Fiber Alignment in Directed Carbon Fiber Preforms – Mechanical Property Prediction

L. T. HARPER,* T. A. TURNER, J. R. B. MARTIN AND N. A. WARRIOR
Division of Materials, Mechanics and Structures, University of Nottingham
NG7 2RD, UK

ABSTRACT: A finite element method is presented for predicting the mechanical performance of discontinuous fiber mesostructures typically produced by directed carbon fiber preforming. High-filament count bundles are modeled using beam elements to enable large representative volume elements to be studied. The beams are attached to a regular grid of 2D continuum elements, which represent the matrix material, using an embedded element technique. The model is validated by comparing simulations with experimental data for random and aligned fiber architectures produced with different tow sizes (6 and 24 K) and fiber lengths (28, 58, and 115 mm). Stiffness and strength predictions are generally within 10% for 6 K preforms, but this error increases up to 40% with increasing tow size because of the assumption that the fiber bundles are circular in cross-section.

KEY WORDS: carbon fiber, mechanical properties, discontinuous reinforcement, preform.

INTRODUCTION

DISCONTINUOUS FIBER COMPOSITES are attractive in their versatility of properties and relatively low manufacturing costs compared with textile-based systems. These composite materials are in high demand for automotive, marine, and aerospace applications, but have historically suffered from low-fiber volume fractions, poor consistency, and low specific properties. Recent advancements in directed carbon fiber preforming (DCFP) have addressed some of these limitations [1], and mechanical properties are now approaching those of continuous, quasi-isotropic laminates [2].

DCFP consists of a robot-mounted mechanical chopper head, which sprays carbon fibers and a powdered binder onto a perforated tool. Air is evacuated from the underside of the tool and the resulting pressure differential holds the deposited fibers in place.

*Author to whom correspondence should be addressed. E-mail: lee.harper@nottingham.ac.uk
Figures 1 and 3–6 appear in color online: http://jcm.sagepub.com
A matched perforated tool is lowered to control the preform thickness as the binder is cured at temperature. Although the process utilizes low cost, high-filament count tows and wastage levels are low (<3% scrap) compared with woven fabrics, the uptake of the technology into industry has been restricted by poor preform repeatability compared with textile systems. The use of high-filament count rovings (≥24 K) has been identified as the main cause of local volume fraction variation and high scatter in material properties [3]. This has previously limited the use of directed fiber preforms to semi-structural and cosmetic applications [4]. The detrimental effects of high-filament count tows can be minimized, to a certain extent, by using shorter fiber bundles to improve macroscopic homogeneity [5]. In addition, a filamentization technique has been developed to fragment the high-filament count tows at the fiber chopping stage [1], to create low-filament count bundles online from low cost, nonaerospace grade tows. Local volume fraction variations are reduced from ±15% to less than ±5% [5], whilst component costs are competitive at intermediate production volumes (1000–10,000 ppa) [6].

In order to make the necessary step change in mechanical properties for structural applications it is recognized that significant levels of fiber alignment are necessary. A mechanical alignment method described in [7] can be used to bias the fiber-orientation distribution in the loading direction. Experimental characterization indicates that up to 94% of fiber bundles can be aligned within ±10° and consequently, tensile stiffness and strength can be increased by 206% and 234%, respectively, over the random fiber case. These alignment properties equate to maximum stiffness and strength retention values of 83% and 31% compared to continuous unidirectional material.

A basic requirement for all new materials and processes is that they can be modeled and their performance predicted within reasonable limits. This article presents a fiber network approach in conjunction with a finite element method (FEM) to predict the tensile properties of both random and aligned discontinuous fiber architectures. The fiber network modeler is used to investigate the heterogeneity induced during the fiber deposition stage and the FEM approach enables the fibers and resin to be modeled as distinct constituents. A multi-scale approach is adopted to account for the complex meso-scale architecture and the potential effects of tow fragmentation. The model is used here to predict the tensile stiffness and strength for a range of fiber architectures. Of particular interest are the influence of increasing fiber length and increasing bundle filament count which are known to affect the preform homogeneity [3,5,8]. Predictions from the model are validated using experimental tensile data from [7]. The model is also utilized to study the effect of parameters that cannot be conducted experimentally. This includes establishing which parameter is the more dominant in terms of in-plane mechanical properties — the level of fiber alignment or the discontinuous nature of the fibers.

**MECHANICAL PROPERTY PREDICTION**

**Modeling of Discontinuous Fibers**

High-filament count bundles (≥6 K) and long fiber lengths (≥23 mm) show the most potential for industrial DFP processes, because large tows offer major cost savings and longer fibers offer better preform rigidity and help to minimize fiber washing during molding. However, it is difficult to predict the mechanical performance of these coarse fiber mesostructures because of the relatively high levels of heterogeneity caused by poor
fiber coverage. A review by the authors in [3] has shown that conventional analytical methods, such as statistical averaging [9,10] or classical laminate theory [11] do not adequately model the mesoscopic fiber architecture exhibited by DCFP. These homogenization methods involve performing a weighted average over a range of fiber orientations, without any detailed consideration for the fiber architecture. An even distribution of filaments is assumed throughout the laminate volume. There is no consideration for the stochastic variation induced by the fiber deposition process and the resin rich regions, which consequently form. Furthermore, high-filament count bundles are of the same scale as the laminate thickness, causing extremely poor fiber dispersion through the thickness of the laminate. Eduljee et al. [12] developed an analytical micromechanical model to distinguish between dispersed and aggregated filaments, demonstrating that dispersed textures provide enhanced reinforcement over bundled aggregate textures. However, each domain (collection of parallel fibers) is considered to be connected for laminates containing small bundles, enabling a uniform strain field (Voigt averaging) to be assumed. The model is, therefore, only applicable for relatively small bundles (400 filaments for SMC) and not for coarse architectures (24,000 filaments for DCFP).

Generally, micromechanical analytical models can produce reasonable estimates for tensile stiffness (typically to within ±20%) [3] because this is essentially a volume averaged quantity, but predictions for ultimate strength are over-predicted by up to 100% [5]. Unlike stiffness, the strength of a discontinuous fiber will never attain that of a continuous fiber. The length at which a plateau in strength occurs may be up to 10 times longer than the corresponding critical length for stiffness [13–15]. The iso-strain condition assumed by the rule of mixtures (ROM) does not hold true for discontinuous fibers because of the presence of fiber ends. Kelly and Tyson [16] derived expressions (the well-known Slip Theory) to account for the effect of fiber length on strength. The failure mechanism of the composite is governed by the fiber length, which is initially dominated by debonding of transverse fibers followed by matrix cracking [17]. Fibers in the loading direction will eventually debond if the length is below a critical threshold; otherwise final laminate failure occurs due to fiber fracture for longer fibers. Kelly and Tyson introduced the critical length term into the ROM in the form of an efficiency factor.

Other work has utilized a laminate analogy [11,18,19] to account for the random or biased fiber orientations, and an efficiency factor to reduce the effective reinforcement properties caused by the discontinuities of the fibers. The success of a laminate approach is strongly dependent upon the assumption of physical volume averaging combined with the ability to accurately estimate the properties of the unidirectional plies. Recent efforts have incorporated progressive failure mechanisms into the laminate approach to account for the accumulation of damage [20,21]. Progressive failure techniques have been limited to the macroscopic level since they only account for the fiber–matrix interface and fiber–fiber contacts in an average sense. Shear lag [22] or slip theory [16] approaches are commonly used to understand the stresses around short fibers and the load transfer between discontinuous fibers. These models are only reliable for noncontacting aligned fibers because it is difficult to predict the axial fiber stress and the interfacial stress at fiber contact points.

The level of macroscopic homogeneity is much more important for strength than for stiffness, since failure is dominated by the weakest link. Wetherhold [23,24] has incorporated stochastic effects relating to fiber orientation, fiber length, fiber volume fraction, and laminate thickness into a modified critical zone model [25,26]. The basic concept assumes that microcracks are most likely to form at fiber ends at microscopic strains, causing the composite to fail in a critical section because of the accumulation of cracks. The strength
of the composite is determined by the relative number of bridging fibers vs. the number of ending fibers within the critical zone. The result is only given in the form of an average value because the model uses statistical methods to describe a representative fiber distribution, rather than modeling the physical architecture. Therefore, there is no consideration for local areal mass variations which strongly influence the ultimate strength.

The most promising path for the development of a comprehensive model is to capture the physical characteristics and failure modes of the discontinuous fiber material. It is important to account for the distinctive fiber architecture in order to achieve realistic results, in particular the heterogeneity of the material [27]. Ionita and Weitsman [28,29] have recently used a fiber network approach in conjunction with laminate theory to investigate the mechanical property variation and specimen size effects for random chopped fiber laminates. This method does account for stochastic fiber coverage variation and stiffness predictions are shown to be within 10% of the experimental values. However, strength prediction is limited because only the load carried by the fiber strands is taken into consideration. Stress concentrations within the matrix (caused by fiber ends or at fiber crossover points) are therefore overlooked and the strength predictions tend to over-estimate the experimental values by 25%. Eason and Ochoa [30,31] overcome these limitations by combining the random fiber network approach with a finite element method. The model is capable of capturing the nonlinear response of discontinuous random composites, by predicting damage initiation and progression within both phases of the composite. The results from their model show that the stress profile along the length of a fiber deviates greatly from the shear lag theory because of the complex fiber/fiber and fiber/matrix interactions occurring within the material. This model shows the importance of modeling the fibers and resin as distinct constituents, in order to describe the heterogeneous nature of the stress and strain fields within the composite.

Modeling Strategy

The current modeling strategy simulates the DCFP fiber architecture at the mesoscopic level, using 1D beam elements in ABAQUS (type B22) to represent tows. Beam elements are computationally inexpensive, and therefore enable larger volumes of material to be studied compared with shell or continuum elements. Inevitably there is a limit to the level of physical detail that can be modeled, but capturing the stochastic fiber distribution is considered to be more important. Each beam is assumed to have a circular cross-section, where the diameter is assigned as a function of the filament count and tow volume fraction ($V_{\text{tow}}$). Tow diameters can be randomly assigned from a probability density function to reflect the filamented mesostructure seen in experimental studies [8]. The internal structure of each bundle contains resin, and therefore the volume of deposited bundles ($V_{\text{deposited}}$) is adjusted to satisfy the target volume fraction of the laminate ($V_f$):

$$V_{\text{deposited}} = \frac{V_f}{V_{\text{tow}}}$$

The matrix material is modeled using a regular array of 2D, plane stress continuum elements (ABAQUS type CPS8R). Beam elements are tied to the solid elements using the embedded element technique, a type of multi-point constraint within ABAQUS/Standard. The translational degrees of freedom of each beam element are eliminated when it is embedded, becoming constrained to the interpolated values of the corresponding
degrees of freedom of the host (resin) element. The resulting fiber architecture is then passed to an intermediate solver interface to generate the meshes. An FE input deck is subsequently created and this is passed to ABAQUS/Standard for solution.

The current modeling strategy has been developed so that it can be universally applied to all discontinuous fiber materials. A process-specific simulation tool is used to generate the fiber network, similar to the one described in [3], since the spatial distribution of random fibers is largely dependent on the nature of the preforming/molding process. For the present work, however tows are deposited at random and, therefore their position is independent of any process induced effects.

Fibers of length \( l \) are deposited over an area \( A'B'C'D' \), which measures two fiber lengths greater in both the \( X \) and \( Y \) directions than the specified region \( ABCD \) (Figure 1). Cartesian coordinates for the centroid of each tow are assigned randomly within the preform perimeter \( A'B'C'D' \) and an orientation is generated between \(-\pi/2\) and \( \pi/2 \) according to the distribution \( f(\theta) \). The coordinates of the two fiber ends and the orientation are used to generate the intersections with the boundaries of \( ABCD \). Fiber deposition and cropping occur concurrently. The target fiber length within \( ABCD \) is determined from the laminate volume fraction and bundle dimensions. As each fiber is deposited, an algorithm checks for intersection with the boundary of \( ABCD \) and then recalculates the fiber length if cropping is performed. A cumulative total of the fiber length within \( ABCD \) is used to establish if the volume fraction requirements have been met. This method ensures that both bridging and ending fibers are captured and that the volume fraction within \( ABCD \) is

![Figure 1](https://example.com/image1.png)

*Figure 1.* (a) Fibers are initially deposited over an area \((A'B'C'D')\) which is two fiber lengths \((2l)\) longer and wider than the specified region of interest. (b) Fibers are then cropped to the ROI boundary \((ABCD)\).
always as specified. Clearly, this does not reflect reality because of the typical levels of areal mass variation across the plaques, but it enables the volume fraction variable to be isolated. Consequently however, the level of variation in the predicted properties is not as great as the level of variation seen in the experimental values. It is assumed that fibers remain straight and that the position of each fiber is exclusive, allowing bundles to intersect within the 2D resin plane. The coordinates for the fiber ends are written to a text file to be imported into the intermediate solver interface.

The random nature of the fiber distribution does not justify the use of periodic boundary conditions, since the unit cell is nonrepeating. The following boundary conditions have been chosen to represent a conventional tensile test: The displacement along edge AB (Figure 1) is zero in both the $X$-direction and $Y$-direction. The displacement along edge CD is zero in the $Y$-direction and a uniform displacement is applied in the $X$-direction. A width of 25 mm, a length of 120 mm, and a thickness of 3 mm have been used to model the section of specimen located between the jaws of a tensile test. A convergence study was performed to determine the mesh densities. Each tow is modeled using a series of 0.4 mm long beam elements, and the matrix material is represented by 0.2 mm × 0.2 mm 2D continuum elements, based on the convergence of the elastic modulus and the ultimate tensile strength for a 24 K DCFP coupon.

**Damage Prediction**

The damage model employed is an extension of the work by Zhao et al. [32]. Both constituent materials are assumed to be linear elastic and isotropic, therefore the Maximum Principal Stress failure criterion is suitable for predicting damage initiation. This simple failure criterion captures the physics of the failure mode of the brittle epoxy system under the loading arrangements observed in experimental studies. For more general purpose modeling with tougher resin systems, it would be less physically representative and would only be used with caution.

At each integration point within the model the current stress state is evaluated (for every strain increment) in order to determine if the failure criterion condition has been met. The maximum and minimum principal stresses are calculated and are compared with the failure criterion to check for damage in tension and compression, respectively. If no damage is present, the stiffness matrix is calculated using the initial elastic constants. However, if the failure stress of the matrix is exceeded in either tension or compression ($\sigma_{\text{matrix}}^t$ or $\sigma_{\text{matrix}}^c$), damage occurs and the Young’s Modulus and shear modulus are degraded to 1% of their initial values.

The material model for the fiber is much simpler than for the resin, since $S_{11}$ is the only relevant stress component for the beam elements. The fibers are also assumed to be linearly elastic and the maximum or minimum value of $S_{11}$ is used to determine whether the fiber fails in tension or compression, respectively. Following failure, the longitudinal modulus at the integration point is reduced to 1% of the initial value. This degradation scheme has been programmed into two user-defined material subroutines (UMAT), one for each constituent, which are subsequently interfaced with ABAQUS/Standard. Stiffness reduction is applied for subsequent increments until final failure of the model, which corresponds to the rupture of the unit cell and the inability to carry any further load.
EXPERIMENTAL METHODOLOGY

Materials

The fiber types modeled during this study are based on standard grades supplied by Toho Tenax GmbH, 24 K STS-J, and 6 K HTA carbon tows. Both of these tow types use 5631 polyurethane sizing, which minimizes the level of fragmentation during DCFP processing [3]. The matrix material is based on a developmental epoxy from Hexcel, designated DLS1692. A global volume fraction of 35% is assumed throughout, which was chosen to enable comparisons with the experimental data reported in [7].

Table 1 shows the calculated material properties for the liquid epoxy and the carbon bundles. Resin values were determined experimentally. The tow properties are calculated from the ROM using manufacturer’s data for carbon fiber (assuming a 60% $V_{tow}$). An average strength value has been assumed for fiber bundles, in order to isolate the geometrical effects from Weibull strength [33] effects for increasing fiber lengths. It is commonly understood that the strength of brittle fibers is a function of gauge length because of the distribution of critical flaws. However, it is not clear how this theory can be applied to parallel fibers arranged in bundle format, due to the local load sharing that takes place as individual filaments fail.

In-plane Fiber Orientation

The in-plane bundle orientation distributions for the random fiber preforms are assumed to be uniform based on tensile properties from previous studies. The in-plane bundle orientations for the aligned fiber preforms are critical to the accurate prediction of the mechanical properties. The orientation distribution $f(\theta)$ was measured using digital image analysis, as outlined in [7]. The discrete probability density functions obtained from the experimental measurements were approximated by a double exponential distribution:

$$f(\theta) = a \exp \left[ - \left( \frac{\mu - \theta}{\beta} \right)^2 \right]$$

where $\mu$ is the location parameter, $\beta$ is the shape parameter, $a$ is the normalization constant and $\theta$ is the fiber orientation in radians between $-\pi/2$ and $\pi/2$. A summary of the parameters for preforms from the present study are presented in Table 2 and the corresponding curves are presented in Figure 2.

<table>
<thead>
<tr>
<th></th>
<th>Carbon tow</th>
<th>Carbon tow</th>
<th>Epoxy resin</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Toho Tenax STS</td>
<td>Toho Tenax HTA</td>
<td>Hexcel DLS1692</td>
</tr>
<tr>
<td>Longitudinal modulus $E_{11}$ (GPa)</td>
<td>144</td>
<td>143</td>
<td>3.2</td>
</tr>
<tr>
<td>In-plane Poisson’s ratio $\nu_{12}$</td>
<td>–</td>
<td>–</td>
<td>0.38</td>
</tr>
<tr>
<td>Tensile elastic limit $\sigma'$ (MPa)</td>
<td>2424</td>
<td>2376</td>
<td>60</td>
</tr>
<tr>
<td>Compressive elastic limit $\sigma^c$ (MPa)</td>
<td>1450</td>
<td>1450</td>
<td>150</td>
</tr>
</tbody>
</table>
Digital Image Correlation

A Limess Vic3D digital image correlation (DIC) system has been used to produce full-field strain measurements of experimental tensile specimens to validate the FE model. The surface of each sample was sprayed white and then a speckle pattern was applied in black paint to obtain randomized gray level distributions. Effort was made to ensure that the diameter of the speckles were as uniform and small as possible to provide a good spatial resolution. A white light (6400 K color temperature) was used to illuminate the area of interest. Images of the specimen surface were recorded throughout the test at an acquisition rate of 0.2 Hz, via two 5.0 Mpixel CCD cameras fitted with Pentax C37500 lenses ($f = 75$ mm, 1:2.8 D). The cameras were mounted on a tripod positioned 1.5 m away from the subject to provide a 120 mm field of view. The local displacement resolution of the system is 0.01 pixels, which corresponds to 0.24 μm for the current set up. An image processing unit was used to calculate a 3D displacement field based on correlation calculations. The method consists of correlating the gray pixels in each deformed image to the counterpart in the undeformed (reference) image. The area of interest of each image was divided into small square subsets of size 20 × 20 pixels, using a step size of 5 pixels (i.e., correlation was only performed on every fifth pixel in the $x$ and $y$ directions).

Specimens for this study were manufactured using four layers of 300 gsm EF6305 epoxy film (supplied by the Advanced Composites Group) and carbon fiber bundles chopped into 58 mm lengths. Bundles were placed manually onto each film to produce a ply of relatively low-fiber volume fraction (15% compared to 35% for the experimental study in [7]). Each epoxy film was then photographed on an illuminated light box to highlight the position of the fibers, before being stacked together to create a preform. Photographs of the four layers were imported into AutoCAD® to determine the center line of each bundle as shown in Figure 3. Accuracy of this methodology is clearly limited to within ~2 mm because of manual image processing and errors associated with aligning each ply of the preform. However, this is considered to be acceptable for the current application, since the DIC is merely used to give a qualitative indicator for the accuracy of the damage model.
Figure 2. Effect of increasing fiber length on the level of fiber alignment for (a) three 6K preforms and (b) three 24K preforms. Double exponential curves are fitted to experimental data taken from [7].
RESULTS AND DISCUSSION

Damage Analysis

An example of a damage map is presented in Figure 4 for a 6K tensile specimen with a fiber length of 28 mm at fiber volume fraction of 35%. The damage is plotted at the peak load point and has been separated into matrix contribution and fiber contribution for clarity. The onset of damage in this sample occurs at a global strain of 0.2% at a number of locations across the specimen. The majority of these sites are at the bundle ends, particularly when the bundle end is isolated in a large resin rich region. These regions grow as the applied strain increases and eventually some coalesce. For the current case, the damage in the matrix progresses across the width of the specimen. As the resin elements fail, load is

Figure 3. (a) Digital photo of a tensile specimen taken from a random fiber plaque positioned on an illuminated box. The white regions indicate resin-only regions. The image was processed in AutoCAD and the yellow lines represent the bundle center lines. (b) The resulting finite element mesh for the bundles generated from the CAD geometry (resin elements not shown). Samples were 120 mm × 50 mm to represent a tensile specimen clamped between the grips of universal testing machine.
redistributed amongst the neighboring fibers until the stress-based failure criterion is violated. Failure of the fibers across the damage zone is synchronized at a global strain of 0.77%, which results in a sudden, catastrophic failure in the marked region.

It is difficult to directly validate the damage progression in random fiber materials using experimental techniques. However, the full-field strain is a good indicator, particularly for brittle materials of an elastic nature. Additional tensile samples were produced at a lower fiber volume fraction (15%) to enable the fiber architecture to be replicated easily, as outlined in the 'Digital Image Correlation section.' A comparison of strain maps produced by DIC and the finite element model are presented in Figure 5. In general, there is very good agreement between the DIC plots and the FE model. The first pair of images corresponds to a global strain of 0.02%. It is clear that a number of high strain regions have started to develop at bundle ends within the FE model, as previously discussed. These higher strain regions of ~0.225% (shown in blue) correspond very well with the location of similar regions on the DIC plot. In the second pair of images, at a global strain of 0.12%, the size of many of these regions have increased and joined together. Strain values reach approximately 1.0% in local regions, which are consistent for both the FE model and the DIC. The DIC plot in the third set of images (0.16% global strain) shows the onset of a crack forming on the left-hand edge of the specimen (A), depicted in red, which corresponds to a local strain of approximately 1.2%. This region is also clearly visible on the FE plot for the same global strain. Final failure (images not shown) occurred at a global strain of 1.6%, as the crack on the left-hand edge (A) progressed upwards at an angle of 45° to the right-hand edge to join to high strain region at (B).

**Mechanical Property Prediction**

Selections of stress/strain curves from the model are presented in Figure 6 and are compared against experimental curves for both random and aligned fiber architectures.

![Figure 4](image-url)
There is generally good agreement between the experimental and simulated data, in terms of the initial elastic modulus, the ultimate tensile strength and the strain to failure.

Average experimental stiffness and strength values are compared with predictions from the FE model in Figures 7 and 8. Tensile stiffness predictions are also included from a multi-scale Eshelby-based model [8]. Second and fourth-order orientation tensors have

Figure 5. A comparison of in-plane principal strain maps produced by the DIC equipment (left image of each pair) and the finite element model (right image of each pair). Images are shown for three different global strains: (a) 0.04%, (b) 0.12%, (c) 0.16%.

Figure 6. Comparison of stress/strain curves for both random and aligned 6 K, 58 mm plaques. Experimental curves are presented alongside predictions from the FE model.
been calculated using the probability density functions from Figure 2, in order to apply the analytical model (Table 3).

Figure 7(a) shows that the FE model is in much better agreement with the experimental stiffness data for the random plaques, compared with the analytical predictions. On average, the FE predictions are within 2% for the 6 K tow size and within 6% for the 24 K tow size.

**Figure 7.** Tensile (a) stiffness and (b) strength for random fiber preforms at a range of fiber lengths and tow sizes. Experimental values are compared against predictions from the finite element model presented in the ‘Modeling Strategy’ section and an analytical model from [8]. The experimental data points have been offset by 0.5 mm to improve clarity and are taken from [7].
size, at the three fiber lengths tested. The analytical model tends to over-predict the experimental values by \(~22\%\) for the 6K tow size and by \(38\%\) for the 24K tow size. The analytical predictions also lie outside the experimental error bounds in each case. The analytical model is very simple to implement and is useful for establishing trends,
but its accuracy is limited because of assumptions concerning the fiber architecture. The Eshelby-based approach uses a homogenization method to model the discontinuous fiber architecture using a single isolated ellipsoidal inhomogeneity embedded in an infinitely extended matrix. Therefore, the high levels of heterogeneity that often results from using a combination of high-filament count bundles (24 K+), long fiber lengths (50 mm+), low-to-intermediate fiber volume fractions (20–40%) are neglected by this method, particularly for thin structures (<3 mm). The main advantage of the FE model is that the fiber architecture is modeled explicitly by the process simulation, therefore fiber coverage effects and resulting areal mass variations are taken into consideration. Fiber–fiber contacts and fiber–matrix contacts are also accounted for Figure 7(a) which indicates that the two predicted analytical curves converge to a common value of 31.5 GPa at longer fiber lengths, whereas the FE predictions for the 6 and 24 K tow sizes converge to separate values of 22 and 27 GPa, respectively. This is also because the analytical model does not account for the high levels of preform heterogeneity.

Figure 7(b) shows a comparison of the FE predictions with the experimental tensile strength for the random preforms. Predictions for the 6 K tow size are generally within 10% of the experimental points, but the predictions for the 24 K case are 43% lower than the experimental values. This is thought to be because the fiber bundles are modeled using circular cross-sectioned beams. The 24 K tow are ribbon-like and are much flatter than is currently assumed. The effective bundle diameter is estimated to be only 1.4 mm, as each bundle is assumed to have 60% fiber content and filament diameters of 7 μm. Consequently, less fiber–fiber contact is modeled in the virtual specimen, limiting the degree of local load sharing amongst neighboring bundles. There is also an artificially high volume of unreinforced areas within the plaque when using circular sectioned beams which limits the stress transfer between bundles. The failure of the virtual specimen is therefore dominated by the matrix material rather than the fibers. Damage tends to initiate in the matrix at bundle ends due to the large stress concentrations caused by the synchronized filaments. These damaged regions coalesce as the global strain increases, circumventing the high-filament count tow by spreading through the resin rich regions. Consequently, the full mechanical potential of the bundles tends to be under-utilized, resulting in minimal fiber damage, but extensive matrix damage. For this reason, the

<table>
<thead>
<tr>
<th></th>
<th>6 K 28 mm</th>
<th>6 K 58 mm</th>
<th>6 K 115 mm</th>
<th>24 K 28 mm</th>
<th>24 K 58 mm</th>
<th>24 K 115 mm</th>
</tr>
</thead>
<tbody>
<tr>
<td>a_{11}</td>
<td>0.7664</td>
<td>0.9278</td>
<td>0.9697</td>
<td>0.8142</td>
<td>0.9560</td>
<td>0.9871</td>
</tr>
<tr>
<td>a_{22}</td>
<td>0.2286</td>
<td>0.0672</td>
<td>0.0253</td>
<td>0.1806</td>
<td>0.0390</td>
<td>0.0080</td>
</tr>
<tr>
<td>a_{33}</td>
<td>0.0050</td>
<td>0.0050</td>
<td>0.0050</td>
<td>0.0050</td>
<td>0.0050</td>
<td>0.0050</td>
</tr>
<tr>
<td>a_{1111}</td>
<td>0.6670</td>
<td>0.8756</td>
<td>0.9431</td>
<td>0.7241</td>
<td>0.9199</td>
<td>0.9746</td>
</tr>
<tr>
<td>a_{2222}</td>
<td>0.1318</td>
<td>0.0193</td>
<td>0.0033</td>
<td>0.0939</td>
<td>0.0074</td>
<td>0.0004</td>
</tr>
<tr>
<td>a_{3333}</td>
<td>0.0001</td>
<td>0.0001</td>
<td>0.0001</td>
<td>0.0001</td>
<td>0.0001</td>
<td>0.0001</td>
</tr>
<tr>
<td>a_{1122}</td>
<td>0.0956</td>
<td>0.0476</td>
<td>0.0218</td>
<td>0.0861</td>
<td>0.0314</td>
<td>0.0076</td>
</tr>
<tr>
<td>a_{1212}</td>
<td>0.0956</td>
<td>0.0476</td>
<td>0.0218</td>
<td>0.0861</td>
<td>0.0314</td>
<td>0.0076</td>
</tr>
<tr>
<td>a_{1133}</td>
<td>0.0038</td>
<td>0.0046</td>
<td>0.0048</td>
<td>0.0040</td>
<td>0.0047</td>
<td>0.0049</td>
</tr>
<tr>
<td>a_{2233}</td>
<td>0.0011</td>
<td>0.0003</td>
<td>0.0001</td>
<td>0.0009</td>
<td>0.0002</td>
<td>0.0000</td>
</tr>
<tr>
<td>a_{1313}</td>
<td>0.0038</td>
<td>0.0046</td>
<td>0.0048</td>
<td>0.0040</td>
<td>0.0047</td>
<td>0.0049</td>
</tr>
<tr>
<td>a_{2323}</td>
<td>0.0011</td>
<td>0.0003</td>
<td>0.0001</td>
<td>0.0009</td>
<td>0.0002</td>
<td>0.0000</td>
</tr>
</tbody>
</table>
model appears to under predict the ultimate strength for specimens containing larger tow sizes.

There is good agreement between the FE predictions and the experimental values for the fiber alignment plaques. On average, the stiffness predictions are within 9% and 10% of the experimental values for the 6 and 24 K tow sizes, respectively (Figure 8(a)). The strength predictions are on average, within 8% and 15% of the experimental values for the 6 and 24 K, respectively (Figure 8(b)). However, the tensile strength of the 24 K, 28 mm random plaque is under-predicted by 43% using the FE model. This is essentially because this plaque is tending to the random fiber cases discussed above, because the level of fiber homogeneity is low due to the large tow size and the level of fiber alignment is poor. The simulated failure of this plaque was, therefore unrealistically dominated by the matrix material because of the assumed circular bundle cross-section. The level of fiber alignment increases with increasing length, which improves the fiber homogeneity, thus providing better load transfer routes between the fibers. There is also an anomaly between the FE prediction and the experimental value for the strength of the 6 K, 115 mm aligned plaque. The FE model over-predicts the ultimate strength by 18% for this plaque, but the experimental data point is thought to be conservative because of the relative increase in stiffness and strength between 58 and 115 mm. The stiffness increases by 10% as the fiber length increases from 58 to 115 mm, compared with only a 7% increase for strength. Previous trends for the random plaques and comments in the literature indicate that the stiffness normally plateaus before the strength with increasing fiber length.

The FE values are compared against stiffness predictions from the analytical model in Figure 8(a). Both models are in very close agreement with the experimental values for the 6 K tow size, but the analytical model is 25% higher than the experimental data for the 24 K case. Again, this can be attributed to the disregard for the poor levels of fiber homogeneity experienced with the 24 K plaques. The analytical model assumes that all of the fibers are distributed uniformly over the specimen to yield a constant local areal mass; therefore the analytical predictions for the 6 and 24 K tow sizes are very similar (within 4%). As discussed previously, the FE model distinguishes between the different levels of homogeneity, and therefore the predictions are within 12% of the experimental values for each fiber length. Figure 8(a) also shows that the analytical predictions for the 24 K tow size are higher than the predictions for the 6 K at the longer fiber lengths. This demonstrates that higher levels of alignment were achieved for the 24 K plaques and further emphasizes the effect of disregarding the fiber coverage issues.

It is clear from Figures 7 and 8 that the standard deviations of the predicted values are considerably lower than for the experimental data points. This is because the model ensures that the fiber volume fraction for each specimen always meets the specified target (see ‘Modeling Strategy’ section), hence the modeled areal mass variations are much lower than in reality. However, the predicted standard deviations do generally increase with both increasing fiber length and increasing tow size, supporting the experimental trends. Plaques made with shorter fibers and smaller tow sizes are generally more repeatable, and therefore the design strength will be higher due to the increased confidence in material properties.

**Fiber Alignment vs. Fiber Length**

This section aims to establish which parameter is the more important in terms of in-plane mechanical properties — the level of fiber alignment or the discontinuous nature of
the fibers. The FE method has been used to simulate a number of scenarios that are not possible using the current experimental setup. These include preforms with the same 6 and 24 K tow sizes and the same 28, 58, and 115 mm fiber lengths, but with perfectly aligned fibers in the loading direction. The properties for UD preforms with continuous fibers are also simulated, in order to determine ceiling values for these materials. These simulations can be used to isolate the length effects from the orientation effects.

Figure 9(a) summarizes the tensile stiffness results from this study. There is a notable increase in stiffness when the fibers are perfectly aligned compared with using the

![Figure 9(a)](image)

---

**Figure 9.** An investigation to study the effects of improving the experimental orientation distributions to unidirectional, vs. the effect of increasing the fiber length to continuous fibers. Tensile stiffness and strength are presented in (a) and (b), respectively.
orientations from the experimental distributions. On average, the 6 K preforms show a 22% increase in stiffness and the 24 K preforms show an 18% increase, since the level of experimental alignment was reported to be higher for the 24 K plaques. Naturally, the improvement in alignment has more of an affect on the stiffness of plaques containing shorter fibers, since the level of experimental alignment was also reported to be lower in this case. The accuracy of the fiber placement is clearly important in terms of the level of stiffness retention achieved. However, the tensile stiffness of the DCFP plaques does not reach that of a continuous UD material, even for 100% alignment. Further increases in stiffness can only be attained if the fibers are longer than 115 mm.

Figure 9(a) indicates that the stiffness of the UD 6 K plaques converges towards the stiffness of the UD continuous case much earlier than the 24 K plaques. The UD 24 K, 115 mm plaque has a 90% stiffness retention at a length of 115 mm, compared with 97% for the UD 6 K, 115 mm plaque. This is because the aspect ratio \((l/d)\) of the 6 K tow is larger than the aspect ratio of the 24 K tow for any given fiber length. This reiterates the importance of using longer fiber lengths to achieve higher stiffness retention when using larger tow sizes. Longer fibers are not only easier to align with the current mechanism, but are also more likely to exceed the critical bundle length [5].

Figure 9(b) shows the influence of the fiber orientation and length on the tensile strength. The increase in strength achieved by improving the fiber alignment is negligible for both tow sizes, compared with the strength achieved when using continuous fibers. Predictions show that on average, the strength of the 6 and 24 K plaques increases by just 10% when the fibers are fully aligned compared with the experimental distributions. The strength retention for the UD 115 mm length is just 24% and 41% for the 24 and 6 K tow sizes, respectively, which is considerably lower than the equivalent stiffness retentions. Figure 9(b) indicates that the tensile strength is influenced much more by the discontinuous nature of the fibers rather than the orientation of the fibers for the range of lengths tested. The plotted curves show no signs of convergence, which implies that the ultimate tensile strength of this material would continue to increase at longer fiber lengths towards the ceiling strength of approximately 1400 MPa. Whilst longer fiber lengths would exceed the length of the tensile specimen, fiber breaks (ending fibers) would still be present within the specimen. The fiber ends are common damage initiation sites, hence it seems more logical to consider the tensile strength to be a function of the total number of fiber ends within the specimen volume, rather than to consider it to be a function of fiber length. The target ceiling strength will be attained as the number of fiber ends approaches zero.

**CONCLUSIONS**

An FE element-based damage model has been presented to predict the in-plane tensile properties for the discontinuous fiber DCFP material. Predicted stress/strain curves appear to be realistic and the engineering properties are generally within the experimental error bars for both aligned and random fiber architectures. On average, tensile stiffness predictions are within 10% and ultimate tensile strength predictions are within 20% of the experimental equivalents. This level of accuracy may not be sufficient for the model to be used directly for design purposes, but the model will serve as a useful tool during material selection and optimization studies. There are considerably more parameters to consider for discontinuous materials than textiles, making it a more daunting task to use such materials for structural applications. The model will be used to minimize the amount of
experimental testing performed, rather than replacing it altogether, by quickly identifying the significance of parameters and discovering their critical range.

The model shows significant improvement over available analytical solutions as it incorporates a better physical representation of the fiber architecture. Fiber coverage effects, local areal mass variations, and fiber/fiber interactions are all taken into consideration to provide a more accurate representation of the fiber mesostructure. A current limitation is the assumption that the fiber bundles have circular cross-sections. The effective width used for larger tow sizes is unrealistic and consequently strength predictions tend to be conservative.

The FE model has been used to further investigate the effect of misalignment and the discontinuous nature of the DCFP material. The most significant discovery is that the strength retention at 115 mm fiber length is restricted to 40%, even with fully aligned fibers. The ultimate strength of the DCFP material is dominated by the stress concentrations at the fiber ends. The study implies that very long fiber lengths are required in order to approach the ultimate strength of plaques containing UD aligned fibers.

Selective fiber alignment is seen as a valuable addition to the standard DCFP process with the potential to significantly increase mechanical properties in a particular direction locally or globally within the part. Clearly the reduced deposition rate is a major limitation from a cost point of view and limits the applicability of the process. This area will be the subject of future investigation. A further limitation is part thickness when used with the current, fan-based, fiber retention mechanism.

ACKNOWLEDGMENTS

This work was funded by the Engineering & Physical Science Research Council (EPSRC), as part of platform grant GR/T18578/01. The authors also gratefully acknowledge the support of the Nottingham Innovative Manufacturing Research Centre (NIMRC). Carbon fibers were kindly supplied by Toho Tenax Europe GmbH for this work.

REFERENCES


