State-of-the-art models for mechanical performance of carbon-glass hybrid composites in wind turbine blades

Yentl Swolfs, Babak Fazlali, Arsen Melnikov, Francisco Mesquita, Vincent Feyen, Christian Breite, Larissa Gorbatikh and Stepan V Lomov

Department of Materials Engineering, KU Leuven, Kasteelpark Arenberg 44 box 2450, 3001 Leuven, Belgium

yentl.swolfs@kuleuven.be

Abstract. Wind turbine blades are a key growth market for fibre-hybrid composites, as they offer the potential for higher turbine efficiency at a reduced cost. This paper therefore overviews the developments of KU Leuven models for mechanical properties relevant to wind turbine blades, including fibre-hybridisation, size scaling effects, flexure, transverse cracking, stress relaxation and fatigue. The strengths and limitations of the models will be highlighted with a particular focus on relevance for wind turbine blades. While significant steps forward have been made, the various developments still need to be incorporated in one all-encompassing model, and further work is needed on gathering the required input data and detailed experimental validation studies.

1. Introduction

Over the last decade the interest in fibre-hybrid composites from academia as well as industry continuously increased [1,2]. New applications are appearing in many different markets, such as automotive and sporting goods [1]. One of the major growth markets, however, is wind turbine blades. The blades are often not only the most expensive component of a turbine [3,4], but also the limiting factor for the lifetime of the turbine [3]. It is therefore important to thoroughly understand how damage evolves over their lifetime, and have simulation tools at hand to help designing the blades taking into account potential degradation.

Improving the stiffness of wind turbine blades is vital to increase the overall turbine efficiency. One of the key methods for improving efficiency is through the use of longer blades. This is particularly important for off-shore turbines [5,6]. Nearly all commercial wind turbine blades are made almost entirely of glass fibre composites [4], which are not viable for turbine blades longer than 80 m. One approach would be to switch completely to carbon fibre composites [7,8], which would lead to lighter blades with improved fatigue performance and lower blade tip deflection. The cost of such all-carbon fibre composite blades was, however, found to be prohibitive. In the meantime, the industry has settled on using carbon/glass hybrid composites as the best compromise [9]. The use of such hybrid composites is particularly helpful in the spar flanges, where the additional carbon fibres can have the largest positive impact. They also have the added benefit of enabling passive actuation, as their higher anisotropy allows the blades to twist away when the wind gets too strong [10,11].

Introducing fibre-hybrid composites into wind turbine blades creates additional complexity. Firstly, the failure mechanisms of fibre-hybrid composites are not as well understood as for non-hybrid composites [1,2]. This has also hampered the development of simulation tools that are industry-friendly, but still capture the relevant failure mechanisms. Secondly, the smaller diameter of carbon fibres may make impregnation more difficult. Thirdly, some non-destructive testing techniques may not work...
anymore. Ultrasonic waves may, for example, be reflected at the carbon/glass interface, which would hamper a reliable ultrasonic C-scan [12].

Delving deeper into the more complex failure behaviour, it should be noted that synergetic effects are commonly observed in fibre-hybrid composites [1,2,13]. These are often referred to as hybrid effects. Many of these synergetic effects only become significant when the fibres are finely dispersed [14-19]. Such fine dispersion is rarely achieved in the fibre-hybrid composites relevant to wind industry. This, however, depends on how narrowly the term hybrid effect is defined. In its broadest sense, a hybrid effect is a deviation from the relevant rule-of-mixtures [1,2,13]. The tensile stress-strain behaviour of a fibre-hybrid composite, for example, commonly deviates from the linear rule-of-mixtures even if the degree of dispersion is low [20,21].

The majority of the published fibre-hybrid research focused on the tensile response [17-24]. Here the development of failure is well understood, and synergetic effects can be predicted with reasonable accuracy [19,25,26]. However, the required predictive models tend to be time-consuming and hence their application is limited to small specimens sizes of a few thousands of fibres of a few millimetres in length [27]. Recently, model predictions have been validated in detail against synchrotron computed tomography experiments [27-29]. These experiments enable tracking of microscopic failure of individual fibres (fibre breaks) as well as the development of clusters. It was concluded that current models are not yet able to capture the real development of clusters. A major contributor to the mismatch is the difficulty in reliably measuring the model input parameters [30,31]. In addition, it also seems that there are still micromechanisms at work that even the most advanced models miss [32]. To increase the applicability of strength prediction models to the load cases in wind turbine blades, the following features require particular attention:

- **Size scaling effect**: It is unclear how well models capture size effects, and whether the computational limitations may be prohibitive to upscale current approaches.
- **Out-of-plane loading, such as flexure**: Most models assume uniaxial tension, and cannot capture the effect of bending loads.
- **Multidirectionality of laminates**: The models that are sufficiently advanced to capture fibre break development assume only load-carrying 0° oriented plies or bundles are present.
- **Time-dependent features**: Only a few models are able to capture time-dependent effects, such as stress relaxation/creep [33] and fatigue [34].

The size scaling effect is the decrease in strength with an increase in specimen size. Its existence is assumed for composites, but has not been thoroughly investigated for unidirectional tensile strength. Since wind turbine blades are much larger than standard coupons, a solid understanding of size scaling of unidirectional composites is vital to predict component failure. Okabe and Takeda [35] performed tensile tests over a wide range of volumes, reporting a reduction of about 10%. Unfortunately, their study misses key details on which dimensions were changed and whether the specimens failed in the gauge section or near the grips. Wisnom et al. [36] measured a reduction of 14% when the specimen size was scaled up in all three dimensions by a factor of 8. They reported gauge length failures for their smallest specimens, but admitted they could not deduce failure initiation for larger specimen sizes. This study deliberately scaled up all specimen dimensions, meaning that the length and cross-sectional scaling effects could not be distinguished. Given its practical importance, there have been several modelling studies to capture the size scaling effect [35,37-39]. Two of these studies did report a good match with the experimental data [35,39]. None of these models analysed how the predicted size scaling effect was affected by the boundary conditions.

Fibre break models typically assume a uniform strain or stress being applied over the entire volume. So far, published models have not attempted to predict the flexural response. This is likely due to the common assumption that the compressive strength is lower than the tensile strength [40]. While this seems true for composites with significant fibre misalignment or with high modulus carbon fibres, it cannot be assumed in general. Czél et al. [41] showed that the compressive failure strain was higher than the tensile failure strain when the experiments are very carefully performed. It is therefore likely that the common idea of compressive strength being lower than tensile strength is due to the difficulties in reliably measuring compressive strength. If true, this would mean that flexural failure is likely governed
by the tensile side. Fibre break models including flexural strain gradients would then become valuable tools for the prediction of bending strength.

Fatigue performance is a vital design driver for wind turbine blades [5,8,42-45]. The stiffness loss occurring during fatigue damage development can be important, as a large blade tip deflection could lead to collision with the tower. Luckily, this stiffness degradation is limited to a few percent for composites with primarily continuous 0° fibres that are well bonded [46]. Many fatigue models in the literature are essentially curve-fitting approaches that yield limited to no insight into the damage mechanisms. Mechanistic models therefore have greater value, as they can offer new understanding of the mechanism and have greater predictive capabilities [47,48]. In recent years, mechanistic models for unidirectional composite fatigue have been developed [34,44,49,50]. These models still have significant limitations:

- Alves and Pimenta [34] capture only a single damage mechanism, which is fibre-matrix debond growth. They assumed a Paris-Erdogan law for the growth [51], which implies that the debond keeps growing until failure.
- Dai and Mishnaevsky [44] developed an advanced finite element (FE)-based model, but still captured only a single fatigue damage mechanism, which is fibre-matrix debond growth. They also assume a Paris-Erdogan law [51].
- Sørensen and Goutianos [49,50] did take a limit on the debond length into account, but only predicted the critical state rather than how that critical state was achieved.

All these models ignore matrix creep and stress relaxation, matrix cracking/fatigue, fibre fatigue and other mechanisms that could potentially contribute.

This paper will provide an overview of developments at KU Leuven towards an all-encompassing model to predict failure modes that are governed by fibre break development, with a particular focus towards aspects relevant in wind turbine blades. After describing the baseline model, the more recent extensions will be briefly explained. The potential of the model will be demonstrated by presenting six different features relevant to wind turbine blades: fibre-hybridisation, size scaling, flexure, transverse cracking, stress relaxation and fatigue. As well as identifying the strengths of the current model, its limitations are reported. Future improvements needed to apply the model to the material design of wind turbine blades are highlighted.

2. Materials and methods

2.1. FE models for stress redistribution around a fibre break

FE models are used to create a library of stress concentrations around individual fibre breaks, which are then used by the strength model described in the next section. These models have been extensively described in earlier papers [52-54]. First, two-dimensional random fibre packings are generated using the basic algorithm of Melro et al. [55]. These packings are then extruded in the fibre direction to create a 3D model of parallel fibres (see Figure 1). Symmetry boundary conditions are applied to the entire top surface, apart from the broken fibre and a tiny matrix ring around the broken fibre. Displacement boundary conditions are applied on the bottom to load the bundle to an axial strain of 2%. The model size is chosen large enough, so that adding additional fibres or extending the length of the fibres does not change the calculated local stress concentrations.

The carbon fibres are transversely isotropic whereas S-glass fibres are isotropic, with the engineering constants reported in Table 1. The matrix properties that were used for the different models were adapted to the investigated scenario with the elastoplastic yielding law described by Okabe et al. [56] as default behaviour.

| Type   | $E_{11}$ (GPa) | $E_{22}$=$E_{23}$ (GPa) | $\nu_{12}$=$\nu_{13}$=$\nu_{23}$ (-) | $G_{12}$=$G_{13}$ (GPa) | $G_{23}$ (GPa) |
|--------|----------------|------------------------|--------------------------------------|------------------------|----------------|
| Carbon | 230            | 15                     | 0.25                                 | 13.7                   | 6              |
| S-glass| 88             | 88                     | 0.22                                 | 36.1                   | 36.1           |
The longitudinal stress of each fibre at a given axial location in the FE model is extracted by averaging the stress in the elements with the same axial position over the fibre cross-section. Dividing this average stress by the average far-field stress, the stress concentration factor (SCF) values are obtained. This was typically repeated for five different packing realisations of the model shown in Figure 1. Initially, the resulting SCF profiles in the broken and intact fibres were transferred through the use of trend lines [15,29,53], but later this was improved towards the use of trend surfaces [57]. For carbon/glass fibre-hybrid composites, two sets of models were built to capture the stress field around broken fibres of each used type [15].

2.2. Longitudinal tensile strength model

The basic strength model was developed between 2010 and 2015, and has been described in a range of publications [15,27,29,31,53]. A simplified version was also developed to help in analysing the effect of dispersion for fibre-hybrid composites [14]. The model was validated in detail using synchrotron computed tomography [29] and compared against two other state-of-the-art strength models [27]. Only the key points and assumptions will be described here, as some of the details have been improved over the years.

Figure 2 displays the flow chart of the strength model. It uses the same generator to create random fibre packings [55] as the FE model. Fibres are then extruded along the 0° direction typically over a length of a few millimetre. Each fibre is split up into elements that are typically one fibre radius long. The unimodal Weibull distribution was scaled down to the element length, and then a randomly generated strength value within that distribution was assigned to each element. During loading, the model gradually increases the strain in steps. Whenever the element stress exceeds the element strength, that element is considered broken. The model then checks for final failure, and if that is not the case yet, it will try to find additional fibre break clusters. Then, the stress concentrations from the FE models are used to update stress concentrations assuming that the breaks do not interact. Interactions between breaks in the same cluster is then accounted for using the enhanced superposition principle described in Swolfs et al. [53]. This process repeats itself until an unstable growth of a cluster is detected within one strain increment, at which point the model is stopped.
2.3. Extensions

The modelling framework is intrinsically able to handle fibre-hybridisation, although it was initially limited to predicting the initial failure strain of the carbon fibres or carbon fibre plies. Recent extensions described in Mesquita et al. [57] also enabled it to predict ply fragmentation in interlayer hybrid composites, taking into account interaction of ply fractures both within the same ply and in different plies. This required developing FE models on the ply-level so that the SCF values due to ply fractures could be incorporated in the longitudinal tensile failure model.

Another recent development was the extension of the model towards non-uniform strain fields. A first approach was to model flexural behaviour, as the global strain distribution can be estimated from simple beam bending equations. This enables the model to predict flexural behaviour on the condition that the failure is governed by the tensile rather than compressive stresses. A second approach was to incorporate strain fields in 0° plies in the vicinity of transverse cracks. The necessary fields were either obtained from analytical equations [58] or from FE models [59].

The baseline model was extended to include viscoplastic deformation of the matrix. Detailed measurements of this viscoplastic behaviour were performed for several epoxy matrices, and an advanced user subroutine was developed to incorporate this behaviour into the FE model. This subroutine captures the time as well as the strain dependency of the stress concentrations. The strength model was then extended to capture time rather than strain-based evolution of fibre breaks and clusters at a given strain rate.

A final extension was to include fatigue damage development by using four key assumptions:

- An initial debond length is present around every fibre break.
- The fibre-matrix debond grows according a Paris-Erdogan law, as measured by Pupurs et al. [60].
- The debond growth stops at a certain length, analogous to the approach of Sørensen and Goutianos [49].
- The fibre fatigues by assuming a strength degradation that is linear in a logarithmic scale, and reaches 40% degradation after 1 million cycles.

After an initial static run of the model, the model uses the concept of fatigue cycle jumps to predict how the debond length grows and how this affects other fibre elements. Using this extension, the model can predict the entire S-N diagram.

3. Results
3.1. Fibre-hybridisation

The most important parameter in hybridisation is arguably the degree of dispersion. It has a direct effect on the hybrid effect, which is the delay in the initial failure strain of the brittle ply in a hybrid composite relative to the failure strain in a composite with only brittle plies. Here, we will limit ourselves to the simple scenario of carbon plies sandwiched in between glass plies.

In a collaboration with the University of Bristol, we showed that our model predicts the correct trend: for thinner and hence better dispersed carbon fibre plies in a carbon/glass hybrid composite, the failure strain of the carbon plies increases (see Figure 3). However, a critical reflection shows that this effect is likely not relevant for wind turbine blades, as they tend to use heavy tow yarns to minimise cost. Blade materials therefore likely do not have sufficient dispersion to yield a measurable hybrid effect.

Initially, the model was stopped when the first critical cluster started to propagate. In a fibre-hybrid composite, this, however, does not necessarily coincide with the final failure. The model was therefore extended to predict the four potential scenarios after the initial failure of a carbon ply:

- Failure of the glass fibre ply;
- Delamination of the entire carbon fibre ply, followed by failure of the glass fibre plies;
- Fragmentation without delamination;
- Fragmentation with small delaminations around each fragment.

Figure 4 shows the predicted stress-strain diagram for a carbon/glass/carbon interlayer hybrid composite, alongside experimental data and a simpler analytical model. This particular material system fragments with very limited delaminations. The analytical model predicts a horizontal plateau when the material starts fragmenting, which is not in line with the slope seen in the experiments. This is because the analytical model assumes a uniform strength of the carbon fibre ply. The fibre break model does not assume this uniform strength, but rather inherently predicts its stochastic nature. Therefore, it correctly approximates the observed slope, but ply fragmentation starts too early. The ultimate failure strain is significantly overpredicted relative to the experiments. The too early fragmentation and too large ultimate failure strain are primarily attributed to uncertainty in the Weibull distribution for the carbon and glass fibres, respectively.

![Figure 3: Failure strain of the carbon fibre ply in a carbon/glass hybrid composite (reprinted from Wisnom et al. [19], with permission from Elsevier).](image-url)
Figure 4: Representative stress-strain behaviour predicted by the fibre break model for glass/carbon/glass hybrid composite, including experimental data from Czél et al. [23] and analytical predictions from Jalalvand et al. [61]. Each “+” marker represents a ply fracture (reprinted from Mesquita et al. [57], with permission from Elsevier).

The model does not only capture the complex stress-strain behaviour of fibre-hybrid composites, but also the underlying mechanisms of fibre break and cluster development and ply fragmentation. This implies the model can be used to guide the design of fibre-hybrid structures that are primarily subjected to tensile loading. At the moment, it is limited to intrayarn and interlayer hybrid configurations. A similar strategy as described in section 2.3 and used for Figure 4 could be used to extend it towards intralayer or bundled hybrid configurations. Such configurations have been much less studied in the literature, but are more relevant for pultruded spar caps.

3.2. Size scaling effects

Wind turbine blades are very large compared most coupons tested in lab conditions. Given the difficulties in testing very large coupons, models are essential to better understand how tensile strength and failure strain scales with size. Boundary effects are a vital feature in size scaling, but are often overlooked in other studies [37-39,56,62]. This is particularly relevant for scaling of the width and thickness.

Figure 5 plots the predicted failure strain for a 10 mm long model with two different boundary conditions applied. In the first case, the model boundary was left free, whereas in the second case, a ring of boundary fibres was added. These boundary fibres can carry stress concentrations, but possess infinite strength. This boundary condition eliminates the preferential cluster development at the perimeter of the model that occurs in the first case. Figure 5 shows a small difference in the predicted failure strain for both cases, but more importantly a large difference in how the failure strain evolves with the number of fibres. Without including boundary fibres, two competing mechanisms occur. On the one hand, an increased total number of fibres increases the likelihood of a combination of weak fibre elements located within each other’s vicinity, which will lead to a decrease in failure strain. On the other hand, more fibres make the preferential cluster development at the perimeter less important, which leads to an increased failure strain. The interaction of these two mechanisms results in a near absence of size scaling in the failure strain predictions. For the models with boundary fibres, the size scaling is much more pronounced, because the second mechanism is eliminated.
Figure 5: The predicted failure strain as a function of the number of breakable fibres, with and without boundary fibres. The length was kept constant at 10 mm (reprinted from Swolfs et al. [31], with permission from Elsevier).

The case without boundary fibres is likely the more representative case for comparing to experimental size scaling studies. It seems to indicate that size scaling is negligibly small. The study should however be extended to even larger specimen sizes to confirm this observation. For longer specimens or specimens with more than 10000 fibres, it is possible that the boundary effects will stop being relevant and more significant size scaling starts appearing. This would also seem to bring the model more in line with the experimental reports mentioned in the introduction. It should however be mentioned that those reports tended to scale all dimensions. In unpublished studies, we have observed that the length scaling in our model is much more significant than the cross-sectional scaling.

3.3. Flexure

With a strain field that varies linearly along the length (X) and thickness (Y) direction, the model can predict flexural strength. For the presented case study, a span length of 28 mm and specimen thickness of 1 mm were considered for estimating the strain gradient, but only the central 1 mm of the length was modelled. Figure 6 shows how this stress varies in the model, and how the fibre breaks (red dots) are concentrated in the regions with higher stress levels. In each model, the location of the largest cluster at the end of the simulation can be tracked. For the simulation run visualised in Figure 6, this happened to be very close to the middle (X=0.01mm). Figure 7 plots how this location varied for 100 simulations. This illustrates that a 1 mm long model is long enough to capture a span length of 28 mm, as the failure is unlikely to occur more than 0.5 mm away from the middle. This model is fully set up to include multiple fibre types needed to model carbon/glass hybrid composites.

Figure 6: Example of a stress field in the middle of the flexural fibre break model with red dots indicating the location of fibre breaks at failure (reprinted from Melnikov et al. [63]).
3.4. Transverse cracking

Similarly to the flexural model in the previous section, transverse cracking in a cross-ply laminate introduces non-uniform strain fields in the 0° plies. These longitudinal strain fields were captured either using analytical equations [58] or finite element calculations, but as Figure 8 shows, both options lead to very similar predictions. Three different cross-ply laminates with varying 90° ply thicknesses were modelled and compared to a UD composite. The predicted reduction in tensile strength is up to 17.1% relative to the UD composite. Interestingly, the experimental and modelling evidence of Noda et al. [64] showed no such reduction.

![Figure 8: Reduction in tensile strength in cross-ply laminates relative to a UD composite where the strain fields were either based on analytical equations or on FE models (reproduced from Melnikov et al. [59]).](image)

Wind turbine blade materials often contain a certain fraction of off-axis bundles or plies to stabilise the fabrics during manufacturing. Such bundles are known to reduce the fatigue strength of the 0° plies or bundles [65,66]. The modelling evidence here is line with computed tomography results [65,66], showing that fibre breaks develop preferentially at the tip of the off-axis cracks.

3.5. Stress relaxation

The stress relaxation model was initially loaded up statically and then held at a constant displacement for 3×10^6 s (about 1 month). Figure 9 shows the outcome of an initial study on stress relaxation. The matrix stress relaxation causes the stress concentrations around fibre breaks to spread out more over time, which can lead to additional fibre break and cluster development (see Figure 9). A later study extending the simulated time frame to 10^9 s (about 30 years). Models that were run at 94% of the mean static failure strain revealed no composite failure over time. For 20 models that were run at 98% of the...
mean static failure strain, however, 4 failed during the initial load-up due to the scatter in composite strength, 11 failed during the relaxation, and only 5 survived the entire modelled period.

![Figure 9: Outcome of the stress relaxation model in terms of fibre break density evolution over 3x10^6 s: (a) as a function of applied strain and (b) as a function of time (reproduced from Breite et al. [67]).](image)

3.6. Fatigue

Now that we have included fibre-matrix debond growth as a function of the number of cycles, the model can predict the entire S-N diagram of a UD composite. Figure 10 plots this diagram with and without the inclusion of fibre fatigue. Without the inclusion of fibre fatigue, the predicted fatigue performance is significantly better than the experimentally measured fatigue performance. By including fibre fatigue, the predictions are more or less in line with the experimental data. However, it should be noted that percentage of fibre fatigue was arbitrarily chosen and that such strong fibre fatigue is likely not realistic for carbon fibres. Some authors even indicate that carbon fibres do not fatigue [68], although this study has not been convincingly replicated yet. Glass fibres are known to fatigue, but there are ongoing discussions on whether this really is fatigue or just stress corrosion. Such corrosion is likely much less significant when the fibres are surrounded by matrix [69]. Another reason for this discrepancy could be in the difficulty to perform reliable fatigue tests on UD composites. Such composites are known to be very sensitive to gripping effects, which can be even worse in fatigue.

![Figure 10: Predicted S-N diagrams for unidirectional carbon fibre composites without and with the inclusion of fibre fatigue, and compared against experimental data sets for unidirectional carbon fibre composites [70-74].](image)

Unfortunately, the fibre-matrix debond growth data used as input to create Figure 10 was based on glass fibre/epoxy measurements. Similar data for carbon fibres is not available in the literature.
The model can be readily adapted for fibre-hybrid composites, provided that the necessary input data are available. This development is currently ongoing.

4. Conclusion and outlook

Fibre break models are powerful tools that can predict a much broader range of properties than just longitudinal tensile strength of unidirectional composites. This paper showed a range of extensions to increase the practical relevance and usefulness of the strength model developed at KU Leuven. A key issue in all of the presented extensions is their experimental validation. While the model seems to capture the key damage mechanisms, it does not capture all of them and hence extrapolations beyond the validated range of materials and behaviours can be unjustified.

In the future, we should focus on better integrating the various features. For example, the flexural and transverse cracking studies have so far limited themselves to quasi-static failure of non-hybrid composites, but can easily be extended to include fatigue, stress relaxation/creep and fibre-hybridisation. Two extensions in particular would be very useful in the modelling of wind turbine blades. The first extension would be to include compressive failure mechanisms. This would enable modelling quasi-static flexure in more detail, and it would open the door towards modelling flexural fatigue. This is the key behaviour driving the design of wind turbine blades. Secondly, the model can be incorporated in a multiscale modelling strategy so that much larger components can be simulated. A proper understanding of size scaling effects in the base model is vital, as they will need to be transferred to the larger scale.

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