

Multi-scale characterization and modelling of the transverse compression response of unidirectional carbon fiber reinforced epoxy

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Abstract

Thick unidirectional fiber reinforced RTM6 epoxy resin specimens were processed by resin transfer molding and tested in transverse compression. Homogenized response from micromechanical analyses on representative volume elements, using a constitutive model for the matrix identified and validated on bulk RTM6 specimens, were assessed towards the experimental results. Large discrepancies were observed in terms of failure initiation and non-linearity in the stress-strain response. Regarding the interfaces, the use of frictional cohesive elements greatly enhanced the damage tolerance of the RVEs to interface damage. Regarding the matrix behavior, in situ tests on UD specimens and microscale digital image correlation highlighted significant gaps between the numerical and experimental strain fields. Excessive strain localization in the simulations is believed to be the main reason for these differences. This study raises fundamental questions on the degree of confidence that can be granted to constitutive models validated at the macroscale to predict strain field at the microscale.

Keywords: Epoxy resin, Micro-mechanics, Polymer matrix composite (PMC), Digital image correlation (DIC), Finite element analysis (FEA)

1. Introduction

Bottom-up approaches dealing with the prediction of the mechanical properties of fiber reinforced polymers (FRP) usually start by the determination of the response at the scale of a laminate ply. In this regard, computational micromechanical analyses have become a widespread tool to characterize the anisotropic failure and the deformation behavior of unidirectional plies. These analyses are often based on finite element (FE) simulations of representative volume elements (RVE) of the composite microstructure and require, as input ingredients, the properties of each constituent. Hence, the proper modelling of the mechanical behavior of FRP heavily relies on an accurate characterization of the constituents behaviour, that is, essentially, the fiber/matrix interfaces, the matrix and the reinforcing fibers.

The characterization of the mechanical response of the interfaces is a challenging task given the small-scale of the mechanisms associated to the regions between the fibers and matrix. As a matter of fact, the first studies on the role played by the interfaces were dedicated to the analysis of the influence of the interface resistance on the macroscopic response of the composite, mainly by varying the fiber surface treatment [1, 2]. The direct determination of the actual properties of interfaces is a more recent subject of interest, mainly favored by the development of new characterization methods, such as fiber fragmentation tests [3, 4], single fiber pull-out [5–7], or more recently the push-out/in of a single fiber in order to probe the resistance of a single interface [8–10]. However, despite the recent progress, the extraction of the actual interfacial properties remains a challenge and is hindered by the strenuous task to clearly identify and distinguish the role played by matrix plasticity and interfacial crack initiation [8, 9]. In computational micromechanics, the fiber-matrix interfaces are almost exclusively represented by cohesive elements through adequate

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traction separation laws. As the number of quantitative fiber-matrix interfacial characterization studies remains scarce, the material parameters that are selected for the interfacial strength and toughness are often very similar, either based on the early modelling work by Varna [11], on the inverse identification of the cohesive parameters to capture tensile failure in unidirectional composites (UD) by Vaughan and McCarthy [12], on the properties proposed by Gonzalez and Llorca [13] or on a combination of them [14–20].

The mechanical behavior of thermoplastic and thermoset polymer matrices has been largely characterized, revealing a complex intermingling of viscoelasticity, viscoplasticity, strain softening, large strain rehardening as well as pressure and temperature dependence [21–25]. Several constitutive models have been developed over the years in an attempt to capture the full range of the post-yield phenomena, at the cost of many parameters, often more than 25 [24, 26–30]. However, in computational micromechanical analyses, more simple models have generally been privileged, neglecting strain softening and/or large strain rehardening [16, 19, 31, 32], or even considering the matrix as elastic perfectly plastic in compression [33]. In conditions where the matrix mainly dictates the mechanical response of the RVE (i.e. in longitudinal and transverse shear or in transverse compression), these assumptions lead to inaccurate actual strain and stress fields at the scale of the fibers.

Despite the development of computational micromechanics, very few of these analyses have been confronted to direct experimental results on UD specimens, particularly in terms of the stress strain response of the RVEs. Hence, there is a lack of validation for results that are considered representative of the mechanical behavior at the ply level. Recent works have highlighted the possibility to use digital image correlation (DIC) to accurately determine strain fields at the microscale by performing in situ tests in scanning electron microscope (SEM) [34, 35], with data relevant at the scale of few fibers in the case of FRP [36–38]. This strain field can then be quantitatively compared to finite element analyses (FEA) predictions using the exact same fiber locations and proper boundary conditions. To the knowledge of the authors, such analyses are still either i) limited to the cases of transverse tension and bending, limiting the extent of plasticity development in the matrix prior to failure, or ii) compared to FE solutions using an elastic model for the matrix. Thus, in either case, their validation capabilities with respect to the choice of proper matrix constitutive models is limited. Moreover, compression tests on micropillars of pure epoxy resin [39] as well as nanoindentation tests within matrix pockets in CFRP [40] have highlighted that the mechanical behavior of epoxy resins is different than when compared to bulk specimens, where the constitutive models are usually identified and validated. As a consequence, recent studies aimed at the identification of the in situ matrix yield stress using nanoindentation tests in the matrix pockets or using micropillars directly milled out of resin pockets in FRP [41].

Previously, a constitutive model for the RTM6 epoxy resin was identified and validated based on a wide experimental campaign on pure resin specimens [25, 30]. The aim of the present work was to perform a two-scale assessment of the matrix model validity in CFRP by directly confronting experimental results at the macro- and microscale and numerical predictions. For that purpose, thick UD slabs have been manufactured using the resin transfer molding (RTM) process, allowing the use of the same curing conditions as for the pure resin bulk specimens. Transverse compression tests provided experimental results in scenarios where the mechanical response is mainly dictated by the matrix. Homogenized responses from FEA on UD RVEs were compared to these macroscale transverse compression tests. Additionally, in situ compression tests in a SEM chamber on the same UD composite were used to carry out DIC analyses on the surface of the in situ specimens. Hence, the validity of the constitutive model is further evaluated by a direct quantitative comparison between the DIC strain field and FEA predictions using the same boundary conditions and fiber distributions.

This paper is organised as follows: i) section 2 describes the UD specimens manufacturing, the testing methods, the constitutive models for FEA, the in situ tests and the DIC method, ii) the experimental results obtained on the thick UD specimens are presented in section 3 where they are compared to the FEA predictions from RVEs, iii) section 4 compares the DIC measurements from the in situ tests and the corresponding FEA, iv) section 5 proposes a correction of the matrix hardening law based on comparing nanoindentation tests in UD matrix pockets and in a pure resin specimen, and v) a general discussion on the results presented in the previous sections is elaborated in section 6.

2. Materials, processing and methods

2.1. Material and macroscopic testing methods

The UD slab was manufactured by RTM using unstitched unidirectional fibers layers from a $+45^\circ / -45^\circ$ non crimp fabric made of HTS carbon fibers (12K), provided by Saertex GmbH. The fabric areal density was 540g/m^2

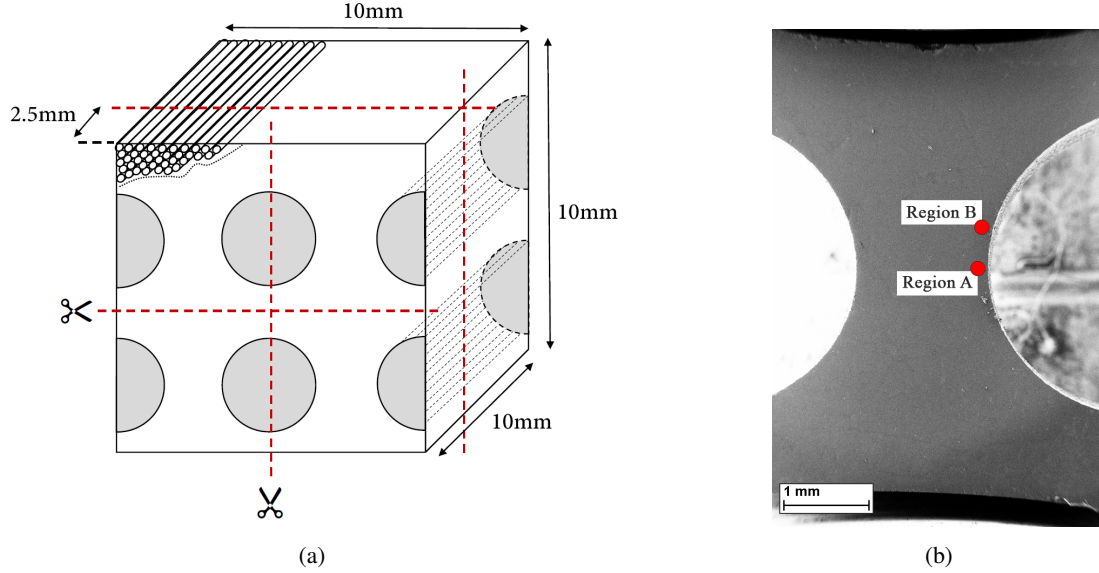


Figure 1: (a) schematic of the extraction of the in situ notched specimens from the cubic specimens; (b) SEM micrograph of a notch specimen, the red dots locate the approximate position of the regions A and B used for the DIC measurements.

(i.e., that is 270g/m^2 for the UD layers). Once unstitched, the UD layers were cut to the dimensions of the mould ($520 \times 30 \text{ mm}$) and 37 layers were stacked into the mould to provide the desired fiber volume fraction. The matrix was the monocomponent HexFlow RTM 6 epoxy resin, from Hexcel. RTM6 is provided as a premixed system composed of tetra-glycidylmethylenedianiline (TGMDA) epoxy polymer and of two amine curing-agents M-DEA and M-DIPA. The curing cycle was identical to the one employed to produce bulk RTM6 specimens for mechanical characterization as described in [30]. The $520 \times 30 \times 10 \text{ mm}$ UD slab was first heated at 2°C/min from 90°C to 130°C , held at 130°C during 3h and then heated again at 2°C/min from 130°C to 180°C for a 3h long post-curing cycle. Then, $10 \times 10 \times 10 \text{ mm}$ cubic specimens were machined out the slab at Cenaero (BE), with particular care to avoid excessive heating and delamination. The resulting fiber volume fraction was 40%.

The transverse compression tests were performed on a screw-driven universal testing machine (Zwick-Roell with an external loading cell of 250 kN). The axial displacement was measured using a compliance corrected crosshead displacement method. All tests were performed at room temperature and three different true strain rates were used : $\dot{\epsilon} = 10^{-4}$, 10^{-3} and 10^{-2}s^{-1} . Three samples were tested for each strain rate. The friction was minimized by inserting PTFE films between the specimens and the compression platens of the testing machine.

2.2. In situ tests and digital image correlation

The in situ tests were performed inside the chamber of a Field Emission Gun Scanning Electron Microscope (FEG-SEM), using a Deben microtester stage with a 2kN load cell. Compression load was chosen in order to favor the development of plasticity in the matrix prior to the appearance of damage and to the final failure of the specimens. In order to obtain the desired geometry for the in situ specimens, holes were carefully drilled in a cubic specimen, in the longitudinal direction. 2.5mm thick cross-sections specimens were then cut out of the cubic specimens using a Accutom cut-off machine from Struers, as illustrated in Fig. 1a. Four notched specimens could be extracted from a single slice. Fig. 1b shows a SEM micrograph of a notched specimen obtained by this method. The notches in the specimens have the double advantage of i) limiting the strength of the specimens to stay below the microtester limit load and ii) knowing a priori the location of macroscopic strain localization in the specimen. The specimens were polished using resin bonded diamond discs with successive grain size of 1200 and 4000, followed by a last polishing using an alumina suspension with particle size of $0.3\mu\text{m}$. For digital image correlation (DIC) measurements, a speckle pattern was deposited on the specimen made of an alumina powder, with an average particle size of $0.3\mu\text{m}$, following

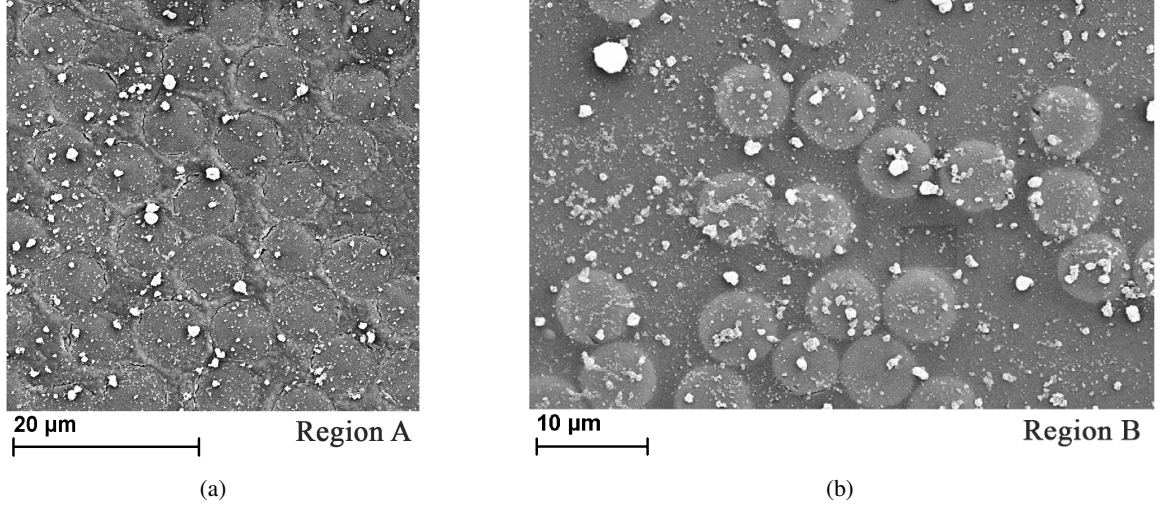


Figure 2: SEM micrographs of the undeformed regions (a) A and (b) B.

the procedure described in [37]. The surface of the in situ specimens was gold coated after the deposition of the pattern to avoid charging effect in the matrix. The test was stopped at regular intervals to take micrographs, inducing a small stress relaxation of the specimen at every pause in the loading phase.

The DIC measurements from two different tests are addressed in this paper. Fig. 2a and 2b show the SEM micrographs of the regions on which DIC was applied, each zone corresponding to a different test. The regions were selected to obtain two different fiber volume fractions: 55% in Fig. 2a and 30% in Fig. 2b, and will be respectively denoted as region A and region B. Region A has a resolution of $1000 \times 1000 \text{pxl}^2$ and region B $764 \times 501 \text{pxl}^2$. The position of each region in the notched specimens is approximately indicated by a red dot in Fig. 1b. The open source software Ncorr [42] was used to perform the DIC analyses. Ncorr builds a correspondence between materials points in a reference image in the undeformed state and in the deformed state. The choice of the DIC parameters (in particular, the subset and the step size) strongly influences the displacement and strain fields from the analyses and is thus a key aspect of any DIC measurement, especially when dealing with composite materials and also the fact that only small material regions are of interest in this work puts additional constraints on the DIC. As demonstrated in [43], there is no ideal set of parameters to simultaneously decrease the noise on strain and to increase spatial resolution. Ultimately, the best set of parameters strongly depends on the application and on the relevant information to extract from the DIC analysis. Increasing the subset size implies lesser loss in spatial resolution than when increasing the step size [43]. Hence, in this case, the latter was kept low while the subset size was increased until the noise on strain was sufficiently low. For region A, a subset size of 20pxl and a step size of 2pxl were selected while, for region B, the subset size was fixed at 30pxl and the step size at 1pxl.

2.3. Nanoindentation tests

The nanoindentation tests in a pure resin specimen and in matrix pockets of the UD composite were performed using a Agilent G200 nanoindenter equipped with a Berkovich diamond tip. A loading-unloading sequence at a constant loading rate $\dot{F}/F = 0.1 \text{s}^{-1}$ was used. The indentation spots in the UD composite were chosen to avoid any contact between the indenter and the fibers, which implied restrictions on the final displacement into the surface. Two different final indentation depths were chosen: $1 \mu\text{m}$ and $2 \mu\text{m}$. For the bulk resin specimen, there was no restriction on the final indentation depth and the tests were conducted until reaching a load of 120mN. At least three tests were conducted in each condition. The Continuous Stiffness Measurement (CSM) mode was activated in order to obtain a continuous measure of the hardness during the tests. Oliver and Pharr method [44] was used to estimate the contact area between the indenter and the specimen surface, assuming sink-in around the indenter. The surface of the nanoindentation specimens was polished using resin bonded diamond discs with successive grain size of 1200 and 4000,

followed by a last polishing using an alumina suspension with particle size of 0.3 μ m.

2.4. Finite element modelling

Finite element analyses were carried out using the commercial code Abaqus [45]. Two types of finite element analyses were performed: i) analyses on RVEs with random fiber distribution and ii) analyses with the same fiber configuration as in Fig. 2a and 2b.

2.4.1. Constitutive models

Fibers. The fibers were considered as linear elastic and transversely isotropic. The elastic properties of the fibers are given in Table 1.

	E_{11} [GPa]	E_{22} [GPa]	E_{33} [GPa]	ν_{12}	ν_{23}	ν_{31}	G_{12} [GPa]	G_{23} [GPa]	G_{31} [GPa]
Fibers	238	28	28	0.28	0.33	0.02	24	7.2	24
Matrix	3	-	-	0.34	-	-	-	-	-
	K [N/mm ³]	τ_1^0 [MPa]	τ_2^0 [MPa]	τ_3^0 [MPa]	G_{Ic} [N/mm]	G_{IIc} [N/mm]	G_{IIIc} [N/mm]		
Interfaces	10 ⁸	75	75	50	0.002	0.006	0.006		

Table 1: Elastic materials properties of the fibers, the matrix and the fiber-matrix interfaces used in the finite element analyses.

Matrix. The matrix was modelled as elastic-plastic with a rate-dependent hardening, following the constitutive model identified and validated by Morelle [30]. Fig. 3a superimposes experimental curves at three different true strain rates and the corresponding hardening laws from the model. The pressure dependent yield surface of the matrix was described by a linear Drucker-Prager model, whose general form writes:

$$F = t + p \tan \beta - d = 0 \quad (1)$$

with t the yield stress, p the pressure defined as $\frac{\sigma_{kk}}{3}$, with σ_{kk} the trace of the stress tensor, β the friction angle and d the cohesion term. The friction angle β consists in the slope of the yield surface in the $p - t$ plane, where t is defined as

$$t = \frac{1}{2} \sigma_{eq} \left[1 + \frac{1}{r} - \left(1 - \frac{1}{r} \right) \left(\frac{I_3}{\sigma_{eq}} \right)^3 \right] \quad (2)$$

with σ_{eq} the von Mises equivalent stress, r the stress ratio between the yield stress under uniaxial tension σ_t and pure compression σ_c and I_3 the third invariant of the deviatoric stress tensor. In the particular case of RTM6, $r = 1$ and thus $t = \sigma_{eq}$. As the hardening law is provided in compression, the yield surface expression must be written as:

$$F = \sigma_{eq} + \frac{\sigma_{kk}}{3} \tan \beta - \left(1 - \frac{\tan \beta}{3} \right) \sigma_c = 0. \quad (3)$$

The corresponding plastic flow rule is expressed as:

$$G = \sigma_{eq} + \frac{\sigma_{kk}}{3} \tan \varphi \quad (4)$$

with G the flow potential and φ the dilation angle. The model was validated under six decades of strain rates with $\beta = 7.86$ and $\varphi = 0$, assuming plastic incompressibility [30]. The elastic properties of the matrix are listed in Table 1. A fracture criterion is added in the description of the matrix model, under the form of a stress triaxiality-dependent critical equivalent plastic strain criterion, which has been identified using uniaxial tensile, pure compression and torsion tests on bulk resin samples. The criterion writes:

$$\varepsilon_{eq,f}^p - ae^{-b\frac{p}{\sigma_{eq}}} - c = 0 \quad (5)$$

with $\varepsilon_{eq,f}^p$ the equivalent plastic strain at failure and a , b and c the parameters to identify. Fig. 3b shows the description of the criterion in the ε_{eq}^p /stress triaxiality plane, with $a = 0.0404$, $b = 7.8152$ and $c = 0.0256$. It is implemented as a ductile damage criterion in Abaqus with a fracture energy set at 1J/m^2 . This value ensures that element failure occurs at the instant when the equivalent plastic strain reaches its critical value.

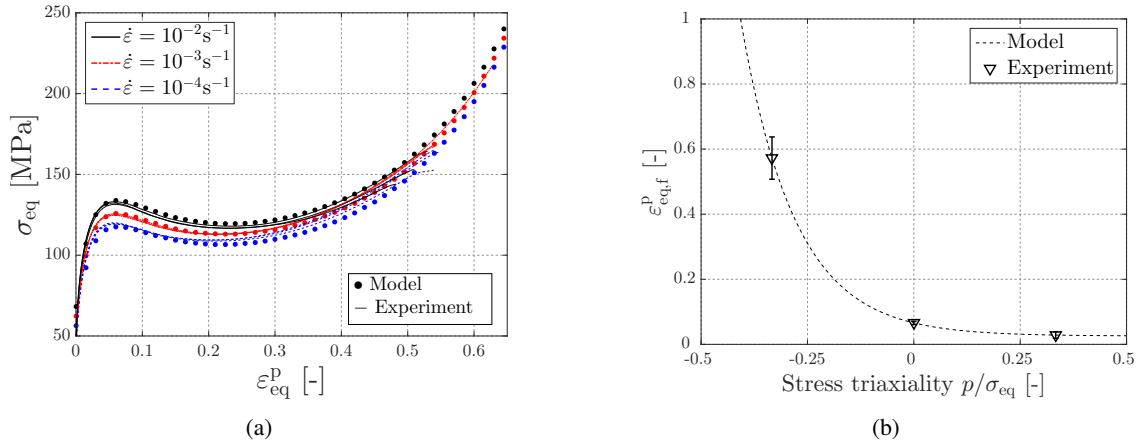


Figure 3: (a) Superimposition of RTM6 constitutive model and of the experimental $\sigma_{eq} - \varepsilon_{eq}^p$ curves at three different strain rate and at room temperature, in pure compression; (b) pressure dependent failure criterion used for RTM6, shown in the $\varepsilon_{eq,f}^p$ - stress triaxiality σ_h/σ_{eq} plane.

Interfaces. The interfaces between the fibers and the matrix were modelled using cohesive elements whose mechanical behavior was described by a linear traction-separation law. The properties of the interfaces are taken from Arteiro et al. [18] and are shown in Table 1. In a first approach, the interfaces were modelled using Abaqus formulation for cohesive elements. In this formulation, it is assumed that once the cohesive elements are completely damaged, they cannot carry any load. However, friction between the matrix and the fibers should allow the interfaces to still carry some load. Hence, in a second stage, a combination of cohesive damage and friction in a cohesive zone model was used, based on the methodology proposed by Alfano and Sacco [46]. It was coupled to the cohesive zone model proposed by Turon [47] and implemented in a user-material subroutine VUMAT [48].

In Alfano and Sacco formulation, the relative displacement between the two regions at the interface can be divided into a damaged part Δ^d and an undamaged part Δ^u . The mechanical response of the latter is assumed to remain linear. Δ^d is divided into an elastic Δ^{de} and an inelastic part Δ^{di} :

$$\Delta^d = \Delta^{de} + \Delta^{di}. \quad (6)$$

As the displacement along a given representative volume area (REA) is assumed to be constant, $\Delta^u = \Delta^d = \Delta$. The interface stresses are also divided into a damaged part τ^d and an undamaged part τ^u . The undamaged part τ^u is expressed as:

$$\tau^u = K\Delta \quad (7)$$

with K the stiffness matrix. K is a diagonal matrix with $K_{11} = K_{22} = K_{sh}$ and $K_{33} = K_n$, K_{sh} and K_n being the shear and the normal interface stiffness, respectively. The relationship between the displacement and the damaged part of the stress τ^d is defined as:

$$\tau^d = \begin{bmatrix} K_{sh} & 0 & 0 \\ 0 & K_{sh} & 0 \\ 0 & 0 & K_n \end{bmatrix} \begin{bmatrix} \Delta_1 - \Delta_1^{di} \\ \Delta_2 - \Delta_2^{di} \\ -\langle -\Delta_n \rangle \end{bmatrix} \quad (8)$$

where $\langle \cdot \rangle$ is the MacAuley bracket defined as $\langle \cdot \rangle = \frac{1}{2}(\cdot + |\cdot|)$. Hence, the damage model is only active for positive values of Δ_n . As defined in [46], Δ^{di} represents the inelastic sliding which has occurred on the damaged part of the REA. Hence, a friction function ϕ governs the evolution of Δ^{di} and is defined as:

$$\phi = \mu \tau_n^d + \tau_{sh}^d \quad (9)$$

with $\tau_{sh}^d = \sqrt{(\tau_1^d)^2 + (\tau_2^d)^2}$. The evolution of Δ^{di} then writes:

$$\dot{\Delta}^{di} = \dot{\lambda} \begin{bmatrix} \frac{\partial \phi}{\partial \tau_1^d} \\ \frac{\partial \phi}{\partial \tau_2^d} \\ 0 \end{bmatrix} = \dot{\lambda} \begin{bmatrix} \frac{\tau_1^d}{\tau_{sh}^d} \\ \frac{\tau_2^d}{\tau_{sh}^d} \\ 0 \end{bmatrix} \quad (10)$$

with λ the displacement jump. Eventually, the stress at the interface writes:

$$\tau = (1 - d) \begin{bmatrix} K_{sh} & 0 & 0 \\ 0 & K_{sh} & 0 \\ 0 & 0 & K_n \end{bmatrix} \begin{bmatrix} \Delta_1 \\ \Delta_2 \\ \Delta_n \end{bmatrix} + d \begin{bmatrix} K_{sh} & 0 & 0 \\ 0 & K_{sh} & 0 \\ 0 & 0 & K_n \end{bmatrix} \begin{bmatrix} \Delta_1 - \Delta_1^{di} \\ \Delta_2 - \Delta_2^{di} \\ -\langle -\Delta_n \rangle \end{bmatrix} \quad (11)$$

with d a damage variable which is defined as the fraction of the damaged area in the REA. The damage activation function and the evolution law are independent from the friction model and are described in detail in [47]. The activation function $F(\lambda)$ reads:

$$F(\lambda) = H(\lambda) - r_d \leq 0 \quad (12)$$

with $H(\lambda)$ a monotonic loading function of the displacement jump λ and r_d a threshold function defined as:

$$H(\lambda) = \min \left\{ \frac{\lambda - \Delta^0}{\Delta^f - \Delta^0} \right\} \quad (13)$$

$$r_d = \max(1, \max_s[H(\lambda)]) \quad 0 < s < t \quad \forall t \quad (14)$$

with t the time of the analysis and Δ^0 and Δ^f the displacement jump corresponding to the onset of damage and the final failure of the interface, respectively, under mixed-mode conditions (see Fig. 4). Mode-mixity is accounted for through the Benzeggagh and Kenane criterion [49], which is used to compute the values of Δ^0 and Δ^f . The damage variable d writes:

$$d = \frac{r_d \Delta^f}{r_d \Delta^f + (1 - r_d) r_d \Delta^0} \quad (15)$$

2.4.2. Micromechanical analyses on RVEs

Three 2D square RVEs of dimension $L \times L$ were generated using a simple script randomly positioning fibers inside the matrix, with $L = 20 \times R_f$. All the fibers were assumed to have the same radius $R_f = 5 \mu\text{m}$. Periodic boundary conditions (PBC) were applied to the edges of the RVEs. While it has already been stated that PBC affect the damage pattern once failure is initiated in the RVE [17], it was not the focus of the present work to extensively characterize this pattern. The PBC are implemented by defining kinematic constraints on the corresponding nodes of opposite edges

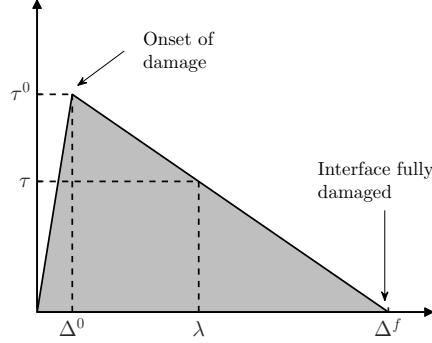


Figure 4: Schematic of the bilinear traction separation law for the cohesive elements.

of the RVEs. The equations relating the degrees of freedom of the corresponding nodes also depend on the applied strain field.

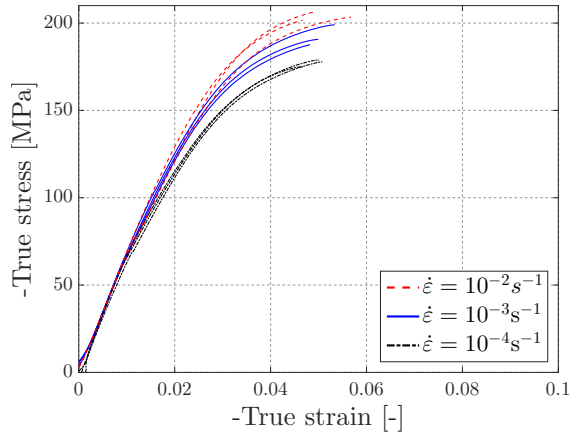
Abaqus Explicit is used instead of the Implicit framework in order to overcome convergence issues caused both by the softening in the strain-stress response of the matrix and by the progressive damage and eventual failure of the matrix and of the cohesive elements. The element size was set to be approximately $0.1 \times R_f$, R_f being the fiber radius. As generalized plane strain elements are not available in Abaqus Explicit, plane strain conditions are used. The interfaces are modelled using 4-node two-dimensional cohesive elements (COH2D4) [45]. Both the matrix and the fibers are modelled using 3-node linear elements (CPE3) [45]. The notation $\bar{\cdot}$ refers to the volume average over the elements of the model. The homogenized responses of the RVEs are shown in terms of true strain (logarithmic strain) versus true stress (Cauchy stress).

To achieve a proper comparison between the strain field predicted by RTM6 constitutive model and the strain field calculated by DIC, other models were built with the exact same geometries (fiber locations and radii) as the ones in Fig. 2a and 2b. The boundary conditions were entirely defined by the displacement field measured by DIC on the edges of zones A and B. As only the strain field at the specimen surface was measured, 2D plane stress analyses were used for these comparisons (CPS3 elements for the matrix and the fibers). In these FEA, no damage law for the interfaces and the matrix was included.

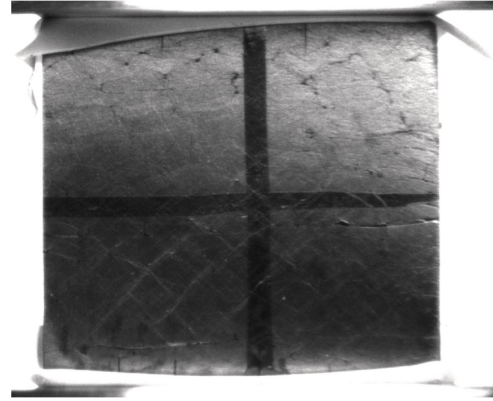
3. Transverse compression results

3.1. Experimental results

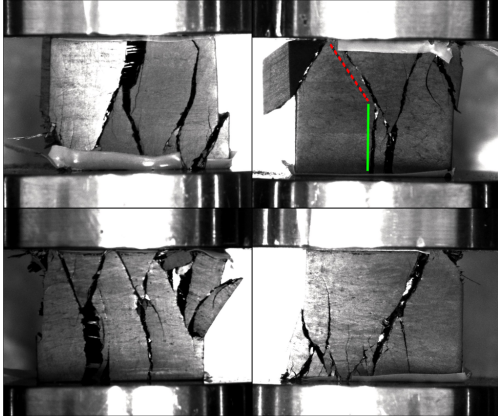
Fig. 5a shows the true stress - true strain curves resulting from the transverse compression tests on the cubic specimens of the UD composite. The significant strain-rate dependency and non-linear behaviour indicates that the matrix dominates the overall mechanical response of the composite under transverse compression. Prior to failure, localization bands were observed on the surface of several specimens, as shown in the bottom part of the specimen in Fig. 5b. These bands are the sign that strain localization is already significant at the macroscale. In terms of fracture mechanism, all specimens failed in a brittle way, without exhibiting any sign of progressive damage, except for small cracks that initiated at the contact between the specimens and the platens. Fig. 5c illustrates the failure modes observed on the specimens. The applied strain rate had no particular impact either on the brittle character of the failure or on the crack density and orientations. In general, failure occurred with the fast propagation of cracks oriented either parallel to the direction of compression or oriented in the direction of the maximum shear stress. Often, the macro-cracks parallel to the direction of loading, possibly initiated at the small cracks produced in the contact region with the platens, deviate in the maximum shear direction. This transition highlights the fact that fracture is initiated first by the presence of tensile stresses, leading to cracks parallel to the direction of loading prior to being deviated in the localization bands directions, such as the ones observed in Fig. 5b. The green and the red dashed lines on the top right image of Fig. 5c refer to the corresponding lines in Fig. 5d. Fig. 5d shows a SEM micrograph of the specimen fracture surface, exhibiting the abrupt shift from tensile stress dominated to shear dominated propagation.



(a)



(b)



(c)



(d)

Figure 5: (a) True stress true strain response of the UD cubic specimens corresponding to the transverse compression tests at three strain rates; (b) surface of a UD specimen instant before failure, the white lines in the bottom part of the specimens show the strain localization occurring at the macroscale; (c) Surfaces of UD specimens after failure, the blue and red lines in the top right image emphasize the transition between a tensile stress dominated to a shear stress dominated crack propagation, which is highlighted by the transition observed on the fracture surfaces, as shown in (d) with a representative SEM micrograph of a fracture surface from the cubic specimens.

3.2. Numerical results

The micromechanical model analyses of the UD RVEs intended to capture the experimental stress-strain response shown in Fig. 5a both in terms of non-linearity and of failure stress $\sigma_{\text{exp,f}}$ and failure strain $\varepsilon_{\text{exp,f}}$. Fig. 6a, 6b and 6c show the superimposition of the experimental curves and of the numerical predictions from the three RVEs, respectively with damage activated both in the cohesive elements (but no friction) and in the matrix (6a), damage only in the cohesive elements (but no friction - 6b) and damage only in the matrix (6c). The numerical curves were stopped when a significant drop in the homogenized axial stress of the RVE was observed. The plane strain conditions are able to properly capture the stiffness of the cubic specimens. However, the finite element model incorrectly predicts the instant of failure initiation, even when removing damage either in the cohesive elements or in the matrix.

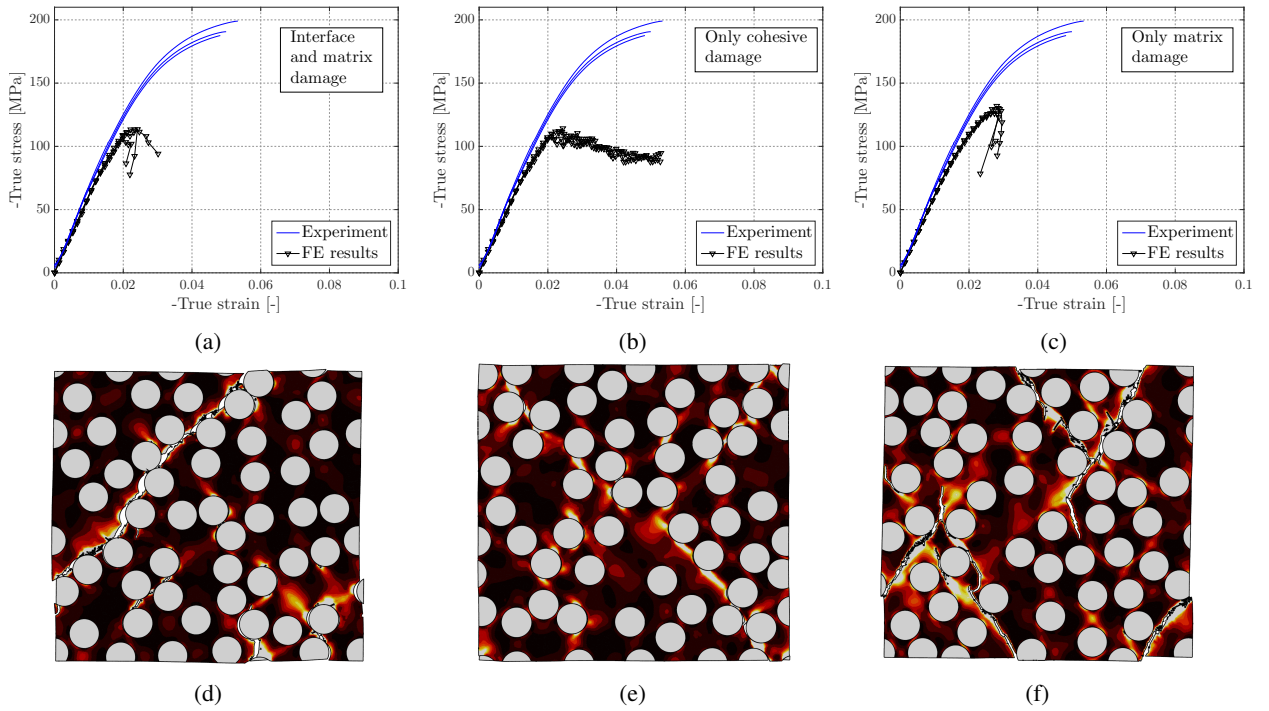


Figure 6: (a), (b) and (c) compare the experimental true stress - true strain curves from the compression tests on the cubic specimens with the FE predictions (at $\dot{\varepsilon} = 10^{-3}\text{s}^{-1}$), respectively with damage at the fiber-matrix interfaces and in the matrix (a), only at the interfaces (b) and only in the matrix (c); (d), (e) and (f) are contour plots of $\varepsilon_{\text{eq}}^p$ from FEA at the end of the simulations, corresponding respectively to the conditions in (a), (b) and (c).

Representative contour plots of the equivalent plastic strain $\varepsilon_{\text{eq}}^p$ at the instant of failure for the three conditions are shown in Fig. 6d, 6e and 6f, respectively corresponding to Fig. 6a, 6b and 6c. In all cases, the plastic strain localizes in bands oriented in the maximum shear direction, as already reported for similar conditions (e.g. [13, 17]). Damage at the interfaces is also mainly located in the regions where $\varepsilon_{\text{eq}}^p$ is high.

In order to fully assess the ability of the matrix constitutive model and hardening law identified on bulk specimens to successfully predict the mechanical response of the UD composite before fracture, the damage and failure models were deactivated, meaning that the non-linearity of the stress-strain curve is entirely associated to matrix plasticity. The broad characterization of the mechanical behaviour of RTM6 [25, 30, 50] has already highlighted the uncoupled nature of plasticity and failure in highly cross-linked epoxy resin, showing no damage growth prior to fracture in RTM6. Regarding the progressive damage at the interfaces and its influence on the mechanical response of the specimens, results of in situ tests described in section 4 show that compression dominated stress states do not seem to trigger extensive fiber-matrix decohesion prior to fracture. The comparison of the experimental results with the FEA without damage is shown in Fig. 7a, while Fig. 7b shows a contour plot of the equivalent plastic strain when $\bar{\varepsilon}_{22} = \varepsilon_{\text{exp,f}}$. The onset of non-linearity in the RVEs response is rapidly followed by a plateau in the stress-strain response of the RVEs, hindering the ability of the FEA to properly capture the experimental stress strain relationship

as further discussed later in the paper. Fig. 7b illustrates the highly localized nature of the strain field. The localization pattern is in line with previously reported micromechanical analyses on UD RVEs under transverse compression, with a preferential alignment in the maximum shear direction [14, 17, 19, 33].

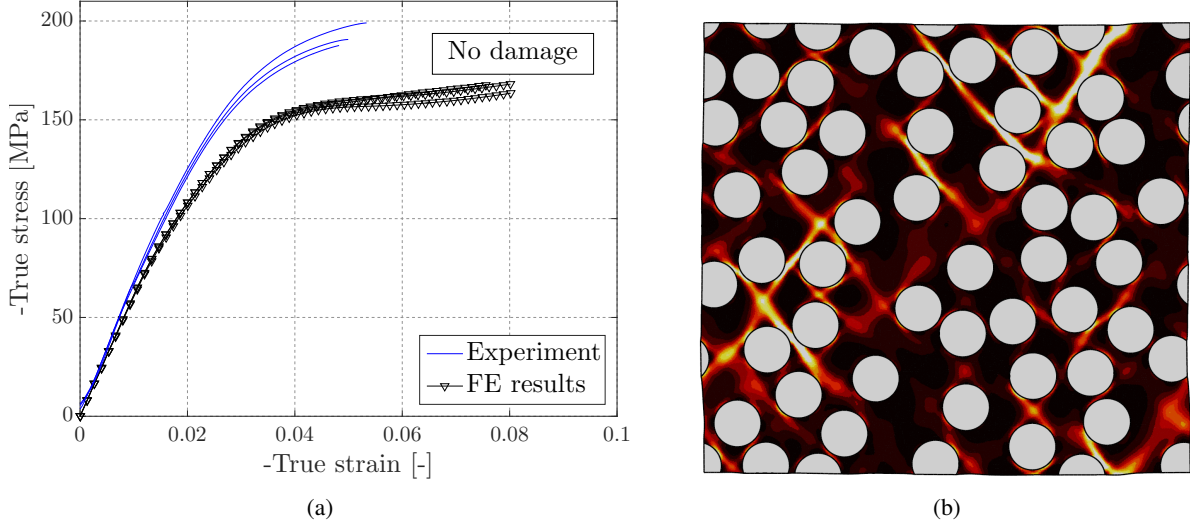


Figure 7: (a) compares the experimental true stress - true strain curves from the compression tests on the cubic specimens with the FE predictions (at $\dot{\epsilon} = 10^{-3}\text{s}^{-1}$), when no damage is included in the simulation; (b) shows a contour plot of ϵ_{eq}^p when $\bar{\epsilon}_{22}$ is equal to ϵ_f , the experimental true strain at failure.

Fig. 6b illustrates an important issue regarding the formulation of cohesive elements in Abaqus when modeling fiber-matrix interfaces in a UD composite. Indeed, as soon as a small amount of the total interface area has failed, localization bands initiate, starting from these broken interfaces, the matrix being free to flow (see Fig. 6e). As a result, only a few partially broken interfaces are sufficient to reduce the load bearing capability of the RVEs. Hence, the Abaqus formulation for the cohesive elements was replaced by a frictional cohesive law implemented in the VUMAT, as described in section 2.4.1. This new cohesive law enables the interface elements that have failed (damage variable $d = 1$) to still carry load through friction. Fig. 8a compares the homogenized response from the simulations with the Abaqus cohesive element formulation and with the frictional cohesive elements using three values of the friction coefficient μ , and this for a particular RVE. The response without any matrix or interface damage is added for the sake of comparison. Clearly, accounting for friction suppresses the abrupt stress drop once the first cohesive elements have failed. Fig. 8b shows, in parallel, the evolution of the relative number of broken cohesive elements and the evolution of the mean shear stress (in absolute value) in the cohesive elements. As illustrated in Fig. 8b, the onset of damage at the interfaces is not delayed when changing the cohesive formulation. When using Abaqus formulation, the stress drop is associated with a sharp increase in the number of broken cohesive elements. These elements mainly lie along the localization bands triggered by the failure of the first cohesive elements. The frictional cohesive elements prevent this localization as the shear stresses still carried by the interfaces partly impede the free flow of the matrix in the vicinity of the broken elements. Fig. 8c and Fig. 8d illustrate the change in localization pattern by showing contour plots of ϵ_{eq}^p after an applied axial strain $\bar{\epsilon}_{22} = -0.035$ for the Abaqus formulation and for the frictional cohesive elements with $\mu = 0.5$, respectively.

4. In situ characterization

4.1. DIC measurements

In an attempt to understand why the numerical model is not capable of capturing the overall stress-strain experimental response, in situ tests on UD specimens originating from the same slab as the cubic specimens used in section 3 were performed in a SEM chamber. The goal was to obtain quantitative information about the strain fields in the

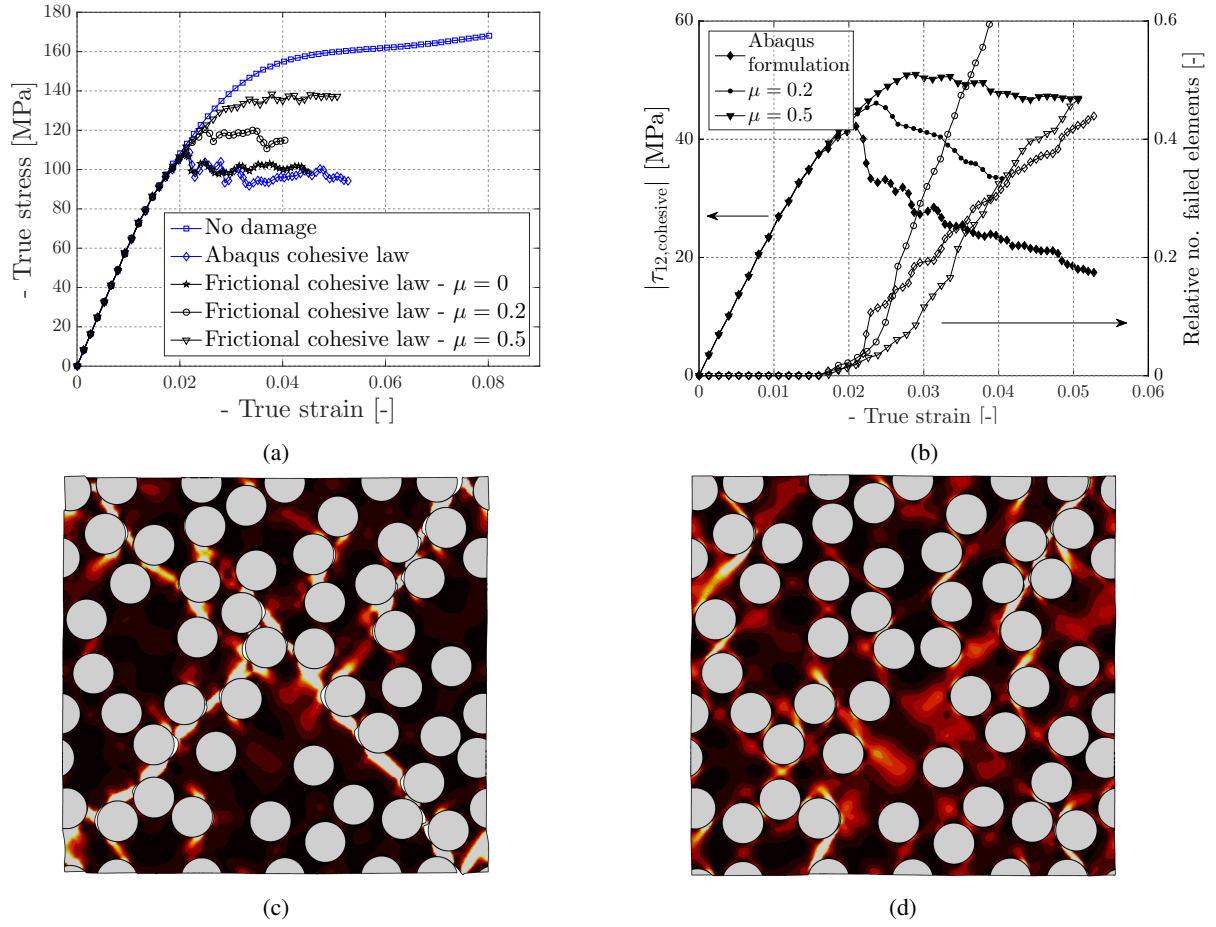


Figure 8: (a) True stress - true strain homogenized response of a UD RVE without any damage (squares) and with damage only at the interfaces using the Abaqus formulation (diamonds) and using the frictional cohesive elements with $\mu = 0$ (stars), $\mu = 0.2$ (circles) and $\mu = 0.5$ (triangles); (b) Evolution of both the mean shear stress carried by the cohesive elements $\tau_{12, \text{cohesive}}$ (filled symbols) and of the relative number of failed elements ($d = 1$) in the model (empty symbols), using Abaqus formulation (diamonds) and frictional cohesive elements with $\mu = 0.2$ (circles) and $\mu = 0.5$ (triangles); (c) and (d) are contour plots of $\varepsilon_{\text{eq}}^p$ at $\bar{\varepsilon}_{22} = -0.035$ with the Abaqus formulation for the cohesive elements and the frictional cohesive elements, respectively.

matrix at the scale of the fibers, in order to enable a direct comparison with the numerical results using the constitutive model for the RTM6 epoxy resin. For this purpose, digital image correlation was used to determine the strain field at the surface of the specimens. DIC has already been proven to be an adequate tool to measure the strain field at the microscopic scale in FRP [36, 37].

The in situ tests revealed the absence of progressive damage either from decohesion at the interfaces or from microcracking in the matrix, both in regions A and B. A post-mortem survey of the specimens surface proved this observation to be true for the majority of the specimens but also indicated several plastic strain localization areas around which decohesion occurred (see Fig. 9a). In both tests presented in this section, failure was initiated suddenly and near the locus of maximum shear stress (close to region B), as shown in Fig. 9b.

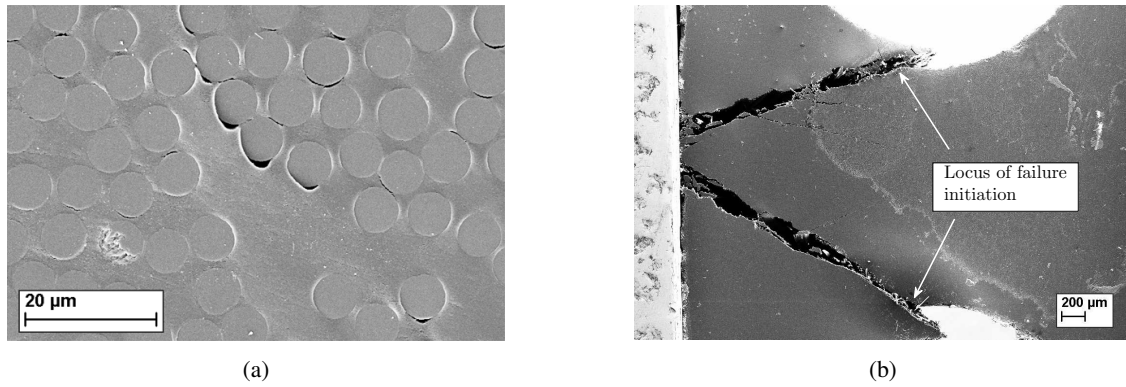


Figure 9: (a) SEM micrograph at the end of an in situ test showing significant fiber-matrix decohesion along a matrix pocket in which strain has localized; (b) SEM micrograph of an in situ specimen after failure; the arrows show the loci of failure initiation, corresponding to the sites of maximum shear stress.

In the remainder of this section, the focus will be on the last displacement and strain fields recorded before the final fracture of the specimens. Fig. 10a and 10b show the displacement fields in region A u_x and u_y calculated by DIC at the end of the test, x and y being respectively the horizontal and the vertical directions. The u_y field already highlights the relative displacement of the matrix with respect to the fibers, as pointed out by arrow 1. Fig. 10c illustrates further the ability of the DIC measurements to properly differentiate the fiber and the matrix phases, showing that the ϵ_{yy} field in region A is mainly localized in the matrix. Overall, the strain field is well distributed in the entire matrix and does not reveal any strong strain localization in this region at the end of the test.

Fig. 10d to 10f show the same results for the region B. On the one hand, the u_y field again highlights that in groups of close fibers, the fibers displacement does not reveal any detectable relative movement with respect to one another (e.g. see arrow 3). Conversely, the zones denoted by the arrows 4 and 5 reveal significant variations in the amplitude of u_x and u_y between adjacent fibers, indicating severe shear strain localization within the thin matrix regions. Fig. 10f shows the ϵ_{xx} field for region B. The first direct observation is the presence of DIC artefacts, in particular in the region indicated by arrow 6, which is caused by the presence of a large alumina particle. As in Fig. 10c, the DIC properly differentiates the matrix and fiber phases, with the strain being mainly concentrated in the matrix. Moreover, the zones of high stress concentration around the fibers are well captured by the DIC, as demonstrated by the higher strain amplitudes calculated in the zones indicated by the arrows 7. Regarding zones 4 and 5, the strain field indeed involves significant strain localization in the small matrix domains between these adjacent fibers.

The ability of standard DIC methods to accurately determine the strain field in confined regions of multiphase or composite materials has already been questioned [36]. Here, the strain field in the zone denoted by arrow 2 seems to be well captured by DIC, even though uncertainty remains regarding the exact strain amplitude. Still, the arrow 6 points toward another zone involving close adjacent fibers where DIC seems to catch relative displacement between adjacent fibers but is unable to locate the strain in the matrix region, while the computed strain level can also be questioned. Fig. 11 provides zooms within zones 3 and 8 from region B. Fig. 11a and Fig. 11d show the initial undeformed state; Fig. 11b and Fig. 11e show the same zones at the end of the test; and Fig. 11c and Fig. 11f consist in the ϵ_{xx} field measured by DIC. The red circles in Fig. 11b and Fig. 11e highlight very confined matrix domains between close fibers in which thin bands whitening progressively appeared during the test. These bands are the sign

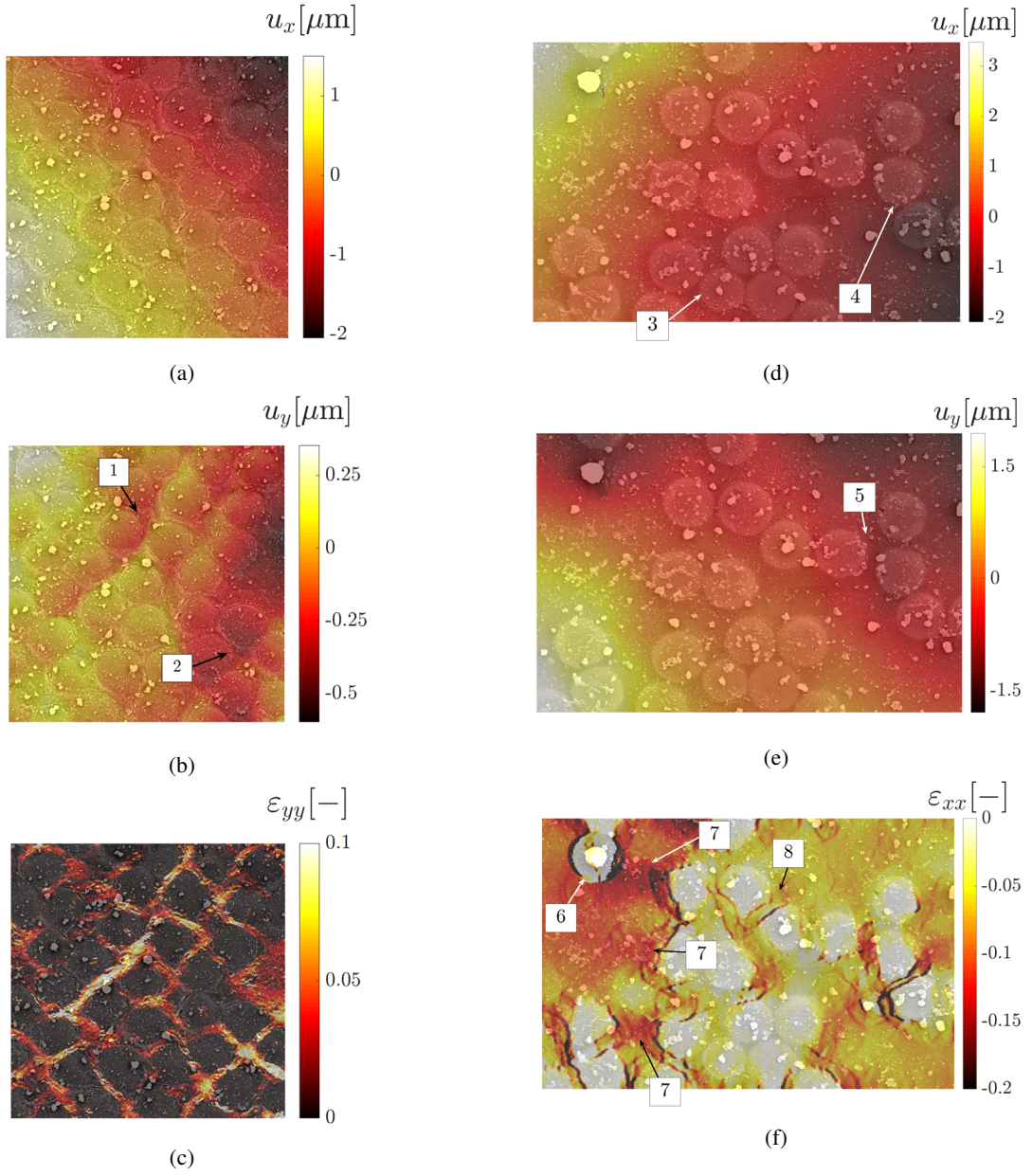


Figure 10: Results from the DIC measurements on regions A - (a), (b) and (c) - and B - (d), (e) and (f); (a) and (d) show the u_x displacement field, (b) and (e) show the u_y displacement field, (c) shows the ε_{yy} strain field for region A and (f) the ε_{xx} strain field for region B, with x and y being respectively the horizontal and the vertical directions.

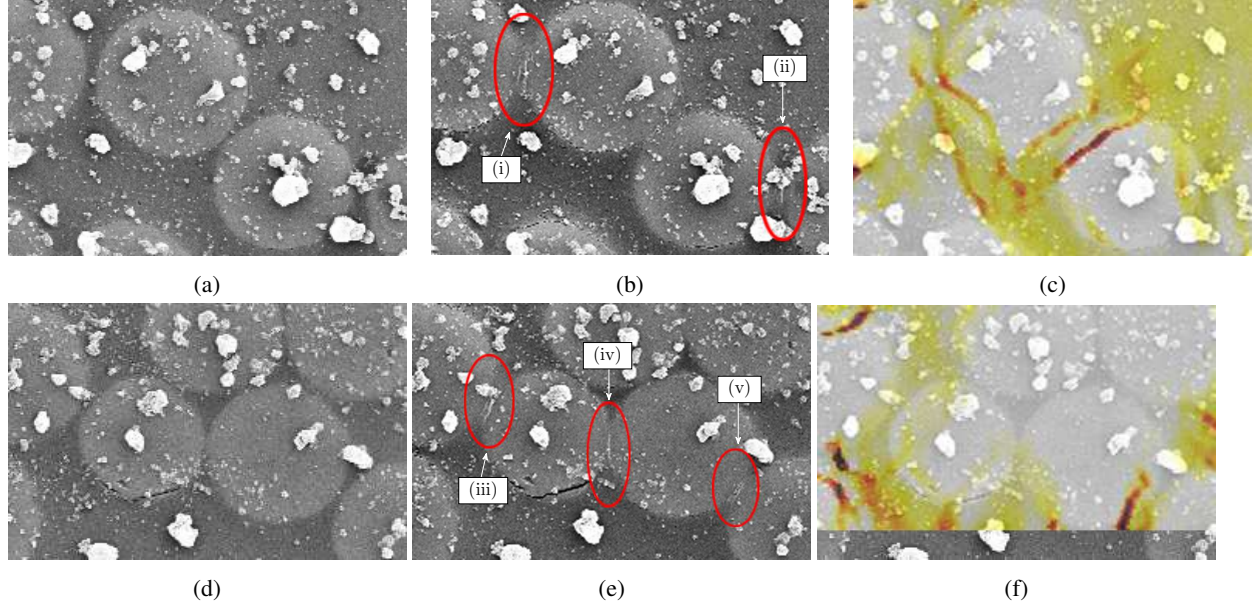


Figure 11: Zoom on the zones denoted by the arrows 3 - (a), (b) and (c) - and 8 - (d), (e) and (f) - in Fig. 10d and 10f; (a) and (d) show the zones in the undeformed state, (b) and (e) just prior to failure, with the red circles highlighting the confined matrix regions in which whitened bands are observed, and (c) and (f) show the corresponding ε_{xx} field measured by DIC (the colorscale is identical as in Fig. 10f).

of extensive strain accumulation in the narrow matrix regions. For the circles (i) and (v), the DIC strain field shows that these highly deformed regions are identified by the software but doubts can legitimately be raised on the accuracy of these measurements. Additionally, for the circles (ii), (iii) and (iv), the DIC strain field does not show any sign of higher strain amplitude in the whitened regions. Eventually, this emphasizes the limitations of the DIC measurements in very small regions, especially in bi-constituents materials where strain estimation in one constituent can be affected by the nearby presence of the other one.

4.2. Comparison with FEA

Fig. 12 shows the displacement and strain fields predicted by FEA, without any damage in the matrix and at the interfaces, using the geometries from regions A (Fig. 12a to 12c) and B (Fig. 12d to 12f). The boundary conditions consisted in imposing the displacements calculated by DIC on the edges of the zones. Fig. 12 must be directly compared with Fig. 10 which presents the same information as obtained by DIC. In both cases, the color code is identical. The displacement fields u_x and u_y are very similar between the DIC analyses and the FEA. Obviously, as the displacement is imposed on the boundaries of the FE model by the displacement on the edges of the DIC regions, the displacement field must at least tend to be identical as getting closer and closer to the edges. Still, the agreement is also satisfactory in the inner part of the FE model. In particular, the zones indicated by arrows 1, 2, 3 and 5 in Fig. 10 exhibit a strong resemblance in both cases.

More significant discrepancies appear when comparing Fig. 12c and Fig. 10c. While the displacement fields seem to correspond when comparing DIC and FEA in zones 1 and 2, the strain amplitudes noticeably differ. More generally, in region A, the strain was not accumulated in the same material regions in the FE model as what is seen experimentally by DIC. Because of the DIC limitations mentioned in section 4.1, region A is more prone to inaccurate DIC measurements, which is related to the high fiber density and the large portion of matrix confined between close fibers. Still, the differences between FEA and DIC strain fields look more related to a difference in terms of the regions where strain accumulates, rather than caused by a lack of precise DIC calculations in the small matrix regions where the FEA predicts higher strain values. Region B offers a different configuration with a much larger fraction of matrix pockets. Globally, both strain distributions are in good agreement. The strain field in the zones indicated by the arrow 7 in Fig. 10f is relatively well captured by the FE model. However, the FEA contour map in Fig. 12f shows a very

localized strain field, especially in the large matrix pocket on the left side, including close to the fibers surrounding the pocket. The DIC analysis was unable to properly capture the highly deformed regions between very close fibers in zones pointed by arrows 3 and 8 and shown in more detail in Fig. 11. Reversely, the FEA results actually predict high compressive strain in these regions in agreement with the visual evidences in Fig. 11b and 11e. Overall, the material regions involving high stress concentration are properly modelled by FEA but the main differences between DIC and FEA lies in the width of the localization bands. These bands either propagate further away from the stress concentration zones or appear in material regions where DIC shows a more homogeneous strain field (e.g. in the matrix pocket on the left side of region B). In other words, DIC results do not show the same intensity of strain localization and the experimental deformation seems to accumulate in the high stress concentration zones, without necessarily involving strongly localized patterns.

A more quantitative comparison between the DIC and FEA strain fields is obtained by superimposing the amplitudes ε_{xx} along cross-sections of region B. Region B is selected to minimize the impact of the lack of accuracy of DIC in confined zones and the focus is on the larger matrix pockets, as shown by the profiles selected in Fig. 13a. Fig. 13b is a contour plot of the ε_{xx} field from FEA in which the matrix was modelled as linear elastic. It is used as a benchmark in the comparison with the DIC results. Fig. 13c to 13f superimpose the strain profiles along the cross-sections displayed in Fig. 13a. The general trend is that, globally, the strain profiles are in a relatively good agreement in all three cases. In Fig. 13c and 13d, the ellipses A, B, D, E and F highlight zones of particular interest. In these zones, a fully elastic FEA surprisingly agrees better to the DIC measurements than the predictions obtained using RTM6 viscoplastic constitutive model. In the latter, strain localization generates strong oscillations in the strain field along these profiles, seemingly over-predicting the actual strain amplitude in the zones of high stress concentration. This observation is in line with the analysis of the localization bands addressed earlier in this section indicating that FEA overestimates the plastic localization amplitude. In Fig. 13e, besides the sign of the DIC artefact indicated by arrow 6 in Fig. 10f, the three strain profiles identically capture the zones of higher stress concentration within ellipse G and between the two fibers. However, in the large matrix pocket, the ellipse H surrounds another zone where the viscoplastic model predicts higher strain fluctuations than in the profiles obtained by DIC and by the elastic simulation. Again, the latter two are in good agreement in this zone. The zone under ellipse I in Fig. 13f reveals the same excessive strain localization in the simulation with the viscoplastic model. Reversely, in zones C in Fig. 13c and K in Fig. 13f, the agreement is much better with a viscoplastic setting compared to simulation.

Overall, despite sparse discrepancies, the strain field obtained using an elastic matrix seems to be in better agreement with the DIC measurements than the predictions using the viscoplastic model.

5. Nanoindentation in resin pockets

The nanoindentation tests in resin pockets of the UD composite were performed in order to characterize the in situ mechanical behaviour of the matrix. Rodriguez et al. [51] developed a method to determine the equivalent yield stress $\sigma_{y,NI}$ of the resin by means of nanoindentation for amorphous materials with cohesive-frictional behaviour. They performed an extensive numerical study to observe the influence of the parameters of a linear Drucker-Prager law on the nanoindentation response of the model material, using a Berkovich indenter. The method relies on estimating the ratio between the elastic energy W_e and the total energy $W_t = W_e + W_p$, with W_p the energy dissipated by plastic deformation. Without going into the details of the methods, a ratio $W_e/W_t > 0.5$ ensures that Oliver and Pharr method [44] provides a good estimate of the contact area. In that case, $\sigma_{y,NI}$ can be determined if the friction angle β of the Drucker-Prager law is known a priori, and vice versa.

As this method does not enable to recover the full stress-strain response of the material, similar tests were conducted into a bulk RTM6 resin specimen that were used to directly compare $\sigma_{y,NI}$ corresponding the UD RTM6 matrix and to the bulk behaviour. As size effects during nanoindentation tests have been reported in epoxy resins, marked by a progressive decrease of hardness with indentation depth, the CSM mode was used to ensure that the beginning of the hardness plateau had been reached far from the end of the tests. Fig. 14a shows the load-displacement curves from the nanoindentation tests both in the bulk RTM6 specimens and in the RTM6 matrix pockets. Fig. 14b shows representative indents in resin pockets. The load-displacement curves indicate the apparently harder behavior of the resin in the UD composite with respect to the pure resin specimen, but also the good reproducibility of the matrix response in the composite. Fig. 14c shows the values of $\sigma_{y,NI}$ obtained using the method described in [51], assuming $\beta = 7.86$ as for the bulk specimens of RTM6. Regarding the mean value obtained for the bulk resin specimen, $\sigma_{y,NI,bulk}$ is very close

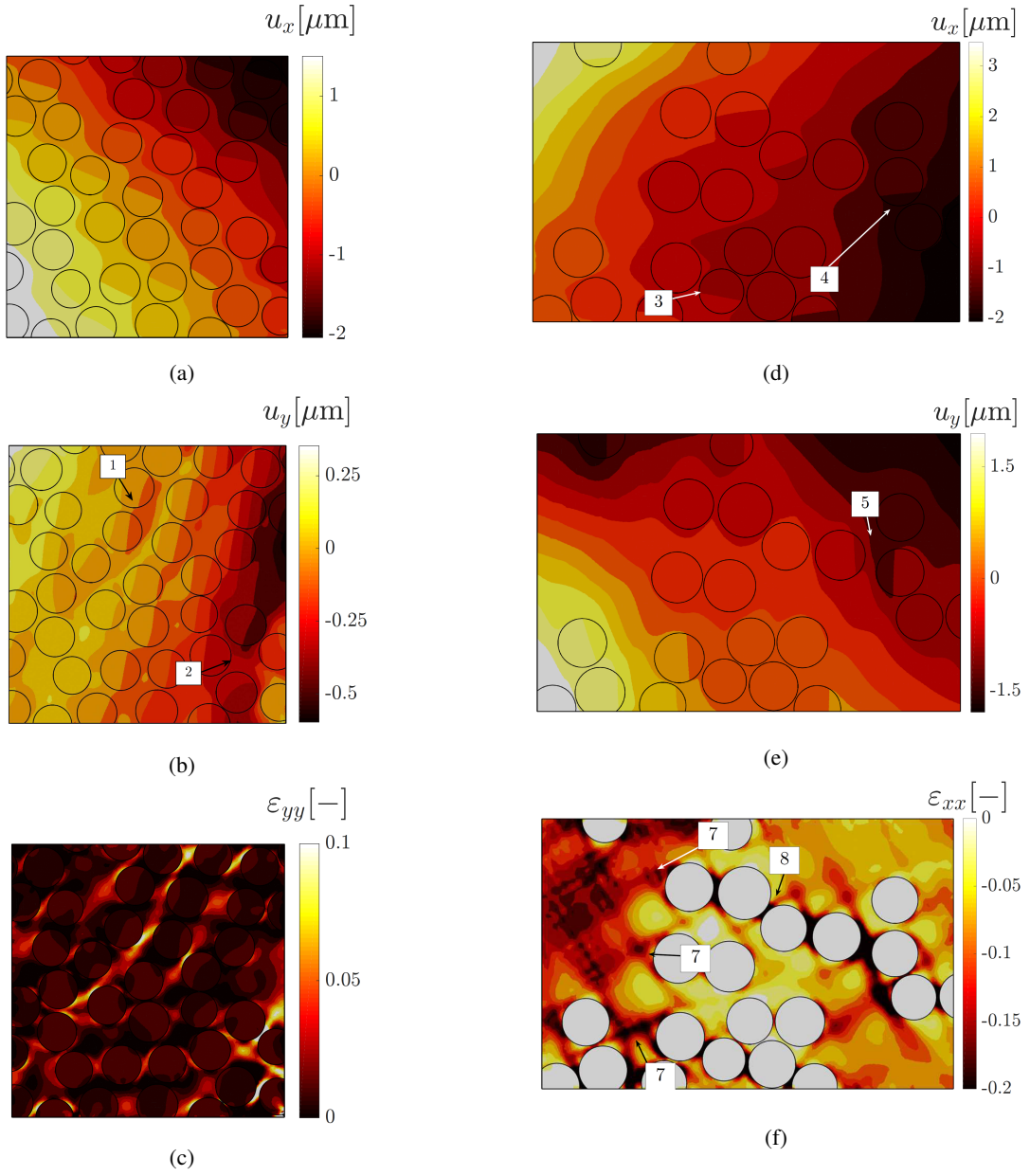


Figure 12: Results from FEA in regions A - (a), (b) and (c) - and B - (d), (e) and (f); (a) and (d) show the u_x displacement field, (b) and (e) show the u_y displacement field, (c) shows the ϵ_{yy} strain field for region A and (f) the ϵ_{xx} strain field for region B, with x and y being respectively the horizontal and the vertical directions.

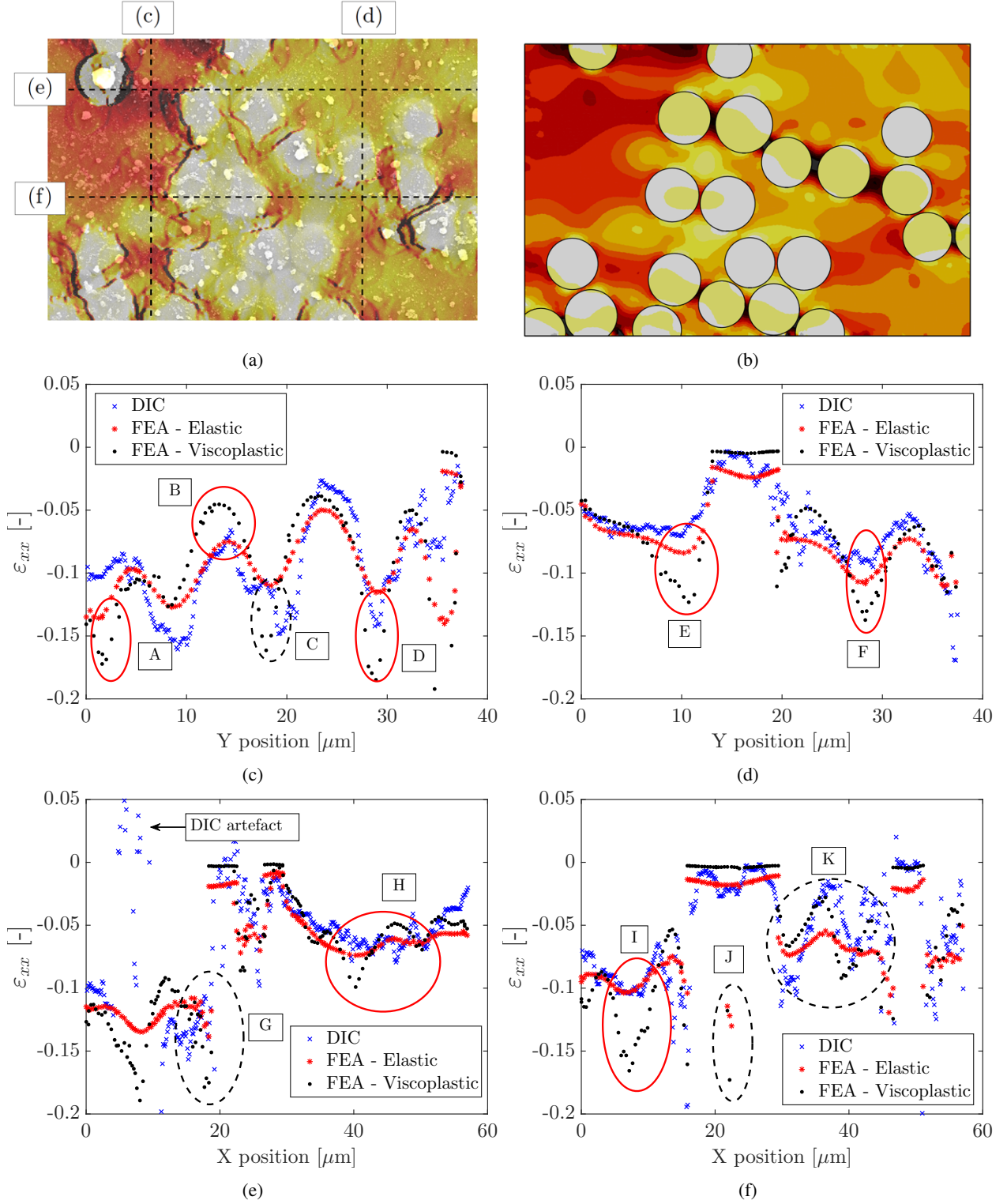


Figure 13: (a) and (b) represent the ε_{xx} strain field in region B, respectively measured by DIC and obtained by FEA with a purely elastic matrix (the colorscale is identical as in Fig. 12f); (c), (d), (e) and (f) compare ε_{xx} profiles along the path defined by the dashed lines in (a) and obtained by DIC (blue crosses), by FEA with RTM6 constitutive model for the matrix (black dots) and by FEA with a purely elastic matrix (red stars); in (c) to (f), the red ellipses highlight zones where the elastic matrix is in better agreement with the experiments while the black dashed lines circle the zones where the viscoplastic model provides better results.

to the maximum stress attained prior to the softening at a similar strain rate in uniaxial macroscopic compression tests - the mean strain rate under the indenter can be estimated by $\dot{F}/2F$ [52] - see Fig. 3a. The mean value of $\sigma_{y,NI,UD}$ for the 1 μm penetration depth in the UD composite is slightly higher than for the 2 μm penetration. This can be explained by the larger influence of the higher hardness at low indentation depths that is observed in nanoindentation tests on the ratio W_e/W_t . The apparent harder response of the UD matrix compared to the bulk resin behaviour noticed in Fig. 14a is indeed translated by a large increase in $\sigma_{y,NI}$.

The ratio $r_y = \sigma_{y,NI,UD}/\sigma_{y,NI,bulk}$ was used as a correcting factor to empirically modify the hardening law of the RTM6 resin. Hence, as a first attempt to correct the behaviour of RTM6 in the UD composite, the flow stress was multiplied by r_y . Fig. 14d compares the experimental results on the cubic UD specimens to the predicted homogenized responses with and without the corrected hardening law. The match between the experiments and the simulations becomes almost perfect when using the hardening law corrected by the ratio $\sigma_{y,NI,UD}/\sigma_{y,NI,bulk}$. Fig. 14e and 14f show comparisons of strain profiles ϵ_{xx} - along the same paths as in Fig. 13c and 13d - obtained from DIC and FEA, using the initial and the corrected hardening laws. While the new hardening laws provide a good correction for the homogenized response of the RVEs, the differences remain between the FEA and the DIC results.

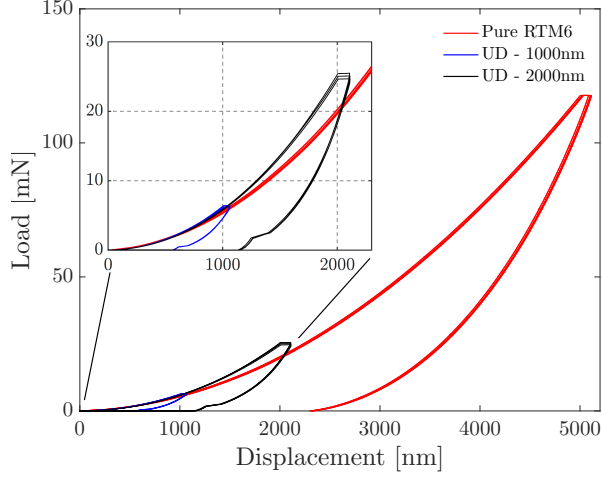
6. Discussion

The direct comparison between the experimental results from the transverse compression tests on the cubic specimens and the FE predictions on RVEs using hardening laws obtained from bulk specimens has highlighted the inability of such approach to properly reproduce the overall experimental stress strain relationship. The error resulting from the FE analyses are related both to shortcomings in the damage models used for the matrix and for the interfaces (Fig. 6) and to the hardening law identified for the RTM6 resin and possibly the constitutive model itself (Fig. 7).

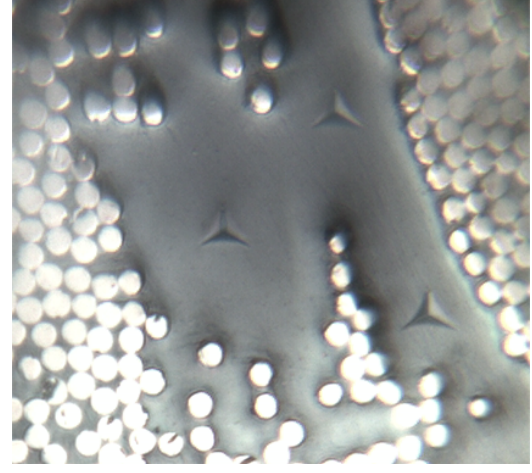
When damage is included, both matrix fracture and cohesive decohesions are heavily connected to regions of maximum equivalent plastic strain (see Fig. 6d, 6e and 6f). Hence, excessive strain localization in the FEA leads to overly strong non-linearity in the overall stress strain response (see below) and to premature failure compared to experiments. For instance, Fig. 6e shows that only a small number of partially failed interfaces along the main localization bands is sufficient to significantly decrease the load bearing capability of the RVEs. Previous micromechanical studies have demonstrated the strong dependence of UD transverse mechanical properties on the strength of the interfaces (e.g. [10, 14, 31]). The widespread use of similar cohesive properties to model fiber-matrix interfaces and the results from Fig. 6e question the accuracy of the interfacial failure predictions in the literature. In an attempt to better capture the progressive damage at the interfaces, a cohesive law enabling failed elements to still carry some load through friction was used. Fig. 8 showed the significant change both in mechanical response and strain distribution in the UD RVEs by modifying the cohesive element formulation. Clearly, including friction improved the predictions by suppressing the sudden stress drop following the first interface failures. Friction hinders the development of localization around the first broken elements. Eventually, the contribution of friction must be taken into account to simulate the progressive damage evolution at the interfaces. Still, even when accounting for friction, the cohesive properties used in the literature seem too weak to properly capture the failure stress of the UD composites. It thus underlines the need of a finer characterization of the interfaces strength to provide reliable cohesive properties.

Regarding the results from Fig. 6f, the early failure of the RVEs associated to matrix damage is in agreement with the common understanding of fracture mechanisms in epoxy resins. As highlighted by studies on the fracture of highly cross-linked epoxy resins [50, 53, 54], the macroscopic stress levels sustainable by a pure matrix is limited by the stress concentrating effect of inherent defects within the thermoset. Hence, when using a failure criterion identified at the macro-scale to predict fracture initiation in FEA at the scale of the fibers, the hypothesis is made that every element of the FE model contains as many critical defects as in a much larger bulk matrix volume element, inevitably leading to earlier failure. No defect similar to the critical ones observed in bulk RTM6 [55] was observed in the matrix of the in situ specimens. Hence, in this case, the critical defects are more likely to be the fibers or the interfaces themselves and a more intrinsic fracture criterion with statistical effects about the probability of the presence of internal defects should be used to predict matrix failure at the microscale (e.g. [50]).

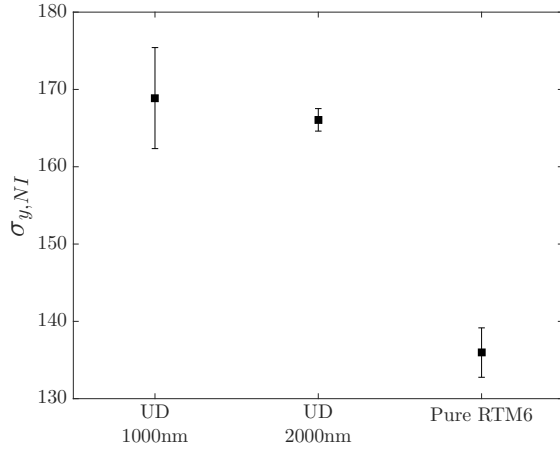
Without damage, the comparison with the DIC strain measurements shows that the FEA lead to excessively strong strain localization in the matrix when based on the viscoplastic model behaviour, while a purely elastic model is sometimes sufficient to get a good agreement with DIC results. At this point, it is worth reminding that the constitutive model and hardening law used here for the description of the RTM6 behaviour have been identified and validated over



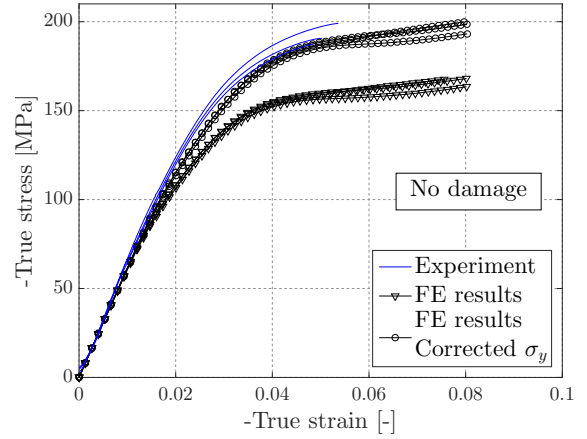
(a)



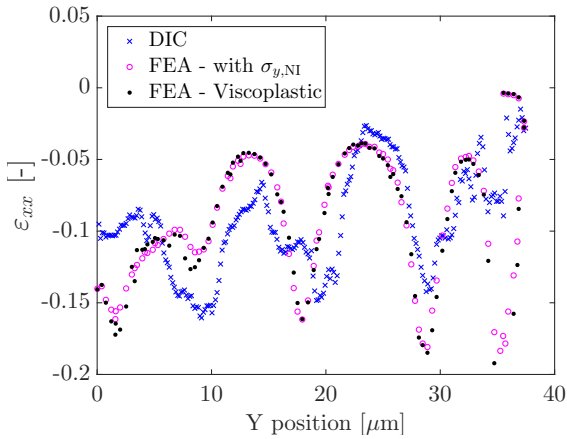
(b)



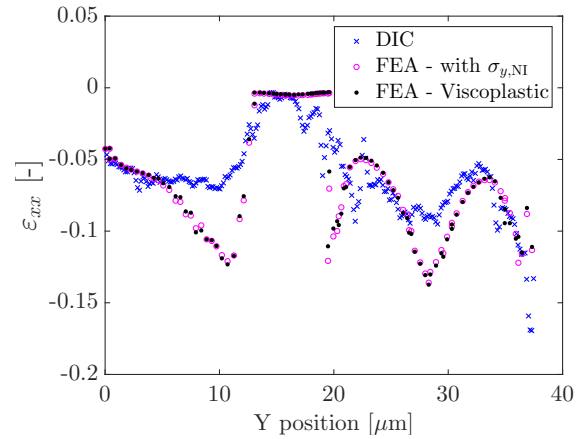
(c)



(d)



(e)



(f)

Figure 14: (a) shows the load-displacement curves from the nanoindentation tests in the matrix pockets of the UD composite and in the pure resin specimens; (b) is an optical micrograph showing representative indents in the matrix pockets; (c) shows the yield stress estimated using the method from [51] for the UD and the pure resin specimens; (d) shows the homogenized true stress true strain behavior of the UD RVEs compared to the experimental results at $\dot{\epsilon} = 10^{-3} \text{s}^{-1}$, both with the initial and the corrected hardening laws and no damage included in the FE model; (e) and (f) compare the strain profiles along two paths in region B ((c) and (d) in Fig. 13a) obtained from DIC and FEA, both with the initial and the corrected hardening laws.

a very large set of macroscopic tests, providing on bulk samples very precise predictions of the overall behaviour under a variety of loading conditions [30]. Hence, the excessively soft behavior observed in FE results can be attributed to the major strain localization triggered in the RVEs in the early stages of the simulations. Indeed, as soon as a material region enters the strain softening regime of the hardening law, the strain immediately localizes in that region, initiating the propagation of a localization band. Fig. 7b indicates the extent of such localization when no damage law is included in the simulation. In this particular case, in the homogenized response of the RVEs, the onset of significant non-linearity directly reflects the initiation and progressive propagation of these bands. In the literature, even though the experimental validation of computational micromechanics in transverse compression is rare, at least one group was able to obtain a good agreement between FEA on RVEs and experimental results [56]. Ultimately, the predictive capability of this type of numerical analysis with matrix constitutive models validated at the macroscale will strongly vary from one epoxy system to another. In [56], both the yield stress and the Young's modulus of the matrix are significantly higher than for RTM6. Hence, the stress strain non-linearity remains limited before the final failure, even in compression. The question is thus why, while the behaviour of the epoxy resin is well captured in bulk samples, it does not work well when applied in micron size pockets.

In general, both chemical shrinkage and thermal residual stresses are mentioned as possible sources of discrepancies. In this regard, modelling strategies were developed to quantify the amplitude of the residual stresses and to assess their impact on the mechanical response of CFRP, often relying on micromechanical analyses on RVEs [13, 31, 57–62]. Eventually, the transverse failure strain and stress are shown to either increase or decrease, respectively under tension or under compression, because of the thermal and/or chemical residual stress field. In addition, the onset of non-linearity is triggered earlier under compression when accounting for the residual stresses. Hence, accounting for this initial stress field would not help to bridge the gap between the experimental and the numerical results in Fig. 7a.

The correction of RTM6 hardening law using nanoindentation tests in the UD resin pockets and in the pure resin specimen lead to almost perfect match between the RVEs homogenized response and the experimental results on the cubic compression specimens. While this solution seems attractive to obtain the in situ properties of the matrix, it does not provide a full answer to the more fundamental questions that arise from this study (see below). Multiplying the whole hardening law by r_y to correct the matrix model is an arbitrary choice that is not guided by a complete understanding of the physics beyond this change in properties. Furthermore, this correction still fails with respect to limiting excessive localization at the scale of the fibers. Eventually, the issue may be that matrix properties are modified homogeneously in the whole composite, which impedes the corrected model ability to capture the microscale strain field.

The more fundamental question raised with these results is thus to what extent macroscopic constitutive models can be used in condition of confined microscale volume of materials involving large strain gradients. The mechanical response of glassy polymers has been the subject of many studies which attempted to unravel the physical origin of plastic flow. For instance, it is now admitted that softening is the consequence of strain localization under the form of micro shear bands in the material. The scale at which this localization occurs is still open to debate and probably depends on the macromolecular characteristics of each polymer. Still, regardless of the exact scale, in composite materials the matrix is often confined into submicron size regions between close fibers. Under loading, the strain gradients developing with respect to the size of the regions becomes extremely large. It is fair to question whether the matrix can accommodate this gradient using all the relaxation mechanisms that explain its macroscopic mechanical behavior. Size effects in glassy polymers have already been reported either from compression tests on micropillars [39] or from nanoindentation tests (e.g. [63–66]), involving the results obtained here on RTM6 where hardness decreases with indentation depth. In nanoindentation, the decrease of the size effect (i.e. lesser influence of the indentation depth on the hardness) when the cross-linking percentage decreases indeed indicates that it may be related to the mobility of the polymer chains [63]. Moreover, the larger size effects observed under pyramidal tip than under spherical tips points toward a direct influence of the strain gradient on the response of the glassy polymers [65]. Additionally, the presence of the fibers strongly modifies the curing conditions in these confined regions, questioning further the actual properties of the matrix in zones of high stress concentration and thus prone to trigger strain localization according to FEA. It is impossible at this stage to deconvolute these two possible sources of discrepancies: true size dependent plasticity and intrinsically different material properties between the fibers. It is the purpose of future developments of the present work to address the present experimental data with a more physical material description based on the so-called shear transformation zone theory, and investigate if size dependent plasticity effects can be captured [67].

7. Conclusion

Computational micromechanics in composite materials heavily rely on the accuracy of the models of the constituents. In this study, a complete multiscale characterization of a UD composite is performed in order to assess the validity of commonly used models for the interfaces and the matrix failure and viscoplastic behavior. This assessment was based on a direct comparison between experimental results and numerical predictions. In this regard, a thick RTM processed long carbon fiber UD slab was manufactured. Two types of tests were performed on the UD: macroscopic transverse compression tests on cubic specimens and in situ transverse compression tests on notched specimens in a SEM chamber. The former was primarily used for a direct assessment towards numerical simulations on RVEs with random fiber distribution, while the latter enabled to quantify the differences between the strain field predicted by FEA and the actual strain field at the specimen surface, measured by DIC. The main findings can be summarized as follows:

- FEA on UD RVEs using a validated constitutive model and hardening law for the bulk matrix behaviour and properties measured in bulk specimens highlight excessive strain localization predictions lead to an overestimation of the non-linearity in the stress strain response compared to the experimental response.
- Damage models for the fiber-matrix interfaces and for the matrix both predict a premature failure of the RVEs, which is related to excessive strain localization in the matrix, to the lack of reliable interface properties characterization and to the use of a macroscopic failure model for the matrix.
- The use of frictional cohesive elements to account for the load carried by friction at the interfaces that have failed has been shown to significantly improve the numerical predictions by suppressing the sudden drop in load bearing capacity of the RVEs after the failure of the first cohesive elements.
- In situ tests can be used to unravel, with the help of DIC, the strain distribution at the microscale, despite limitations related to the DIC ability to capture accurately the strain amplitude in highly confined material regions.
- The agreement between the DIC strain fields and the corresponding FEA predictions is better when the matrix is considered as linear elastic than when the constitutive model identified for the bulk RTM6 specimens is used, the latter showing excessive strain localization.
- Correcting the hardening laws of RTM6 based on the comparison between nanoindentation results in pure resin and in matrix pockets of the UD composite lead to a perfect match between the RVE and the cubic specimens mechanical responses but proved to be insufficient to properly capture all the details of the local strain fields.

Overall, through a multiscale analysis, this study highlights the need to validate experimentally computational micromechanics analyses and the differences existing between the failure and deformation behavior in the matrix at the micro- and macroscale. In loading configurations that favor early damage accumulation (e.g. transverse tension), the predictions will essentially rely on accurate models for the failure of both the interfaces and the matrix. Conversely, in other cases (e.g. transverse compression), it was shown that progressive damage accumulation at the interfaces and matrix cracking are negligible and the response is mainly governed by the amount of plasticity which develops in the matrix. Given the discrepancies between FE predictions, macroscopic experimental tests and DIC measurements at the microscale, the assumption to consider the matrix macroscopic behavior as identical to the microscopic one is questionable, at least for the epoxy system under consideration. Physically, the reason beyond these differences is still not unravelled and can be linked to different intrinsic or extrinsic factors: modification of the curing process in the vicinity of the fibers, size effect or strain gradient influence in confined regions.

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