MULTI-SCALE MODELING OF FAILURE OF CONTINUOUS CARBON FIBER COMPOSITES
APPLICATION TO COUPON TESTS

Benoît Bidaine, Laurent Adam, Marc Duflot, Bender Kutub, Emmanuel Lacoste,
Roger Assaker
e-Xstream engineering
Hanson Chang
MSC Software

Abstract
In the steady quest for lightweighting solutions, continuous carbon fiber composites come
down to the ground, serving now not only the aerospace but also the automotive industries. This
category of Carbon Fiber Reinforced Plastics (CFRP) has recently taken a step in car body
structures for its high stiffness and strength.

Continuous carbon fiber composites are much more complex than metal, with respect to
failure in particular. If they are so-called unidirectional, they involve stacks of several plies, each
ply characterized by a single fiber orientation. Hence they fail because of various mechanisms
taking place at the ply level (matrix cracking, fiber breakage, fiber-matrix debonding) or between
the plies (delamination). These mechanisms remain not fully understood and are investigated
through experimental and virtual testing.

To predict composite failure, we have developed advanced simulation strategies combining
finite element analysis (FEA) and nonlinear micromechanical material modeling. In particular,
we implemented progressive failure models such as Matzenmiller-Lubliner-Taylor to prevent the
analysis to become unrealistic after the first elements have failed. In addition, we enriched these
models with inputs computed at the micro scale. Indeed multi-scale modeling decomposes the
macroscopic mechanical state between fibers and resin, enables the definition of per-phase
failure criteria and provides access to macroscopic or microscopic stiffness degradation. In this
paper, we will cover the application of micro-mechanically-based progressive failure models to
simple demonstrator structures such as coupons.

Introduction
The automotive industry has recently increased its use of continuous carbon fiber
composites. These carbon fiber reinforced plastics (CFRP) are replacing traditional materials
such as extruded metals for structural parts and thermoplastic composites for body parts. CFRP
are traditionally used in the aerospace industry due to their light weight, high strength and
stiffness properties. The recent shift in the CFRP use by the automotive industry is due to the
high demands for reduction of C02 emissions, fuel efficiency and high performance vehicles.
The new BMW i3 and i8 models have achieved a 25% reduction in weight compared to
conventional thermoplastic car designs through the use of CFRP composite materials [1].

CFRP provide substantial weight reduction and increased performance in automotive
designs but also increased complexity. A crucial step for implementing CFRP into production
parts is the ability to accurately predict and characterize the failure mechanisms of the material:
it actually constitutes the bottleneck. In particular, the microstructure of laminates creates very
complex failure mechanisms such as matrix cracking, fiber breaking, and delamination [2]. Each of those failure mechanisms and others can have a more or less dominating effect based on the different types of loading conditions the material is subject to. Therefore the intricacies of these failure mechanisms remain not fully understood and are very difficult to characterize. The two methods used to describe and characterize CFRP strength and failure are physical testing and virtual simulation. Although the most accurate estimates of the strength of a composite is achieved by means of physical coupon testing, such as open hole tension and compression tests. The tests required are often extensive, costly and only give raise to a macro level understanding of the material strength with not a lot of information about the mechanisms that caused the failure itself. Hence virtual finite element analysis (FEA) of experimental tests such as open hole tension and compression can prove very valuable.

The development of accurate virtual testing can help dramatically reduce the cost and time needed by physical testing. The key to accurate strength and failure predictions using FEA lies in the proper development of robust micromechanical material models and failure criteria. The use of progressive failure modeling has proven to be very effective in capturing the actual end behavior of CFRP laminate failure. The use of multi-scale micromechanical material modeling has proven to enhance the predictions of the stiffness of composites and provides micro-per phase level material responses.

The purpose of this paper is to illustrate how multi-scale material modeling can be coupled with progressive failure analysis and provide the ability to develop robust failure models for continuous carbon fiber composites. This paper first presents the main characteristics of such composites, focusing on one particular unidirectional (UD) tape example. Then it describes how their behavior can be modeled thanks to micromechanics and progressive failure. Finally it addresses the ability to exploit corresponding material models in the framework of multi-scale structural simulations.

**Continuous Fiber Composites**

Continuous fiber composites consist of polymer matrices, typically epoxy, reinforced with continuous fibers, often made of carbon. These materials are stiffened by the fibers when the latter are aligned with the direction of loading. Hence these materials are usually engineered in stacks of several plies exhibiting various fiber alignments, called UD laminates. They are also characterized by a certain mass or volume fraction of fibers.

Due to the large spread in constituent properties, continuous fiber composites exhibit very different failure behaviors depending on the angle between loading and fiber directions. In particular, UD laminates host various failure mechanisms in different regions (cf. Figure 1). In aligned plies, where loading and fiber directions correspond, the stresses are mainly transmitted through the fibers. Hence fiber breakage initiates failure in these plies. In transverse plies, both the matrix and fibers support the loading but the matrix or the fiber-matrix interface get damaged first. Hence matrix cracking mainly accounts for failure in those plies. Taking into account this disparity in mechanical behavior, a separation between plies of different fiber orientation can also appear leading to delamination.
Figure 1: Failure a UD laminate under tensile loading. Various mechanisms occur in and between plies.

Composite properties are characterized through experimental testing [3]. Various tests are performed, triggering different failure modes: on single plies or laminae or on laminates; at various loading angles for laminae (e.g. in the fiber – or warp – and transverse – or fill – directions) or for various layups for laminates; with or without structural characteristics such as notches or holes; in tension, compression or shear.

By way of example, the material system 8552/IM7 by Hexcel exhibits the typical anisotropy of continuous carbon fiber composites, both in terms of stiffness and strength (cf. Figure 2). Such properties have been characterized by the National Institute for Aviation Research and are publicly available [4].

- The tensile modulus of a lamina is more than 15 times larger in the warp direction than in the fill direction. It adopts an intermediate value for a so-called quasi-isotropic (unnotched) laminate i.e. with an equal of plies hosting fibers aligned at 0°, 45°, 90° and -45° with respect to the loading direction. It is generally slightly larger than the corresponding compressive modulus.

- The tensile strength of a lamina is more than 40 times larger in the warp direction than in the fill direction. It adopts an intermediate value for a quasi-isotropic laminate (unnotched or open hole). It is generally larger than the corresponding compressive strength, apart for the fill test for which it is much smaller.

Figure 2: Mechanical properties of the material system 8552/IM7 by Hexcel. The properties exhibit a large variability depending on the loading direction (along fiber – or warp – direction or transversely – fill – considering a single ply or lamina) or type (tension/compression) among others. They reach their maximum values for the warp direction.
Micromechanical Material Modeling

The simulation of continuous carbon fiber composites advantageously combines micromechanics, deriving composite properties from constituent properties e.g. through mean-field homogenization, and progressive failure.

Mean-Field Homogenization

As composite properties depend on the material microstructure including fiber amount and orientation, they are adequately modeled from micromechanics. In particular, mean-field homogenization combines the properties of the underlying constituents of a multi-phase material so that the original heterogeneous material is represented by an equivalent homogeneous one. Implemented in the Digimat software [5], this technology has proven effective for a broad range of materials. For CFRP, it represents the matrix material as isotropic elastic (or even elastoplastic), the fiber material as transversely isotropic elastic and accounts for the actual fiber volume fraction.

Mean-field homogenization provides a means to investigate the origin of the experimental variability of composite properties. In particular, it reveals their sensitivity to micromechanical parameters (cf. Figure 3). In turn, these parameters can be considered as effective parameters enabling fits of sets of different composite measurements. For instance, the experimental variability of the quasi-isotropic tensile modulus (labeled “Unnotched” in Figure 2) can be compared to the corresponding simulated variability from 10% variations of different matrix or fiber properties: varying the fiber longitudinal modulus or the fiber volume fraction yields similar modulus ranges.

![Figure 3: Sensitivity of the quasi-isotropic tensile modulus to micromechanical parameters](image)

Mean-field homogenization provides access to per-phase properties. It enables a finer interpretation of simulation results as it distinguishes the matrix and fiber behaviors. In particular, it enables the definition of failure criteria at the phase level while they are usually or at first defined at the composite level.
Progressive Failure

Several strategies can be used to deal with the failure of quasi-brittle materials. The simplest method consists in abruptly degrading the material stiffness when a failure criterion, i.e. a given combination of stress/strain components, reaches a critical value. A drawback of this method is that the material stiffness is reduced in every direction, which is unrealistic for laminate composites: a UD ply that fails in the transverse direction (due to matrix cracking) still exhibits a significant stiffness in the fiber direction. Some models, such as the Chang and Chang model [6], were developed in order to account for this kind of anisotropic degradation. An enhancement of this method was formalized by Talreja [7] through the Continuum Damage Mechanics (CDM) framework, which uses damage state variables in order to apply a gradual (and not instantaneous) degradation of the material, translated in a softening of the stress-strain curve before failure.

Progressive failure consists in linking failure criteria to stiffness degradation through damage variables. A popular application of this formalism is the Matzenmiller-Lubliner-Taylor (MLT) model [8], in which the stress-strain behavior of the composite material (considered at the macroscopic scale) is represented by the equation

$$
\begin{bmatrix}
\varepsilon_{11} \\
\varepsilon_{22} \\
\varepsilon_{12}
\end{bmatrix} =
\begin{bmatrix}
\frac{1}{(1-D_{11})} & \frac{1}{E_1} & -\frac{\nu_{12}}{E_1} & 0 \\
-\frac{\nu_{12}}{E_1} & \frac{1}{(1-D_{22})} & \frac{1}{E_2} & 0 \\
0 & 0 & \frac{1}{(1-D_{12})} & \frac{1}{G_{12}}
\end{bmatrix}
\begin{bmatrix}
\sigma_{11} \\
\sigma_{22} \\
\sigma_{12}
\end{bmatrix}
$$

where $D_{11}$, $D_{22}$ and $D_{12}$ denote damage variables. These variables are often expressed from failure criteria, e.g. for the Hashin tape failure criterion:

$$
D_{11} = \varphi(f_F), \quad D_{22} = \varphi(f_M) \quad \text{and} \quad D_{12} = 1 - (1 - D_{11}) \times (1 - D_{22}),
$$

with

$$
f_F = \begin{cases}
\left(\frac{\sigma_{11}}{X_t}\right)^2 + \left(\frac{\sigma_{12}}{S}\right)^2 & \text{if } \sigma_{11} > 0 \\
-\frac{\sigma_{11}}{X_t} & \text{otherwise}
\end{cases}
\quad \text{and} \quad
f_M = \begin{cases}
\left(\frac{\sigma_{22}}{Y_t}\right)^2 + \left(\frac{\sigma_{12}}{S}\right)^2 & \text{if } \sigma_{22} > 0 \\
0 & \text{otherwise}
\end{cases}
$$

where $\sigma$ stands for the effective (or undamaged) stress tensor. $\varphi(f)$ stands for the damage law, which is activated only when the failure criterion increases (irreversible damage).

The damage law shapes the stress-strain behavior between damage initiation and ultimate failure. It can follow several generic formulations implemented in Digimat, among which the 2 following simple examples:

$$
\varphi(f) = \begin{cases}
0.99 & \text{if } f \geq 1 \\
0 & \text{otherwise}
\end{cases} \quad \text{(instantaneous damage)}
$$

$$
\varphi(f) = \begin{cases}
0 & \text{if } f \leq f_{\text{min}} \\
\frac{f - f_{\text{min}}}{f_{\text{max}} - f_{\text{min}}} & \text{if } f_{\text{min}} < f \leq f_{\text{max}} \\
1 & \text{otherwise}
\end{cases} \quad \text{(linearly increasing damage)}
$$

The first example eventually switches the stiffness almost off, however in a single direction (cf. Figure 4 showing results of the MLT model applied to a UD 8552/IM7 lamina loaded in the fiber direction, labeled “Warp” in Figure 2). The second example yields a gradual stiffness decrease after the maximum stress has been reached (with $f_{\text{min}} = 1$ and $f_{\text{min}} = 3$) or even introduces a non-linear behavior before the maximum stress (with $f_{\text{min}} = 0.8$ and $f_{\text{min}} = 4.7$).
Multi-scale Structural Simulation

Above-described material models bring more realistic material knowledge to structural simulation. In the framework of coupled FEA, they provide homogenized material properties at the integration point level based on the local microstructure. When these properties vary over a part because of the manufacturing process, micromechanical material models enable thus a more accurate description of the part performances: the multi-scale approach reveal the influence of the microscopic properties on the macroscopic performances.

In particular, progressive failure improves the realism of structural simulations after damage initiation, a local event in both space and time. In a classical implicit FEA, damage initiation is inferred from failure criteria. However the analysis becomes unrealistic after this moment as the material behavior is not modified in the damaged region. In an explicit FEA, elements where a failure criterion has reached a critical value can be deleted. However such element deletion actually corresponds to mass removal and, unless the elements are very small, induces structural instability and precipitated failure path propagation. On the contrary, progressive failure accounts for material damage within an element by gradually decreasing the stiffness in the corresponding direction.

Such tools apply particularly well to coupons and other simple structural components. Such components, e.g. an overmolded beam made of two different composite materials (cf. Figure 5), are indeed extensively tested experimentally to guarantee the reliability of more complex structures, as long as the materials are concerned. Hence improving the FEA realism for such parts would reduce the need for experimental testing, as well as time and cost likewise.
Summary and Next Steps

Continuous carbon fiber composites have rapidly spread across automotive components for their lightweighting capabilities but pose new design challenges because of their complex properties. In particular, their failure behavior is not easily characterized and requires new tools to be realistically simulated. In that respect, micromechanical material models allied to progressive failure provide an in-depth understanding of the composite behavior at the constituent – matrix or fiber – level and a directionally selective stiffness degradation. Hence they pave the way for a reduction in experimental testing in favor of virtual testing.

We have implemented the Matzenmiller-Lubliner-Taylor model in Digimat for traditional laminate problems and are currently extending such advanced simulation tools to a wider range of solutions. The tools are interfaced to the most popular FE software. They have been successfully applied for a macroscopic damage description but will benefit even more from a microscopic damage formulation, relating per-phase failure criteria to constituent stiffnesses.

Bibliography