Composite Interlaminar Shear Fracture Toughness,
$G_{\text{ilc}}$:
Shear Measurement or Sheer Myth?

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Abstract

The concept of $G_{\text{ilc}}$ as a measure of the interlaminar shear fracture toughness of a composite material is critically examined. In particular, it is argued that the apparent $G_{\text{ilc}}$ as typically measured is inconsistent with the original definition of shear fracture. It is shown that interlaminar shear failure actually consists of tension failures in the resin rich layers between plies followed by the coalescence of ligaments created by these failures and not the sliding of two planes relative to one another that is assumed in fracture mechanics theory. Several strain energy release rate solutions are reviewed for delamination in composite laminates and structural components where failures have been experimentally documented. Failures typically occur at a location where the mode I component accounts for at least one half of the total $G$ at failure. Hence, it is the mode I and mixed-mode interlaminar fracture toughness data that will be most useful in predicting delamination failure in composite components in service. Although apparent $G_{\text{ilc}}$ measurements may prove useful for completeness of generating mixed-mode criteria, the accuracy of these measurements may have very little influence on the prediction of mixed-mode failures in most structural components.
Introduction

One of the most common failure modes for composite structures is delamination. The remote loadings applied to composite components typically get resolved into interlaminar tension and shear stresses at discontinuities that create mixed mode I & II delaminations. Over the past 10-20 years, it has become accepted practice to characterize the onset and growth of these mixed-mode delaminations using fracture mechanics. The strain energy release rate, G, and the mode I component due to interlaminar tension, $G_I$, and mode II component due to interlaminar shear, $G_{II}$, are calculated using the virtual crack closure technique\[1\]. In order to predict delamination onset or growth, these calculated G components are compared to interlaminar fracture toughness properties measured over a range from pure mode I loading to pure mode II loading.

Examples of mixed-mode delamination criteria for carbon fiber reinforced composites with either a brittle epoxy (AS4/3501-6) or a toughened epoxy matrix (IM7/E7T1-2), are shown in figures 1a and 1b, respectively \[2,3\]. The critical G for delamination is plotted as a function of the mode II percentage compared to the total G. Hence, at $G_{II}/G = 0$, the loading at the delamination front is a pure opening mode I, whereas at $G_{II}/G = 1$, the loading at the delamination front is a pure shear mode II. As figure 1 indicates, the apparent mode II toughness for a graphite epoxy material is typically much greater than the mode I toughness and has significantly more variability or scatter. In this paper, some of the reasons for the differences in magnitude and scatter between $G_{IC}$ and $G_{IIC}$ will be examined. The significance of these differences on the prediction of mixed-mode delamination in composite structural configurations will be discussed.

Background

In the 1950's, Irwin proposed a general theory of fracture \[4,5\], based on the method of Westergaard \[6\], that postulated the existence of three unique fracture modes that could occur at the tip of a crack (fig. 2). These fracture modes included: (1) an opening mode I, where the crack faces underwent opening displacements relative to one another as the crack grew, (2) an in-plane sliding
shear mode II, where the crack faces slid over one another in the direction of the crack growth, and (3) an out-of-plane scissoring (or tearing) mode III where the crack faces slid relative to one another in a direction normal to the direction of crack growth. The elasticity solution for stress intensity factors associated with these three postulated fracture modes \( (K_{Ia}, K_{IIa}, K_{III}) \) were derived yielding a mathematically complete and consistent theory for fracture of materials and structures. Strain energy release rates may be related to these stress intensity factors squared through coefficients consisting solely of material properties. Solutions for cracked bodies with specific configurations and loadings were developed and applied to structural problems [7]. However, most of these problems consisted of cracks in homogeneous materials (typically structural metallic materials) where cracks that may originally have all three fracture modes when loaded would typically turn immediately as the crack grew to assume a pure opening mode I orientation. Hence, the resistance of these materials to fracture could be adequately described in terms of the opening mode fracture toughness, \( K_{IC} \), alone.

With the advent of adhesively bonded structures and laminated composite materials in the 1960's and 1970's, the problem of debonding of adhesive bonds, and delamination in composite materials, created a class of problems where cracks were constrained to grow in bond lines, or resin rich regions between composite plies, such that macroscopically the cracks could not assume a pure opening mode I orientation. Therefore, for this class of materials, the mixed-mode fracture problem that was resolved mathematically in the 1950's posed a challenge in terms of fracture toughness characterization.

The opening mode I characterization proved relatively straightforward with the advent of the Double Cantilever Beam (DCB) test configuration, although complexities involving the influences of bondline and insert thicknesses, precracking techniques, and fiber bridging in composites delayed the standardization of this test method until the 1990's [8,9]. Development of test methods for characterizing the interlaminar shear fracture toughness, however, has proven to be a difficult task, both in terms of achieving an adequate configuration to yield a pure shear loading at the crack tip, and in the interpretation of the test results [10]. It is the latter issue that will be the focus of this paper. As Shakespeare might
have said, at issue is whether \( G_{\text{IIc}} \) is to be, or not to be, considered a generic property of the composite material. Since most of the data generated to date has been for the sliding shear mode II fracture, the discussion will be limited to the measurement and interpretation of \( G_{\text{IIc}} \).

**Mode II Fracture Toughness measurement results**

In the 1980's several test methods were proposed for measuring \( G_{\text{IIc}} \). However, to-date none of these have been standardized. The most popular methods are the End-notched Flexure (ENF) and End-loaded split beam (ELS) shown in figure 3. The ENF test involves a simple three point bend loading, but it results in an unstable delamination growth unless the initial crack is very long [11] or the test is controlled with a special shear displacement gage [12]. The ELS test involves a more complicated clamped boundary condition but results in a stable delamination growth [13]. Both of these test configurations have been analyzed and have been demonstrated to yield a pure sliding shear fracture mode II at the delamination front [13,14].

Several interlaboratory "round robin" test programs have been conducted using both of these test methods [3,10]. However, interpretation of the test results has proven to be difficult to resolve. This difficulty may be illustrated by examining some typical results found in the literature [2,3,15-34]. For this review, papers were chosen that compared the influence of precracking versus testing from the embedded insert on \( G_{\text{IIc}} \) values measured on the same specimen. In addition, papers were chosen that reported the ratio of precracked, or insert, \( G_{\text{IIc}} \) values to \( G_{\text{c}} \) values measured only from the insert to avoid the complication due to fiber bridging. These data are summarized in Table 1.

**Mode II Precracking Effects**

One difficulty in measuring the interlaminar shear fracture toughness is the apparent inconsistency between \( G_{\text{IIc}} \) values measured by growing the crack from a thin midplane insert versus \( G_{\text{IIc}} \) values measured by growing the crack from an initial shear precrack. For \( G_{\text{c}} \) values measured using the DCB test, a single generic
toughness value may be obtained as the insert thickness is decreased [24], thereby achieving a composite material property where the insert successfully simulates an initial delamination crack without having to resort to precracking which induces fiber bridging. However, \( G_{ic} \) values decrease with insert thickness, but never reach a single value that may be considered a generic property of the composite material [24]. Furthermore, \( G_{ic} \) values measured from the insert are sometimes greater, and sometimes less than, \( G_{ic} \) values measured from a shear precrack. Figure 4 compares results from studies where \( G_{ic} \) values were determined from both the insert and from precracking on the same specimen. In most cases the precrack value is lower than the insert value. However, for two materials (S2/SP250 glass epoxy and IM7/F3900 graphite epoxy) the reverse was true.

\( G_{ic} \) Comparison to \( G_{ic} \)

One result that is consistent in the literature for composite materials over a large range of toughnesses is that \( G_{ic} \) always exceeds \( G_{ic} \). Figure 5 shows a plot of the ratio of \( G_{ic} \) to \( G_{ic} \) for composite materials with a large range of opening mode interlaminar fracture toughnesses. There is a general trend that the more brittle materials (lower \( G_{ic} \) values) have \( G_{ic} \) values that are much greater than the corresponding \( G_{ic} \), whereas the tougher matrix materials (higher \( G_{ic} \) values) have \( G_{ic} \) values that are close to, but still greater than, the corresponding \( G_{ic} \).

Some other interesting trends are noted when the total data set in figure 5 is separated into two plots, one for relatively brittle thermoset (epoxy) matrix composites (figure 6) and the rest for relatively tough thermoplastic matrix composites, which for this literature search consisted solely of AS4/PEEK (figure 7). The \( G_{ic}/G_{ic} \) ratios for the epoxy matrix composite materials all exceeded 2 to 1, which gives some credence to the mechanistic fracture theory of reference 35 (described in a later section) as an explanation of a lower bound for \( G_{ic} \) values. However, for all but one of the results in the literature for the tough thermoplastic composite (AS4/PEEK), the \( G_{ic}/G_{ic} \) ratios range between 2 to 1 and 1 to 1. Hence, the micromechanisms at the delamination front may be quite different for these two classes of materials.
Mode II Fracture Mechanisms

Figure 8 compares scanning electron microscope (SEM) photographs of mode I and mode II fracture surfaces for a brittle T300/5208 graphite epoxy composite [22]. For the brittle epoxy material, the mode I fracture surface appears to have a fairly clean cleavage plane, whereas the mode II fracture surface exhibits a very rough fracture plane with the characteristic “hackles” observed in mode II delamination [16,20,22,33,36]. The sketch in figure 8 under the mode II SEM image illustrates the principal tension stress at a 45 degree angle to the delamination plane that results from mode II loading and is responsible for creating tension microcracks in the resin rich region between plies. Once these cracks appear, the ligaments formed by them are forced to bend until they fracture and coalesce, creating the perceived extension of the original delamination via mode II. Figure 9 shows an edge view generated by in-situ testing in a scanning electron microscope [36] that illustrates the formation and coalescence of microcracks forming the final interlaminar shear fracture surface.

Figure 10 compares SEM photographs of mode I and mode II fracture surfaces for a tough AS4/PEEK graphite thermoplastic composite. In contrast to the brittle composite, for the tough thermoplastic matrix composite both the mode I and II fracture surfaces are similar, with extensive evidence of matrix yielding at the delamination front.

In both the brittle and tough matrix composite materials, the complex failure process under mode II loading is far removed from the idealized sliding of two crack planes relative to one another as postulated in the fracture mechanics elasticity solutions. Hence, a mechanistic explanation of mode II fracture is needed.

Mode II Fracture Theory

In reference 35, a fracture mechanics based failure criteria was postulated based on the observation that for brittle homogeneous materials subjected to mixed-mode loading, “fracture occurs when the total mode I component which is experienced is equal to a critical value, \(G_0\).” This idea was expressed mathematically as
\[ G_I + \sin^2 \omega G_{II} = G_0 \]  

(1)

where \( \omega \) corresponds to "the slope of the surface roughness." A worse case was postulated for omega of 45 degrees corresponding to the "shear microcracks that form ahead of the main crack" in a composite interface that is delaminating. The authors noted that if \( G_{II} = G_{IIc} \) when \( G_I = 0 \), then

\[ \sin^2 \omega = G_{II}/G_{IIc} \]  

(2)

Equation 2 implies that \( G_{IIc} \) may be estimated based on a measured \( G_{IC} \) and omega, which if assumed to be the worst case of 45 degrees, would yield a maximum value of \( G_{IIc} \) equal to twice \( G_{IC} \). This is consistent with the initial damage mechanism postulated in the previous section, i.e. the formation of tension microcracks ahead of the delamination front. However, it does not account for the bending and fracture of the ligaments formed by the microcracks that releases much of the energy required to create the perceived macroscopic mode II delamination growth. Indeed, the authors cited an epoxy matrix composite example where \( G_{IIc} = 2.22 \ G_{IC} \). Further examples from the literature, as summarized in figure 5, also yield \( G_{IIc} \) values much greater than twice the corresponding \( G_{IC} \). Hence, the basic premise set forth in reference 35 was sound, i.e., that the actual failure mechanism is due to tension, but a complete description of failure was missing. Therefore, equation (2) at best represents only a lower bound on \( G_{IIc} \).

The theory postulated in reference 35 implies that the macroscopic shear fracture corresponds to the growth of the tension microcrack as an extension of the initial delamination due to the mode I component of the mixed mode loading at the delamination front. For composite delamination under shear loading, however, a more realistic sequence of failure may be as follows: (1) a tension microcrack initiates ahead of the original delamination front wherever the weakest flaw exists in the resin rich layer ahead of the crack and forms at a 45 degree angle to the original delamination, (2) further microcracks accumulate in front of the initial microcrack, with a spacing defined by classical shear lag considerations and the distribution of inherent flaws in the ply.
interface, and (3) these microcracks finally coalesce to form an extension of the delamination.

For the initial tension microcracking, a transverse tension strength criteria, using Weibull statistics to characterize the flaw sensitivity [37], may prove useful. This may also help to account for the increased scatter in mode II toughness values compared to mode I values as shown in figure 1. In fact, it may be argued that the transverse tension strength and mode I interlaminar fracture toughness are the only true generic material properties that control the failure of a brittle material.

Beyond the initiation phase, however, a complex model of ligament bending and fracture for brittle materials, and ligament bending, yielding, and rupture for ductile materials, would be needed to adequately characterize the actual failure mechanism at the delamination front under mode II loading. Such a model was recently proposed to account for this sequence of events [38]. The model was based on an adhesive bond containing a crack analogous to the composite delamination in the resin rich layer between plies. A shear lag model was utilized to predict a regular spacing of microcracks ahead of the delamination front that depended on the opening mode fracture toughness of the resin, the shear modulus and yield strength of the resin, and the thickness of the resin rich layers between the plies. A quantitative relationship was derived relating these constituent properties to the apparent composite $G_{IC}$ by utilizing a curve fitting parameter, $m$, determined by plotting the effective shear modulus of the resin layer with multiple matrix cracks as a function of the crack spacing normalized by the resin layer thickness.

Although this model provides some useful insights into the relative influence of constituent properties on the apparent composite $G_{IC}$, the predicted toughness is very sensitive to the curve fitting parameter chosen. Even if a model could be developed without introducing a new curve fitting parameter, it would be difficult to demonstrate its value as a predictive tool because of the extreme variability of apparent $G_{IC}$ values relative to corresponding $G_{IC}$ values (figure 5). In addition, requiring a local characterization of the material, i.e., measuring transverse tension strength, resin fracture toughness, resin moduli, resin tension and yield strengths, etc. needed for modeling crack tip micromechanisms discreetly will be extremely labor intensive compared to simply measuring the
apparent $G_{\|c}$. Furthermore, experience with mixed-mode delamination failures that occur in structural configurations under realistic loadings typically result in problems where the mode I component is predominant, as illustrated in the next section.

Relevance to Mixed-mode Delamination Onset Prediction

Since the mid 1970’s, several strain energy release rate solutions have been developed for delaminations in flat uniform thickness composite laminates and in more realistic structural configurations (such as tapered, curved, and stringer reinforced laminates) where failures have been experimentally documented. These studies all seem to indicate that it is the $G_i$ component that controls the onset of delamination much more than the $G_{\|}$ component.

Initial studies on edge delamination of flat laminates with straight edges indicated that the onset of delamination depended strongly on the stacking sequence. Stacking sequences having large $G_i$ components associated with delamination prone interfaces yielded lower failure strains at delamination onset [39]. Failure occurred when the mode I component reached $G_{\|c}$ for the composite regardless of the $G_{\|}$ component. Flat laminates with open holes, modeled as a series of laminates with varying layups using a rotated straight edge concept, yielded similar results where the $G_i$ component appeared to control the onset of delamination around the hole boundary [40].

Recently, a number of studies have begun to appear in the literature for a variety of more typical structural configurations. These include studies on tapered laminates containing internal ply drops to vary thickness [41-43], curved laminates [44,45], and stringer reinforced laminates subjected to out-of-plane “pull-off” loading [46,47]. The material, configuration, and mixed-mode ratio ($G_{\|}/G$) for these studies is summarized in Table 2. In the majority of these cases, the failure occurs at a location where the $G_i$ component accounts for at least one half of the total $G$ at failure, i.e, where $G_{\|}/G$ is less than 50%. Indeed, for the tapered and curved laminates studied, $G_{\|}/G$ never exceeded 10%.

In addition to these models of complex loadings on structural components, the problem of delamination resulting from low velocity impact has been shown to consist of a sequence of events beginning with a matrix crack induced by transverse shear (principle...
tension stresses at 45 degrees to the laminate axis) followed by a mixed mode I & II delamination [48]. Hence, the typical assumption that \( G_{IIC} \) is the most important parameter for characterizing low velocity impact damage [38] is questionable.

These structural configuration case studies indicate that the mode I and mixed-mode interlaminar fracture toughness data will be most useful in predicting delamination failure in composite components in service. Even though the apparent \( G_{IIc} \) values will typically be measured for completeness of generating the mixed-mode criteria illustrated in figure 1, the accuracy of these measurements may have very little influence on the prediction of mixed-mode failures in most structural components.

**Summary**

The concept of \( G_{IIc} \) as a measure of the interlaminar shear fracture toughness of a composite material was critically examined. An extensive literature review was conducted to identify studies where Mode I and Mode II fracture toughness measurements were compared. This review indicated that for composite materials over a large range of toughnesses, \( G_{IIc} \) always exceeds \( G_{ic} \). In addition, the more brittle materials had \( G_{IIc} \) values that were often much greater than the corresponding \( G_{ic} \), whereas the tougher matrix materials had \( G_{IIc} \) values that are close to, but still greater than, the corresponding \( G_{ic} \). However, it was noted that \( G_{IIc} \) values measured from the insert are sometimes greater, and sometimes less than, the \( G_{IIc} \) values measured from a shear precrack. Furthermore, examination of the micromechanisms at the tip of the delamination front documented in the literature for a wide range of composite materials, indicated that interlaminar shear failure actually consists of tension failures, followed by the coalescence of ligaments created by these failures, and not the sliding of two planes relative to one another that is assumed in fracture mechanics theory. Hence, the apparent \( G_{IIc} \) as typically measured is inconsistent with the original definition of shear fracture.

For the initial tension microcracking, a transverse tension strength criteria, using Weibull statistics to characterize the flaw sensitivity, may help to account for the increased scatter in mode II toughness values compared to mode I. Beyond the initiation phase, however, a complex model of ligament bending and fracture for
brittle materials, and ligament bending, yielding, and rupture for ductile materials, would be needed to adequately characterize the actual failure mechanism at the delamination front under mode II loading. Even if this could be achieved, it would be of questionable value because of the extreme variability of apparent $G_{IIc}$ values relative to corresponding $G_{Ic}$ values. In addition, requiring a local characterization of the material needed for modeling crack tip micromechanisms discreetly will be extremely labor intensive compared to simply measuring the apparent $G_{IIc}$.

Several strain energy release rate solutions have been developed for both composite laminates and structural components where failures have been experimentally documented. In the majority of these cases, the failure occurs at a location where the mode I component accounts for at least one half of the total $G$ at failure. Hence, it is the mode I and mixed-mode interlaminar fracture toughness data that will be most useful in predicting delamination failure in composite components in service. Although apparent $G_{IIc}$ measurements may prove useful for completeness of generating mixed-mode criteria, the accuracy of these measurements may have very little influence on the prediction of mixed-mode failures in most structural components. Therefore, as Shakespeare might have said, the controversy over mode II fracture toughness measurement may turn out to be much ado about nothing.
References


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<td>0 and 45 degree plies</td>
<td>0-5</td>
</tr>
<tr>
<td>42</td>
<td>Tapered Laminate</td>
<td>IM6/18271 Graphite-Epoxy</td>
<td>0 and 45 degree plies</td>
<td>0-5</td>
</tr>
<tr>
<td>43</td>
<td>Tapered Laminate</td>
<td>S2/SP250 Glass-Epoxy</td>
<td>0 degree plies</td>
<td>0-10</td>
</tr>
<tr>
<td>43</td>
<td>Tapered Laminate</td>
<td>S2/CE9000 Glass-Epoxy</td>
<td>0 degree plies</td>
<td>0-10</td>
</tr>
<tr>
<td>43</td>
<td>Tapered Laminate</td>
<td>IM6/18271 Graphite-Epoxy</td>
<td>0 degree plies</td>
<td>0-5</td>
</tr>
<tr>
<td>44</td>
<td>Curved Laminate</td>
<td>AS4/3501-6 Graphite-Epoxy</td>
<td>0 degree plies</td>
<td>5-10</td>
</tr>
<tr>
<td>45</td>
<td>Curved Laminate</td>
<td>AS4/3501-6 Graphite-Epoxy</td>
<td>0 and 90 degree plies</td>
<td>5-10</td>
</tr>
<tr>
<td>46</td>
<td>Stringer Reinforced Laminate</td>
<td>IM6/3501-6 Graphite-Epoxy</td>
<td>0, 45, and 90 degree plies</td>
<td>13-40</td>
</tr>
<tr>
<td>47</td>
<td>Stringer Reinforced Laminate</td>
<td>IM7/E7T1-2 Graphite-Epoxy</td>
<td>45 degree plies</td>
<td>28-55</td>
</tr>
</tbody>
</table>
Fig. 1a - Mixed Mode Delamination Criterion for AS4/3501-6
Fig. 1b - Mixed Mode Delamination Criterion for IM7/E7T1-2
Mode I
Crack opening or tensile mode

Mode II
In-plane shearing mode

Mode III
Out-of-plane shearing mode

Fig.2 - Fracture Modes
Fig. 3 - Mode II Test Configurations
Fig. 4 - Ratio of Precrack $G_{\text{llc}}$ to Insert $G_{\text{llc}}$ for Polymer Matrix Composites
Fig. 5 - Ratio of Mode II to Mode I Toughness for Polymer Matrix Composites
Graphite and Glass Epoxy Composites

Figure 6 - Ratio of Mode II to Mode I Toughness for

\[ G^c, \text{ J/m}^2 \]

\[ G^c / G^c \]

Pre-crack \( G^c \)

Insert \( G^c \)

\( G^c = 2G^c \)
Fig. 7 - Ratio of Mode II to Mode I Toughness for AS4/PEEK Composites

\[ G_{IIc} / G_{Ic} \]

\[ G_{IIc} = 2G_{Ic} \]

\[ G_{IIc} \]

\[ \text{Precrack } G_{IIc} \]

\[ G_{IIc} = G_{Ic} \]
Fig. 8 SEM photographs of delamination fracture surfaces in T300/5208 Graphite Epoxy laminates
Formation of tension microcracks

Coalescence of microcracks forming hackles

Fig. 9 Formation and Coalescence of Tension Microcracks
in AS4/PEEK laminates

FIG. 10  SEM photograpahs of delamination fracture surfaces

Mode II

Mode I
The Concept of GIIc as a measure of the interlaminar shear fracture toughness of a composite material is critically examined. In particular, it is argued that the apparent GIIc as typically measured is inconsistent with the original definition of shear failure. It is shown that interlaminar shear failure actually consists of tension failures in the resin rich layers between plies followed by the coalescence of ligaments created by these failures and not the sliding of two planes relative to one another that is assumed in fracture mechanics theory. Several strain energy release rate solutions are reviewed for delamination in composite laminates and structural components where failures have been experimentally documented. Failures typically occur at a location where the mode I component accounts for at least one half of the total G at failure. Hence, it is the mode I and mixed-mode interlaminar fracture toughness data that will be most useful in predicting delamination failure in composite components in service. Although apparent GIIc measurements may prove useful for completeness of generating mixed-mode criteria, the accuracy of these measurements may have very little influence on the prediction of mixed-mode failures in most structural components.