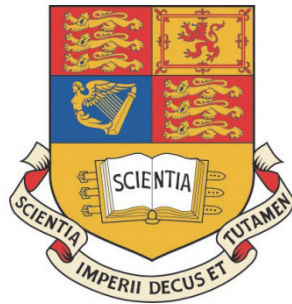


Development of translaminar fracture toughness testing methods for composite materials

by

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Abstract

The work presented within this thesis concerns measurement of the fracture toughness associated with the translaminar, fibre-breaking, failure modes of composites. Loading cases of mode I tension, mode I compression and mixed mode I/II tension and shear are considered.

Fracture toughness measurement for translaminar tensile failure is investigated using the compact tension specimen. A detailed analysis of data reduction schemes concludes that a modified compliance calibration technique is the most appropriate, in terms of reproducibility of results and simplicity. Investigation of specimen in-plane size, thickness and lay-up effects indicates a thickness dependence. Specimen fracture surfaces reveal that the increase in measured toughness for specimens with thicker 0° plies is due to an increased amount of fibre pull-out. Furthermore, it is found that true measurements of fracture toughness are obtained from specimens with initial notch radii less than $250\text{ }\mu\text{m}$, for the IM7/8552 material system. Fractography reveals that the critical notch root radius is dictated by the 0° plies alone.

A four-point bend specimen is used to measure the fracture toughness associated with translaminar compressive failure. This method enables the failure mode to be triggered in isolation, making it superior to other methods in the literature. Microscopy of failed specimens reveals a shear driven fibre failure at 45° to the initial notch; the cause of this failure mode is investigated using a micromechanical finite element model.

Mixed-mode translaminar tensile/shear fracture is investigated using a modified compact tension configuration and a specially developed fixture. SEM of the fracture surfaces of failed specimens reveals damage mechanisms specifically caused by the introduction of a mode II component of loading which accompany a significant increase in the damage zone size. R-curves are generated which suggest that the specimen configuration used here is appropriate for characterising fracture toughness at low proportions of mode II.

Declaration

The work presented within this thesis is the author's own, unless referenced to the contrary in the text, and is the product of research carried out at Imperial College London. No part of this thesis has been submitted elsewhere for any other degree or qualification.

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Notation

Lower case Roman letters

a	crack length
a_{eff}	effective crack length
Δa	change in crack length
c_i	$i = 0, 1, 2, 3$ constants obtained from polynomial curve fit
d	displacement
f_p	peak frequency of acoustic emission hit
g_I	local mode I strain energy release rate
g_{II}	local mode II strain energy release rate
h	height of compact tension specimen
k_I	local mode I stress intensity factor
k_{II}	local mode II stress intensity factor
l	length of compact tension specimen
t	specimen thickness
t_0	thickness of 0° plies
t_{90}	thickness of 90° plies
w	specimen width as defined in relevant figure

NOTATION

Upper case Roman letters

C	compliance
E_{11}	Young's modulus in material 1 direction
E_{22}	Young's modulus in material 2 direction
G_{12}	shear modulus
G	strain energy release rate
$\bar{G}$	normalised strain energy release rate
G_c	critical strain energy release rate
G_{Ic}	mode I critical strain energy release rate
G_{Ic}^{lam}	mode I critical strain energy release rate of a laminate
G_{Ic}^0	mode I critical strain energy release rate of 0° plies
G_{Ic}^{90}	mode I critical strain energy release rate of 90° plies
G_{IIc}	mode II critical strain energy release rate
G_{IIc}^{90}	mode II critical strain energy release rate of 90° plies
J	J-integral
K	stress intensity factor
K_{Ic}	mode I critical stress intensity factor
K_{Ic}^0	mode I critical stress intensity factor of 0° plies
P	load
P_c	critical load

NOTATION

Lower case Greek letters

α constant derived from best fit of compliance–crack length data

β constant derived from best fit of compliance–crack length data

χ constant derived from best fit of compliance–crack length data

η factor relating laminate and 0° ply level stresses/fracture toughness

θ loading angle of mixed mode fixture

ν_{12} major Poisson’s ratio

ρ notch tip radius

ρ_c critical notch tip radius

σ stress

σ_c critical stress

σ_c^0 critical stress in 0° plies

Upper case Greek letters

Δa change in crack length

Δd_n displacement of ECT specimen

Chapter 1

Introduction

When a structure contains a flaw, or some pre-existing damage, it is necessary to be able to predict the conditions under which a crack will begin to propagate and subsequently lead to failure of the structure. Fracture mechanics testing addresses this problem by providing information in terms of critical stress intensity factors, which define the local state close to the crack tip at fracture, and critical strain energy release rates, which give a global description of the energy dissipated during the fracture process.

Today, continuous fibre reinforced composite materials are used extensively in the vast array of engineering situations where high strength and stiffness to weight ratios are demanded. With the advantageous properties offered by these material systems comes the associated penalty of increased design complexity and the fact that the failure mechanisms they exhibit are still not yet fully characterised and understood.

The laminated nature of these material systems, produced by stacking layers of reinforcement (either pre-impregnated with resin or impregnated during the forming process), means that the fracture characteristics can be divided into three areas: interlaminar, intralaminar and translaminar failure. Delamination has been under extensive investigation for several years, and the need to characterise

this behaviour has led to the development of several standardised testing procedures.

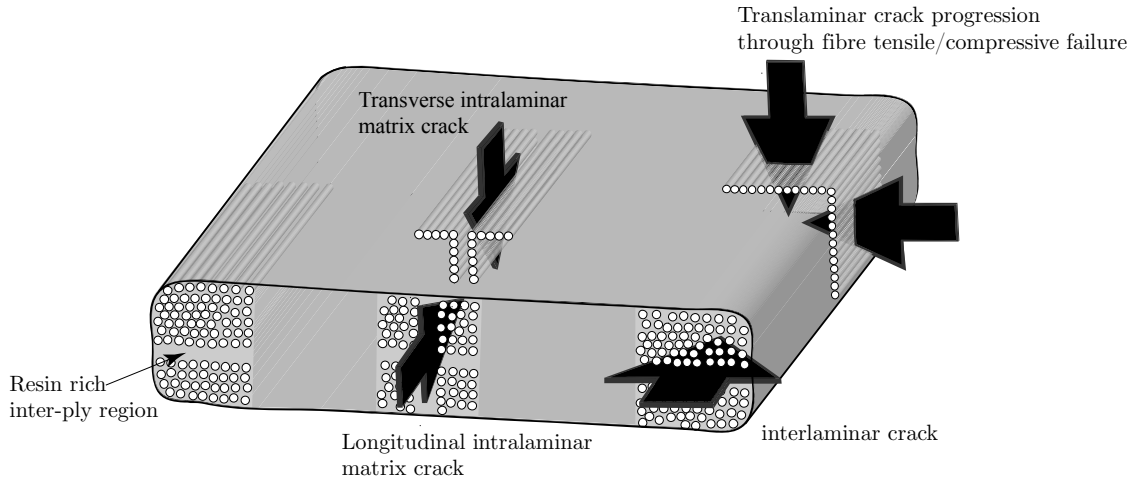


Figure 1.1. Directions of crack growth in laminated CFRP.

Cracks can propagate through the thickness of a laminate, rather than between plies, via the failure mechanisms illustrated in Figure 1.1: longitudinal and transverse intralaminar matrix failure, and fibre breaking translaminar failure. The fracture toughnesses associated with each of these modes of failure needs to be measured in order to fully characterise a given material system. Methods for characterisation of these properties are relatively immature, when compared with test methods for interlaminar failure (a current state of the art of translaminar fracture toughness testing is presented in Chapter 2), and the broad objective of the work presented within this thesis has been to develop and advance testing procedures for characterisation of translaminar fibre failure.

Testing methods for the measurement of the fracture toughness associated with mode I tensile and compressive loadings as well as mixed mode I/II tensile and shear loadings are investigated. The characterisation of these properties provides data that is useful for engineers during the material selection and design processes, as well as for numerical simulation of failure in laminated composites.

Chapter 3 presents an extensive investigation of the use of the compact tension (CT) specimen for the characterisation of the in situ toughness of the 0°

plies within a laminate. Suitable methods of data reduction are investigated with the aim of identifying the most appropriate. Using an identified optimum method, the effects of specimen in-plane size, thickness, and lay-up are then investigated; lastly the effects of notch-tip radius are studied to identify what manufacturing process must be used to obtain true measurements of fracture toughness.

In Chapter 4, a method for measurement of the fracture toughness associated with mode I fibre compressive failure is presented. Unidirectional four-point-bend specimens are tested and the critical strain energy release rate for initiation of damage in the form of shear driven fibre failure is obtained. This method is superior to other methods in the literature that are not able to achieve the desired failure mode in unidirectional specimens.

Combined tensile and shear loadings are applied to a CT type specimen in Chapter 5 to investigate mixed mode I/II translaminar failure. The objective here is to investigate the fracture mechanisms controlling mixed-mode failure and to characterise R-curves where possible.

Conclusions derived from Chapters 3 - 5 are presented in Chapter 6 and, finally, suggestions for further work are presented in Chapter 7.

Chapter 2

Translaminar fracture toughness testing of composites

2.1. Introduction

The desire for efficient design, coupled with an ever-increasing understanding of composite failure, is moving industry towards a more damage tolerant approach to design with composites. Methods for predicting the onset and subsequent propagation of damage in composite components are therefore highly desirable.

There are several approaches to model damage propagation, such as using cohesive elements [1], smeared crack models [2] or X-FEM [3]. Each of these requires the fracture toughness for the failure mode they are simulating. Development of experimental techniques for accurate characterisation of the toughness is therefore critical.

The failure modes exhibited by laminated composites can be divided into delamination, intralaminar fracture and translaminar fracture. Delamination has been under extensive investigation for several years. Round-robin exercises [4-7] have paved the way to the standardised procedures for measurement of mode I [8-

10], mode II [10] and mixed mode I/II [11] fracture toughness. The state of the art of interlaminar fracture toughness testing methods has been comprehensively reviewed by several authors over the past years (e.g. [12-16]).

A significant number of published studies have been dedicated to translaminar fracture toughness measurement; however, these are often isolated pieces of work with some dating back to the late '70s. Whilst the importance of translaminar fracture toughness measurement was recognised many years ago, it has received relatively little attention from the scientific community until now. This is at least partially due to (i) a lack of confidence in composites resulting in them not being used in primary structures where this type of characterisation is useful, and (ii) the lack of modelling capabilities which can use the parameters effectively.

This has now changed: large composite primary structures can be found on the latest aircraft, for example, and finite element analysis (FEA) is a common tool used for design. It is envisaged that the fracture toughnesses associated with the translaminar failure modes will play an increasingly important role over the coming years; therefore a review of the current literature is now opportune. This review draws together the conclusions of these studies, and presents a discussion to serve as a platform from which further work in the field can move forward. Work concerning characterisation of multidirectional laminates, as well as the individual translaminar ply failure modes, is presented in sections specific to specimen configuration. Methods of data reduction are noted and important experimental observations are highlighted.

2.2. Overview of composite failure modes

For a given loading condition, formation and propagation of damage within a laminate (and therefore its fracture toughness) will be lay-up dependent. Ultimately, failure will be governed by any one, or a combination of the ply-level failure mechanisms illustrated in Figure 1.1. The failure modes which can arise through direct in-plane loading are:

Translaminar fibre tensile failure. Although the critical strain energy release rate of a single carbon fibre has been estimated to be as low as 7.4 J/m^2 [17], this mode of failure within a composite is characterised by the dissipation of large amounts of strain energy. The large amounts of fibre-matrix debonding and subsequent fibre pull-out, visible on the fracture surface shown in Figure 2.1(a), result in a homogenised ply-level fracture toughness that has been measured to be over 3 orders of magnitude higher than that of the fibre alone.

Translaminar fibre compressive failure. Under an applied compressive load, failure of the fibres aligned with the loading axis can initiate as either shear driven fibre failure, Figure 2.1(b), or fibre kinking, Figure 2.1(c). Which failure mode occurs is dependent on the presence of shear stresses that can arise through features such as localised fibre misalignment. The relative motion of the crack faces during shear failure induces bending ahead of the crack tip, subsequently, a transition to the fibre kinking failure will always occur [18].

Intralaminar matrix failure. This is characterised by matrix cracking either longitudinally or transversely with respect to the fibres, as shown in Figure 1.1. The fracture surface resulting from longitudinal intralaminar matrix failure is highlighted as ❷ in Figure 2.1(a). The measured toughnesses of these failure modes are comparable to their interlaminar counterparts [19, 20].

The critical strain energy release rates associated with these modes of failure are properties that are intrinsic to the material system and need to be measured for complete characterisation of the damage tolerance of the material system in question.

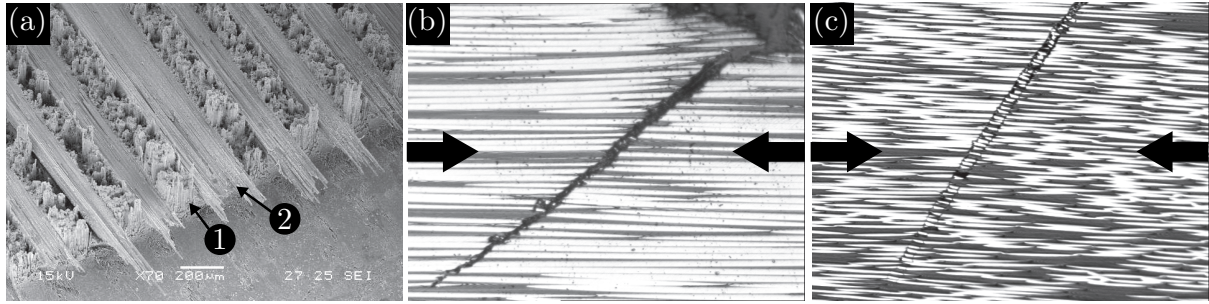


Figure 2.1. Failure mechanisms in FRP: (a) fracture surface including ❶ translaminal fibre tensile failure and ❷ longitudinal matrix failure, (b) shear driven fibre compressive failure (the arrows indicate the loading direction), (c) fibre kinking (the arrows indicate the loading direction).

2.3. Characterisation of Multidirectional laminates

The majority of work in the field of translaminal fracture characterisation has been in measuring the toughness of multidirectional laminates manufactured from unidirectional or 2D woven material systems. In these cases, the measured toughness is dependent on lay-up and will be the result of the constituent plies failing in a combination of any of the aforementioned failure modes. Of the specimen configurations found in the literature, shown in Figure 2.2, only the compact tension, three/ four point bend and extended compact tension specimens exhibit stable crack growth. Crack growth stability is necessary to capture the change in critical strain energy release rate with the evolution of damage in the form of a resistance curve (R-curve).

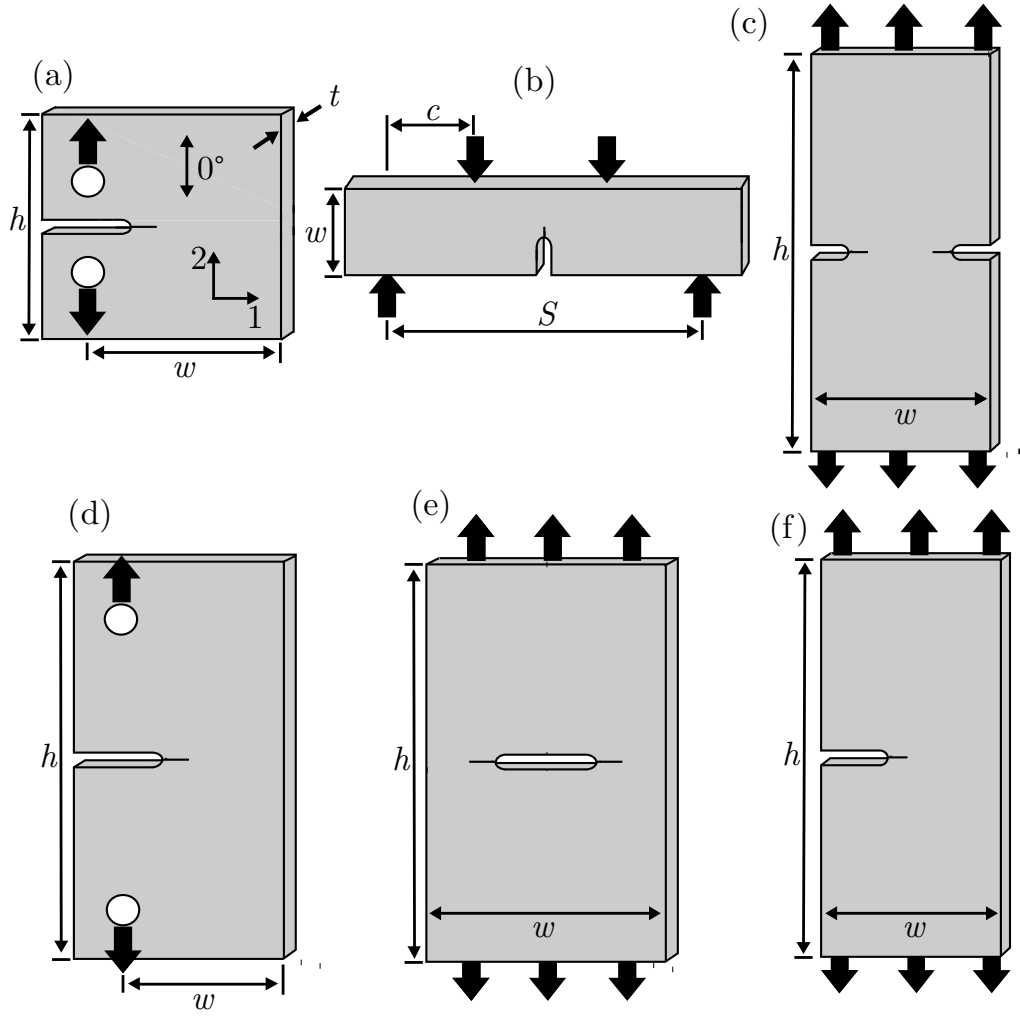


Figure 2.2. Specimen configurations used for translaminar fracture toughness measurement: (a) compact tension (has also been used for compressive tests), (b) four-point bend (three point bend configuration also possible but not shown), (c) double edge notched tension, (d) extended compact tension, (e) centre notched tension (has also been used for compressive tests), (f) single edge notched tension.

2.3.1. Compact Tension

The compact tension (CT) configuration, shown in Figure 2.2(a), is perhaps the most widely used specimen configuration for translaminar fracture toughness measurement of composites. It has been used for characterisation of carbon/epoxy [21-30], carbon/carbon [31], carbon/PEEK [32] and Boron/Aluminium [33, 34] laminates.

A range of approaches to data reduction have been used; but by far the most widely used [23, 25, 33-36] has been through the determination of the critical stress intensity factor, K_{Ic} , as recommended by the ASTM E399 for metallics [37]:

$$K_{Ic} = \frac{P_c}{t\sqrt{w}} f(a/w) \quad (2.1)$$

where P_c is the critical load causing crack extension, t and w are defined in Figure 2.2(a), a is the crack length and $f(a/w)$ is a finite-width correction factor. A number of studies [23, 25, 33-36] have used $f(a/w)$ values obtained from expressions developed for isotropic materials, however, it has been highlighted that errors can stem from inappropriate use of such correction factors with composites [26, 27].

The critical strain energy release rate, G_{Ic} , can be obtained from the critical stress intensity factor as [38]

$$G_{Ic} = \frac{K_{Ic}^2}{\sqrt{2E_{11}E_{22}}} \sqrt{\sqrt{\frac{E_{11}}{E_{22}} + \frac{E_{11}}{2G_{12}}} - \nu_{12}} \quad (2.2)$$

where E_{ii} are the laminate elastic moduli in the 1 and 2 directions (shown in Figure 2.2(a)), G_{12} is the laminate shear modulus, and ν_{12} the Poisson's ratio.

Alternatively, the critical strain energy release rate can be determined directly from experimental data, if the rate of change of compliance, C , with crack length is known:

$$G_{Ic} = \frac{P_c^2}{2t} \frac{dC}{da} \quad (2.3)$$

Two methods of compliance calibration have been used: the ‘‘Compliance Calibration method’’ (CC) makes use of direct measurement of the specimen compliance from load-displacement curves of tested specimens correlated with associated values of crack length (measured optically [39]). The ‘‘Modified experimental Compliance Calibration method’’ (MCC) uses measurement of the compliance of specimens with machined cracks of known length [34].

The area method has also been used [32], whereby the energy dissipated during fracture, equivalent to the area underneath the specimen load-displacement curve, is divided by the fractured area associated with crack growth.

Data reduction using FEA is an alternative [26]; the CT specimen can be modelled with 1 mm thickness and a 1 N load applied at the point of the centre of the loading pin to obtain the J-integral, J , for a range of crack lengths. By defining an energy release rate, normalised to the specimen of unit thickness with unit applied load, as

$$f(a) = J \cdot \left(\frac{1 \text{ mm}}{1 \text{ N}} \right)^2 \quad (2.4)$$

The laminate toughness can be obtained from

$$G_{Ic} = \left(\frac{P}{t} \right)^2 f(a) \quad (2.5)$$

An extensive investigation of all of the aforementioned data reduction schemes [27] found that the methods that relied on optical crack growth measurement were sensitive to errors in its measurement. It was mainly this factor that led to the conclusion that the MCC method was the most appropriate of the data reduction schemes investigated.

More recently, Catalanotti et al. [30] used a data reduction scheme that relies on digital image correlation to identify both the location of the crack tip and to compute the J -integral, enabling real-time generation of R-curves. Results from this method compare well with those produced via the MCC method for the same material system [29] thus making it a valid alternative provided the necessary equipment is available.

Prewo [33] investigated the effects of lay-up on the measured toughness values for fracture initiation in Boron/Aluminium composites and found that blocking together of 0° plies had a significant effect on the measured laminate toughness. The measured K_{Ic} for the $[90_6/0_6]_s$ lay-up was found to be 20% higher than that measured for the $[90/0]_{6s}$ lay-up. Investigation of the specimen fracture surfaces using scanning electron microscopy revealed significantly more fibre pull-

out in the specimens with blocked 0° plies. It is thought that this was due to a reduced constraining effect of the 90° plies in confining the crack to a single plane. Similar observations have been made for carbon-epoxy composites [27].

Harris and Morris [22] compared fracture toughness measurements obtained from three specimen configurations: compact tension (CT), Figure 2.2(a), centre notched tension (CNT), Figure 2.2(e), and three-point bend (3PB), Figure 2.2(b). Initiation values of K_{Ic} obtained from $[0/\pm 45/90]_{ns}$ and $[0/90]_{ns}$ lay-ups with thicknesses of 64, 96 and 120 plies were very similar for the three specimen configurations. For a $[0/\pm 45]_{ns}$ laminate, results obtained from CT and 3PB specimens were approximately 15% lower than those obtained via the CNT. The authors suggested that for CNT specimens the “pop-in” load (the load corresponding to the first load-displacement discontinuity) should be used, rather than the maximum load, P_{max} , as is historically the case with metals. CNT results calculated using this method correlated more favourably with those obtained from CT and 3PB specimens using P_{max} .

Masters [25] tested two sizes of CT specimen and reported a size effect. K_{Ic} values measured for 43×45 mm specimens were found to be approximately 20% lower than those obtained from specimens with the in-plane dimensions doubled. Crack growth in the smaller specimens was seen to be perpendicular to the intended growth direction, which questions the interpretation of the quoted values for K_I ; this problem has been reported elsewhere [24, 40]. A more recent study on size effects [28] of carbon epoxy composites, with self-similar crack growth, observed no effect when the in-plane size (area) of the specimen is doubled.

2.3.2. Extended Compact Tension

The extended compact tension (ECT) specimen, shown in Figure 2.2(d), is another design originating from toughness testing of metallics [41]. It was adopted by Underwood et al [24] for testing of carbon/epoxy laminates, to resolve problems with undesirable failure modes. Damage beneath the central loading point was encountered during three-point-bend testing and crack growth perpendicular to the

desired growth direction was observed for some laminate configurations during CT testing. Kongshavn and Poursartip [40] also adopted a configuration similar to the ECT, when the same problem with the CT specimen was encountered.

The ECT is the only specimen configuration that has become standardised for translaminar testing: ASTM E1922-04 [42]. El-Hajjar and Haj-Ali [43] assessed the applicability of the E1922 ASTM standard for use with pultruded thick-section composites; the conclusions of the study extended the scope of the standard to include these materials.

The ASTM E1922 standard is limited in that it only covers fracture toughness determination for damage initiation. An R-curve effect, where the propagation toughness has been measured to be significantly larger than that at initiation, is exhibited by at least some [26, 27, 29], if not all material systems, and results have been successfully used to numerically reproduce experimentally measured R-curves [44].

Calculation of K_{Ic} for fracture initiation uses equation (2.1), but the term $f(a/w)$ is strictly valid only for isotropic materials. The measured K_{Ic} , deduced from the maximum load during testing, is deemed to give a consistent value of laminate fracture toughness if the following relationship is satisfied:

$$\Delta d_n / d_{n-0} \leq 0.3 \quad (2.6)$$

where $d_{n-0} = d_n$ at $P = P_{max}$ on the extension of the initial linear portion of the load-displacement plot, and Δd_n is the additional notch-mouth displacement up to the P_{max} point, as shown in Figure 2.3. If the criterion outlined in equation (2.6) is not satisfied, the damage zone that has formed at the notch tip may be too large to obtain a valid measure of K_{Ic} [42].

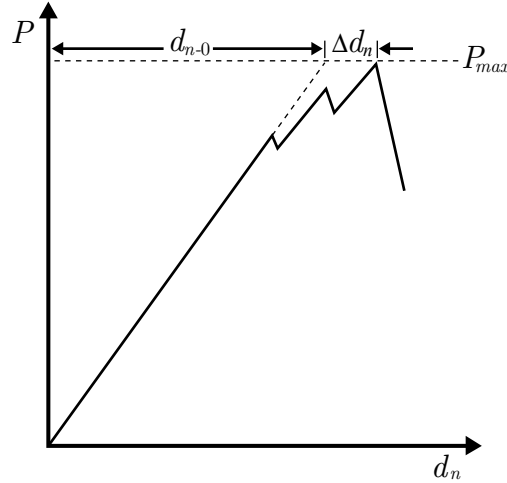


Figure 2.3. Typical load-displacement plot for ECT specimen.

2.3.3. Centre Notched Tension/Compression

The centre-notched tension specimen, shown in Figure 2.2(e), has been used for the testing of carbon/epoxy [21, 22, 25, 45-47], carbon/bismaleimide [47] and boron/aluminium material systems [33, 48]. It has also been used for characterisation of the compressive fracture toughness of carbon/epoxy laminates by reversal of the loading configuration shown in Figure 2.2(e). Measurements of the critical stress intensity factor have been obtained from [25, 33, 45, 48, 49]

$$K_{Ic} = \sigma_c \sqrt{\pi a} f(a/w) \quad (2.7)$$

where σ_c is the applied stress at fracture initiation.

Prewo [33] found that for boron/aluminium laminates with a $(90/0)_{6s}$ lay-up, the value of K_{Ic} obtained from CNT tests were considerably lower than that obtained using the CT specimen. No explanation was given for this discrepancy; however, it was postulated that the difference in results could lie in the methods of data reduction. This is most likely the case as the term $f(a/w)$ assuming isotropic values was used in obtaining K_{Ic} .

Yeow [46] compared fracture toughness measurements resulting from four different data reduction methods. The inherent-flaw method, the point stress and average stress methods all arise from modifications of equation (2.7). (For details,

the reader is referred to [46]). A boundary integral approach was favoured over these approaches as it could be more readily applied to various geometries and load cases.

Using the CCT specimen configuration, Harris and Morris [21] found that for $[0/\pm 45/90]_{ns}$ and $[0/90]_{ns}$ laminates, K_{Ic} for fracture initiation decreased with increasing laminate thickness. In both cases, an asymptotic value of K_{Ic} was approached above a laminate thickness of 64 plies. For the quasi-isotropic case, the observed thickness effect was attributed to differences in subcritical damage at the crack tip, as the 8 ply specimens exhibited more delaminations and matrix cracking than in the thicker laminates. For the cross-ply case, the decrease in toughness with increasing laminate thickness was attributed to longitudinal splitting of the 0° plies. The 8 ply laminate exhibited extensive splitting, evident by X-ray examination, which subsequently reduced the strength of the stress concentration at the notch tip. The splitting was shown to be a surface effect, having less influence on the measured K_{Ic} with increasing laminate thickness.

Underwood et al. [47] measured the translaminar fracture toughness for initiation of a carbon/epoxy system with a $(0/\pm 60/0/\pm 60/0)_s$ lay-up; three different initial notch radii were used: $\rho = 0.02, 0.3$ and 3 mm. It was found that the toughness was at a lower bound at $\rho = 0.02$ and 0.3 mm, with the maximum measured value of K_{Ic} being 28.8 % greater for the specimens with $\rho = 3$ mm.

Masters [25] measured the K_{Ic} of 51×203 mm 2D woven carbon/epoxy CNT specimens to be approximately 19% lower than 102×406 mm specimens of the same (6 ply) thickness. No comment was made on the source of this difference, which is also likely to result from the use of finite width correction factors based on isotropic material assumptions.

Soutis et al. [49] measured the compressive fracture toughness of various multidirectional carbon/epoxy laminates. The results indicated that toughness increased with increasing number of 0° plies, confirming that they make a major contribution to the resistance of compressive fracture.

2.3.4. Three Point Bend

Similarly to the CT arrangement, the three-point bend (3PB) specimen design also exists as an ASTM standardised specimen for metals [37] and has been used for translaminar toughness characterisation of composites [21, 22, 24].

There have been varied approaches to the data reduction process with the 3PB specimen, with the early work of Harris and Morris [21, 22] using an FE based method of data reduction similar to that outlined in equations (2.4) and (2.5). In the paper presented by Underwood et al. [24], K_{Ic} was calculated using equation (2.1) with $f(a/w)$ being replaced by $f(S/w, a/w)$, where S is defined in Figure 2.2(b). Damage formation beneath the central loading point in quasi-isotropic 3PB specimens led to the investigation of the ECT as an alternative configuration.

2.3.5. Single/Double Edge Notched Tension

The single (SENT) and double (DENT) edge notched tension specimens are shown in Figure 2.2(c) and (f) respectively. The SENT specimen has been used to characterise carbon/epoxy systems [25, 50, 51] as well as E-glass, Kevlar and boron/epoxy [50] and the DENT glass/epoxy systems [52].

Typically, the critical stress intensity factor for both configurations has been obtained from [25, 50, 52]

$$K_{Ic} = \sigma_c \sqrt{af(a/w)} \quad (2.8)$$

though the MCC method (section 2.3.1) has also been used for toughness determination of carbon/epoxy SENT specimens [51].

Mandell et al. [50] investigated the effects of specimen geometry as well as the effect of reinforcing fibres (carbon, E-glass, Kevlar and Boron). Increasing the ply thickness was seen to increase the size of the damage zone, observed through micrographs of specimens having the damage zone stained with a fluorescent dye placed under a fluorescent light. The lengths of intralaminar cracks in 90° plies and longitudinal splits in 0° plies of cross-ply, $\pm 15^\circ$, and $\pm 30^\circ$ laminates were observed

to be approximately proportional to K_I^2 . Also, the lengths of longitudinal splits in the 0° plies at fracture were found to be proportional to the ply thickness, with the resulting laminate K_{Ic} , measured for cross-ply carbon/epoxy, also increasing linearly with ply thickness.

Hallett and Wisnom [52] presented K_{Ic} values for a baseline specimen of $h = 100$ mm and $w = 20$ mm; these in-plane dimensions were also scaled such that specimens of $w = 10$ and 40 mm were also tested. For both $[90/0]_s$ and $[+45/90/-45/0]_s$ lay-ups, K_{Ic} was observed to increase with specimen size. It was noted that the results seemed to asymptote towards a constant value at larger specimen sizes.

2.4. Characterisation of Translaminar Ply Failure Modes

The work reviewed thus far has been concerned with measurement of fracture toughness of laminates. The resulting critical strain energy release rates, or stress intensity factors, are essentially a homogenisation of the damage mechanisms being exhibited in each constituent ply. These lay-up dependent properties are of limited use as they are not representative of any property intrinsic to the material system.

Vaiyda and Sun [53] first demonstrated the concept of an in situ 0° ply fracture toughness, a material property intrinsic to the material system, regardless of lay-up. Using classical lamination theory, they related the stress carried by the 0° plies in a laminate, σ_c^0 , to the stress applied to the laminate, σ_c , via a parameter η such that

$$\sigma_c^0 = \eta \sigma_c \quad (2.9)$$

with η being dependent only on the lay-up and elastic properties of the constituent plies in the laminate in question. It was then shown that the critical stress intensity factor of a notched laminate, K_{Ic} , and the critical stress intensity factor associated with the 0° plies, K_{Ic}^0 , are related via the same parameter:

$$K_{Ic}^0 = \eta K_{Ic} \quad (2.10)$$

Using this model, the authors were able to predict the fracture toughness of various multidirectional laminates to a high degree of accuracy, thus illustrating that K_{Ic}^0 , or equally G_{Ic}^0 , is an intrinsic material property. Methods that have been used to characterise the in situ fracture toughness of 0°plies, for both tensile and compressive failure modes, are discussed in the following sections.

2.4.1. Tensile Failure

Vaidya and Sun [53] measured K_{Ic}^0 using carbon/epoxy CNT specimens, however the use of the CT specimen [26-30] has prevailed more recently owing to the ability to capture the full R-curve behaviour. Leach and Moore [54] used a unidirectional 3PB arrangement for testing of a carbon/PEEK material system, however, high ductility at the crack tip, longitudinal splitting and failure of the compressive side of the specimens, called into question the suitability of this testing arrangement.

Two approaches have been used for determination of the fracture toughness of the 0° plies. Vaidya and Sun [53] used equation (2.10) to determine the property from the fracture toughness of 9 multidirectional laminates of differing lay-ups. Values of K_{Ic}^0 were found to be approximately equal for all laminate configurations. Pinho et al. [26] measured the fracture toughness of cross-ply CT specimens and used a rule of mixtures type approach to obtain G_{Ic}^0 . It was assumed that the laminate fracture toughness can be partitioned into the separate contributions made by the 0° plies and 90° plies such that G_{Ic}^0 could be obtained from

$$G_{Ic}^0 = \frac{t_{lam}}{t_0} G_{Ic}^{lam} - \frac{t_{90}}{t_0} G_{Ic}^{90} \quad (2.11)$$

where t_{lam} is the laminate thickness, and t_0 and t_{90} are the total thicknesses of 0° and 90° plies within the laminate respectively. G_{Ic}^{lam} is the critical energy release rate for the laminate and G_{Ic}^{90} represents the fracture toughness associated with mode I longitudinal intralaminar matrix failure.

The fracture toughness associated with the 0° plies in a laminate has been measured to dramatically increase with ply thickness. By doubling the 0° ply thickness, G_{Ic}^0 was observed to almost double for a T300/920 material system [28], the fracture surfaces of failed specimens indicated that increasing lengths of pulled out fibre accounted for the increased amount of dissipated energy.

The effect of notch root radius, ρ , on G_{Ic}^0 has been investigated for cross-ply IM7/8552 CT specimens [29]. The translaminar toughness, measured at initiation, was found to be independent of notch root radius for $\rho \leq 250 \mu\text{m}$. Above this value, an apparent initiation toughness was measured. It was concluded from fractographic analysis of specimens that there is no significant interaction between the 0° and 90° ply failure modes and that the critical notch root radius of the tested laminate was a property of the 0° plies.

Published fracture toughness data for translaminar tensile failure is summarised in Table 2.1.

2.4.2. Compressive Failure

Attempts have been made to measure the fracture toughness associated with mode I fibre compressive failure of carbon/epoxy systems using the previously discussed CNC specimen [49], the compact compression specimen [26, 30, 55, 56] (which is essentially the CT specimen loaded in compression) and the 4PB arrangement [57].

The method of equations (2.4) and (2.5) has been used to determine laminate critical energy release rates [26, 57]. For the case where cross-ply laminates were tested [26], equation (2.11) was modified such that G_{Ic}^{90} became G_{IIc}^{90} , the mode II matrix cracking intralaminar fracture toughness.

Soutis et al. [49] tested unidirectional CNT specimens. The specimens failed by cracks propagating parallel to the fibres, orthogonally to the initial notch thus no meaningful fracture toughness data could be deduced. This problem was overcome by Sutcliffe and Fleck [58] who preconditioned specimens by adding initiation sites

for intralaminar splitting away from the notch tips intended for fibre compressive failure, so that splitting and fibre failure occur simultaneously.

Pinho et al. [26] noted that although meaningful initiation values could be obtained from compact compression tests, the presence of contact forces across the kink-band during propagation meant that the R-curve could not be derived from the data observed during testing.

A unidirectional 4PB specimen [57] has allowed for the compressive failure mode to be obtained in isolation, something that has not been possible thus far with the other specimen configurations. The use of acoustic emission and monitoring of the specimen surfaces with a microscope revealed that compressive failure initiates long before any signs are visible on the load displacement curve.

Published fracture toughness data for translaminar compressive failure is summarised in Table 2.1.

Table 2.1. Summary of experimental data obtained from characterisation of the tensile and compressive translaminar failure modes

Study	Specimen configuration	Material System	G_{Ic}^0 (kJ/m ²)	
			Initiation	Propagation
Tension				
Vaiyda and Sun [53]	CNT	AS4/3501-6	220.6	-
Pinho et al. [26]	CT	T300/913	91.6	133.0
Laffan et al. [27]	CT	T300/920	56.1	65.2
Catalanotti et al. [30]	CT	IM7/8552	97.8	133.3
Laffan et al. [29]	CT	IM7/8552	112.7	147.2
Compression				
Pinho et al. [20]	CC	T300/913	79.9	-
Catalanotti et al. [30]	CC	IM7/8552	47.5	-
Laffan et al. [57]	4PB	IM7/8552	25.9	-

2.5. Discussion

2.5.1. Tension

A range of specimen configurations have been used for translaminar tensile testing, however, as already mentioned, only the CT, ECT, 3PB and 4PB exhibit stable crack growth which is necessary for R-curve characterisation. Of these, only the ECT has yet to exhibit any experimental problems; it was introduced as a result of crack propagation perpendicular to the desired growth direction in CT specimens, and loading point damage in 3PB specimens [24]. It is evident from the literature, however, that use of the CT has still prevailed since the inception of the ASTM standard [42] in 1997. Various publications have illustrated that its use is still perfectly valid for the testing of material systems/lay-ups that do not present undesired failure modes. It also has the advantage of using far less material than the ECT (one ECT specimen is typically around twice the size of a CT), which is perhaps the reason why the use of the CT specimen has still prevailed

Nevertheless, Table 2.1 illustrates a wide range of measured fracture toughness data for different material systems and it is likely that in the future, newer systems will be developed to be tougher still. These tougher systems will require ECT type specimens if tests are to be performed without the undesirable failure modes observed in several studies [24, 25, 40].

The data reduction scheme suggested in the ASTM E1922 standard is suitable for quasi-isotropic laminates, but the results become less accurate for highly orthotropic lay-ups. Given the fact that cross-ply specimens are often used to characterise G_{Ic}^0 , an appropriate data reduction scheme should be used. Generally, approaches to the data reduction process have been varied; and the danger of over-reliance on methods intended for isotropic materials has been highlighted [27]. For the specimen configurations suitable for this type of test there are no simple analytical formulae for dC/da or the J -integral, however the use of FE [26, 27] or DIC [30] has provided researchers with the quickest routes to accurate results.

Compliance calibration curve generation using specimens with known (machined) crack lengths has proved to be the optimum alternative to FE [27].

An important material property not currently addressed by the ASTM E1922 standard is the R-curve. Full characterisation of the R-curve for various carbon/epoxy material systems shows that the difference between the initiation and propagation toughness is not trivial [26-28, 30]. This is something which should be addressed in any future revisions of the testing standard.

Finally, it should be noted that whilst G_{Ic}^0 has been found to be a ply-level material property, the measured values have been observed to increase dramatically with ply thickness [28, 33]. Given the availability of different thickness grades for a given material system, care must be taken to ensure the appropriate values of G_{Ic}^0 are characterised.

2.5.2. Compression

Measurement of the fracture toughness associated with translaminal compressive failure poses a greater challenge than for the tensile failure. Obtaining the fibre failure mode in isolation has been a problem; the CC specimen [26] exhibits simultaneous 90° and 0° ply failure, and the centre cracked compression [58] specimen exhibits simultaneous splitting between the fibres of the 0° plies and failure of the fibres themselves. These issues have been overcome by the use of the 4PB configuration.

A detailed study on the initiation of fibre compressive failure [57] found that initiation occurred long before any signs of its occurrence were visible on the load-displacement curve. This calls into question the validity of results from studies which have not monitored failure with such accuracy; the fracture toughness measured in [30] is almost double the value presented in [57], as can be seen from Table 2.1. This discrepancy suggests that the compact compression specimen is unsuitable for characterisation of compressive fracture toughness in its current form.

Data reduction options are limited. A lack of a simple analytical formulation of the strain energy release rate of the specimens necessitates the use of FE. Any form of compliance calibration is made difficult by the effect of contact of the crack faces during damage propagation; it was seen that damage initiation has negligible effect on specimen load-displacement curves [57].

Measurement of an initiation value has been adequately addressed, however, these are of limited use without any propagation data. The ultimate goal is the characterisation of an R-curve, however this is hampered by several factors. Monitoring crack growth is not trivial as the embedded damage can be masked by out of plane failure by the plies at the surface [30] and an accurate effective crack length cannot easily be derived from compliance.

2.6. Conclusions

The work presented here illustrates the diversity of approaches which have been used to characterise translaminar fracture. A wide range of specimen configurations, sizes and data reduction strategies have been used. This diversity has arisen from the innovations of many researchers, all of whom are motivated by the difficulty in triggering the desired failure mode and obtaining accurate results. Having reviewed the work available in the literature, a platform for future testing has been built based on the identification of best practice.

For tensile testing, matters of data reduction appear to be solved. A specimen suitable for all material systems and lay-ups can be derived from the current CT and ECT configurations, and its geometry will be dictated by the toughest available material systems.

Characterisation of toughness for compressive failure poses more of a problem; current methodologies are unable to capture the R-curve behaviour at this stage.

Although significant steps have been taken towards solid, well defined procedures for translaminar toughness characterisation, it is clear that the volume of work is lagging significantly behind that in the interlaminar field. Translaminar

fracture testing is not yet a mature field; however, with an increasing need for high quality experimental data, the best practice identified here provides a solid baseline from which to continue further investigation.

2.7. Publications

The literature review presented within this chapter has resulted in the following publication:

Translaminar fracture toughness testing of composites: A review. M. J. Laffan, S. T. Pinho, P. Robinson, A. J. McMillan. Polymer Testing 2012.

Chapter 3

Mode I translaminar tensile failure

3.1. Introduction

Numerous studies have measured the translaminar fracture toughness of composite laminates in tension [21-24, 34]. A range of approaches in terms of both specimen configuration and data reduction schemes have been used. Perhaps the most commonly used specimen is the compact tension (CT) configuration, which also exists as a recommended configuration for fracture toughness characterisation of metallics [37].

The use of the CT specimen for toughness measurement in composites has previously been disregarded, as longitudinal splitting perpendicular to the notch tip was observed in highly orthotropic laminates [24]. The results of this study favoured an extended compact tension (ECT) specimen configuration, and subsequently led to the creation of the ASTM E1944 “Standard test method for translaminar fracture toughness of laminated and pultruded polymer matrix composite materials” [42]. The standard recommends the use of the ECT and calculation of the laminate critical stress intensity factor using a finite width correction factor developed for isotropic materials. For the measurement of toughness at fracture initiation, the finite width correction factor has been deemed valid for a wide range of composite laminates. However, the standard only covers measurement of toughness for fracture initiation and so the full R-curve behaviour cannot be determined.

There have been few studies that have specifically attempted to measure the fracture toughness associated with fibre tensile failure. In 1997, Vaidya and Sun [53] developed a fracture criterion based on the 0° plies in a laminate; the authors stated that the critical stress intensity factor for fibre failure, K_{Ic}^0 , is a material constant

and should be independent of laminate configuration. The initiation values of K_{Ic}^0 measured using the centre-cracked tension specimen geometry, calculated using a finite width correction factor developed for isotropic materials, were found to be approximately equal for the range of laminates tested.

The CT specimen is another configuration that has been used to measure fracture toughness associated with mode I fibre tensile failure [26]. The critical energy release rate of the fibre tensile failure mode, G_{Ic}^0 , at both fracture initiation and propagation were calculated using a finite element (FE) based method of data reduction; fractographic analysis of the test specimens indicated that the measured values may be lay-up dependent.

The work presented in this chapter aims to take a step towards defining accurate methods for measurement of the toughness of mode I fibre tensile failure. Firstly, six methods of data reduction that have been used throughout the literature to calculate toughness are investigated in section 3.2. Having identified an appropriate method of data reduction, the effect of specimen geometry and laminate stacking sequence on the measured translaminar fracture toughness is investigated in section 3.3. Finally, in section 3.4 the effect of notch tip radius on the measured initiation value of toughness is investigated with the aim of identifying the critical value below which true measurements of toughness can be obtained.

3.2. Investigation of Data Reduction Schemes

3.2.1. Materials and Manufacture

The material system used for the experiments presented in this section was a T300/920 unidirectional carbon/epoxy pre-preg. The material properties used for the data reduction were obtained using standard tests and are presented in Table 3.1 in the principal material axes.

The method of data reduction used to extract the 0° ply-level toughness from the laminate toughness necessitates that the energy dissipation in non 0° plies is known. A cross-ply laminate was chosen for this purpose, as will be discussed later in section 3.2.3.6. Composite panels with a lay-up of $[(90/0)_s/90]_s$ were manufactured using the hand lay-up method and cured to the pre-preg manufacturer's instructions. It has been postulated that the 0° ply toughness is dependent on ply thickness [26], in the light of this it was decided to place 90° plies either side of the laminate symmetry plane to avoid blocking of 0° plies. Furthermore, the laminate was designed to have 90° surface plies so that a self-similar crack could be optically monitored on the specimen surface during the tests. The compact tension specimen geometry, shown in Figure 3.1, was taken from a previous study [26]. A wet saw was used to cut the specimens and the 8 mm holes were made using a carbide tipped drill.

A three-step procedure was followed to obtain a sharp crack tip in the CT specimens: (i) a 4 mm wide notch of approximately 30 mm length was machined using a diamond coated disk-saw, (ii) the notch was then extended to 40 mm using a 0.2 mm thick razor saw, and (iii) finally, a 0.1 mm thick razor blade was used to further sharpen the notch-tip in a sawing motion.

Prior to testing, a scale with a 1 mm increment was drawn on the CT specimens in order to optically measure crack growth and the individual specimen dimensions were measured and recorded.

Table 3.1. Mechanical properties of T300/920 unidirectional lamina.

Modulus (GPa)			Major Poisson's ratio	Propagation G_{Ic}^{delam} (kJ/m ²)
Longitudinal	Transverse	Shear		
135.10	8.88	4.54	0.32	0.456

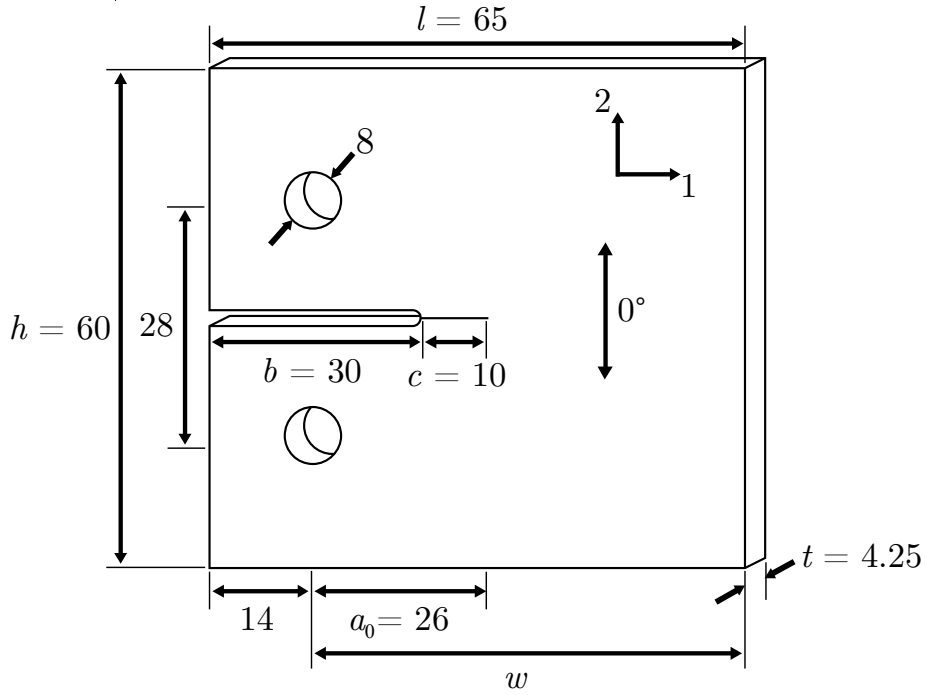


Figure 3.1. CT test specimen nominal dimensions (in mm).

3.2.2. Test method and Experimental Setup

Six CT specimens were tested using an Instron machine with a 10 kN load cell; each specimen was loaded under displacement control at a rate of 0.5 mm/min. The experimental setup is pictured in Figure 3.2(a). Measurements of load and crosshead displacement were recorded using a data logger. After the test, the values of crosshead displacement were corrected to account for compliance of the test machine and fixtures, which had been previously measured to be 0.06 mm/kN.

A crack-growth scale was prepared by hand on the specimen surface using correction-fluid and a pen, as shown in Figure 3.2(b). A CCD camera, which can be seen in Figure 3.2(a), was used to view a magnified image of approximately 12 mm

of the area of the specimen containing the crack-growth scale. This magnified image was used together with an event marker connected to the data logger to monitor crack growth.

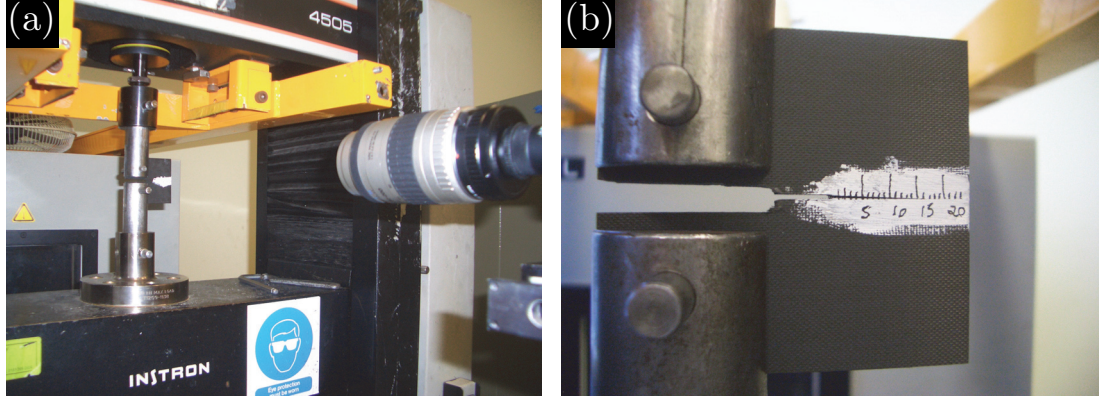


Figure 3.2. (a) Picture of specimen setup in the test machine with CCD camera in place, (b) close up of specimen during a test.

3.2.3. Data Reduction

3.2.3.1. ASTM E399

The ASTM E399 testing standard [37], valid for isotropic metals, gives the critical stress intensity factor as

$$K_{Ic} = \frac{P_c}{t\sqrt{w}} f(a/w) \quad (3.1)$$

$$\text{with } f(a/w) = \frac{2 + a/w}{(1 - a/w)^{1.5}} \left[0.886 + 4.64(a/w) - 13.32(a/w)^2 + 14.72(a/w)^3 - 5.6(a/w)^4 \right]$$

$$(3.2)$$

where P_c is the measured critical load that causes fracture, t is the thickness of the specimen, w is the dimension from the load line to the right hand edge of the specimen, as indicated in Figure 3.1, and a is the crack length, whose initial value a_0 is also indicated in Figure 3.1. The critical strain energy release rate of the laminate can then be calculated from the laminate critical stress intensity factor as [59]

$$G_{Ic}^{lam} = \frac{K_{Ic}^2}{\sqrt{2E_1E_2}} \sqrt{\sqrt{\frac{E_1}{E_2}} + \frac{E_1}{2G_{12}} - \nu_{12}} \quad (3.3)$$

where E_1 and E_2 are the Young's moduli in the 1 and 2 directions respectively (see Figure 3.1), G_{12} the shear modulus, and ν_{12} the Poisson's ratio.

A FE analysis was carried out using the commercial package Abaqus to investigate the accuracy of equations (3.1)-(3.3) when applied to orthotropic laminates. A half CT specimen was modelled using uniform square 8-noded elements (S8R5), with side $l = 0.5$ mm as shown in Figure 3.3. Models were 1 mm thick and a 1 N load was applied at the position of the loading pin. Several models with elastic properties equivalent to lay-ups with varying proportions of 0° and 90° plies were run to obtain the J-integral (J) around the crack tip. The obtained J values were then compared with the strain energy release rates obtained from equations (3.1)-(3.3) for the same geometry and loading. The results, for ranges of a/w used in this study, are shown in Figure 3.4.

It can be inferred that equations (3.1)-(3.3) cannot be readily applied to highly orthotropic laminates. For the $[(90/0)_8/90]_s$ lay-up used in this study, the maximum error is 13.2%; this maximum error increases as the proportion of 0° plies increases.

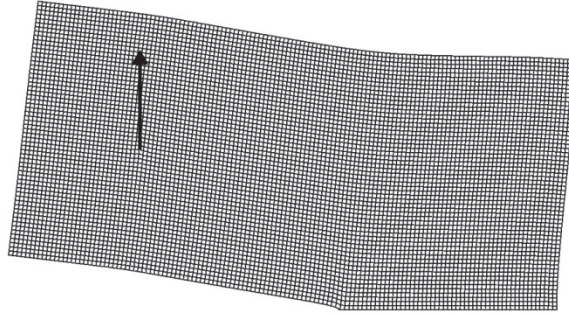


Figure 3.3. FE mesh of a half CT specimen.

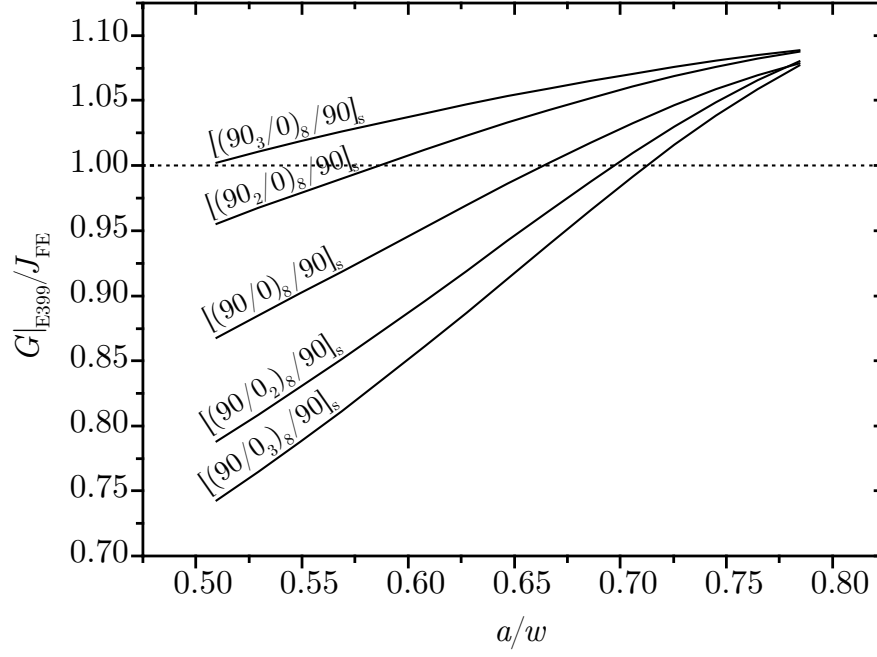


Figure 3.4. Comparison of strain energy release rates obtained from FE and from ASTM E399 method.

3.2.3.2. Area method

The area method is among the simplest methods of data reduction. The critical strain energy release rate can be calculated as

$$G_{Ic} = \frac{1}{2t\Delta a} (P_1 d_2 - P_2 d_1) \quad (3.4)$$

where the critical loads, P_i , and corresponding displacements, d_i , are defined in the representative load vs. displacement plot shown in Figure 3.5; the shaded region indicates the energy consumed for a crack growth of Δa .

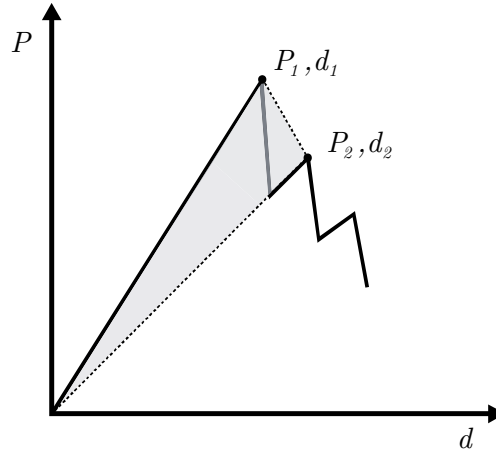


Figure 3.5. The area method: definition of terms in equation (3.4).

3.2.3.3. Compliance Calibration

The critical strain energy release rate can be calculated using the change in compliance, C , with crack length:

$$G_{Ic} = \frac{P_c^2}{2t} \frac{dC}{da} \quad (3.5)$$

The compliance calibration method (CC) requires the elastic compliance of the specimen at each optically measured crack length, a_{vis} , to be determined directly from the load displacement curve.

The experimental C vs. a_{vis} data was plotted and fit with a function of the form [44]

$$C = (\alpha a + \beta)^\chi \quad (3.6)$$

where α , β , and χ were calculated to best fit the experimental data. The critical strain energy release rate at each optically measured crack length was then obtained from the following expression

$$G_{Ic} = \frac{P_c^2}{2t} \cdot \alpha \chi (\alpha a + \beta)^{\chi-1} \quad (3.7)$$

3.2.3.4. Modified Compliance Calibration

A modified compliance calibration (MCC) method that eliminates the need to optically measure crack growth during testing was used to calculate the fracture toughness. In this case, the elastic compliance of the CT specimen was measured at various machined cracks of known length. These measurements were taken across the whole of the potential crack growth range, in approximately 3 mm increments, to capture the full compliance versus crack length response. Note that a compliance calibration curve can also be obtained by using FE.

The calibrated C vs. a data was plotted and fitted with equation (3.6). For the fracture toughness tests, an effective crack length, a_{eff} , could be determined using the elastic compliance measured from the load displacement curve of the test specimens, rearranging equation (3.6):

$$a_{eff} = \frac{C^{1/\chi} - \beta}{\alpha} \quad (3.8)$$

The critical strain energy release rate was then determined using equation (3.7). Fracture toughness measurements were also obtained by using a_{vis} with the data obtained from the MCC.

3.2.3.5. J-Integral/VCCT

For this case, the FE model of the half CT specimen already discussed, shown in Figure 3.3, was used to compute J for different values of crack length. It should be noted, however, these values of strain energy release rate could have alternatively been obtained using the virtual crack closure technique (VCCT). A normalised energy release rate $\bar{G}_I(a)$ was defined as [26]

$$\bar{G}_I(a) = J \cdot \left(\frac{1 \text{ mm}}{1 \text{ N}} \right)^2 \quad (3.9)$$

where the function $\bar{G}_I(a)$ can be approximated by a third order polynomial

$$\bar{G}_I(a) = c_3 a^3 + c_2 a^2 + c_1 a + c_0 \quad (3.10)$$

Finally, the critical strain energy release rate can be defined as

$$G_{Ic} = \bar{G}_I(a) \left(\frac{P_c}{t} \right)^2 \quad (3.11)$$

3.2.3.6. Determination of the fracture toughness associated with the 0° plies within the laminate

Once the critical energy release rate for the laminate is obtained, the critical energy release rate corresponding to fibre tensile failure is obtained by accounting for the toughness corresponding to matrix cracking in the 90° layers. This procedure neglects other damage modes such as delamination, as well as any interaction between matrix cracking and the fibre dominated failure modes, and assumes that a single matrix crack parallel to the pre-crack occurs in the 90° layers. This approximation seems reasonable as the toughness of mode I fibre tensile failure is much higher than that of the matrix. For a general [90/0] cross-ply laminate, the critical energy release rate for the fibre tensile failure mode is thus expressed as

$$G_{Ic}^0 = \frac{t_{lam}}{t_0} G_{Ic}^{lam} - \frac{t_{90}}{t_0} G_{Ic}^{90} \quad (3.12)$$

where t_{lam} is the specimen thickness, and t_0 and t_{90} are the total thicknesses of 0° and 90° plies within the laminate respectively. G_{Ic}^{lam} is the critical energy release rate for the laminate, as obtained from CT testing and G_{Ic}^{90} represents the mode I matrix cracking intralaminar fracture toughness. G_{Ic}^0 is the fracture toughness for the fibre tensile failure mode; this property represents a ply level toughness and is a homogenisation of the energy dissipated due to the various failure mechanisms happening within the ply during damage propagation, these being: fibre fracture, fibre-matrix debonding and subsequent pull-out. It is believed that the latter two account for most of the dissipated energy as the fracture toughness of carbon fibre has been estimated to be as low as 7.4 J/m^2 [17].

The mode I intralaminar critical energy release rate for through-the-thickness matrix crack growth has been found to be very similar to the interlaminar toughness between the 0° plies [20] so that for materials where the matrix cracking intralaminar critical energy release rate is unknown, the interlaminar critical energy release rate is thought to be a good approximation. As already noted, for carbon/epoxy systems, the toughnesses of the matrix failure modes are much lower than that of the fibres so the last term in equation (3.12) could be omitted without any significant loss of accuracy.

3.2.4. Results

All the CT specimens tested exhibited stick-slip crack growth during testing, with several 2-5mm crack growths typically being observed. A typical load-displacement curve is shown in Figure 3.6. Only the peak loads, after which the crack propagated unstably, were used for fracture toughness calculations.

Specimens were subjected to maximum loads typically in the region of 2.5 kN. For this reason, each specimen was optically inspected for signs of hole damage. Furthermore, it was possible to X-ray some specimens to verify that hole damage

was not present. A typical X-ray is shown in Figure 3.7, as can be seen, there are not obvious signs of damage at the specimen holes.

The elastic compliance was measured directly from the load-displacement curve after each crack growth; the results are shown in Figure 3.8. Also shown in the figure is the compliance versus crack length behaviour obtained from FE, and the results of the MCC measurements.

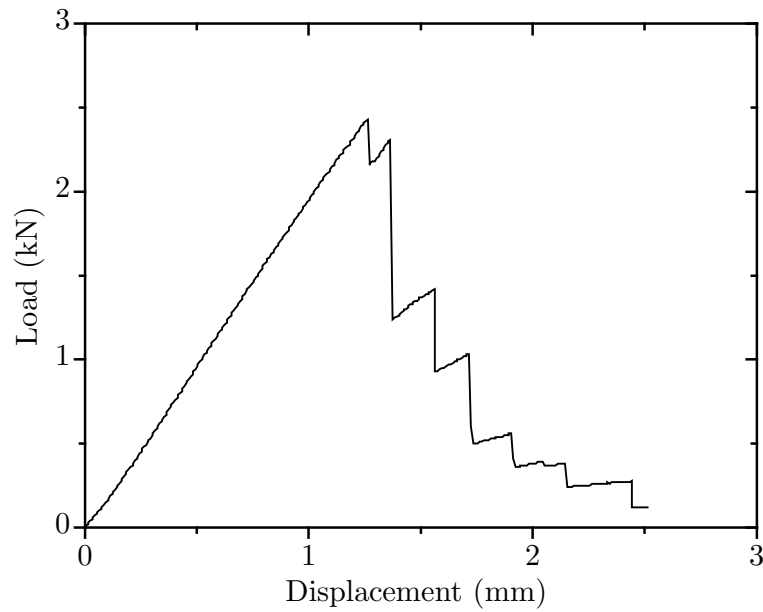


Figure 3.6. Typical load vs. displacement curve for a CT specimen.

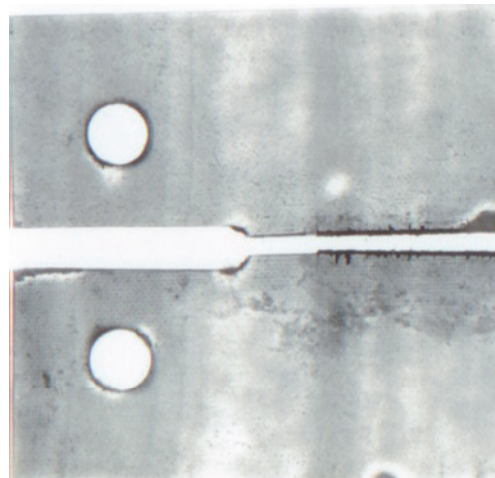


Figure 3.7. X-ray of tested specimen. No damage can be observed around the specimen holes.

Fracture toughness data was obtained from the MCC data using both the optically measured crack length, a_{vis} , and the effective crack length, a_{eff} , calculated from the measured elastic compliance of the specimen using equation (3.8). A comparison of these crack lengths is shown in Figure 3.9.

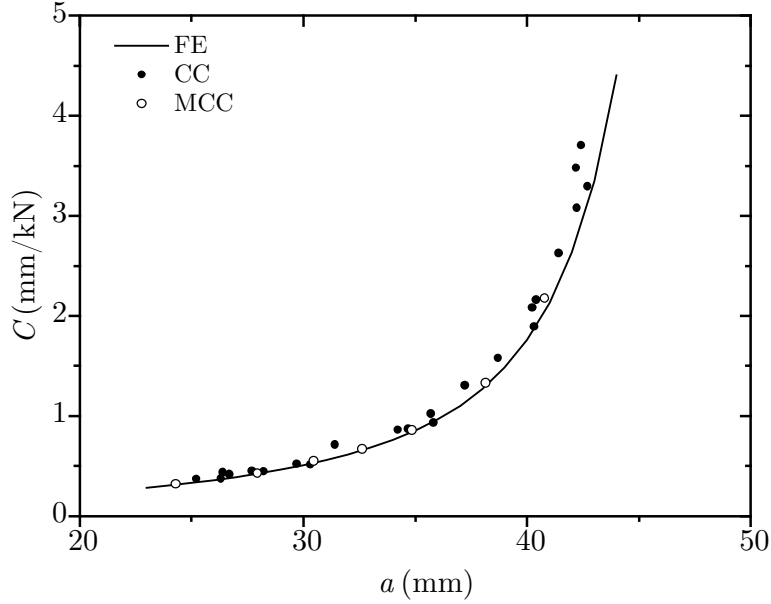


Figure 3.8. Compliance vs. crack length curves obtained from FE, CC method and MCC method.

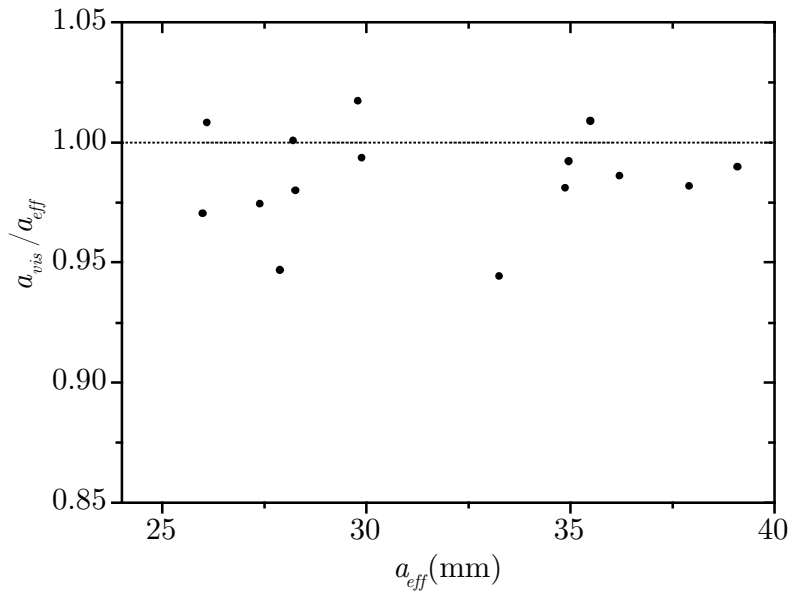


Figure 3.9. Comparison of crack lengths: optically measured a_{vis} and effective a_{eff} from MCC method.

The R-curves obtained using the 6 methods of data reduction are shown in Figure 3.10(a)-(f); a comparison of the measured initiation and propagation values of critical strain energy release rate are presented in Figure 3.11, also shown are the coefficients of variation, c_v . In the case of toughness values obtained via the area method, the initiation value shown in Figure 3.11(a) corresponds to the first toughness measurement. Due to the lack of a strong R-curve effect and lack of data points in the region where the R-curve may be rising, propagation values for all data reduction schemes were defined simply as any toughness value obtained for crack lengths greater than 0 mm, it is the average of these values presented in Figure 3.11(b).

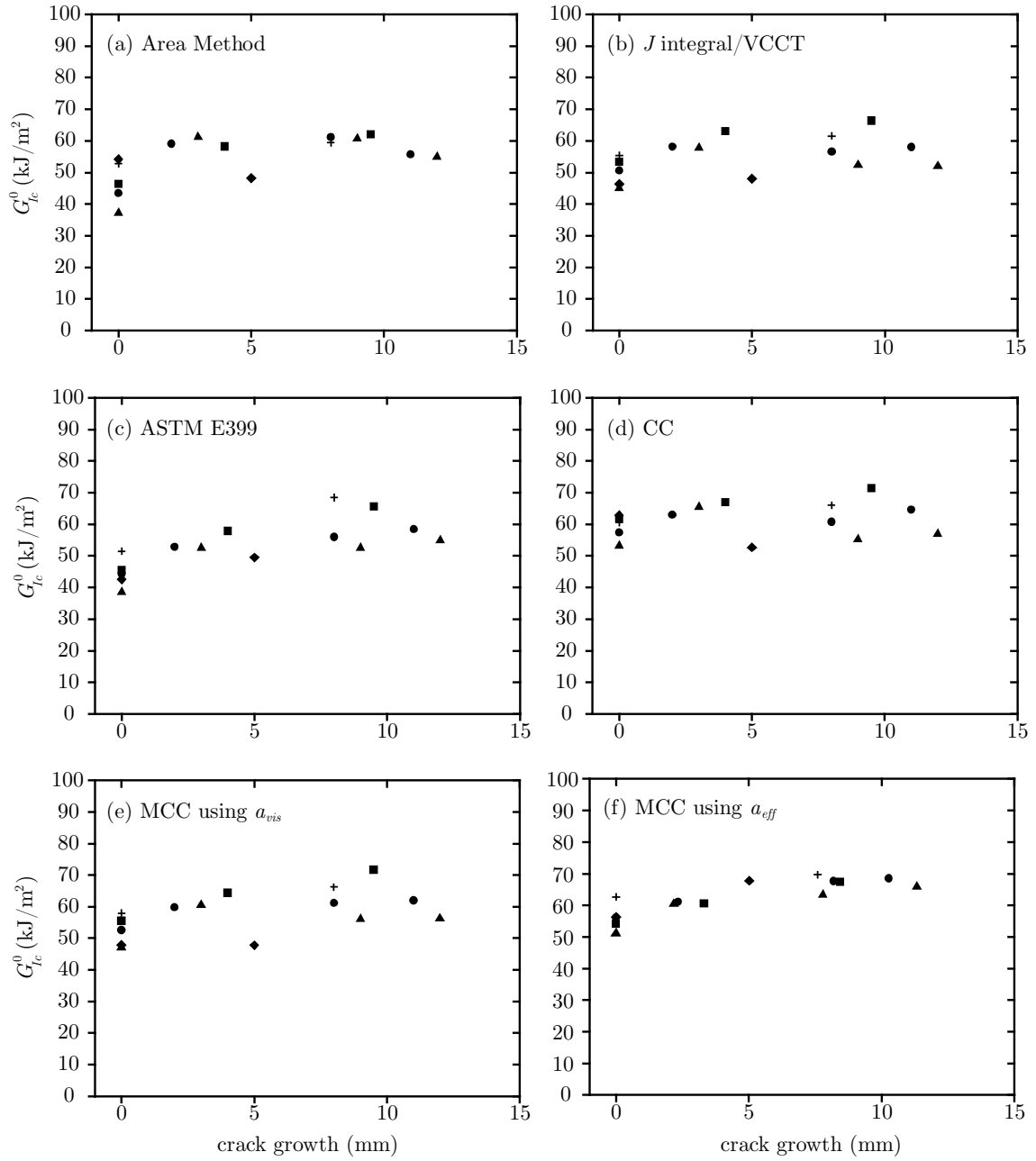


Figure 3.10. R-curves obtained from CT testing: (a) Area method, (b) J-integral/VCCT, (c) ASTM E399, (d) ECC, (e) MECC using a_{vis} , and (f) MECC using a_{eff} .

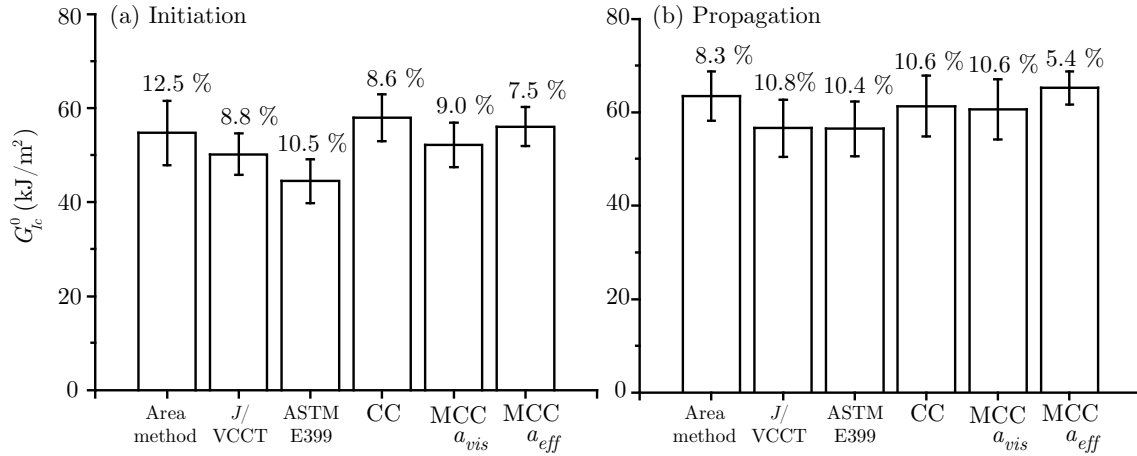


Figure 3.11. Comparison of strain energy release rates obtained via different data reduction schemes (coefficients of variation are provided in %) (a) initiation and (b) propagation.

3.2.5. Discussion

Six data reduction schemes have been presented to determine the critical strain energy release rate at fracture initiation and propagation. One notable feature of all the results is the significant scatter; it is believed that a major contribution to this (apart from the case of MCC using a_{eff}) lies in the use of an optically measured crack length. As previously mentioned, crack growth was tracked and measured using a mm scale drawn onto the specimen. The accuracy attainable using this method is only within ± 0.5 mm. The dC/da vs. a response obtained via FE for the CT specimen geometry used in this study is shown in Figure 3.12 ($w = 65$ mm), as can be seen the curve becomes steep at higher values of crack length which indicates the sensitivity of data reduction schemes to crack length measurement. This error is quantified in Figure 3.13, which shows the possible error in dC/da , obtained by FE, introduced through crack length measurements made using a measurement resolution of ± 0.5 mm. No measurements were taken for $a > 40$ mm where the rate of change of dC/da becomes large, and also due to the proximity of the crack tip to the back edge of the specimen.

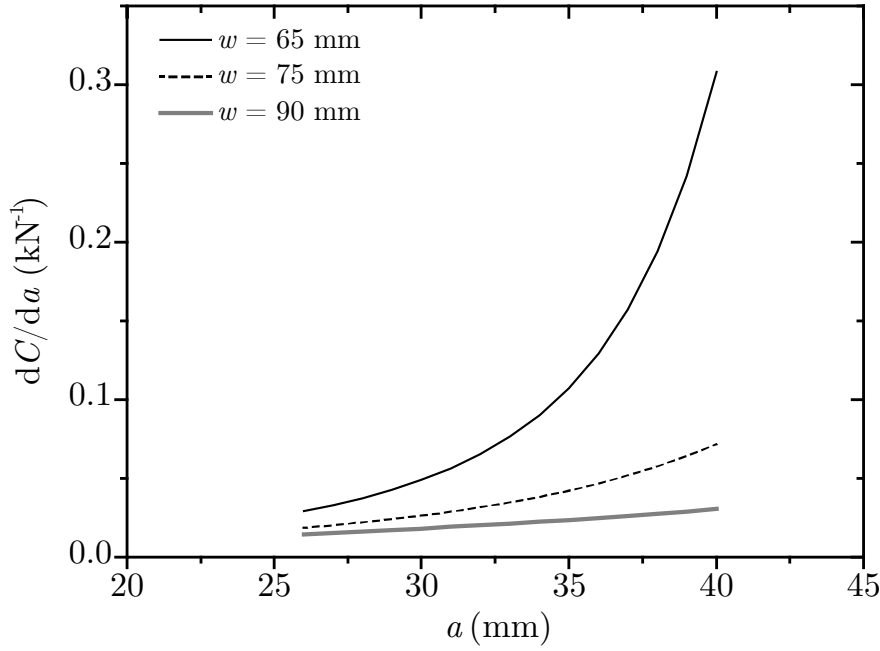


Figure 3.12. dC/da vs. a curve for CT specimen with various values of w obtained from FE.

The area method exhibits the largest amount of scatter for fracture initiation. Initial crack growth increments exhibited by the specimens were measured to be between 1 and 8 mm; the sensitivity of the area method to Δa means that the method is not suited to the use of an optically measured crack length. Despite this, the propagation value for critical strain energy release rate has the lowest c_v of the methods; if it were not for the poor correlation of initiation values, the agreement of propagation results coupled with the method's simplicity would make it the ideal method for measurement of G_{Ic} .

It is thought that the J-integral/VCCT method gives accurate results under the used test conditions as Figure 3.8 shows the excellent agreement between the experiment and FE. This agreement suggests that the use of a_{vis} contributes significantly to the resulting scatter in the propagation values of critical strain energy release rate, and that the results obtained should be very similar to those when using the MCC method with a_{vis} . Although the J-integral/VCCT method can produce as consistent results as possible whilst using a_{vis} , it does involve the added complexity of having to use FE in the data reduction process. For this reason it would be deemed unsuitable for any standard testing procedures.

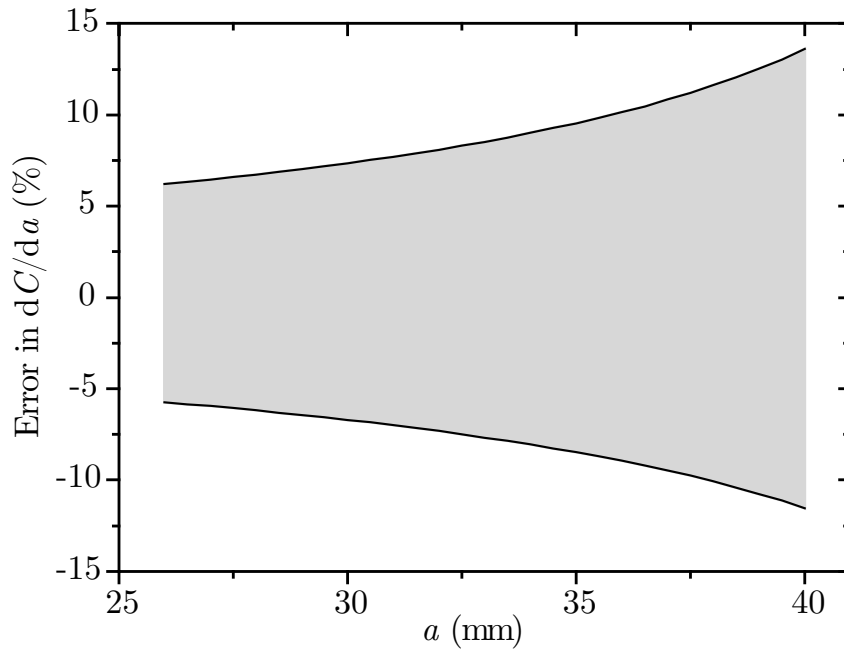


Figure 3.13. Possible error in dC/da with increasing a for ± 0.5 mm crack measurement resolution used in this study: the shaded region indicates the possible error.

As previously mentioned, the method outlined in ASTM E399 is only strictly valid for isotropic materials, and the errors introduced by its application to orthotropic laminates are illustrated in Figure 3.4. By examining the figure, it is unsurprising to discover that the measured initiation value has the lowest average out of all the methods being investigated, as the crack length at initiation corresponds to $a/w = 0.51$ —the point on the plot with the maximum error. When looking at Figure 3.11(b), the good agreement of the results with the other data reduction schemes is noticeable. Again this is unsurprising; it would be expected that fracture toughness measurements taken over the a/w range used in these tests, for this particular lay-up, would average out to be near the correct value, as the curve in Figure 3.4 indicates. This would not be the case for more orthotropic lay-ups with higher proportions of 0° or 90° plies. If only initiation values of toughness were required, then initial crack length, a_0 , shown in Figure 3.1, could be adjusted to produce accurate results (see Figure 3.4).

Using the CC method, results were calculated using the compliances measured at each optically measured crack length. The compliance versus crack

length behaviour is shown in Figure 3.8. It can be seen that for $a < 40$ mm there is a very good agreement between the C values measured directly from the load-displacement curve, and those obtained via FE and MCC methods. For $a > 40$ mm the results become more scattered, which is a result of the sensitivity of the optically measured crack length to errors in the region where dC/da increases sharply (see Figure 3.12), and the fact that the measured compliances were taken over small ranges on the load displacement curve (see Figure 3.6). For tests requiring measurements of compliance, unloading the specimen after crack propagation can lead to more accurate results as the measurement can be taken over a wider range. However, for the testing of this type of material system, this method is not recommended as the unloading/reloading curve will be influenced by pulled-out fibres contacting the crack faces during crack closure.

Values of critical strain energy release rate were obtained from the MCC data using 2 approaches; the first uses the optically measured crack length, a_{vis} , and the second uses an effective crack length, a_{eff} , inferred from the compliance measured directly from the load vs. displacement curve obtained from testing and equation (3.8). Figure 3.9 shows how the measured a_{vis} and inferred a_{eff} values compare. Generally the correlation is very good considering that the accuracy achievable by optically measuring the crack length is ± 0.5 mm. Clearly the use of a_{eff} is more reliable, as measurements of a_{vis} taken from the unconstrained outer 90° layers will not be perfectly representative of the damage through the entire thickness of the specimen.

The R-curves obtained using the two approaches of crack length measurement are shown in Figure 3.10(e) and (f). There are distinct differences between the two plots. Firstly, the scatter at initiation is lower when using a_{eff} . This is believed to be a result of the accuracy of the MCC method in predicting the C vs. a behaviour of the specimens, and the initial compliance of the specimens being measured over a wider range on the load vs. displacement curve. The advantage of obtaining C measurements over wider ranges on the load vs. displacement curve is reflected in the difference in the c_v values for the toughness at initiation and propagation when using a_{eff} .

The R-curve produced using a_{vis} is qualitatively very similar to those produced by the other data reduction schemes, where the general trend shows increasing fracture toughness after initiation and a constant value as the crack length increases. When using a_{eff} , the R-curve is again very similar; however, the data points are shifted due to the differences between a_{vis} and a_{eff} . Overall, the MCC method using a_{eff} gives the most consistent results of the data reduction schemes.

A comparison of the experimental scatter in the propagation values produced by the method using a_{eff} suggests that the method is almost as equally susceptible to errors in compliance measured from the load vs. displacement curves as the other methods are to errors in the measured a_{vis} . The accuracy of this method could be improved by obtaining more data points on the MCC curve, especially in the region where dC/da increases rapidly (see Figure 3.8). It is thought that this method poses the ideal data reduction solution for the current specimen geometry, as it eliminates the need for an optically measured crack length and use of FE.

The advantages and disadvantages of each data reduction method have been summarised in Table 3.2. As has been already mentioned, the MCC method using a_{eff} inferred from specimen compliance is considered to be the most appropriate data reduction scheme to use. It has been shown to experimentally measure C vs. a response accurately, and also eliminates the need to rely on an optically measured crack length. The downfall of this method is that it is laborious, requiring a specimen to be machined and tested several times to capture the full dC/da behaviour, and for further improvement of results this process should be performed on several specimens.

Although a most appropriate data reduction scheme was identified, there is still an undesirable amount of scatter within the results. Eliminating the area method due to excessive scatter at initiation, the J-integral/VCCT method due to the necessity of FE and ASTM E399 for reasons already discussed, the CC is left as a possible scheme. Both the CC and MCC methods require compliance measurements to be taken from the load vs. displacement curve obtained from tests, and it is assumed that this is the major contributing factor to scatter in the MCC

results using a_{eff} as the use of an optical crack length has been eliminated, and that the toughness values are more consistent for initiation where C measurements were taken over wide ranges of the load vs. displacement curve. Compliance measurements were made at the best available accuracy; however, after fracture initiation they were sometimes taken over small regions of the load vs. displacement curve as shown in Figure 3.6. The inability to unload and reload specimens makes this contribution to error unavoidable. For the data reduction schemes requiring an optically measured crack length, a significant amount of the scatter is due to errors in its measurement. The effect of errors in a_{vis} on G_{Ic} is related to the dependence of dC/da upon a , as shown in Figure 3.12 and Figure 3.13. While it is believed that the accuracy of compliance measurement cannot be easily improved, the effect of any error resulting from an optically measured crack length could be minimised by reducing the rate of change of dC/da as the crack length increases. The curves shown in Figure 3.12 indicate that this could be achieved by widening the specimen by increasing the dimension w in Figure 3.1; by selecting a suitable value of w it is envisaged that the CC method could pose as the best data reduction scheme. Its advantages are that it could potentially obtain accurate C vs. a response for several specimens (i.e. total number of test specimens) without any increase in testing time and without the use of any extra material.

Finally, it should be noted that the laminates tested during this study were chosen so that the in situ ply toughness associated with mode I fibre tensile failure could be extracted from the laminate toughness given assumptions about the damage in non-zero degree plies. For fracture toughness tests of multidirectional laminates it has been observed that a large damage zone (made up of delaminations, longitudinal splits and fibre breaks) can be formed [21]. In contrast to [90/0] specimens, the fracture is not observed to be self-similar at the macroscale because the damage zone increases in size as the crack tip advances. Given the size of this damage zone relative to the specimen it is clear that characterisation of a propagation property lies outside the realm of linear elastic fracture mechanics thus the discussion presented here cannot be directly extrapolated to laminates containing angle-ply. Measurement of the translaminar fracture toughness for

initiation of multidirectional laminates has already been addressed in the ASTM E1922 standard [42].

Table 3.2. Summary of the advantages and disadvantages of data reduction methods.

Method	Advantages	Disadvantages
Area	Simplicity	Must use optically measured crack growth, Δa . The result of equation (4) becomes more sensitive to errors for low Δa and insensitive to R-curve effects for large Δa . For stick-slip crack growth, the method requires interpolating P vs. d between P_1 and P_2 as shown in Figure 3.5.
ASTM E399	Initial (a/w) (see Figure 1.) can be chosen to give accurate initiation values.	Relies on optically measured crack length. Use of $f(a/w)$ for isotropic materials inappropriate.
J-integral / VCCT	Eliminates any error from differentiating a fitted curve such as that in the CC and MCC methods.	Relies on optically measured crack length. Use of FE adds complexity.
CC	Works well if C vs. a curve can be sufficiently populated.	Relies on optically measured crack length. Points for compliance calibration curve are dictated by specimen crack jumps, may result in areas of curve not being populated.
MCC	Does not require optical crack length measurement.	Requires additional experimental work or adds the complexity of using FE to obtain C vs. a .

3.2.6. Conclusions

Six methods of data reduction for CT testing have been investigated with the aim of finding the ideal strategy: the area method, J-Integral/VCCT, ASTM E399, compliance calibration, and modified compliance calibration (using a_{eff} and a_{vis}). Cross-ply laminates were tested and the results for all data reduction schemes were fairly consistent, a modified experimental compliance calibration (MCC) method, using an effective crack length determined through compliance measurements, was deemed to be the most appropriate data reduction scheme, as it gives consistent results and does not require the use of an optically measured crack length or FE.

An undesirable amount of scatter (up to 10.2 % for the MCC method) was still present within the results. This can be improved upon by changing the proportions of the specimen, namely by increasing the width.

3.3. Investigation of Size and Lay-up Effects

Much work has been performed in measuring the fracture toughness of composite laminates using the compact tension (CT) specimen [21, 24, 25, 34], though very little has been done to investigate specimen size or lay-up effects.

Harris and Morris [21] used the centre-crack tension specimen with a $[0/90]_{ns}$ configuration to investigate the effects of thickness on the measured laminate critical stress intensity factor, K_{Ic} , for fracture initiation. Tested specimens were between 8 and 120 plies thick; it was found that initially the measured K_{Ic} decreased with increasing thickness, reaching a plateau value at thicknesses above 64 plies. The higher apparent toughness at lower laminate thicknesses was attributed to longitudinal splitting, observed to occur in the outer 0° plies, affecting the intensity of the singularity at the crack tip and the effective fracture toughness. Interior 0° plies, being constrained by a 90° ply on either side, were not observed to split extensively; thus the effect of longitudinal splitting was deemed to be a surface phenomenon, becoming less influential as the laminate thickness increased.

Prewo [33] investigated the effect of ply lay-up sequence on the measured toughness for various cross-ply Boron-Aluminium CT specimens. By measuring K_{Ic} for fracture initiation, it was found that the blocking together of plies, i.e. for a $[90_6/0_6]$ laminate, resulted in a 20% increase in toughness over the alternating cross-ply ($[90/0]_{ns}$) sequence. Study of the failed specimen fracture surfaces using scanning electron microscopy (SEM) revealed significantly more fibre pull-out in the specimens with blocked plies. It was thought that increasing the thickness of the 0° layer reduces the ability of the 90° plies to constrain the crack to a single plane.

The work presented in this section investigates factors affecting the accurate measurement of the fracture toughness associated with the in situ fracture toughness of the 0° plies within a laminate. In section 3.2. various methods of data reduction were assessed and the most appropriate scheme determined. This section uses the proposed data reduction scheme to investigate specimen size and lay-up effects on the measured initiation and propagation values of fracture toughness associated with 0° ply-level fracture toughness.

3.3.1. Materials and Manufacture

The material system used for the work presented in this section and methods of specimen preparation were outlined in section 3.2.1. of this chapter. The geometry of the baseline specimen, CT-base, is shown in Figure 3.1. The lay-ups and geometries of the tested CT specimens are outlined in Table 3.3.

In this study, size effect is investigated by scaling specimen dimensions in accordance with the definition given by Bažant in [60], similarly to the experimental program presented in [61]. Specimen in-plane size effect was studied by using CT-50% and CT-100%, where the dimensions l and h (see Figure 3.1) were scaled as shown in Table 3.3 so that the areas ($l \times h$) were increased by 50% and 100% with respect to the baseline. For these specimens the dimensions of the notch and the location of the holes were also scaled. The effect of thickness size effect is investigated by using specimens CT-base and CT-thick. The effects of lay-up, namely blocking together of 90° and 0° plies, are investigated by using specimens CT-base, CT- 90° , and CT- 0° .

Table 3.3. CT specimen configurations investigated.

Specimen code	Lay-up	l (mm)	h (mm)
CT-base	$[(90/0)_8/90]_s$	65	60
CT-50%	$[(90/0)_8/90]_s$	80	74
CT-100%	$[(90/0)_8/90]_s$	92	84
CT- 90°	$[(90_2/0)_8/90]_s$	65	60
CT- 0°	$[(90/0_2)_8/90]_s$	65	60
CT-thick	$[(90/0)_{16}/90]_s$	65	60

3.3.2. Test method and Experimental Setup

Five specimens were tested for the baseline configuration (CT-base) and six specimens were tested for each of the other CT configurations shown in Table 3.3.

Each specimen was tested under displacement control at a rate of 0.5 mm/min using an Instron machine with a 10 kN load cell. Measurements of load and crosshead displacement were recorded using a data logger. After the test, the values of crosshead displacement were corrected to account for the compliance of the test machine and fixtures, which has been measured as 0.06 mm/kN.

3.3.3. Data Reduction

Section 3.2 of this chapter investigated several data reduction schemes. A modified compliance calibration (MCC) technique, described in section 3.2.3.4., was proposed as the preferred method as it simultaneously gives results with low scatter, eliminates the need to measure the crack length optically (important to reduce operator-dependence), and eliminates the need for finite element (FE) analysis for the data reduction (important for industrial use).

The method of calibration requires the elastic compliance of a CT specimen to be determined at several machined cracks of known length. For the baseline specimen (CT-base in Table 3.3) tested in Section 3.2, Figure 3.8 shows that the experimentally-obtained compliance calibration curve agrees well with that predicted by the corresponding FE model.

Therefore, obtaining compliance vs. crack length data experimentally or numerically is just a matter of convenience, as both agree very well. In this section, numerically obtained data was used, as results could easily be obtained by modifying the dimensions and laminate elastic properties of the FE model shown in Figure 3.3. The numerically obtained compliance calibration curves for the specimen configurations used in this study are shown in Figure 3.14.

Once the laminate fracture toughnesses had been obtained using the MCC method, the ply-level fracture toughnesses were obtained as outlined in section 3.2.3.6. of this chapter.

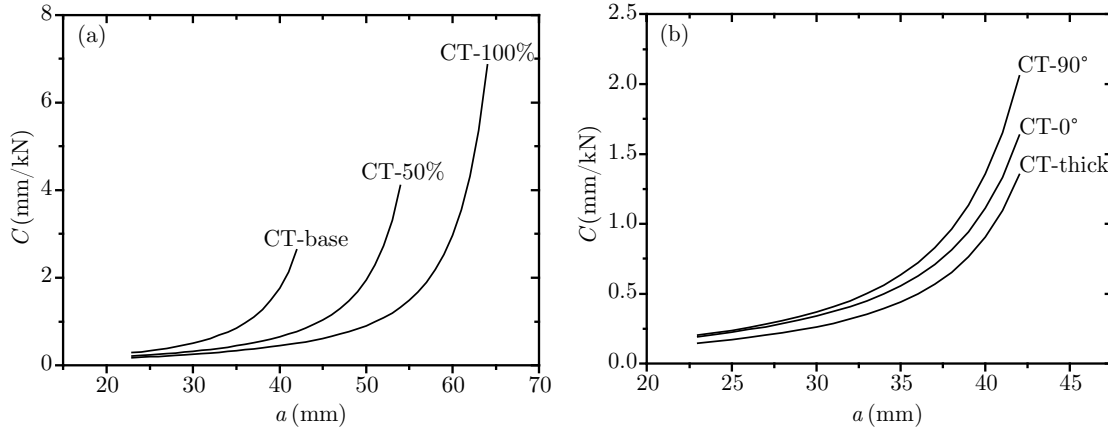


Figure 3.14. Numerically obtained compliance vs. crack length data for (a) CT-base, CT-50%, CT-100% and (b) CT-90°, CT-0°, and CT-thick.

3.3.4. Results

Typical load-displacement curves obtained from the various configurations of CT specimen are shown in Figure 3.15(a)-(f); as can be seen all specimens exhibit stick-slip crack growth.

The R-curves obtained from each specimen configuration are shown in Figure 3.16(a)-(f), and it can be seen that only configurations CT-0° and CT-thick exhibited an increasing trend notable over a few millimetres of crack growth. For comparative purposes, the measured initiation and propagation values of critical strain energy release rate are presented in Figure 3.17(a) and (b) respectively. In the case of CT-0° and CT-thick, propagation values are defined by the plateau reached in Figure 3.16(e) and (f) respectively, for all other specimen configurations the propagation toughness was defined simply as the average of the toughness values obtained for crack extensions greater than 0 mm, the average of these values is presented in Figure 3.17(b).

Typical scanning electron microscopy (SEM) images of the fracture surfaces from tested specimens are shown in Figure 3.18 - Figure 3.23.

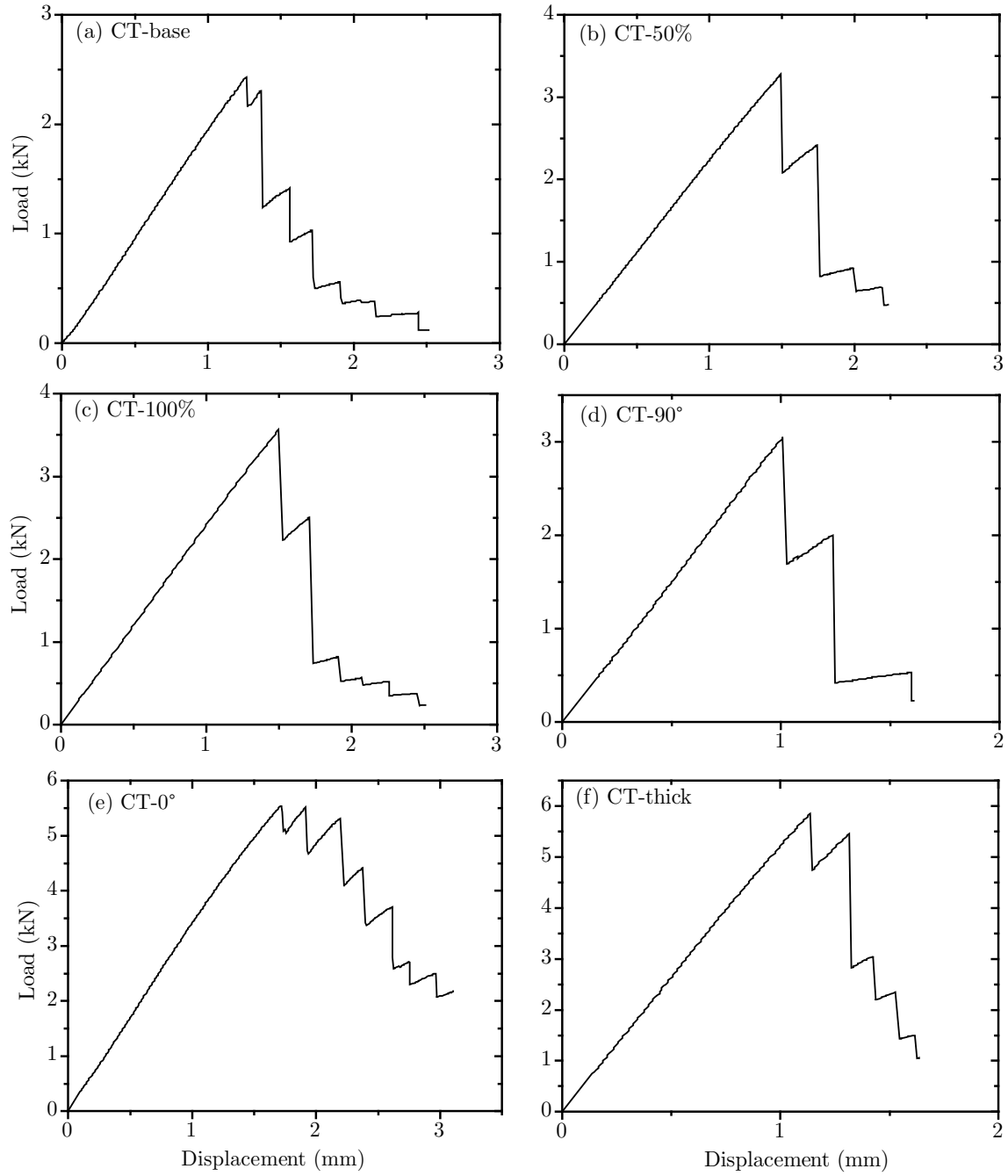


Figure 3.15 Representative load vs. displacement curves for (a) CT-base, (b) CT-50%, (c) CT-100%, (d) CT-90°, (e) CT-0°, and (f) CT-thick.

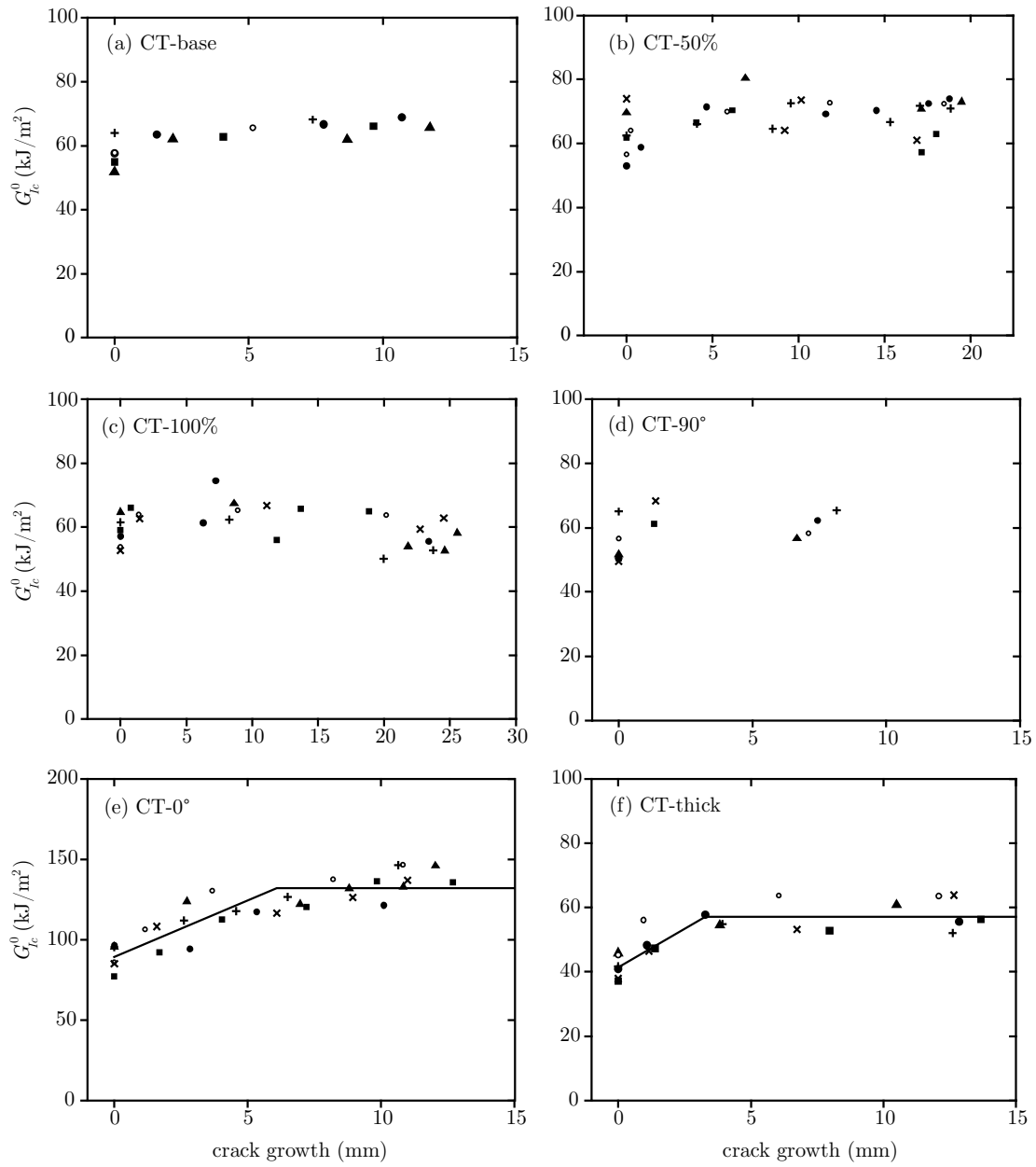


Figure 3.16. R-curves obtained from CT testing: (a) CT-base, (b) CT-50%, (c) CT-100%, (d) CT-90°, (e) CT-0°, and (f) CT-thick. Each symbol represents a different specimen.

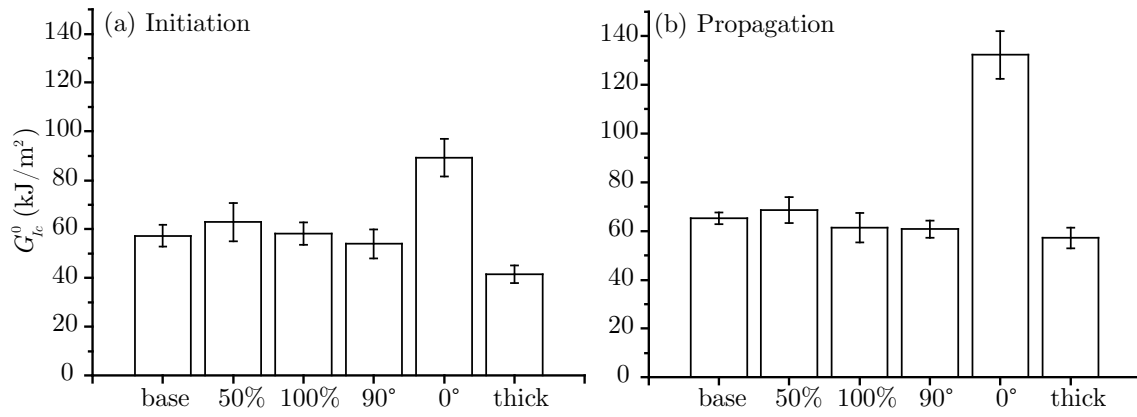


Figure 3.17. Comparison of critical strain energy release rates obtained for different specimens: (a) initiation, (b) propagation.

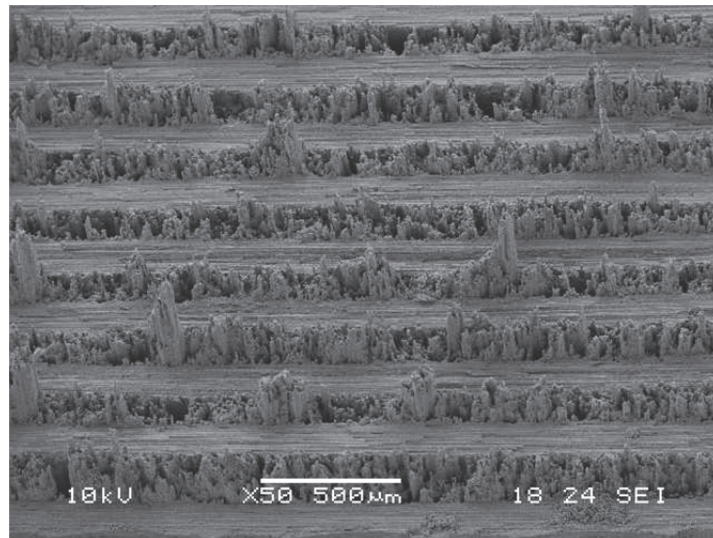


Figure 3.18. Fracture surface of [(90/0)_s / 90]_s (CT-base) laminate. Direction of crack propagation is right-to-left in the image.

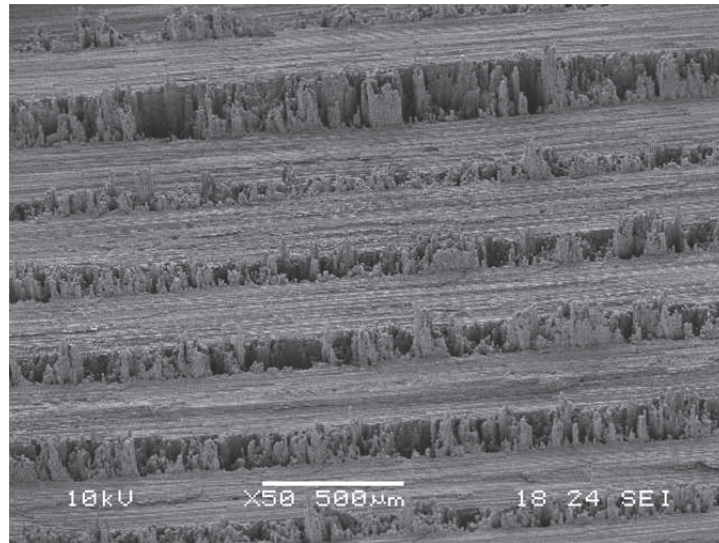


Figure 3.19. Fracture surface of $[(90_2/0)_8/90]_s$ (CT-90°) laminate. Direction of crack propagation is left-to-right in the image.

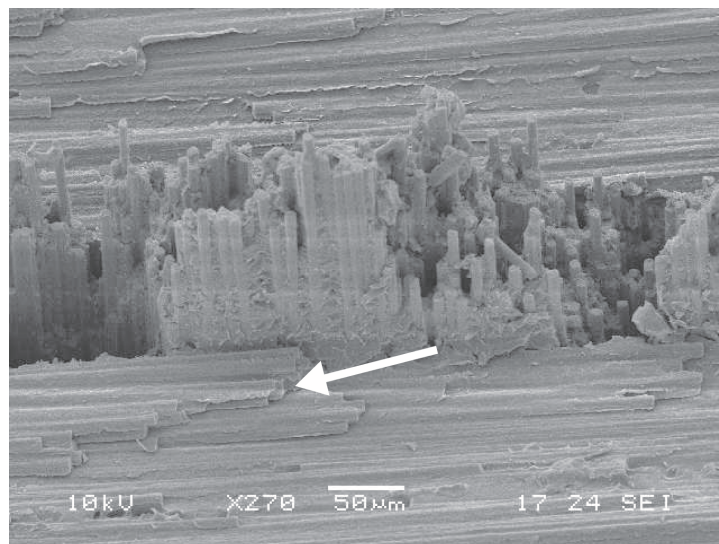


Figure 3.20. Close up of Fracture surface of $[(90_2/0)_8/90]_s$ (CT-90°) laminate. Broken fibres in the 90° plies are highlighted. Direction of crack propagation is left-to-right in the image.

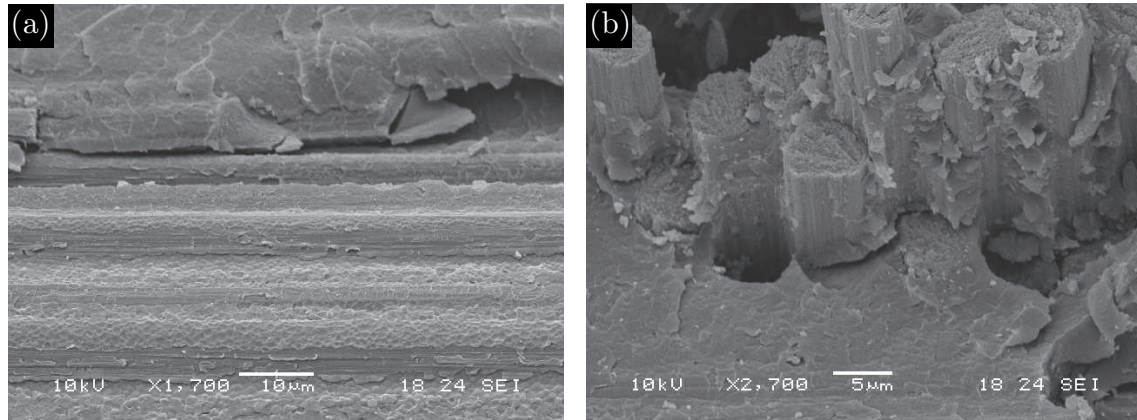


Figure 3.21. Fracture surface at 90/0 interface of $[(90/0)_s/90]_s$ (CT-base) laminate (a) close to 90° layer (b) first broken fibres at interface. Direction of crack propagation is right-to-left in the images.

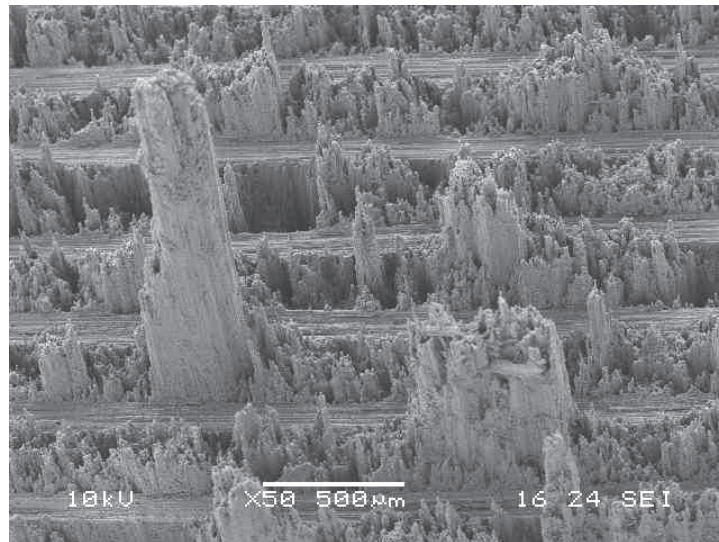


Figure 3.22. Fracture surface of $[(90/0_2)_s/90]_s$ (CT-0°) laminate. Direction of crack propagation is right-to-left in the image.

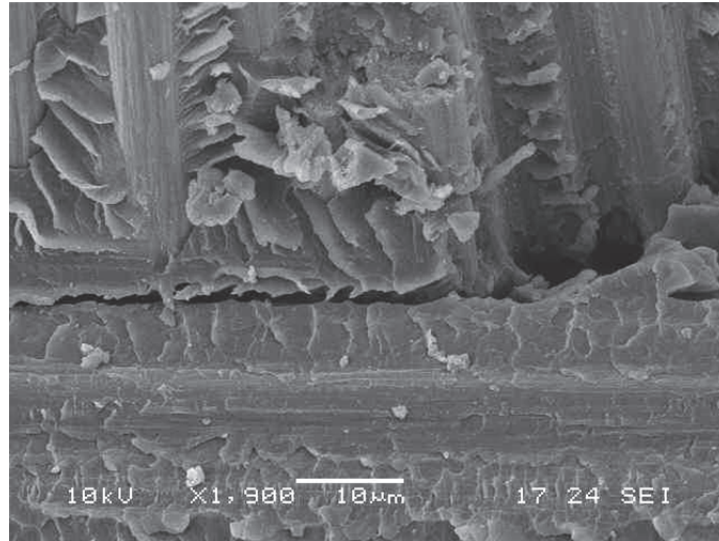


Figure 3.23. Fracture surface at 90/0 interface of $[(90/0_2)/90]_s$ (CT-0°) laminate. Direction of crack propagation is right-to-left in the image.

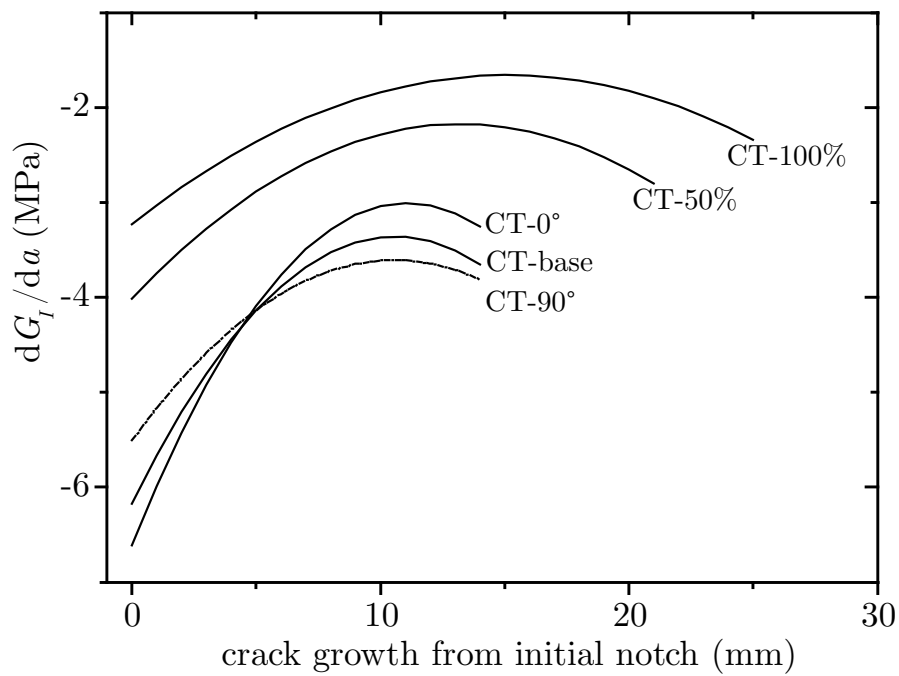


Figure 3.24. dG_I/da vs. a under load-control obtained from FE. Models were run for 1mm crack length intervals.

3.3.5. Discussion

3.3.5.1. In-plane size effects

From the results presented for configurations CT-base, CT-50% and CT-100% in Figure 3.16(a)-(c), respectively, it can be inferred that there is no significant effect of scaling the specimen size.

Qualitatively, the R-curves for CT-base, CT-50% and CT-100% are similar, though the amount of scatter present at initiation for CT-50% (Figure 3.16(b)) is noticeable and probably reflects the quality of the prepared crack tips. Another feature apparent from the R-curves is that the scaled specimens (CT-50% and CT-100%) show noticeably larger crack jumps relative to CT-base. For all CT tests presented within this study, the crack growth increment is determined by two things; the stability of the specimen (i.e. dG_I/da) and material variability (i.e. dG_{Ic}/da). For each of the specimen configurations dG_I/da is negative under displacement control (see Figure 3.24) and thus stable crack growth should occur. However, the variation of G_{Ic} with crack length is unknown, but this factor, combined with the magnitude of dG_I/da determines the resulting crack growth increment.

As can be seen from Figure 3.24, dG_I/da for a given crack length (under displacement control) becomes less negative as the specimen is scaled up in size and as a result, generally larger crack growths are observed in the resulting R-curves.

The baseline geometry, shown in Figure 3.1 and the largest specimen (CT-100%), with the area $l \times h$ increased by 100%, could be representative of the lower and upper boundaries of practical specimen size. Specimens significantly smaller than the baseline configuration could result in undesirable failure modes (e.g. shear out of the loading holes) and too few points on the R-curve due to a small crack growth range. Given that the specimen size has negligible effect on results, a specimen larger than CT-100% would seem impractical in terms of material usage, though not unfeasible.

3.3.5.2. Thickness size effect

The effects of thickness can be assessed by comparing the results from specimens CT-base (34 plies) and CT-thick (66 plies). The measured initiation and propagation values of G_{Ic}^0 for the thicker CT-thick specimen were the lowest of the tested specimen configurations. A similar phenomenon has previously been observed by Harris and Morris [21] where the toughness measured at initiation was seen to asymptote towards a lower bound value for thicker laminates. However, the study used a cross-ply laminate with 0° outer plies, extensive splitting was observed in these plies, and this contributed to the observed thickness effect. Some longitudinal splitting and delamination was still observed in embedded 0° plies. In this present study, a cross-ply lay-up is used, however, with 90° outer plies, thus avoiding the problems encountered with the outer plies in [21]. The lower toughness values obtained from the thicker (CT-thick) laminate could be attributed to a similar effect if longitudinal splitting in the 0° plies near the surface significantly affected the stress intensity at the notch-tip. However, longitudinal splitting in the outer 0° plies should be evident in the form of longer fibre pull-out lengths in these plies. SEM images show that there is not any significant difference between the lengths of pulled-out fibres from ply-to-ply within the laminate and that the fracture surfaces of CT-base and CT-thick are indistinguishable. The observed differences in results between CT-base and CT-thick are 28% at initiation and 14% for propagation, and to make any firm conclusion as to the source of this difference would be inappropriate without further investigation.

It is also interesting to note the increasing R-curve trend shown in Figure 3.16(f); the reason for this feature is unclear as investigation using SEM revealed fracture surfaces indistinguishable to those of the CT-base specimen shown in Figure 3.18.

3.3.5.3. Lay-up effects

3.3.5.3.1. Blocking of 90° plies

The results shown in Figure 3.17 indicate that blocking 90° plies (compare CT-base and CT-90°) has no effect on the measured toughness values at initiation and propagation. Similarly to CT-100%, a noticeable feature of the laminates with blocked 90° plies (CT-90°) is the relatively large crack jumps when compared to other laminates; this feature is apparent on the load vs. displacement curve shown in Figure 3.15(d) and in the R-curve shown in Figure 3.16(d) and, similarly to specimens CT-50% and CT-100%, is attributed to the stability of the specimen, as shown in Figure 3.24, and material variability. As a result of this, the propagation toughness is the average of relatively few values when compared to other laminates (see Figure 3.16). Figure 3.18 and Figure 3.19 show representative fracture surface images taken using scanning electron microscopy from $[(90/0)_8/90]_s$ (CT-base) and $[(90_2/0)_8/90]_s$ (CT-90°) specimens respectively. The images are qualitatively similar, with some degree of fibre pull-out, indicating that the measured toughness of the 0° plies should be the same in each case.

Broken fibres can also be seen in the 90° layers in both Figure 3.18 and Figure 3.19, indicating that fibre bridging has occurred in these plies, and is a toughening mechanism contributing to the laminate toughness. Because mode I matrix failure only is assumed in the 90° plies, the effects of this bridging phenomenon also contribute towards the toughness values in the G_{Ic}^0 R-curves shown in Figure 3.16. This feature was observed to be more prominent in the CT-90° specimens, where small misalignments between the blocked 90° plies occur, as illustrated SEM image in Figure 3.20. The contribution of this mechanism to the R-curve is however negligible: if this were not the case, it would be expected that the measured G_{Ic}^0 in the case of CT-90° would be higher than for CT-base.

3.3.5.3.2. Blocking of 0° plies

The blocking of 0° plies is investigated by comparing specimens CT-base and CT- 0° . The R-curve obtained from the CT- 0° specimens is shown in Figure 3.16(e); the plot starkly contrasts the R-curves obtained from testing of the other configurations. There is an increased number of data points relative to the other specimen configurations tested; due to the issues of specimen stability (see Figure 3.24) and material variability, CT- 0° specimens typically exhibited several short crack growths before complete specimen failure, as illustrated in Figure 3.15(e).

A large increase in toughness is seen by blocking 0° plies together; by increasing the thickness of the 0° layer, as in the case of CT- 0° , the apparent propagation toughness of the 0° plies is almost 100% greater than in the case of the baseline configuration. This fact suggests that the toughness associated with the fibre tensile failure mode is not simply a material property, but rather an in situ material property.

The reasons for these differences can be explained using fractography. Figure 3.21 shows SEM images taken at the $90/0$ ply interface of a $[(90/0)_8/90]_s$ laminate; river-lines in the resin-rich region of Figure 3.21(a) indicate that fracture initiates and propagates from the 90° plies towards the 0° plies. Figure 3.21(b) shows propagation into the 0° layer where fibre fracture initially occurs in the plane of the 90° ply before cracks propagate longitudinally along the fibres, resulting in additional energy being dissipated during fibre pull-out. The lengths of pulled-out fibres increase as the crack propagates towards the centre of the 0° plies. This means that the thinner the 0° layer, the less fibre pull-out and thus the lower the apparent toughness.

Figure 3.22 shows the fracture surface of a laminate with blocked 0° plies (CT- 0°). A much higher degree of fibre pull-out is seen than for the baseline laminate in Figure 3.18, with pulled-out fibres being up to millimetres in length, absorbing much more energy during the fracture process. The larger pulled-out fibres indicate more extensive splitting and therefore increased crack-tip blunting.

Figure 3.23 shows a SEM picture of the 90/0 ply interface; again it can be seen that river lines indicate that the crack has progressed from the 90° ply. In contrast to the un-blocked case there is no fibre fracture in the plane of the 90° ply, instead, mode II features associated with the pull-out process can be observed on the residual matrix left on the block of pulled-out fibres.

3.3.6. Conclusions

This study has investigated the size and lay-up effects on the measured fracture toughness associated with mode I fibre tensile failure in FRP using the compact tension specimen. Scaling specimen in-plane dimensions had no significant effect on measured 0° ply toughness. Doubling the specimen thickness, by doubling the number of [90/0] sub-laminates, was found to slightly decrease the toughnesses measured at initiation and propagation, though no firm conclusion was made as to the cause of this.

Lay-up was investigated by testing laminates with blocked 0° and 90° plies; the apparent toughness of the fibre tensile failure mode was found to increase when the thickness of the 0° layer was increased. The increase in toughness was attributed to an increase in fibre pull-out for blocked 0° specimens, as observed by SEM imaging.

3.4. Investigation of notch tip radius

3.4.1. Introduction

In practice, it is not possible to manufacture specimens for fracture toughness testing with pre-cracks that are as sharp as assumed in classical linear elastic fracture mechanics. For translaminar fracture toughness measurements of FRP, crack tip radii typically between $\rho = 10 \text{ }\mu\text{m}$ [54] and $\rho = 400 \text{ }\mu\text{m}$ [23] have been used. The ASTM standard for translaminar testing of multidirectional laminates [42] suggests the use of any tool that produces a narrow slit; a diamond impregnated copper slitting saw and jeweller's saw are both quoted as suitable for this purpose. It is also stated that the profile of the notch tip is unimportant.

Given that the stress concentration ahead of a sharp crack would be much greater than that of a crack of several orders of magnitude larger, the theoretical driving force required to cause growth of a blunt crack, and subsequently apparent fracture toughness, will be much larger. In fact, true measurements of fracture toughness have been found to be dependent on some critical notch root radius, ρ_c . Toughness measurements obtained from specimens with $\rho \leq \rho_c$ will yield the true fracture toughness of the material, whereas when $\rho > \rho_c$, an apparent toughness is measured which is found to increase with increasing ρ . This well-known phenomenon has been widely observed and values of ρ_c have been published for a range of materials e.g. [62, 63].

Underwood et al. [47] published data obtained from testing of cross-ply carbon epoxy laminates for $\rho = 150 \text{ }\mu\text{m}$ and $\rho = 300 \text{ }\mu\text{m}$; the results showed that the toughness increased with increasing ρ . However, it was not possible to infer from the results whether or not the true toughness of the material had been measured i.e. if $150 \text{ }\mu\text{m} \leq \rho_c$. The same study showed that the measured toughness for initiation in $(0/\pm 60)$ carbon/bismaleimide laminates is constant for $\rho \leq 300 \text{ }\mu\text{m}$,

this result cannot, however, be extrapolated to a completely different material system and layup.

The experimentalist must have a priori knowledge of ρ_c for a given material in order to achieve accurate measurements of fracture toughness. In this section, the effect of ρ on the translaminar fracture toughness measured at initiation for carbon/epoxy IM7/8552 cross-ply laminates is investigated. The compact tension (CT) specimen which is selected for this purpose, has commonly been used to measure the translaminar fracture toughness of fibre reinforced composite laminates [21-23, 34] and also the in situ toughness of 0° plies [26-28]. The aim of this work is to provide a guideline on required notch tip radii for researchers wishing to measure the translaminar fracture toughness of composite laminates.

3.4.2. Materials and Manufacture

The material system used for this study was IM7/8552 unidirectional carbon/epoxy pre-preg. Panels with a lay-up of $[(90/0)_8/90]_s$ were manufactured using the hand lay-up method and cured to the pre-preg manufacturer's instructions. The panels were C-scanned as a quality check before being cut to the geometry shown in Figure 3.1 using a wet saw; the 8 mm holes were made using a carbide tipped drill. Four sizes of notch tip radius were investigated; the methods of preparation and resulting radii are summarised in Table 3.4.

For the 750 μm radius configuration, a 40 mm long notch was prepared using a bandsaw over the distance $b+c$ in Figure 3.1. For the remainder of the notch configurations, the specimen was notched to 30 mm over distance b using a diamond coated disk saw then cut a further 10 mm over the distance c with the corresponding method given in Table 3.4. Representative micrographs of the notch tips corresponding to the configurations listed in Table 3.4 are shown in Figure 3.25.

Table 3.4. Summary of CT notch configurations tested in this study.

ρ (μm)	Method of preparation	Image
15	Carbide coated disk saw	Figure 3.25(a)
125	Razor saw	Figure 3.25(b)
250	Diamond coated wire saw	Figure 3.25(c)
750	Band saw	Figure 3.25(d)

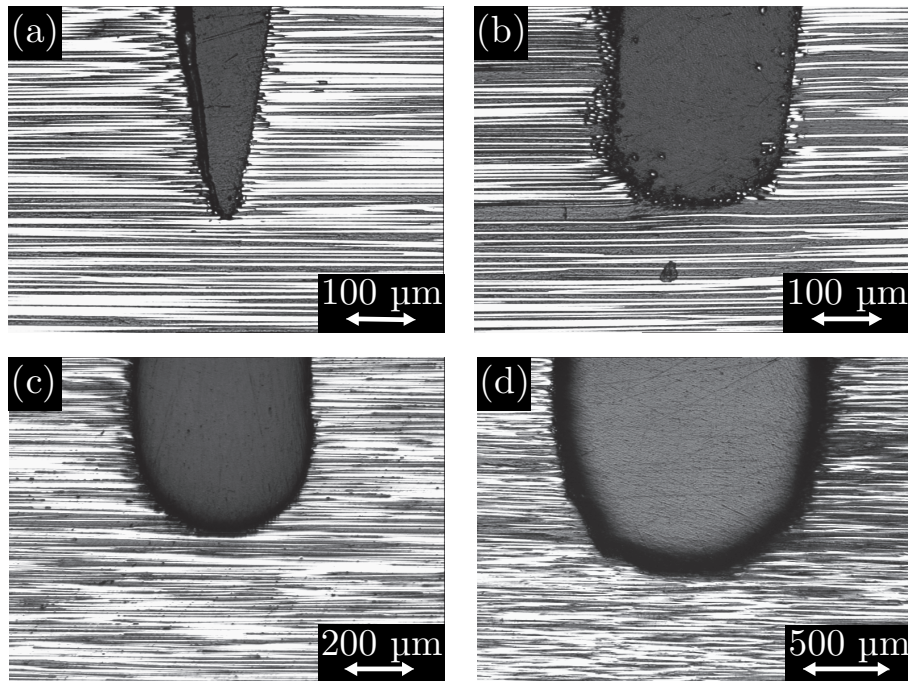


Figure 3.25. Micrographs of notch tips with radii of (a) 15 μm , (b) 125 μm , (c) 250 μm , (d) 750 μm .

3.4.3. Test method and Experimental Setup

Six specimens of each configuration were tested using an Instron machine with a 10 kN load cell. Each specimen was loaded under displacement control at a rate of 0.5 mm/min. Measurements of load and displacement were recorded using a data logger, and after the test the measurements of displacement were corrected to account for the compliance of the testing machine and fixture.

3.4.4. Data Reduction

The critical strain energy release rate of the laminate was determined using the MCC method outlined in section 3.2.3.4. of this chapter. The compliance calibration curve that was used, shown in Figure 3.26, was obtained experimentally and uses the results of 6 CT specimens.

Ply-level fracture toughness data was subsequently obtained from the laminate toughness following the method outlined in section 3.2.3.6. with the fracture toughness associated with intralaminar matrix failure for IM7/8552 taken to be $G_{Ic}^{90} = 311 \text{ J/m}^2$ [19].

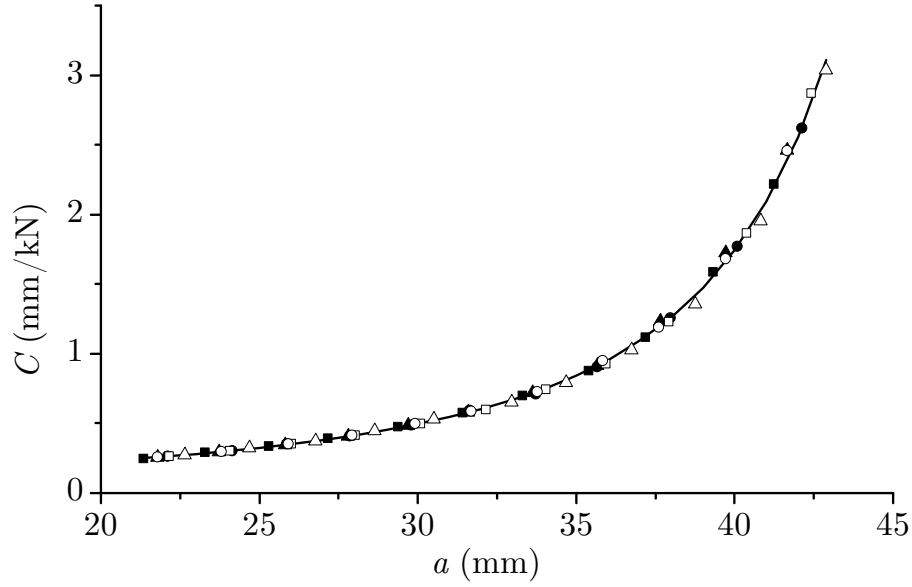


Figure 3.26. Curve obtained by measuring the compliance, C , of 6 CT specimens with known crack lengths, a (each symbol represents a different specimen), and fit obtained using equation (3.6).

3.4.5. Results

A representative load-displacement curve obtained from CT testing is shown in Figure 3.27. The variation of fracture toughness, measured for initiation, with initial notch root radius is shown in Figure 3.28. R-curves obtained from all tested specimens are plotted together in Figure 3.29.

Representative SEM micrographs taken at the point of initiation for each specimen configuration are shown in Figure 3.30 and Figure 3.31.

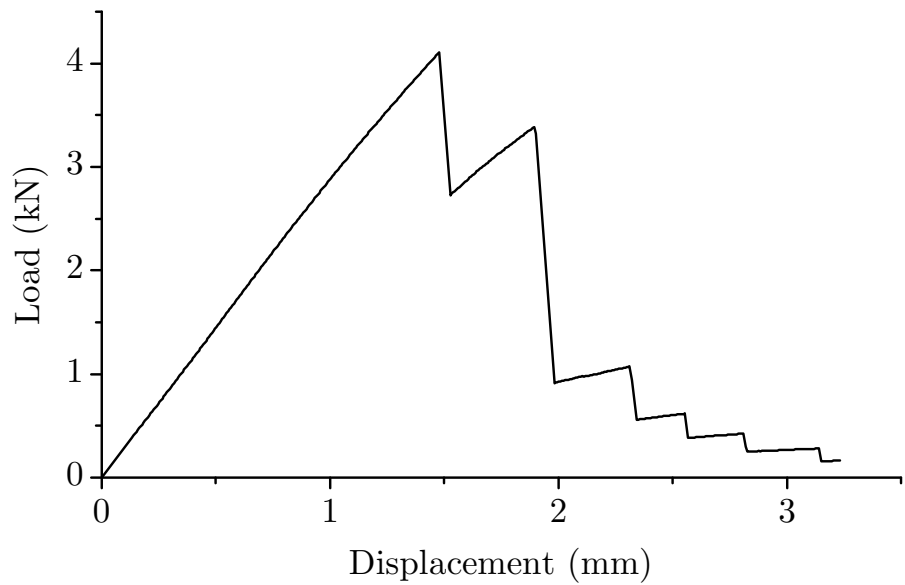


Figure 3.27. Representative load-displacement curve (for specimen with $\rho = 250 \mu\text{m}$).

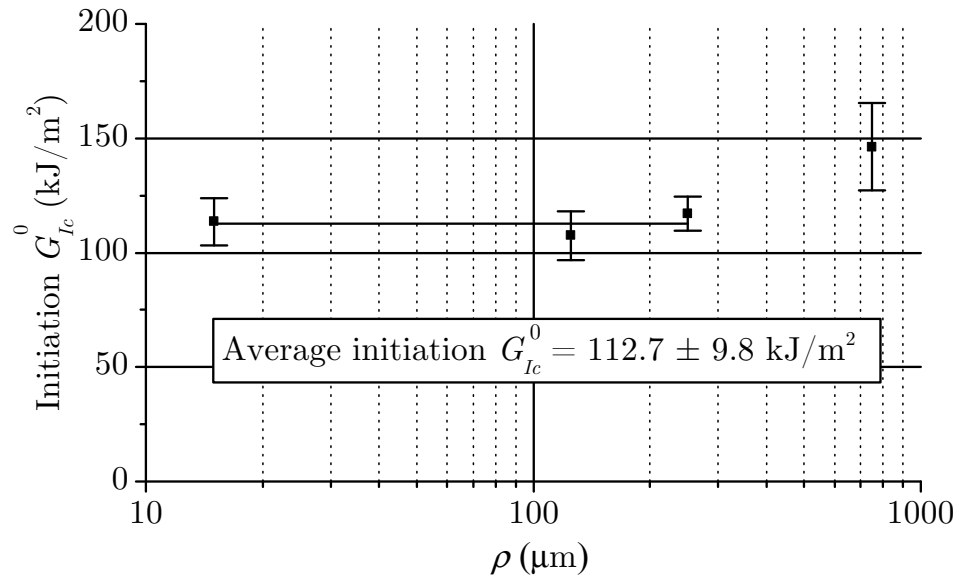


Figure 3.28. Variation of G_{Ic}^0 measured at initiation with notch root radius, ρ .

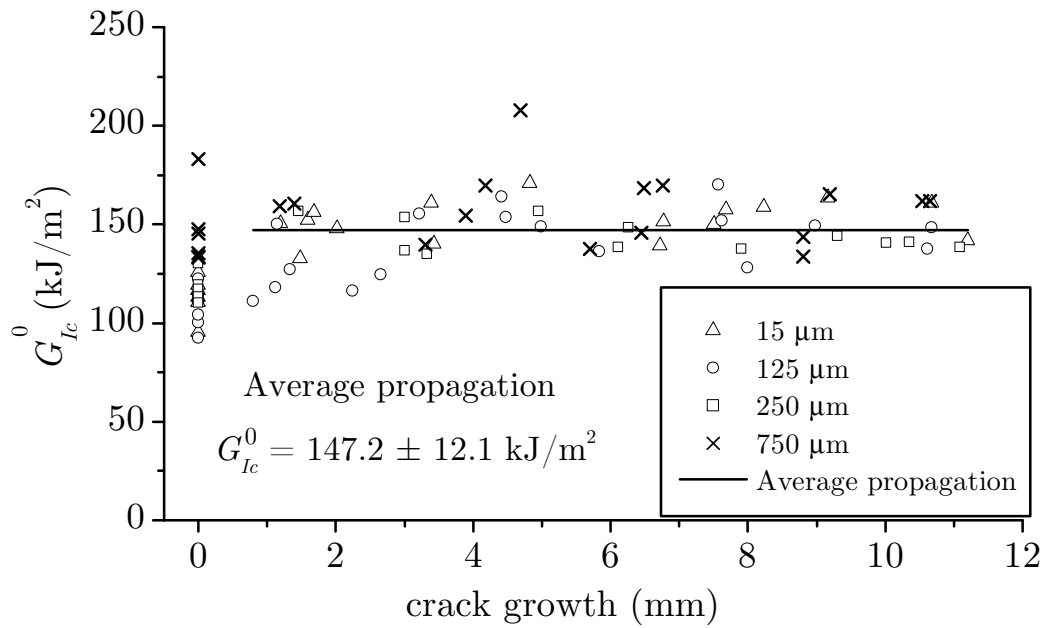


Figure 3.29. R-curves obtained from specimens with different notch tip radii.

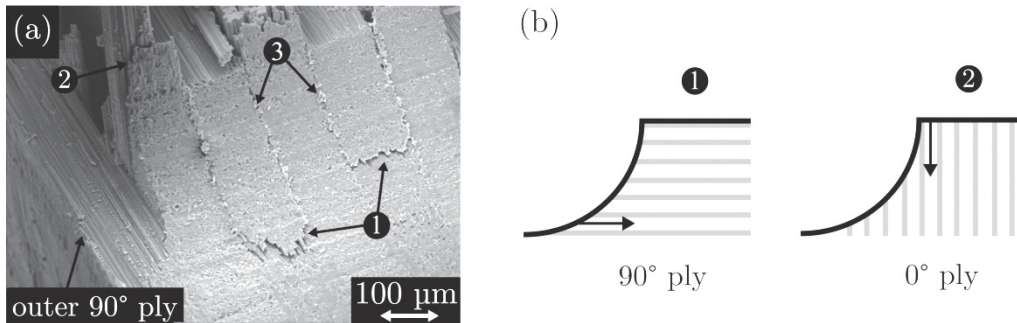


Figure 3.30. (a) SEM image taken at the notch tip of a $\rho = 250 \mu\text{m}$ specimen. Features of the damage zone prior to fibre failure are indicated as splitting in (1) 90° and (2) 0° plies, (3) delamination, (b) schematic of ply splitting.

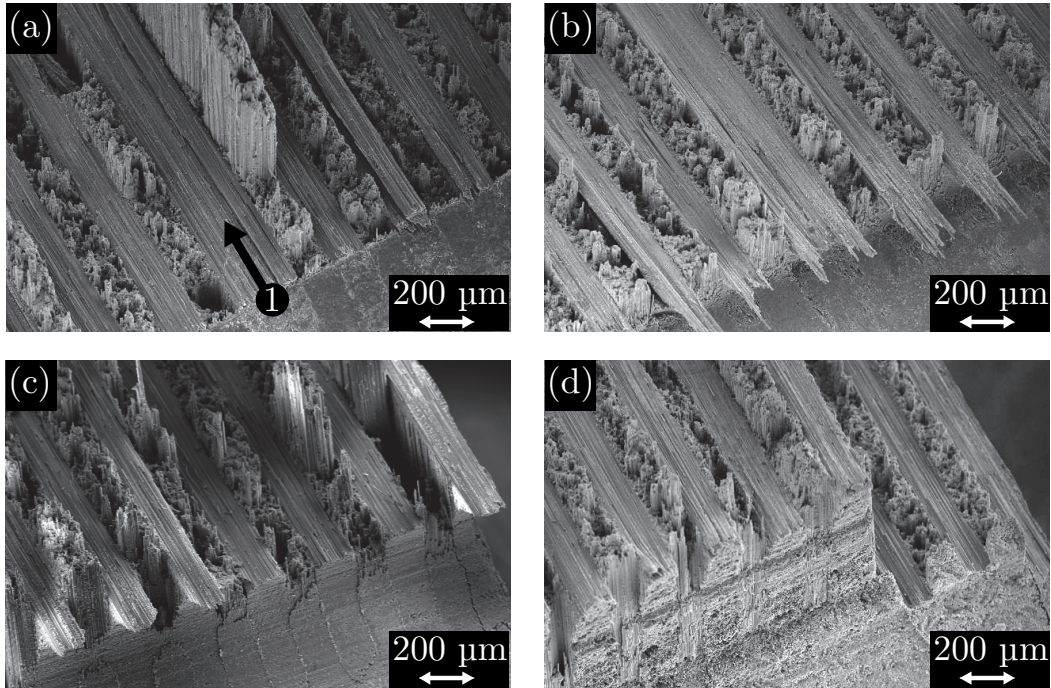


Figure 3.31. SEM of fracture surfaces of specimens with initial notch tip radii of (a) 15 μm , (b) 125 μm , (c) 250 μm and (d) 750 μm . Images are taken at the specimen initial notch tips, the damage propagation direction in each image is as indicated by ❶ in Figure (a).

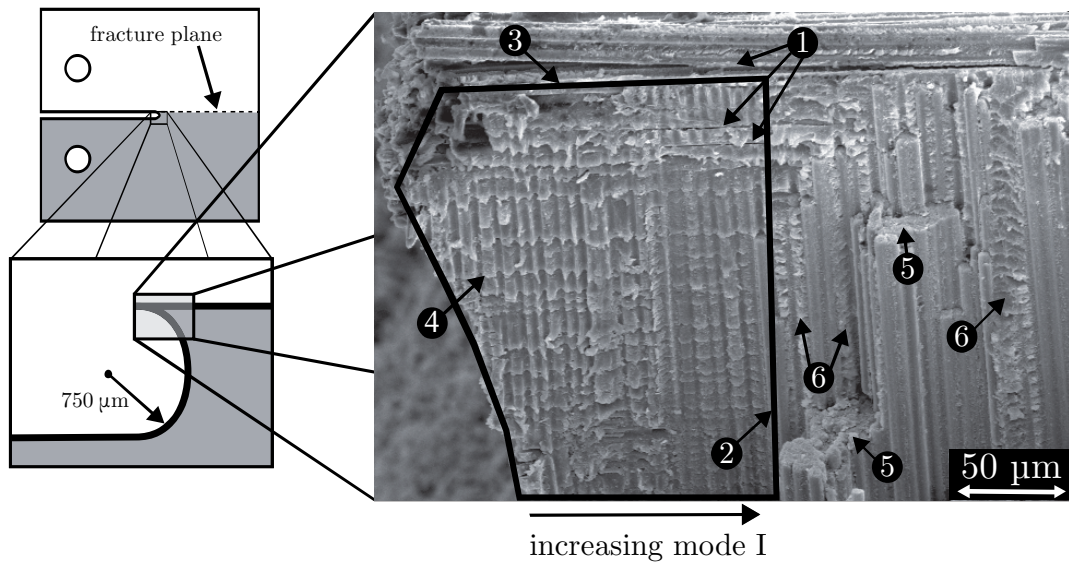


Figure 3.32. SEM image of 90°/0° ply interface at region of notch tip indicated in schematic.

Features of interest highlighted are ❶ 90° ply splits, ❷ line indicating location of 0° ply split, ❸ box indicating delaminated region and ❹ gouges associated with mixed mode delamination, ❺ fractured fibre ends, and ❻ shear cusps associated with fibre pull-out.

3.4.6. Discussion

The load-displacement curve shown in Figure 3.27 is representative of all tested specimens; the unstable manner in which translaminar fracture propagates is characterised by the large load drops, making it difficult to capture data on the rising part of the R-curve. The only marked differences in specimen behaviour were observed for the 750 μm configuration, which was found to have $\rho > \rho_c$, where P_c was usually higher and the resulting initial crack jumps were subsequently longer than in the other specimens; this is reflected in the R-curves shown in Figure 3.29.

The critical strain energy release rate, G_{Ic}^0 , measured at initiation was measured to be constant for specimens tested with $\rho \leq 250 \mu\text{m}$, as shown in Figure 3.28, indicating that $250 \mu\text{m} \leq \rho_c < 750 \mu\text{m}$. This lower bound plateau, averaged over the 18 specimens tested with $\rho \leq \rho_c$, is measured to be $G_{Ic}^0 = 112.7 \pm 9.8 \text{ kJ/m}^2$. Catalanotti et al. [14] obtained a slightly lower value of 97.8 kJ/m^2 for the same material system, though given their result is the average of three specimens it is believed that the results do correlate reasonably well. No details of the notch preparation are given in [14] though their result indicates that it is likely that the specimens were tested with $\rho \leq \rho_c$.

The plateau in the R-curve shown in Figure 3.29 indicates that the process zone is fully developed after a crack growth of approximately 1.5 mm. An increased number of points on the rising part of the R-curve is desirable, however, this is not possible given the unstable nature of the fracture mechanism. It should, however, be noted that other carbon/epoxy material systems can exhibit a more progressive R-curve effect, such as that characterised in [11]. The rising portion of the R-curve could be interpreted in several ways, so to avoid the subjectivity of the present author the measured propagation toughness is taken to be the average of all propagation measurements for specimens tested with $\rho \leq \rho_c$; $G_{Ic}^0 = 147.2 \pm 12.1 \text{ kJ/m}^2$. Again, this is larger than the result of Catalanotti [14] et al, who measured $G_{Ic}^0 = 133.3 \text{ kJ/m}^2$, though their result falls within the experimental scatter measured here and again is the result of only three specimens. Although specimens

tested with $\rho = 750 \mu\text{m}$ yielded an initiation toughness 30% higher than the other configurations it can be seen from Figure 3.29 that the propagation toughnesses are unaffected by ρ . The considerable difference between the initiation and propagation values of G_{Ic}^0 illustrate the importance of measuring both the initiation and propagation toughness for complete material characterisation, something that is not addressed by the current translaminar fracture toughness testing standard [4].

The main features of the damage zone associated with translaminar failure that were visible through fractography were, in accordance with Figure 3.30(a), found to be intralaminar splitting in the (1) 90° plies and (2) 0° plies, schematics of which can be seen in Figure 3.30(b), and (3) delamination at ply interfaces. These observations are in accordance with X-ray images made by Mandell et al [17] and account for the slight softening behaviour exhibited before the first load drop due to fracture of the specimen in Figure 3.27.

These features can be observed in more detail at the $90/0$ ply interface of a specimen tested with $\rho = 750 \mu\text{m}$, shown in Figure 3.32, where the damage is spread over a larger area. The bulk of the 0° ply has been pulled-out leaving fibre imprints in the matrix still stuck to the 90° ply that once formed the inter-ply region. It can be seen that 90° ply splits (1) formed first, followed by splitting in the 0° plies (2), these correspond to the failure mechanisms illustrated in Figure 3.30(b). Following this, the region indicated by (3) delaminated, with the initiation site corresponding to the plane of fracture in the 90° ply. Gouges (4) indicate a mixed mode delamination [64] with a slight increase in the proportion of mode I as indicated. Finally, fibre fracture occurred, indicated by the fractured fibre ends (5), the process of fibre pull-out is apparent from the highlighted shear cusps.

Inspection of the fracture surfaces at the notch tip of failed specimens using SEM revealed that despite there being a constant value of fracture toughness measured for specimens tested with $\rho \leq 250 \mu\text{m}$, the apparent damage zone size varied considerably between configurations, as can be seen from Figure 3.31(a)-(c). However, all specimens exhibited identical features in the wake of the advancing crack: fibre-matrix debonding, fibre breakage and subsequent pull-out can be

observed in the 0° plies with the plane of fracture in the 0° plies being dictated by the plane of intralaminar matrix failure in the 90° plies. Given that there seems to be no apparent change in the fracture morphology between regions very close to the initiation site and those further away that have been left in the wake of the advancing crack, it is unsurprising that the R-curves in Figure 3.29 indicate that the damage zone is fully developed after a very short amount of crack growth.

One mechanism contributing to the apparent size of the damage zone is the intralaminar splitting of the 90° plies that occurs prior to failure of the 0° plies. Splits are observed to occur preferentially at the corners of the initial notch tip, between continuous fibres that extend into the specimen and fibres that have been cut during the manufacturing process. The result is that the initiation site and subsequent plane of fracture in the 0° plies is at the corner of the original notch tip as can be seen in Figure 3.31(b)-(d). The ‘V’ shaped notch of the specimens tested with $\rho = 15 \mu\text{m}$ (Figure 3.25(a)) conditions fibre failure to occur centrally, as shown in Figure 3.31 (a). Given that G_{Ic}^0 for initiation is measured to be the same for Figure 3.31(a) – (c) it is suggested that the profile of the initial notch tip is unimportant provided $\rho \leq \rho_c$.

For both the $\rho = 250 \mu\text{m}$ and $750 \mu\text{m}$ configurations (Figure 3.31(c) and (d) respectively) the plane of fracture can be observed to vary across the width of the laminate as the result of the formation of 90° ply splits. Energy dissipation over a larger surface area in these cases would suggest an increased G_{Ic}^0 for both initiation and propagation, however, the results plotted in Figure 3.28 and Figure 3.29 indicate that the effect of this phenomenon has negligible effect on the measured G_{Ic}^0 .

Specimens with $\rho = 15 \mu\text{m}$ (Figure 3.31(a)) and $\rho = 250 \mu\text{m}$ (Figure 3.31(c)) represent the upper and lower bounds of the specimens tested with $\rho \leq \rho_c$. Although the same value of G_{Ic}^0 was measured in both cases, the fracture surfaces exhibit strikingly different morphologies. For the larger radius, the damage is spread over a much larger area, multiple splits can be seen in the 90° plies, the plane of

fracture varies across the width of the laminate and delaminated regions are visible at the ply interfaces. These mechanisms, which develop prior to 0° ply failure, serve to reduce the stress concentration promoting fracture of the 0° plies, but the fact that none of these are present in Figure 3.31(a) indicates that they are not factors controlling ρ_c .

The results plotted in Figure 3.28, coupled with the fracture surfaces shown in Figure 3.31(a)-(c) suggests that the extent of damage in the 90° plies prior to 0° ply failure has no effect on the resulting G_{Ic}^0 , and that there is no apparent interaction between 0° and 90° ply failure modes. In this case ρ_c is a property controlled by the 0° plies alone.

Failure in the 0° plies is controlled by the quality of the fibres and the fibre-matrix interface. Fibre-matrix debonding, initiated by large shear stresses at the fibre-matrix interface, occurs until the local strength of the fibre is reached at some point along its length where it breaks and is pulled-out. This process of debonding, fibre failure and subsequent pull-out is the result of the variable strength of the fibre along its length [18], meaning that ρ_c for the translaminar failure mode will be ultimately be controlled by quality of the fibres.

The sub-critical values of ρ tested here can be used as a broad guide for characterisation of G_{Ic}^0 ; Figure 3.28 illustrates that a notch tip with a sufficiently sharp radius can be prepared without having to use any specialist tools; a cut of $\rho = 250 \text{ }\mu\text{m}$ will suffice. For higher performance fibres and matrix ρ_c would be expected to increase, thus it is expected that the results presented within this study are applicable to the majority of carbon/epoxy materials systems.

3.4.7. Conclusions

The effect of specimen notch root radius, ρ_c , on the mode I translaminar fracture toughness, G_{Ic}^0 , has been investigated for the IM7/8552 carbon/epoxy system. The toughness measured at initiation was found to be independent of notch root radius for $\rho \leq 250 \mu\text{m}$. Above this value, an apparent initiation toughness was measured, implying that the critical notch root radius was in the range $250 \mu\text{m} \leq \rho_c < 750 \mu\text{m}$.

Propagation toughness was found to be independent of initial notch root radius for the tested configurations ($15 \mu\text{m} \leq \rho \leq 750 \mu\text{m}$), despite the fact that for larger values of ρ an increase in the apparent size of the damage zone was observed through SEM of the fracture surfaces of failed specimens.

Inspection of the fracture surfaces of failed specimens revealed that there is no interaction between the 90° and 0° ply failure modes, and that for the translaminar failure mode, ρ_c is a property of the 0° plies.

The results of this work serve as a guide to experimentalists. When preparing specimens, meaningful values of fracture toughness can be obtained for this material system provided that a notch with $\rho = 250 \mu\text{m}$ or smaller is used.

3.5. Publications

The work presented within this chapter has resulted in the following publications:

Measurement of the in situ ply fracture toughness associated with mode I fibre tensile failure in FRP. Part I: Data reduction. M. J. Laffan, S. T. Pinho, P. Robinson, L. Iannucci. *Composites Science and Technology* 2010.

Measurement of the in situ ply fracture toughness associated with mode I fibre tensile failure in FRP. Part II: Size and lay-up effects. M. J. Laffan, S. T. Pinho, P. Robinson, L. Iannucci. *Composites Science and Technology* 2010.

Translaminar fracture toughness: The critical notch tip radius of 0° plies in CFRP. M. J. Laffan, S. T. Pinho, P. Robinson, A. J. McMillan. *Composites Science and Technology* 2011.

Chapter 4

Mode I translaminar compressive failure

4.1. Introduction

Thus far, fracture toughness testing methods have been developed and standardised for interlaminar, e.g. [8], and translaminar [42] testing. Procedures, although unstandardised, have also been developed for longitudinal [65] and transverse [20] intralaminar matrix failure. However, as yet, no methods have successfully been developed to address the fibre compressive failure mode.

Soutis et al. [49] measured the compressive fracture toughness of carbon/epoxy centre-cracked specimens containing various proportions of 0° , 90° and $\pm 45^\circ$ plies. The results indicated that the laminate toughness increased with increasing proportion of 0° plies. Fracture testing of the 0° plies in isolation, however, resulted in failure by splitting parallel to the fibres perpendicular to the notch.

Sutcliffe and Fleck [58] found that fibre compressive failure could be initiated in unidirectional centre-cracked specimens if the specimens were pre-conditioned by

adding initiation sites for intralaminar splitting away from the notch tips intended for fibre compressive failure, so that splitting and fibre failure occur simultaneously. Another approach [26], that avoids any problems due to splitting of 0° plies, has been to measure the toughness of cross-ply laminates and account for energy dissipated due to intralaminar failure in the 90° plies using a rule of mixtures type approach.

Whilst the aforementioned studies provided measurements of toughness associated with fibre compressive failure, in both cases [20, 58] the energy consumed by intralaminar splitting was assumed to be very small or negligible and the interaction between failure modes was also neglected. To the author's knowledge, there are currently no fracture toughness data in the literature that have been measured for the longitudinal fibre compressive mode in isolation.

In the case of cross-ply laminates [26], compressive failure of the fibres can be masked by the 90° plies and it is unclear from the load displacement curves exactly when compressive failure initiates – thus calling into question whether or not the critical loads used in the data reduction process are the true values. A unidirectional specimen is therefore desirable for fracture toughness measurement of the fibre compressive failure mode; assumptions about other damage modes can be eliminated and the critical loads causing failure can be accurately measured.

The first part of this chapter, section 4.2, presents a new approach for characterisation of the fracture toughness associated with longitudinal fibre compressive failure. Notched unidirectional four-point-bend (4PB) specimens are tested without the problem of splitting observed in compression specimens elsewhere. By eliminating non 0° plies, the added complexity of failure mode interaction is also avoided.

In the second part of this chapter, section 4.3, a micromechanical FE model is used to investigate the nature of the observed failure mechanism.

4.2. Experimental

4.2.1. Materials and Manufacture

The material system used in this study is IM7/8552 unidirectional carbon/epoxy prepreg with a 0.13 mm nominal ply thickness. The material properties used for the data reduction were obtained using standard tests and are presented in Table 4.1 in the principal material axes.

Table 4.1. Moduli (in GPa) and Poisson's ratio of IM7/8552

E_{11}	E_{22}	G_{12}	ν_{12}
176.6	8.6	4.48	0.34

Composite panels with a lay-up of $[0]_{40}$ were manufactured using the hand lay-up method and cured to the instructions of the prepreg manufacturer. The panels were subsequently C-scanned to check for quality and cut to the geometry shown in Figure 4.1 using a wet saw.

In order to investigate the effect of initial notch root radius, ρ , the tip of the 5 mm long notch was prepared using two different methods. A razor saw was used to produce a notch radius of $\rho = 125 \mu\text{m}$ and a razor blade to give $\rho = 15 \mu\text{m}$, representative micrographs are shown in Figure 4.2. Prior to testing, the area around the notch tip on one side of the specimen was polished using a fine grade of sand paper so that damage initiation could be better observed during the test using a microscope lens attached to a CCD camera.

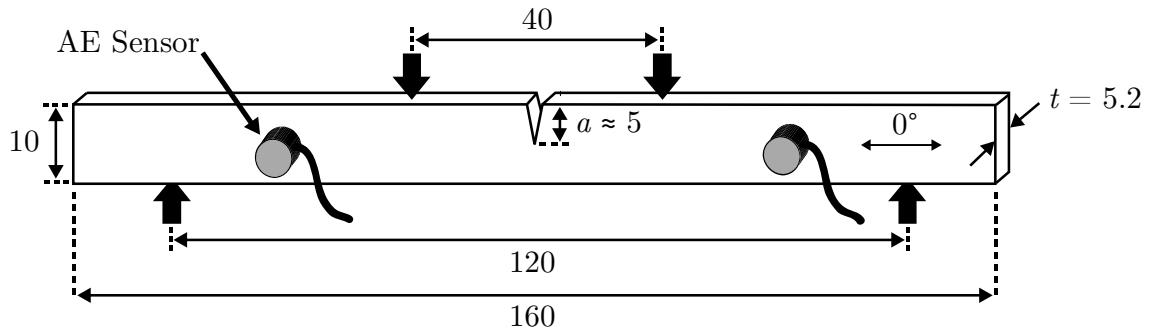


Figure 4.1. 4PB specimen dimensions (in mm) and location of AE sensors.

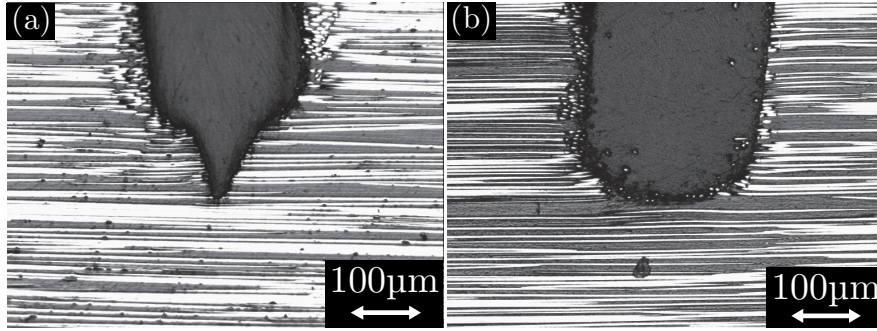


Figure 4.2. Micrographs of (a) $\rho = 15 \mu\text{m}$, (b) $\rho = 125 \mu\text{m}$ notch tips.

4.2.2. Experimental Procedure

Six 4PB specimens of each configuration were tested using an Instron machine with a 10 kN load cell; each specimen was loaded under displacement control at a rate of 0.1 mm/min. Measurements of load and crosshead displacement were recorded using a data logger.

A CCD camera with an attached microscope lens was used to view a highly magnified image of the polished notch tip on a monitor. This magnified image was used together with an event marker connected to the data logger to identify the point where damage initiation was optically detected.

An acoustic emission (AE) system was used to monitor damage evolution within the specimen. Sensors were positioned on the specimens as shown in Figure 4.1.

4.2.3. Data Reduction

4.2.3.1. Mode I Critical Energy Release Rate

The method used in this study to derive the mode I critical energy release rate from the raw experimental data relied on the use of the finite element (FE) method. A half specimen of unit thickness, $t = 1 \text{ mm}$, was modelled under a unit load, $P/2 = 0.5 \text{ N}$, with the boundary conditions shown in Figure 4.3. The virtual crack closure

technique (VCCT) was used to obtain the mode I energy release rate, G_I , from the model and a normalised mode I energy release rate was defined as

$$\bar{G}_I(a) = G_I(a) \left(\frac{1 \text{ mm}}{1 \text{ N}} \right)^2 \quad (4.1)$$

for crack lengths, a , between 4.6 and 5.4 mm in 0.2 mm increments as shown in Figure 4.4. The model was run using $0.2 \text{ mm} \times 0.2 \text{ mm}$ shell elements (S8R5) owing to the constraint of the 0.2 mm crack growth increment. Prior to this analysis, a convergence study was performed to confirm the accuracy of the solution with this element geometry.

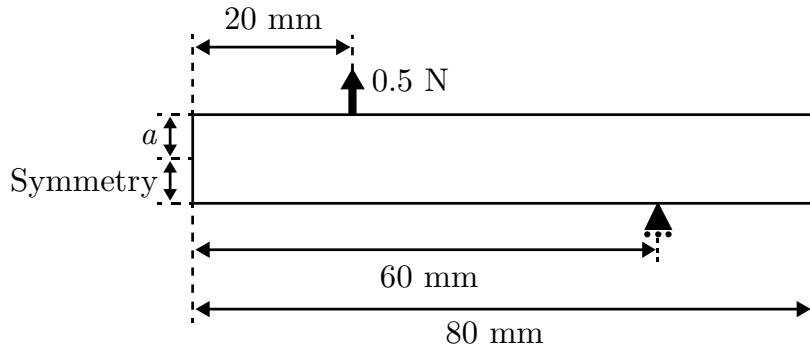


Figure 4.3. FE model used including loads and boundary conditions.

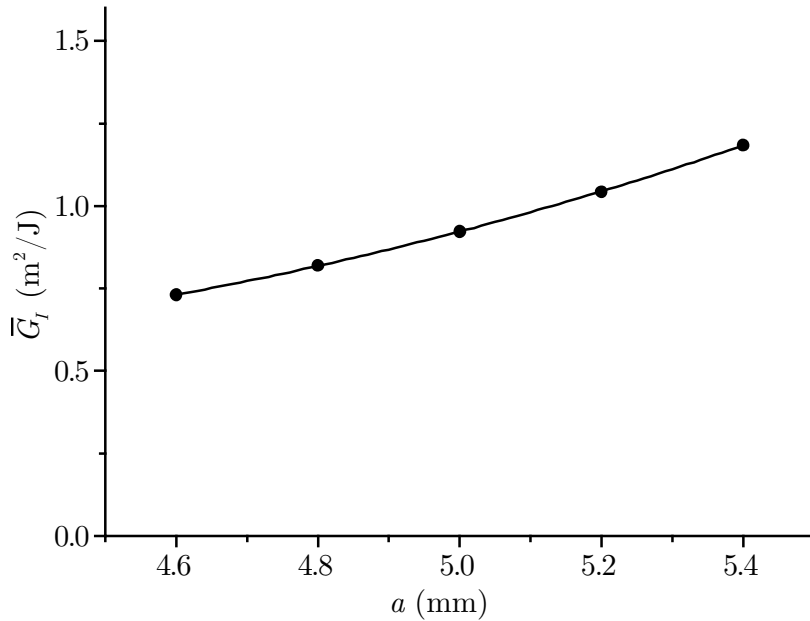


Figure 4.4. Values of $\bar{G}_I$ for $4.6 \text{ mm} \leq a \leq 5.4 \text{ mm}$.

The critical energy release rate associated with the onset of fibre compressive failure was then determined using [26]:

$$G_{Ic} = \bar{G}_I(a) \left(\frac{P_c}{t} \right)^2 \quad (4.2)$$

where P_c is the critical load at which failure was optically detected, and $\bar{G}_I(a)$ is obtained from a 2nd order polynomial fit to the data presented in Figure 4.4.

4.2.3.2. Mode II Critical Strain Energy release rate of a 45° crack

As will be discussed in section 4.2.5, failure consistently initiated as a mode II crack at 45° to the loading direction. This failure mode has also been observed post-mortem by Crasto and Kim [66], whilst Gutkin et al. [18] obtained micrographs from loaded specimens that detailed its formation in-situ. An intrinsic material property in this case would therefore be the mode II critical energy release rate associated with the 45° shear crack.

An approach similar to that described in the previous section was used to obtain G_{IIc} , however, the whole specimen was modelled with a 45° crack emanating from the initial notch, as shown in Figure 4.5(a), in order to obtain $\bar{G}_{II}(a)$. The mesh used is shown in Figure 4.5(b); a mesh convergence study lead to the adoption of square elements (S8R5) with a side length of 0.005 mm.

Models were run for three values of l (see Figure 4.5(a)), and $\bar{G}_{II}(l=0)$ was inferred by regressing linear lines of best fit, as shown in Figure 4.6. Values of $\bar{G}_{II}(l=0)$ were obtained for initial notch lengths $4.6 \text{ mm} \leq a \leq 5.4 \text{ mm}$, as shown in Figure 4.6 and Figure 4.7, so that $\bar{G}_{II}(l=0)$ for a particular specimen could be determined from a 2nd order polynomial fit to the data as shown in Figure 4.7.

This method was validated by performing the analysis for an isotropic case, the value of $\bar{G}_{II}(l=0)$, inferred by linearly regressing data obtained using VCCT, correlated well with an analytical solution. This can be found in Appendix A of this thesis.

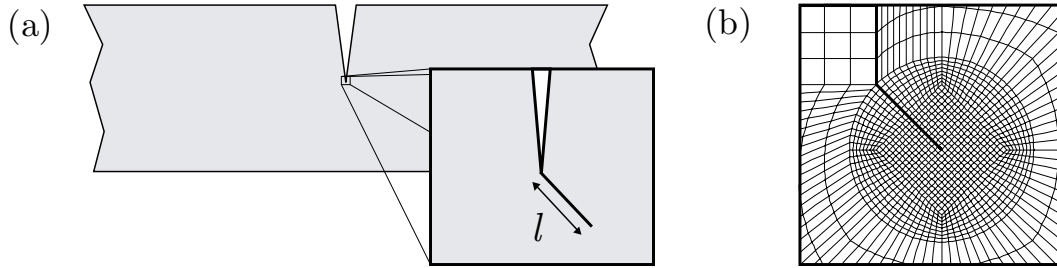


Figure 4.5. (a) Definition of 45° shear crack length, l , (b) mesh used to determine $\bar{G}_{II}$ via the VCCT.

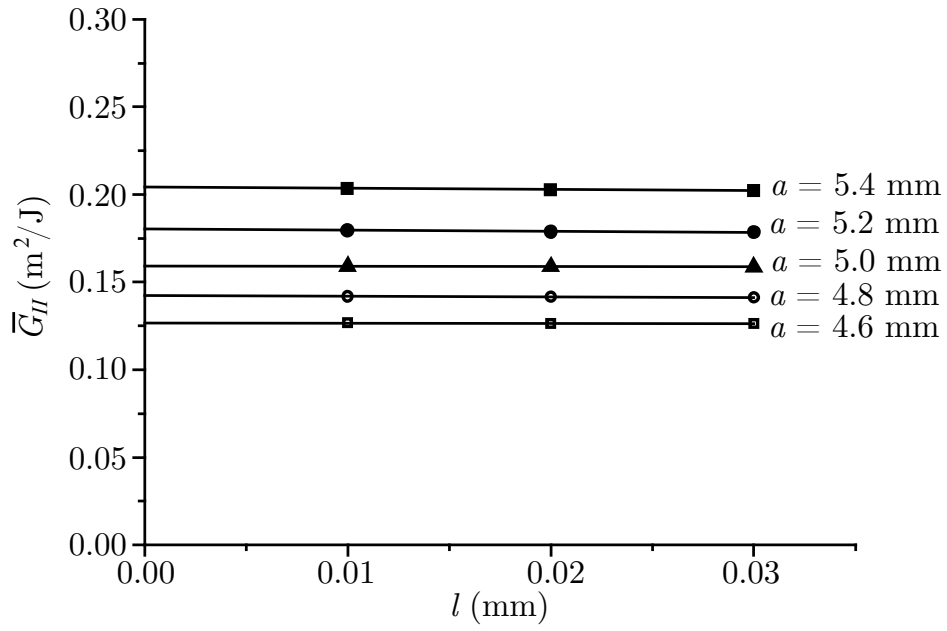


Figure 4.6. Values of $\bar{G}_{II}$ for 45° degree cracks of length l obtained from VCCT. Each symbol corresponds to a different value of a (see Figure 4.1).

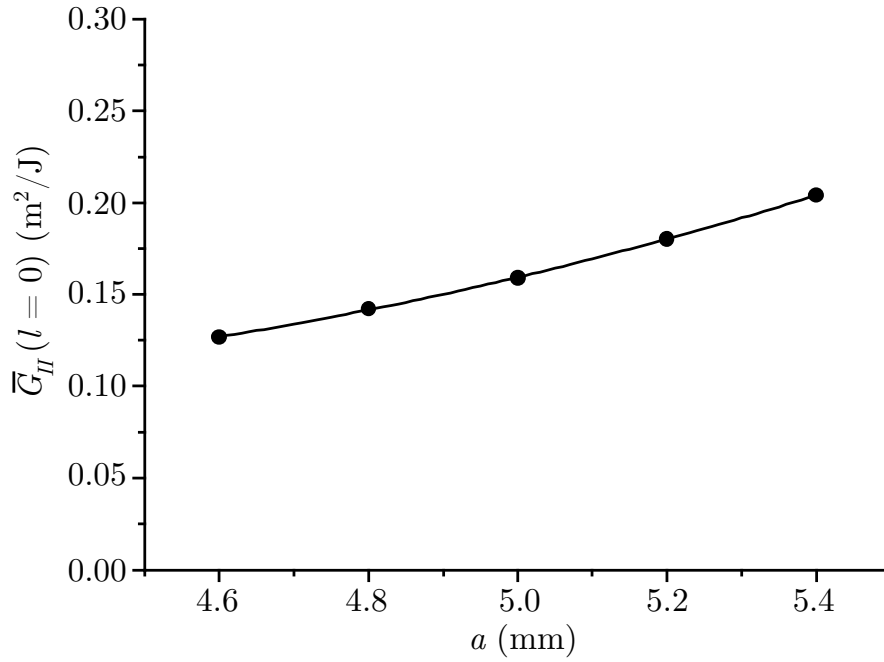


Figure 4.7. $\bar{G}_{II}(l=0)$ obtained from Figure 4.6, for various values of a fit with 2nd order polynomial.

4.2.4. Experimental Results

Typical load-displacement curves obtained from 4PB testing are shown in Figure 4.8 together with peak frequency data obtained from AE monitoring. The apparent mode I fracture toughnesses measured for initiation for the two notch configurations and the mode II toughnesses associated with the initiation of a 45° shear crack are shown in Table 4.2.

After testing, each specimen was cut and prepared for microscopy. Representative micrographs are shown in Figure 4.9, Figure 4.10 and Figure 4.11.

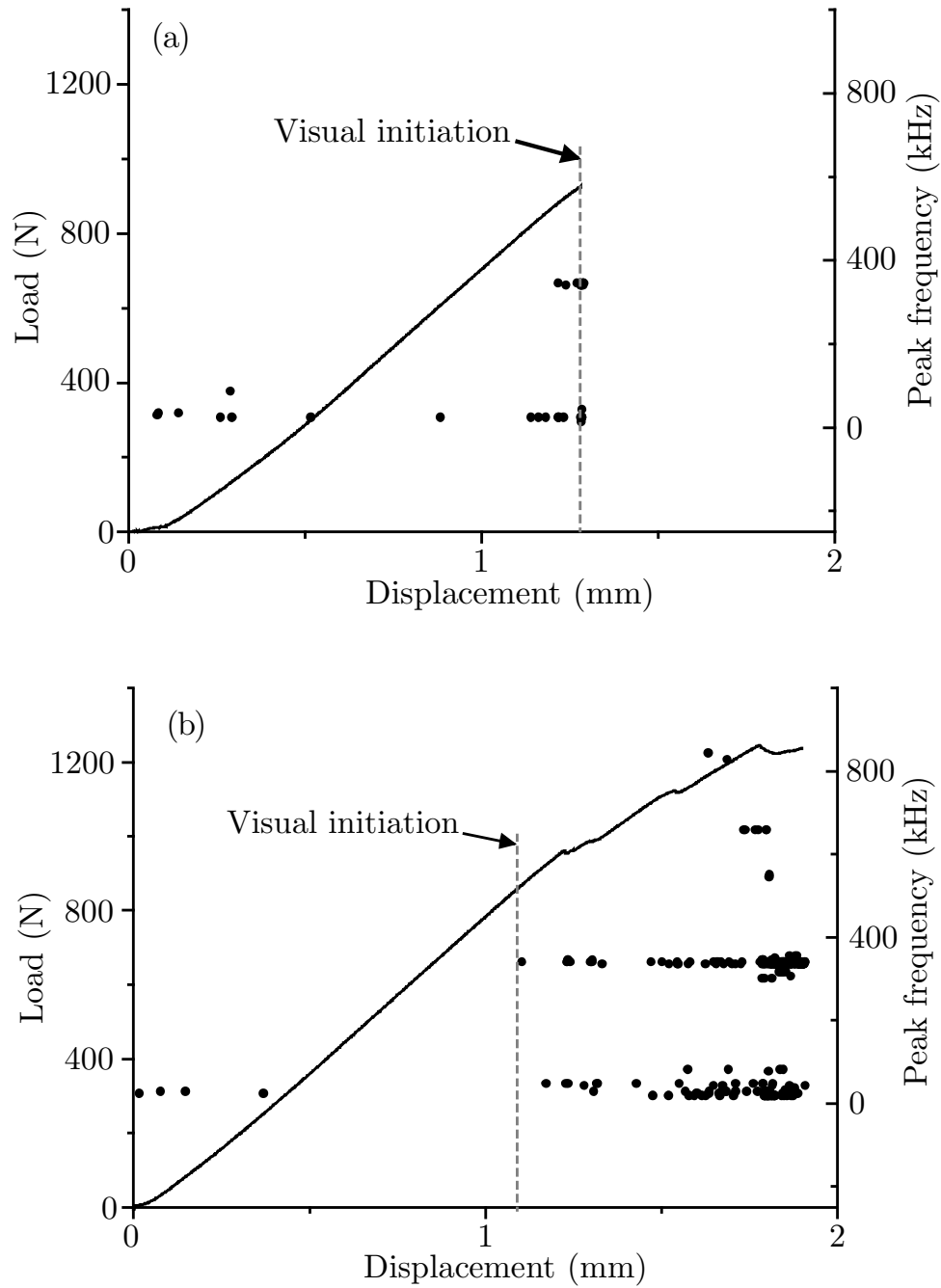


Figure 4.8. Representative load displacement curves showing peak frequencies of AE hits.
 (a) Test stopped after damage initiation, (b) specimen loaded until initial notch faces almost touching.

Table 4.2. Critical strain energy release rates obtained from 4PB testing. Six specimens were tested at each value of ρ .

ρ (μm)	G_{Ic} (kJ/m^2)	G_{IIc} (kJ/m^2)
15	25.1 ± 4.3	4.3 ± 0.7
125	26.7 ± 5.5	4.6 ± 1.0
All specimens	25.9 ± 4.8	4.5 ± 0.8

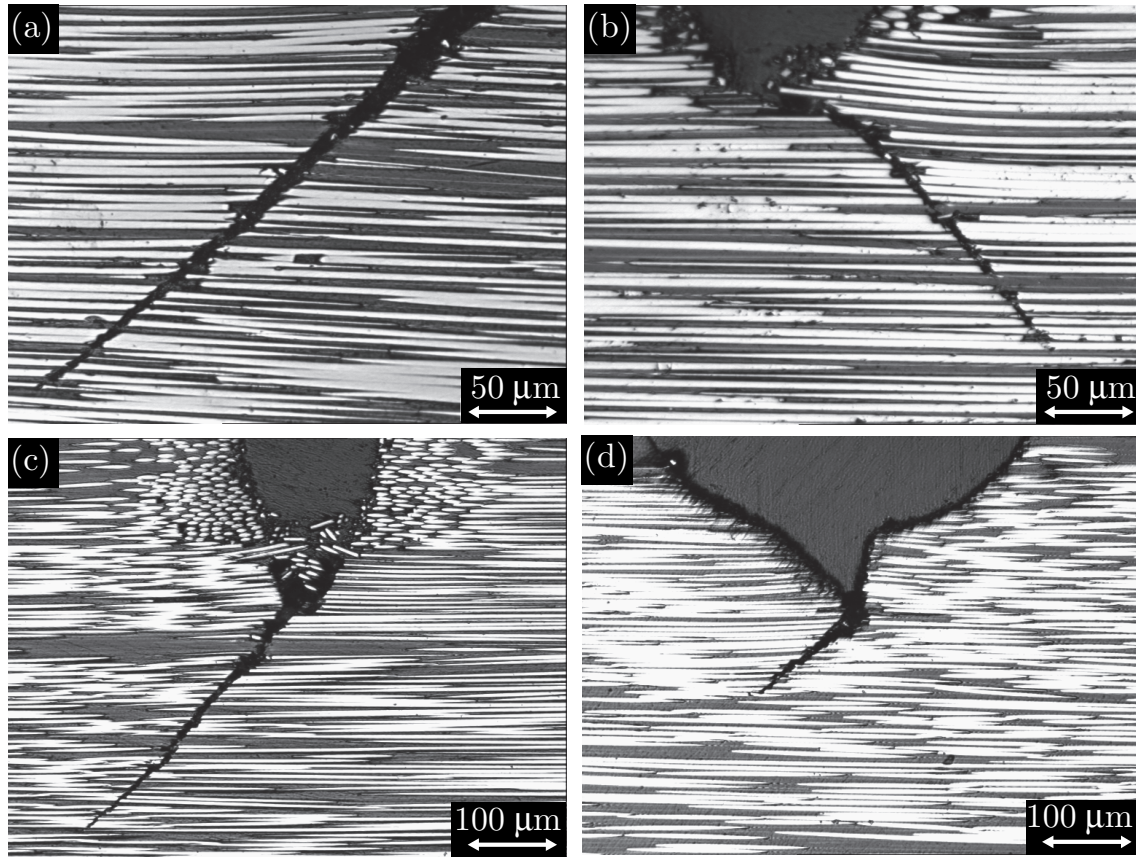


Figure 4.9. Microscopy of different specimens with 45° shear cracks, the initial stage of failure in 4PB specimens.

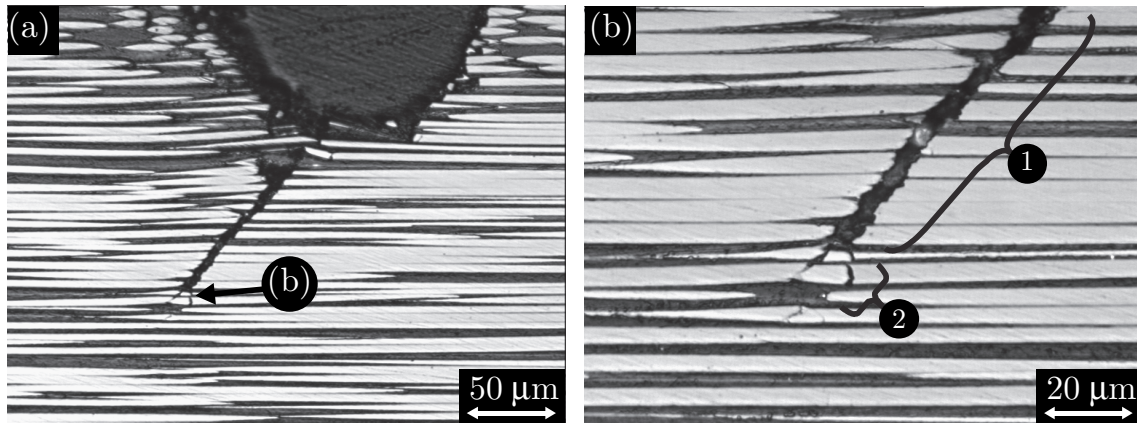


Figure 4.10. (a) Shear crack with kink band forming at tip, (b) Close up of shear crack/kink band transition region showing (1) shear crack and (2) first broken fibres of kink band.

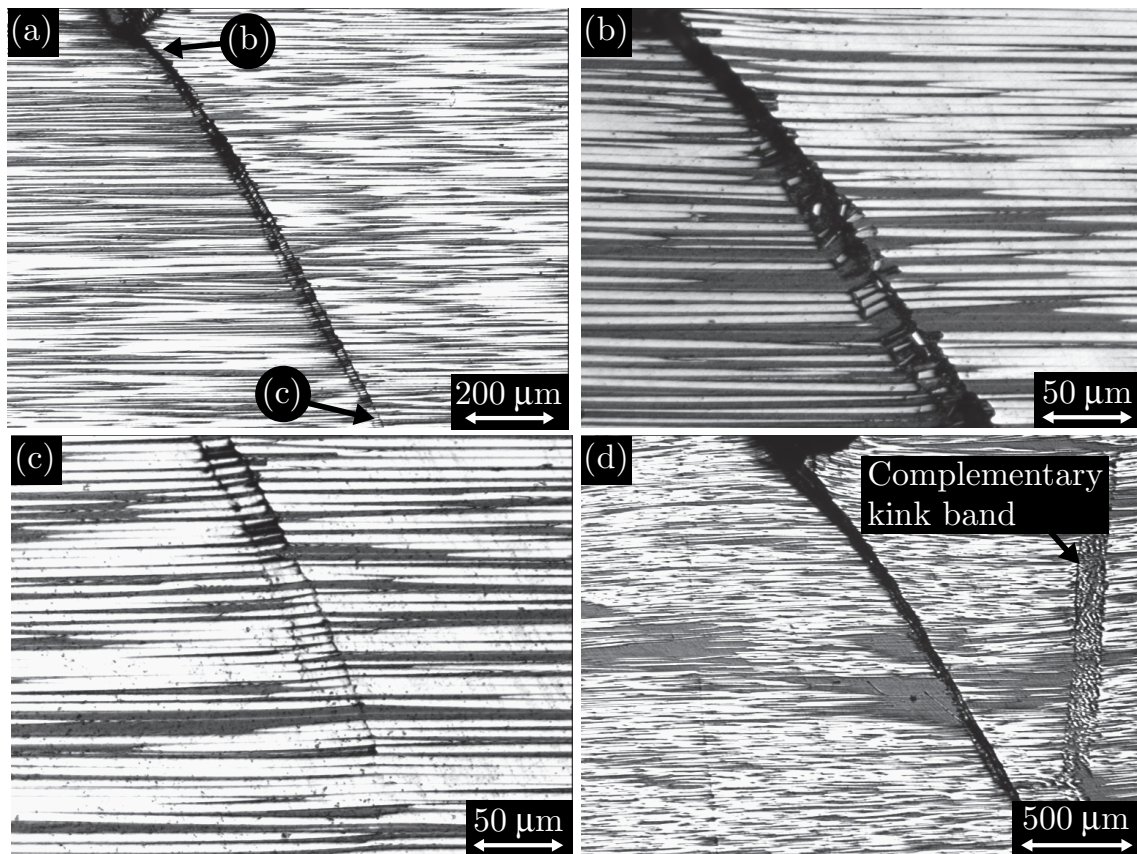


Figure 4.11. (a) Fully developed kink band within a 4PB specimen, (b) shear crack/kink band transition region, (c) kink band tip, (d) Fully developed kink band and subsequent complementary kink band.

4.2.5. Discussion

Figure 4.8 shows representative load-displacement curves from two specimens, one stopped almost immediately after failure has been detected and one loaded significantly past the point where the onset of damage initiation occurred. As can be seen, there are no strong discontinuities or clear indicators of failure in the region where damage initiation has been detected both optically and by the AE sensors, illustrating the importance of the methods used here to closely monitor failure.

The results of this type of test can be sensitive to the radius of the notch tip if it has been prepared such that it is greater than the critical notch root radius, ρ_c , e.g. [29]. In order to investigate this phenomenon, specimens were tested with two notch tip configurations, $\rho = 10 \mu\text{m}$ and $\rho = 125 \mu\text{m}$; given the negligible difference of results shown in Table 4.2, it can be inferred that all specimens were tested with notch radii smaller than the critical radius, ρ_c , for this material system. The average mode I critical energy release rate of all 12 specimens is measured to be $G_{Ic} = 25.9 \text{ kJ/m}^2$ with a standard deviation of 4.8 kJ/m^2 .

The measured G_{Ic} (25.9 kJ/m^2) is considerably lower than the value of $G_{Ic} = 47.5 \text{ kJ/m}^2$ reported by Catalanotti et al. [30] for the same material system, obtained using the compact compression specimen. It is thought that this is related to the fact that, in the current study, a UD specimen was used which enabled a detailed observation of the onset of failure. The data shown in Figure 4.8 illustrate that, in this case, the correct critical load cannot be obtained by inferring the onset of failure from load-displacement curves.

The two micromechanisms of longitudinal fibre compressive failure were clearly identified in each failed 4PB specimen using microscopy. Figure 4.9 illustrates the damage present in specimens that were stopped immediately after optical detection of damage initiation. The micrographs illustrate that compressive failure of the composite always initiates as fibre shear failure from the notch, propagating at approximately 45° to the pure mode I growth direction. The shear crack progresses for some distance before a transition occurs to fibre kinking failure; this transition has been captured in Figure 4.10.

The magnified image of the crack tip in Figure 4.10 (b) clearly shows the end of the shear crack (1) and the first few broken fibres of the newly formed kink band (2). Figure 4.9 and Figure 4.10 illustrate that the length of the shear crack region is not consistent from specimen to specimen; Figure 4.9 details shear cracks between approximately $150\text{ }\mu\text{m} - 300\text{ }\mu\text{m}$ in length with no signs of fibre kinking. The shear-kinking failure transition has occurred after just approximately $100\text{ }\mu\text{m}$ of shear crack growth in Figure 4.10. It is likely that this length is controlled by factors such as specimen alignment, and defects such as localised fibre misalignment at the crack tip.

Specimens with fully developed in-plane kink bands are shown in Figure 4.11. In Figure 4.11(a), after approximately $100\text{ }\mu\text{m}$ of shear crack growth, a kink band has developed and propagated approximately $1000\text{ }\mu\text{m}$ across the width of the specimen, the associated shear-kinking transition region and kink band tip shown are at a higher magnification in Figure 4.11(b) and (c). Despite the significant amount of damage present within the specimens, large stresses still exist across the failed region due to the contact between crack faces. Not only do these stresses serve to propagate the original kink band further, but also can initiate further damage such as the complementary kink band shown in Figure 4.11(d); features such as this account for the increased intensity of AE hits in the region above $P = 1200\text{ N}$ in Figure 4.8(b).

An apparent mode I critical strain energy release rate, G_{Ic} , has been measured which describes fracture initiation in a notched 0° ply under longitudinal compression. However, through microscopy, it is clear that the damage always initiates and propagates for some distance as a mode II shear crack. Therefore, the mode II critical strain energy release rate associated with a 45° fibre shearing crack is a more insightful intrinsic material property. In this case, the average of all 12 specimens is $G_{IIc} = 4.5\text{ kJ/m}^2$ with a standard deviation of 0.8 kJ/m^2 . The G_{IIc} value measured in this case is low, however, considering the micromechanisms dissipating energy during damage initiation (some plastic deformation of the matrix, subsequent fracture of both the matrix and fibres (the fracture toughness of an

individual carbon fibre has been estimated to be as low as 7.4 J/m^2 in tension [17])) such a low value would be expected.

Experimental investigations of the acoustic signature of various failure modes of carbon/epoxy composites [67, 68] have resulted in the peak frequency classifications shown in Figure 4.12. The data obtained from the 4PB tests, shown in Figure 4.8, exhibit two characteristic bands. The lowest acoustic emission band of peak frequencies, f_p , can be identified with matrix cracking. In both cases, the onset of fibre shear failure, and subsequent propagation of damage through fibre shear failure and kinking, have an associated band of AE data in the range $300 \text{ kHz} < f_p < 400 \text{ kHz}$. This range could be correlated with fibre compressive failure in all specimens tested and as the specimens were UD i.e. no other failure modes present besides matrix cracking, the grounds for matching this frequency range with fibre compressive/shear failure are indeed strong. It can thus be proposed that this frequency range corresponding to fibre compressive failure be added to the existing classification ranges as shown in Figure 4.12.

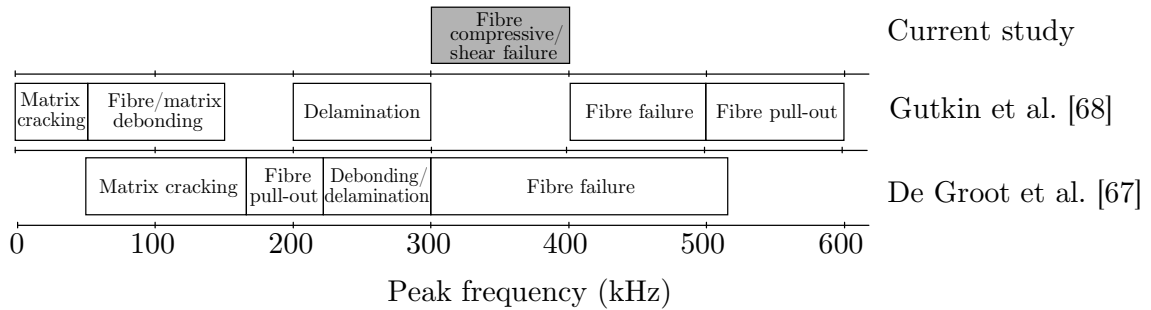


Figure 4.12. Peak frequency range classifications of failure modes in carbon/epoxy composites from modified to include fibre compressive/shear failure.

The critical strain energy release rates presented in this chapter are for initiation only; for complete characterisation of the damage tolerance of a material, information is required as to how the energy dissipation is affected with increasing crack length; thus a full R-curve is highly desirable. However, its measurement in this case is not trivial. Complete characterisation of compressive failure would require testing the shear and kinking failure modes in isolation to obtain separate R-

curves, which poses a problem. Accounting for the friction between and compressive stress across the contacting faces of the crack, as well as energy absorbed during the later stages of kink band development (e.g. complementary kink bands and kink band broadening), and monitoring crack growth are all problems that would be encountered, thus obtaining a full R-curve is not a trivial problem and is the subject of continuing efforts.

4.3. Micromechanical FE model

Failure of individual carbon fibres in compression via fracture orientated at 45° to the fibre axis is a widely reported phenomenon [69-71] thought to be associated with the internal architecture of the fibre [69]. Compression failure of composites, however, is more often observed in the form of fibre kinking, rather than the shear driven fibre failure shown in the micrographs of Figure 4.9. This is likely to be due to the difficulty in obtaining the necessary stress state at the notch tip as well as the constituent fibres and matrix needing a specific combination of material properties in order to facilitate the failure mode.

Post-mortem micrographs, such as those presented in Figure 4.9 - Figure 4.11, can be misleading, as the damage morphology can change upon unloading of the specimens. However, micrographs of loaded specimens [18] exhibiting the characteristic 45° cracks have provided conclusive evidence of the shear-driven fibre compressive failure. Following this work, Gutkin et al. [72] generated failure envelopes for unidirectional composites under combined compression and shear that account for the failure mode. A schematic of a failure envelope is presented in Figure 4.13; two distinct regions, labelled I and II, are apparent which are representative of the two modes of compressive failure. In region I, the compressive strength of the composite is dominated by the compressive strength of the fibres; the envelope is almost parallel to the vertical axis, indicating that the contribution of shear stress has little or no effect. This region corresponds to the shear driven fibre failure observed in this present work. Region II corresponds to fibre kinking, and eventually splitting as the vertical axis is approached.

The aim of the work in this present section is to verify, with the aid of a micro-mechanical FE model, that the failure observed in this present work does indeed lie in region I of Figure 4.13. By showing that the strength of the fibres dominates compressive failure of the composite, the possibility that failure has arisen through a microbuckling type process can be further ruled out.

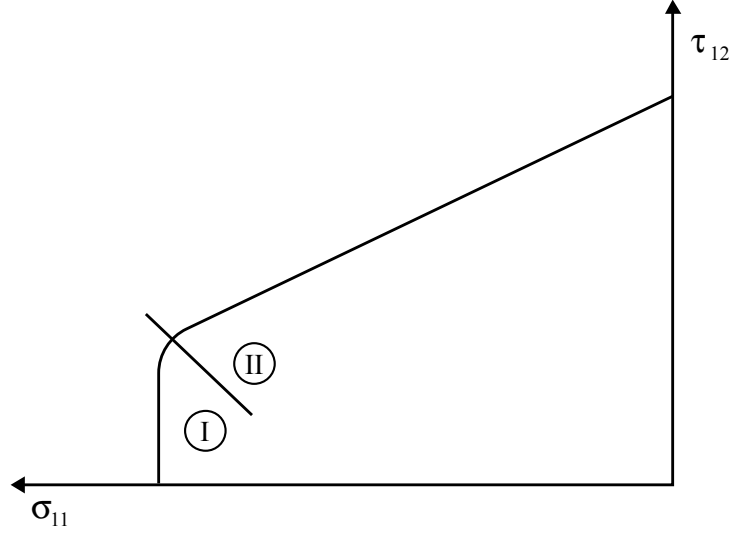


Figure 4.13. Schematic showing main features of composite compression/shear failure envelope generated by Gutkin et al. [72].

4.3.1. Model setup

The 4PB specimen was modelled with a thickness of $t = 5$ mm and a maximum applied load of $P = 2000$ N as shown in Figure 4.14(a). A 0.6×1.2 mm portion at the notch tip was refined such that the fibres and layers of matrix were individually discretised. The mesh of S4R elements used in this 0.6×1.2 mm region is illustrated in Figure 4.14(b), three elements were used across the width of each fibre and 1 element was used to represent the thickness of the matrix layer. The fibre volume fraction in 2D was chosen to be representative of hexagonal packed fibres such that

$$v_f^{2D} = \sqrt{\frac{2\sqrt{3}}{\pi}} v_f \quad (4.3)$$

where v_f is the volume fraction of the composite (in this case $v_f = 0.6$). The fibre and matrix properties used for the model are shown in Table 4.3.

The matrix was modelled using a Drucker-Prager plasticity model with the required model parameters calculated as per reference [73].

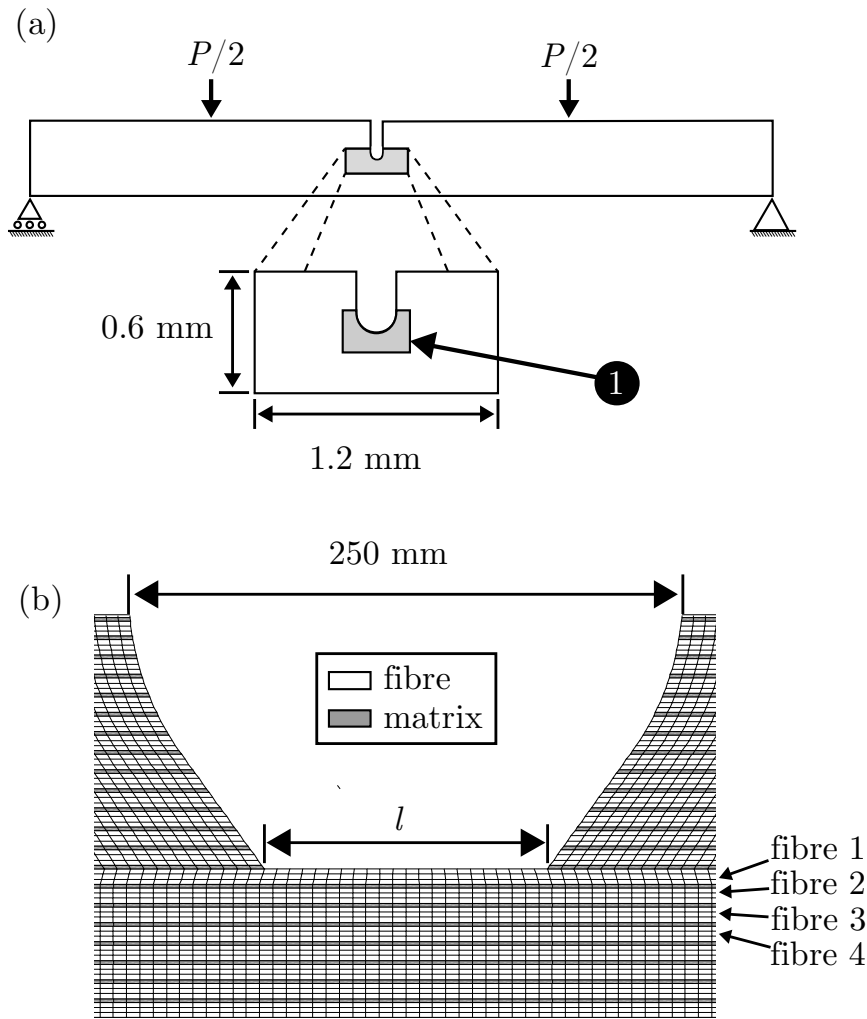


Figure 4.14. (a) FE model with loading and boundary conditions highlighting the refined portion where individual fibres and layers of matrix are modelled. (b) Mesh at notch tip corresponding to the shaded area indicated by 1 in (a).

Table 4.3. Fibre and matrix properties used for micromechanical FE model [74]

Matrix properties	
E (GPa)	4.08
ν	0.38
Compressive strength (MPa)	57
Shear strength (MPa)	130
Fibre properties	
Diameter d_f (μm)	7
E_{11} (GPa)	276
E_{22} (GPa)	19
G_{12} (GPa)	27
ν_{12}	0.2
Compressive strength (MPa)	3200

4.3.2. Results

The relative longitudinal displacement of the four fibres directly below the notch-tip, see Figure 4.14(b), are plotted against the global load applied to the 4PB specimen in Figure 4.15. The distance over which the relative longitudinal displacement was measured is indicated as l in Figure 4.14.

The compressive stress in the central element of the fibre 1 (Figure 4.14) at the symmetry axis is plotted against the relative longitudinal displacement of the fibre in Figure 4.16. The compressive strength of the fibre is reached at 66% of the buckling load.

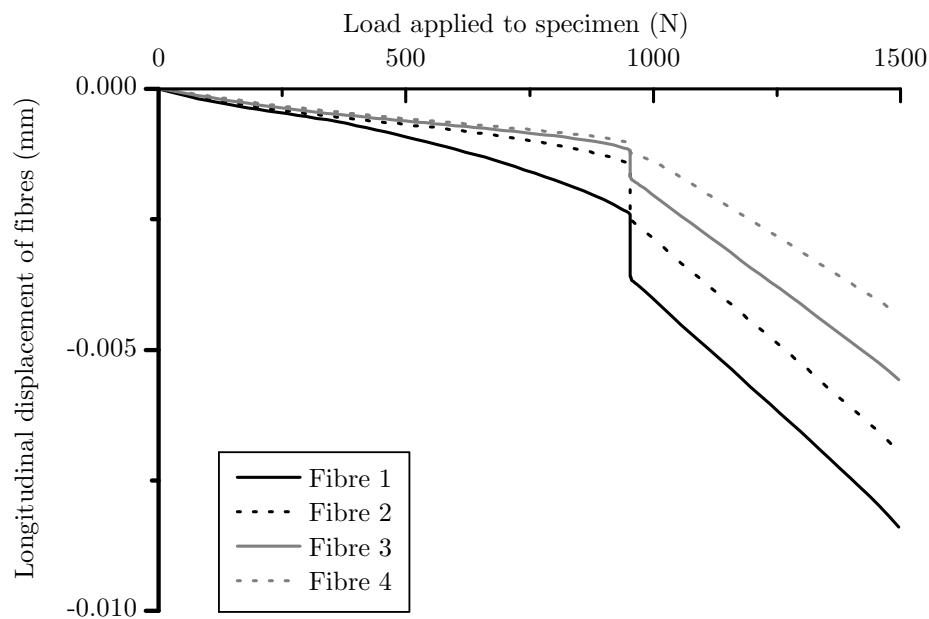


Figure 4.15. Longitudinal displacement of four fibres adjacent to notch-tip plotted against load applied to the 4PB specimen.

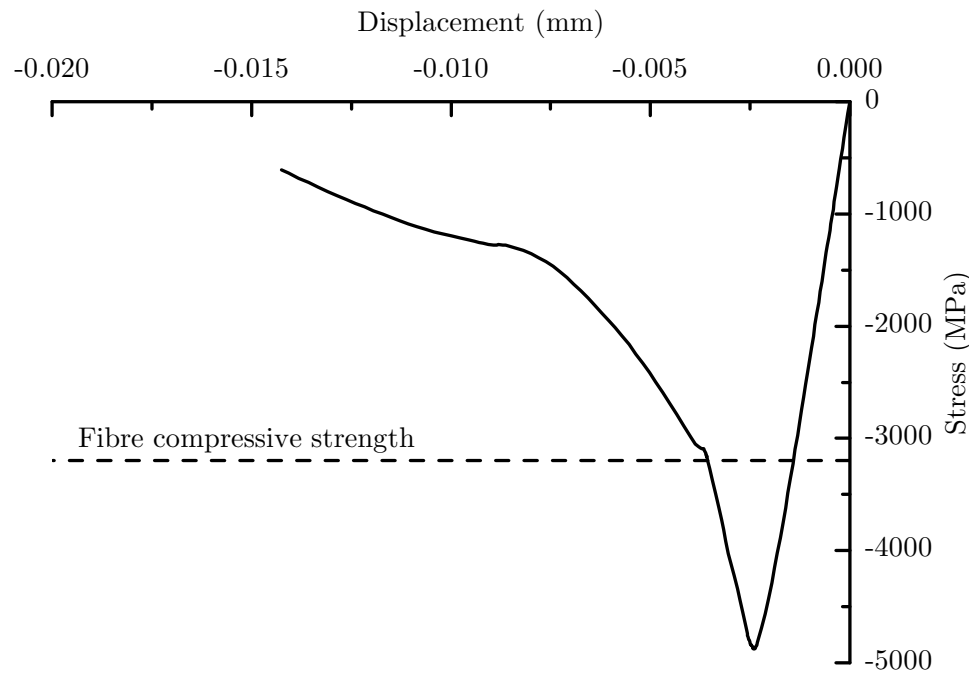


Figure 4.16. Stress experienced by the fibre adjacent to the notch tip plotted against its longitudinal displacement. The compressive strength of the fibre is highlighted.

4.3.3. Discussion

The micro-buckling phenomenon exhibited by composites loaded in compression can be clearly observed in Figure 4.15. The instability at a global applied load of just under 1000 N is representative of fibre buckling. If compression failure of the fibre had not yet occurred, further loading of the specimen after this point would result in fibre kinking failure, as the necessary conditions are provided by high bending stresses within the fibre.

The conditions under which shear driven fibre compressive failure occur are illustrated in Figure 4.16, where the compressive stress experienced by the fibre adjacent to the notch tip is plotted against its longitudinal displacement. Again the micro-buckling instability is visible, this time in the form of a sharp decrease in compressive stress at a displacement of approximately 0.0025mm, but it can now be seen that the compressive strength of the fibre is reached well before this occurs, at a stress that is 66% of that which causes the fibre to buckle. The results of this model therefore strongly suggest that failure in this case is governed by the strength of the fibre and would lie in the region denoted by I in the failure envelope of Figure 4.13.

In contrast, similar experiments on notched specimens by Gutkin et al. [18] and Hapke et al. [75] resulted in fibre kinking initiating directly at the notch tip. In the former case, it is probable that the method of load introduction resulted in shear stress combined with compressive stress at the notch tip which provided favourable conditions for fibre kinking. In the latter case, an inclined notch and splitting at the notch tip provided favourable conditions for kink-band initiation. The benefit of the 4PB setup presented here is that the state of compression induced at the notch tip is relatively in-sensitive to imperfections in geometry and loading. Composite failure is therefore dictated by the strength of the fibres, rather than the shear properties of the matrix, allowing for shear driven fibre compressive failure to occur.

4.4. Conclusions

A simple method is presented for the measurement of the fracture toughness associated with the initiation of fibre compressive failure. The fracture toughness can be directly determined from unidirectional specimens without any intralaminar splitting and without having to make assumptions about interaction of damage in non 0° plies.

Fibre compressive failure was seen to initiate in the form of a shear crack at 45° to the direction of pure mode I crack growth; the shear crack later forms a single in-plane kink band which propagates further into the specimen. Information about energy dissipation during propagation of these fracture mechanisms cannot be readily derived from the data obtained from testing due to their interaction and the contact stresses between the crack faces. The data obtained from these tests enables the prediction of damage initiation for composite structures under pure mode I compressive loadings.

Furthermore, through a micromechanical FE model it has been shown that the 45° shear driven fibre failure mode exhibited in the present experiments is brought about by the lack of a significant shear stress component along the plane defined by the initial notch. This condition allows the compressive failure strength of the fibre to dominate and dictate the failure process of the composite.

4.5. Publications

The work presented within this chapter has been submitted to Composites Part A as a paper entitled: “Measurement of the fracture toughness associated with longitudinal fibre compressive failure of FRP”.

Chapter 5

Mixed-mode I/II translaminar tensile/shear failure

5.1. Introduction

Mode I translaminar tensile failure has been studied extensively in Chapter 3 of this thesis, however, this represents a very narrow portion of the design spectrum. Typically, an in-service structure would be subject to a much broader range of loading regimes; it is therefore necessary that mode II and mixed-mode failure be investigated so that a more complete understanding of material behaviour is obtained.

Thus far, investigations of mixed-mode translaminar fracture have been limited to fracture toughness measurement for quasi-isotropic laminates [25, 76-78]. Whilst these studies have provided critical stress intensity factors for fracture initiation of the respective laminates, no information is available in the literature concerning details of the failure mechanisms introduced alongside the mode II component of loading. Furthermore, the centre-cracked tension specimens used in past studies suffer from unstable crack growth with subsequent crack propagation

under mode I conditions, such that information regarding damage zone development and progression under mixed-mode conditions cannot be gleaned from post-mortem analyses of tested specimens.

At present, little is known about the micro-mechanisms associated with mixed mode translaminar failure of CFRP. With this as motivation, the aim of this study is to present a fractographic analysis of CFRP specimens which have had translaminar cracks initiated and propagated through them under several mixed mode ratios. The failure mechanisms introduced alongside the mode II component of loading are identified, and where possible, R-curves are generated to provide an understanding of how these mechanisms affect the behaviour of the composite during damage propagation.

5.2. Materials and Manufacture

A mixed mode compact tension specimen (MMCT), shown in Figure 5.1, was selected for this study so that damage could be stably propagated under several mixed mode ratios.

It was anticipated that the mixed-mode toughness could be considerably higher than the pure mode I value, for this reason it was decided to have only 4 0° plies in the laminate. A cross-ply laminate was chosen so that the data reduction process outlined in section 3.2.3.6 of thesis could be used to determine the 0° ply-level toughness by accounting for the energy dissipated in the 90° plies. Furthermore, having observed a slight specimen thickness effect in Chapter 3 of this thesis, it was decided to test a laminate of similar thickness to the CT-base arrangement (34 plies) so a lay-up of $[90_4/0/90_4]_{2s}$ was chosen for the mixed-mode translaminar tests.

Composite panels made up of 0.125 mm thick unidirectional plies of IM7/8552 were laid up by hand and cured to the instructions of the pre-preg manufacturer. The cured panels were C-scanned as a quality check before being cut into the dimensions shown in Figure 5.1 using a wet saw. The 6 mm holes were prepared using a carbide coated dagger drill, a 45 mm length of the initial notch

was cut using a diamond coated disc saw and the remaining 10 mm were prepared as a V shaped notch using a carbide coated disc saw.

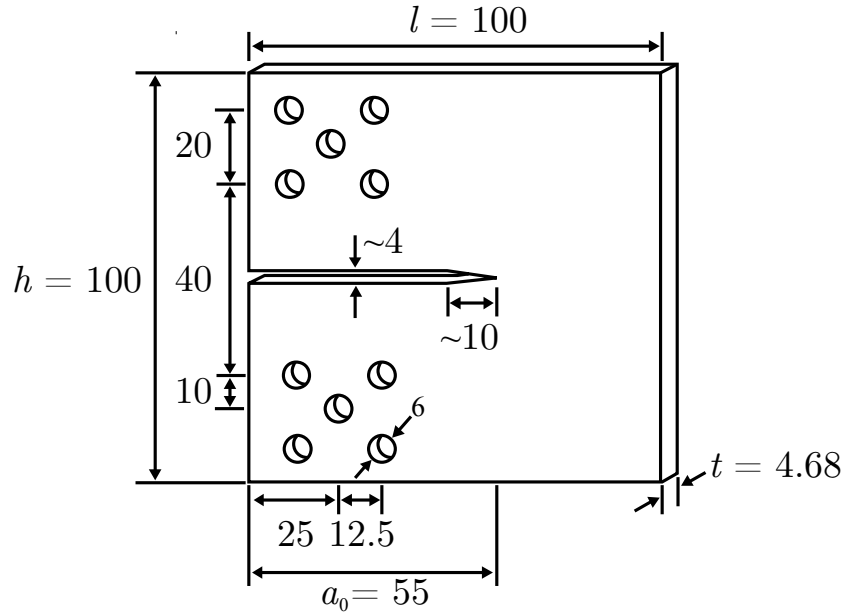


Figure 5.1. Dimensions (in mm) of the mixed mode compact tension specimen.

5.3. Experimental Procedure

Tests were performed using the fixture shown in Figure 5.2 which is based on a design originally proposed by Richard and Benitz [79]. The mode-mixes obtainable using the given testing configuration were calculated using finite element (FE) analysis as will be outlined in section 5.6. Four mixed-mode ratios were used: $G_{II}/G_{total} = 0.12, 0.24, 0.46$ and 1.00 , which, in accordance with Figure 5.2, corresponded to loading angles of $\theta = 34^\circ, 24^\circ, 15^\circ$ and 0° respectively.

Five MMCT specimens were tested for each of the mixed-mode ratios under displacement control at 0.5 mm/min using an Instron machine with a 100 kN load cell. Measurements of load and crosshead displacement were recorded using a data logger. After the test, the displacement measurements were corrected to account for the machine and testing fixture compliance.

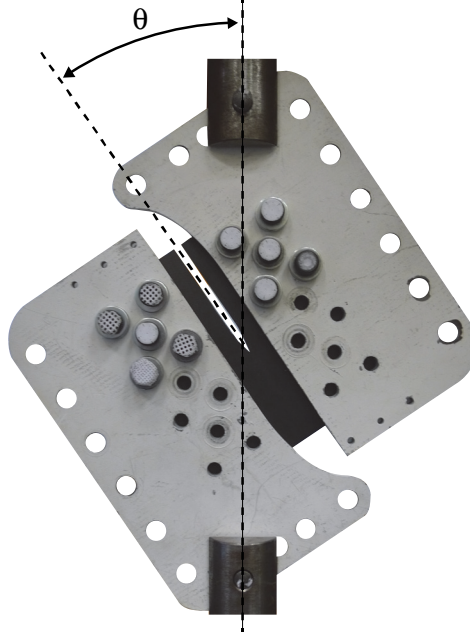


Figure 5.2. Mixed mode fixture at a loading angle of $\theta = 34^\circ$ corresponding to $G_{II}/G_{total} = 0.12$.

5.4. Results

Representative load-displacement curves for each tested mixed-mode ratio are shown in Figure 5.3.

Specimens were C-scanned after testing and then dissected so that their fracture surfaces could be observed using a scanning electron microscope (SEM). Representative images of failed specimens tested with $G_{II}/G_{total} = 0.12$, 0.24 and 0.46 are shown in Figure 5.4. Each image details the fracture surface immediately adjacent to the notch tip of tested specimens as indicated in Figure 5.4(a). Several features associated with mixed-mode translamellar failure were identified; Figure 5.5 highlights the uniform lengths of fractured 0° ply, Figure 5.6 indicates that the 0° plies were in fact delaminated before fibre fracture occurred, and Figure 5.7 indicates splitting within pulled-out fibre bundles.

No fibre fracture occurred for specimens tested with $G_{II}/G_{total} = 1.00$. Instead, extensive ply splitting and delamination occurred. Figure 5.8 details the damage observed at the notch tip of a specimen tested in pure mode II. The outer 90° plies

had completely delaminated and were removed in the dissection process, leaving the 0° ply below exposed.

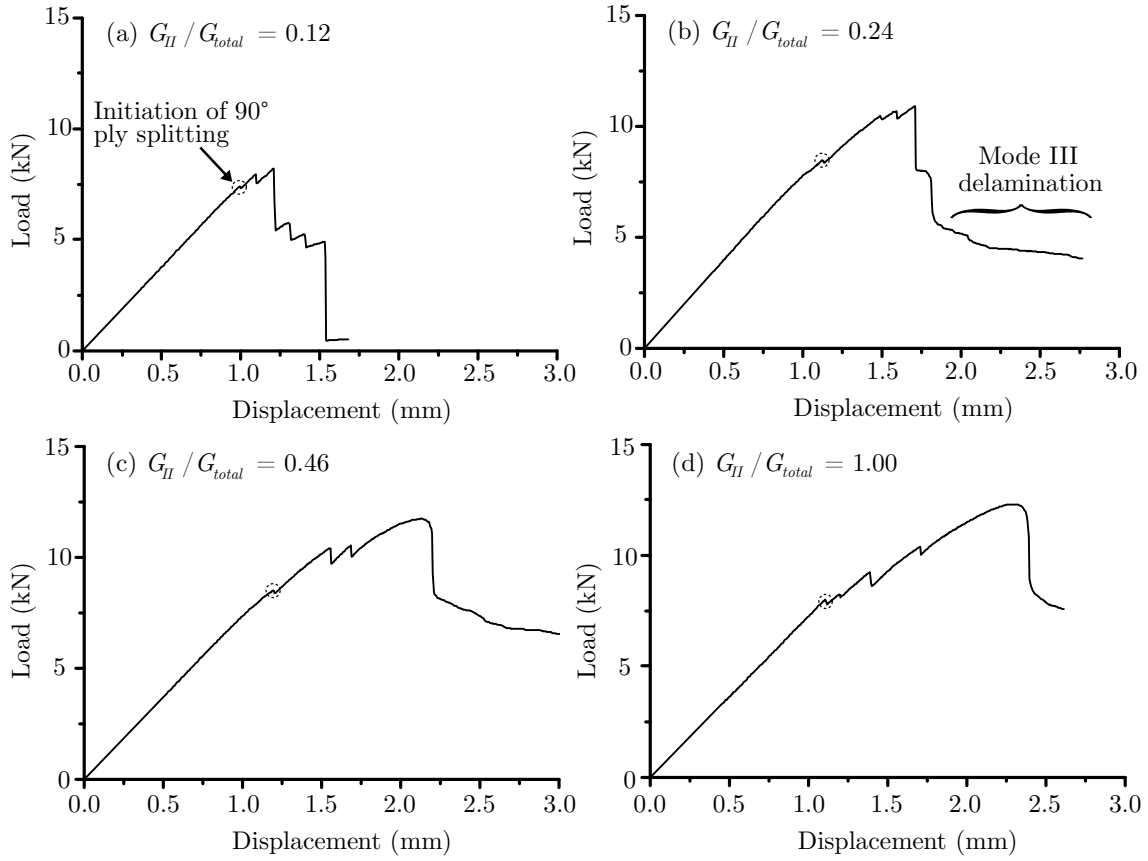


Figure 5.3. Load displacement curves of specimens with initial G_{II}/G_{total} ratios of (a) 0.12, (b) 0.24, (c) 0.46 and (d) 1.00.

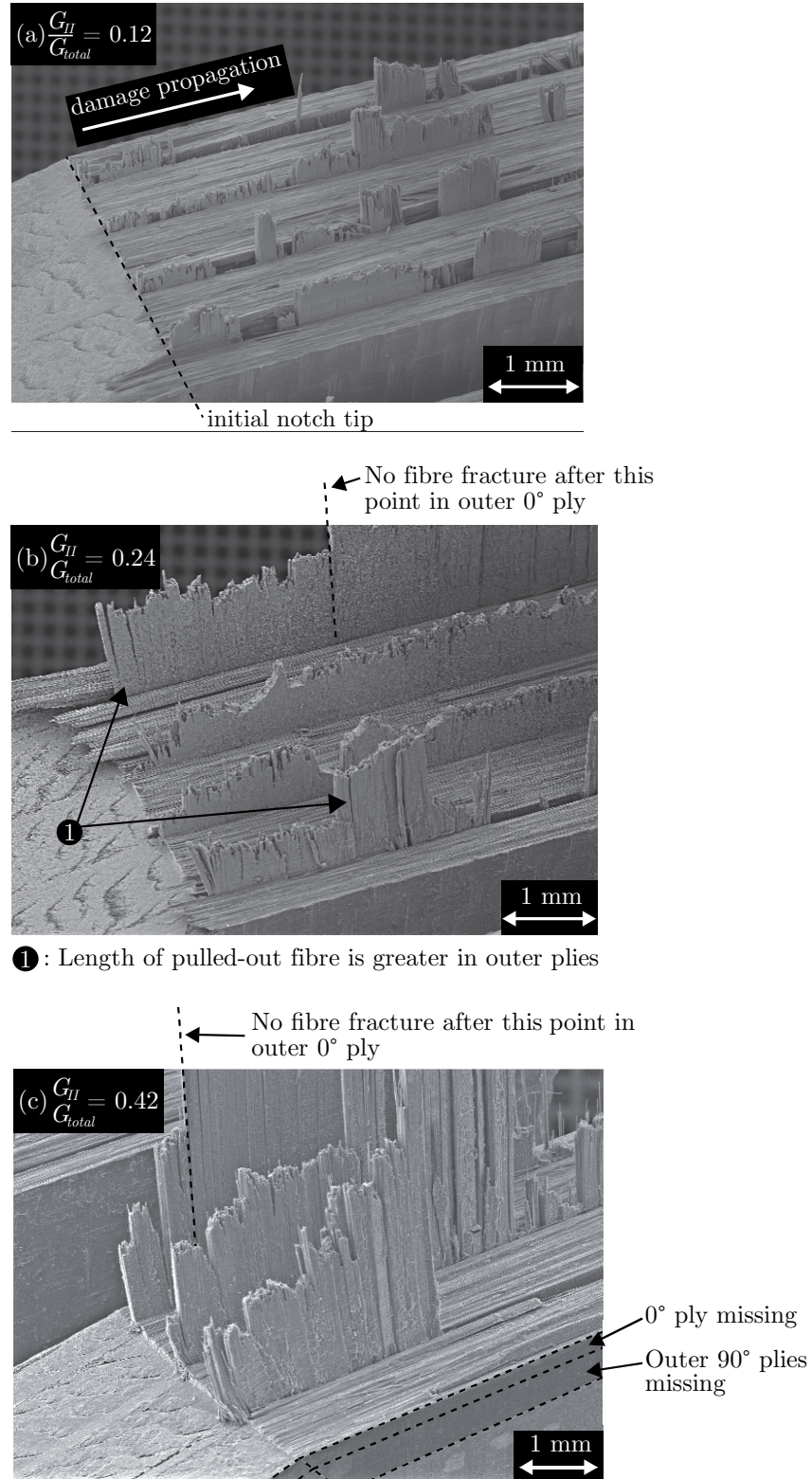


Figure 5.4. Fracture surfaces representative of specimens tested with increasing mode II component.

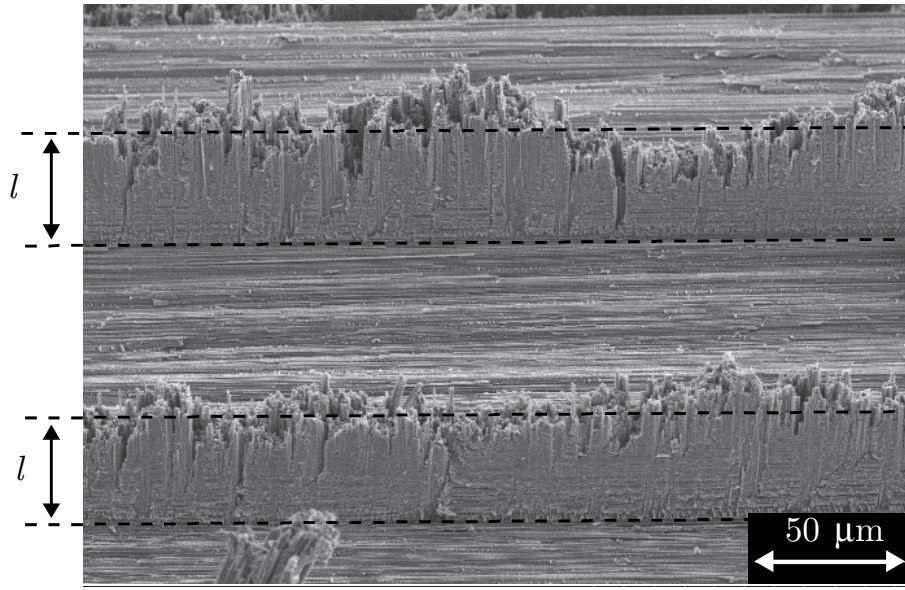


Figure 5.5. SEM image taken from a specimen tested with $G_{II}/G_{total} = 0.12$ highlighting regions where the length of pulled out fibres, indicated as l , is observed to be constant.

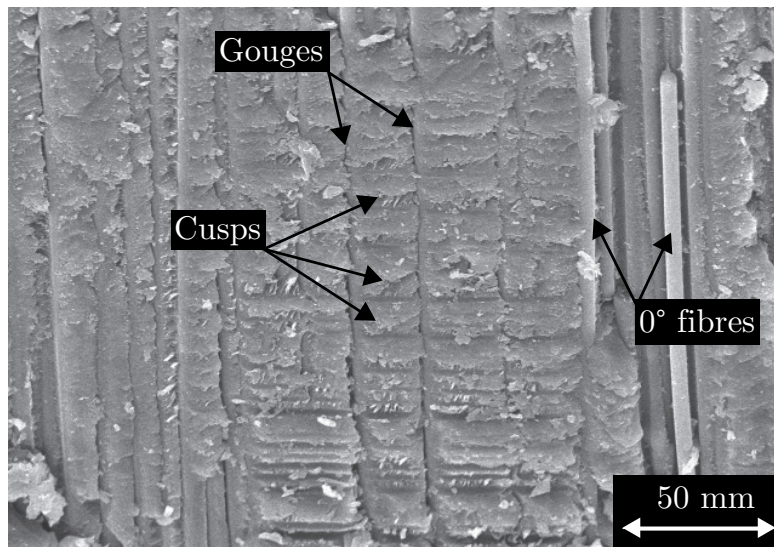


Figure 5.6. Higher magnification image of a block of pulled out fibres from a specimen tested with $G_{II}/G_{total} = 0.12$. The highlighted gouges indicate delamination propagation from left to right in the image.

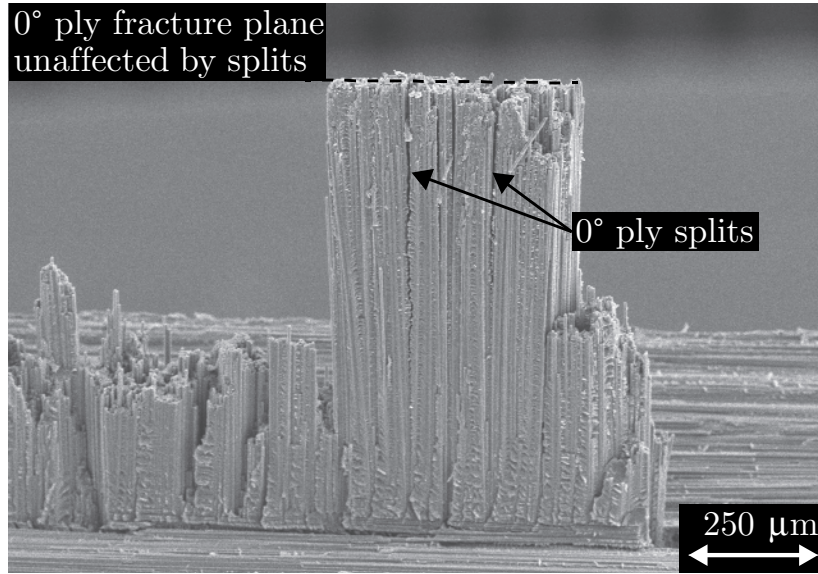


Figure 5.7. Splitting is observed within fibre bundles. The bundle length is unaffected by their presence indicating that they were formed after fibre fracture.

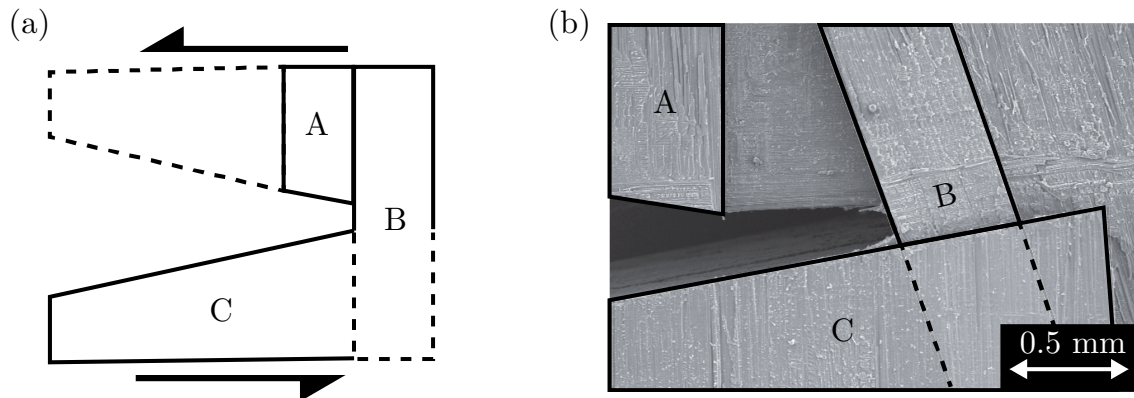


Figure 5.8. Details of the outer 0° ply in specimen tested with $G_{II}/G_{total} = 1.00$ where no fibre fracture has occurred. (a) Schematic showing intact ply before testing. (b) SEM image of 0° ply after test.

5.5. Failure analysis and fractography

The load-displacement curve of each specimen exhibited a small discontinuity well before the maximum load was reached, as indicated in Figure 5.3; which was accompanied by an audible indication of damage initiation during the test. Inspection of specimens revealed that this corresponded to the formation of splits in the outer 90° plies that propagated whilst leaving the embedded 0° plies below to delaminate whilst still intact, as illustrated in the schematic shown in Figure 5.9(a).

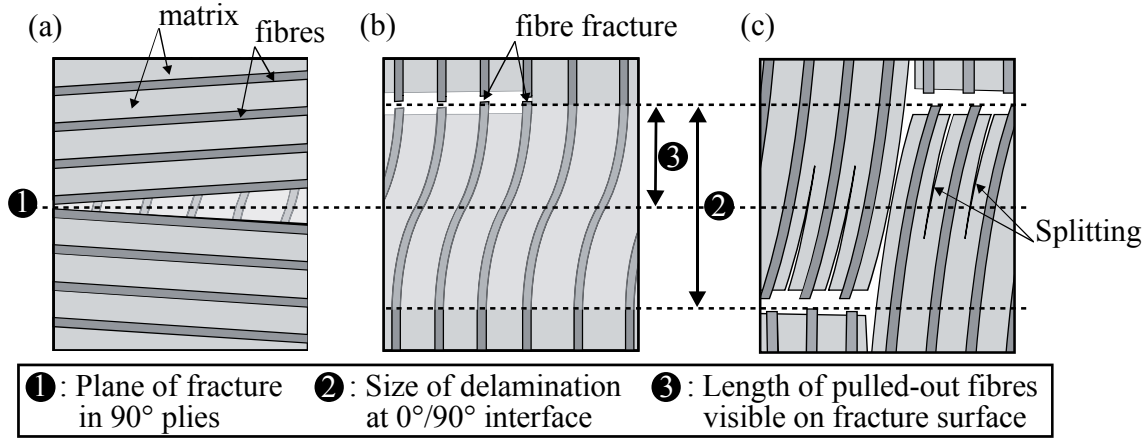


Figure 5.9. Schematics detailing (a) 90° ply fracture visible during testing with 0° ply intact, (b) proposed mechanism of fibre fracture in region of high bending stress at edges of delaminated region, and (c) mechanism of 0° ply splitting in the wake of advancing damage.

Also apparent from the specimen load-displacement curves illustrated in Figure 5.3 is a change in the mechanism governing ultimate failure of the specimen as the proportion of mode II loading is increased. For the lowest mixed mode ratio, shown in Figure 5.3(a), unstable bursts of crack growth typically associated with translaminar failure are observed; above this however, a more stable process of energy dissipation dominates. In fact, only the specimens tested under conditions of $G_{II}/G_{total} = 0.12$ (corresponding to Figure 5.3(a)) exhibited complete translaminar fracture of the 0° plies such that they became completely separated into two halves during the test. Final failure of all other specimens, namely $G_{II}/G_{total} = 0.24, 0.46$ and 1.00 , was governed by mode III delamination at the $0^\circ/90^\circ$ interfaces such that they would be progressively torn apart, resulting in the almost horizontal part of the curve highlighted in Figure 5.3(b). This delamination failure resulted in all four of the once embedded 0° plies of each laminate becoming exposed during the test as

shown in Figure 5.10. Dissection of specimens tested at $G_{II}/G_{total} = 0.24$ and 0.46 revealed fibre fracture had in fact occurred and arrested after a very short distance before the delamination failure became dominant, thus translaminar fracture surfaces could be inspected for all but $G_{II}/G_{total} = 1.00$, where no fibre fracture occurred.

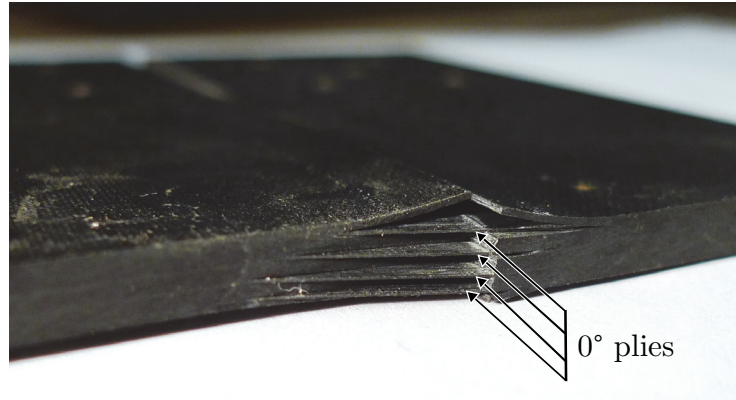


Figure 5.10. All four 0° plies visible at back of specimen.

Perhaps the most striking feature visible on the fracture surfaces of the failed specimens shown in Figure 5.4 is the increase in the length of pulled-out fibre bundles with increasing proportion of mode II loading. Matrix gouges and cusps [64], typically observed on delamination fracture surfaces between plies with a relative orientation of 90° , can be seen upon closer inspection of the pulled-out fibre bundles in Figure 5.6. These features reveal that delaminations form at $0^\circ/90^\circ$ interfaces prior to fibre fracture and therefore suggest that the lengths of fractured fibre visible in the SEM micrographs of Figure 5.4 are a direct indication of the delamination size. The lengths of fractured fibre are observed to be constant along the crack path in several regions, as highlighted in Figure 5.5, indicating that their failure was conditioned by the presence of the delamination front in accordance with Figure 5.9(b): as the $0^\circ/90^\circ$ interface is delaminated over the length indicated by ②, the once embedded 0° ply is left unconstrained and subject to large shear stresses as it acts as a bridge between the two halves of the specimen. The lateral displacement of the specimen halves induces locally high bending stresses in the region where the fibre is no longer constrained to its neighbouring ply, as indicated in Figure 5.9(b);

this stress concentration eventually causes failure of the fibre leaving fibre blocks of length indicated by ❸ protruding from the fracture plane.

The size of the delaminations forming at $0^\circ/90^\circ$ ply interfaces prior to fibre fracture, evident from the lengths of pulled out fibre blocks on the fracture plane, is observed to increase with the increased proportion of mode II loading (Figure 5.4), furthermore delaminations become more extensive in the outer plies as indicated by ❶ in Figure 5.4(b). This would be expected if the fibres fractured following the mechanism outlined in Figure 5.9(b), as an increase in the mode II component of loading would increase the shear deformation that the specimen is subject to, subsequently driving the delamination further before the necessary stress concentration in the fibre is reached to cause its fracture. Longer lengths of fibre are subsequently left on the fracture plane, particularly in the unconstrained surface plies which have a higher propensity to delaminate.

In Figure 5.4(b) and (c), it can be seen that translaminar failure has arrested at the points indicated and thereafter the remaining length of the test-section has failed via mode III delamination. This indicates a strongly rising R-curve associated with the 0° plies alone, the plateau of which cannot be reached due to the mode III delamination failure process being more energetically favourable.

SEM micrographs indicate that there are further mechanisms of energy dissipation in the wake of the advancing damage. Numerous splits can be observed within the blocks of pulled-out fibre bundles in the images of Figure 5.4; these are highlighted in more detail in Figure 5.6. The fact that the fracture plane of bundles of fibres is unaffected by the presence of these splits indicates that they occur after the fibres have fractured, in the wake of the advancing damage. It is thought that this is another damage mechanism introduced by the mode II component of loading and is caused through the unavoidable sliding and contact of neighbouring blocks of pulled-out fibres, as shown in Figure 5.9(c).

In summary, the total size of the process zone associated with translaminar fracture is observed to increase with the introduction of a mode II component of loading. At the very tip of the advancing damage front, 90° ply splits propagate as

illustrated in Figure 5.9(a). These splits subsequently act as initiation sites for delaminations which propagate until fibre fracture occurs in accordance with Figure 5.9(b). In the wake of the advancing damage, the opposing crack faces constrain the movement of pulled-out fibre blocks causing further splitting as shown Figure 5.9(c). In the next section, measurements of fracture toughness are obtained for the tested specimens, where possible, to investigate what effect these mechanisms of energy dissipation have on the laminate R-curves compared to those obtained from pure mode I testing.

5.6. Fracture Toughness Characterisation

The fracture toughness for specimens which exhibited fracture in all 0° plies, namely $G_{II}/G_{total} = 0.12$ and 0.24 , is characterised in this section and also compared to the mode I results for IM7/8552 that were obtained in Chapter 3 of this thesis. The goal is to quantify the impact that the failure processes described in the previous section, introduced by the mode II component of loading, have on the measured R-curves.

The lack of any analytical solutions for the mechanical response and energy release rates of the MMCT specimen required that finite element (FE) analysis be used for the data reduction process. A specimen with the boundary conditions shown in Figure 5.11 was modelled using FE [80] to generate compliance calibration curves. The applied proportions of the energy release rates, G_I and G_{II} , were extracted using the virtual crack closure technique for a range of crack lengths under each loading configuration. The mesh used consisted of square shell elements (S8R5) of length 0.5 mm, the actual geometry of the notch and holes were not modelled as their presence has been found to not affect the accuracy of the data reduction scheme for a similar specimen arrangement [27]. The loading fixture was represented using rigid body constraints shown as dashed lines in Figure 5.11. The material properties used in the model are presented in Table 4.1.

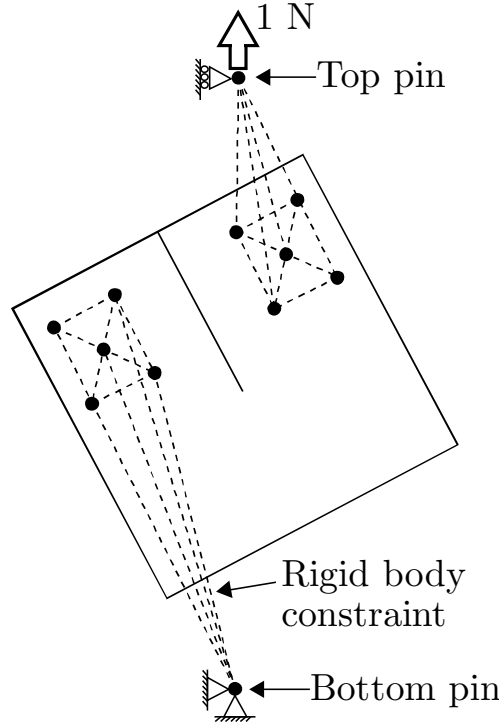


Figure 5.11. FE model used to simulate the experimental setup shown in Figure 5.2.

The critical strain energy release rate of the laminate, G_c , can be calculated if the rate of change of specimen compliance, C , with crack length, a is known:

$$G_c = \frac{P_c^2}{2t} \frac{dC}{da} \quad (5.1)$$

where P_c is the critical load causing extension of the main crack and t is the specimen thickness.

Compliance calibration curves, shown in Figure 5.12, for a specimen with thickness $t = 1$ mm with an applied load of $P = 1$ N were generated such that the critical strain energy release rates of tested specimens could be calculated from their compliance, critical loads, and geometry alone. The data fits shown in Figure 5.12 were produced by using a function of the form

$$t \cdot C = (\alpha a + \beta)^\chi \quad (5.2)$$

where α and β and χ were calculated to best fit the data. By differentiating equation (5.2) with respect to a and substituting into equation (5.1), the critical strain energy release rate of the laminate can be calculated as:

$$G_c = \frac{P_c^2}{2t^2} \chi \alpha (tC)^{\frac{\chi-1}{\chi}} \quad (5.3)$$

where C is the compliance of the specimen at $P = P_c$. For plotting R-curves, an effective crack length was determined from C by rearranging equation (5.2).

A disadvantage of the MMCT configuration is that the proportions of G_I and G_{II} change slightly with crack length, as shown in Figure 5.13, therefore the values of G_{II}/G_{total} mentioned in this section correspond to initiation only.

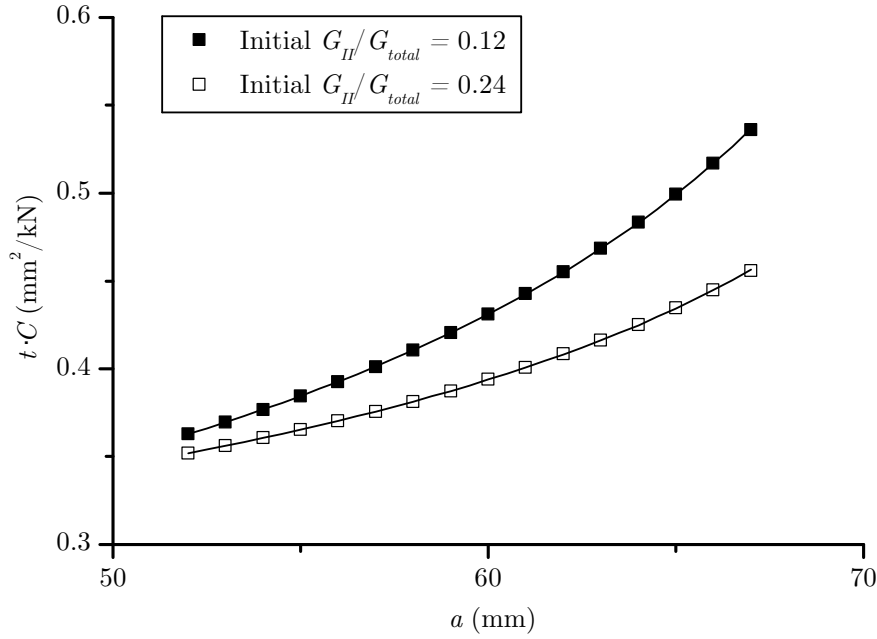


Figure 5.12. Compliance calibration curves obtained from FE using specimen of thickness $t = 1$ mm corresponding to initial $G_{II}/G_{total} = 0.12$ and 0.24 .

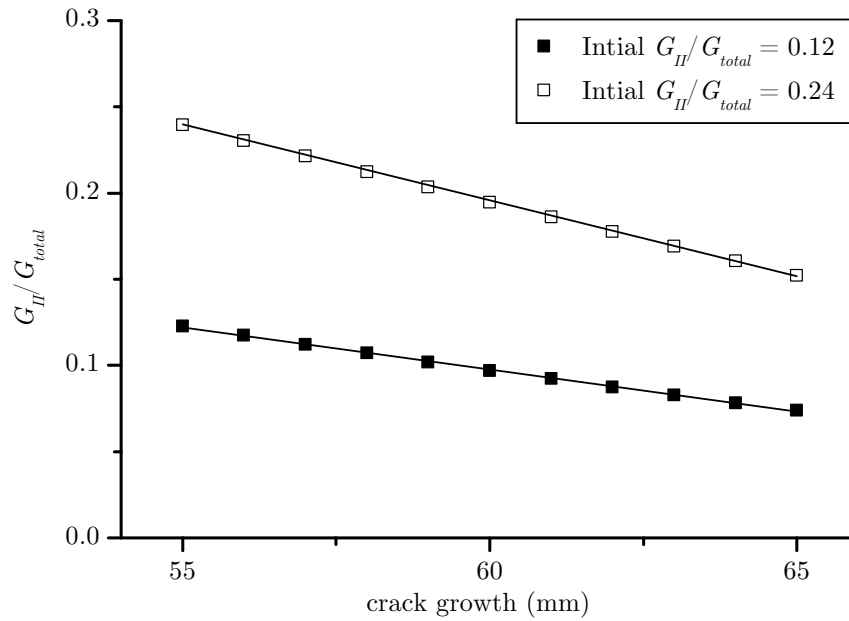


Figure 5.13. Variation of G_{II}/G_{total} with crack length, a , for each of the loading configurations used.

5.7. R-curves

The validity of data obtained from fracture toughness testing depends on, amongst other factors, the size of the damage zone associated with crack growth relative to the size of the specimen being tested. Post-mortem C-scans, shown in Figure 5.15 for both of the G_{II}/G_{total} configurations for which toughness data is calculated, reveal that the size of the damage zone becomes relatively large after small amounts of crack growth. Only data points obtained early on in the tests, where the damage zone is still of a reasonable size, were plotted on the R-curves.

In order to extract the valid data from each specimen, a portion of its post-mortem specimen C-scan, as indicated in Figure 5.14, was superimposed upon its corresponding R-curve, as shown in Figure 5.15. Each data point was then individually assessed, using the size of the damage zone as a reference. Data points were disregarded if deemed to be associated with an area of the specimen exhibiting extensive damage or an area where the damage can be observed to be growing considerably (admittedly, this method has some degree of subjectivity).

The resulting R-curves, containing data deemed to be valid for $G_{II}/G_{total} = 0.12$ and 0.24 , are shown in Figure 5.16. In order to directly compare the results obtained here with those obtained from mode I tests of the same material system published elsewhere [29], the 0° ply-level toughness was extracted using the method outlined in [26]. The R-curve trends for $G_{II}/G_{total} = 0.12$ and 0.24 are plotted in Figure 5.17.

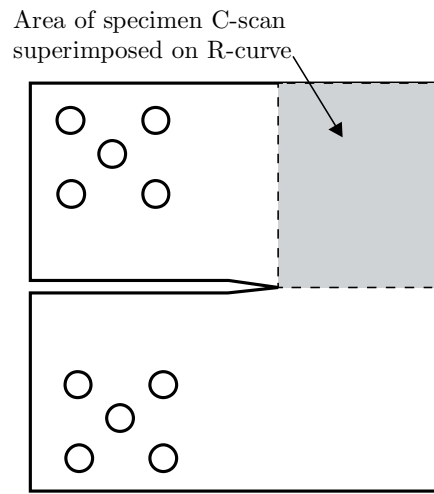


Figure 5.14. Specimen schematic detailing area of specimen C-scan superimposed on the R-curves of Figure 5.15

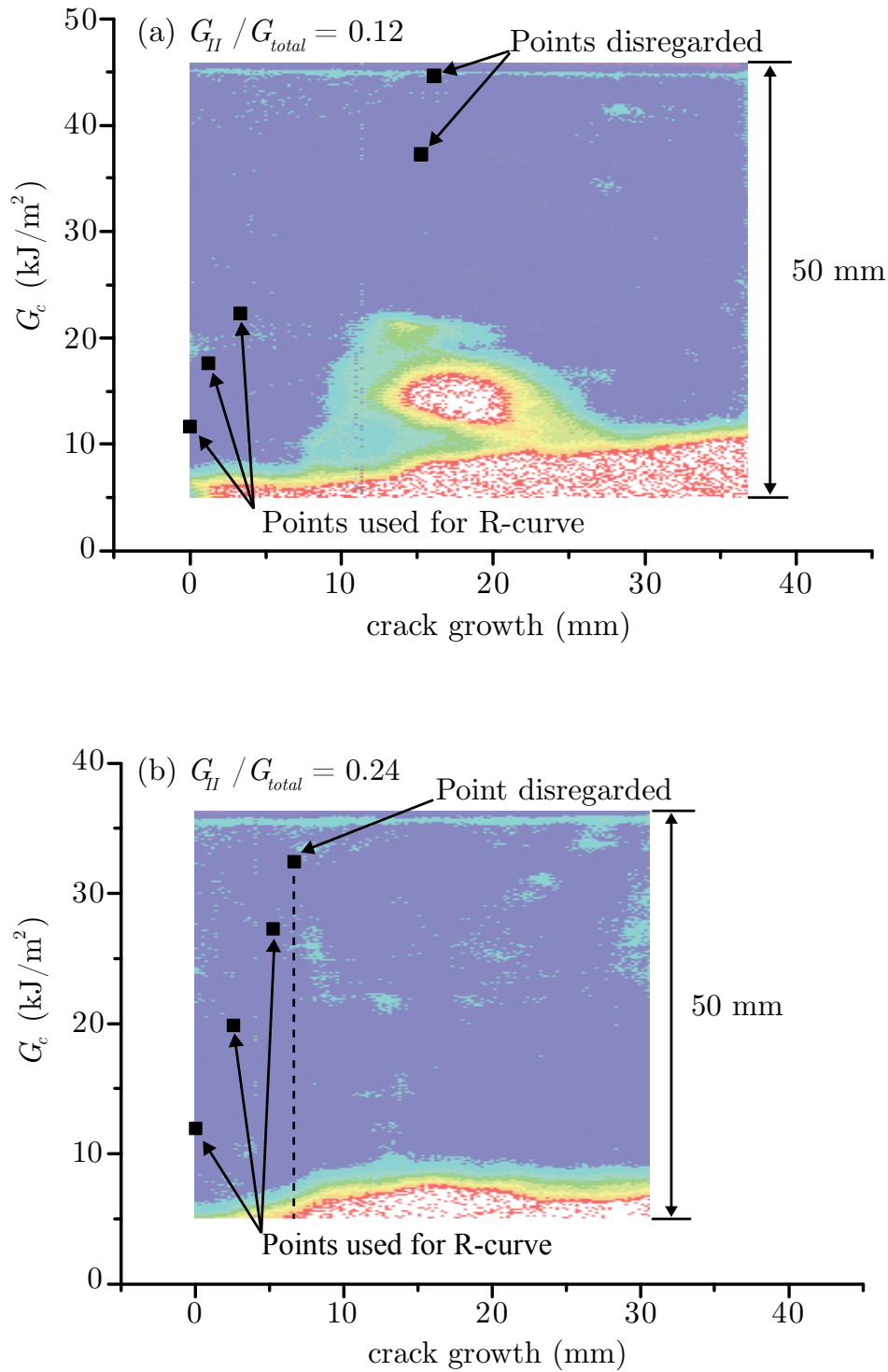


Figure 5.15. R-curve of single specimens superimposed on C-scans detailing damage formed during test for initial G_{II}/G_{total} ratios of (a) 0.12, (b) 0.24. The points used in Figure 5.16 are indicated. The C-scans have been resized to fit the plots, the actual dimensions are indicated.

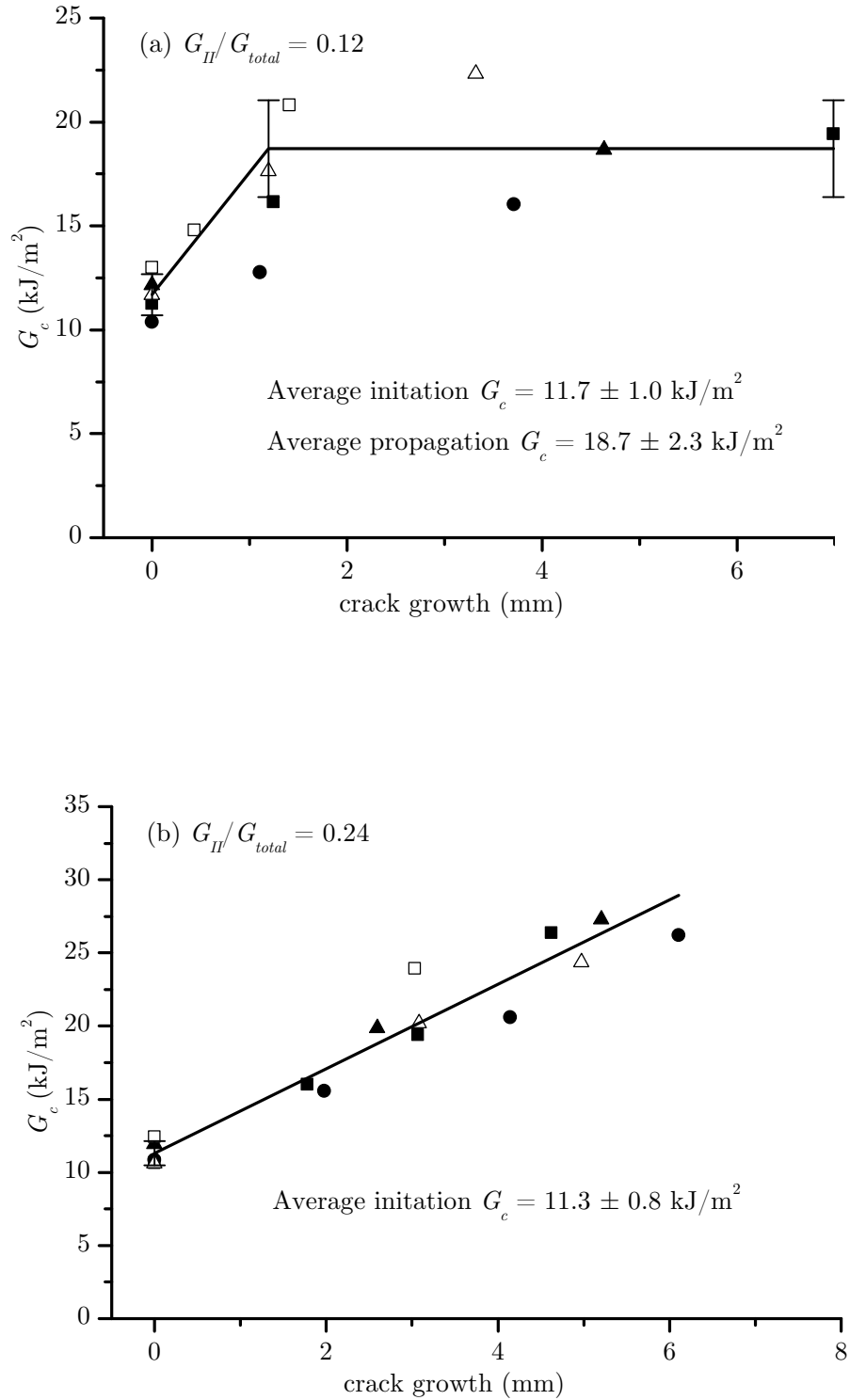


Figure 5.16. R-curves, with data obtained using method detailed in Figure 5.15, obtained from testing specimens with initial G_{II}/G_{total} of (a) 0.12 and (b) 0.24.

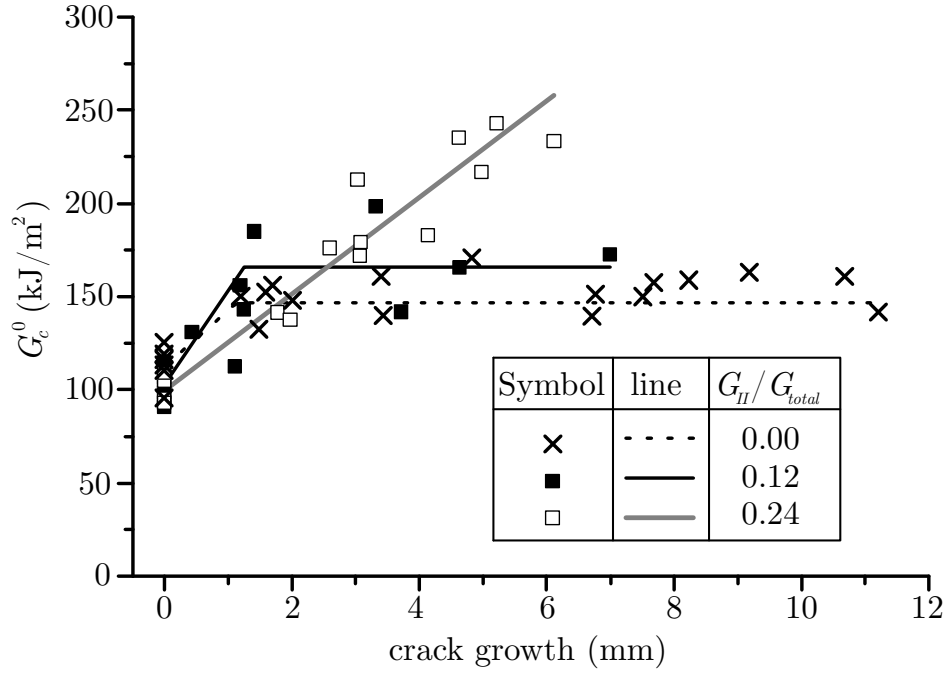


Figure 5.17. Comparison of R-curve data fits for mode I and mixed mode cases. The plotted fracture toughness values correspond to those for the 0° plies alone.

5.8. Discussion

The R-curve for $G_{II}/G_{total} = 0.12$, shown in Figure 5.15(a) suggests that the damage zone develops over approximately 1.5 mm before a plateau propagation toughness is reached. This is consistent with the SEM image of Figure 5.4(a), where the pull-out lengths (and therefore the extent of delamination) can be observed to have reached a consistent size within this amount of crack growth. The fact that a plateau of critical strain energy release rate has been reached suggests that the method for R-curve generation outlined in Figure 5.15 has provided sensible values in this case. Figure 5.16(a) indicates that despite more extensive delaminations and the presence of energy dissipating mechanisms such as the fibre splitting indicated in Figure 5.7 for $G_{II}/G_{total} = 0.12$, no significant increase in fracture toughness relative to the pure mode I case is observed.

For $G_{II}/G_{total} = 0.24$, the R-curve shown in Figure 5.16(b) suggests the damage zone develops more gradually, relative to the mode I and $G_{II}/G_{total} = 0.12$ cases, over several millimetres and with no plateau being reached. However, the SEM observations made for $G_{II}/G_{total} = 0.24$ contradict this interpretation; the long blocks of pulled out fibres present on the fracture surface suggest that delaminations much larger than those observed for lower proportions of mode II loading form at the notch tip prior to fibre failure. These damage mechanisms should be reflected in the initial portion of the R-curve being steeper for $G_{II}/G_{total} = 0.24$ than for the $G_{II}/G_{total} = 0.00$ and 0.12 . The disparity between what is observed on the fracture surfaces of failed specimens and the measured R-curve suggests that the critical strain energy release rate in this case should be considered as a first approximation.

5.9. Conclusions

Mixed mode translaminar failure of CFRP laminates was investigated using modified compact tension specimens. The introduction of a mode II component of loading saw the damage zone associated with translaminar fracture grow significantly due to an increased amount of delamination at $0^\circ/90^\circ$ ply interfaces. Fibre fracture was found to initiate at the delamination fronts between 0° and 90° plies, resulting in long bundles of fractured fibre being observed on the fracture surfaces of failed specimens. The size of the damage zone was also observed to extend in the wake of the advancing failure process as the opposing crack faces constrained blocks of fibre bundles causing them to split.

For low proportions of mode II loading ($G_{II}/G_{total} = 0.12$), R-curves were obtained that indicated a similar damage zone size to that observed in mode I testing. The mechanisms of energy dissipation observed via SEM that are not exhibited in mode I tests did not account for an increase in measured G_c .

Reliable R-curves could not be obtained from specimens tested with higher proportions of G_{II} , due to the size of the damage zone becoming inappropriately large for fracture mechanics testing using the chosen specimen configuration.

5.10. Publications

The work presented within this chapter has been submitted to Composite Structures as a paper entitled: “Mixed-mode translaminar fracture of CFRP: Failure analysis and Fractography”.

Chapter 6

Conclusions

6.1. Mode I translaminar tensile failure

6.1.1. Data Reduction

Six methods of data reduction for CT testing have been investigated with the aim of finding the ideal strategy: the area method, J-Integral/VCCT, ASTM E399, compliance calibration, and modified compliance calibration. The results for all data reduction schemes were fairly consistent, a modified experimental compliance calibration (MCC) method, using an effective crack length determined through compliance measurements, was deemed to be the most appropriate data reduction scheme, as it gives consistent results and does not require the use of an optically measured crack length or FE.

An undesirable amount of scatter (up to 10.2 % for the MCC method) was still present within the results. It was thought that this can be improved upon by changing the proportions of the specimen, namely by increasing the width.

6.1.2. Size and lay-up effects

Size and lay-up effects on the measured translaminar fracture toughness of the 0° plies in CFRP laminates were investigated using the compact tension specimen. Scaling specimen in-plane dimensions had no significant effect on measured 0° ply toughness. Doubling the specimen thickness, by doubling the number of $[90/0]$ sub-laminates, was found to slightly decrease the toughnesses measured at initiation and propagation, though no firm conclusion was made as to the cause of this.

Lay-up was investigated by testing laminates with blocked 0° and 90° plies; the apparent toughness of the fibre tensile failure mode was found to increase when the thickness of the 0° layer was increased. The increase in toughness was attributed to an increase in fibre pull-out for blocked 0° specimens, as observed by SEM imaging.

6.1.3. Notch tip radius

The mode I translaminar fracture toughness associated with the initiation of fracture in the 0° plies within a laminate was found to be independent of notch root radius for $\rho \leq 250 \text{ }\mu\text{m}$. Above this value, an apparent initiation toughness was measured, implying that the critical notch root radius was in the range $250 \text{ }\mu\text{m} \leq \rho_c < 750 \text{ }\mu\text{m}$.

Propagation toughness was found to be independent of initial notch root radius for the tested configurations ($15 \text{ }\mu\text{m} \leq \rho \leq 750 \text{ }\mu\text{m}$), despite the fact that for larger values of ρ the position of the fracture plane differed from ply-to-ply suggesting an increased energy dissipation over a larger surface area.

Comparison of the fracture surfaces of failed specimens reveals that the extent of damage within the 90° plies has no effect on the laminate critical strain energy release rate measured at initiation thus there is no interaction of failure modes between neighbouring 90° and 0° plies.

The results of this work serve as a guide to experimentalists. When preparing specimens, meaningful values of fracture toughness can be obtained for this material system provided that a notch with $\rho = 250 \text{ }\mu\text{m}$ or smaller is used.

6.2. Mode I translaminar compressive failure

A simple method is presented in chapter 4 for the measurement of the fracture toughness associated with the initiation of fibre compressive failure. The fracture toughness can be directly determined from unidirectional specimens without any intralaminar splitting and without having to make assumptions about interaction of damage in non 0° plies.

Fibre compressive failure was seen to initiate in the form of a shear crack at 45° to the direction of pure mode I crack growth; the shear crack later forms a single in-plane kink band which propagates further into the specimen. Information about energy dissipation due to propagation of these fracture mechanisms cannot be readily derived from the data obtained from testing due to their complex nature. The data obtained from these tests enables the prediction of damage initiation for composite structures under pure mode I compressive loadings.

6.3. Mixed mode I/II translaminar tensile/shear failure

Mixed mode translaminar failure of CFRP laminates was investigated using modified compact tension specimens. The introduction of a mode II component of loading saw the damage zone associated with translaminar fracture grow significantly due to an increased amount of delamination at $0^\circ/90^\circ$ ply interfaces. Fibre fracture was found to initiate at the delamination fronts between 0° and 90° plies, resulting in long bundles of fractured fibre being observed on the fracture surfaces of failed specimens. The size of the damage zone was also observed to extend in the wake of the advancing failure process as the opposing crack faces constrained blocks of fibre bundles causing them to split.

For low proportions of mode II loading ($G_{II}/G_{total} = 0.12$), R-curves were obtained that indicated a similar damage zone size to that observed in mode I testing. The mechanisms of energy dissipation observed via SEM that are not exhibited in mode I tests did not account for an increase in measured G_c .

Reliable R-curves could not be obtained from specimens tested with higher proportions of G_{II} , due to the size of the damage zone becoming inappropriately large for fracture mechanics testing using the chosen specimen configuration.

Chapter 7

Future work

7.1. Mode I translaminar tensile failure

The aim of chapter 3 was to produce a well-defined testing procedure for the measurement of fracture toughness associated with mode I translaminar tensile failure. This has been achieved, however, only two material systems have been characterised; IM7/8552 and T300/920. The measured toughness of the former was over twice that measured for the latter; this fact indicates that the loads that the specimen is subject to can vary considerably between material systems. Testing of much tougher material systems could result in specimen buckling, hole failure, or potentially compressive failure at the specimen back. Furthermore, the damage zone associated with translaminar failure could be much larger in a tougher material system, making the current geometry inappropriate for fracture toughness characterisation.

In order for the test method towards closer to standardisation it would be necessary to have a universal specimen geometry that would work for any material system. For this reason, the next step in this area of research would be to characterise the toughest known materials systems to see if the specimen is

appropriate. The results of these tests may result in the need to perform a parametric study to ensure that failure occurs as intended and not via any of the aforementioned failure modes.

7.2. Mode I translaminar compressive failure

Measurement of the fracture toughness associated with the onset of fibre failure has been successfully accomplished, however, a method for characterisation of the R-curve is still lacking.

Characterisation of the R-curve associated with fibre compressive failure is not a trivial task, the main obstacle to overcome at this stage is accurate quantification of the extent of damage within the specimen and correlation of this with experimental load-displacement data. Scaling down the specimen size may result in the failure process having a more discernible effect on the load displacement curve, however, any compliance based method of data reduction must be able to account for the fact that a load is still being transmitted across the faces of the advancing crack.

Perhaps the best way to proceed would be to recreate the test within the chamber of the SEM using a scaled down specimen. Damage formation and propagation may have more of an obvious effect on the load displacement curve of a smaller specimen, and the length of the failed zone could be accurately monitored during the test. If it is possible to correlate crack length with the load-displacement curve in this case, an R-curve could be generated. The main problem anticipated here is that fibres at the surface of a specimen will have a tendency to deflect out of plane during the failure process; so what is being monitored at the specimen surface may not necessarily be indicative of the damage within the specimen. The degree to which the surface damage reflects the interior damage could, however, be checked after the test by polishing and optical microscopy of the failed specimens.

7.3. Mixed mode I/II translaminar tensile/shear failure

Preliminary results from characterisation of the translaminar fracture toughness of the 0° plies under mixed mode loading has shown that the method can only be considered to be in its developmental stages.

It has been seen that the damage zone that forms prior to extension of the main crack greatly increases with increasing proportion of mode II loading, and that the current specimen size and lay-up may be inappropriate for measurement of the critical strain energy release rate. A suitable specimen size and lay-up needs to be determined before the mixed mode fracture toughness can be accurately characterised. This would necessitate a size study similar to that performed in Chapter 3 of this thesis. Furthermore, a specimen with unchanging proportions of G_I and G_{II} is highly desirable for R-curve determination.

It is clear that as the mode II component of loading is increased, there is a change in failure mode, instead of translaminar fracture, mode III delamination at $0^\circ/90^\circ$ ply interfaces dominates. It would be necessary from further tests to identify the loading conditions under which this failure mode transition dominates, which would represent the limit after which mixed-mode translaminar fracture toughness is no longer a useful material property.

Appendix A.

Analytical and numerical solution for the mode II energy release rate of a 45° crack in an isotropic material

If an infinitesimally small crack forms at an angle α from the plane of the main crack, as shown in Figure A.1, the local stress intensity factor differs from that of the main crack. For an isotropic material, the local mode I and mode II stress intensity factors, k_I and k_{II} respectively, are related to the global stress intensity factors by, K_I and K_{II} respectively, by

$$k_I(\alpha) = K_I \left[\frac{3}{4} \cos\left(\frac{\alpha}{2}\right) + \frac{1}{4} \cos\left(\frac{3\alpha}{2}\right) \right] - K_{II} \frac{3}{4} \left[\sin\left(\frac{\alpha}{2}\right) + \sin\left(\frac{3\alpha}{2}\right) \right] \quad (\text{A.1})$$

$$k_{II}(\alpha) = K_I \frac{1}{4} \left[\sin\left(\frac{\alpha}{2}\right) + \sin\left(\frac{3\alpha}{2}\right) \right] - K_{II} \left[\frac{1}{4} \cos\left(\frac{\alpha}{2}\right) + \frac{3}{4} \cos\left(\frac{3\alpha}{2}\right) \right] \quad (\text{A.2})$$

In order to validate the method used to produce Figure 4.6 and Figure 4.7, the model shown in Figure 4.3 was run using isotropic material properties ($E = 1000$ MPa, $\nu = 0.3$) to obtain the J -integral for $a = 5$ mm. Because the loading is pure

mode I, the second term in both equations (A.1) and (A.2) disappears, thus the local mode I and mode II energy release rates, g_I and g_{II} can be obtained from

$$g_I(\alpha) = J \left[\frac{3}{4} \cos\left(\frac{\alpha}{2}\right) + \frac{1}{4} \cos\left(\frac{3\alpha}{2}\right) \right]^2 \quad (\text{A.3})$$

$$g_{II}(\alpha) = J \frac{1}{16} \left[\sin\left(\frac{\alpha}{2}\right) + \sin\left(\frac{3\alpha}{2}\right) \right]^2 \quad (\text{A.4})$$

Figure A.2 illustrates $g_{II}(45^\circ)$, in terms of $\bar{G}_{II}$ (the normalised energy release rate outlined in Chapter 4), obtained using VCCT for various 45° cracks of length l . Also shown is the point corresponding to $l = 0$, obtained using equation (A.4). As can be seen, the VCCT values correlate well with the analytical solution, with a linear regression providing a value of $\bar{G}_{II}(l = 0)$ 4.5 % less than that obtained from equation (A.4).

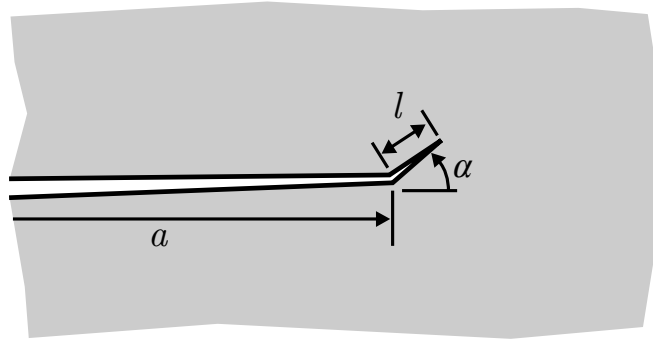


Figure A.1. Infinitesimally small kink crack at an angle α to the main crack.

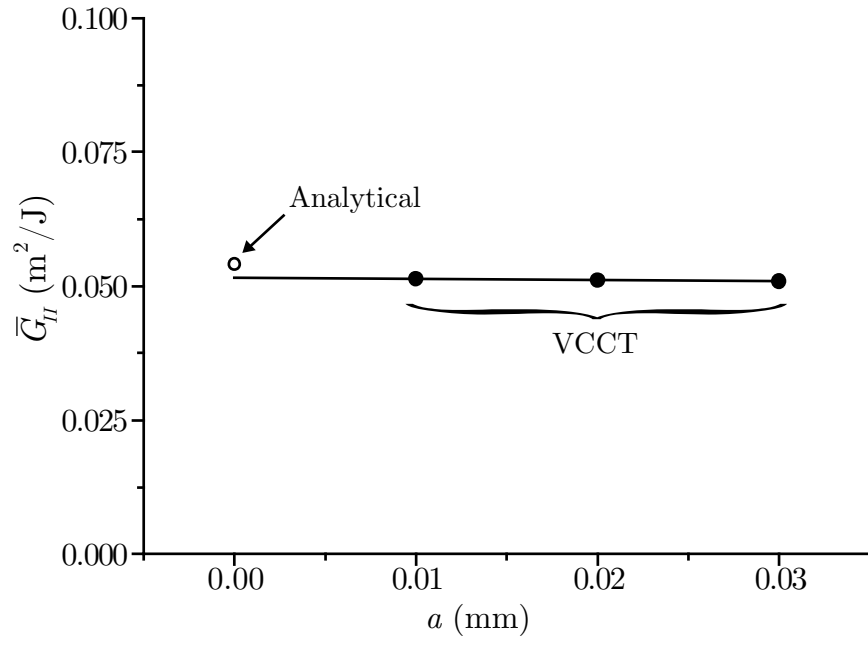


Figure A.2. Mode II energy release rates for a 45° kink crack of length l obtained from VCCT and equation (A.4).

List of references

- [1] Pinho ST, Iannucci L, Robinson P. Formulation and implementation of decohesion elements in an explicit finite element code. *Composites Part A: Applied Science and Manufacturing*. 2006;37(5):778-789.
- [2] Pinho ST, Iannucci L, Robinson P. Physically based failure models and criteria for laminated fibre-reinforced composites with emphasis on fibre kinking. Part II: FE implementation. *Composites Part A: Applied Science and Manufacturing*. 2006;37(5):766-777.
- [3] Moës N, Belytschko T. Extended finite element method for cohesive crack growth. *Engineering Fracture Mechanics*. 2002;69(7):813-833.
- [4] O'Brien TK, Martin RH. Round robin testing for mode I interlaminar fracture toughness of composite materials. *Journal of composites technology and research*. 1993;15(4):269-281.
- [5] Davies P. Round Robin analysis of G_{Ic} interlaminar fracture test. *Applied Composite Materials*. 1996;3(3):135-140.
- [6] Davies P, Kausch HH, Williams JG, Kinloch AJ, Charalambides MN, Pavan A, et al. Round-robin interlaminar fracture testing of carbon-fibre-reinforced epoxy and PEEK composites. *Composites Science and Technology*. 1992;43(2):129-136.
- [7] Hojo M, Kageyama K, Tanaka K. Prestandardization study on mode I interlaminar fracture toughness test for CFRP in Japan. *Composites*. 1995;26(4):243-255.
- [8] ASTM. D5528 Standard test method for mode I interlaminar fracture toughness of unidirectional fiber-reinforced polymer matrix composites. 2007.

- [9] ISO. ISO 15024:2001 Determination of mode I interlaminar fracture toughness, GIC, for unidirectionally reinforced materials. 2001.
- [10] JIS. K 7086-1993 Testing methods for interlaminar fracture toughness of carbon fibre reinforced plastics. 1993.
- [11] ASTM. D 6671/D 6671M-06 Mixed Mode I-Mode II Interlaminar Fracture Toughness of Unidirectional Fiber Reinforced Polymer Matrix Composites. 2006.
- [12] Martin RH. Interlaminar fracture characterization: A current review. NASA contractor report 187573. 1991.
- [13] Sela N, Ishai O. Interlaminar fracture toughness and toughening of laminated composite materials: a review. *Composites*. 1989;20(5):423-435.
- [14] O'Brien TK. Interlaminar fracture toughness: the long and winding road to standardization. *Composites Part B: Engineering*. 1998;29(1):57-62.
- [15] Szekrényes A. Overview on the experimental investigations of the fracture toughness in composite materials, MET-020507-A. *Hungarian Electronic Journal of Sciences*, <http://hejszehu>. 2002.
- [16] Davies P, Blackman BRK, Brunner AJ. Standard Test Methods for Delamination Resistance of Composite Materials: Current Status. *Applied Composite Materials*. 1998;5(6):345-364.
- [17] Honjo K. Fracture toughness of PAN-based carbon fibres estimated from strength-mirror size relation. *Carbon*. 2003;41:979-984.
- [18] Gutkin R, Pinho ST, Robinson P, Curtis PT. On the transition from shear-driven fibre compressive failure to fibre kinking in notched CFRP laminates under longitudinal compression. *Composites Science and Technology*. 2010;70:1223-1231.
- [19] Gutkin R, Laffan MJ, Pinho ST, Robinson P, Curtis PT. Modelling the R-curve effect and its specimen-dependence. *International Journal of Solids and Structures*. 2011;48(11-12):1767-1777.
- [20] Pinho ST, Robinson P, Iannucci L. Developing a four point bend specimen to measure the mode I intralaminar fracture toughness of unidirectional laminated composites. *Composites Science and Technology*. 2009;69:1303-1309.

- [21] Harris CE, Morris DH. Fracture behaviour of thick, laminated graphite/epoxy composites. NASA Contractor Report 3784. 1984.
- [22] Harris CE, Morris DH. A comparison of the fracture behaviour of thick laminated composites utilizing compact tension, three-point bend, and center-cracked tension specimens. *Fracture Mechanics: Seventeenth Volume*, ASTM STP 905. 1986:124-135.
- [23] Underwood JH, Kortschot MT. Notch-tip damage and translaminar fracture toughness measurements from carbon/epoxy laminates. US Army armament research, development and engineering centre, technical report ARCCB-TR-94010. 1994.
- [24] Underwood JH, Kortschot MT, Lloyd WR, Eidinoff HL, Wilson DA, Ashbaugh N. Translaminar fracture toughness test methods and results from interlaboratory tests of carbon/epoxy laminates. *Fracture Mechanics: 26th Volume*, ASTM STP 1256. 1995.
- [25] Masters JE. Translaminar fracture toughness of a composite wing skin made of stitched warp-knit fabric. NASA Contractor Report 201728. 1997.
- [26] Pinho ST, Robinson P, Iannucci L. Fracture toughness of the tensile and compressive fibre failure modes in laminated composites. *Composites Science and Technology*. 2006;66(13):2069-2079.
- [27] Laffan MJ, Pinho ST, Robinson P, Iannucci L. Measurement of the in situ ply fracture toughness associated with mode I fibre tensile failure in FRP. Part I: Data reduction. *Composites Science and Technology*. 2010;70(4):606-613.
- [28] Laffan MJ, Pinho ST, Robinson P, Iannucci L. Measurement of the in situ ply fracture toughness associated with mode I fibre tensile failure in FRP. Part II: Size and lay-up effects. *Composites Science and Technology*. 2010;70(4):614-621.
- [29] Laffan MJ, Pinho ST, Robinson P, McMillan AJ. Translaminar fracture toughness: The critical notch tip radius of 0° plies in CFRP. *Composites Science and Technology*. 2011;72(1):97-102.
- [30] Catalanotti G, Camanho PP, Xavier J, Dávila CG, Marques AT. Measurement of resistance curves in the longitudinal failure of composites using

digital image correlation. *Composites Science and Technology*. 2010;70(13):1986-1993.

[31] Dassios KG, Kostopoulos V, Steen M. Intrinsic parameters in the fracture of carbon/carbon composites. *Composites Science and Technology*. 2005;65(6):883-897.

[32] Reber R, De Haan J, Mayer J, Petitmermet M, Wintermatnel E. Intralaminar Fracture of Weft-Knitted Carbon Fiber Reinforced Peek. *Fracture of Polymers, Composites and Adehsives ESIS Publication*. 2000;27:123-133.

[33] Prewo KM. The effect of ply lay-up sequence on the fracture toughness of boron aluminum. *Journal of Composite Materials*. 1978;12(1):40-52.

[34] Sun CT, Prewo KM. The fracture toughness of boron aluminum composites. *Journal of Composite Materials*. 1977;11(2):164-175.

[35] Jose S, Ramesh Kumar R, Jana MK, Venkateswara Rao G. Intralaminar fracture toughness of a cross-ply laminate and its constituent sub-laminates. *Composites Science and Technology*. 2001;61(8):1115-1122.

[36] Ramesh Kumar R, Jose S, Venkateswara Rao G. Evaluation of intralaminar fracture toughness of angle ply laminate. *Indian Journal of Engineering and Materials Sciences*. 2002;9(4):269-274.

[37] ASTM. E399-06 Standard test method for linear-elastic plane-strain fracture toughness K_{Ic} of metallic materials. 2006.

[38] Tada H, Paris P, Irwin G. *The stress analysis of cracks handbook*. 3rd ed: ASME; 2000.

[39] Slepetz JM, Carlson L. Fracture of Composite Compact Tension Specimens. *Fracture Mechanics of Composites*. 1975;ASTM STP 593:143-162.

[40] Kongshavn I, Poursartip A. Experimental investigation of a strain-softening approach to predicting failure in notched fibre-reinforced composite laminates. *Composites Science and Technology*. 1999;59(1):29-40.

[41] Piascik RS, Newman JC. An extended compact tension specimen for fatigue crack growth and fracture testing. *International Journal of Fracture*. 1996;76(3):R43-R48.

- [42] ASTM. E1922-04 Standard test method for translaminar fracture toughness of laminated polymer matrix composite materials. 2004.
- [43] El-Hajjar R, Haj-Ali R. Mode-I fracture toughness testing of thick section FRP composites using the ESE(T) specimen. *Engineering Fracture Mechanics*. 2005;72(4):631-643.
- [44] Dávila C, Rose C, Camanho P. A procedure for superposing linear cohesive laws to represent multiple damage mechanisms in the fracture of composites. *International Journal of Fracture*. 2009;158(2):211-223.
- [45] Morris DH, Hahn HT. Fracture Resistance Characterization of Graphite/Epoxy Composites. *Composite Materials: Testing and Design (Fourth Conference)*. 1977:5-17.
- [46] Yeow YT, Morris DH, Brinson HF. A correlative study between analysis and experiment on the fracture behavior of graphite/epoxy composites. *Journal of testing and evaluation*. 1979;7(2):117-125.
- [47] Underwood JH, Burch IA, Bandyopadhyay S. Effects of Notch Geometry and Moisture on Fracture Strength of Carbon/Epoxy and Carbon/Bismaleimide Laminates. *Composite Materials: Fatigue and Fracture (Third Volume)*, ASTM STP 1110. 1991:667-685.
- [48] Poe CC, Sova JA. Fracture Toughness of Boron/Aluminium Laminates With Various Proportions of 0° and $\pm 45^\circ$ Plies. NASA TP-1707. 1980.
- [49] Soutis C, Curtis PT, Fleck NA. Compressive failure of notched carbon fibre composites. *Proceedings of the Royal Society*. 1993;440(1909):241-256.
- [50] Mandell JF, Wang SS, McGarry FJ. The Extension of Crack Tip Damage Zones in Fiber Reinforced Plastic Laminates. *Journal of Composite Materials*. 1975;9(3):266-287.
- [51] Toygar ME, Toparli M, Uyulgan B. An Investigation of Fracture Toughness of Carbon/Epoxy Composites. *Journal of Reinforced Plastics and Composites*. 2006;25(18):1887-1895.
- [52] Hallett SR, Wisnom MR. Experimental Investigation of Progressive Damage and the Effect of Layup in Notched Tensile Tests. *Journal of Composite Materials*. 2006;40(2):119-141.

- [53] Vaidya RS, Sun CT. Fracture criterion for notched thin composite laminates. *AIAA Journal*. 1997;35(2):311-316.
- [54] Leach DC, Moore DR. Toughness of aromatic polymer composites reinforced with carbon fibres. *Fibre Science and Technology*. 1985;23(2):131-161.
- [55] Ratcliffe J. Compression Strength Prediction of Impact-Damaged Composite Sandwich Panels. *American Helicopter Society Annual Forum Proceedings*. 2004;60(1):656-672.
- [56] Jackson W, Ratcliffe J. Measurement of fracture energy for kink-band growth in sandwich specimens. *CompTest*. 2004;Conference Proceedings.
- [57] Laffan MJ, Pinho ST, Robinson P, Iannucci L, McMillan A. Measurement of the fracture toughness associated with longitudinal fibre compressive failure of FRP. Submitted 2011.
- [58] Sutcliffe MPF, Fleck NA. Microbuckle propagation in carbon fibre-epoxy composites. *Acta Metallurgica et Materialia*. 1994;42(7):2219-2231.
- [59] Tada H, Paris PC, Irwin GR. *The stress analysis of cracks handbook*. 2000.
- [60] Bažant ZP. Scaling laws in mechanics of failure. *Journal of engineering mechanics*. 1993;119(9):1828-1844.
- [61] Bažant ZP, Daniel IM, Li Z. Size effect and fracture characteristics of composite laminates. *Journal of engineering materials and technology*. 1996;118:317-324.
- [62] Kamat SV, Hirth JP, Zok FW. Effects of notch root radius on crack initiation and growth toughnesses of a cross-ply Ti6Al4V/SiC composite. *Acta materialia*. 1996;44(5):1831.
- [63] Paradkar A, Gogia AK, Kamat SV. Effect of notch root radius on fracture toughness of Ti18Al8Nb alloy. *Materials Science and Technology; MST*. 2009;25(4):515.
- [64] Greenhalgh ES. *Delamination-dominated failures in polymer composites. Failure analysis and fractography of polymer composites*, Abington Hall, Granta Park, Great Abington, Cambridge.: Woodhead Publishing Limited; 2009.

- [65] Sørensen BF, Jacobsen TK. Large-scale bridging in composites: R-curves and bridging laws. *Composites Part A*. 1998;29A:1443-1451.
- [66] Crasto A, Kim R. Compression strengths of advanced composites from a novel mini-sandwich beam. *Sampe Quarterly*. 1991;22:29-39.
- [67] de Groot PJ, Wijnen PAM, Janssen RBF. Real-time frequency determination of acoustic emission for different fracture mechanisms in carbon/epoxy composites. *Composites Science and Technology*. 1995;55(4):405-412.
- [68] Gutkin R, Green CJ, Vangrattanachai S, Pinho ST, Robinson P, Curtis PT. On acoustic emission for failure investigation in CFRP: Pattern recognition and peak frequency analyses. *Mechanical systems and signal processing*. 2011;25(4):1393-1407.
- [69] Melanitis N, Tetlow PL, Galiotis C, Smith SB. Compressional behaviour of carbon fibres. Part II: Modulus softening. *Journal of materials science*. 1994;29(3):786-799.
- [70] Hayes G, Edie D, Kennedy J. The recoil compressive strength of pitch-based carbon fibres. *Journal of materials science*. 1993;28(12):3247-3257.
- [71] Dobb MG, Johnson DJ, Park CR. Compressional behaviour of carbon fibres. *Journal of materials science*. 1990;25(2):829-834.
- [72] Gutkin R, Pinho ST, Robinson P, Curtis PT. Micro-mechanical modelling of shear-driven fibre compressive failure and of fibre kinking for failure envelope generation in CFRP laminates. *Composites Science and Technology*. 2010;70(8):1214-1222.
- [73] Dean G, Mera R. Determination of material properties and parameters required for the simulation of impact performance of plastics using finite element analysis. NPL; 2004.
- [74] Kaddour AS, Hinton MJ. Instructions to contributors of the second world-wide failure exercise (WWFE-II): Part A. 2007.
- [75] Hapke J, Gehrig F, Huber N, Schulte K, Lilleodden ET. Compressive failure of UD-CFRP containing void defects: In situ SEM microanalysis. *Composites Science and Technology*. 2011;71(9):1242-1249.

- [76] Daniel I. Mixed-mode failure of composite laminates with cracks. *Experimental Mechanics*. 1985;25(4):413-420.
- [77] Seif MA, Shahjahan M. Mixed-mode failure of graphite/epoxy composites. *Journal of Engineering materials and technology*. 2001;123:371-376.
- [78] Seif MA, Dasari NB. Effect of combined loading on cracks in graphite/epoxy composites. *Composite Structures*. 2001;52(2001):539-544.
- [79] Richard HA, Benitz K. A Loading Device for the Creation of Mixed Mode in Fracture Mechanics. *International Journal of Fracture*. 1983;22:R55-R58.
- [80] Abaqus 6.9-1. Dassault Systemes Simulia Corp. Rising Sun Mills, 166 Valley Street, Providence, RI 02909-2499, USA.