Exploring the pseudo-ductility of aligned hybrid discontinuous composites using controlled fibre-type arrangements

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Abstract

Pseudo-ductility presents a potential means for preventing catastrophic failure in composite materials; large deformations will prevent brittle fracture and provide warning before final failure. This work explores how the pseudo-ductility and strength of aligned hybrid discontinuous composites can be controlled by manipulating the arrangement of different fibre types. Aligned carbon/glass hybrid specimens with different fibre arrangements are manufactured and tested using a modification to the High Performance Discontinuous Fibre (HiPerDiF) method. Experimental results are complemented by an improved virtual testing framework, which accurately captures the fracture behaviour of a range of hybrid discontinuous composite microstructures. With a randomly \textit{intermingled} fibre arrangement as a baseline, a 27\% increase in strength and a 44\% increase in pseudo-ductility can be achieved when low elongation fibres are completely \textit{isolated} from one-another. Results demonstrate that the HiPerDiF method is the current state-of-the-art for maximising the degree of intermingling and hence the pseudo-ductility of hybrid composites.

\textit{Keywords:} Pseudo-ductility, hybrid, microstructures, fracture,

1. Introduction

Composite materials are widely used in aerospace and automotive applications due to their high specific strength and specific stiffness [1]. However, composite materials often fail in a brittle manner, which may lead to catastrophic failure; consequently, composite structures are often over-designed and overweight. There is therefore a strong desire for...
pseudo-ductile composites that fail more gradually and offer more warning before final failure [2].

One means of achieving pseudo-ductility in composite materials is through the use of aligned discontinuous reinforcements, which exploit the progressive failure of the matrix or fibre-matrix interface to promote gradual failure. The pseudo-ductility of aligned discontinuous composites was verified both analytically [3] and experimentally [4] at the ply-level, before being further demonstrated at the fibre-level using the HiPerDiF (High Performance Discontinuous Fibre) method to produce aligned discontinuous fibre compos-
Hybrid fibre composites can enhance pseudo-ductility via progressive fragmentation of low elongation fibres or plies and their subsequent partial debonding or delamination from the high-elongation material [6–8]. Yu et al. [9] modified the HiPerDiF method [5] to combine hybridisation with an aligned discontinuous microstructure to create an *intermingled* aligned hybrid discontinuous composite; this material contained well-aligned hybrid fibre types which are randomly placed within the cross-section of the specimen. The *intermingled* aligned hybrid discontinuous composite demonstrated pseudo-ductile strains of over 1%, both experimentally [9] and analytically [10].

Previous research suggests that the arrangement of hybrid fibre types has a significant influence on the structural performance of continuous fibre hybrid composites [11–14]. Most of the authors [11–13] reported that an increased dispersion of the hybrid fibre types (i.e. a highly *intermingled* fibre arrangement) increases the ultimate strain of the composite, although Swolfs et al. [14] argued that an *interlaminated thin-ply* arrangement leads to a larger hybrid effect on the apparent failure strain of the low-elongation fibres. However, similar studies have not been carried out for aligned hybrid *discontinuous* composite materials.

This paper aims to explore whether the arrangement of hybrid fibre types can influence the mechanical performance of aligned hybrid discontinuous composite materials. The scope of this work will cover both highly *intermingled* hybrid fibre arrangements, as well as microstructures with more grouped fibre-type arrangements (i.e. *intraply* [9] hybrids). Section 2 describes the techniques used to manufacture the specimens and to gather the experimental data. Section 3.1 describes a method that simulates the formation of the same type of microstructures that were created during the manufacture of the experimental specimens. An existing semi-analytical model [10] to predict the response of aligned hybrid discontinuous composites will be further improved in Section 3.2 with a new fracture toughness-based failure criterion, which is formulated specifically to analyse the fracture behaviour of both *intermingled* and *intraply* aligned hybrid discontinuous composite materials. Finally, results are shown and discussed in Section 4, and conclusions are drawn in Section 5.
2. Experimental development

2.1. HiPerDiF method for intraply aligned short fibre hybrid composites

The fibre orientation mechanism for the HiPerDiF method is described in previous work by Yu et al. [5, 9]. In summary, a suspension of fibres (dispersed in water) is supplied by nozzles to the fibre orientation head, which is composed of two thin-parallel plates that then align the fibres by a sudden momentum change, provided that the fibre length is less than the gap distance between the two parallel plates (see Figure 1).

A lab-scale prototype machine is capable of aligning discontinuous fibres, drying the remaining water on the fibres, and delivering the tape-type plies to the resin film impregnation stage, all in an in-line process. The desired ply width is achieved by placing a number of fibre orientation units next to each other (see Figure 1b), which results in a ply up to 5 mm wide. This process enables a high throughput, making it a strong candidate for industrialisation.

More importantly for this paper, the HiPerDiF method also allows for the manufacture of various types of hybrid discontinuous fibre arrangements. For instance, intermingled hybrid fibre arrangements can be easily manufactured if two or more types of fibres are dispersed in the water suspension [9]. Since the HiPerDiF method uses numerous nozzles, each one corresponding to a fibre orientation unit, intraply hybrids can also be generated, which feature a much lower level of intermingling, as shown in Figure 1. The HiPerDiF method was deemed the most suitable for this application as other manufacturing methods have limitations in controlling group (or bundle) size [15] and fibre dispersion [16] and are incapable of incorporating discontinuous reinforcements.

![Diagram](image1.png)

(a) Intraply hybrids are manufactured using separate nozzles and orientation heads, each supplying a suspension containing a single fibre type.

(b) The separate sub-regions combined to form a ply, which contains the different sub-regions across its width.

Figure 1: The HiPerDiF process, here shown in the modified version to produce intraply hybrid discontinuous composites with distinct carbon or glass subregions.
2.2. Materials and manufacturing condition

High modulus carbon (HMC, subscript ‘c’, represented in light grey in all figures) and E-glass (EG, subscript ‘g’, represented in dark grey in all figures) fibres were used in this paper; the properties of both fibre types are listed in Table 1. This fibre combination was selected because pitch-based high-modulus carbon fibres have a much lower failure strain than E-glass fibres, which promotes fibre fragmentation and increases the pseudo-ductility of the hybrid composite [9]. On the contrary, PAN-based discontinuous fibres feature a failure strain that is too similar to that of the E-glass fibres, and hence would not enable a pseudo-ductile response [9]. The HiPerDiF method enabled the manufacture of the intermingled and intraply hybrid intended cross sections shown in Figures 2a to 2e, with different intended HMC fibre group sizes and distributions, while keeping the intended carbon volume ratio in the composites at $v_c = 0.33$ (the carbon ratio ($v_c$) is defined such that $v_c + v_g = 1$).

Five intended cross-section fibre arrangements were manufactured; these are shown (not to scale) in Figures 2a to 2e. Each intended hybrid specimen is composed of four hybrid plies laid-up in the through-thickness direction, z. Each hybrid ply consists of a dry hybrid preform impregnated with an epoxy resin film (Cytec MTM49-3), the properties for which are shown in Table 1. The specimens are then laid-up in a mould, placed in a vacuum bag and then cured in an autoclave at 135°C for 135 minutes at 6 bar pressure.

Each intraply hybrid ply was manufactured using nine nozzles in parallel, with each nozzle connected individually to the HMC or EG suspension tanks, as shown in Figure 1a. Five unique intraply hybrid plies were manufactured, as shown in Figure 3a. A single stripe of a specific fibre type, which originates from a single nozzle (see Figure 3b), is hereby designated a sub-region (subscript ‘r’). The plate spacing defines the width of each sub-region ($w_r$, set as 0.5 ±0.01 mm for 3 mm long fibres), while the areal density of each sub-region is defined by [17]:

$$\rho_{A_r} = \frac{q_r \cdot b_{f_r} \cdot \rho_{f_r} \cdot w_r \cdot V}{w_r \cdot V},$$

where $q_r$ is the flow rate per nozzle, $b_{f_r}$ is the fibre concentration in the suspension (water), $\rho_{f_r}$ is the fibre density, and $V$ is the velocity of the conveyor belt.

The machine parameters were selected, as described in a previous paper [9], to obtain HMC and EG sub-regions with an areal density of 58 and 68 g/m² respectively. This allows the same overall fibre volume ($v_f = 0.3$) in each of the HMC and EG intraply hybrid composite cross sections. Figure 3c shows an example of intraply hybrid plies.
Table 1: Material properties used for this paper. It should be noted that the fibre volume fraction \( (v_f) \) is set to 0.30 throughout this paper. Values with a † were estimated from work by Pardini and Manhani [18], while values with a ‡ were calculated using the method described by Henry and Pimenta [19].

<table>
<thead>
<tr>
<th>Fibre-specific properties</th>
<th>High modulus carbon fibre</th>
<th>E-glass fibre</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fibre name</td>
<td>NGF XN90</td>
<td>Vectrotex C100</td>
</tr>
<tr>
<td>No. fibres in cross-section, ( n_f )</td>
<td>1348</td>
<td>5474</td>
</tr>
<tr>
<td>Diameter, ( \phi ) (( \mu )m)</td>
<td>10 [20]</td>
<td>7 [9]</td>
</tr>
<tr>
<td>Length, ( l ) (mm)</td>
<td>3 [9]</td>
<td>3 [9]</td>
</tr>
<tr>
<td>Young’s modulus, ( E ) (GPa)</td>
<td>876 [20]</td>
<td>73 [9]</td>
</tr>
<tr>
<td>Weibull reference strength, ( X_f ) (MPa)</td>
<td>3460 [20]</td>
<td>2400 [9]</td>
</tr>
<tr>
<td>Weibull modulus, ( m )</td>
<td>5 [20]</td>
<td>5.65†</td>
</tr>
<tr>
<td>Weibull gauge length, ( l_w ) (mm)</td>
<td>1 [20]</td>
<td>25 [9]</td>
</tr>
<tr>
<td>Composite fracture toughness, ( G_I ) (kN/m²)</td>
<td>3.50†</td>
<td>4.37‡</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Matrix / interface properties</th>
<th>MTM 49-3 epoxy</th>
</tr>
</thead>
<tbody>
<tr>
<td>Shear modulus, ( G_m ) (GPa)</td>
<td>1.5 [10]</td>
</tr>
<tr>
<td>Interlaminar shear strength, ( X_m ) (MPa)</td>
<td>80 [21]</td>
</tr>
<tr>
<td>Mode-II fracture toughness, ( G_{II_m} ) (kJ/m²)</td>
<td>0.8 [22]</td>
</tr>
<tr>
<td>Pull-out frictional stress (SFPO), ( \tau_0 ) (MPa)</td>
<td>10 [23, 24]</td>
</tr>
</tbody>
</table>

Figure 2: Different cross-section fibre arrangements investigated as part of this study (not to scale). Throughout this paper, glass fibres are shown in dark grey, while carbon fibres are shown in light grey (to match the fibres’ appearance in micrographs).
(a) Five distinct hybrid plies are made, which are combined to make the cross-sections shown in Figures 2b to 2e.

(b) In this paper, nine nozzles are used to create a ply with nine distinct sub-regions.

(c) Examples of several striped intraply hybrid plies.

Figure 3: The intraply HiPerDiF process shown in more detail for the striped fibre arrangement.

made using the HiPerDiF method.

2.3. Experimental testing

A schematic of the specimen used for tensile testing is shown in Figure 4. GFRP end-tabs were bonded with Huntsman Araldite 2014-1. Tensile tests were carried out using a servo-electric testing machine (Shimadzu, 10 kN, Japan) with a cross-head displacement speed of 1 mm/min. A white speckle pattern over a black background was painted on the specimens to allow strain measurement with a video extensiometer (IMETRUM, UK).

Figure 4: Specimen dimensions for experimental testing and analytical modelling. All dimensions are in mm.

Specimen cross-sections in epoxy pots were polished and prepared for microscopy analysis (Carl Zeiss, BG).

3. Model development

Stress vs. strain curves for the specimens described in Section 2 were modelled using a virtual testing framework with the fibre and matrix properties shown in Table 1. The virtual testing framework used in this paper was based on that recently proposed to model
*intermingled* hybrid discontinuous composites [10], but with two main modifications to account for the *intraply* (rather than *intermingled*) fibre arrangement, as described in Sections 3.1 and 3.2.

### 3.1. Model for realistic intraply fibre arrangements

The previously proposed *intermingled* model [10] arranged the fibres in a square cross-section, assuming a square fibre packing, and imposed a random arrangement of fibre types. The previous *intermingled* model was modified for this paper to enable a rectangular cross-section, of width $w$ and thickness $t$; the rectangular cross-section changed the dimensions of the fibre type matrix ($T$) from an $n$ by $n$ matrix to a $n_{f_i}$ by $n_{f_j}$ matrix, where $n_{f_i}$ and $n_{f_j}$ are the number of fibres across the thickness and width of the specimen:

$$
n_{f_i} = \text{round} \left( \frac{t \sqrt{v_f}}{\sqrt{\pi} \left( \frac{v_c \phi_c + (1-v_c) \phi_g}{2} \right)} \right), \quad n_{f_j} = \text{round} \left( \frac{w \sqrt{v_f}}{\sqrt{\pi} \left( \frac{v_c \phi_c + (1-v_c) \phi_g}{2} \right)} \right). \quad (2)$$

The newly developed virtual testing framework was used to investigate the *intermingled*, chequered, striped, double-blocked, and centre-blocked fibre arrangements, which are shown in Figure 2. The intended *intraply hybrid* fibre arrangements were created in the virtual testing framework in a similar way to the experiments, by dividing the matrix of fibre types ($T$) into $n_{r_i} = 4$ by $n_{r_j} = 9$ sub-regions, each with their own fibre type, as shown in Figure 5. Because different fibre types have different diameters, the number of fibres within each sub-region was selected to preserve the size ($w_r$ and $t_r$) and fibre volume fraction ($v_f$) of each sub-region (as shown in Figure 5). The intended fibre arrangement was then stored in the matrix of fibre types ($T$), where $T[i,j] = 0$ when fibre $(i,j)$ is a glass fibre, and $T[i,j] = 1$ when fibre $(i,j)$ is a carbon fibre.

It will be shown in Figures 8 to 12 that the intended fibre arrangements were significantly different from the fibre arrangements that were achieved in the experiments; this difference is because the fibres migrated from their intended position during the HiPerDiF manufacturing and alignment process. Consequently, once the intended fibre arrangement was stored in the matrix of fibre types ($T$), an empirical ‘fibre migration algorithm’ was created to capture the movement of fibres types during the HiPerDiF process. This algorithm works by swapping carbon and glass fibres within $T$, with the number of fibre swaps being governed by a ‘fuzzy factor’, $\psi$: a fuzzy factor of $\psi = 0$ implies no fibre migration, while a fuzzy factor of $\psi = 1$ implies that every carbon fibre has migrated some distance.
around the cross-section.

The fibre migration algorithm follows the flowchart shown in Figure 6. For each swap of a carbon and glass fibre, the target location for each carbon fibre is defined by stochastic vertical and horizontal migrations:

- The vertical migration, \( \Delta i \), is sampled from a normal distribution \( N(0, \sigma_v^2) \), where \( \sigma_v \) is a pre-defined vertical migration standard deviation;

- The horizontal migration, \( \Delta j \), is sampled from a normal distribution, \( N(\mu_h, \sigma_h^2) \), where \( \sigma_h \) is a pre-defined horizontal migration standard deviation; \( \mu_h \) is a stochastic vector of mean horizontal migrations, which are defined independently for every couple of lines in the matrix \( T \), and which are sampled from a random distribution defined as \( U(-0.5m_h,+0.5m_h) \), where \( m_h \) is a pre-defined maximum horizontal migration interval.

This process is repeated until the correct level of fibre migration (defined by the migration “fuzzy factor” \( \psi \)) is achieved. Figure 7 shows the cumulative effects of the various features of the fibre migration algorithm.

The inputs for the fibre migration algorithm were adjusted empirically until the experimental fibre arrangements visually matched with the migrated fibre arrangements (seen by comparing Figures 8b to 12b with Figures 8c to 12c). The inputs for the fibre migration algorithm are normalised by the relative dimensions of the cross-section \( (n_f = 18 \text{ fibres} \) and \( n_j = 387 \text{ fibres} \)) and of the sub-regions \( (n_r = 4 \text{ sub-regions} \) and \( n_i = 9 \text{ sub-regions} \)), as shown in Table 2.

Table 2: Inputs for the fibre migration algorithm for the four *intraply* hybrid cross-sections.

<table>
<thead>
<tr>
<th>Cross-section</th>
<th>( \sigma_v )</th>
<th>( m_h )</th>
<th>( \sigma_h )</th>
<th>( \psi )</th>
</tr>
</thead>
<tbody>
<tr>
<td>Chequered</td>
<td>( n_f / (4n_i) = 1.125 )</td>
<td>2( n_f / n_i ) = 86</td>
<td>3( n_f / (4n_i) ) = 32.25</td>
<td>0.96</td>
</tr>
<tr>
<td>Striped</td>
<td>( n_f / (4n_i) = 1.125 )</td>
<td>2( n_f / n_i ) = 86</td>
<td>3( n_f / (4n_i) ) = 32.25</td>
<td>0.96</td>
</tr>
<tr>
<td>Double-blocked</td>
<td>( n_f / (4n_i) = 1.125 )</td>
<td>2( n_f / n_i ) = 86</td>
<td>3( n_f / (4n_i) ) = 32.25</td>
<td>0.80</td>
</tr>
<tr>
<td>Centre-blocked</td>
<td>( n_f / (4n_i) = 1.125 )</td>
<td>2( n_f / n_i ) = 86</td>
<td>3( n_f / (4n_i) ) = 32.25</td>
<td>0.80</td>
</tr>
</tbody>
</table>

3.2. Fracture-based failure criterion for intermingled and *intraply* hybrid composites

The average stress-strain curve of hybrid discontinuous composites can be predicted using a previously-developed virtual testing framework [10]. This framework decomposes a complete composite specimen into Representative Volume Elements (RVEs), each with the
Figure 5: Definition of the sub-regions (intended to have a single fibre type) within a virtual specimen cross-section. Different fibres have a different diameter, therefore the sub-regions are all the same physical size ($t_r = 0.06$ mm and $w_r = 0.56$ mm) and have the same overall fibre volume fraction ($v_f = 0.30$), but the number of fibres within the sub-region varies by fibre type.

The virtual testing framework for *intermingled* discontinuous composites [10, 19] proposed that the full stress-strain curve of a hybrid discontinuous composite specimen can be calculated by combining the stress-strain curves of its individual RVEs in series. Subsequently, a non-linear fracture mechanics criterion is used to determine the ultimate strength and failure strain of the specimen; this criterion aims to model unstable fracture of an entire cross-section, triggered by the formation of a broken-fibre cluster of a critical size. To do so, the criterion first identifies the weakest square-shaped cluster of a given size within the specimen; fracture will occur if the strain energy release rate of the corresponding penny-shaped crack is greater than the average fracture toughness of the composite; if not, then the size of the square cluster is increased, until fracture is predicted.

Henry and Pimenta’s fracture-based failure criterion mentioned above [10, 19] works
I. Definition of input variables and initialisation

I.1: Cross-section variables
\[ n_{fi}, n_{fj}, v_c, n_c, T \]

I.2: Migration input variables
\[ \psi, m_h, \sigma_v, \sigma_h \]

I.3: Initialise matrix of changes, \( D \), and mean horizontal migration, \( \mu_h \)
\[
D = \text{zeros} \left(n_{fi}, n_{fj}\right)
\]
for \( i = \left\{ 1, \frac{n_{fi}}{2} \right\} \), \( \mu_h^{[2i-1,2i]} = \text{round} \left( \text{U}(-0.5,0.5) m_h \right) \)

II. Change a carbon fibre to a glass fibre

II.1: Select random fibre location
\[ i = \left[ \text{U}(0,n_{fi}) \right], \quad j = \left[ \text{U}(0,n_{fj}) \right] \]

II.2: Check whether the fibre at position \( i, j \) is of carbon type and has not been migrated before. If so, change the carbon fibre to a glass
\[
T_{i,j} = 0, \quad D_{i, j} = 1
\]

III. Find a nearby glass and change it to a carbon fibre

III.1: Find vertical migration
\[
\Delta_i = \mathcal{N} \left(0, \sigma_v^2\right)
\]

III.2: Find horizontal migration
\[
\Delta_{j \mid i} = \mathcal{N} \left( \mu_h^{[i]}, \sigma_h^2 \right)
\]

III.3: Change the glass to a carbon if it has not been changed before
\[
T_{i + \Delta_i \mid j + \Delta_{j \mid i}} = 0 \quad \text{and is} \quad D_{i + \Delta_i \mid j + \Delta_{j \mid i}} = 0 ?
\]
\[
T_{i + \Delta_i \mid j + \Delta_{j \mid i}} = 1, \quad D_{i + \Delta_i \mid j + \Delta_{j \mid i}} = 1
\]

IV. Repeat the process until the required number of fibres have migrated

\[
\sum_{i=1}^{n_{fi}} \sum_{j=1}^{n_{fj}} D_{i, j} = 2 \cdot n_c \cdot \psi \cdot \left(1 - v_c \right) ?
\]

Finish: matrix of fibre types \( T \) updated

Figure 6: Flow chart of the fibre migration algorithm.
(a) Intended *double-blocked* cross-section; $\sigma_v = 0$, $m_h = 0$, and $\sigma_h = 0$.

(b) *Double-blocked* cross-section with only a vertical migration applied. The vertical migration introduces vertical spread in the fibre location; $\sigma_v = 1.125$, $m_h = 0$, and $\sigma_h = 0$.

(c) *Double-blocked* cross-section with a vertical migration, and a row-based horizontal migration applied. The row-based horizontal migration introduces the layered effect in the cross-section; $\sigma_v = 1.125$, $m_h = 86$, and $\sigma_h = 0$.

(d) *Double-blocked* cross-section with a vertical migration, a row-based horizontal migration, and a fibre-based horizontal migration applied. The fibre-based migration introduces horizontal spread into the cross-section; $\sigma_v = 1.125$, $m_h = 86$, and $\sigma_h = 32.25$.

Figure 7: *Double-blocked* cross-sections showing the cumulative effects of each aspect of the fibre migration algorithm.

(a) Intended *intermingled* cross-section.

(b) Experimental *intermingled* cross-section.

(c) *Intermingled* cross-section (note that for the *intermingled* case only, zero migration occurs as the cross-section is already intermingled). The critical cluster is also highlighted.

Figure 8: The intended *intermingled* cross-section, the experimental cross-section after manufacture, and the cross-section after the fibre migration model was applied. Small critical clusters are formed before fracture of the *intermingled* specimen (Section 4.1).
(a) Intended chequered cross-section.

(b) Experimental chequered cross-section.

(c) Chequered cross-section with fibre migration model applied. The critical cluster is also highlighted.

Figure 9: The intended chequered cross-section, the experimental cross-section after manufacture, and the cross-section after the fibre migration model was applied. Medium-sized critical clusters are formed before fracture of the chequered specimen (Section 4.1).

(a) Intended striped cross-section.

(b) Experimental striped cross-section.

(c) Striped cross-section with fibre migration model applied. The critical cluster is also highlighted.

Figure 10: The intended striped cross-section, the experimental cross-section after manufacture, and the cross-section after the fibre migration model was applied. Medium-sized critical clusters are formed before fracture of the striped specimen (Section 4.1).
(a) Intended \textit{double-blocked} cross-section.

(b) Experimental \textit{double-blocked} cross-section.

(c) \textit{Double-blocked} cross-section with fibre migration model applied. The critical cluster is also highlighted.

Figure 11: The intended \textit{double-blocked} cross-section, the experimental cross-section after manufacture, and the cross-section after the fibre migration model was applied. Medium critical clusters are formed before fracture of the \textit{double-blocked} specimen (Section 4.1).

(a) Intended \textit{centre-blocked} cross-section.

(b) Experimental \textit{centre-blocked} cross-section.

(c) \textit{Centre-blocked} cross-section with fibre migration model applied. The critical cluster is also highlighted.

Figure 12: The intended \textit{centre-blocked} cross-section, the experimental cross-section after manufacture, and the cross-section after the fibre migration model was applied. Large critical clusters are formed before fracture of the \textit{centre-blocked} specimen (Section 4.1).
well for non-hybrid discontinuous composites, and for hybrid discontinuous composites with *intermingled* cross-sections. Although it is recognised that a square cluster may contain unbroken fibres or may neglect adjacent broken fibres, the effects of these two assumptions will effectively cancel one-another out for an *intermingled* cross-section, due to the random fibre arrangement; however, these effects will be much more significant for *intraply* hybrids, as the cross-section is more ordered. In an *intraply* hybrid, the propagation of failure from a given cluster will also be associated with a fracture toughness that is dependent on the properties of the fibres surrounding that broken cluster, something that is also neglected in Henry and Pimenta’s method for *intermingled* hybrids [10, 19].

We thus propose an improved fracture-based failure criterion (see the flowchart in Figure 13), which accurately calculates the size, shape, position, and fracture toughness associated with clusters of broken fibres; this fracture criterion is therefore suited for both *intermingled* and *intraply* hybrid discontinuous composites. In order to determine the fracture-based failure strain, the following steps are carried out for every RVE:

- Inputs for the fracture criterion are first gathered; this includes the matrix of fibre types ($T$, from Section 3.1), full RVE stress-strain curve ($\sigma_{\text{RVE}}$ vs. $\varepsilon_{\text{RVE}}$), matrix of fibre stresses ($\sigma_f$), RVE initial stiffness ($E_{\text{RVE}}$), and the ultimate strength of the RVE ($X_{\text{RVE}}$, from Henry and Pimenta [10], without considering the possibility of unstable fracture).

- At each strain increment ($\varepsilon_{\text{RVE}}$), all fully-failed fibres (which carry zero stress due to either fibre fragmentation or matrix debonding) are identified and stored in the matrix of failed fibres $B$. Clusters of fibre breaks are identified using the ‘bwlabel’ function in MATLAB (see Figure 14 for details).

- The fracture toughness of the material surrounding each cluster is determined (see Appendix A). For each of the $n_k$ clusters of broken fibres, the equivalent penny-shape crack radius is determined, and the associated strain energy release rate is calculated. The fracture stress of the RVE is then determined when the maximum strain energy release rate $J$ (associated with fracture propagating from the most critical cluster) exceeds the fracture toughness $G_{\text{f}}$ of the material surrounding the critical cluster.

- This process is repeated for all of the RVEs and the minimum fracture strength of all the RVEs is determined.
Once this process is repeated for every RVE, the specimen stress-strain curve is formed from the average of all of the RVE stress-strain curves. The specimen stress-strain curve is then trimmed up to the minimum fracture strength.
Figure 13: Flow chart to demonstrate the implementation of the fracture-based failure criterion. The ‘bwlabel’ function is shown in further detail in Figure 14, along with the ‘==’ symbol, which denotes the logical operation ‘if equal to’.
II. Find individual clusters of broken fibres for a given strain increment, $\varepsilon$

II.1: Find which fibres are broken.

$$\mathbf{B}^{[\varepsilon]} = (\sigma_f^{[\varepsilon]} == 0) = \begin{bmatrix} 1 & 1 & 0 & 0 & 0 & 1 & 1 & 1 & 1 \\ 0 & 0 & 0 & 1 & 1 & 0 & 0 & 1 & 0 \\ 1 & 1 & 0 & 1 & 1 & 0 & 0 & 0 & 0 \\ 1 & 1 & 0 & 1 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 1 & 0 & 1 \end{bmatrix}$$

II.2: Apply the 'bwlabel' function to identify clusters of broken fibres (cluster co-ordination number = 4)

$$\mathbf{C}^{[\varepsilon]} = \text{bwlabel}(\mathbf{B}^{[\varepsilon]}) = \begin{bmatrix} 1 & 1 & 0 & 0 & 0 & 4 & 4 & 4 & 4 \\ 0 & 0 & 0 & 3 & 3 & 0 & 0 & 4 & 0 \\ 2 & 0 & 0 & 3 & 3 & 0 & 0 & 0 & 0 \\ 2 & 2 & 0 & 3 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 5 & 0 & 0 & 6 \end{bmatrix}, \ n_k = 6$$

Figure 14: Flow chart which shows an example of the “bwlabel” function in practice. The “bwlabel” function finds failed fibres that also have neighbouring failed fibres to the top, bottom, left, or right of themselves, and numbers each cluster with a unique identifier. The “bwlabel” function is used in MATLAB code and can be found in the documentation [25].

4. Results and discussion

4.1. Model validation

The stress-strain curves for the experiments and virtual testing framework predictions are in good agreement, as can be seen in Figure 15. Due to the stochastic assignment of fibre types, fibre ends, and fibre strengths, outputs from the virtual testing framework are stochastic; consequently, three model runs are shown for each fibre arrangement (the same as the number of experimental runs), with the point of fracture for each model run marked with a red cross. Although the model was only run for a small number of times, the results indicate a similar level of variability between the virtual testing framework and the experiments.

The level of fibre alignment in the experimental specimens was not measured, although the HiPerDiF method has previously shown that very good levels of fibre alignment can be achieved [5], and the effects of fibre misalignment from the HiPerDiF method have been proven to be small [26]. This evidence, coupled with good correlation between the initial stiffness of the experimental and modelling curves, suggests that the level of fibre misalignment in the specimens was negligible, and had little or no effect on the structural performance of the manufactured specimens.

The virtual testing framework over-predicts the initial stiffness and pseudo-yield strength of the centre-blocked fibre arrangement; a significant load drop is also predicted that is not shown in the experiments (Figure 15e). The high initial stiffness indicates that there might
be a lower number of carbon fibres in the experimental tests than in the model, as the
carbon fibres are much stiffer than the glass fibres; a higher number of carbon fibres also
leads to a larger load drop, as more of the low elongation fibres will fragment at low strain.
The model is therefore re-run in Figure 15f) for a carbon ratio of $v_c = 0.28$ to reduce the
number of carbon fibres within the cross-section, which gives a much better agreement for
the pseudo-yield strength, initial stiffness, and load drop behaviour for both experiments
and modelling. Another possible reason for this lower initial stiffness in the *centre-blocked*
fibre arrangement may be due to poorer alignment of the carbon fibres when large groups
of these fibres are formed, as previously discussed by Yu et al. [9]; however, this is unlikely
to be the case, because a significant level of fibre misalignment would also lead to a strain
hardening initial response (due to the re-alignment of the fibres with increasing strain),
which was not seen in the experiments. Poor control of the carbon ratio or HMC fibre
alignment may explain why there is also a small mismatch between the stiffness and load-
drop of the experimental and modelling results for the *double-blocked* fibre arrangement
in Figure 15d.

Figures 8c to 12c and Figure 15 indicate that fibre arrangements with the coarsest
fibre-type grouping (i.e. the *double-blocked* and *centre-blocked* cross-sections) have a lower
strength and strain to failure. This is because larger clusters of broken fibres promote
overall fracture of the specimen at lower strains, because the cluster strain energy release
rate (Stage III.3 of Figure 13) increases significantly with cluster size. The strength and
failure strain of a hybrid discontinuous composite material could therefore be improved by
maximising the level of intermingling within the cross-section, as intermingling will reduce
the size of the low elongation fibre groups and, consequently, will delay the formation of
large broken clusters.

The modifications to the virtual testing framework proposed in this paper are essential
in order to accurately model *intraply* hybrid discontinuous composites. Without the ap-
lication of the fibre migration algorithm, the virtual testing framework predicts a brittle
stress-strain response for the *intraply centre-blocked* cross-section (Figure 16a), while in
reality a small amount of ductility was observed during the experiments. The proposed
fracture criterion is also necessary to predict an accurate failure strain of the *intraply* fibre
arrangements (Figure 16b).
(a) Stress-strain curve for the *intermingled* cross-section.

(b) Stress-strain curve for the *chequered* cross-section.

(c) Stress-strain curve for the *striped* cross-section.

(d) Stress-strain curve for the *double-blocked* cross-section.

(e) Stress-strain curve for the *centre-blocked* cross-section.

(f) Stress-strain curve for the *centre-blocked* cross-section with the carbon ratio scaled to match experiments.

Figure 15: Stress-strain curves for the various hybrid cross-section arrangements.
Figure 16: Two sets of stress-strain curves demonstrating the importance of the proposed fibre migration algorithm and the proposed fracture-based failure criterion for the centre-blocked cross-section.

4.2. Effect of intermingling

Increasing levels of random intermingling (i.e. randomly swapping carbon and glass fibres within the cross-section) were applied to the centre-blocked cross-section, which resulted in smaller groups of low elongation fibres (see Figure 17). Smaller groups of low elongation fibres caused smaller clusters of fibre breaks to form, because the fragmented low elongation fibres remained more isolated from one-another. Increasing intermingling is therefore an effective means to increase the strength and pseudo-ductility of hybrid discontinuous composites (see Figure 18), as the critical cluster size and resulting strain energy release rate is kept small.

Figure 17: The centre-blocked microstructure with increasing levels of random intermingling. Increasing levels of intermingling reduces the size of the critical cluster (shown in red) before fracture. The size of the critical cluster can be minimised by isolating low elongation fibres from one-another.
Figure 18: Effect of increasing intermingling on the stress-strain curve of intraply hybrid discontinuous composites with a centre-blocked microstructure. Increasing intermingling results in an increase in pseudo-yield strength, ultimate strength, and pseudo-ductile strain, but a reduction in initial stiffness.

It should be noted that the centre-blocked cross-section with the least intermingling ($\psi = 0$) showed the highest initial modulus (as also shown in Figures 15a to 15f and Figure 18). Henry and Pimenta [26] showed that the maximum modulus in a hybrid discontinuous composite is achieved when the stiffnesses of the interacting (i.e. neighbouring) fibres are similar to one-another. The modulus of the whole specimen is therefore maximised when the fibre types are arranged such that those interactions mainly consist of the same fibre type (i.e. when intermingling is minimised). Intermingling therefore presents a clear trade-off between (i) strength and pseudo-ductility or (ii) stiffness, although the positive effect of intermingling on the strength and ductility appears to be much more significant than the negative effect on stiffness (for the fibre diameters and stiffnesses considered here).

4.3. Quantifying the effect of fibre-type grouping using controlled fibre arrangements

Several fibre arrangements with controlled levels of intermingling (shown in Figure 19) were investigated using the virtual testing framework, in order to quantify how grouping low elongation fibres influences the structural performance of hybrid discontinuous composites. The isolated fibre cross-section features single carbon fibres that are surrounded entirely by glass fibres, in order to minimise the size of clusters of broken fibres and delay fracture for as long as possible. The 4-block, 9-block, 16-block, and 25-block cross-sections feature increasingly large groups of low-elongation carbon fibres, so that the influence of fibre grouping could be quantified. Finally, the thin-ply fibre arrangement features layers of single carbon fibres across the entire width of the cross-section; this cross-section was
suggested by Swolfs et al. [14] to maximise the hybrid effect for the failure strain of the low elongation fibres. It should be noted that carbon fibres were added randomly to the *thin- ply* cross-section in order to maintain a consistent carbon ratio ($v_c = 0.33$) with the other fibre arrangements.

![Image](a) The *isolated fibre* fibre arrangement.

![Image](b) The *4-block* fibre arrangement.

![Image](c) The *9-block* fibre arrangement.

![Image](d) The *16-block* fibre arrangement.

![Image](e) The *25-block* fibre arrangement.

![Image](f) The *thin- ply* fibre arrangement. Please note that carbon fibres were added randomly to the *thin- ply* cross-section in order to maintain a consistent carbon ratio ($v_c = 0.33$) with the other fibre arrangements.

Figure 19: Highly controlled fibre arrangements for detailed analysis of the effects of fibre-type grouping.

Pseudo-ductile strains were calculated for all fibre arrangements using the method proposed by Wisnom et al. [27], while pseudo-yield strengths were calculated using a 0.1% strain offset. Using these methods the *isolated fibre* arrangement demonstrates the optimal structural performance, with a 27% increase in ultimate strength, a 21% increase in pseudo-yield strength, and a 44% increase in pseudo-ductile strain when compared to the *intermingled* fibre arrangement (see Figure 20). The critical cluster in the *isolated*
Figure 20: A comparison between the stress-strain curves of the intermingled and highly controlled fibre-arrangements. If the low elongation fibres are isolated from one another (i.e. if the isolated fibre arrangement is used), the pseudo-yield strength, ultimate strength, and pseudo-ductile strain are maximised, with a minor reduction in stiffness.

The initial modulus increases for the more highly-grouped fibre arrangements (Figure 20), with the 25-block fibre arrangement showing the largest modulus increase of 18% relative to the isolated fibre arrangement; this modulus increase is in line with observations in Section 4.2. However, the more highly-grouped fibre arrangements also show significantly lower ultimate strength, pseudo-ductility, and pseudo-yield strength as the size of the fibre groups increases; the most significant case is the 25-block fibre arrangement, which shows an almost brittle response, a 44% reduction in ultimate strength, and a 96% reduction in pseudo-ductile strain, relative to isolated fibre arrangement. Larger groups of low elongation fibres enable larger clusters of broken fibres to form (as shown in Figures 19b to 19d), hence fracture occurs at increasingly lower strains, as mentioned in Section 4.1.

The intermingled, 4-block, and thin-ply arrangements all show similar critical cluster sizes, ultimate strengths, and failure strains, however the 4-block, and thin-ply arrangements feature a significant load-drop after yielding (Figure 20). When a fibre breaks, the stress from the broken fibre is passed to the neighbouring fibres; therefore, failure of the
Low elongation fibres generate significantly higher stress concentrations in neighbouring low elongation fibres than in neighbouring high elongation fibres, due to the larger diameter and modulus of the low elongation fibres. These higher stress concentrations lead to further failures of low elongation failures propagating through the whole specimen, which leads to a significant global load drop. For the fibre types considered in this study, intermingling is an effective way to prevent the propagation of fibre breaks, as it results in more of the fibre breaks being shielded by high elongation fibres; these high-elongation fibres are not only less susceptible to breaking at low applied strains, but they also experience a lower stress concentration and hence are less likely to fail in the vicinity of a fibre break. It should be noted that this argument does not hold when the hybrid reinforcements have equal diameters; in this case the thin ply has a higher ultimate strength pseudo-ductile strain than the intermingled case, in line with predictions made by Swolfs et al. [14], however the isolated fibre arrangement remains the strongest and most ductile.

5. Conclusion

This paper combines experiments and a virtual testing framework to explore how grouping low elongation fibres affects the pseudo-ductility of aligned hybrid discontinuous composites. The main conclusions are shown below:

- Both intermingled and intraply aligned hybrid discontinuous composite materials were successfully manufactured using an improved High Performance Discontinuous Fibre (HiPerDiF) method.

- Significant migration of the fibres was present during the intraply HiPerDiF process. A new fibre migration model was therefore developed to simulate fibre migration in the virtual testing framework; the migration algorithm was proven to be essential in order for the virtual testing framework to capture the correct stress-strain response of the intraply hybrid discontinuous composites.

- A new fracture criterion was also developed for the virtual testing framework, in order to predict final failure of both intermingled and intraply hybrids. The proposed fracture criterion proved to be necessary for an accurate prediction of the failure strain of intraply aligned hybrid discontinuous composites.

- Both the experiments and the new virtual testing framework indicated that the intermingled fibre arrangement had the highest strength and pseudo-ductility out
of the arrangements tested in the experiments; the *intermingled* fibre arrangement was successful because it featured small groups of low elongation fibres, which delayed fracture by preventing large clusters of broken fibres from forming. These results suggest that the HiPerDiF process is the current state-of-the-art manufacturing method for maximising pseudo-ductility in hybrid discontinuous composites, because the HiPerDiF process produces the most dispersed fibre-type intermingling of all manufacturing processes for hybrid composites in the literature so far [11, 16].

- A further theoretical study using the virtual testing framework showed that a cross-section with full-scale isolation of low elongation fibres could present a further improvement on the strength and ductility of the *intermingled* fibre arrangement. A 27% increase in ultimate strength, a 21% increase in pseudo-yield strength, and a 44% increase in pseudo-ductile strain were achieved for the *isolated fibre* arrangement when compared to the *intermingled* arrangement.

The qualitative conclusions drawn in this paper are valid for any hybrid discontinuous material system where the failure strains of the two fibres are dissimilar and the ratio between the two fibre types is adequately selected. High modulus / high strength carbon hybrid discontinuous composites would qualitatively agree with the conclusions drawn here, although the virtual testing framework would have to be re-used with the correct inputs for those two fibre-types, in order to quantify any changes in overall strength and ductility of the materials. The virtual testing framework developed in this paper is capable of modelling a large variety of material systems; this framework will therefore be used to further investigate different hybrid (or non-hybrid) fibre types, fibre volume fractions, matrix behaviours, and fibre arrangements to find an optimal material for strength and pseudo-ductility. One example of this Future work will also look at optimising the strength and ductility of aligned hybrid discontinuous composites by implementing an *isolated fibre* arrangement using the HiPerDiF method and by using higher fibre volume fractions.

6. Acknowledgements

This work was funded under the UK Engineering and Physical Sciences Research Council (EPSRC) programme grant EP/I02946X/1 on *High Performance Ductile Composite Technology*. S. Pimenta acknowledges the support from the Royal Academy of Engineering for her Research Fellowship on *Multiscale discontinuous composites for large scale and...*

All underlying data to support the conclusions are provided within this paper.

7. References


**Appendices**

**Appendix A  Calculation of the fracture toughness of fibres surrounding a cluster**

The fracture toughness associated with a cluster of broken fibres within a composite material is dependent on the fracture toughness of the material surrounding the broken
cluster. For discontinuous composites [19] and intermingled hybrid discontinuous composites [10], the fracture toughness can be approximated as a constant value throughout the cross-section, because the arrangement of fibre types is homogeneous (or quasi-homogeneous).

However, this approximation of an average fracture toughness does not hold when applied to intraply hybrid discontinuous composites, as the fibre arrangement is no longer homogeneous. It is thus necessary to evaluate the fracture toughness associated with each cluster as the toughness of the material surrounding each individual cluster.

Figure 21 shows a detailed flowchart of how to evaluate the toughness of the surrounding material of a specific cluster. The process starts by identifying the $k$th cluster of broken fibres within the cross section, whose location is stored in a matrix $K$. The matrix $K$ is then translated around itself to the left, right, top, and bottom to identify all of the fibres surrounding the cluster (in this case a co-ordination number of four is used); the locations of the surrounding fibres are then stored in the matrix $S$. The matrix of fibre types $T$ and the matrix of surrounding fibres $S$ are used to determine how many fibres of each fibre type surround the broken cluster; the translaminar fracture toughness of a single glass fibre composite and a single carbon fibre composite are calculated using the method described by Henry and Pimenta [19] (these values are shown in Table 1). Finally, the fracture toughness of the surrounding material $G_I$ is calculated as the weighted average of each individual fibre type fracture toughness, based on the relative area occupied by each fibre type.
III.2 Calculate the fracture toughness of fibres bordering each cluster

III.2A: Select cluster $k$ in the cross-section and store in matrix $K^{[k,\varepsilon]}$

$$K^{[k,\varepsilon]} = \{C^{[\varepsilon]} = k\} = \begin{bmatrix} 0 & 0 & 0 & 0 & 0 & 1 & 1 & 1 & 1 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 1 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \end{bmatrix}$$

III.2B: Create a matrix $\bar{R}$, which will be used to translate $K^{[k,\varepsilon]}$

$$\bar{R} = \begin{bmatrix} 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 1 & 1 & 1 & 1 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 1 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \end{bmatrix}$$

III.2C: Translate $K^{[k,\varepsilon]}$ around $\bar{R}$ and sum the elements. A co-ordination number of 4 is used

$$K^{[k,\varepsilon]}_{[i=1:end-2;j=2:end-1]} = K^{[k,\varepsilon]}_{[i=1:end-2;j=2:end-1]} + K^{[k,\varepsilon]}_{[i=3:end];[j=2:end-1]} + K^{[k,\varepsilon]}_{[i=1:end];[j=2:end-1]} + K^{[k,\varepsilon]}_{[i=1:end-1];[j=3:end]}$$

$$\bar{R} = \begin{bmatrix} 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 1 & 2 & 3 & 4 \\ 0 & 0 & 0 & 0 & 0 & 1 & 2 & 2 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 1 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \end{bmatrix}$$

III.2D: Trim the contents of $\bar{R}$ to match the dimensions of the cross-section and set any positive values to 1

$$\bar{R} = (R_{[i=2:end-1;j=2:end-1] > 0}^T) = \begin{bmatrix} 0 & 0 & 0 & 0 & 1 & 1 & 1 & 1 \\ 0 & 0 & 0 & 0 & 0 & 1 & 1 & 1 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 1 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \end{bmatrix}$$

III.2E: Subtract $K$ from $\bar{R}$ to give the matrix of fibres surrounding the cluster, $S^{[k,\varepsilon]}$

$$S^{[k,\varepsilon]} = (\bar{R} - K^{[k,\varepsilon]}) = \begin{bmatrix} 0 & 0 & 0 & 0 & 1 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 1 & 1 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \\ 0 & 0 & 0 & 0 & 0 & 0 & 0 & 0 \end{bmatrix}$$

III.2F: Find the number of carbon fibres, $n_{bc}^{[k,\varepsilon]}$, and glass fibres, $n_{bg}^{[k,\varepsilon]}$, that surround the cluster

$$n_{bc}^{[k,\varepsilon]} = \sum_{j=1}^{n_{bc}} \sum_{i=1}^{n_{bc}} (S^{[k,\varepsilon]} = 1) \cdot \mathbf{T}$$

$$n_{bg}^{[k,\varepsilon]} = \sum_{j=1}^{n_{bg}} \sum_{i=1}^{n_{bg}} (S^{[k,\varepsilon]} = 1) \cdot (1 - \mathbf{T})$$

III.2G: Calculate the fracture toughness of the surrounding fibres

$$g^{[k,\varepsilon]} = \frac{\sigma_{bc}^{[k,\varepsilon]} \cdot n_{bc}^{[k,\varepsilon]} + G_{bg}^{[k,\varepsilon]} \cdot n_{bg}^{[k,\varepsilon]} + \phi_g}{n_{bc}^{[k,\varepsilon]} \cdot \phi_g^{[k,\varepsilon]} + n_{bg}^{[k,\varepsilon]} \cdot \phi_g}$$

Figure 21: Flow chart to demonstrate how the fracture toughness of the material surrounding a cluster is calculated.