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The effect of temperature on stress transfer between a broken fibre and the adjacent fibres in unidirectional fibre composites

S. Behzadi, F. R. Jones¹

Ceramics and Composite Laboratory, Department of Engineering Materials, Sir Robert Hadfield building, University of Sheffield, S1 3JD, UK

Abstract

The effect of yielding of a polymer matrix on stress transfer between a broken fibre and an adjacent fibre has been studied using a three-dimensional (3D) finite element model. The full stress-strain curves of Araldite MY0510 epoxy resin have been measured as a function of temperature. The elasto-plastic behaviour of the epoxy resin has been included in a 3D multi-fibre FE model. The yielding of the matrix affects the axial strain development along both the fractured and the neighbouring fibre. The intensified axial strain at the intact neighbouring fibre has been found at the immediate radial distance from the broken fibre. Strain concentration Factor (SCF) of nearest neighbouring fibre decreased as a result of an increase in temperature.

Keywords: A. Polymer-matrix composites (PMCs), Matrix Yielding, B. Stress/strain curves, C. Finite element analysis (FEA), D. Stress concentrations.

1. Introduction

It is known [1,2] that the “Rule Of Mixture” (ROM) overestimates the strength of composites containing high performance brittle fibres such as carbon. This method assumes a uniform distribution of strength along the fibres and simultaneous breakage of all fibres as well as a linear elastic matrix. The scatter in fibre strength along the fibres results in a sequential fibre fractures. This in turn induces an intensified stress on the adjacent intact fibres to a fibre-break. Therefore, the rule of mixtures is not reliable

¹ Corresponding author. Tel:+44-114-2225477; fax: +44-114-2225943.
E-mail address: f.r.jones@shef.ac.uk (f.r.jones)

because a precise failure strength cannot be predicted [3,4]. Several studies [5-7] have demonstrated that the durability and performance of fibre composites is determined by micromechanisms occurring at the fibre/matrix interface level. As the stored energy in the region around a fractured fibre-end is released, it is required to be redistributed among its intact neighbouring fibres to re-establish an equilibrium of load [8,9]. Depending on their distance from the fibre-break, the local intact fibres are introduced to a stress concentration factor. Curtis [2] demonstrated that a fibre-break in UD composite is most likely to be followed by propagating fibre fractures as a result of the local intensified stress. The intensified stress increases with the number of fractures and hence increases the probability of the failure of the intact fibres and modifies the mode of failure of composites [10,11]. The term of the stress concentration factor (SCF) is the ratio of the maximum (or local) to the applied (or undisturbed) stress of the intact fibre far away from the break. Hedgepeth and van Dyke [8] believed that the role of redistributed overload from a fibre-break is negligible compared to the axial load on the enormous number of fibres in a real composite. At some distant away from the broken fibre-end, the axial tensile stress recovers in the broken fibre and develops a length over which the stress rises to reach the applied stress. The distance over which the broken fibre is ineffective to support the applied load termed as “ineffective length” (IL) by Rosen [12] and shown schematically in Fig. 1. The transfer length or ineffective length determines the redistribution of the overload between neighbouring fibres [7,13,14]. It is believed [8,15-19] that the SCF of the neighbouring fibres together with ineffective length of the broken fibre are the main parameters to govern the ultimate failure mode of composites. In a UD composite, the load carried by a fibre is thrown onto the matrix when the fibre breaks. The strength of the interfacial bond also influences the stress

transfer length because depending on shear stresses generated, either debonding or interphasal shear can occur. Therefore, when the interface remains intact, the matrix plays a significant role in determining the magnitude of the stress concentration in adjacent fibres.

However, the estimation of the stress concentration factor does not lend itself to experimental determination. In order to deal with the complicated 3D nature of real fibre composites, Finite Element Methods (FEM) were developed in the literature as the best alternative for obtaining stress distributions in the fibres. Nedele *et al* [18,20] introduced a 3D FE model containing elastic fibres and matrix and reported that the stress concentration factor was not significant in the ultimate failure of a UD composite. In another approaches, Fiedler *et al* [21] and Lane *et al* [22] studied the load transfer between a fractured fibre and its surrounding intact fibres through linear elastic and linear elastic-ideal plastic matrices. They [21,22] strongly demonstrated the importance of matrix plasticity on the reinforcing efficiency of a broken fibre and on the SCF values. Their values are in a good agreement with the analytical values reported by Wagner *et al* [17]. Van den Heuvel *et al* [23] introduced a 3D finite element model which included an element of full UD composite around a fibre-break similar to the model of Nedele *et al* [20] and obtained considerably lower stress concentration factor compared to their previous 2D planar model [24].

Epoxy resins are one of the most important thermosetting polymers which have been extensively used in aerospace composites materials over last few decades because of their high yield strengths, Young's moduli and glass transition temperatures. These materials exhibit brittle tensile behaviour in the bulk. However, in a real composite the large difference between the elastic modulus of the fibre and the thin layer of matrix

results in interfacial shear yielding at a fibre-break [10,25-30]. This develops a plastic region as the interfacial shear stress in the matrix exceeds the yield strength of the bulk epoxy resin. The measurement of fibre stress profiles using Raman spectroscopy [31] have confirmed that an extensive localised yield region occurs in the matrix at the vicinity of a fibre-break. Other studies [27,32-34] have demonstrated the occurrence of transverse matrix cracking or interfacial debonding, or mixed mode where both of these are coincident. Zhao *et al* [35] have demonstrated using photoelasticity that the sequence of events involves fibre fracture, which propagates into the matrix and delays the stress transfer back to the fibre through shear. The shear stress can then initiate debonding which can propagate back towards the fibre-break and away from it. Kettle *et al* [36] have demonstrated that all of these micromechanisms from debonding to transverse matrix cracking are a function of the strength of the interfacial bond.

Therefore, linking the experimental or modelled micromechanical results to the macroscopic mechanical properties of UD fibre reinforced composites is the main goal in characterisation of the composite materials. In this study, the elasto-plastic behaviour of a high performance resin matrix was identified as the principal matrix property which is a key mechanism for controlling the matrix-to-fibre stress transfer. In the literature, no attempt has been made to characterize the stress-strain behaviour of epoxy resins used in high performance fibre composites while most work is focused on crosslinked difunctional (DGEBA) epoxy resins [37-40]. In general, the yield behaviour of the viscoelastic materials is highly dependent on the temperature. Therefore, the stress-strain behaviour of the epoxy resins was studied in a wide range of temperatures.

The elasto-plasticity behaviour of the epoxy resin is then used to predict the stress transfer from a broken fibre to the adjacent fibres and in particular stress concentration

factor in the neighbouring fibres. Using statistical methods, the finding of this study has provided a predictive tool to evaluate the failure strain of an UD fibre composite [41] and to optimize the composite performance through matrix selection.

2. Experimental

The epoxy resin used for this study is triglycidyl *p*-aminophenol (TGAP), Araldite MY0510[®] cured with 4-4'diaminodiphenyl sulphone, DDS hardener. While thermosetting resins fracture at very low tensile strains, they undergo considerable plastic deformation under uniaxial compressive load. The cured cylindrical test specimens were machined to achieve an accurate *Length/Diameter* ratio of 1:1 (~ 10 mm/10 mm). Uniaxial compression testing of the specimens was performed with a *Hounsfield* universal testing instrument. The crossheads and the specimen ends were smeared with a touch of petroleum jelly before alignment to minimise the frictional forces at the junctions. Fig. 2 monitors the deformation of a typical epoxy resin specimen at different compressive strains over its corresponding true stress-strain curve. It can be seen that there is no visible barrelling in the specimen after the yield strain or even at higher strains before the fracture. Extensometers or strain gauges could not be mounted on the test specimens because of their small dimensions. Therefore, for obtaining the actual deformation of the specimen, the compliance of the testing machine had to be measured precisely and deducted from the total deformation monitored by the actuator. In the *non-sample* technique demonstrated by Kalidindi *et al* [42], the load-displacement relationship for the machine was directly measured with and without the specimen. This compliance correction [42] assumes that the specimen and loading parts are two springs in series. In this way, the functionality of the compliance displacement with applied load at a given crosshead speed can be obtained for the testing instrument. This data can be

used to correct the recorded load-displacement data for any specimen tested under uniaxial compression by the testing machine at the same crosshead speed.

A fan-assisted oven mounted on the testing machine, delivered a range of temperatures from 22 °C to 180 °C during the uniaxial compression testing. The test was performed at a crosshead speeds of 0.1 mm/min which corresponds to initial strain rate of $1.67 \times 10^{-4} \text{ s}^{-1}$, considering the initial lengths of the test specimens. However, the true strain rate increases as the uniaxial compressive deformation of the specimen proceeds. The true longitudinal strain, ϵ_T was obtained by:

$$\epsilon_T = \ln\left(\frac{L_i}{L_t}\right) \quad (\text{eq. 1})$$

where L_i and L_t are the initial and instantaneous (corrected for the machine compliance) lengths. Because the continuous monitoring of the lateral expansion of the small test specimens was not practical, the instantaneous diameter of the specimens was obtained using the longitudinal true strain ($\epsilon_T = \epsilon_{yy}$) and Poisson's ratio. The schematic illustration of three-dimensional deformation of a cylinder under uniaxial compression is given in Fig. 3. The Poisson's ratio in uniaxial compression can be defined as:

$$\nu = -\frac{\epsilon_{xx}}{\epsilon_{yy}} = -\frac{\epsilon_{zz}}{\epsilon_{yy}} \quad (\text{eq. 2})$$

where ν is the Poisson's ratio and ϵ_{xx} and ϵ_{zz} are the transverse strains which may be considered as (using eq. 1):

$$\epsilon_{xx} = \epsilon_{zz} = \ln\left(\frac{D_i}{D_t}\right) \quad (\text{eq. 3})$$

where D_i and D_t are initial and instant diameter of the test specimen. Eq. 3 can be rearranged, using eq. 2, to:

$$D_t = \frac{D_i}{\exp(-\nu\epsilon_{yy})} \quad (\text{eq. 4})$$

In this way, the instant cross-sectional area of the specimen and hence true compressive stress can be obtained using the applied load. The yield was defined as the point where the load passed through a maximum or an apparent ‘knee’ in the stress-strain curve. The Poisson’s ratio of thermoset polymers is a viscoelastic aspect of the material which is increased by temperature and strain rate [43,44]. A recent theoretical study by Foreman *et al* [45] using Group Interaction Modelling (GIM), reported the variation in Poisson’s ratio of the epoxy resin of interest over a range of temperatures. However, in this study, no significant change was observed in the true compressive stress-stress curves of the epoxy resin using the reported ranges of Poisson’s ratio [43-45] for the change in temperature and strain rate. Therefore, a constant Poisson’s ratio of 0.35 is considered for the epoxy resin at all temperatures and strain rates.

3. Multi fibre finite element model

A 3D model of a square array of fibres was generated using ANSYS 8.1 and is shown in Fig. 4. The model is consisted of four quarter fibres of 100 μm long placed at the corners of a rectangular prism similar to the models of Fiedler *et al* [21] and Lane *et al* [22]. The properties of the transversely orthotropic carbon fibres (Table 1) used in this study were taken from elsewhere [46] and listed in Table 1. The properties of the carbon fibre were assumed to be temperature-independent. The fibres are enclosed in a volume of the resin matrix. The fibre volume fraction was chosen to be 50 %, which is realistic of actual material. The coinciding nodes all along the interfaces in the FE model ensured perfect interfacial bonding between the fibres and matrix. The 8-noded solid brick structural SOLID45 element was employed to mesh both the fibres and the matrix. In order to introduce a fibre-break the top layer of the elements can be removed from the fibre-end as shown in Fig. 4. It is also shown that the concentration of the elements was increased towards the front face of the model where the fibre-break occurred. An axial

displacement of 1 % was applied on the front face of the model in z-direction while the back face was constrained with $z = 0$. The applied strain of 1 % is practically around a strain level, at which a fibre fractures. The model is constrained from motion on its other four faces regarding the symmetry boundary conditions to satisfy the model's stability.

The tensile failure of a bulk cured epoxy resin is dominated by the concentration of flaws. This results in premature failure of the resin specimens at very low tensile strains. However, in a composite, the epoxy resin matrix is uniformly distributed between the fibres and its thickness is not higher than 1 μm . Thus, it can yield under a tensile load between the stiff fibres. This is consistent with the recent study of Fiedler *et al* [47] on small volumes of an epoxy resin. They reported a large plastic deformation and necking for the resin fibre of a diameter of 2 mm. In this study, we assumed that the tensile stress-strain behaviour of such a small volume of the resin matrix can be considered identical to its compressive one. The true experimental compressive stress-strain curve of the epoxy resin tested at different temperatures was digitized and input into the model. Using the von Mises yield criterion, a reliable calculation of the 3D matrix yielding properties can be obtained. Because of the large plastic strain at the fibre/matrix interface, post-yielding phenomena namely strain softening and strain hardening could not be ignored [24]. The resin matrix was considered as an isotropic material with Poisson's ratio of 0.35. It is emphasized in Section 2 that the changes in Poisson's ratio within the range of temperatures and strain rates have not altered the true stress-strain curves of the epoxy resin. In order to ensure the accuracy of the FEM results, the corresponding Poisson's ratio [45] was input into the model at each temperature. However, no considerable variations in the findings of the model was observed over the range of temperature studied.

4. Results

4.1 Uniaxial compression test

The full compressive stress-strain curves of the epoxy resin as a function of temperature are shown in Fig. 5. These glassy polymers proved to be extensively ductile up to a compressive strain of 40 %. It is obvious that increasing the test temperature led to a significant decrease in the compressive yield strength as well as the yield strain. Furthermore, although the range of test temperatures was selected so that the resin remained in a glassy state, the compressive modulus also showed a considerable reduction with temperature. The variations of the yield strength, yield strain and modulus of the epoxy resin with temperature are tabulated in Table 2. It has been reported elsewhere [48] that the yield properties decrease linearly with increase in test temperature.

4.2 Finite element modelling

Following the solution of the finite element model, the strain (or stress) in both the fibres and the matrix can be obtained. The effect of matrix yielding on the load transfer characteristics and particularly the SCF in the presence of a fibre-break has been studied as a function of the temperature. Fig. 6(a) shows the effect of ambient temperature on the development of the axial tensile strain along the broken fibre in the model composite containing the resin matrix MY0510/DDS. In this study, a strain analysis was carried out because of the known failure response of these composites. The distance from the fibre-break is normalized along the fibres with the length of the FE model. It can be observed that there is an exponential increase in axial tensile strain in the broken fibre after a fibre-break. However, at higher temperatures this exponential strain build-up becomes effectively linear. Fig. 6(a) also exhibits that because of the elasto-plastic behaviour of the matrix (Fig. 5), the applied strain cannot be fully recovered even at the far-field from

the broken fibre so it falls below the applied strain of 1 %. Moreover, as the degree of plasticity of the matrix increased with temperature, the far-field strain decreased. The effect of a fibre-break on the adjacent fibre is studied by monitoring the development of axial tensile strain along the centre of the neighbouring fibre and shown in Fig. 6(b). It can be seen that the introduction of a break results in a considerable increase in the axial strain at the vicinity of the break. After passing through a maximum in the immediate vicinity of the fibre-break, the axial strain reduced along the neighbouring fibre to below the applied strain at the far-field. It can also be noted that as the degree of plasticity of the matrix increases with temperature, the maximum and far-field axial tensile strains decreased. Fig. 7 shows the variation in interfacial shear stress in the interface of neighbouring fibre/matrix as a function of temperature in a model composite containing the resin matrix. It indicates that the strain transfer from the matrix to the neighbouring fibre behind a fibre-break is caused by a local matrix yielding. As the temperature increases, the matrix becomes more plastic in nature so that a higher volume is involved in the strain transfer mechanism. Table 3 lists the SCF values in the neighbouring fibre to a broken fibre for a composite containing the resin matrix MY0510/DDS over a range of temperatures. In this study, the values of SCF are representative of the maximum strain concentration factors in-plane to the fibre-break along the centre of the neighbouring fibre. Therefore, these values are not dependent on the length of the FE model.

5. Discussion

On the occurrence of a fibre-break in a UD fibre composite, a large fraction of the load which used to be carried by the fibre is transferred to the matrix in the plane of the crack. This results in a high shear stress at the interface between the fibre and the matrix. With a perfect interfacial bond, the overload is redistributed between the intact

neighbouring fibres through the matrix shear yielding. The maximum axial tensile strain in the neighbouring fibre, SCF, demonstrates the increased probability of the failure of the intact neighbouring fibre. Most of the models [18,20] for predicting the strain (or stress) concentration factor neglect the elasto-plasticity of the matrix. The assumption of a linear elastic material for the matrix limits the predictive accuracy of the models. Hedgepeth and van Dyke [8] included the matrix plasticity in their work which resulted in a decrease in the stress concentration factor with increase in load. Ochiai *et al* [49] also concluded that the inclusion of such local matrix yielding in the model decreases the SCF in the adjacent fibres and results in a better agreement between the experimental and analytical results. The elasto-plasticity of the matrix has been incorporated into finite element models by other authors [21,22] and led to a significantly lower strain concentration factor in comparison to the elastic case. In contrast, it has been shown elsewhere [23] that matrix plasticity resulted in a slight decrease in the SCF and it was concluded that the SCF was mainly affected by the number of surrounding fibres.

It was also shown that the increase in temperature led to a decrease in axial tensile strain along the broken fibre. As the matrix yield strength decreases with increase in temperature, the exponential development of the axial tensile strain along the centre of the broken fibre was replaced by a linear behaviour. This results in a less efficient strain transfer back into the broken fibre since more of the stored energy released on fibre fracture is already absorbed by the matrix, which undergoes a larger plastic deformation. Because the plasticity of the matrix increases with temperature, less strain is transferred back to the other end of the broken fibre. Thus, the broken fibre cannot recover the applied strain at its far-field. Table 4 shows the variation in far-field strain in the broken fibre as a function of temperature. It can be seen that the discrepancy between the far-

field and applied strains becomes more pronounced at higher temperatures. This is consistent with a longer ineffective length of the broken fibre at higher temperatures. Table 4 also lists the changes in the ineffective length in the broken fibre with temperature. In the literature, the ineffective length is known as the length of fibre over which 90% recovery in axial strain (or stress) occurs. However, in this study, the ineffective length is considered as the distance between the fibre-break and an axial strain plateau. As the modulus of the matrix decreases with an increase in temperature (Table 2), the mismatch between the moduli of the fibre and matrix increases and hence a larger volume of the matrix contributes to the shear yielding mechanism. Furthermore, it is shown in Fig. 7 that the interfacial shear yielded zone at the neighbouring fibre/matrix interface expands towards the middle of the fibre in the axial direction, with increase in the matrix plasticity. Thus, the stress-transfer capacity of the matrix is reduced more and more accordingly and as a result, a lower overload is redistributed between the neighbouring fibres. This is also associated with the decrease in maximum and far-field axial strains in the neighbouring fibre with increase in temperature as shown in Fig. 6(b). Hence, the fibre adjacent to a broken fibre experiences a reduction in SCF as a result of a higher temperature (Table 3). In this study, the magnitude of the calculated SCF value obtained at room temperature (Table 3) is higher than those reported in other studies [8,21-23] where matrix elasto-plasticity was included in the models. This can be explained by the redistribution of the overload from a broken fibre among a larger number of intact neighbouring fibres in those studies. On the other hand, the high modulus and yield strength of the high performance resin matrix used in the current FE model can be responsible for the increase in the SCF values in the neighbouring fibre.

6. Conclusion

The yield behaviour and properties of a cured epoxy resin which forms one of the main components of high performance commercial matrix systems have been investigated. The true compressive stress-strain curves of the resin have been measured as a function of test temperature. A 3D FE model of UD fibre composite containing a fibre-break was introduced. The properties of the matrix were input into the model in the form of true compressive stress-strain curves of the epoxy resin. The relationship between the matrix yielding and the redistribution of the stress overload from the fibre-break to the neighbouring fibres was studied. In high volume fraction composites, the yielding of the matrix governs the strain redistribution between the intact fibres at the vicinity of a broken fibre. Interfacial shear yielding of the matrix reduces the efficiency of load transfer and thus reduces the SCF in neighbouring fibres to the broken fibre. At a higher temperature, the reduced stress transfer efficiency results in a lower SCF in the neighbouring fibre. Thus, the probability of multiple in-plane fibre-breaks (and hence a flaw of critical dimensions) is reduced at higher temperatures. Therefore, the durability of an epoxy resin composite is strongly influenced by the elasto-plastic behaviour of the matrix.

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Figures

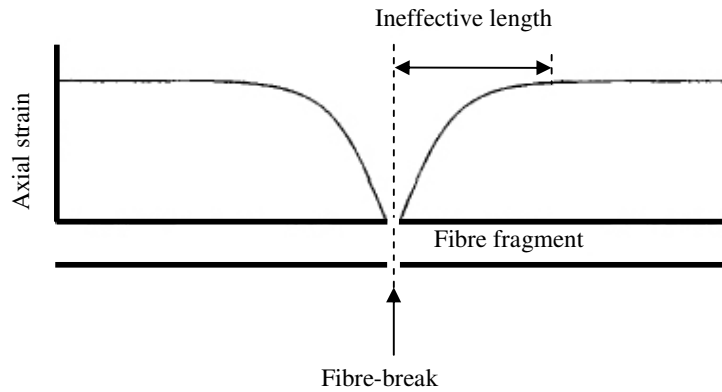


Fig. 1. Schematic illustration of the strain build-up in a broken fibre along the ineffective length.

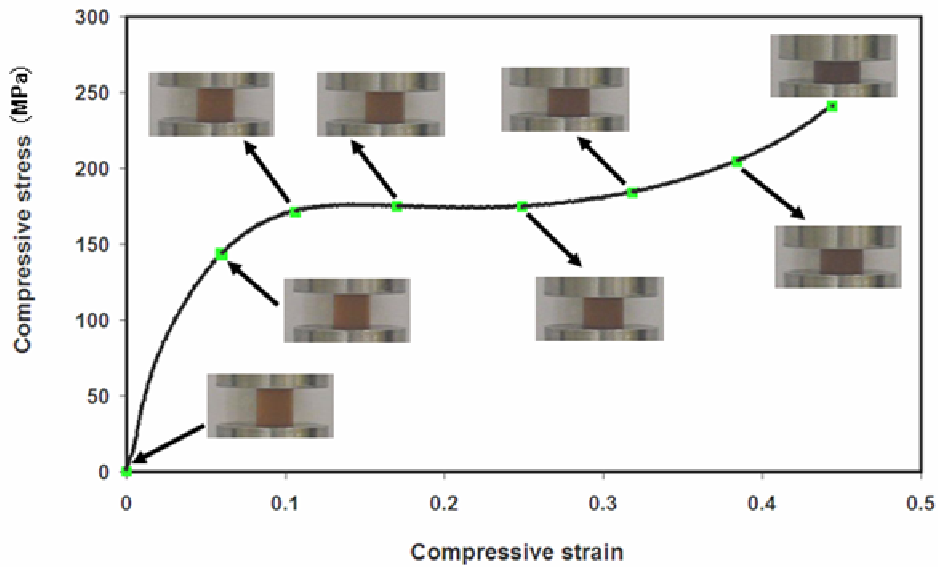


Fig. 2. Photographs of deformation of an epoxy resin MY0510/DDS specimen under uniaxial compression load at different strains shown on the corresponding true stress-strain curve at room temperature and a strain rate of $1.67 \times 10^{-3} \text{ s}^{-1}$.

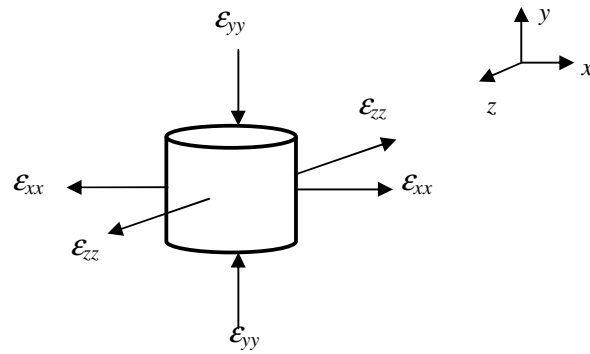


Fig. 3. Schematic illustration of longitudinal and transverse strains in the epoxy resin specimen under uniaxial compressive load.

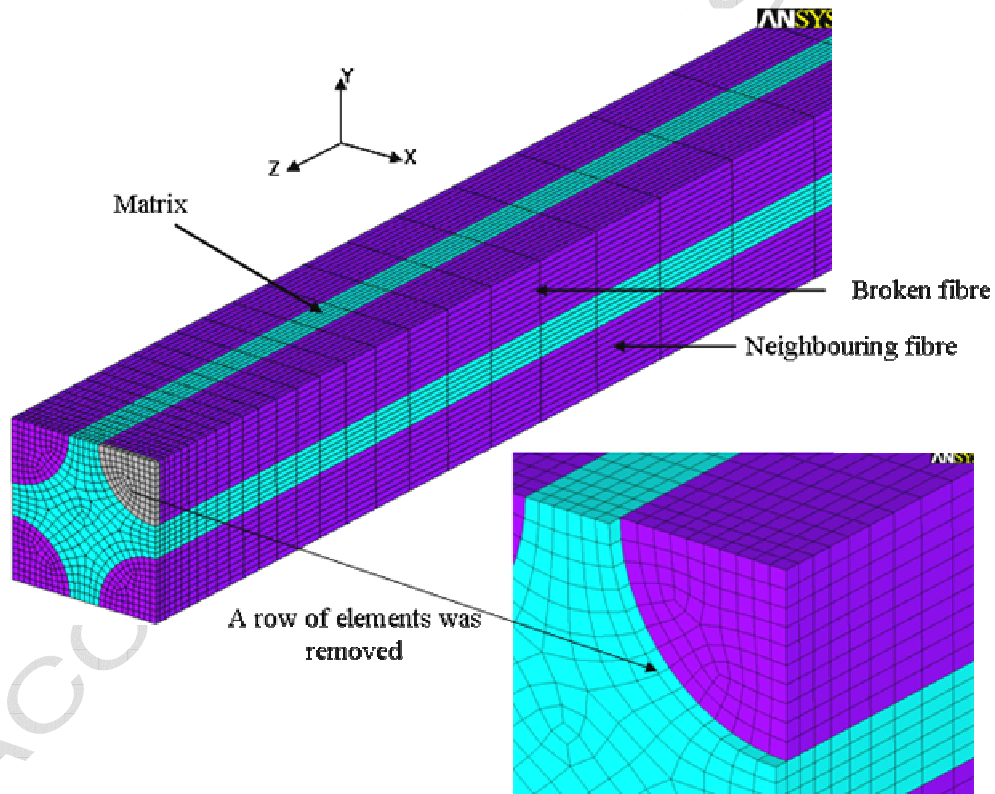


Fig. 4. Illustration of the geometry and meshing of the 3D FE model of a unidirectional high volume fraction composite used in this study. The first row of elements was removed to represent a fibre-break on the broken fibre. Neighbouring fibre is the immediate adjacent fibre to the broken fibre.

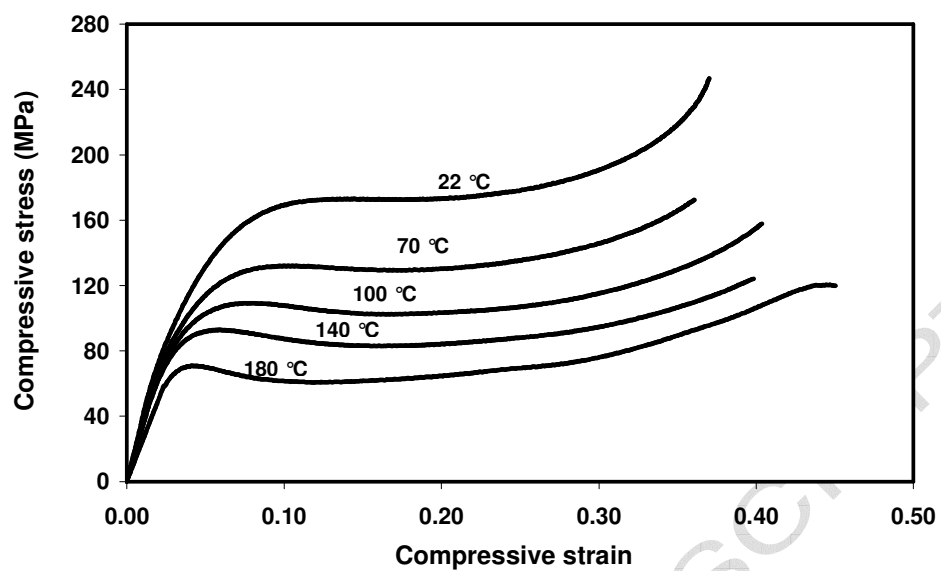


Fig. 5. True compressive strain-stress curves of epoxy resin MY0510/DDS at a range of temperatures and the initial strain rate of $1.67 \times 10^{-4} \text{ s}^{-1}$.

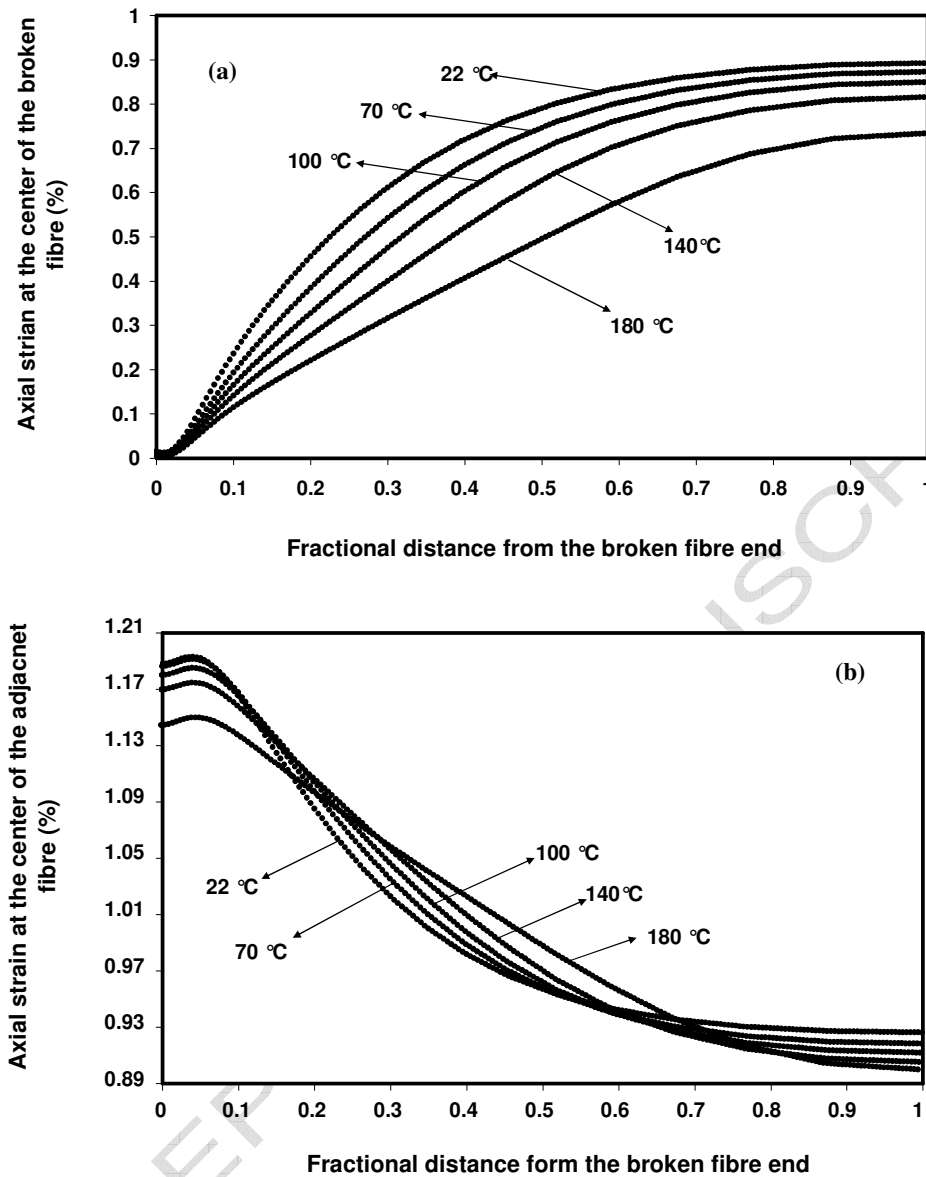


Fig. 6. Profiles of the axial tensile strains at the centre of (a) the broken fibre and (b) the neighbouring fibre to a fibre-break in a UD composite containing the resin matrix MY0510/DDS at an applied strain of 1% and over a range of temperatures.

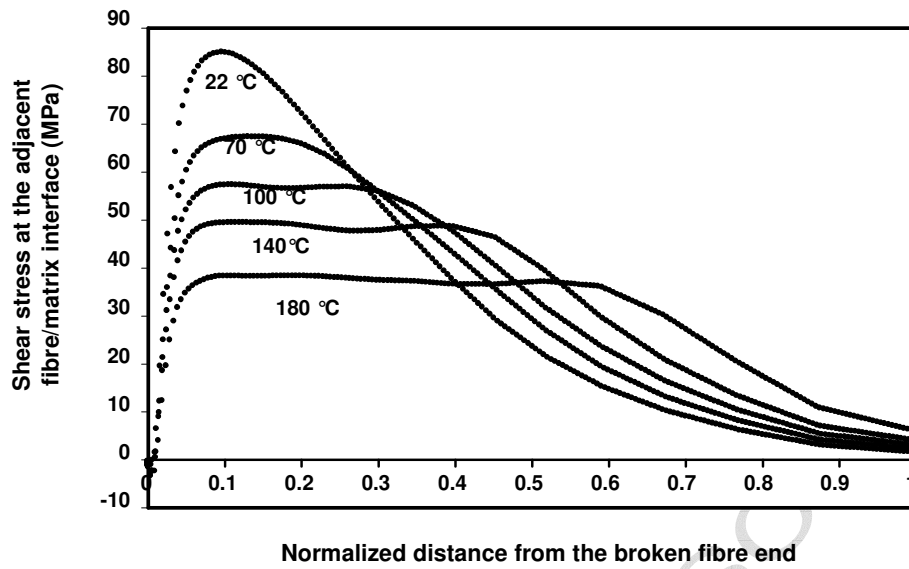


Fig. 7. Profiles of the shear stress at the neighbouring fibre/matrix interface adjacent to a fibre-break in a UD composite containing the resin matrix MY0510/DDS at an applied strain of 1 % and over a range of temperatures.

Tables

Table 1 Mechanical properties of the carbon fibre employed in the FE model taken from elsewhere [46]

Type	HTA5131	
Diameter (μm)	7.0	
Elongation to break (%)	1.43	
Young's modulus (GPa)	E_z	235
	$E_x = E_y$	13.8
	ν_{yx}	0.35
Poisson's ratio	ν_{zy}	0.2
	ν_{zx}	0.2
	G_{xy}	5.11
Shear modulus (GPa)	$G_{yz} = G_{xz}$	18

Table 2 Compressive yield strength (σ_{cy}), yield strain (ϵ_{cy}) and compressive modulus (E_c) of epoxy resin MY0510/DDS over a range of temperatures and the strain rate of $1.67 \times 10^{-4} \text{ s}^{-1}$ (\pm standard error).

Temperature ($^{\circ}\text{C}$)	σ_{cy} (MPa)	ϵ_{cy}	E_c (GPa)
22	167.4 ± 0.4	0.09	3.82 ± 0.01
70	127.3 ± 0.3	0.07	3.42 ± 0.01
100	106.8 ± 0.1	0.06	3.33 ± 0.00
140	88.7 ± 0.1	0.04	2.94 ± 0.03
180	67.1 ± 0.2	0.03	2.49 ± 0.02

Table 3 The strain concentration factors (SCF) in the neighbouring fibre to a broken fibre in a UD composite containing the resin matrix MY0510/DDS as a function of temperature.

Temperature (°C)	22	70	100	140	180
SCF	1.193	1.191	1.183	1.172	1.146

Table 4 The variation in the fractional ineffective length and far-field strain in the broken fibre with the changes in the matrix yield strength as a result of the temperature in a UD composite containing the resin matrix MY0510/DDS.

Temperature (°C)	Matrix yield strength (MPa)	Fractional ineffective length	Far-field strain (%) [Max = 1 %]
22	167.4	0.85	0.89
70	127.3	0.89	0.87
100	106.8	0.92	0.85
140	88.7	0.94	0.82
180	67.1	0.96	0.73