Numerical modeling on compressive behaviors of 3-D braided composites under high temperatures at microstructure level

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Abstract

This paper reports a numerical modeling on the compressive behaviors of 3-D braided carbon / epoxy composites with different braided angles under high temperatures at microstructure level. The coupling effect of temperature and braided angle on the compressive failure mechanisms and damage morphologies have been calculated with finite element method (FEM) and compared with experimental results. In the FEM model, the imperfect interfacial bonding and the fibers with waviness were introduced into the FEM model. The ductile/brittle transition temperature was investigated and the ductile failure criterion was employed in the FEM model at high temperatures. It was found that braided angle affects the compressive failure. The compressive failure changes from ductile to brittle when the braided angle changes from larger to smaller one. The high temperature has almost no influence on the damage mode but it leads to the failure modes more ductile. The thermal stress improves the longitudinal compressive properties. The higher temperature and larger braided angle lead to the ductile behaviors of the 3D braided under longitudinal compression.

Keywords: 3D braided composite; high temperature; ductile/brittle transition; failure mechanism; finite element method (FEM).

1. Introduction

Braided composites are often used in high temperature applications and there is very limited literature on mechanical behaviors of 3D braided composite. The study of mechanical behaviors of the 3D braided composites under high temperatures is very important in practical and in theoretical applications.

The influences of temperature on compressive behaviors of 3D braided composites have been investigated at high and low temperature in experimental [1-5]. In numerical, representative volume cell (RVC) method [6, 7], multi-unit cell model (MUCM) [8-10] and multi-scale structure model [11] were often employed to predict mechanical behaviors of 3D braided composites. However, the damage morphologies and failure mechanisms cannot be revealed from the three kinds of geometrical models because the geometrical structures of braided preform and resin were homogenized into a RVCs or unit cells.

Here we report a full-scale microstructure model of 3D braided composites for the investigation of the compressive damage under high temperature environment. The influence of temperatures and the braided angles on the compressive behaviors will be studied from experimental and finite element analyses (FEA) approach. The coupling effect of the thermo-mechanical effect was introduced in the FEA model. The compressive behaviors and damage morphologies will be presented from the experimental results and numerical simulations. We hope the comparisons between the experimental and numerical will verify the FEA model and furthermore will find the compressive failure mechanisms from the microstructure level.

2. Materials and tests

Three-dimensional (3D) braided carbon fiber tows (T700-12k) preform was prepared with 4-step 1×1 braided method in yarn array of 10×10 as shown in Fig. 1(a). Epoxy resin was injected into the 3D braided preform with vacuum assisted resin transfer molding (VARTM) technique as shown in Fig. 1(b). The resin is composed of epoxy resin (JC-02A) and hardener (JC-02B) with a weight ratio of 10:8. Fig. 1(c) presents the cured samples. Table 1 lists the coupon specifications.

Fig. 2 shows a self-designed high-temperature chamber. TC1 and TC2 are thermocouples used to monitor temperature. TC1 detects the internal temperature of the reference specimen and TC2 detects the environment temperature. The compression tests were conducted on an MTS810 servo-hydraulic materials tester. All the specimens were compressed at a velocity of 2 mm/min. The test was repeated three times at each condition.

3. Full-scale microstructure model

3.1. Modeling approach

Braided composites consist of yarn and resin matrix. The yarn, which is used as reinforcement in the braided composites, is composed of fiber and resin. Usually, the interface between fiber and resin is assumed perfect and the fibers inside the yarn are aligned perfectly but in reality, the interfaces are not perfectly bonded which influence the mechanical properties of composites. Here the imperfect interfacial bonding and the fibers with waviness were introduced into the full-scale microstructure model. Fig. 3 shows the framework of the FEM model for 3D braided composites.

3.2. Geometry model

The geometry model of braided preform is established based on the real structure as shown in Fig. 4. This model ensures that the adjacent faces among yarns are parallel due to the compaction among yarns and the surface of the braided preform is flat because of the pressure of mould during the composite consolidation. The cross section shape of the braiding yarns is assumed to be hexagon as shown in Fig. 4(b). The matrix is cut from the geometric model of the braided preform. The related geometric parameters for the hexagon section are derived as follows:

$$\begin{aligned} Y_a &= V(y) \times \frac{\sqrt{1 + \cot^2 \alpha}}{4\sqrt{2 + \cot^2 \alpha}} W \\ Y_w &= V(y) \times \frac{\sqrt{2}}{4} W \\ Y_d &= 2Y_a \end{aligned} \quad (1)$$

where $V(y)$ is a coefficient related to the yarn volume fraction; W is the repeating unit

width; α is the braided angle. In this article, the value of $V(y)$ is 0.7.

3.3. Material model

It is assumed that the braided yarns are transversely isotropic unidirectional composites and the resin is isotropic material. A unique local coordinate system for each yarn is defined corresponding to the yarn axial direction because of the variation of the yarn space trajectory.

The carbon fiber is assumed as a linear elastic material. The resin is assumed to be a elastic-plastic solid based on a Von Mises yield criterion [12]. The braided yarn is constructed as an elastic-plastic material obeying the Hill's potential function [13].

Hill's potential function is an extension of the Von Mises function.

$$f(\sigma) = \sqrt{F(\sigma_{22} - \sigma_{33})^2 + G(\sigma_{33} - \sigma_{11})^2 + H(\sigma_{11} - \sigma_{22})^2 + 2L\sigma_{23}^2 + 2M\sigma_{31}^2 + 2N\sigma_{12}^2} \quad (2)$$

For the yarns, which are orthotropic materials, the Hill's potential function is simplified into [14]:

$$\begin{aligned} f(\sigma) &= \sqrt{F(\sigma_{22} - \sigma_{33})^2 + 2L\sigma_{23}^2 + 2M(\sigma_{31}^2 + \sigma_{12}^2)} \\ F &= \frac{(\sigma_0)^2}{2} \left(\frac{1}{\bar{\sigma}_{22}^2} + \frac{1}{\bar{\sigma}_{33}^2} - \frac{1}{\bar{\sigma}_{11}^2} \right) \\ L &= \frac{3}{2} \left(\frac{\tau_0}{\bar{\sigma}_{23}} \right)^2 \\ M &= \frac{3}{2} \left(\frac{\tau_0}{\bar{\sigma}_{13}} \right)^2 \end{aligned} \quad (3)$$

where $\bar{\sigma}_{ij}$ is the yield stress when σ_{ij} is applied as the only nonzero stress component.

Fig. 5 is a multi-scale structural model of non-ideal yarns. The imperfect interface and the fiber waviness were considered in this model. The generalized method of cell (GMC) was adopted to introduce the imperfect interface into the yarn.

3.3.1 Imperfect bonding interface

Constitutive equations for the unidirectional fiber-reinforced composites with

imperfect bonding is given in Reference [15]. The braided yarn can be taken as a unidirectional fiber-reinforced composite. Moreover, the following assumptions are made in the continuum model of yarn:

- (1) The continuous fibers are arranged at equal spacing l_2 periodically in the transverse direction;
- (2) Square sections of fibers;
- (3) Void free braided yarns.

Fig. 5 exhibits the representative cell of a yarn fragment, having four sub cells. The sub cell ($\beta=\gamma=1$) represents the carbon fiber and the others represent the resin matrix. The interfaces exist only between the fiber and the resin matrix. The weak interface bonding is submitted at the fiber/matrix interface. The linear jump of displacements across the interface accord with the following relations:

$$\begin{aligned} u_n &= R_n \sigma_n \\ u_t &= R_t \sigma_t \end{aligned} \quad (4)$$

where u , σ are the displacement and traction respectively; and the subscript n , t denotes the normal component and the tangential component, respectively. R_n , R_t are the interface separation factors. If the interface is a perfect bonding, both the value of R_n , R_t are zero. Based on the GMC method, the constitutive equations of the representative cell of yarn in the elastic range can be derived [15].

3.3.2 Fiber waviness

A model of unidirectional composite with uniform waviness is applied to represent the fiber waviness in yarns as shown in Fig. 5. The matrix of yarn with fiber waviness can be derived according to the reference [16, 17].

3.4. Progressive damage model

The Tsai-Wu strength criterion [18] is used to judge the damage of yarns.

$$F_i \sigma_i + F_{ij} \sigma_i \sigma_j = 1 \quad (5)$$

where F_i and F_{ij} are the factors of strength tensors. For the yarn, the expanded form is:

$$F_1 \sigma_1 + F_2 (\sigma_2 + \sigma_3) + F_{11} \sigma_1^2 + F_{22} (\sigma_2^2 + \sigma_3^2) + 2F_{12} (\sigma_1 \sigma_2 + \sigma_3 \sigma_1) + 2F_{23} \sigma_2 \sigma_3 + F_{44} \sigma_4^2 + F_{55} (\sigma_5^2 + \sigma_6^2) = 1$$

(6)

The F_i and F_{ij} are determined by the tensile, compressive or shear strength of yarn along the i - j direction or plane:

$$\begin{aligned} F_1 &= \frac{1}{X_T} - \frac{1}{X_C}; F_2 = \frac{1}{Y_T} - \frac{1}{Y_C}; F_{11} = \frac{1}{X_T X_C}; F_{22} = \frac{1}{Y_T Y_C}; \\ F_{44} &= \frac{1}{S_{23}^2}; F_{55} = \frac{1}{S_{31}^2}; F_{12} = \frac{-\sqrt{F_{11} F_{22}}}{2}; F_{23} = \frac{-F_{22}}{2}; \end{aligned} \quad (7)$$

where T, C denote the tensile or compressive strength respectively; and X, Y stand for the axial or transverse direction respectively; S_{ij} are pure shear strengths. The axial strengths can be estimated by the modified rule of mixture as reference [19].

The transverse strength of yarn is difficult to determine because of the nonlinear features of matrix near failure and the uniform distribution of stress in matrix. Therefore the following predictions are made based on the reference [19]. If the matrix exhibits brittle failure or the premature interface fails ahead of the yield of matrix, the transverse tensile strength will always be less than the matrix tensile strength. Otherwise, if the matrix is ductile, the transverse tensile strength can be greater than the matrix tensile strength.

In the case of high temperature, the compressive behaviors of the epoxy matrix present obvious ductile failure. The interfacial failure does not dominate in 3D braided composites at high temperatures, especially beyond the glass transition temperature. Therefore, here the transverse strength of yarn is assumed to be greater than the matrix tensile strength. The transverse compressive strength is four times as high as the transverse tensile strength. In addition, the mechanisms of shear failure are similar to the transverse tension failure.

If the left side of the equation (6) exceeds 1, the failure occurs and the mechanical properties of braided yarns will be degraded.

The epoxy matrix is isotropic and strongly temperature-dependent. The failure of matrix may be ductile or brittle depending upon the temperature [20]. Furthermore, the failure theory of matrix should be applied corresponding to the temperature. So the determination of ductile/brittle transition is an essential part of the failure theory.

A thermomechanical theory of ductile/brittle failure behavior is proposed and the ductile/brittle transition temperature can be predicted by [21]

$$\begin{aligned}\Theta_{D/B} &= \Theta_0 + \frac{C_m}{9\alpha_m k_m} \left[1 - 3 \frac{T_m}{C_m} + \sqrt{1 - 3 \frac{T_m}{C_m} + 3 \left(\frac{T_m}{C_m} \right)^2} \right], \frac{T_m}{C_m} \geq \frac{1}{2} \\ \Theta_{D/B} &= \Theta_0 + \frac{C_m}{6\alpha_m k_m} \left(1 - 2 \frac{T_m}{C_m} \right), \frac{T_m}{C_m} \leq \frac{1}{2}\end{aligned}\quad (8)$$

Θ : temperature; Θ_0 : the ambient temperature; α_m : the linear CTE; k_m : the bulk modulus; C_m : the compressive strength; T_m : the tensile strength.

For the environmental temperature not exact at the ductile/brittle transition, the ductile versus brittle states of failure are specified by

$$\begin{aligned}\Theta &< \Theta_{D/B}, \text{ brittle} \\ \Theta &> \Theta_{D/B}, \text{ ductile}\end{aligned}\quad (9)$$

In this article, the ratio of T_m/C_m of epoxy matrix is 2/3, the ambient temperature is 20°C, the bulk modulus is 2.44GPa, the compressive strength is 97.6MPa and the CTE of epoxy is $52.5 \times 10^{-6} / ^\circ\text{C}$. As a result, the prediction for the ductile/brittle transition temperature of epoxy is -15.8°C. Therefore, the ductile failure theory can be applied for the epoxy matrix at high temperatures fields (80°C, 110°C, 140°C). In addition, the epoxy matrix owns a higher tensile strength than the compressive strength due to the influence of hydrostatic pressure, and a modified version of the von Mises failure criterion in terms of the initial tensile strength and compressive strength is given [22]:

$$\left(\frac{1}{T_m} - \frac{1}{C_m} \right) \sigma_{ii} + \frac{3}{2C_m T_m} s_{ij} s_{ij} \leq 1 \quad (10)$$

where s_{ij} is the deviatoric stress tensor. To ensure computational stability, the maximum value of degradation for matrix was set to 0.9 in this article.

3.5. Thermoelastic expansion coefficients

The braided yarn is assumed as a unidirectional composite. Through the Rosen-Hashin's CTE formulas [23], the effective CTE of yarn is derived as

$$\begin{aligned}\bar{\alpha}_{ij}^y &= V_f \alpha_{ij}^f + (1 - V_f) \alpha_{ij}^m + (\alpha_{kl}^f - \alpha_{kl}^m) P_{klmn} (\bar{S}_{mnij}^y - V_f S_{mnij}^f - (1 - V_f) S_{mnij}^m) \\ P_{klmn} (S_{mnij}^f - S_{mnij}^m) &= I_{klij}\end{aligned}\quad (11)$$

where α is the CTE; the superscript '-' means the effective component; f, m and y denote the fiber, matrix and yarn, respectively; S is the compliance matrix and the I_{klij} is a fourth rank symmetric unit tensor.

4. Results and discussion

Hexahedral mesh grids with C3D8R were developed for the regular braided yarns. The tetrahedral meshes with C3D4 were adopted for the mesh discretization of the matrix due to the structural complexity. At room temperature (20°C) no residual stress in composites was assumed. The temperature distributions of composites were uniform and stable during the tests at high temperature fields. The experimental stress-strain curves are presented in Fig. 6. The mechanical parameters used in this article are listed in Tables 2-5.

4.1. The stress-strain curves and mechanical properties

Fig. 7 presents the experimental and numerical calculated stress-strain curves of the longitudinal compression of 3D braided composites with different braided angles at various temperatures. The compression modulus and failure stress of the 3D braided composites are presented in Fig. 8 and Fig. 9 respectively. Through the comparison of the experiment and the FEM results, the FEM model is proved to be effective and the discrepancies are acceptable. It can be concluded from the results that the high temperature and large braided angle decrease the modulus and failure stress of 3D braided composites. From the shape of stress-strain curves, we can see that the 3D braided composites for 48° braided angle possess the characteristic of ductile failure and own a wide yielding plateau after the yielding point. However, the 3D braided composites for 35° or 26° braided angle possess the characteristics of fracture failure and exhibit a decline stage in the stress-strain curves after peak point. Therefore, from the large braided angle to the small braided angle, there is a ductile/brittle failure

transition angle in the longitudinal compression. Above the transition angle, the longitudinal compression behavior of 3D braided composites will exhibit ductile features. Otherwise, the brittle failure will dominate in the process of longitudinal compression.

In order to find the influence of temperature on the ductile/brittle states of failure, the strain energy density was extracted from the stress-strain curves before the peak point and the plastic part as well as elastic part was distinguished out as shown in Fig. 10. The percentage on the side stands for the ratio of the elastic or plastic part to the total. The proportion of plastic part can be adopted to characterize the ductile/brittle states of failure to some extent. Moreover, greater proportion of plastic part results in more ductile failure. Although both the angles of 35° and 26° exhibit a brittle failure, the 35° presents more ductile compared to the 26° . Although the effect is not obvious as that of the braided angle, the high temperature also makes the failure of 3D braided composites more ductile.

4.2. Stress distributions at various temperatures

Fig. 11 shows the stress distributions at the same states of the 3D braided composites for 35° braided angle at various temperatures. Although, the stress distribution modality is similar at various temperatures, the temperature makes much difference with regard to the level of stress value of matrix. The matrix of composite at high temperature fields bears a low stress with little damage. On the other hand, as compared to resin matrix the stress properties of reinforcement yarns are less affected by change in temperature. Therefore, the way that the high temperature influences the bearing capacity of the composites is mainly through changing the mechanical properties of the matrix.

4.3. Thermal stress

4.3.1. Thermal stress distribution

Fig. 12 presents the thermal stress distribution along an edge of yarn and the neighboring matrix at various high temperatures, respectively. The thermal stress at

140°C owns a higher value. The same interval of temperature results in the similar change of thermal strain. However, due to the decrease of the modulus, the increase of thermal stress is not even and the increase of thermal stress at 110°C to 140°C is less than that at 80°C to 110°C. The increasing range of thermal stress decreases with the increase of temperature. Owing to the edge effect, the value of the thermal stress at the end is a little different from the internal part. The thermal stress of matrix at the internal part was averaged for the convenience of comparison with the matrix failure stress as presented in Fig. 12(b). The thermal stress of matrix with different braided angle is almost the same. However, the high temperature increases the thermal stress of matrix. Conversely, the high temperature decreases the failure stress of matrix. As a result, the ratio of the thermal stress to the failure stress of matrix increases from $8.2\% \pm 0.3\%$ at 80°C to $19.7\% \pm 0.9\%$ at 140°C. It means, the impact of thermal stress on the failure behaviors of composites may be more apparent at 140°C temperature.

4.3.2. The influence of thermal stress on longitudinal compression

In order to further investigate how the thermal stress impacts the longitudinal compression behaviors of 3D braided composites, the comparison of FEM results with normal CTE and zero CTE was conducted as shown in Fig. 13. The stress distributions of the epoxy matrix with the normal CTE owns a slightly higher stress than that with zero CTE at the same strain. Conversely, the yarns with zero CTE own a slightly higher stress than that with normal CTE. The non-uniformity of yarn and matrix in CTE leads to the thermal stress at the interface. Moreover, the continuous deformation and the non-uniformity in CTE will result in that one medium endures pressure stress and the other endures pulling stress. The longitudinal stress curves of a node on yarn at the center of composite and the Von Mises stress curves of a node on the adjacent matrix were extracted from the FEM results as presented in Fig. 13(b). The initial thermal stress has different influence on the yarn and the matrix. The matrix with normal CTE yields ahead of that with zero CTE due to the pre-generated thermal stress. However, the thermal stress on the yarn act a role of puller which will offset some stress in the latter compression process. Therefore, the yarn stress with normal CTE will be less than that with zero CTE. This difference between the normal

CTE and zero CTE will influence the longitudinal compression behaviors around the failure as shown in Fig. 13(c). The failure of composite with zero CTE is slightly ahead of the failure of composite with normal CTE. To some extent, the thermal stress does benefit to the longitudinal compressive properties of 3D braided composites in high temperature fields.

4.4. Damage analysis

4.4.1. Damage evolution versus stress-strain curves with different braided angles

In order to reveal the different failure mechanisms with different braided angles, the damage status of the yarns and the matrix at three moments around the peak stress point are listed in Fig. 14-16. For 26° braided angle, the matrix of 3D braided composites has a similar damage status around the peak stress point. However, for 26° braided angle the surface yarns are more damaged after the peak stress point which results in the sharp drop of the stress after the peak stress point. The phenomenon manifests that the failure mechanism for 26° braided angle is mainly the fracture of the yarns. For 35° braided angle, the damage of matrix obviously gets a lot more and forms a zigzag damage zone on profile after the peak stress point. For 35° braided angle, the damage of yarns gets slightly worse after peak stress point. The damage of the matrix dominates the failure for 35° braided angle, and this is combined with some fracture of yarns. For 48° braided angle, the damage of the matrix is similar around the peak stress point and less pronounced as compared to damage at 35° and 26°. The yarns for 48° braided angle almost have no obvious damage. The damage of the yarns or the matrix may not be the main failure mechanism of the 3D braided composites for 48° braided angle but the plastic yielding of the matrix. Therefore, For 48° braided angle, the failure of 3D braided composites is more ductile and for 35° and 26° braided angle failure is more brittle.

4.4.2. Damage spatial distribution with different braided angles

The damage spatial distributions of the 3D braided composites at 80°C are presented by semi transparent figures as shown in Fig. 17. The damage for 26° braided angle is mainly concentrated at the load end which is also the free end of yarns. The damage for 35° braided angle is not only at the load end but also around the surface yarn and

inside the composites. Moreover, a shear damage band was formed that could be seen from the transverse view. However, relatively little damage only appears around the surface yarns for 48° braided angle.

4.4.3. Surface yarn deformation on the 3D braided composites

Six surface yarns at the free end of the 3D braided composites are chosen to investigate the deformation in order to deeply understand the damage morphologies as shown in Fig. 18. The red lines stand for the original distance between the two sides of surface yarns. The blue lines stand for the distance between the two sides of surface yarns under longitudinal compression. To facilitate comparison, the distance between the red lines with the different braided angles are adjusted to the same size as presented in Fig. 18. It is apparently that the distance between the two sides of surface yarns for 26° and 35° braided angle gets longer than that for 48° braided angle after compression. For 48° braided angle, the distance between two sides of surface yarns is almost unchanged. That means the constraint of yarns at the end of composites with 48° braided angle is much better and the relative movement of yarns is small. In addition, the larger change of distance signifies the larger dislocation of yarns, and consequently, the matrix between the yarns is apt to form the stress concentration. The stress concentration will lead to the damage of the matrix. That means, the 3D braided composite with a small braided angle is apt to generate damage around the surface yarns at the free end under longitudinal compression.

Fig. 19 presents the damage evolution process of the 3D braided composite for 35° braided angle from the transverse view. In the beginning, the damage occurs from a side of the composite end. Then, the damage extends along the yarn trajectory direction and the damage also appears around the surface yarns at the same time. Finally, the damage runs through the composite and a zigzag damage zone is formed.

4.4.4. Experimental photographs of damage morphologies

Fig. 20 exhibits the optical photographs of the damage morphologies of the 3D braided composites with different braided angles at various temperatures. There is an obvious fracture of yarns for 26° braided angle and the damage direction is along the braided angle. In addition, the damage of the matrix appears at the end of the

composites for 26° and 35° braided angle. However, the composites for 48° braided angle have no obvious damage. Apart from the above, the damage of the composites at a higher temperature of 140°C becomes slightly less compared with the composites at 80°C. All of these manifest the consistency of the FEM with the experiments.

5. Conclusions

The coupling effect of the braided angle and temperature on the longitudinal compression damage of 3D braided composite was tested and analyzed from quasi-static compression tests and numerical simulation. The ductile/brittle transition temperature is investigated and the ductile failure criterion is adopted for the matrix at the high temperature fields. The ductile/brittle states of failure and the damage modes of the 3D braided composites are strongly dependent on the braided angle. From the large braided angle to the small braided angle, the failure of composites changes from ductile to brittle. The high temperature hardly changes the damage mode but makes the failure of composites more ductile. In addition, the existence of the thermal stress at high temperature fields somewhat improves the longitudinal compressive properties of 3D braided composites. In conclusion, the higher temperature and larger braided angle increase the ductile features of the 3D braided carbon fiber/epoxy composites under longitudinal compression.

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Fig. 1 Sample preparation (a) 4-step 1×1 braided method; (b) curing by VARTM; (c) samples with different braided angles

Fig. 2 Experimental setup and the schematic diagram of high-temperature chamber

Fig. 3 Framework of FEM model for 3D braided composites

Fig. 4 Geometric model

Fig. 5 Multi-scale structures of yarns

Fig. 6 Experimental stress-strain curves (a) 80°C; (b) 110°C; (c) 140°C

Fig. 7 Comparison between experimental and FEM (a) 48°; (b) 35°; (c) 26°

Fig. 8 Compression modulus of 3D braided composites at various temperatures

Fig. 9 Failure stress of 3D braided composites at various temperatures

Fig. 10 Strain energy density at various temperatures

Fig. 11 Stress distributions of 3D braided composites for 35° braided angle at various temperatures

(a)

(b)

Fig. 12 Thermal stress distribution along yarn by axial location at various temperatures (a) yarn; (b) matrix

(a)

(b)

(c)

Fig. 13 Comparison of FEM results with normal CTE and zero CTE for 35° braided angle at 140°C (a) stress distributions; (b) stress evolution curves of the nodes on the yarn and the adjacent matrix at the center part of composite; (c) stress-strain curves of composites

Fig. 14 Damage evolution of the 3D braided composite for 26° braided angle at 80°C

Fig. 15 Damage evolution of the 3D braided composite for 35° braided angle at 80°C

Fig. 16 Damage evolution of the 3D braided composite for 48° braided angle at 80°C

Fig. 17 Damage spatial distribution of the 3D braided composites with different braided angles at 80°C

Fig. 18 The surface yarn deformation on the edge of the 3D braided composites with different braided angles at 80°C

Fig. 19 Evolution of the matrix damage and the stress distribution of yarns for 35° braided angle at 80°C

Fig. 20 The damage morphologies of the 3D braided composites with different braided angles at various temperatures

Table 1 Coupon specifications

braided angle (°)	pitch length (mm)	specimen size (mm)	fiber volume fraction (%)	temperature (°C)
48	2.2	$12 \times 12 \times 12$	32 ± 2	80, 110, 140
35	3.4			
26	4.8			

Table 2 Mechanical parameters of carbon fiber

E_{11} (GPa)	E_{22} (GPa)	ν_{12}	ν_{23}	G_{12} (GPa)	α_{11} ($10^{-6}/^{\circ}\text{C}$)	α_{22} ($10^{-6}/^{\circ}\text{C}$)	XT (MPa)
230	11.8	0.2	0.25	9	-0.54	10.1	4900

Table 3 Mechanical parameters of epoxy resin

E_m (MPa)	ν_m	α_m ($10^{-6}/^{\circ}\text{C}$)	T_m (MPa)	C_m (MPa)	temperature (°C)
2192	0.35	52.5	65	98	20
2029	0.35	52.5	58	87	80
1927	0.35	52.5	52	78	110
1616	0.35	52.5	39	58	140

Table 4 Equivalent engineering constants of yarn

E_{11} (MPa)	E_{22} (MPa)	ν_{12}	ν_{23}	G_{12} (MPa)	G_{23} (MPa)	temperature (°C)
89485	1234	0.2722	0.2816	612	482	80
89245	1229	0.2723	0.2814	608	480	110
80211	1149	0.2773	0.2804	560	448	140

Table 5 Equivalent CTE of yarn

longitudinal CTE (10^{-7})	transverse CTE (10^{-5})	temperature ($^{\circ}\text{C}$)
-3.75	1.53	80
-3.80	1.53	110
-4.10	1.52	140