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Fracture Behaviour of Polyetherimide (PEI) and Interlaminar Fracture of CF/PEI Laminates at Elevated Temperatures

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**Abstract**

To investigate effects of environmental temperature on fracture behaviour of a polyetherimide (PEI) thermoplastic polymer and its carbon fibre (CF/PEI) composite, experimental and numerical studies were performed on compact tension (CT) and double cantilever beam (DCB) specimens under mode-I loading. The numerical analyses were based on 2-D large deformation finite element analyses (FEA). Elevated temperatures greatly released the crack tip triaxiality (constraint) and promoted matrix deformation due to low yield strength and enhanced ductility of the PEI matrix, which resulted in the greater plane-strain fracture toughness of the bulk PEI polymer and the interlaminar fracture toughness of its composite during delamination propagation with increasing temperature. Furthermore, the high triaxiality was developed around the delamination front tip in the DCB specimen, which accounted for the poor translation of matrix toughness to the interlaminar fracture toughness by suppressing the matrix deformation and reducing the plastic energy dissipated in the plastic zone. Especially, at delamination initiation, the weakened fibre/matrix adhesion at elevated temperatures leaded to premature failure of fibre/matrix interface, suppressing matrix deformation and preventing the full utilization of matrix toughness. Consequently, the low interlaminar fracture toughness was obtained at elevated temperatures.
1 Introduction

Delamination is the most serious limiting factor in the full utilization of polymer matrix composite laminates. For this reason, extensive research has been carried out with regard to various aspects including testing variables, processing conditions and material properties in order to characterize and enhance the interlaminar fracture toughness. Matrix toughness has been recognized as a very important factor to control delamination toughness in composite laminates. A comprehensive review on the relationship of matrix fracture toughness to interlaminar fracture toughness was given by Hunston et al. (1) and Bradley (2). Mode-I delamination toughness of composites made from a brittle epoxy was often greater than the neat-resin toughness when additional energy absorbing fracture mechanisms such as fibre bridging and breakage were involved. On the other hand, composites with a ductile matrix showed poor translation of matrix toughness to delamination toughness of the composites. The reason was partially attributed to the constraint effect imposed by the neighboring rigid fibres that limit the evolution of the plastic zone within the matrix resin, experimentally confirmed via observations of delamination micromechanisms and measurements of strain fields around the crack tip (2). Besides, numerical simulations were carried out using finite element analyses (FEA) to evaluate the contribution of matrix toughness to interlaminar fracture toughness by comparing crack-tip stress-strain fields between bulk matrices and composite laminates (3, 4).

The effect of environmental temperature on mode-I and mode-II interlaminar fracture toughness for a carbon fibre reinforced polyetherimide matrix composite (CF/PEI), $G_{IC}$ and $G_{IIC}$, was studied with a special emphasis on the relationship between matrix toughness and delamination fracture toughness at elevated temperatures (5). It was found that matrix toughness was one of important factors in defining temperature-dependent delamination
toughness. In this study, numerical stress analyses in line with experimental investigations were carried out for compact tension (CT) and double cantilever beam (DCB) specimens to examine effects of environmental temperature on fracture behaviour of the polyetherimide (PEI) polymer in the bulk specimen and composite laminate. The numerical study of the near-crack tip fields provided an appropriate description of crack-tip plastic zones in the CT and the DCB specimens, and a better understanding of the translation of matrix toughness to the interlaminar fracture toughness of the composite as a function of temperature.

2 Experimental

2.1 Tensile and Compact Tension Tests for bulk PEI polymers

Commercial grade PEI pellets (ULTEM 1000 supplied by GE Plastics) were used to manufacture 4 and 22 mm thick panels using compression moulding for tensile and compact tension (CT) tests, respectively. Tensile tests were performed in accordance with ASTM D 638-99 (6), except that the uniform section specimen was replaced with an hour-glass specimen of 4 mm thickness, shown in Figure 1a. The hour-glass shaped specimen was designed for determining local stress-strain data for polymers after necking (7). The tensile tests were carried out at a crosshead speed of 2 mm/min at temperatures of RT, 80 and 130°C using an environmental chamber on an Instron 5567 universal testing machine with a computer data acquisition system. At least three specimens were tested for each temperature condition. Strains were measured using a clip gauge of 50 mm in gauge length at RT. At the elevated temperatures, strains measured from the crosshead displacement were calibrated with those from the clip gauge measurement at RT. To measure inhomogeneous or local deformation after necking, a stripe pattern was printed on the specimen surface. The relative displacement in the pattern was measured from the video images recorded during the tests, providing local strains in the waist of the specimen. Engineering stresses were calculated via
dividing the load by the original cross-sectional area. Young’s modulus and the yield strength were determined from the slope of the initial straight line and the maximum stress of the stress-strain curves, respectively, which are summarised in Table 1.

The consolidated PEI panels with a thickness of 22 mm were machined into compact tension (CT) specimens, shown in Figure 1b. The CT specimens were carefully precracked from the machined notch using a razor blade tapping method to produce a sufficiently sharp notch (8), and the length of crack, a/W, was controlled between 0.45 and 0.55. According to ASTM D 5045-99 (9), the CT tests were performed at temperatures of RT, 80 and 130°C with a crosshead speed of 10 mm/min.

2.2 Mode-I Interlaminar Fracture Tests

The composite laminates were prepared using carbon fibre fabric reinforced polyetherimide (CF/PEI) prepreg, which had a 5 harness satin weave configuration with a resin content of 42 wt % (supplied by Ten Cate Advanced Composites, Netherlands). The 22 plies of the prepreg were symmetrically laid up with the side of the fabric of more exposed warp yarns to face each other at the mid-plane, being parallel to the longitudinal direction of the specimens to be cut. Two UPLEX®-R-50 films were placed between the stacked prepreg and aluminum plates to allow easy release from the tooling surface and smooth laminate surfaces after consolidation. A UPLEX®-R-25 film (25μm in thickness) was inserted at the mid-plane of the stacked prepreg to produce an initial delamination in the consolidated laminate panels. The stacked prepreg was preheated at 320°C for 25 minutes, and then consolidated at 320°C for 25 minutes with an applied pressure of 2.5 MPa using a hot press. The laminate was cooled under the pressure to the temperature far below the glass transition temperature (217°C) using cooling water. The 22 layers of the prepreg gave a subsequent panel thickness
of approximately 7.2 mm after consolidation. The double cantilever beam (DCB) specimens were cut from the laminates, shown in Figure 1c.

The mode-I interlaminar fracture toughness test was carried out in a temperature range from 25°C to 130°C at a crosshead speed of 1 mm/min. To introduce loading to the DCB specimens, aluminum T-tabs were bonded to both sides of the pre-delaminated end using a high-strength commercial epoxy-based adhesive, and the bonding was more secured with aid of screws to prevent bonding failure at elevated temperatures. The critical strain energy release rates, $G_{IC}$, were defined using the modified beam theory (10).

3 Numerical Approach

2-D large deformation finite element analyses (FEA) were carried out using a commercial finite element code ABAQUS (Version 5.8). The plane strain condition was assumed for both CT and DCB specimens. Only one half of the specimen was modeled because of symmetry. The mesh configuration around the crack tips is shown in Figure 2. Rate independent plasticity and associated flow rule (11) were used for the material constitutive data. The values of true stress and true strain of bulk PEI polymers at different temperatures were calculated from the following (12):

\[
\text{True stress } \sigma = S(1 + e) \quad (1)
\]

\[
\text{True strain } \varepsilon = \ln(1 + e) \quad (2)
\]

where $S$ and $e$ are the engineering stress and strain, respectively. The true stress and strain curves are shown in Figure 3. The tensile properties were used in the FEA modeling for the CT and DCB specimens. The elastic properties of the laminates used in the FEA are listed in Table 2, which were estimated using the unit cell model (13). The finite element model for
the DCB specimen consists of a resin rich area of 25 \( \mu \text{m} \) in thickness around the middle plane. The resin rich layer was treated as isotropic while the laminate orthotropic.

4 Results and Discussion

4.1 Fracture Toughness

The fracture toughness (strain energy release rate, \( G_{IC} \)) for the CT and DCB specimens is summarised in Table 3. To represent delamination resistance of the composite laminate, two different toughness values were obtained from the R-curves of delamination growth in Figure 4. The mode-I interlaminar fracture toughness at delamination initiation, \( G_{IC,\text{ini}} \), was determined using the load-displacement curves at the deviation from linearity (NL) and/or the points when delamination visually initiated on the edge (VIS) (5). The toughness, \( G_{IC,\text{prop}} \), for delamination propagation was defined as the mean value of \( G_{IC} \) beyond the delamination increment length of 10 mm in the R-curves. For the PEI CT specimen, \( G_{IC} \) slightly increases with increasing testing temperatures whereas \( G_{IC,\text{ini}} \) for the composite displays a reverse trend although \( G_{IC,\text{prop}} \) increases.

In order to numerically evaluate fracture toughness, the \( J \)-integral concept was adopted. The \( J \)-integral can be a more appropriate parameter than critical strain energy release rate, \( G_{IC} \), if extensive plastic deformation of matrix around the crack tip is expected at elevated temperatures (4). The critical \( J \)-integral (\( J_{IC} \)) is equal to the critical strain energy release rate (\( G_{IC} \)) for linear elastic materials, and \( J_{IC} \) is an equivalent value of fracture toughness for non-linear materials (3). The experimentally measured critical fracture loads were employed to calculate the critical \( J \)-integral (\( J_{IC} \)). The fracture loads are the maximum load for the CT specimen, and the NL (or VIS) load point at delamination initiation with the initial delamination length (\( a_0 \)) of 50 mm, and the load at the delamination length of 62 mm in the
load-displacement curves from the DCB tests. The far field \( J \)-integral was evaluated using the domain integral method, which is directly implemented into ABAQUS (11). The method is an energy approach based on a contour integral along any arbitrary path enclosing the crack tip, where the \( J \)-integral is defined by:

\[
J = \int_{\Gamma} \left( W dy - T_i \frac{\partial u_i}{\partial x} ds \right)
\]  

(3)

where \( \Gamma \) is an arbitrary contour path around the crack, \( W \) is the total strain energy density, \( T_i \) are the components of the traction vector, \( u_i \) are the components of the displacement vector, and \( ds \) is the length of a line element along \( \Gamma \). The strain energy density is divided into elastic \( (W^e) \) and plastic \( (W^p) \) components as (3):

\[
W = \int_0^\varepsilon \sigma_{ij} \varepsilon_{ij} = W^e + W^p = \int_0^\varepsilon \sigma_{ij} \varepsilon^e_{ij} + \int_0^\varepsilon \sigma_{ij} \varepsilon^p_{ij}
\]  

(4)

where \( \sigma_{ij} \) the stress tensor, \( \varepsilon_{ij} \) is the total strain increment, and \( \varepsilon^e_{ij} \) and \( \varepsilon^p_{ij} \) are the elastic and plastic strain increments, respectively. The values of critical \( J \)-integral obtained from the 2-D large deformation FEA for the CT and DCB specimens are summarised in Table 4.

Tables 3 and 4 show there is a reasonable agreement between \( G_{IC} \) and \( J_{IC} \) values of the fracture toughness for the bulk PEI matrix and its composite over the entire temperature range. The \( G_{IC} \) and \( J_{IC} \) values for the bulk PEI polymer were increased with increasing temperature from room temperature (RT) to 130\(^\circ\)C. The initiation toughness, \( J_{IC,ini} \), for the CF/PEI composite decreases with increasing temperature while the propagation value, \( J_{IC,prop} \), increases, similar to those of \( G_{IC,ini} \) and \( G_{IC,prop} \), respectively. It should be noted that \( J_{IC,ini} \) is much smaller than \( J_{IC,prop} \). That correlates well with the rapid increase in \( G_{IC} \) of the initial R-curves shown in Figure 4. The cause of the R-curve behaviour was directly related to the evolution of process zone around the crack tip, which includes a number of different failure
processes such as the fibre bridging and breakage, multiple side matrix cracking, fibre/matrix debonding and plastic deformation of matrix contributing to delamination growth resistance (4, 14). Although the FEA used for the composite laminate cannot completely simulate the actual delamination behavior because of the assumption of perfect adhesion between fibre and matrix, the model represents sufficiently well for the purpose of this study to investigate the influence of matrix toughness on composites. Since fibre bridging effect is not considered, the increase in $J_{IC,prop}$ can be associated with matrix deformation and fracture, that will be discussed later along with fracture mechanisms via fractographic observations.

4.2 Fracture Mechanisms in Bulk PEI Polymers

Three distinct regions were microscopically identified on brittle fracture surfaces of the PEI polymer, i.e. crack initiation, mist and mirror/end band regions (15). These regions were characterized by different fracture features such as river patterns, patch patterns and end-banded structures, respectively. Especially, the crack initiation region was usually associated with a slow stable crack growth because the formation of the river pattern requires relatively higher energy absorption for developing crazing when the crack propagates (16). For glassy thermoplastic polymers, crazing is one of primary localized plastic deformation mechanisms. In general, crazing initiates with the nucleation of microvoids in areas of stress concentration, but prevents the coalescence of these voids and fracture by stretching fibrils from the bulk material and creating a crack like void-fibril structure. The fibril breakdown leads to unstable brittle fracture and hence crazing can be considered as a precursor to unstable brittle fracture (17).

Figure 5 shows the comparison in the crack initiation region between the specimens tested at RT and 130°C. The initiation region at 130°C is characterised by more distinctive and thicker
river patterns with the apparent whitened front zone. Compared to the specimen tested at RT, the initiation region includes the additional shear lips near the edges of the surface area. The microscopic examination of the fracture region shows a very rough surface with extensive tearing-type shear deformation (15). Elevated temperature promotes yielding and crazing, because yield and crazing stresses decrease with increasing temperatures, and hence the PEI polymer would be expected to undergo more significant deformation, depending on the stress state (18). Thus, extensive plastic deformation of combined yielding and crazing in the crack initiation region would consume more energy, and can be responsible for greater $G_{IC}$ and $J_{IC}$ values at elevated temperatures. However, to determine the influence of yielding and crazing independently or synthetically on fracture toughness, three-dimensional stress states at the crack tip through the thickness should be examined, and a more adequate craze-yield initiation criterion is necessary. Unfortunately, it is difficult to experimentally establish a craze-yield initiation criterion and the mechanical response of the crazed material in a three-dimensional stress state (17). Due to such difficulties, this study only accounts for the influence of bulk plasticity (yielding) on fracture toughness, neglecting the crazing micromechanism.

Figure 6a shows the triaxiality ($\sigma_m/\sigma_e$) in the vicinity of the crack tip corresponding to the fracture loads for the bulk matrices, where $\sigma_m$ and $\sigma_e$ are the mean stress and equivalent stress, respectively. The triaxiality (constraint) is released at the crack-tip to different extent with increasing temperature and the reduction is shown largest at 130°C. This is due to the low yield strength of the PEI polymer at the elevated temperature. Figure 6b shows the distributions of crack-tip plastic strain, and a high strain is observed in the specimen tested at 130°C. Figure 6c gives the distribution of crack-tip equivalent stress ($\sigma_e$), which increases with increasing the testing temperatures. As the equivalent stress is the driving force for
plastic yielding, the specimen tested at RT has the smallest plastic zone but the largest at 130°C. Therefore, the high fracture toughness at 130°C is due to a significant amount of constraint release and plastic deformation near the crack-tip, leading to the large plastic zone.

### 4.3 Fracture Mechanisms in Composite Laminates

Table 4 shows the interlaminar fracture toughness values ($J_{IC}$) of the DCB specimens, determined at the fracture loads, are lower than the corresponding toughness of the bulk matrix in the entire temperature range. Especially, the $J_{IC}$ values at delamination initiation are significantly reduced in comparison with the matrix toughness. Also, $J_{IC,ini}$ decreases as temperature increases, being irrespective of increasing matrix toughness. The result is consistent with the variation of $G_{IC,ini}$ with temperature. The fracture surfaces of the laminates near the delamination initiation area are shown in Figure 7. It can be seen that the dominating fracture mechanisms are fiber/matrix interfacial debonding along with matrix failure between fibres at ambient and elevated temperatures. The fracture surfaces with stepwise topography at the end of the insert film indicates that the delamination initiated and propagated through the fibre/matrix interface rather than from the resin rich area induced by the insert film, and hence the delamination initiation was dominated by the interfacial failure (19). It was found that the degree of adhesion between the PEI polymer and carbon fibers measured from the single fibre fragmentation method was strongly influenced by temperatures, and showed a remarkable decrease with increasing temperatures (20). As the interface adhesion strength decreases and the matrix toughness increases with increasing temperatures, the delamination is more likely to initiate at the interface rather than from the matrix-rich area ahead of the film insert. Therefore, the weakened fibre/matrix interface plays an overriding role in the delamination initiation and is unable to transfer the enhanced matrix toughness to composites. Such a phenomenon is reflected on the reduced $G_{IC,ini}$ or $J_{IC,ini}$ at elevated temperatures.
Figure 8 illustrates the distributions of stress triaxiality and equivalent stress ahead of the crack tip corresponding to the fracture loads at delamination initiation for the DCB laminates. No apparent release of triaxiality can be seen for the DCB specimens, compared with the bulk specimen (Figure 6). Although the highest $\sigma_e$ and the largest plastic zone are obtained in the laminate tested at 130°C due to the lowest yield strength, it has the lowest $J_{IC}$. The unexpected result is attributed to the decreased elastic strain energy dissipated at delamination initiation with increasing temperature because of premature failure of fibre/matrix interface, and the relatively small amount of plastic energy in the plastic zone is unlikely to compensate for the decrease in the elastic energy, as shown in Figure 9. It can be concluded that the weakened fibre/matrix interface adhesion at elevated temperatures leads to the delamination initiation at the fibre/matrix interface, which impedes the transfer efficiency of matrix toughness by preventing extensive deformation of the matrix.

Figure 10 illustrates SEM micrographs of mode-I delamination propagation areas. The fracture surface of CF/PEI composites at room temperature reveals fibre/matrix interface debonding, leaving flakes of matrix without any significant matrix deformation. On the other hand, at 130°C, more extensive matrix plastic deformation (or drawing) occurred in matrix fracture regions between fibres along with fibre/matrix interface debonding. The significant matrix deformation is one of major energy contributions to enhance the interlaminar fracture toughness of thermoplastic composites at elevated temperatures (21). Although no apparent release of triaxiality ahead of the crack tip corresponding to the fracture loads during delamination propagation can be seen for the DCB specimens tested at all temperatures in Figure 11, the highest $\sigma_e$ for the specimen at 130°C is indicative of the large plastic zone, where the elastic energy density is similar to that around the crack tip at delamination initiation (Figures 9 and 12) but the plastic energy density is an order of magnitude greater
than that at delamination initiation (Figures 9 and 13). Hence the greater $J_{IC,prop}$ compared to $J_{IC,ini}$ can be attributed to the increased plastic energy dissipated in the plastic zone. As a result of the increased plastic energy with increasing temperature, $J_{IC,prop}$ has the highest value at 130°C. This result supports the conclusion drawn from the experimental observation that the enhanced matrix toughness at elevated temperatures can lead to the greater interlaminar fracture toughness during delamination propagation through extensive matrix deformation.

The reduction of $J_{IC}$ values for both delamination initiation and propagation in the laminate in comparison to those for the bulk matrix is also attributed to the high triaxiality developed in the laminates; $\sigma_m/\sigma_e$ is about 3 and 1.2 for the laminate and the bulk matrix, respectively, ahead of crack tip at the distance of 0.011 mm (Figures 6, 8 and 11). The high triaxiality for the laminate indicates the constrain effects of rigid fibres, and suppresses the plastic deformation and crack-tip blunting. Consequently, the plastic energy density around the crack tip considerably decreases in comparison to that of the bulk PEI matrix as shown in Figure 13. The reduced plastic energy density for the laminate was suggested to account for poor translation of matrix toughness to delamination toughness (3). On the other hand, the $J_{IC,prop}$ value at 130°C is lower than the matrix fracture toughness ($J_{IC}$) in Table 4, and this is not in agreement with the fact that $G_{IC,prop}$ is slightly higher than the matrix toughness ($G_{IC}$) at 130°C (Table 3). The weakened fibre/matrix interface at elevated temperatures promoted the formation of fibre/yarn bridging and the crack branching around the debonded weft yarn, compared to the specimen tested at room temperature [5]. These additional toughening fracture mechanisms apparently increased the fracture toughness by increasing fracture surface area and damage zone, lowering fibre constraint in matrix and relieving triaxiality of stress, thereby facilitating more matrix deformation. However, these fracture mechanisms are
not taken into account in the current model, that may be attributed to the low translation of matrix toughness to $J_{IC,prop}$ at elevated temperatures.

5 Conclusions

Effects of testing temperatures on fracture toughness of a PEI thermoplastic polymer and its relationship with interlaminar fracture toughness of its carbon fibre composites were investigated. It has been shown that the fracture toughness for the PEI matrix and the laminates based on the $J$-integral approach is in good agreement with that based on the conventional approach of strain energy release rate.

For the bulk matrix, both $G_{IC}$ and $J_{IC}$ increased with increasing temperatures. The high fracture toughness for the bulk PEI matrix at elevated temperatures was due to the large plastic zone near the crack-tip with a significant amount of the plastic deformation and release in stress triaxiality. Although no apparent release of the triaxiality was found in the DCB specimen, the plastic zone increased with increasing temperatures. However, the plastic zone size did not show the correlation with the fracture toughness at delamination initiation, and the plastic energy density in the plastic zone was very small and the elastic energy density is still the dominant factor. In this case, the overriding role of the weakened fibre/matrix adhesion at elevated temperature impeded the transfer efficiency of matrix toughness without exercising excessive plastic deformation at delamination initiation, leading to the low interlaminar fracture toughness. However, the large plastic zone with the greater magnitude of the plastic energy at elevated temperature accounted for the increased composite toughness during delamination propagation, combined with other fracture mechanisms.
In comparison of the computed stress-strain and plastic energy density distributions around the crack tips between the CT and DCB specimens from 2-D large deformation finite element analyses (FEA), the matrix toughness was not fully transferred to the interlaminar fracture toughness due to the high triaxiality and reduced plastic energy density of the DCB specimen relative to those of the bulk matrix. The high triaxiality indicated the constraint effects of rigid fibres and suppression of the plastic deformation and crack-tip blunting.
References


Caption of Figures

Figure 1 Specimen geometry for (a) tensile specimen, (b) compact tension (CT) specimen and (c) double cantilever beam (DCB) specimen (all dimensions in mm), where a is the crack length.

Figure 2 Finite element mesh around the crack-tip in 2-D large deformation FEA analyses.

Figure 3 True stress-strain relationships for PEI polymer at different temperatures (solid circles in dotted lines indicate stress and strain from image analyses).

Figure 4 Typical R-curves depicting the variation of mode-I interlaminar fracture toughness, $G_{IC}$, with delamination crack growth at different testing temperatures.

Figure 5 Microflow lines and river patterns in rough whitened front zone (initiation region) of fracture surfaces of PEI polymer at different testing temperatures, (a) RT and (b) 130°C. Crack propagation direction is from left to right.

Figure 6 Stress distributions at fracture in CT specimen, (a) triaxiality, (b) plastic strain and (c) equivalent stress.

Figure 7 Mode-I interlaminar fracture surfaces in delamination initiation areas of CF/PEI composite laminate at different testing temperatures.

Figure 8 Stress distributions ahead of crack-tip at delamination initiation in DCB specimen, (a) triaxiality and (b) equivalent stress.

Figure 9 Elastic and plastic energy density ahead of crack-tip at delamination initiation in DCB specimen.

Figure 10 Mode-I interlaminar fracture surfaces in delamination propagation areas of CF/PEI composite laminates at different testing temperatures.

Figure 11 Stress distributions ahead of crack-tip for delamination propagation in DCB specimen, (a) triaxiality and (b) equivalent stress.

Figure 12 Elastic energy density ahead of crack-tip for DCB specimen at delamination propagation.

Figure 13 Plastic energy density ahead of crack-tip for CT specimen and DCB specimen at delamination propagation.
**Table 1** Tensile properties of PEI polymer at different temperatures

<table>
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<tr>
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<th>Young’s modulus [GPa]</th>
<th>Yield strength [MPa]</th>
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<tr>
<td>25°C</td>
<td>3.3</td>
<td>111</td>
</tr>
<tr>
<td>80°C</td>
<td>2.7</td>
<td>64</td>
</tr>
<tr>
<td>130°C</td>
<td>2.5</td>
<td>64</td>
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**Table 2** Effective elastic properties of CF/PEI laminates at different temperatures

<table>
<thead>
<tr>
<th></th>
<th>E₁</th>
<th>E₂</th>
<th>E₃</th>
<th>G₁₂ [GPa]</th>
<th>G₁₃</th>
<th>G₂₃</th>
<th>ν₁₂</th>
<th>ν₁₃</th>
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<tr>
<td>25°C</td>
<td>57.6</td>
<td>57.6</td>
<td>8.7</td>
<td>3.1</td>
<td>2.8</td>
<td>2.8</td>
<td>0.03</td>
<td>0.4</td>
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<td>57.0</td>
<td>7.7</td>
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<td>2.5</td>
<td>2.5</td>
<td>0.03</td>
<td>0.4</td>
<td>0.4</td>
</tr>
<tr>
<td>130°C</td>
<td>56.8</td>
<td>56.8</td>
<td>7.3</td>
<td>2.5</td>
<td>2.3</td>
<td>2.2</td>
<td>0.03</td>
<td>0.4</td>
<td>0.4</td>
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**Table 3** Fracture toughness, $G_{IC}$, for CT and DCB specimens

<table>
<thead>
<tr>
<th></th>
<th>CT specimen $G_{IC}$ [kJ/m²]</th>
<th>DCB specimen $G_{IC,ini}$ [kJ/m²]</th>
<th>$G_{IC,prop}$ [kJ/m²]</th>
</tr>
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<tbody>
<tr>
<td>25°C</td>
<td>3.17</td>
<td>1.55</td>
<td>2.69</td>
</tr>
<tr>
<td>80°C</td>
<td>3.75</td>
<td>1.26</td>
<td>3.35</td>
</tr>
<tr>
<td>130°C</td>
<td>4.07</td>
<td>1.06</td>
<td>4.23</td>
</tr>
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**Table 4** Fracture toughness, $J_{IC}$, for CT and DCB specimens

<table>
<thead>
<tr>
<th></th>
<th>CT specimen $J_{IC}$ [kJ/m²]</th>
<th>DCB specimen $J_{IC,ini}$ [kJ/m²]</th>
<th>$J_{IC,prop}$ [kJ/m²]</th>
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<tr>
<td>25°C</td>
<td>3.08 (1800)*</td>
<td>1.17 (190)</td>
<td>2.41 (240)</td>
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<tr>
<td>80°C</td>
<td>3.36 (1700)</td>
<td>0.85 (158)</td>
<td>2.94 (255)</td>
</tr>
<tr>
<td>130°C</td>
<td>3.65 (1650)</td>
<td>0.75 (149)</td>
<td>3.41 (265)</td>
</tr>
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* The values in parentheses are fracture loads (N) used in FEA
Figure 1 Ki-Young Kim, Lin Ye and Cheng Yan
Figure 1 Cont
Figure 2 Ki-Young Kim, Lin Ye and Cheng Yan
Figure 3 Ki-Young Kim, Lin Ye and Cheng Yan
Figure 4 Ki-Young Kim, Lin Ye and Cheng Yan
Figure 5 Ki-Young Kim, Lin Ye and Cheng Yan.
Figure 6 Ki-Young Kim, Lin Ye and Cheng Yan

(a) 

(b)
Figure 6 Cont
Figure 7 Ki-Young Kim, Lin Ye and Cheng Yan
Figure 8 Ki-Young Kim, Lin Ye and Cheng Yan
Figure 9 Ki-Young Kim, Lin Ye and Cheng Yan
Figure 10 Ki-Young Kim, Lin Ye and Cheng Yan
Figure 11 Ki-Young Kim, Lin Ye and Cheng Yan
Figure 12 Ki-Young Kim, Lin Ye and Cheng Yan
Figure 13 Ki-Young Kim, Lin Ye and Cheng Yan