head speed of 0.1 mm/min generated von Mises strain rates of $5.6 \times 10^{-5}$ and $1 \times 10^{-5}$ for extreme loading arm lengths of $a=40$ mm to $a=110$ mm, respectively (Fig. 5.1). A change in the strain rate of this magnitude will produce negligible changes in the mechanical properties of solder as indicated by [18, 19]. For example, it was found that an order of magnitude decrease in the strain rate of SAC305 solder, from $1 \times 10^{-4}$ to $1 \times 10^{-5}$, produced only a 6% decrease in the yield stress and the ultimate stress [18]. Moreover, it was observed in [14] that for a similar range of arm lengths and cross-head speeds, the steady-state mode I fracture properties of SAC305 were unchanged. Hence, the effect of changes in the strain rate was assumed to be negligible in the current experiments.
Fig. 5.3 Schematic of the DCB specimen mounted in the mixed-mode load jig [16].

Table 5.1 Number of specimens tested at each phase angle.

<table>
<thead>
<tr>
<th>Specimen type</th>
<th>Mode ratio of loading, $\psi$</th>
<th>Number of specimens tested</th>
<th>Number of joints tested</th>
</tr>
</thead>
<tbody>
<tr>
<td>Continuous joint DCB</td>
<td>0° (mode-I)</td>
<td>9</td>
<td>9</td>
</tr>
<tr>
<td></td>
<td>25°</td>
<td>4</td>
<td>4</td>
</tr>
<tr>
<td></td>
<td>45°</td>
<td>5</td>
<td>5</td>
</tr>
<tr>
<td>Discrete $l=5$ mm joints</td>
<td>25°</td>
<td>3</td>
<td>6</td>
</tr>
<tr>
<td></td>
<td>45°</td>
<td>3</td>
<td>3</td>
</tr>
<tr>
<td>Discrete $l=2$ mm joints</td>
<td>20°-30°</td>
<td>5</td>
<td>11</td>
</tr>
<tr>
<td></td>
<td>35°-49°</td>
<td>4</td>
<td>10</td>
</tr>
</tbody>
</table>
5.2.3 Measurement of Displacement

Opening displacements near the beginning of the solder layer in some mode-I DCBs was measured using two different methods: 1) a clip gage, and 2) digital image correlation (DIC). Only the clip gage method was used with the discrete joint specimens. The clip gage arms were mounted against knife edges secured with screws located as indicated in Figs. 5.1a and 5.1b. The contribution of Cu deformation was insignificant, and it was assumed that the measured opening displacement was entirely from the solder layer.

In the DIC method, the displacement of the solder layer was measured directly using images of the distortion of a black and white speckle pattern deposited on the solder layer as a very diffuse spray paint (Fig. 5.4). The deformation history was recorded during loading using a video camera (progressive scan CMOS sensor with 1280x1024 pixels) with a zoom lens (VMZ1000i, Edmund Optics Inc. NJ, USA). The video images were synchronized with the load cell output so that the captured images could be associated with the corresponding load.

The images were processed using the open-source image correlation software of [20]. All the images starting from zero force to final fracture, captured at 5 s intervals, were analyzed using a 30x30 pixel sampling grid in a region enclosing the leading edge of the solder layer (rectangle in Fig. 5.4) to track the feature shifts between successive images and calculate the strain field evolution. The software can lead to errors if the specimen undergoes rigid body rotations of more than 3° [20]. To eliminate such rotations, the DIC measurements were only done under mode-I conditions without the load-jig arrangement of Fig. 5.3. Furthermore, care was taken to minimize errors that can occur if the lighting conditions or lens focus change during the loading. The measured solder opening displacement corresponding to crack initiation was used to estimate the $J$-integral as in [21-24],

$$ J = d \sigma_y \delta, $$

where $\sigma_y$ and $\delta$ are the yield stress of the solder and the crack tip opening displacement, respectively, and $d$ is a constant that depends on the type of material and loading. These experimentally determined $J$-integral values were used to validate the $J$-integral obtained from the finite element calculations.
5.3 Finite Element Models

Both elastic and elastic-plastic finite element (FE) models were evaluated for the calculation of the fracture parameters corresponding to crack initiation. The elastic model calculated the critical strain energy release rate at fracture using the stress intensity factors at the solder crack tip, while the elastic-plastic model used the J-integral.
5.3.1  Elastic Energy Release Rate and Mode Ratio Calculations

The mesh and the boundary conditions of the DCB FE model are shown in Fig. 5.5, with the crack in the middle of the 0.4 mm thick solder layer with its tip denoted as A. The region near the crack tip was meshed with singular elements (i.e. 6-noded triangular elements formed by collapsing 8-noded PLANE183 elements) available in ANSYS capture the $r^{1/2}$ singularity near the crack tip [25]. The length of the singular elements was maintained at 0.02 mm. Everywhere else, the model was meshed with PLANE183 elements; the solder in plane strain and the copper bars in plane stress. The mesh was graded to ensure that the solder layer contained at least 20 elements in the thickness direction. The solution was a linear elastic static case with the material properties of Table 5.2.

The total strain energy release rate, $G$, was calculated using

$$G = \left( \frac{1-v^2}{E} \right) (K_I^2 + K_{II}^2),$$

(7)

where $v$ and $E$ are the Poisson ratio and tensile modulus of the solder, respectively. The $K_I$ and $K_{II}$ are the mode-I and mode-II stress intensity factors, respectively, obtained directly from the FE model for the given loading and constraint. Figure 5.6 shows a typical mesh for the $l=2$ mm discrete solder joint specimen. A 250 μm long crack was used in both the $l=2$ mm and 5 mm discrete models to simulate the conditions corresponding to experimental fracture initiation.

Table 5.2  Mechanical properties of copper and solder used in FEA [26, 27].

<table>
<thead>
<tr>
<th>Material</th>
<th>Tensile modulus (GPa)</th>
<th>Poisson ratio</th>
<th>Shear modulus (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cu C110</td>
<td>124</td>
<td>0.35</td>
<td>45.9</td>
</tr>
<tr>
<td>Sn3Ag0.5Cu</td>
<td>51</td>
<td>0.4</td>
<td>18.2</td>
</tr>
</tbody>
</table>
Fig. 5.5 Finite element mesh of the continuous joint DCB specimen showing the overall view with boundary conditions and the details near the crack tip, denoted as A.
Fig. 5.6 Finite element mesh of $l=2$ mm discrete joints and the crack tip details.

5.3.2 Elastic-plastic $J$-integral Calculations

The same mesh and material properties were used for the $J$-integral calculations, the only difference being that the solder was modeled as an isotropic elastic perfectly plastic material with a yield stress of 40 MPa [28] and elastic properties as in Table 5.2. As the solution procedure was non-linear static, the applied loads were increased in small increments.

To avoid the effects of the stress singularity at the crack tip, paths along the first two rows of elements nearest the crack tip were excluded, and to assess any crack-path dependence the $J$-integral was evaluated along each of the four different paths shown in Fig. 5.7. As expected, the $J$-integral for these four paths was almost same, and an average of these four values
was considered as the strain energy release rate. This procedure provided only the total energy dissipation associated with a given loading, but not the phase angle, \( \psi \). Hence, the phase angle for a given loading was estimated as explained in the previous section using the elastic FE model.

![Crack in the middle of solder layer and J-integral paths around crack tip](image)

Fig. 5.7 Contours surrounding the solder crack tip used for \( J \)-integral evaluation in the FEA.

### 5.4 Results and Discussions

#### 5.4.1 Crack Initiation in Continuous Joint DCBs

A typical load-displacement response of a DCB specimen under mode-I loading is depicted in Fig. 5.8. The reaction force on the DCB arms continued to increase with the displacement beyond crack initiation until it reached a maximum. This increase of the fracture load beyond crack initiation was due to the evolution of a damage zone near the crack tip which
reached a steady-state size approximately at the maximum load corresponding to the steady-state energy release rate [14]. The load decreased with crack length after attaining steady-state to maintain almost constant energy release rate (note from Eq. 5 that $G$ is directly proportional to both crack length, $a$, and force). Similar R-curve toughening was also observed under the various mixed-mode conditions, and can result in stable crack growth in solder joints as discussed below.

The crack initiation loads of the continuous joint DCBs, under mode-I and the mixed-mode conditions of Table 5.1, were used to calculate the fracture parameters at initiation, $G_{ci}$ and $J_{ci}$, using the elastic and elastic-plastic FE models, respectively. Figure 5.9(a) shows that $G_{ci}$ was almost constant until a phase angle of 25° but increased after that from an average $G_{ci}=480 \text{ J/m}^2$ at $\psi=25^\circ$ to 690 J/m$^2$ at $\psi=45^\circ$. This dependence on the mode ratio is attributed to the larger plastic zone near the crack tip at higher phase angles due to the presence of shear [29]. Similar behavior has been observed in adhesives and in metal/ceramic joints [29]. These $G_{ci}$ values in Fig. 5.9(a) for different mode ratios obtained from the FEA and Eq. (7) matched very well with the $G_{ci}$ calculated analytically using Eq. (5). Figure 5.9(b) shows that for $\psi \leq 25^\circ$, $G_{ci}$ and $J_{ci}$ were both approximately 490 J/m$^2$, but that the predictions of the two models diverged as the mode ratio increased; e.g. at $\psi =45^\circ$, the elastic FE model predicted, $G_{ci}=690 \text{ J/m}^2$, whereas the elastic-plastic FE model resulted in $J_{ci}=825 \text{ J/m}^2$. This deviation can be attributed to the increasing amount of plastic dissipation at higher mode ratios, which was not captured by the crack tip stress intensity factors used in the elastic model.

In addition to the higher phase angle, lower levels of solder constraint can cause larger plastic deformation near the crack tip (e.g. large height-to-diameter ratio in a BGA solder ball). Skipor et al. [30] studied the effect of mechanical constraint on flow and fracture of Sn-Pb solder using a Cu/Sn-Pb/Cu compact tension specimen by varying the thickness of the solder from 28 mm to 0.25 mm. It was observed that the maximum load of the specimen decreased and the ductility of the joint increased with the thickness of the solder layer and the resulting reduced constraint from the Cu members of the joint. In other words, a specimen with a 0.25 mm thick solder layer showed very little macroscopic ductility (i.e. plastic deformation after yield) and failed suddenly as a result of rapid crack growth. However, a 28 mm thick solder layer specimen showed gross plastic deformation. Hence, the elastic-plastic parameter, $J_{ci}$, is more suitable for predicting initiation at higher phase angles and lower constraint levels. Therefore, the average $J_{ci}$
(ψ) (critical strain energy release rate as a function of the mode ratio of loading) from the measured fracture initiation loads for the continuous joint DCBs was used to predict the mixed-mode strength of the discrete joints (l=2 mm and l=5 mm).

Fig. 5.8 Load vs crosshead displacement response of a continuous joint DCB under mode-I loading to crack initiation and beyond. The open circle marks the point of crack initiation at the start of the solder layer.
Fig. 5.9  (a) Initiation strain energy release rate, $G_{ci}$, as a function of the loading phase angle of continuous joint DCBs. The symbols represent data points from individual specimens and the solid curve represents the mean values. (b) Comparison of $G_{ci}$ and $J_{ci}$ from the elastic and elastic-
plastic FE models, respectively. Error bars represent ± standard deviation for the number of specimens shown in Table 5.1.

### 5.4.2 Mixed-mode Fracture of Discrete Joints

Figure 5.10 shows how the force on the load jig, $F$ (Fig. 5.3) varied with the crosshead displacement for a typical 5 mm discrete joint specimen for a loading arm length of 75 mm ($a$ in Fig. 5.1) tested at $\psi=25^\circ$. The initial non-linearity up to approximately 0.1 mm of crosshead movement was due to the clearance in the pins of the jig. Afterwards, the force increased linearly with crosshead displacement, continuing to increase beyond crack initiation due to R-curve toughening. The toughening occurred over a crack propagation distance of approximately 2-3 mm following crack initiation, after which the crack-tip damage zone reached its steady-state size corresponding to the maximum force. Further crosshead displacement led to rapid crack growth and complete joint separation (Fig. 5.10). Similar R-curve toughening was observed in all the discrete $l=5$ mm joints tested at $\psi=25^\circ$ and $45^\circ$, and for all loading arm lengths.

In contrast, the discrete $l=2$ mm joints were too short to show appreciable toughening, and the maximum load was essentially equal to the crack initiation load. Figure 5.11 illustrates this for an $l=2$ mm joint tested at $\psi=28^\circ$ with a loading arm length, $a=40$ mm. This behavior can be understood using a stability analysis for small crack extension at a constant load equal to the measured maximum in Fig. 5.11. Figure 5.12 shows the predicted increase in the applied strain energy release rate with crack length compared with the R-curve toughening that was measured in [14] for the same solder. Although, the toughness in the DCBs of [14] could be slightly different from that in the present discrete $l=2$ mm joints, what is key is that the rate of change of the applied strain energy release rate, $J$, with respect to crack length is much greater than the rate at which the toughness of the joint increases. Therefore, the crack initiating from the edge of the solder joint soon propagates unstably at the maximum force of Fig. 5.11.
Fig. 5.10 Load jig force vs crosshead displacement for an \( l=5 \) mm discrete joint at a loading arm length of \( a=75 \) mm (Fig. 5.1(b)) tested at \( \psi = 25^\circ \).
As was mentioned previously, the applied phase angle became a function of the loading arm length, \(a\) (Fig. 5.1), when the DCBs were made with the \(l=2\) mm joints. The effect of this on \(J_{ci}\) is shown in Fig. 5.13 for four different specimens tested under a load jig configuration which gave a constant \(F_2/F_1=-0.5\) (\(\psi=15^\circ\) in a DCB made with a continuous solder layer). It is evident that the \(J_{ci}\) values at shorter arm lengths were slightly greater than those at longer loading arm lengths, although not significant statistically (95% confidence). This is explained in Fig. 5.14, which shows that in discrete \(l=2\) mm joint specimens at a constant \(F_2/F_1\) ratio, \(\psi\) decreased with increasing loading arm length, causing \(J_{ci}\) to decrease as in Fig. 5.9. The change of phase angle
with loading arm length was also observed in $l=5$ mm joint specimens, but it was less than $1^\circ$ compared to a change of $9^\circ$ from $a=40$ to $110$ mm in the $l=2$ mm joint specimens. In general, this effect will be proportional to the compliance of the DCB arms between the solder joints, so that it will grow as the discrete joints become shorter or the spacing between them increases. As explained earlier, the changes in the applied strain rate due to changes in the loading arm length were negligible in these specimens. Consequently, the small effect observed in Fig. 5.13 was due only to the change of the phase angle with arm length, and the loading arm length had no other independent effect on $J_{ci}$.

Fig. 5.13 The initiation energy release rates, $J_{ci}$, of $l=2$ mm joints as a function of loading arm length ($a$ in Fig. 5.1b) for four different specimens tested with the load jig configuration that gives a constant $F_2/F_1$ ratio of -0.5. Each symbol corresponds to a different specimen.
Fig. 5.14 Variation of phase angle with the loading arm length in discrete $l=2$ mm joint specimens loaded under $F_2/F_1 = -0.5$ using the load jig of Fig. 5.3.

The $J_{ci}$ values for the discrete $l=2$ mm joints are plotted in Fig. 5.15 as a function of the local phase angle at each joint. Each symbol represents $J_{ci}$ for the joints in a different specimen calculated using the maximum load as in Fig. 5.11 (9 specimens in total). As expected, the $J_{ci}$ values increased with phase angle (95% confidence level), consistent with the results for the continuous solder joint DCB specimens of Fig. 5.9. Figure 5.16 shows that a similar statistically significant (95% confidence) trend existed in the fracture data for the discrete $l=5$ mm joints.
Fig. 5.15  The initiation energy release rate, $J_{ci}$, of discrete $l=2$ mm joints as a function of phase angle from nine specimens, each represented by a different symbol. The solid curve shows the average $J_{ci}$ values obtained from continuous joint DCBs as shown in Fig. 5.9(b).

Fig. 5.16  The initiation energy release rate, $J_{ci}$, of discrete $l=5$ mm joints as a function of phase angle compared with the average data from continuous joint DCBs (Fig. 5.9(b)).
5.4.3 Fracture Surfaces and Crack Path

Figure 5.17(a) shows both sides of the DCB fracture surface tested at $\psi=25^\circ$, illustrating the three-dimensional nature of the crack path which simultaneously grew very close to either interface, causing locally smooth fracture surfaces. Note that even at initiation the crack front was extending along both interfaces at lower phase angles ($\psi \leq 25^\circ$). The frequent crack jumps between the two interfaces across the width and along the crack growth direction were caused by the local intermetallic microstructure variations at the interfaces [14]. At higher phase angles ($\psi =45^\circ$), the crack path was more planar due to the asymmetric nature of the loading which tends to drive the crack to one interface rather than causing jumps from one interface to the other (i.e. the path of maximal mode-I loading is along the more highly-strained loading arm [29]). Figure 5.17(b) shows that the fracture surfaces of $l=5$ and $l=2$ mm joints were similar to those of the continuous solder joint of Fig. 5.17(a). However, because there was very little stable, subcritical R-curve crack growth in the $l=2$ mm joints, these cracks tended to propagate very quickly along only one interface, producing a macroscopically smooth, planar surface under all mode ratios, including mode-I [15].
Fig. 5.17  (a) Fracture surfaces of a continuous solder joint DCB tested at $\psi=25^\circ$ illustrating the three-dimensional nature of crack initiation and growth. (b) Comparison of fracture surfaces from continuous solder DCB, $l=5$ mm, and 2 mm joints tested at $\psi=25^\circ$.

5.4.4  **Mixed-mode Fracture Load Predictions for Discrete $l=2$ mm and $l=5$ mm Joints**

The average $J_{ci}(\psi)$ calculated using the FEA and the measured crack initiation loads from the continuous joint DCBs is compared with the values calculated from the initiation loads of the discrete $l=2$ mm and $l=5$ mm joints in Figs. 5.15 and 5.16, respectively. Although the average $J_{ci}$ values from the continuous joint DCBs tended to be smaller than those for the discrete joints, they both followed the same trend with respect to the phase angle. This means that the critical strain energy release rate at crack initiation was not a strong function of the length of the solder joint and that the fracture properties of the continuous joint DCBs can be used to predict the
failure of the smaller joints. Table 5.3 summarizes the data of Figs. 5.15 and 5.16, showing that the mean $J_{ci}$ from the continuous joint DCB provides a lower bound strength prediction for the shorter joints; i.e. the mean $J_{ci}$ of the continuous joint DCBs at various mode ratios was approximately 13% and 20% smaller than the values for $l=2$ mm and $l=5$ mm joints, respectively. One possible reason for these differences is the reduced constraint on the discrete joints compared to continuous joint DCBs. As in [30], reduced constraint would tend to increase the toughness of the smaller joints. Similarly, changes in the substrate stiffness may also influence the level of constraint; however, this is likely to have only a relatively small effect since initiation occurs near the free surface of the solder joint.

Note from Eq. 6 that the energy release rate is in general a quadratic function of the applied loads, and hence the difference between the measured fracture load and the prediction based on the $J_{ci} (\psi)$ failure criterion will be smaller than the corresponding difference in $J_{ci}$ (i.e. 13% for $l=2$ mm and 20% for $l=5$ mm joints). This is illustrated in Table 5.4 which shows that the differences between the measured and predicted fracture loads at different arm lengths and for various phase angles were within 11% and 8% for $l=2$ mm and $l=5$ mm discrete joints, respectively. In the previous study of mode-I loading [14], the fracture loads of 2 mm and 5 mm discrete joints were predicted to within 11% and 12%, respectively, using the mode-I initiation fracture energy from a continuous joint DCB. One source of error in the present calculations is the assumed crack length at fracture initiation, corresponding to the ultimate load of $l=2$ mm joints (since there was negligible R-curve toughening in these short joints). For example, if a crack length of 400 μm is assumed to define the point of maximum strength in the $l=2$ mm joints instead of 250 μm (Fig. 5.6), the predicted a fracture load would be 748 N for an arm length $a=40$ mm at $\psi=25^\circ$; i.e. 6% smaller than that predicted using the 250 μm prediction. Hence, for this case a 60% error in the crack length (400 μm instead of 250 μm) resulted in only a 6% increase in the fracture load prediction error. It remains to be seen whether this energy release rate, $J_{ci} (\psi)$, criterion continues to hold true for joints on the scale of BGA solder balls, which are an order of magnitude smaller than the joints used here.
Table 5.3  Comparison of average $J_{ci}$ from continuous joint DCBs with the average $J_{ci}$ from discrete $l=2$ mm and $l=5$ mm joints under various mode ratios. % difference is relative to the continuous joint DCBs. Data for mode ratios from 20°-27° were averaged.

<table>
<thead>
<tr>
<th>$\psi$ (deg)</th>
<th>Average $J_{ci}$ of DCBs (J/m²)</th>
<th>$l=2$ mm joints</th>
<th>$l=5$ mm joints</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Average $J_{ci}$ % difference</td>
<td>Average $J_{ci}$ % difference</td>
<td>Average $J_{ci}$ % difference</td>
</tr>
<tr>
<td>20°-27°</td>
<td>490</td>
<td>552 11</td>
<td>580 16</td>
</tr>
<tr>
<td>45°</td>
<td>825</td>
<td>948 13</td>
<td>1027 20</td>
</tr>
</tbody>
</table>

Table 5.4  Comparison of (mean ± absolute range) fracture loads for $l=2$ mm and $l=5$ mm discrete joints from experiments with the FEA predictions based on $J_{ci}$ ($\psi$) criterion. A minimum of three specimens were tested in each condition.

<table>
<thead>
<tr>
<th>$l=2$ mm</th>
<th>$\psi$=25°-28°</th>
<th>Load jig force, $F$ (N)</th>
<th>Experiment</th>
<th>Prediction</th>
<th>%difference</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>$a=40$ mm</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>Experiment</td>
<td>838±85</td>
<td>792</td>
<td>-6</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Prediction</td>
<td>839±71</td>
<td>814</td>
<td>-3</td>
</tr>
<tr>
<td></td>
<td></td>
<td>%difference</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>$a=75$ mm</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>Experiment</td>
<td>1254±142</td>
<td>1157</td>
<td>-8</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Prediction</td>
<td>1012±22</td>
<td>1019</td>
<td>1</td>
</tr>
<tr>
<td>$l=5$ mm</td>
<td>$\psi$=25°</td>
<td>Load jig force, $F$ (N)</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>$a=75$ mm</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>Experiment</td>
<td>943±111</td>
<td>921</td>
<td>-2</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Prediction</td>
<td>--</td>
<td>--</td>
<td>--</td>
</tr>
</tbody>
</table>
5.4.5 Validation of $J_{ci}$ with Measured Opening Displacement

The opening displacements measured using the clip gage in two continuous joint DCBs tested under mode-I are plotted as a function of applied force in Fig. 5.18. The displacement increased linearly with the applied force, but became nonlinear as the load approached the crack initiation force. Figure 5.19 shows the average normal strain in the opening direction at the end of the solder layer (i.e. at the location of crack initiation, Fig. 5.4) measured using DIC in two different continuous joint DCBs loaded under mode-I. This average strain was multiplied by the thickness of the solder layer (400 μm) to get the opening displacements of 9 μm and 10.6 μm. These values are close to the values of 9.1 μm, 10.9 μm and 13.7 μm measured from clip gage. The estimated $J$-integral from these measured opening displacements using Eq. 6 are shown in Fig. 5.20 and Table 5.5 for continuous joint DCBs loaded under mode I, assuming the constant $d=1$. It is seen that there was good agreement between the values calculated from the measured opening displacements at crack initiation and the values calculated using the FE model based on the measured loads at initiation. It is noted that the constant $d$ was reported to vary from 1 to 2.9 depending on the characteristics of the solder such as the work hardening exponent, stress state, specimen configuration and loading conditions [21, 22]. However, for a crack in a non-hardening metal foil constrained by stiff elastic substrates, such as the present case of solder constrained by Cu bars, Varias et al. [21] showed that $d$ varies between 0.9 and 1.3 depending on the loading. Hence, the choice of $d=1$ was reasonable.

The total opening displacement in discrete $l=2$ mm and $l=5$ mm joints tested under mixed-mode loading of $\psi=25^\circ$ are shown in Figs. 5.21 (a) and (b), respectively. The response was similar to that seen with continuous joint DCBs except that the force decreased immediately after crack initiation in $l=2$ mm joints, because the solder layer was too short to support the subcritical crack growth and R-curve toughening seen in the $l=5$ mm and continuous joint DCB specimens. The opening displacement values corresponding to crack initiation in the discrete joints are given in Table 5.5. According to [23, 24], Eq. 6 is also valid in mixed-mode fracture, but the constant $d$ should decrease with the mode ratio [24]. However, since $\psi=25^\circ$ is a relatively small mode ratio, $d=1$ was used again to estimate $J_{ci}$ from the measured opening displacements at crack initiation. It can be seen from Table 5.5 that the $J_{ci}$ from FEA (for $l=2$ mm and $l=5$ mm) is only 1.5% different from the $J_{ci}$ estimated from the measured opening
displacement. This provides additional support for the FEA calculations shown in Figs. 5.15 and 5.16, and Table 5.3, and hence to the application of the critical strain energy release rate as a solder joint crack initiation criterion for joints of different length loaded under different mode ratios.

Fig. 5.18 The opening displacement measured by clip gage as a function of applied force in two continuous DCBs tested under mode-I loading.
Fig. 5.19 Normal strain to crack plane measured using DIC plotted against applied force in two continuous DCBs loaded under mode-I.

Table 5.5 Comparison of $J_{c1}$ estimated from the opening displacements measured using the clip gage and $J_{c1}$ obtained from FEA for different joints (mean±standard deviation).

<table>
<thead>
<tr>
<th>Type of joint</th>
<th>Phase angle, $\psi$ (deg)</th>
<th>Total opening displacement $\delta$ (μm)</th>
<th>$J_{c1}$ (J/m²) from Eq. 6 and measured $\delta$ with $d = 1$ and $\sigma_y = 40$ MPa</th>
<th>$J_{c1}$ (J/m²) from FEA (Fig. 5.9 and Table 5.3)</th>
<th>% difference</th>
</tr>
</thead>
<tbody>
<tr>
<td>DCB</td>
<td>0 (mode-I)</td>
<td>11.23±2.31</td>
<td>449±93</td>
<td>480±30</td>
<td>-7</td>
</tr>
<tr>
<td>l=2 mm</td>
<td>25</td>
<td>13.72±4.00</td>
<td>549±160</td>
<td>552±72</td>
<td>-1</td>
</tr>
<tr>
<td>l=5 mm</td>
<td>25</td>
<td>14.73±3.48</td>
<td>589±140</td>
<td>580±121</td>
<td>1.5</td>
</tr>
</tbody>
</table>
Fig. 5.20 Comparison of average $J_{CI}$ from FEA and measured initiation load with values estimated from measured opening displacement for continuous joint DCB loaded in mode I.
5.5 Conclusions

Continuous and discrete SAC305 solder joints of different lengths were made between copper bars under standard surface mount (SMT) processing conditions, and then fractured under mode-I and various mixed-mode loading conditions. The load-displacement behavior corresponding to crack initiation and the subsequent toughening before ultimate failure were recorded.

The loads corresponding to crack initiation in the DCB specimen with a continuous solder joint were used to calculate the critical strain energy release rates, $G_{ci}$ and $J_{ci}$, using elastic and elastic-plastic finite element models, respectively. Both these values were approximately 490 J/m$^2$ for phase angles $\psi < 25^\circ$, but deviated as $\psi$ increased so that at $\psi = 45^\circ$ $G_{ci}=690$ J/m$^2$ and $J_{ci}=825$ J/m$^2$. This deviation at higher mode ratios was attributed to the additional plastic dissipation which was not captured by the elastic finite element model used to calculate $G_{ci}$. 
From the mixed-mode fracture of discrete joints of lengths 2 and 5 mm, it was observed that R-curve toughening could increase the ultimate strength of the joint beyond the value corresponding to crack initiation value at $J_{ci}$. The longer joints ($l=5\,\text{mm}$) experienced some toughening beyond crack initiation similar to that seen with the continuous solder joint DCBs. The shorter discrete joints ($l=2\,\text{mm}$) did not show any toughening and failed soon after crack initiation at $J_{ci}$. The mean $J_{ci}$ of the continuous joint DCBs at various mode ratios was approximately 13\% and 20\% smaller than the values for $l=2\,\text{mm}$ and $l=5\,\text{mm}$ joints, respectively; i.e. the average $J_{ci}$ from the continuous joint DCBs provided a lower bound strength prediction for the shorter joints. This corresponded to differences in the predicted fracture loads of about 11\% and 8\%, respectively for $l=2\,\text{mm}$ and $l=5\,\text{mm}$ joints.

Additionally, the opening displacements at crack initiation in continuous and discrete solder joints were measured using a clip gage and digital image correlation. These were then used to provide an independent estimate of the critical strain energy release rate at fracture. It was found that these critical $J$-integral values agreed well with those calculated using the FEA and the measured fracture loads.

It is concluded that the critical strain energy release rate as a function of the mode ratio of loading provides a useful fracture criterion for solder joints of intermediate lengths subject to arbitrary combinations of tension and shear. The next step will be to assess whether the criterion is applicable to much shorter solder joints typically found in microelectronic packages such as ball grid arrays.

5.6 References


[18] Che FX, Poh EC, Zhu WH, Xiong BS. Ag content effect on mechanical properties of Sn-xAg-0.5Cu solders. 9th Electron Packaging Technol Conf, 2007, p.713-18.


[28] Jenq ST, Chang HH, Lai YS, Tsai TY. High strain rate compression behavior for Sn-37Pb eutectic alloys, lead-free Sn-1Ag-0.5Cu and Sn-3Ag-0.5Cu alloys. Microelectron Reliability 2009; 49: 310-17.


Chapter 6

6 Prediction of Pad Cratering Fracture at the Copper Pad – Printed Circuit Board Interface

6.1 Introduction

Epoxy-based printed circuit board (PCB) laminates that are compatible with higher lead-free reflow temperatures can be more brittle than earlier materials, raising new reliability issues. Among these is the increased propensity for PCB surface epoxy cracking beneath the copper pads of solder joints, also known as pad cratering [1]. This mode of failure has been observed widely in both low and high strain rate mechanical loading conditions, such as the quasi-static bending of PCBs and board-level drop tests [2, 3].

At present, there are no widely accepted standards for fracture testing of PCBs. Pad-crater cracking is generally assessed qualitatively using the pin pull test [4, 5] in which a pin is soldered to the Cu pad on the PCB surface and pulled to measure the breaking force. Although the data from this test can be useful for assessing the relative quality of the pad-epoxy bond strength in a specific board, it does not provide fundamental fracture data such as the critical strain energy release rate. Therefore, the measured strength is not useful for predicting the pad-crater failure load in other boards with different pad sizes and under different loading. As well, the loading in this test is predominantly tensile which induces only mode-I fracture and may not represent the actual loading in service where a combination of tensile and shear loads generates a mixed-mode loading. In general, cracks constrained by tough interfaces experience mixed-mode loading conditions, and the fracture strength will depend on ratio ($\psi$) of the mode I (crack opening mode) and mode II (shear mode) loading components [6]. This is the case in pad-cratering where cracks propagate in the surface epoxy layer between a copper pad and the epoxy-glass fiber composite. The strain energy release rate as a function of the mode ratio, $J_c(\psi)$, has been used successfully in many studies to predict the mixed-mode fracture strength of epoxy adhesive joints [7-10]. A similar approach was adopted here to treat pad-crater fracture in PCBs.
Under certain conditions the weak link in the solder connection is in the solder itself or in the intermetallic compound layer at the copper interface, rather than in the surface epoxy layer between the copper pad and the epoxy-glass fiber composite of the PCB. In this case, fracture can also be analyzed in terms of a critical strain energy release rate as a function of the mode ratio of loading, \( J_c(\psi) \) \[11-13\].

In this chapter, pad-crater failure was characterized in terms of the critical strain energy release rate, \( J_{ci} \), measured at various mode ratios, \( \psi \). Fracture specimens were prepared from a commercial PCB rated for lead-free assembly and assembled with “chip array thin core ball grid array” (CTBGA) packages. The specimens were fractured at low and high loading rates in various bending configurations to generate a range of mode ratios. \( J_{ci} \) for pad cratering was calculated from the measured fracture strength and specimen deformation using a linear elastic finite element analysis (FEA). The predictive capability of the approach was then demonstrated by measuring the fracture loads of single lap-shear specimens made from these same PCB-CTBGA assemblies and comparing with the strength predictions made using the FEA and the \( J_c(\psi) \) failure criterion.

### 6.2 Specimen Preparation and Fracture Testing

The fracture test specimens were manufactured from assemblies containing a CTBGA 228 I/O package mounted on a printed circuit board (Practical Components Inc., USA) as shown in Fig. 6.1. The package (12x12 mm) contained 228 Sn3.0Ag0.5Cu (SAC305 alloy) solder balls arranged in a perimeter array with 0.5 mm pitch that matched the pattern of the copper pads on the PCB (Fig. 6.1). The components contained solder mask defined (SMD) pads while the PCB contained non-solder mask defined (NSMD) pads, and the surface finish on both these pads was organic solderability preservative (OSP or copper finish). The diameter of the solder balls after assembly was 300 µm and the height was 200 µm. The CTBGA components contained a silicon die (10x10x0.27 mm) attached to 0.2 mm thick substrate, encapsulated in a mold compound, and the total height of the mounted package, including the solder height, was 1 mm from PCB surface. The PCB was 1.5 mm thick and was made of a high-performance FR-4 epoxy laminate.
(phenolic-cured, non-filled, IS410 from Isola Inc.), with a glass transition temperature, $T_g$, of 180°, compatible with higher lead-free reflow temperatures.

The components and boards were assembled using a standard reflow oven with nine heating zones. Two different time-temperature reflow profiles were used for the assembly process: one with a time above solder liquidus temperature (TAL) of 60 s and the other with a TAL of 120 s. In both these cases, the peak temperature and cooling rates were maintained at 245-250°C and approximately 2°C/s, respectively. Some boards with a gold finish (electroless nickel and immersion gold) were also assembled with the TAL60 profile.

![Fig. 6.1](image)

Fig. 6.1 The printed circuit board and CTBGA assembly is shown along with the basic specimen prepared by cutting the middle of the assembly such that 3 rows of solder balls were isolated at each end.
The specimens were cut from the package-PCB assembly as shown in Fig. 6.1 using a Struers Accutom-2 precision cut-off machine with a diamond blade. These specimens contained a 6x3 array of solder joints at either end with six joints across the 3 mm width direction. The specimens were then loaded in four different ways as shown in Fig. 6.2, thereby generating a range of loading phase angles, $\psi$, from $7^\circ$ to $33^\circ$. The phase angle is a measure of the relative amounts of mode I and mode II loading applied to the solder joint, and can be defined in terms of the stress intensity factors as,

$$\psi = \arctan\left(\frac{K_{II}}{K_I}\right)$$  \hspace{1cm} (1)

where $K_I$ and $K_{II}$ are the mode-I and mode-II stress intensity factors. The double cantilever beam and single lap-shear specimens shown in Figs. 6.2b and 6.2d, respectively, were prepared by gluing a PCB arm to the package using a cyanoacrylate adhesive (Instant Adhesive, Loctite, USA). The specimens were loaded at a cross-head speed of 1.52 mm/min in all tests. The applied force was measured using a 222 N load cell and the displacements of the 3-point bend specimens in Figs. 6.2a and 6.2c during loading were measured with a resolution of 0.2 µm using a laser displacement sensor (LK-G82, Keyence Corp.).

The appearance of cracks in the test specimens was recorded using a video camera with a zoom lens (VZM1000i, Edmund Optics Inc., USA). The load corresponding to the creation of a crack of approximately 50 µm length was defined as the initiation fracture load in the 3-point bend specimens of Figs. 6.2a and 6.2c. In the other specimens (Figs. 6.2b and 6.2d), the maximum recorded load was considered to be the crack initiation load at the critical solder joint, since the entire package failed soon after crack initiation with no significant increase in load due to crack toughening.

The critical strain energy release rate of initiation, $J_{ci}$, and the phase angle of loading, $\psi$, were calculated using the critical load and the critical displacement in the finite element models explained in Section 6.3. The testing configurations depicted in Figs. 6.2a-6.2c were intended to characterize the fracture behavior of the pad-PCB interfacial region as a function of the phase
angle; i.e. the fracture envelope, $J_{cA}(\psi)$ characteristic of the epoxy layer within which the cracks grew.

The geometry of Fig. 6.2d was included as a test case of the ability to predict the pad-PCB fracture strength in other loading configurations using the failure criterion $J = J_{cA}(\psi)$; i.e. joint fracture was predicted when the applied strain energy release rate, $J$, was equal to the measured critical value at the specific phase angle created by the loading and specimen. The applied strain energy release rate, $J$, is a function of the applied loads and the stiffness of the PCB-component-solder ball assembly, while the phase angle, $\psi$ is mainly a function of the applied loads and the stiffness of the materials on either side of the crack [6].

(a) $\psi = 33^\circ, N = 9$
Fig. 6.2 Schematic of specimen of Fig. 6.1 tested in four configurations: (a) 3-point bending, (b) double cantilever beam, (c) reversed 3-point bending, and (d) single lap shear. The critical location where cracks initiated is indicated along with the corresponding mode ratio, $\psi$, of the...
critical solder joint. \( N \) is the number of specimens tested. Figures not to scale; dimensions in mm.

### 6.2.1 Fracture Testing at Higher Loading Rates

It was of interest to determine qualitatively how the fracture load and crack path changed at higher rates of loading. It was suspected that the mode of failure might change as the solder joint itself became relatively weaker with increasing strain rate. Such behavior has been observed in [14-16]. Figure 6.3 depicts the schematic of a drop test that was applied to another configuration of the basic specimen of Fig. 6.1. The component was glued to a rigid support, and the solder joints were loaded by attaching a 109.62 g mass to the PCB arm with nylon thread and dropping it from a height of 20 mm. The displacement response was measured at a sampling rate of 1 kHz using the laser displacement sensor.

![Schematic of drop test](image)

**Fig. 6.3** Schematic of drop test. Figure not to scale; all dimensions in mm.
6.3 Finite Element Models: $J$ and $\psi$ Calculation

As will be discussed in the following section, the specimens shown in Fig. 6.2 all failed by epoxy cracking between the copper pad and the reinforcing glass fibers in the PCB. This was simulated using ANSYS 12® finite element (FE) software in order to calculate the critical strain energy release rate and phase angle at the moment of fracture using the measured fracture load. For example, Fig. 6.4 shows the finite element mesh of the CTBGA package and PCB assembly under 3-point bending (Fig. 6.2a). Based on the experimental observations, the load corresponding to the onset of fracture in the specimen was assumed to correspond to a 50 $\mu$m crack. Nevertheless, the calculation of $J_{ci}$ was not a strong function of crack length in these specimens, and changed only 1.5% for a 50% change in the assumed crack length. Based on crack path measurements in failed joints described in the following section, the crack was modeled in the epoxy matrix at a depth of 20 $\mu$m from the PCB surface as shown in Fig. 6.5. As with the assumed crack length the strain energy release rate was quite insensitive to this depth and changed only 2% for a 75% change in the depth of the assumed crack path. Similar cracks were modeled at the observed critical locations for the other loading configurations as indicated in Fig. 6.2.

The region near the crack tip in all the FE models was meshed with singular elements (i.e. 6-noded triangular elements formed by collapsing 8-noded PLANE183 elements) available in ANSYS [17] to capture the $r^{1/2}$ singularity near the crack tip. The length of the singular elements was 0.001 mm. Everywhere else the models were meshed with PLANE 183 8-node structural elements in plane strain. The element size was varied smoothly from a size of 0.003 mm near the crack to 0.1 mm in the areas far from crack tip. It was established that $J_{ci}$ was not sensitive to the mesh size at this level.

The constitutive behavior of the specimen materials was defined as isotropic linear elastic with the properties of Table 6.1, provided by the supplier of the boards and components (Practical Components Inc., USA). Figure 6.6 shows that the FE model with these properties produced a load-displacement response in 3-point bending that compared reasonably well with
the experimentally measured curve. A similar comparison was performed for the bare PCB in bending, and the predictions agreed very well with the experimental observations.

The FE models omitted certain detailed features such as the adhesive layer that bonded the PCB arm to the package (Figs. 6.2b and 6.2d), and the layer joining the Si die to the package substrate (Fig. 6.4), because their contribution to the overall stiffness of the specimen was negligible and did not affect the $J_{ci}$ calculations. Most of the Cu pads on both the component side and the PCB side were also not modeled for the same reason. However, the Cu pad near the critical location was modeled (see Cu pad in Figs. 6.4 and 6.5) as it could affect the phase angle of loading, $\psi$. The residual stresses due to the manufacturing reflow process were also not modeled in the FE simulations.

Table 6.1  Mechanical properties used in FE models.

<table>
<thead>
<tr>
<th>Material</th>
<th>Tensile modulus (GPa)</th>
<th>Poisson ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>PCB (Isola 410)</td>
<td>22</td>
<td>0.2</td>
</tr>
<tr>
<td>SAC305 solder alloy</td>
<td>51</td>
<td>0.4</td>
</tr>
<tr>
<td>Substrate</td>
<td>14.5</td>
<td>0.11</td>
</tr>
<tr>
<td>Die (silicon)</td>
<td>130</td>
<td>0.278</td>
</tr>
<tr>
<td>Mold compound (G770)</td>
<td>16.7</td>
<td>0.25</td>
</tr>
<tr>
<td>Copper</td>
<td>124</td>
<td>0.34</td>
</tr>
</tbody>
</table>
The total strain energy release rate at the crack tip, $J$, was calculated using

$$ J = \left( \frac{1-v^2}{E} \right) (K_I^2 + K_{II}^2) $$ (2)

where $v$ and $E$ are the Poisson ratio and tensile modulus of the PCB material. $K_I$ and $K_{II}$ are the mode-I and mode-II stress intensity factors, respectively, obtained directly from the FE models, calculated from the crack flank displacements for a given loading and constraint. The phase angle, $\psi$, was calculated using these stress intensity factors in Eq. (1). This energy release rate calculation was verified by evaluating the $J$-integral along three different paths around the crack tip as shown in Fig. 6.5. As expected from the path independence of the $J$-integral, the values obtained from these three paths were almost identical and matched the values obtained from Eq. 2 to within 1%.

Fig. 6.4 Finite element mesh of CTBGA package under 3-point bending (Fig. 6.2a), and details of the FE model near the critical solder joints compared with a photograph of the specimen edge. Note that copper pad was modeled only near the critical location in the FE model.
Fig. 6.5 FE mesh of solder joint and near the crack tip. Right figure shows the J-integral contours surrounding the crack tip within the epoxy surface layer of the PCB.

Fig. 6.6 Verification of specimen mechanical properties by comparing experimental measurements of 3-point bending with FE predictions.
6.4 Results and Discussion

6.4.1 Mechanical Response and Failure Mode Under Quasi-static Loading

The load-displacement response of a CTBGA package tested under 3-point bending as in Fig. 6.2a is shown in Fig. 6.7. The failure process of the specimen is depicted in Fig. 6.8 at different load levels, with the corresponding locations indicated in Fig. 6.7 as points (a) and (b). At 3.6 N/mm (load per unit specimen width) the first solder joint started failing by pad-cratering. The second joint began to break at 4.4 N/mm, while the third solder joint failed at a load level close to point (b) in Fig. 6.7. The failure of the solder joints did not produce any noticeable effect on the load-displacement curve because the contribution from these joints to the overall stiffness of the specimen (Fig. 6.2a) was far smaller than that from PCB. Hence, the failure load was identified visually. Consequently, the load-deflection response was essentially linear until approximately 6 mm of deflection, after which the onset of damage in the PCB material caused the response to become nonlinear. The load-displacement response and the failure behavior of the other 3-point bend specimen (Fig. 6.2c) were similar.

Figure 6.9 shows that the response of the DCB specimen (Fig. 6.2b) was also linear until the onset of cracking at the first solder joint, which happened very near the maximum force when the specimen failed completely. The initial non-linearity in the curve until approximately 2 mm was due to the initial slack in the wire used to load the DCB arm. The loads corresponding to the onset of cracking in the specimens, for example, point (a) in Fig. 6.7 and the maximum load in Fig. 6.9 were used in FE models of Section 6.3 to calculate $J_{ci}$ and the phase angle.
Fig. 6.7 Load-displacement behavior of CTBGA package-PCB assembly in 3-point bending.
Point (a) 3.6 N/mm, failure of first joint by pad cratering, and (b) 9.75 N/mm, complete failure of PCB.

Fig. 6.8 Bending of PCB-CTBGA assembly loaded in 3-point bending as in Fig. 6.2a at: zero load (left image), 3.6 N/mm, failure of first joint (middle image), and 9.75 N/mm, complete failure of PCB (right image). The middle and right images correspond to points (a) and (b) on the load-displacement curve of Fig. 6.7.
Figure 6.10 shows the failure mode of the solder joint interconnects of the CTBGA package loaded in 3-point bending under quasi-static loading (Fig. 6.2a). As mentioned previously, the crack path was in the epoxy surface layer of the PCB between the copper pad and the reinforcing glass fibers. This was the crack path seen in all of the quasi-static tests of the specimens of Fig. 6.2, including both TAL 60 s and 120 s joints. Therefore, under quasi-static conditions (low to moderate loading rates; von Mises strain rates in the solder from $5 \times 10^{-5}$ s$^{-1}$ to $3 \times 10^{-3}$ s$^{-1}$) the epoxy surface layer of the PCB was weaker than the solder alloy. This is consistent with recent studies on lead-free compatible assemblies which revealed similar failure modes in board-level bend tests and board-level drop tests with PCBs formulated to withstand higher reflow temperatures [2,3,18,19].

The schematic of Fig. 6.10b illustrates the crack path typically found in the specimens of Fig. 6.2. Cracks initiated near the edge of the Cu pad-epoxy interface and penetrated into the PCB until they reached the reinforcing glass fibers where they turned to run parallel to the fibers. As a result, the thickness of the cracked epoxy layer of each joint in Fig. 6.10a was slightly different, being dependent on the depth of the nearest glass fibers.
Fig. 6.10 (a) Low magnification optical photograph demonstrating the failure mode of the CTBGA package of Fig. 6.2a after quasi-static testing. (b) Schematic of crack path illustrating the crack path within the epoxy surface layer of the PCB. Woven bundles of glass fibers are shown oriented in the plane of the sketch and normal to it.
The fracture surfaces were scanned with an optical profilometer to measure the thickness of the cracked epoxy layer or the depth of crater. Figures 6.11 and 6.12 show surface profile scans of the PCB before assembly and after fracture, respectively. The Fig. 6.11 shows that the thickness of the Cu pad and the solder mask (the dark green surface layer on the PCB) were 30 µm and 40 µm, respectively. Therefore, anything deeper than 40 µm from the PCB surface (top of solder mask) represented the fractured epoxy layer. In other words, the values below zero in Fig. 6.12 represent the cracked epoxy that is pulled from the PCB surface (also shown in Fig. 6.12) which created approximately a 20 µm crater in this case. Similarly, several fracture surfaces were scanned and the average thickness of the cracked epoxy layer was calculated to be approximately 22 µm. Hence, a value of 20 µm was used as cracked epoxy thickness (Fig. 6.5) in all the FE models.
Fig. 6.11 Profilometer scan of PCB with Cu pads before reflow showing the height of Cu pad and solder mask from PCB epoxy surface.
Fig. 6.12 Fracture surfaces on PCB and CTBGA after quasi-static 3-point bend test (Fig. 6.2a). The profilometer scan of the fracture surface on the PCB side shows the crater depth under the copper pad as values below zero.
6.4.2 Mixed-mode Fracture Characterization of Epoxy Surface Layer Under Quasi-static Loading

Figure 6.13 shows the critical energy release rate, $J_{ci}$, for pad cratering (PCB epoxy layer fracture) as a function of the phase angle, $\psi$. The average $J_{ci}$ value at $\psi = 7^\circ$ was approximately 87 J/m$^2$ and increased slightly with phase angle to a value of 127 J/m$^2$ at $\psi = 26^\circ$, after which it increased sharply to reach an average $J_{ci} = 330$ J/m$^2$ at $\psi = 33^\circ$. The scatter in the $\psi = 33^\circ$ data was greater because the greater relative amount of shear loading decreased the crack opening, making it more difficult to detect the onset of cracking. This is similar to the mixed-mode fracture behavior observed in brittle and toughened epoxy adhesive joints [7-10, 20]. $J_{ci}$ is smallest at low phase angles because this is predominantly mode I or a cleavage-type loading and is similar to that found in the pin pull test [4, 5]. In comparison with this epoxy fracture, the SAC305 solder behaves similarly with increasing phase angle, but has a mode-I $J_{ci} = 480$ J/m$^2$ [13]. Thus the unreinforced epoxy surface layer on the PCB was significantly weaker in fracture than the solder.

The increased toughness with phase angle (Fig. 6.13) is attributed to the mixed-mode stress state at the crack tip. As mentioned previously, such mixed-mode crack propagation is typical of situations where cracks are deflected by tougher interfaces, such as in adhesive joints and at the epoxy-glass fiber boundary in pad cratering. Under these constrained mixed-mode conditions, the damage or fracture process zone ahead of the crack grows with the phase angle, thereby increasing the critical strain energy release rate [9].

The data of Fig. 6.13 were measured on specimens having an OSP finish. Identical results were obtained on 3-point bend specimens (Fig. 6.2a) having an ENIG finish. This was expected since the crack propagated in the epoxy and not in the solder. Similarly, the data of Fig. 6.13 were also independent of the solder reflow profile (TAL 60s and 120 s), because this only affected the solder fracture and not the PCB surface epoxy. Therefore, the fracture strength governing this type of pad cratering is independent of the pad finish, the solder and the soldering conditions.
6.4.3 Effect of Strain Rate on the Failure Mode of Solder Interconnects

Figure 6.14 shows the displacement response of the PCB arm with respect to time in four repetitions of the drop test (Fig. 6.3). After approximately 60 ms the displacement increased rapidly as the mass impacted the loading arm (Fig. 6.3). In these higher strain rate tests (von Mises strain rates of approximately 0.2 to 1 s\(^{-1}\) in the solder), the joints failed by crack propagation on the board side within the solder intermetallic layer adjacent to the copper pads (Fig. 6.15). The change in the failure mode with loading rate was believed to be due to the strain-rate sensitivity of the solder and epoxy. It is known that the yield and fracture strength of solder and epoxy increases with loading rate [14,21-23]. However, the intermetallic layer at the Cu/solder interface is brittle and its fracture strength is insensitive to the strain rate; hence, as the loading rate increased, the solder and epoxy tended to become stronger than the intermetallic layer making it the weak link for crack propagation. Similar observations have been reported for
a flip-chip mounted PCB with ENIG finish that was subjected to four-point bending [3]; i.e. the failure mode changed from epoxy cracking to intermetallic layer fracture as the loading rate increased. Therefore, the failure mode of an electronic package due to mechanical loading depends not only on the materials and their relative strengths, but also on the loading rate (or strain rate).

Fig. 6.14 Response of CTBGA package during drop weight test.
6.4.4 Fracture Load Prediction Using $J_{ci}(\psi)$

If the pad cratering fracture strength $J_{ci}$ is known as a function of the phase angle of loading as in Fig. 6.13, it is possible to predict the load at which pad cratering will occur in other types of loading and in other components made with the same PCB epoxy. This was demonstrated with the present PCB using a lap shear specimen (Fig. 6.2d) that was loaded in tension. The measured failure loads from five different lap-shear specimens are given in the second column of Table 6.2. The load-displacement response of these single lap-shear specimens was similar to that of the DCB specimen of Fig. 6.2b; i.e., the maximum recorded force corresponded closely to the onset of fracture in the outermost joint.

The FE models of Section 6.3 were then used with the average $J_{ci}(\psi)$ fracture envelope of Fig. 6.13 to predict the fracture loads for pad cratering. The phase angle of loading in this specimen was $\psi = 32^\circ$, so that the corresponding average $J_{ci} = 330$ J/m$^2$ (Fig. 6.13). The predicted fracture loads were then calculated assuming, as before, a crack length of 50 $\mu$m and a
crack depth of 20 µm. Table 6.2 shows that the predicted pad cratering fracture loads were within 17% of the measured values, being consistently smaller than the actual values. One possible reason for this conservative prediction could be the method used to measure the failure load of the lap-shear specimens; i.e. the maximum load was taken to be the failure load, but video observation showed that this could be slightly higher than the actual crack initiation load.

Table 6.2 Comparison of measured pad cratering fracture loads for lap-shear specimens with FE predictions based on an average $J_{ci} = 330 \text{ J/m}^2$ from the fracture envelope of Fig. 6.13. Average difference was -10%.

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Experimental failure load (N)</th>
<th>Predicted (FE) failure load (N)</th>
<th>% Difference</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>72</td>
<td>66</td>
<td>-9</td>
</tr>
<tr>
<td>2</td>
<td>66</td>
<td>61</td>
<td>-7</td>
</tr>
<tr>
<td>3</td>
<td>51</td>
<td>50</td>
<td>-2</td>
</tr>
<tr>
<td>4</td>
<td>65</td>
<td>57</td>
<td>-13</td>
</tr>
<tr>
<td>5</td>
<td>67</td>
<td>56</td>
<td>-17</td>
</tr>
</tbody>
</table>

6.5 Conclusions

Pad-crater fracture loads were measured in lead-free PCB-CTBGA assemblies tested at low and high loading rates in various bending configurations to generate different ratios of tensile and shear loads (mode ratios). The intrinsic fracture strength of the crack path region between the pad and the PCB was then characterized in terms of the critical strain energy release
rate, $J_{ci}$, calculated from the measured fracture loads and displacements at the different mode ratios.

It was observed that all of the specimens tested at low strain rates failed by pad cratering, but the specimens tested at a higher loading rate failed by crack propagation within the solder intermetallic layer adjacent to the copper pad.

The critical energy release rate, $J_{ci}$, for pad cratering (PCB epoxy layer fracture) depended strongly on the mode ratio or phase angle, $\psi$. The average $J_{ci}$ value at $\psi = 7^\circ$ was approximately 25% of the value at $\psi = 33^\circ$, indicating the pad-cratering fracture loads are a strong function of the ratio of tensile and shear stresses at the tip of cracks growing between a pad and the PCB. This in turn means that the onset of pad cratering will depend on the directions of the loads applied to the PCB-component assembly as well on the stiffness of the PCB and the component.

Because the crack path for pad cratering was within the epoxy layer on the surface of the PCB, the critical energy release rate, $J_{ci}$, was confirmed to be independent of pad finish (OSP and ENIG) and the solder reflow profile (TAL 60s and 120 s). Finally, the generality of the $J = J_{ci}(\psi)$ failure criterion to predict pad cratering fracture was demonstrated by predicting the fracture loads of single lap-shear specimens made from the same lead-free assemblies.

### 6.6 References


Chapter 7

7 Conclusions and Future Work

7.1 Conclusions

A fracture mechanics approach was followed to characterize and predict the failure loads of solder joints. Two different modes of solder joint failure were studied: cracking within solder and pad crater fracture beneath the Cu pad of solder joints. It was demonstrated by experiments and complimentary modeling techniques that the critical energy release rate as a function of the mode ratio of loading or $J = J_c(\psi)$ is a fracture criterion that can be used to predict the strength of solder joints of arbitrary geometry subject to combined tensile and shear loads. The approach works regardless of the crack path within the solder-pad-circuit board assembly. The detailed observations of this work are presented below.

7.1.1 Fracture Behavior of Lead-free Solder Joints

The fracture behavior of a Cu/Sn3Ag0.5Cu solder joint system was studied under mode I and mixed-mode conditions using DCB specimens with different solder layer thicknesses and Cu beam heights; manufactured using standard industrial processing conditions and three time-above-liquidus (TAL) values. The microstructures of the specimens were found to be similar to those seen in commercial solder joints, and all joints exhibited a qualitatively similar R-curve behavior, but with different values of toughness.

1. The phase angle had little effect on the critical strain energy release rate, $G_c$, between $\psi=0^\circ$ (mode I) and $25^\circ$, but caused a 35% increase at $\psi=45^\circ$ in both the initiation value, $G_{ci}$, and the steady-state value $G_{cs}$. This behavior was analogous to that seen in adhesive joints.

2. The solder joint toughness decreased as the time-above-liquidus increased. This was attributed to the intermetallic compound (IMC) layer thickness at the joint interface, which increased from an average value of 3 μm at TAL 60 s to 5 μm at TAL 240 s.
The thinner IMC layer produced a tougher, more ductile fracture, while the thicker IMC layer caused brittle Cu6Sn5 cleavage at a lower critical strain energy release rate. These differences were reflected in the roughness of the fracture surfaces and in the degree of crack bridging behind the macro-crack tip. The other R-curve parameters, such as rising length and slope were not affected significantly by the phase angle, but were affected by TAL.

3. The initiation strain energy release rate, which governs the failure of small joints such as BGAs, was largely independent of the geometry of the solder fillet at the free end of the solder layer.

4. The effect of substrate stiffness (or Cu beam height) on the initiation strain energy release rate, $G_{ci}$, was negligible, which implies that the $G_{ci}$ measured from DCB specimens can be used to predict the fracture loads of joints of much smaller joints such as BGAs.

5. Similarly, the variation of the $G_{ci}$ with solder thickness was statistically insignificant over the range $t=200 - 400 \mu m$ for the three phase angles $\psi=0^\circ$, $25^\circ$, and $45^\circ$, respectively. In contrast, the steady-state critical strain energy release rate, $G_{cs}$, did increase significantly with the solder thickness.

6. These observations imply that the effect of solder layer thickness can be neglected when predicting the strength of relatively short solder joints using $G_{ci}$. However, longer joints may support sufficient subcritical crack growth to realize appreciable toughening which will increase with the solder layer thickness.

7. The crack path was found to be influenced by the mode ratio of loading and its effect on the stress state at the crack tip. The crack path was highly three-dimensional for phase angles below $\psi=25^\circ$ and predominantly planar for higher phase angles $\psi=45^\circ$. The crack paths followed the contour of maximum von Mises strain rather than the maximum principal stress.

8. Consistent with earlier work on solder joint strength, some preliminary results indicated that the loading rate had a significant effect on $G_{ci}$. 
7.1.2 Fracture Load Predictions

7.1.2.1 Mode-I Fracture of Discrete \( l = 2 \text{ mm and 5 mm} \) Solder Joints

Discrete SAC305 solder joints of different lengths \( (l = 2 \text{ mm and } l = 5 \text{ mm}) \) were made between copper bars under standard surface mount (SMT) processing conditions, and then fractured under mode-I loading. The load-displacement behavior corresponding to crack initiation and the subsequent toughening before ultimate failure were recorded.

The load corresponding to crack initiation in the DCB specimen with a continuous solder joint was identified using two different methods: one based on visual inspection and the other based on the onset of nonlinearity in load-displacement behavior. Based on these values, strain energy release rates at crack initiation, \( G_{ci} \), were calculated.

1. It was found that the \( G_{ci} \) based on the onset of nonlinearity was, 380 J/m\(^2\), while that from the visual method was 480 J/m\(^2\). The larger value corresponding to the appearance of a crack at the edge of the joint was attributed to the state of plane stress at that location compared with the state of plane strain existing over most of the crack front.

2. From the discrete joint experiments, it was observed that R-curve toughening could increase the joint ultimate strength beyond the crack initiation value at \( G_{ci} \). The longer joints \((l = 5 \text{ mm})\) experienced some toughening beyond crack initiation similar to that seen with the continuous solder joint DCB.

3. The smaller discrete joints \((l = 2 \text{ mm})\) did not show any toughening and failed as soon as a crack initiated at \( G_{ci} \).

The fracture of these discrete solder joints was simulated using finite elements with two different failure criteria: one in terms of the critical strain energy release rate at initiation, \( G_{cis} \), and another based on a cohesive zone model at the crack tip (CZM). The parameters of the CZM for the case of mode-I loading were obtained from an iterative procedure using the measured crack initiation load and load-displacement curve of the continuous solder joint DCB.

4. Both models predicted the fracture loads reasonably well (i.e., to within 12\% accuracy).
5. In addition, the CZM was able to predict accurately the overall load-displacement behavior of the discrete joint specimens and predict the load sharing that occurred between neighboring solder joints as a function of joint pitch and adherend stiffness.

This has application in the modeling of the strength of solder joint arrays such as those found in ball grid array packages. These observations imply that the $G_{ci}$ obtained from a continuous solder joint DCB can be used to predict the ultimate strength of short joints with lengths less than 2 mm.

### 7.1.2.2 Mixed-mode Fracture of Discrete $l=2$ mm and $l=5$ mm Solder Joints

Discrete SAC305 solder joints of different lengths were fractured under mode-I and various mixed-mode loading conditions. The load-displacement behavior corresponding to crack initiation and the subsequent toughening before ultimate failure were recorded.

The loads corresponding to crack initiation in the DCB specimen with a continuous solder joint were used to calculate the critical strain energy release rates, $G_{ci}$ and $J_{ci}$, using elastic and elastic-plastic finite element models, respectively.

1. Both these values ($G_{ci}$ and $J_{ci}$) were approximately 490 J/m² for phase angles $\psi < 25^\circ$, but deviated as $\psi$ increased so that at $\psi = 45^\circ$ $G_{ci}=690$ J/m² and $J_{ci}=825$ J/m².
2. This deviation at higher mode ratios was attributed to the additional plastic dissipation which was not captured by the elastic finite element model used to calculate $G_{ci}$.
3. From the mixed-mode fracture of discrete joints of lengths 2 and 5 mm, it was observed that R-curve toughening could increase the ultimate strength of the joint beyond the value corresponding to crack initiation value at $J_{ci}$. The longer joints ($l=5$ mm) experienced some toughening beyond crack initiation similar to that seen with the continuous solder joint DCBs. The shorter discrete joints ($l=2$ mm) did not show any toughening and failed soon after crack initiation at $J_{ci}$.
4. The mean $J_{ci}$ of the continuous joint DCBs at various mode ratios was approximately 13% and 20% smaller than the values for $l=2$ mm and $l=5$ mm joints, respectively; i.e. the average $J_{ci}$ from the continuous joint DCBs provided a lower bound strength prediction.
for the shorter joints. This corresponded to differences in the predicted fracture loads of about 11\% and 8\%, respectively for \(l=2\) mm and \(l=5\) mm joints.

Additionally, the opening displacements at crack initiation in continuous and discrete solder joints were measured using a clip gage and digital image correlation. These were then used to provide an independent estimate of the critical strain energy release rate at fracture. It was found that these critical \(J\)-integral values agreed well with those calculated using the FEA and the measured fracture loads.

It is concluded that the critical strain energy release rate as a function of the mode ratio of loading provides a useful fracture criterion for solder joints of intermediate lengths subject to arbitrary combinations of tension and shear.

7.1.2.3 Pad-crater Fracture in Chip-Scale-PCB Assembly

Pad-crater fracture loads were measured in lead-free PCB-CTBGA assemblies tested at low and high loading rates in various bending configurations to generate different ratios of tensile and shear loads (mode ratios). The intrinsic fracture strength of the crack path region between the pad and the PCB was then characterized in terms of the critical strain energy release rate, \(J_{ci}\), calculated from the measured fracture loads and displacements at the different mode ratios.

It was observed that all of the specimens tested at low strain rates failed by pad cratering, but the specimens tested at a higher loading rate failed by crack propagation within the solder intermetallic layer adjacent to the copper pad.

The critical energy release rate, \(J_{ci}\), for pad cratering (PCB epoxy layer fracture) depended strongly on the mode ratio or phase angle, \(\psi\). The average \(J_{ci}\) value at \(\psi = 7^\circ\) was approximately 25\% of the value at \(\psi = 33^\circ\), indicating the pad-cratering fracture loads are a strong function of the ratio of tensile and shear stresses at the tip of cracks growing between a pad and the PCB. This in turn means that the onset of pad cratering will depend on the directions of the loads applied to the PCB-component assembly as well on the stiffness of the PCB and the component.
Because the crack path for pad cratering was within the epoxy layer on the surface of the PCB, the critical energy release rate, $J_{ci}$, was confirmed to be independent of pad finish (OSP and ENIG) and the solder reflow profile (TAL 60s and 120 s). Finally, the generality of the $J = J_{ci}(\psi)$ failure criterion to predict pad cratering fracture was demonstrated by predicting the fracture loads of single lap-shear specimens made from the same lead-free assemblies.

### 7.2 Future Work

1. One of the main contributions of this thesis is a solder joint fracture criterion $J = J_{ci}(\psi)$. However, the focus of the current study was fracture under low strain rate conditions or quasi-static conditions. This methodology can be extended to fracture prediction of solder joints under impact conditions by measuring $J_{ci}(\psi)$ properties under higher loading rates. Although a preliminary attempt was made in Chapter 3 to quantify the effect of strain rate on the solder fracture properties, it was not comprehensive. Hence, the experimental techniques presented here can be extended to measure fracture properties under impact conditions.

2. The microstructure of solder joints is often not stable and evolves with time under temperature cycles during service. The experimental methods established in the second chapter can be used to quantify the effect of aging on the fracture behavior of solder joints. Along with the proposed fracture criterion, these aging studies would enable the prediction of joint strength as a function of its aging.

3. The cohesive zone model presented in Chapter 3 captured the mechanics of joint failure very well and can be very useful for fracture modeling in electronic packages. However, the present model considered only mode-I fracture and did not include the R-curve toughening behavior. Future work can extend this model to treat mixed-mode fracture predictions and can incorporate the toughening behavior into this model.

4. The FE models used in the study did not consider the residual stresses that could exist in the solder joints due to the joining processes. A future study can investigate how these residual stresses affect the failure predictions.
5. The discrete joint specimens designed in Chapter 4 can be used to understand the effect of joint pitch on the fracture properties of array type electronic packages. Although cohesive zone model was used here to study the effect of pitch, they were not compared with experimental data.

6. The pad crater failure studied in Chapter 5 was characterized using specimens prepared from a commercial package which is a more practical approach and has an immediate use in the electronics industry. It would be interesting to see if the properties measured from a bulk epoxy specimen matches with these properties. The high strain rate fracture characterization can be simpler with bulk epoxy specimens or with the DCB specimens developed in Chapter 2 by replacing the solder layer with epoxy.