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Revealing the effects of matrix behavior on low-velocity impact response of continuous fiber-reinforced thermoplastic laminates

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Abstract

Matrix behavior is expected to widely influence the impact response of composites, but detailed conclusions in the case of thermoplastic laminates are still needed. In this paper, we investigated the effect of using either ductile homopolymer PP or less-ductile impact copolymer PP matrices on the low-velocity impact responses of continuous glass fiber-reinforced polypropylene (PP) laminate. These PP types represent two variants in the same family of thermoplastic matrix. A thorough experimental campaign was first performed to provide the tensile properties (for PP and glass/PP) and fracture toughness (Mode-I and Mode-II, glass/PP only) of the employed materials. Then, low-velocity impact tests where the energy levels are ranging from 12 to 30 J were performed. Using ductile PP in glass/PP laminates reduces the energy dissipated during impact as well as the impact damage area. The effect of selected stacking sequences on the resistance to impact was also studied as a way to reveal the difference between ductile and less-ductile glass/PP. Stacking sequence with thin plies shows better impact properties than other sequences regardless of the matrix ductility, which can be explained by micromechanics for both grades of material. Finally, as quasi-static indentation (QSI) is usually used to quickly access the resistance of laminates towards out-of-plane impact, we systematically compared our impact results with QSI results. We found that the prospective use of QSI in forecasting impact properties and damage is very limited in glass/PP composite due to strain-rate sensitivity.

Keywords: Glass/polypropylene, damage, impact, quasi-static indentation

1. Introduction

Composite laminates are widely used to manufacture lightweight components in automotive, electronic and aircraft structures. However, the use of composites for making external parts is still limited because these parts may be subjected to a number of severe impacts arising from operational events, e.g., collision, flying debris, tool drop or bird strike. On one side, since composites exhibit limited plastic deformation the impact damage can be beneficial as it becomes the primary energy dissipation mechanism [1]. Nonetheless, impact damage strongly compromises the residual strength of composite parts, e.g. compression-after-impact performance. Understanding the damage phenomenology under out-of-plane impact conditions is thus essential.

Impact damage phenomenology has been studied for thermoset-based [2, 3, 4, 5, 6, 7, 8] and thermoplastic-based composites [9, 10, 11, 12, 13, 14, 15, 16]. Both the fiber type [17] and the matrix type should be carefully chosen to obtain the desired impact properties. Thermoplastic matrix is generally preferred over thermoset for improving impact resistance. For example, selecting polypropylene (thermoplastic) over epoxy (thermoset) was found to improve the maximum impact force, absorbed energy and peak energy in glass fiber-reinforced composites due to strain hardening and resistance to matrix cracking [14]. Impact loading on thermoplastic composite systems (carbon/PEEK, carbon/Torlon) prompted less amount of damage than that on thermoset composite (carbon/bismaleimide) [9]. The impact damage resistance of PEEK was associated with its high Mode-I interlaminar fracture toughness (G_{Ic}) [18]. A recent investigation by Vieille et al. [13] suggested that matrix plasticization at the crack tip and fiber-bridging during Mode-I fracture (opening mode) were responsible for the reduction of damage in thermoplastic composites.

One of the widely-used thermoplastic matrices for the impact-prone component is polypropylene (PP). PP is abundant and can be economically produced at large scale [19]. However, the relatively brittle nature of PP is limiting its applications, particularly when it contains a notch and tested at (or below) room temperature [20]. Thus, further improvement of

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PP, especially its fracture toughness under impact, resulted in the development of “impact copolymer polypropylene” obtained by adding rubber into the PP homopolymer [19]. Subjected to impact loading, rubber within PP phase create micro-voids that allow larger matrix deformation during yielding process [21, 22]. PP has also been continuously developed by reinforcing it with glass fibers (viz. glass/PP). The low-velocity impact response of glass/PP was reported for various studies: the comparative study between thermoplastic and thermoset matrices [14], effect of fiber/matrix interface [23], validation of intralaminar and interlaminar damage models [24], modeling of glass/PP with long fibers [25], sandwich composite application [26], lay-up optimization [27], thermal response study of self-reinforced PP composites [28]. Nevertheless, the effect of thermoplastics matrix types (homopolymer PP and copolymer PP) on the low-velocity impact response of glass/PP laminates with some important stacking sequences has not been reported so far. The effect of matrix types in that context was reported for thermoset composites, e.g., bismaleimide [9], epoxy [2, 29, 30], showing that toughening thermoset matrices improved the impact damage resistance. Moreover, the understanding of low-velocity impact response of thermoplastic composites, particularly PP-based laminates, is still not at the same level with thermoset composites. In fact, such understanding is critical in the material development process aiming at cost-efficient impact resistant thermoplastic-based structures.

In this paper, our first objective was to investigate the effect of PP matrix type on impact properties and damage of continuous glass fiber-reinforced PP laminates. Here, we compared glass-reinforced homopolymer PP (referred to as “GF-PP”) with glass-reinforced impact copolymer PP (“GF-IPP”). Homopolymer PP was found to be more ductile than impact copolymer PP [31]. Our baseline lay-up for impact test was $[90_2/0_2]_s$, a simple lay-up applicable for automotive or electronic enclosures. Secondly, we investigated how GF-PP and GF-IPP perform with respect to some classical lay-up modifications, namely (i) reducing the ply thickness of the baseline lay-up into $[90/0]_{2s}$ (similar to ply-block scaling [32], but with a constant thickness), (ii) changing the relative angle between plies of the baseline lay-up into $[90_2/\pm 45]_s$. Thirdly, we used GF-PP and GF-IPP as thermoplastic material systems to study the relevance of using quasi-static indentation (QSI) technique, reported earlier in Ref.

[31], to forecast the impact properties. Thus far, the comparison between QSI and impact damage was performed for thermoset composites, showing that both techniques resulted in a comparable damage pattern [5, 33, 34, 30, 35, 36]. For thermoplastic composites, QSI test produces quite different damage than impact, e.g. glass/polyester [37], carbon/polyamide-6 [38]. To the best of authors' knowledge, this is the first systematic comparative study for glass/PP with various lay-ups.

2. Experimental Details

2.1. Materials

We used continuous E-glass fiber-reinforced PP in a unidirectional tape form (110 mm width, 0.25 mm thickness) produced by SABIC [39]. The glass fibers have an average diameter of $16.3 \mu\text{m}$, density of 2920 kg/m^3 , ultimate tensile strength of $1799 \pm 823 \text{ MPa}$, tensile modulus of $78.2 \pm 10.0 \text{ GPa}$, and failure strain of $2.3 \pm 1.1\%$. Two polypropylene matrices were used, i.e., homopolymer PP ("PP") and impact copolymer PP (rubber particles added to the homopolymer PP phase; "IPP"). The PP and IPP are 905 and 910 kg/m^3 , respectively; the melt flow index of PP and IPP is 100 and 120 dg/min, respectively. Correspondingly, we used two glass/polypropylene composites called "GF-PP" (homopolymer-based) and "GF-IPP" (impact copolymer-based).

2.2. Manufacture of impact specimens

Three lay-ups were prepared for each type, i.e. $[90_2/0_2]_s$, $[90/0]_{2s}$ and $[90_2/\pm 45]_s$. Here, $[90_2/0_2]_s$ was a baseline cross-ply laminate with thicker plies; $[90/0]_{2s}$ having a reduced ply thickness was expected to affect the transverse cracking mechanism as well as the transition from transverse cracks to delamination; $[90_2/\pm 45]_s$ activates the delamination that is strongly related to a smaller angle of ply interface in comparison to $[90_2/0_2]_s$. The laminates were manufactured by compression molding (static press Pinette Emidecau Industries 15 Tons, PEI 15T) using metallic molds with processing parameters prescribed in Table 1. PP/mold adhesion was prevented using a release agent, while PP leakage from the mold was prevented using polyimide (Kapton) tapes. The resulting plates were cut into rectangular

specimens (110 mm × 110 mm) with a thickness of 2 mm using a water-cooled cutting machine (Struers).

Table 1: Processing parameters for manufacturing the laminates [39, 31].

<i>Parameter</i>	GF-PP	GF-IPP
Mold material	Aluminum	Steel
Plate size	250×110 mm ²	275×110 mm ²
Pressure (equivalent force)	6.8 bar (18 kN)	7.5 bar (22 kN)
Initial temperature	25°C	25°C
Heating rate	10°C/min	10°C/min
Dwell temperature	230°C	210°C
Dwell time	20 min	20 min
Cooling rate	40°C/min	40°C/min

2.3. Impact test methods

The low-velocity impact test was performed using an impact test system Instron CEAST 9350 shown in Fig. 1a. The impactor displayed in Fig. 1b has a hemispherical tip (diameter of 16 mm) and the total mass of 5.392 kg (impactor and tup holder masses were 1.092 kg and 4.3 kg, respectively). Fig. 1c shows that the test system was equipped with a velocity sensor and an anti-rebound system to avoid the repeated impact (after the first impact) on the plate. The specimen was compressed at 5.5 bar using a pneumatic clamping system (see Fig. 1d) that has a circular cut-out (diameter = 76 mm) at the upper and lower clamps. The impactor was then placed at a prescribed falling height ranging from 227 mm to 569 mm to achieve a velocity range of 2.1-3.4 m/s that corresponds to impact energy range of 12-30 J. The test was performed at 25 °C. Table 2 shows the impact test parameters for testing totally 72 specimens.

The software in Instron CEAST 9350 recorded the following impact test parameters: time (t), force (F), impactor velocity (v_i), displacement (s), energy (E). The force was measured using a load cell embedded within the impactor. v_i was measured using a velocity sensor (Fig. 1c) with the double-prong system:

$$v_i = \frac{l_{12}}{t_2 - t_1} + g \left(t_i - \frac{t_1 + t_2}{2} \right) \quad (1)$$

Figure 1: (a) Impact test system, (b) weight and impactor, (c) velocity sensor and anti-rebound system, (d) clamping system.

Table 2: Impact test parameters.

Specified impact energy (J)	Falling height (mm)	Impact velocity (m/s)	Specimen quantity
12	227	2.11	18
18	345	2.60	18
24	453	2.98	18
30	569	3.34	18

where l_{12} is the distance between two flag prongs; t_1 and t_2 are the time needed for the first

and the second flag prongs, respectively, to pass the detector; t_i is the time of initial contact determined from the force-time curve; g is gravity of 9.81 m/s^2 . The velocity of impactor $v(t)$ at time t was calculated according to the ASTM D7136 Standard as follows:

$$v(t) = v_i + gt - \int_0^t \frac{F(t)}{m} dt \quad (2)$$

where $F(t)$ is the force at time t , and m is total impactor mass. The displacement $s(t)$, i.e. the deflection of the impacted point in the sample, was calculated as follows:

$$s(t) = v_i t + \frac{gt^2}{2} - \int_0^t \left(\int_0^t \frac{F(t)}{m} dt \right) dt \quad (3)$$

Since the time was fully synchronized between force and displacement, the energy (E) can be calculated as follows:

$$E = \int_0^s F(s) ds \quad (4)$$

The relationship between energy and time can then be schematically depicted in Fig. 2. From the energy-time curve, three components of energy can be derived. E_i is the maximum energy at maximum force. The end-point in the energy-time curve can be used to estimate the elastic energy (E_e that is non-dissipated during the impact) and absorbed energy (E_a that is dissipated during the impact). The relationship among E_i , E_e and E_a can then be written as follows:

$$E_i = E_a + E_e \quad (5)$$

2.4. Damage evaluation methods

We evaluated the global impact damage area using the backlight method, which was developed to identify damage progression in glass/PP under QSI load; readers may refer to Ref. [31] for the QSI test set-up. A similar method was also used in the damage analysis of glass/epoxy [40]. The impacted face of the specimen was subjected to a light source (Direct Current Regulated Light Source DCR III from Schott-Fostec; 4" Maxi Ringlights from Moritex) from 80 mm distance. The shadow of damage was captured by the digital

Figure 2: Schematic of typical energy-time curve obtained from the low-velocity impact test.

camera (Canon 700D) from the opposite, distal face at 300 mm distance, and the damage size was measured using *ImageJ* software [41]. To identify the internal damage modes, we used scanning electron microscope (SEM) Quanta 600 (FEI). The specimen's impacted region was cut along 90°-direction, and the cross-section was polished and sputtered with 4-nm thick iridium (Ir) using Q150TS (Quorum Technologies).

3. Results and Discussion

3.1. Overview of tensile and fracture properties

Tensile properties of impact copolymer polypropylene (IPP) at various strain rates (10^{-3} , 10^{-2} , 10^{-1} /s) were obtained from testing dumb-bell specimens (ISO 527-2 1BA Standard) using Instron 5944 (2 kN load cell). The details can be found in our earlier works [39, 31]. Here, we added the tensile properties of homopolymer PP that were obtained using the same standard and test machine. Figs. 3a-b show that PP (homopolymer) is stiffer and stronger than IPP (copolymer) for all studied rates. This is an expected outcome as our particular PP has a higher molecular weight than the employed IPP. Both PP types similarly exhibit rate-dependent stiffness (modulus increases with an increase of strain rate; see Fig. 3a) and strength (strength increases with an increase of strain rate; see Fig. 3b). The sensitivity

of polypropylene towards the strain rate is maintained regardless of the rubber particle addition.

Figure 3: Tensile behavior of polypropylene matrix: (a) tensile modulus, (b) tensile strength [39].

Table 3 provides a summary of the tensile and fracture properties of GF-PP and GF-IPP, which was already reported in Refs. [39, 31, 42]. GF-PP is stronger and stiffer than GF-IPP in longitudinal, transverse and in-plane shear directions. In-plane shear properties of GF-IPP were found to be strain-rate dependent [39], while that of GF-PP is expected to behave the same way. GF-PP and GF-IPP exhibit a similar transverse crack behavior since their Mode-I intralaminar fracture toughness is comparable. In contrast, their delamination behavior is expected to be different since interlaminar fracture toughness (Mode-I, Mode-II) of GF-PP is significantly higher than that of GF-IPP.

3.2. Impact properties and damage phenomenology

The representative force-displacement curves of GF-PP and GF-IPP under various impact energies (12, 18, 24, 30 J) for [90₂/0₂]_s, [90/0]_{2s} and [90₂/±45]_s are given in Figs. 4a-c, respectively. GF-PP and GF-IPP at lower impact energy (12 or 18 J) exhibited similar hysteresis loop, indicating that their energy absorption is also similar. As the impact energy is increased up to 30 J, GF-IPP experienced a more drastic load drop (which was followed by the unloading path) than GF-PP. Substantial load drop in GF-IPP indicates a loss of

Table 3: Tensile and fracture properties of GF-PP and GF-IPP [31, 42].

<i>Properties</i>	GF-PP	GF-IPP
Fiber volume fraction, %	44.1 ± 1.3	38.7 ± 1.9
Longitudinal strength, MPa	846 ± 87	744 ± 73
Longitudinal modulus, GPa	36.1 ± 1.1	34.3 ± 1.6
Longitudinal failure strain, %	2.53 ± 0.16	2.32 ± 0.13
Transverse strength, MPa	25.3 ± 1.2	16.8 ± 0.6
Transverse modulus, GPa	5.59 ± 0.28	4.43 ± 0.19
Transverse failure strain, %	0.59 ± 0.11	0.50 ± 0.05
In-plane shear strength, MPa	55.8 ± 2.7	32.6 ± 2.1
In-plane shear modulus, GPa	1.35 ± 0.06	0.99 ± 0.04
In-plane shear failure strain, %	50.8 ± 3.8	32.8 ± 1.8
Mode I intralaminar fracture toughness, kJ/m^2	0.37 ± 0.07	0.33 ± 0.04
Mode I interlaminar fracture toughness, kJ/m^2	0.74 ± 0.05	0.21 ± 0.02
Mode II interlaminar fracture toughness, kJ/m^2	3.92 ± 0.41	0.92 ± 0.02

elastic energy and its vulnerability towards fiber fracture and perforation.

The energy-time curves obtained from impact tests are shown in Fig. 5a where we derived the absorbed energy and elastic energy. Fig. 5b shows that GF-PP is not sensitive to the stacking sequence. For all stacking sequences in GF-PP, the same level of impact energy produced a comparable energy absorption. The trend of energy absorption in GF-PP also appears to be linear regardless of the stacking sequence. This linearity suggests that the change in impact energy level does not necessarily modify the damage phenomenology. In contrast, the energy absorption in GF-IPP is dependent on the stacking sequence. The relationship between energy absorption and impact energy in GF-IPP $[90_2/0_2]_s$ exhibits a linear trend, while that in GF-IPP $[90/0]_{2s}$ or $[90_2/\pm 45]_s$ is non-linear. The non-linearity of energy absorption displayed by GF-IPP $[90/0]_{2s}$ or $[90_2/\pm 45]_s$ suggests that these lay-ups are prone to penetration and degradation of elastic energy (see Fig. 5c), particularly at higher impact energy levels.

The effect of matrix behavior was not evident in $[90/0]_{2s}$ as the trend of energy absorption is quite similar for GF-PP and GF-IPP (see Fig. 5b). Transverse crack and transition-to-delamination in $[90/0]_{2s}$ were difficult to develop, which corroborates the finding by Yokozeki

Figure 4: Force-displacement curves of GF-PP and GF-IPP at various energy levels: (a) $[90_2/0_2]_s$, (b) $[90/0]_{2s}$, (c) $[90_2/\pm 45]_s$.

et al. [43]. This observation is also consistent with the micro-mechanics of laminated systems, such as the one proposed in Ref. [44]. Lubineau [44] suggested that the transition from transverse crack to delamination is a thickness-dependent process, i.e. thin plies experience fewer transverse cracks and induce smaller delamination at higher strain than the thick plies. The fact that the matrix behavior has a limited effect on transverse crack mechanism has been shown in Ref. [45] as this mechanism is mainly guided by the geometry of the inter-fiber arrangement. Thus, the stacking sequence with thin plies ($[90/0]_{2s}$) triggered a very limited delamination and mainly experienced transverse cracking instead. Hence, the

Figure 5: (a) Energy-time curves, (b) impact energy and absorbed energy, (c) impact energy and elastic energy.

slope of energy absorption and impact energy in GF-PP and GF-IPP for $[90/0]_{2s}$ is similar up to 30 J when the penetration occurred. This is consistent with Fig. 6b where the damage area in GF-PP and GF-IPP for $[90/0]_{2s}$ is comparable.

On the other hand, the effect of matrix behavior is evident in $[90_2/0_2]_s$. Fig. 5b shows that in $[90_2/0_2]_s$ GF-IPP clearly absorbed more energy than GF-PP at the same impact energy level, meaning that GF-IPP exhibits a more severe damage than GF-PP. Table 3 already shows that GF-IPP has a lower interlaminar fracture toughness than GF-PP, suggesting that the former is more prone to develop delamination and experience larger damage size than the latter (see damage area vs. impact energy in Fig. 6a).

In terms of damage resistance, selecting $[90/0]_{2s}$ for GF-PP and GF-IPP over $[90_2/0_2]_s$ or $[90_2/\pm 45]_s$ seems to be beneficial. Fig. 6b shows that $[90/0]_{2s}$ has a much smaller damage area than two other lay-ups (see Figs. 6a and c for $[90_2/0_2]_s$ and $[90_2/\pm 45]_s$, respectively) since the transverse crack and delamination are very limited within $[90/0]_{2s}$ for both materials (see Fig. 8). In contrast, Figs. 7 and 9 show that $[90_2/0_2]_s$ and $[90_2/\pm 45]_s$, respectively, exhibit a large number of transverse cracks and large delamination. $[90_2/\pm 45]_s$ was found to exhibit a large number of shear-induced transverse cracks and delamination. The non-linearity of energy absorption as displayed in GF-IPP $[90_2/\pm 45]_s$ (see again Fig. 5b) was caused by the fact that this less-ductile system induces delamination at all ply interfaces, which triggers early fiber fracture. Thus, laminate with thin plies is favorable in suppressing impact damage area regardless of the matrix behavior.

The understanding of impact-induced damage phenomenology in thermoplastic laminates with different matrix ductility levels is very critical, especially in structural design, because impact damage may reduce their residual strength, e.g., compression-after-impact (CAI) strength [46, 47], of up to 48% [48]. A more ductile composite (carbon/PEEK) could have an improved CAI strength by 25% in comparison to the less ductile one (carbon/PPS) because carbon/PEEK showed a better resistance to transverse cracking and delamination than carbon/PPS [49]. Furthermore, Vieille et al. [48] suggested that the compressive failure of impacted-composite with a more ductile matrix is characterized by micro-buckling that is accompanied by extensive plastic deformation, causing plastic buckling phenomenon, which then delays the transverse cracking and improves CAI strength. In terms of lay-up, sub-laminate scaling strategy using $[45/0/90/-45]_{4s}$ rather than ply-blocking strategy $[45_2/0_2/90_2/-45_2]_{2s}$ was found to improve CAI strength because sub-laminate scaling could reduce impact-induced delamination and dent depth [50]; this strategy is analogous with the comparison between $[90/0]_{2s}$ and $[90_2/0_2]_s$ in the present work. Nonetheless, the effect of matrix ductility levels and lay-ups on CAI strength of glass/PP could be of future interest.

Figure 6: Impact damage obtained using backlight method and the measured damage area in GF-PP and GF-IPP at 12 J and 30 J impact energy: (a) $[90_2/0_2]_s$, (b) $[90/0]_{2s}$, (c) $[90_2/\pm 45]_s$.

Figure 7: Comparison of internal damage in GF-PP and GF-IPP of $[90_2/0_2]_s$ lay-up: force-displacement curves, backlight images and SEM of Section A-A'.

Figure 8: Comparison of internal damage in GF-PP and GF-IPP of $[90/0]_{2s}$ lay-up: force-displacement curves, backlight images and SEM of Section A-A'.

Figure 9: Comparison of internal damage in GF-PP and GF-IPP of $[90_2/\pm 45]_s$ lay-up: force-displacement curves, backlight images and SEM of Section A-A'.

3.3. Comparison between impact and QSI

The comparison of force-displacement between impact and QSI for GF-PP and GF-IPP was made by selecting the displacement limit of around 7 mm. The impact case considers 12 J impact energy, while QSI data employs the load-unload results at a displacement limit of 7 mm that were published earlier [31]. The results containing the peak force (F_{max}), absorbed energy (E_a), damage area (A) as well as the backlight images are given in Fig. 10. It is noteworthy that we applied similar clamping condition (albeit different mechanisms employed) to the specimen tested under impact and QSI loads. The main difference is that the loading speed (v) was 2.1 m/s and 2.1×10^{-5} m/s for impact and QSI, respectively.

Our finding suggests that QSI generally failed to reproduce force-displacement curve of impact for all stacking sequences as well as for both GF-PP and GF-IPP. In the early stage of the curve, QSI produced lower force-displacement slope (i.e., stiffness) than impact. Stiffer response during impact shows that glass/PP laminate is a strain-rate sensitive system, and this behavior is determined by the strain-rate sensitivity (i.e., viscoelasticity) of the matrix. The peak force obtained from QSI is also lower than that from impact tests although the absorbed energy appeared to be similar (see, for instance, GF-PP $[90/0]_{2s}$ or GF-IPP $[90_2/\pm 45]_s$ in Fig. 10). This peak force is strongly related to fiber strength that determines the onset of fiber breakage. Since the strength of glass fiber is increasing with the increase of strain rate [51], fiber breakage during impact is likely to occur at a higher magnitude than that during QSI. This finding agrees quite well with impact-QSI comparative study for glass/polyamide-6 thermoplastic composite [38]. In terms of damage, Fig. 10 shows that QSI produced relatively larger damage area than impact. The invalidity of QSI in predicting impact damage may be associated with the strain sensitivity of the interlaminar fracture behavior, particularly Mode-II fracture, which dominantly controls the growth of delamination in the laminate under out-of-plane loading [52, 53]. Therefore, despite the fact that QSI may be a cost-effective method in obtaining the out-of-plane properties and stage-by-stage damage progression, one should be cautious to use QSI results in forecasting impact test results of thermoplastic composites (with and without rubber particles in the matrix), which are strain-rate dependent.

Figure 10: Comparison of force-displacement curves and backlight damage zone between QSI and impact for GF-PP and GF-IPP.

4. Conclusions

Low-velocity impact tests have been performed on continuous glass-reinforced PP laminates having two different PP types representing two different mechanical behaviors. Ductile homopolymer PP and less-ductile copolymer PP were used to produce GF-PP and GF-IPP laminates, respectively. The effect of stacking sequence is also studied as a way to compare two PP types. PP behavior modifies the magnitude of absorbed energy, elastic energy and damage area in glass/PP laminates, especially at a high impact energy of 30 J. The matrix type has mainly an effect on stacking sequences in which different ply thicknesses are considered. Indeed, thick plies are more prone to transverse cracking and transverse cracking-to-delamination transition where the matrix behavior plays a major role. In addition, the current study shows that the use of QSI to forecast impact damage in glass/PP is very limited due to the strain rate sensitivity of the matrix and interlaminar damage behavior.

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