LOCAL STRAIN RATE EFFECT ON DAMAGE IN GLASS FIBER REINFORCED ETHYLENE-PROPYLENE COMPOSITE

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1 General Introduction

Glass fiber reinforced polymers generally exhibit a diffuse damage during loading leading. This leads then to improved properties of energy absorption during impact and crash. However, several previous works showed that damage threshold and kinetic are widely sensitive to the strain rate effect [1-2]. The strain-rate effects have been extensively investigated on in-plane shear behavior mainly for thermoset matrix composites [3-6]. In a relevant work by Jerabek et al. [7] on particulate filled PP composites, the authors evidenced a modification in the dominant micro-mechanism that depends on the strain rate and the temperature. It can occur as a local cavitation mechanism and particle debonding for high strain rate and room temperature, or as a more significant contribution of local shear yielding for low strain rate or high temperature. Reiss et al. [8] reported that the glass-fiber reinforced polymer (GFRP) laminates exhibited strain rate and temperature sensitivities. The authors analyzed the coupled effects of strain rate and temperature under tensile tests. These authors showed that strain rate and temperature affect the ultimate strengths and the elastic modulus. Their effects appear as an elastoviscoplastic behavior of the GFRP.

In a previous paper [1], an experimental methodology devoted to investigate strain-rate effects on overall composite behavior for moderate rates (up to 200s\(^{-1}\)) has been proposed and optimized. Interrupted tensile tests were also optimized to study the influence of damage on the mechanical properties under high strain rate. This experimental methodology is successfully was applied to emphasize the strain rate effects on the microscopic and macroscopic behavior of SMC-R26, carbon/epoxy woven laminate composites and glass-polyester long fiber composites [1, 9]. The increase of the overall properties of reinforced thermoplastic composites observed for higher strain-rate is described as positive strain-rate dependence [10-11].

In the present work, a multi-scale experimental approach is developed to investigate the origin of the strain rate effect on the mechanical behavior of a discontinuous glass fiber reinforced Ethylene-Propylene Copolymer (EPC) matrix composite. High speed tensile tests up to failure and specific interrupted tensile tests have also been performed on glass fiber reinforced EPC composite and on neat ethylene-propylene copolymer (EPC) matrix specimens. Microscopic observations have been performed to identify the deformation processes and the damage mechanisms at the microscopic scale. The deformation processes and the damage mechanisms are then related to the material overall mechanical properties. These tests contributed to a better understanding of the strain rate effect which has been quantified and discussed at both microscopic and macroscopic scales in terms of damage threshold and the kinetic. This paper aims at quantifying the effect of the strain rate (from quasi-static to 200s\(^{-1}\)) on the overall response of EPC matrix composite at the macroscopic scale in terms of longitudinal Young’s modulus, damage threshold and ultimate properties.

Moreover, analysis of SEM observations showed a localized deformation around the fiber-matrix interface zone which occurs particularly at high strain rates. This local viscoplastic deformation directly affects the visco-damage behavior of the
composite material. In fact, increasing the strain rate induces a local viscoplastic dissipation affecting the damage kinetic observed at both microscopic and macroscopic scale: the damage initiation is delayed and the related kinetic is reduced with respect to the quasi-static loading case. It is established that the strain rate effect on the damage behavior of the EPC matrix composite is mainly due to the influence of the viscous behavior of the EPC matrix.

2 Material and microstructure

The material of this study is an injected molded ethylene-propylene copolymer matrix reinforced by discontinuous glass fiber. Average fiber length is up to 1.2 mm whereas the average value fiber weight content is $39 \pm 0.6$ % corresponding to a fiber volume fraction up to 19.7 %. The composite material is presented in the form of 3.2 mm thickness plates, which have been extracted from specific planar zones of actual structural automotive parts. The orientation distribution of fibers in the thermoplastic matrix has been investigated by image analysis of SEM micrographs performed upon a representative element volume whose the observed surface was $2.5 \times 3.2$ mm$^2$ (Fig. 1.). It has been shown (Fig. 2.) that due to the injection molding process, most of the fibers remains into the plane of the plate and are oriented parallel to the Mold Flow Direction (MFD). The process induced orientation distribution confers to the material an overall orthotropic behavior, which can be characterized by a longitudinal direction (X) parallel to the MFD and a transversal direction (Y).

It should be noticed that in this work, the high speed tensile tests and the experimental investigations of the strain rate effect focused exclusively on the longitudinal direction.

3 Experimental approach

3.1 Methods

Dynamic tensile tests up to failure and specific interrupted dynamic tensile tests have been optimized and performed for high strain rates up to $200 \text{s}^{-1}$ to quantify the strain rate effect at different scales. High-speed tensile tests have been conducted upon a servo-hydraulic machine equipped with a launching system [1, 11]. The latter can reach a crosshead speed range from $10^{-4}$ m/s to 20 m/s.

The developed experimental approach is widely detailed elsewhere [1]. The experimental methodology devoted to high-speed tensile tests has been developed by optimizing two main aspects: damping joint and specimen geometry. It combines experimental and numerical analysis aimed at optimizing the sample geometry and a damping joint. These enabled minimizing the perturbation due to the propagation of dynamic shock waves as detailed in a previous paper [1].

The tested composite specimen has dumbbell-shaped geometry and the optimization procedure was achieved by numerical computation using the ABAQUS finite element code to determine the optimal dimensions. The optimization criterion consists of reaching a stabilized stress/strain distribution within the specimen narrow parallel-sided portion. For the glass-fiber reinforced EPC matrix composite, the optimized dumbbell-shaped specimen geometry is given in Fig. 3. The strain is measured using strain gages positioned at the central zone of the specimen as well as by means of a laser extensometer measuring the relative displacement of two points located at the edges of the sample’s central zone. The sampling frequency of the acquisition system utilized for the laser extensometer and for the strain gages can be set from 1 MHz to 50 MHz depending on the nominal test velocity (crosshead speed). The strain rate is then determined directly by a temporal derivation of these strain measurements.

High speed tensile tests were carried out at different strain rates according to two experimental procedures: tensile tests until rupture and successive interrupted tensile tests.

3.2 High-speed tensile tests until rupture

The first type of high-strain rate tensile tests has been performed at different strain-rates until the composite specimen total failure. The optimization described above and detailed in [1] allows ensuring submitting the specimen to a quasi-constant load-rate.

3.3 Interrupted high-speed tensile tests

Interrupted tensile test consists of interrupting the high speed loading at fixed stress levels before the total failure [1]. Indeed, due to inertial effects during a rapid tensile test it is difficult to interrupt a high
speed loading before the total failure of the specimen. An original solution consists in loading the specimen simultaneously with a double-edge notched (DEN) fuse sample characterized by brittle fracture [1, 9, 11]. The fuse ligament width is associated with the suitable failure load level to be reached when the test should be interrupted. The interruption of the high speed tensile test then brings about a stress release in the specimen. This procedure is repeated several times for the same specimen by changing the fuse ligament width before reloading each specimen. Using this procedure, one can reach a load level greater than the previous one. The interrupted test methodology thus enables to monitor different damage level for the studied material.

4 Experimental results

4.1 Strain rate analysis

High speed tensile tests have also been performed on the neat copolymer matrix and the composite material. Specific optimization procedure described in [1] was developed in order to avoid perturbations due to the choc wave during the early stages of the dynamic tensile test. This procedure aimed at rapidly stabilizing the strain rate and to obtain a homogeneous strain during the dynamic loading. In order to reduce system ringing [12], a damping joint enabled a partial absorption of the generated stress wave and the related oscillations. The damping joint should be completely compressed before 30% to 50% of the total elastic deformation of the specimen. The compression time define the rise time ranging from $10^{-3}$ s to $10^{-6}$ s which depends on the adopted joint thickness and the nominal crosshead speed. Fig. 4a shows an example of the evolution of the strain measured by the laser extensometer at a moderate strain rate of $15$s$^{-1}$. For a strain rate of $15$s$^{-4}$, the strain-time curve shows that the slope corresponding to the strain rate is rapidly stabilized after a rise time of about $1.5 x 10^{-4}$s when the total testing time is $1.5 x 10^{-3}$ s. The strain rate is then estimated on the linear part of the curve.

However, for high strain rates, it is more difficult to get a constant strain rate. In fact, the inertial effects may require a thicker damping joint. Consequently, the damping stage is relatively longer with regard to the total time of the test. Thereby, the strain rate measured on the specimen becomes really constant relatively later. Fig. 4b shows the evolution of the strain rate during a high speed test. The average strain rate is approximately $208$s$^{-1}$ and the corresponding rise time is $6 x 10^{-3}$ s whereas the total testing time is $2 x 10^{-4}$ s.

4.2 Strain rate effect on EPC composite

Experimental results established that the overall behavior of the studied composite is strongly strain-rate dependent. Typical EPC matrix composite stress–strain tensile curves are shown in Fig.5. Strain rate effect on the overall mechanical behavior of the glass fiber reinforced EPC matrix has been assessed by plotting in Fig. 6, the evolution of the principal mechanical properties versus strain rate from quasi static to $200$s$^{-1}$. Young’s modulus, damage threshold strain ($\varepsilon_{\text{threshold}}$), damage threshold stress ($\sigma_{\text{threshold}}$), ultimate strain and ultimate stress. All the characteristics vary significantly under rapid straining as compared to quasi-static loading. The dynamic behavior of the studied material is widely strain-rate dependent. One can notice the scattered experimental results obtained which is mainly due to the variation of the microstructure induced by the injection process which characterizes typically the short glass fiber reinforced thermoplastics. Indeed, it is worth noting that tested specimens in this study were cut from plates which have been extracted from specific planar zones of actual structural automotive parts. However, at the local scale, as shown in a recent paper [13], the microstructure has a strong effect on the stress-wave propagation at the local scale. In addition, due to the damping stage during the rise time, the strain rate is not exactly constant in particular for high speed tests. Thus, the indicated value of strain rate is an average value. Furthermore, one can observe that the scattering increases for high strain rate.

Fig6. shows clearly that an increase of the strain rate gives rise to improved overall mechanical characteristics in terms of damage threshold and kinetic. However, it should be mentioned that this increase is noticeably more pronounced after a strain rate beyond to $10$s$^{-1}$. Similar experimental results have been obtained by Schoßig et al. [10]. The effect of strain rate appears through a significant increase.
in the elastic modulus, the damage thresholds and ultimate strain and stress. The ultimate properties and the damage threshold increase significantly at high strain rate. One can notice that the strain rate effect is more visible and marked for the ultimate stress than for the ultimate strain.

4.3 Strain rate effect on pure matrix

The thermal properties of the composite and those on the pure EPC matrix have been measured in terms of melting temperature, temperature of crystallization and crystallinity ratio. The comparison of the two types of measurements showed similar values for unreinforced (neat) and reinforced EPC matrix [11]. Thus, it can be stated that the thermoplastic matrix thermo-mechanical behavior is not significantly affected by the presence of the reinforcement. Then, it can be reasonably assumed that the strain rate effect measured on the pure matrix should be quite representative, at least qualitatively, of that observed for the composite. Figure 7 shows the pure matrix tensile behavior for two different strain rates: quasi-static and a moderate strain rate (25s⁻¹). An important evolution is observed in terms of matrix Young’s modulus, strain and stress damage thresholds which increase significantly. It should be noticed that a high strain rate leads to a significant increase of the matrix deformation (from 4.5% for a quasi-static loading to 10% for a dynamic loading (@25s⁻¹). As it will be discussed later, this is an important point to figure out the inelastic deformation mechanisms of the composite when submitted to high strain rates.

4.4 Multi-scale damage analysis

Optimized interrupted tensile tests [1, 11] have been performed to investigate the main damage and deformation mechanisms occurring at the microscopic scale of EPC composite subjected to high strain-rate loading. The threshold and the kinetic of damage have been quantified at both microscopic and macroscopic scales. The multi-scale analysis combined to microscopic observations showed that the strain rate effect on the overall behavior and damage is basically due to the fiber-matrix interface debonding. The propagation of this damage mechanism leads to matrix micro-cracking [2]. In this procedure, macroscopic damage can be quantified by tensile modulus reduction using the well-known continuum damage mechanics using the damage scalar variable \((D_{macro})\). Interrupted high-speed tensile tests have been carried-out for two strain rates: 0.0005 s⁻¹ and 15 s⁻¹. The inelastic (permanent) strain measured after unloading is caused by the subsequent crack opening. For each “unloading/re-loading” loop (Fig. 8(a)), the evolution of the macroscopic damage parameter \((D_{macro})\) is experimentally estimated and plotted as a function of the strain level reached for both strain rates as shown in Fig. 8(b). Experimental results show that the stiffness reduction is rate dependent. Nevertheless, since the slopes of the both curves of \(D_{macro}\) as a function of the strain are the same, one can state hence that the damage kinetic is relatively insensitive to the strain rate.

Moreover, the evolution of a microscopic damage indicator representing the fiber-interface debonding during loading, it is expressed as:

\[
d_{micro} = 1 - \frac{f_d}{f_v},
\]

where \(f_d\) is the volume fraction of debonded fibers and \(f_v\) is the fiber volume content in the RVE. In order to measure the evolution of \(d_{micro}\) before each re-loading stage, a microscopic image analysis is achieved by means of SEM micrographs performed on a representative volume element (RVE): the observed surface was of 2.5x3.2 mm² which is large enough to contain all the heterogeneities of the material microstructure. So, the local investigation can be assumed as statistically representative of the damage accumulation in the studied material composite.

Under quasi-static loading, damage mainly begins simultaneously with the first non-linearity on the stress-strain curve \((\varepsilon_{threshold} = 0.18\%)\). It appears first as fiber-matrix interface debonding which occurs on the more disoriented fibers (\(\theta=90°\)). Then, it progressively propagates to the less disoriented fibers. Besides the interface debonding, one can observe matrix micro-cracking around the debonded fibers. For a strain of 0.4% which corresponds to the end of the knee point on the stress-strain curve (Fig. 5), it is worth mentioning that the fiber-matrix interface failure is generalized for all fiber orientations on the whole volume of the specimen (Fig. 10). A diffuse damage over the whole
investigated zone is observed until the final stages of the specimen behavior which occurs after the coalescence of the fiber-matrix interface debonding and the matrix micro-cracks.

4.5 Local viscous effect and damage at high strain rate

For high strain rates, no damage is observed before the strain threshold of 0.45%. A large influence of the matrix viscosity (rheology) on the damage evolution is observed. Conversely to the quasi-static loading, no matrix micro-cracking is observed around the debonded interfaces. Moreover, it was established that the strain rate brings about a viscous nature of damage evolution due to the matrix rheology. In fact, it has been shown that the viscoelastic-viscoplastic behavior of the EPC matrix has an important contribution into the strain rate effect. Specifically for high strain rates, it has been observed that a local zone around fibers is highly strained due to local matrix rheology (Fig. 11).

Fig. 9 shows the comparison at the microscopic scale of the damage threshold and evolution for two strain rates. For a strain rate of 15 s⁻¹, the damage occurs firstly through the interfacial decohesion and the related deformation zone. With the increase of the strain, beyond a strain of 1.53% the interfacial decohesion starts to propagate through the matrix from one fiber to its closer neighbor. These microscopic damage mechanisms intensify through the accumulation of micro-defects leading to macroscopic failure. The capability of the matrix to undergo this large strain localization around the fibers explains also the reduced damage kinetic observed at the microscopic scale. This statement confirms that this strained zone around the debonded interface dissipates a part of the strain energy and accordingly hinders the interfacial crack propagation through the matrix. Consequently, it reduces the degradation and its kinetics at the microscopic scale. Thus, for high speed loading the damage initiation is delayed and the related kinetic is reduced with respect to the quasi-static loading case [1, 11].

Furthermore, one can see figure 9, that the microdamage level and its kinetic decrease when the strain rate increases. These results lead to the conclusion that, to investigate the degradation kinetic, the microdamage indicator (d_micro) is more relevant than the macroscopic damage variable expressed through the stiffness reduction. Indeed, as shown in figure 6b, for a strain rate of 15s⁻¹, the evolution of the longitudinal Young’s modulus is not sensitive to the damage occurring at disoriented fibers due to its low value. That can explain why the studied composite material exhibits similar macroscopic damage kinetic for quasi-static and dynamic loading (Fig. 8(b)).

5 Conclusion

In this paper, high strain rate tensile tests up to failure and interrupted tensile tests coupled with SEM observations lead to a better comprehension of the main local mechanisms at the origin of the strain rate effect observed at the macroscopic scale. Qualitative and quantitative relations between the microscopic phenomena and the macroscopic properties have been established. It has been demonstrated that the tensile behavior of short fibers reinforced EPC matrix composite is strongly rate dependent. Damage threshold and ultimate properties, are highly sensitive to the strain rate. As the strain rate increases, noticeable effects consist of an increase of the Young’s modulus and a delayed damage onset. These effects are followed by a slightly reduced damage accumulation.

It has been established that the predominant damage mechanism is the fiber-matrix interface debonding which propagates through the whole volume of the sample and for quasi-static loading matrix micro-cracking. It was demonstrated that the strain rate brings about a viscous nature of damage evolution due to the matrix rheology. The viscoelastic-viscoplastic behavior of the EPC matrix has an important role in the strain rate effect. For high strain rate, it has been observed that a local zone around the fibers is highly strained due to local matrix rheology. The capability of the matrix to undergo this large strain localization around the fibers explains also the reduced damage kinetic observed at the microscopic scale and leads to the absence of the matrix micro-cracking mechanism for high strain rates. One can conclude that the strain rate effect can be explained by the interaction of two main phenomena: the local and the global viscous behavior of the matrix and fiber-matrix interface debonding. These two phenomena are coupled and
lead to: an increase of the composite stiffness, a delay of the damage threshold, a diminution of the fiber-matrix interface failure kinetic and reduction of the matrix micro-cracking.

The comparison of damage kinetics at microscopic and macroscopic levels for quasi-static and dynamic loading confirms that the micro damage indicator is more relevant than the macroscopic stiffness reduction for investigating damage accumulation. Indeed, the macroscopic damage investigation based on the longitudinal Young’s modulus reduction cannot reflect the effect of the strain rate on the damage kinetic. Finally, the proposed multi-scale analysis highlights clearly that the strain rate effect observed at the macroscopic scale is especially controlled by the interfacial debonding of disoriented fibers and the EPC matrix viscous rheology. The latter affects the deformation and the damage history at the microscopic scale. In addition, the time-dependent damaged-behavior can explain readily the accommodation exhibited at the macroscopic scale.

In a further communication, all these results will be introduced in a multi-scale approach [14] integrating, at the microscopic scale, the couple effect of the damage and of the viscous behavior of the EPC matrix..

Fig. 1. Typical microstructure of the composite according to two views.

Fig. 2. Definition of the fibers orientation angles and fiber orientation distribution measured by image analysis.

Fig. 3. Optimized specimen geometry for high speed tensile tests of short glass fiber reinforced composites.

Fig. 4 (a): Measured strain versus time curve for a moderate speed tension test ( $\dot{\varepsilon} = 15 \text{s}^{-1}$ )
Fig. 4 (b): Measured strain versus time curve for a high speed tension test ($\varepsilon_{\text{max}} = 208 \text{s}^{-1}$)

Fig. 5. Typical EPC matrix composite stress–strain tensile curves.

Fig 6. (a): Strain rate effect on the EPC matrix composite longitudinal Young’s modulus

Fig 6. (b): Strain rate effect on the EPC matrix composite damage threshold and ultimate strain
Fig. 6. (c): Strain rate effect on the EPC matrix composite damage threshold and ultimate stress.

Fig. 7.: EPC pure matrix (without reinforcement) stress-strain tensile curves.

Fig. 8.: (a) Loading-unloading stress-strain curves for a strain rate of $15 \text{ s}^{-1}$.
(b) Comparison of macroscopic damage ($D_{macro}$) evolution for a quasi-static and a dynamic test.

Fig. 9.: Comparison between dynamic and quasi-static overall microscopic damage kinetic at the microscopic scale.
LOCAL STRAIN RATE EFFECT ON DAMAGE IN GLASS FIBER REINFORCED ETHYLENE-PROPYLENE COMPOSITE

References

Fig. 10: Microscopic observation of the diffuse damage of glass fiber reinforced EPC under quasi-static loading.

Fig. 11: Matrix local viscoplastic deformation around a debonded fiber