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RESEARCH ARTICLE

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Impact response of interphase-modified CFRP laminates: an assessment across PEI, PES, PSU, and PC

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ABSTRACT

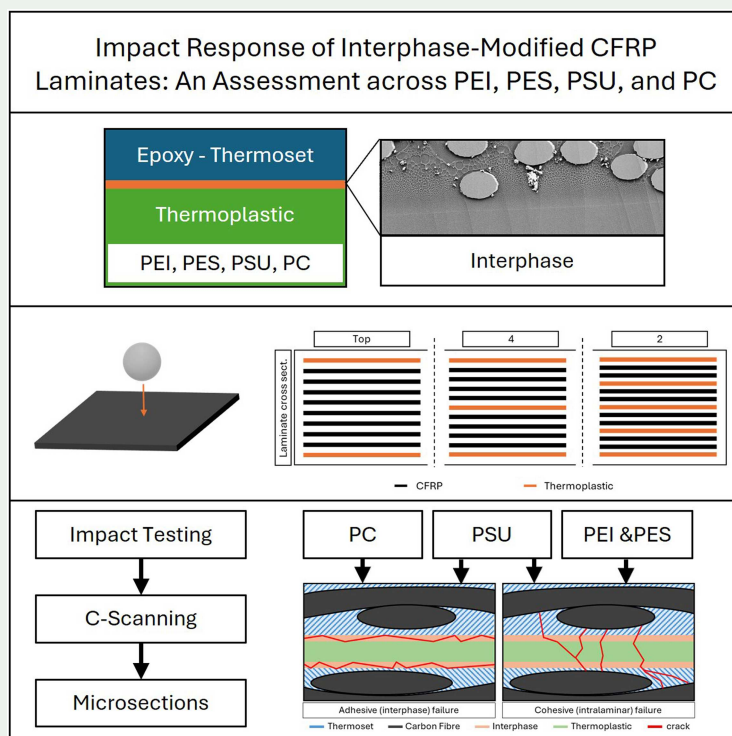
This study evaluates the low-velocity impact response of aerospace-grade epoxy laminates (M18/1) interleaved with four distinct thermoplastic films (PEI, PES, PSU, PC). Impact performance was correlated with quasi-static fracture toughness to assess the transferability of intrinsic polymer properties to dynamic loading. The results reveal a fundamental trade-off governed by chemical compatibility. Chemically compatible interphases (PEI, PES) successfully bridged the stiffness mismatch, increasing the damage initiation threshold by up to 23%, although total energy absorption plateaued due to matrix saturation. Conversely, the incompatible Polycarbonate (PC) system exhibited a mechanism shift: despite high intrinsic ductility, low Mode I adhesion (G_{Ic}) caused immediate decoupling, triggering a pseudoplastic sliding mode that increased energy absorption (+10%) at the cost of structural integrity. Polysulfone (PSU) revealed an inefficient compromise, suffering from premature adhesive failure without providing significant energy dissipation. These findings challenge established propagation-based models, identifying Mode I initiation toughness as the critical 'gatekeeper' for structural toughening in hybrid laminates.

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1. Introduction

Carbon Fiber Reinforced Polymers (CFRP) have become the material of choice for lightweight structures in the aerospace industry due to their exceptional specific stiffness and strength [1–4]. However, a persistent limitation of this material class remains its sensitivity to transient loads, particularly low-velocity impact [5–10]. Even low energy levels, which cause Barely Visible Impact Damage (BVID) on the surface, can induce severe internal delamination, significantly reducing the residual load-bearing capacity of the structure [8–10]. To enhance damage tolerance, interleaving, the integration of tough thermoplastic films between prepreg plies, has established itself as an effective toughening strategy [11–26]. The primary objective is to create a functional interphase capable of dissipating fracture energy through plastic deformation [10–13,27–35]. While the fundamental concept is widely adopted, the current state of the art indicates that its success is not universally guaranteed but critically depends on material compatibility, processing conditions, and the specific nature of the loading regime [16–21,27,36].

The efficiency of thermoplastic interleaves in improving quasi-static fracture toughness (G_{IC} , G_{IIC}) is well documented. Recent studies by WEIDENMANN et al. (2025) on PEI-modified epoxy laminates demonstrated that compatible interphases can achieve Mode I fracture toughness values exceeding 1500 J/m², with mechanisms such as ‘crack jumping’ contributing significantly to energy dissipation [21]. Similarly, Too et al. (2025) showed that optimizing process parameters for PA-6 interleaves can drastically increase G_{IC} [19]. These findings confirm the immense potential of compatible, ductile interphases under static loading conditions.

However, transferring these static properties to high-strain-rate impact events is complex. While impact resistance is theoretically governed by crack propagation energy, the dynamic and multi-axial nature of an impact fundamentally alters the material response. This discrepancy was explicitly highlighted by FOTOUHI et al. (2019), who observed that the toughening effect of PES interleaves is highly strain-rate dependent [17]. Their findings demonstrate that interphases can behave significantly differently under higher impact velocities than statically assumed.

At the same time, the rate-dependent response of the interleaf material itself must be distinguished from the integrity of the interphase it forms with the thermoset. A dynamically ductile film does not necessarily improve structural impact resistance if the interface fails before that ductility can be engaged at the laminate level. This distinction is consistent with the broader interphase perspective summarized by Liu et al. (2015), who identified the interphase as a key mediator of stress transfer and macroscopic composite performance [37]. It also aligns with the impact study of Lai et al. (2025), where improvements in interlaminar resistance enhanced laminate impact performance only when structural continuity across the interlaminar region was maintained [38].

Furthermore, a specific vulnerability of interleaved laminates lies within the interphase boundary itself. As de SOUZA et al. (2021) [18] warned, without sufficient chemical bonding, a ductile film can act as a structural defect rather than a toughening agent, causing premature adhesive failure [18]. This poses a critical challenge for engineering thermoplastics like Polycarbonate (PC) and Polysulfone (PSU). A recent static characterization by Taubert et al. (2026) revealed that while PC and PSU possess high intrinsic ductility and perform well in certain static fracture metrics (such as G_{IIC}), they exhibit distinct weaknesses in Mode I adhesion G_{IC} [27]. Therefore, it is necessary to investigate how these specific static vulnerabilities translate to dynamic crash performance and whether the intrinsic plasticity of the films can compensate for moderate interfacial adhesion under impact.

Investigating the impact behavior of PC and PSU is particularly relevant for the advancement of multi-material structures. Compared to established high-temperature references like PEI and PES, PC and PSU offer significantly lower processing temperatures [39,40]. This thermal characteristic makes them highly suitable candidates for functional integration [41–45]. Successfully establishing an impact-stable interphase with these materials would unlock highly efficient post-cure manufacturing routes, enabling processes such as ultrasonic welding or overmolding directly on the co-cured thermoplastic surface after the primary laminate consolidation [46–51].

Consequently, a distinct research gap exists regarding the dynamic performance threshold of these functionally promising, yet adhesion-limited thermoplastics. This study addresses this gap through a comparative low-velocity impact study within a controlled M18/1 prepreg framework, benchmarking PC and PSU directly against the established PEI and PES references. The specific objectives are defined as follows:

- **Transferability:** to analyze whether the quasi-static fracture toughness (G_{IC} , G_{IIC}) established in prior characterization translates to low-velocity impact performance or if a mechanism shift occurs.
- **Failure morphology:** to correlate the impact damage footprint (C-Scan, microscopy) with the chemical compatibility of the interphase (adhesive vs. cohesive failure).
- **Scalability:** to investigate whether increasing the interleaf density (from every 4th to every 2nd ply) results in a proportional increase in damage tolerance ('Additive Hypothesis') or if saturation effects dominate

2. Materials and methods

In the terminology used throughout this manuscript, interleaf refers to the discrete thermoplastic film inserted between prepreg plies prior to cure, whereas interphase refers to the graded thermoplastic epoxy transition zone formed during curing. The term interlaminar region is used for the geometric location between adjacent plies.

2.1. Thermoset composite baseline

The base laminate employed for this study was HexPly® M18/1, a high-performance carbon fiber/epoxy prepreg supplied by Hexcel [52]. This system was selected due to its representative character for the aerospace sector [53,54]. The chemical formulation of the M18/1 matrix is closely related to the RTM6 resin (used in Resin Transfer Molding) and the HexMC system (used in compression molding) [27,54]. Consequently, results obtained on this prepreg platform offer high transferability to a wider range of structural epoxy manufacturing processes.

2.2. Thermoplastic interleaves and resulting interphases

Four amorphous thermoplastic films were selected as interleaves to functionalize the interlaminar region. Based on the 'compatibility vs. initiation' framework established in the introduction, these materials are categorized into two groups:

Reference group (compatible)

Polyetherimide (PEI) and Polyethersulfone (PES) were chosen as benchmark materials. Both are high-Tg polymers known for their good solubility in epoxy precursors, enabling the formation of graded, inter-diffused interphases during the cure cycle [39,47,54–59].

Challenger group (non-compatible/functional)

Polysulfone (PSU) and Polycarbonate (PC) were selected to represent systems with lower processing temperatures, which is advantageous for post-cure functionalization processes such as ultrasonic welding or overmolding [46–51,60–67]. However, they pose a challenge regarding interfacial adhesion, serving to test the impact response of potentially 'non-compatible' interfaces.

Films were used with a nominal thickness of 125 μm and 175 μm for neat resin characterization, and fixed at 175 μm for laminate integration to ensure a consistent interleave volume fraction.

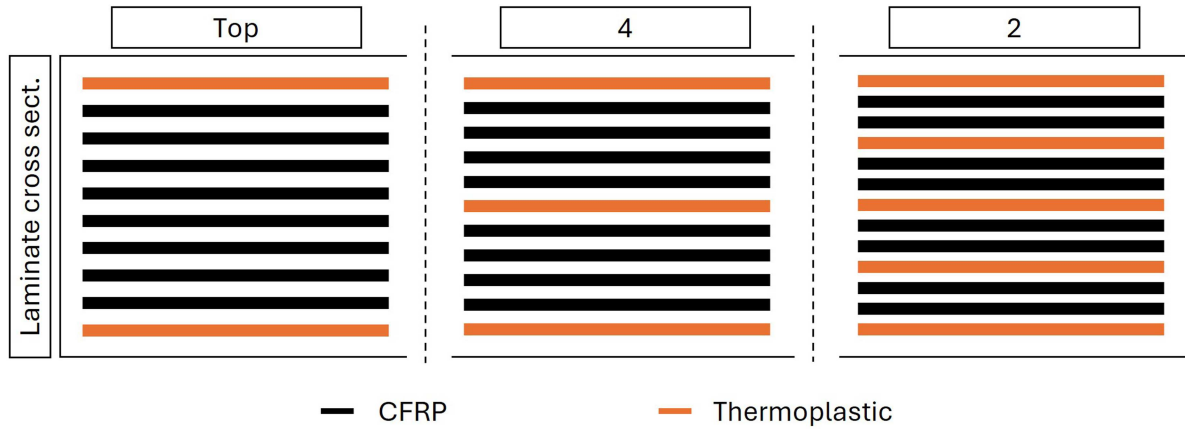
2.3. Preliminary interlaminar characterization

To interpret the impact results in the context of fracture mechanics, the quasi-static baseline properties of the hybrid laminates were established in a prior characterization study by Taubert et al. using the same M18/1 system [27]. Table 1 summarizes the Mode I (G_{IC}) and Mode II (G_{IIC}) fracture toughness values.

As shown in Table 1, while PEI and PES exhibit high toughness driven by cohesive failure, the PC interphase showed limitations in static adhesion (G_{IC}) despite the high intrinsic ductility of the polycarbonate film. The G_{IIC} values for PC, however, remain competitive due to mechanical interlocking effects [27,39].

Table 1. Quasi-static interlaminar fracture toughness G_{IC} and G_{IIC} and failure morphology of the unmodified M18/1 baseline and the interphase-modified laminates [27].

Interphase material	G_{IC} initiation [J/m ²]	G_{IIC} propagation [J/m ²]	Fracture morphology (static)
Reference (M18/1)	~371 [Ref_Lit]	~999 [Ref_Lit]	Brittle/Baseline
PEI (Reference)	590 ± 37	1707 ± 174	Cohesive
PES (Reference)	440 ± 23	6180 ± 334	Cohesive
PSU (Challenger)	161 ± 39	1053 ± 200	Mixed
PC (Challenger)	42 ± 34	1770 ± 274	Adhesive

**Figure 1.** Laminate Layup for 'top', '4' and '2' configuration.

2.4. Laminate configurations and design rationale

Four distinct laminate configurations were manufactured to investigate the influence of interleaf position and interleaf density on the impact response (see Figure 1):

Reference ('Ref') : a pure CFRP laminate without thermoplastic films, serving as the baseline for normalized energy absorption and force comparisons.

Surface integration ('Top') : thermoplastic films were placed exclusively as the outer skins. This configuration assesses the film's ability to suppress surface damage (e.g. fiber breakout) and its suitability for subsequent welding operations.

Distributed integration ('4' and '2') : thermoplastic films were embedded periodically within the stack—either after every 4th ply ('4') or after every 2nd ply ('2'). These two internal configurations were specifically designed to test the 'Additive Hypothesis'. By doubling the interleaf density from configuration '4' to '2', the study investigates whether damage tolerance scales linearly with the number of modified interlaminar regions, or if saturation effects and structural discontinuities dominate the response.

2.5. Manufacturing

For each test series, five specimens were manufactured. The panels were cured in an autoclave. The cure cycles were not chosen arbitrarily but were adopted directly from the process windows defined and optimized in a previous characterization study by Taubert et al. to ensure the reproduction of the graded interphases described above [37]. While a constant heating ramp of 3 K/min was applied to all specimens, the maximum cure temperature and hold time were specifically adapted:

- **High-Tg group (PEI/PES):** cured for 2 h at 180 °C.
- **Low-Tg group (PSU/PC):** cured for 3 h at 160 °C.

Apart from these tailored holding temperatures and times, all other processing parameters were kept consistent and strictly followed the manufacturer's specifications for the M18/1 prepreg system [52]. This

tailored processing ensures that each polymer reaches a comparable state of integration, preventing degradation of the lower-melting films while ensuring sufficient diffusion.

2.6. Impact testing

Low-velocity impact testing was conducted based on the general guidelines of DIN ISO 18352:2017-11, see Figure 2.

However, specific modifications to this standard protocol were deliberately implemented to better capture the global damage response of the interphase-modified laminates rather than highly localized failure modes.

First, a hemispherical steel tup with a 20 mm diameter was employed instead of the standard 16 mm. This slight increase reduces localized contact pressure, promoting a more global bending response of the panel rather than purely localized indentation. Second, the clamping method differed from the ISO standard procedure, which typically requires a fully closed continuous clamping frame. Instead, the specimens were secured using a four-point clamping configuration. This boundary condition is widely established and formally recognized in other prominent aerospace composite testing standards, most notably ASTM D7136/D7136M, where specimens are fixed over a standard cutout using four toggle clamps. This setup provides sufficient restraint to prevent specimen bounce during the impact event while allowing for realistic out-of-plane flexural deformation of the composite panel.

2.7. Ultrasonic inspection (non-destructive testing)

To evaluate the lateral extent of impact-induced delamination, all specimens were subjected to ultrasonic inspection prior to mechanical sectioning. Pulse-echo C-scans were acquired using a HILLGER ImmersionScan FB3-1000 UG system. The scans were performed in reflection mode to map the attenuation of the backwall echo.

As illustrated by the reference laminate (see Figure 3), an adhesive hole reinforcement ring was bonded to the back surface of the panels to act as a known reflector, confirming that the ultrasound signal



Figure 2. Drop tower setup—Instron 9400.

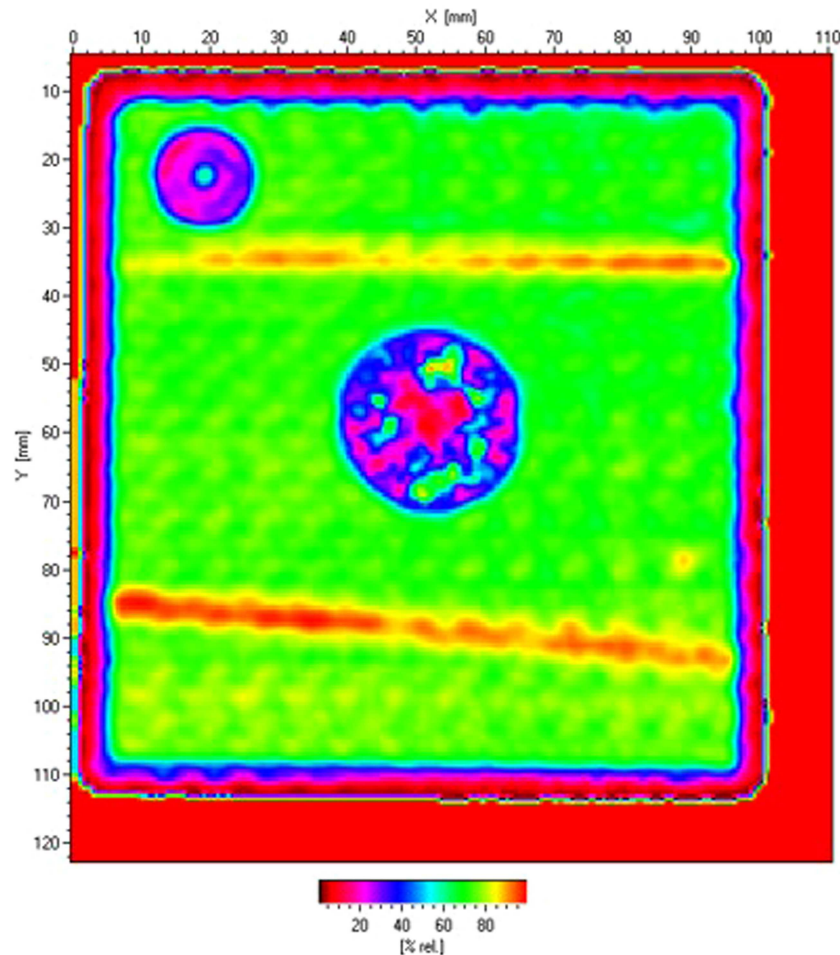


Figure 3. Reference C-Scan image of an impacted M18/1 thermoset panel without thermoplastic interleaf.

successfully passed through the plate. The circular feature at the center corresponds to the impact area and serves as the geometric reference for aligning all subsequent specimens.

In this setup, planar defects that are roughly normal to the sound path, such as interlaminar delamination, disbonding, and layer separations, produce significant amplitude loss and can be reliably detected and sized. However, the signal level can also vary due to fiber anisotropy, local thickness changes, and coupling efficiency. Furthermore, overlapping defects parallel to the propagation direction or fine matrix microcracks below the first reflector make the depth ambiguous due to acoustic shadowing. Consequently, the reported quantity in this study is strictly the projected damaged area ('damage footprint') and not a depth profile.

Finally, the C-scans utilize pseudocolors to encode the relative signal intensity and attenuation. Because these values are normalized for each scan to its specific minimum and maximum values, the colors are only directly comparable within a single scan. They serve to define local defect boundaries but should not be interpreted across different scans as absolute amplitudes.

2.8. Microstructural analysis (destructive testing)

To correlate the macroscopic damage footprint with the local failure mechanisms, cross-sectional analysis was conducted. Specimens were sectioned radially through the impact crater center using a water-cooled diamond saw. To prevent preparation-induced damage, the samples were embedded in a cold-curing epoxy resin system prior to grinding. The cross-sections were ground and polished using a progressive abrasive sequence (up to 1 μm diamond suspension) to achieve a smooth surface for optical microscopy. Analysis focused on

distinguishing between adhesive failure (delamination strictly along the thermoplastic-thermoset boundary) and cohesive failure (cracks propagating through the thermoplastic film or the surrounding matrix), as this distinction is critical for evaluating the chemical compatibility of the interphase.

2.9. High-speed tensile testing (strain rate sensitivity)

In addition to the laminate tests, the strain-rate dependent behavior of the neat thermoplastic films was characterized. This characterization is essential to determine whether the impact performance of the hybrid laminates is driven by the intrinsic viscoelasticity of the film or by the interface quality. Rectangular film strips (15 mm width, 100 mm gauge length) were tested in a pendulum tensile impact configuration using an Instron CEA5T 9050 system. To execute this, one end of the thermoplastic specimen was secured in a stationary clamp, while the opposite end was attached to a movable crosshead.

The swinging pendulum hammer then strikes this crosshead, subjecting the strip to an instantaneous high-strain-rate tensile loading until rupture. This setup imposes a high-speed tensile pulse at strain rates comparable to the low-velocity impact event (approx. $30\text{--}70\text{ s}^{-1}$). These data serve as a comparative baseline to assess the dynamic tensile response of the neat films and to distinguish the intrinsic deformability of the thermoplastic films from the structural effects of laminate integration.

Force time and displacement time signals were recorded. The displacement at maximum force $s_{F_{\max}}$ and the time to maximum force $t_{F_{\max}}$ were obtained from the recorded signals. The nominal strain rate at maximum force was estimated as

$$\dot{\epsilon} = \frac{s_{F_{\max}}}{t_{F_{\max}} l_0}, l_0 = 100 \text{ mm}.$$

Averaged values over repeated tests are reported.

2.10. Data reduction and analysis

The impact energy (E) was calculated based on the drop mass (m), gravitational acceleration (g), and drop height (h) according to $E = mgh$. From the force-time data recorded by the piezoelectric load cell, the force-displacement curves were derived via double integration of the acceleration signal. Key performance indicators were extracted from these curves to quantify the structural response and the toughening effect, see Figure 4.

- **First failure load:** defined as the first significant load drop in the rising branch of the force-displacement curve, indicating the onset of macroscopic damage, such as matrix cracking or initial delamination (damage resistance).

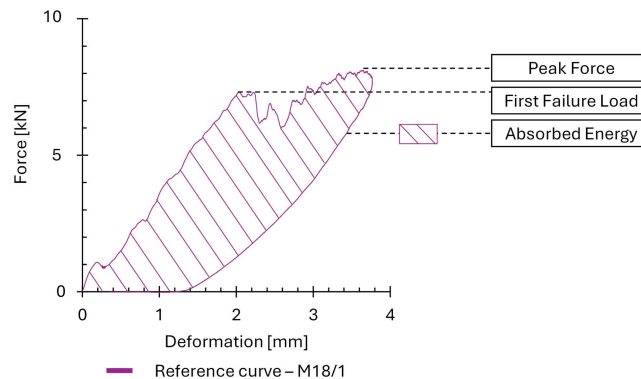


Figure 4. Derivation of key performance parameters from the experimental force-deformation history of the baseline laminate.

- **Peak force:** the maximum force recorded during the impact event, serving as an indicator of the ultimate load-bearing capacity of the laminate.
- **Absorbed energy:** calculated as the area enclosed by the force-deformation hysteresis loop. This is determined by continuously integrating the force over the displacement over the entire impact cycle, which inherently subtracts the elastic energy recovered during the rebound phase from the total energy introduced during the loading phase. For highly compliant configurations undergoing global delamination (such as PC), where the prolonged rebound phase exceeds the machine's recording time window or triggers the anti-rebound catch mechanism, the integration is evaluated up to the end of the recorded event. In all cases, this value accurately represents the energy dissipated through irreversible processes like interlaminar fracture, friction, and plastic deformation.

3. Results

3.1. Impact energy absorption

The impact energy absorption (referenced to the unmodified baseline) reveals a clear material-dependent grouping, as summarized in Figure 5.

Compatible systems (PEI, PES)

Overall, the established reference interphases (PEI and PES) exhibit a near-baseline response. For PES, the 'top' configuration showed negligible changes (+2% to +3%), remaining essentially at the unmodified baseline level. The internal configurations (Layup '4') resulted in a slightly higher energy absorption of approximately +4% to +5%. Similarly, PEI produced minor increases of +4% to +5% for its internal layers. These modest variations confirm that these cohesive interphases do not significantly alter the global energy dissipation mechanism of the laminate, but maintain its structural integrity.

Challenger systems (PC, PSU)

In contrast, the 'Challenger Group' shifts towards higher energy dissipation. PC internal layers ('2' and '4') resulted in the highest absorption gains of approximately +10%. Notably, the 'top' configuration for PC showed no significant change (near 0%), confirming that the effect is driven by internal delamination. PSU occupies an intermediate position with an increase of about +6% for internal layers. While it absorbs more energy than the stiff reference systems, it does not reach the levels of PC, indicating a less effective activation of plastic reserves.

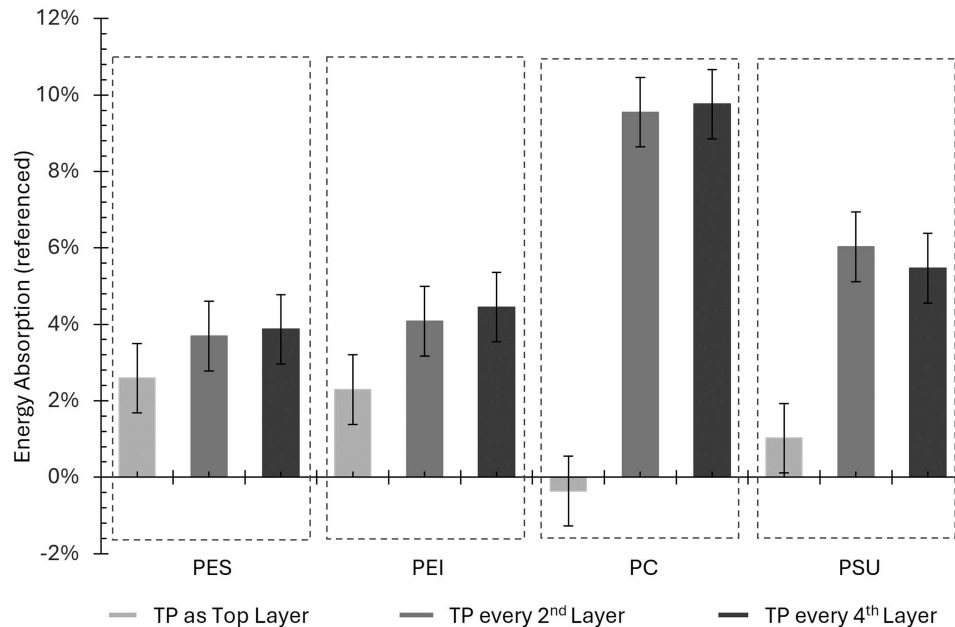


Figure 5. Energy absorption in contrast to the reference configuration for hybrid laminates of CFRP + PES/PEI/PC/PSU.

3.2. Peak force feedback

The peak force analysis (Figure 6) highlights the trade-off between stiffening and softening.

Reinforcement (PEI)

PEI demonstrates a distinct reinforcing effect. Internal placement ('2' and '4') increased the peak force by +9% to +10% compared to the reference. Even the 'top' layer showed a slight positive trend. This indicates that the PEI interphase effectively delays the onset of damage.

Neutral/saturation (PES)

PES behavior is sensitive to layer density. While the '4' layup showed a positive trend (+4% to +5%), the denser '2' configuration dropped slightly below the reference (−1% to −2%). This suggests a saturation effect. The additional PES interleaves do not appear to destabilize the interface itself, but they likely increase the compliance of the interlaminar architecture. As a result, part of the local toughening benefit is offset by a reduction in the effective flexural stiffness and load transfer efficiency of the laminate.

Softening (PC, PSU)

For the functional thermoplastics, a significant drop in load-bearing capacity was measured. PC internal layers caused a drastic reduction in peak force by −23% to −26%, effectively softening the laminate. PSU showed a distinct reduction of −11% to −17% for internal layers. Importantly, the 'top' configurations for both PC and PSU remained close to the reference (near 0%), proving that the loss of strength is a result of internal interlaminar failure and not an intrinsic material property of the surface ply.

3.3. Force-deformation history—all systems

The force-displacement curves allow to pinpoint the damage initiation threshold (First Failure Load), see Figure 7.

For PEI, the reinforcement effect is robust. In the '4' layup, the force at first failure increased by ~23% (from ~7200 N to ~8850 N) compared to the reference. The deformation at failure also extended by ~15% (to 2.3 mm). The '2' layup achieved a nearly identical peak force (~8800 N) but allowed for slightly larger deformation (+25%, ~2.5 mm). PES showed its best performance in the '4' layup (+18% force, ~8500 N). However, in the '2' layup, the first failure load dropped to ~7950 N (+10%), confirming the saturation tendency observed in the peak force data. Both materials show a stepwise post-peak decay rather than a collapse.

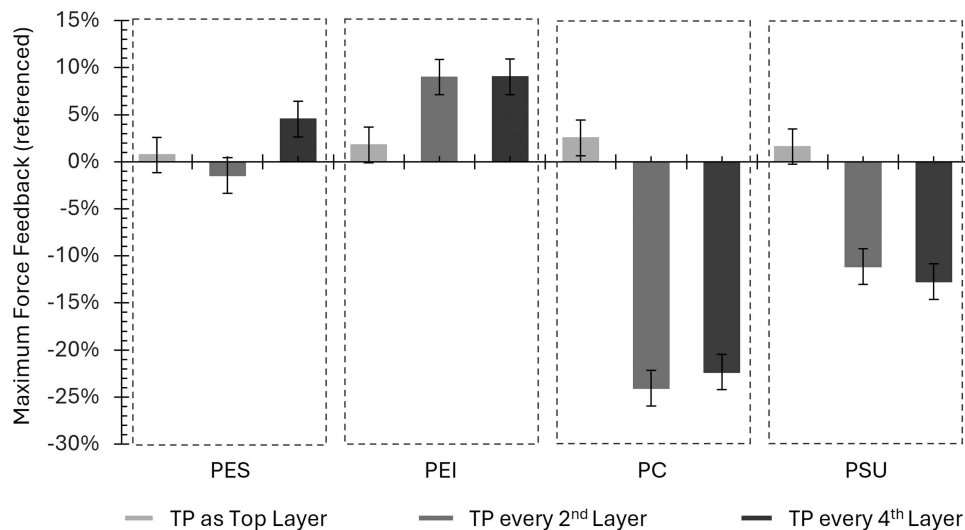


Figure 6. Maximum force feedback in contrast to the reference for hybrid laminates of CFRP + PES/PEI/PC/PSU.

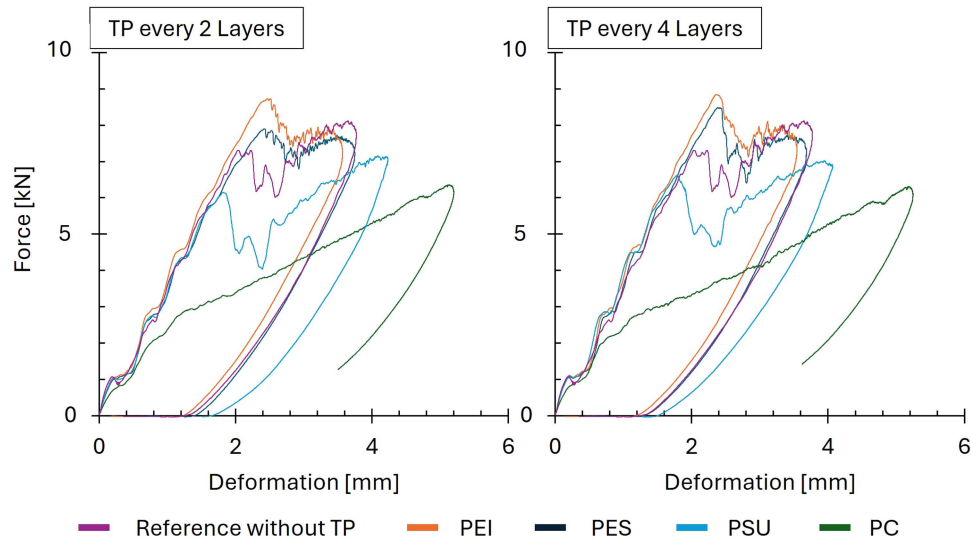


Figure 7. Impact force versus deformation for laminates with thermoplastic films every two or four CFRP plies.

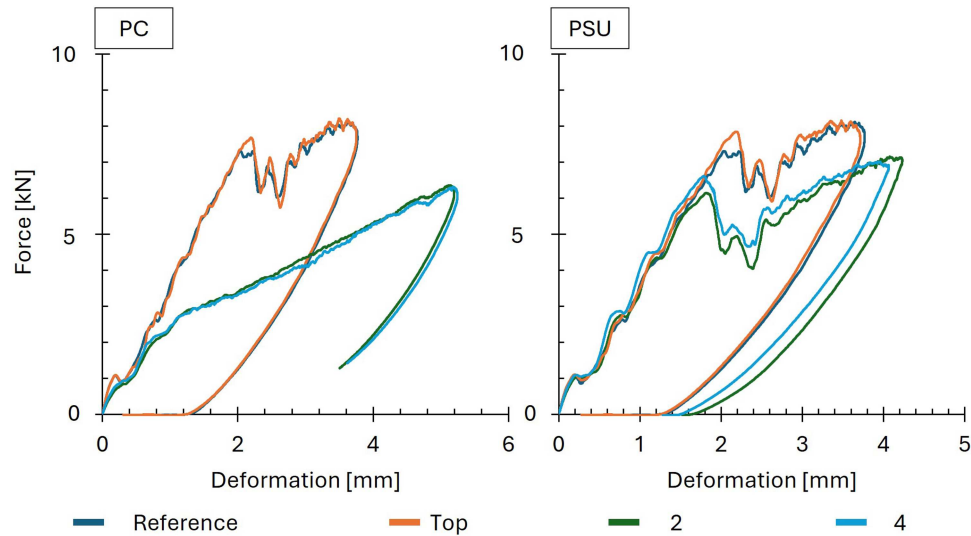


Figure 8. Impact force versus deformation for hybrid laminates with CFRP + PC or PSU with TP layers on top or every two or four CFRP plies.

3.4. Force-deformation history—incompatible systems

PC and PSU exhibit fundamentally different curve signatures characterized by premature initiation and increased compliance, see [Figure 8](#).

PC leads to a very early first damage event (at ~50%–60% of reference deformation), followed by a long, low-slope rising branch. This ‘pseudoplastic’ plateau explains the high energy absorption despite the low peak force. PSU reveals a critical weakness: The first force drop occurs early at ~6500 N (–10% vs. Reference) and at a low deformation of 1.8 mm. The denser ‘2’ layup further reduced the initiation load to ~6100 N (–15%). Unlike PC, which recovers into a high-energy plateau, PSU shows only moderate recovery. This confirms that PSU exhibits an inefficient performance compromise in this configuration: It fails earlier than the reference (lower initiation) and carries less load than PC (lower propagation resistance).

3.5. Damage morphology comparison

The C-Scan analyses (Figure 9) quantify the lateral extent of the damage.

PEI & PES (containment)

Both reference materials exhibit compact, sharply defined damage footprints. The affected area is barely larger than in the unmodified reference, confirming successful damage containment.

PC (global failure)

PC panels show area-wide signal attenuation (loss of backwall echo) across almost the entire scan field. This indicates global delamination and explains the ‘soft’ global response.

PSU (intermediate)

PSU displays a damage area significantly larger than PEI/PES, with diffuse ‘lobes’ extending from the center. This correlates with the reduced peak force and confirms that the interphase fails to contain the crack as effectively as the reference materials.

3.6. Layup influence on damage

Comparing the ‘2’, ‘4’, and ‘Top’ layups (Figure 10) confirms that internal placement is the driver for the observed effects.

For all materials, the ‘Top’ configuration shows a C-scan image very similar to the reference, confirming that surface films alone do not drastically alter the internal damage mechanics. For the compatible systems PEI and PES, the damage footprint remains remarkably consistent when increasing the interleave density from the ‘4’ to the ‘2’ configuration. This lack of damage expansion indicates that the interfacial adhesion is highly stable. It suggests that the surrounding thermoset matrix acts as the limiting factor, failing prior to the interface and thereby dictating the ultimate size of the damage area. In combination with the reduced force response of the PES ‘2’ configuration, this indicates that the performance decline is more likely related to increased structural compliance than to a loss of interfacial stability.

In contrast, for PSU, both ‘2’ and ‘4’ layups show large-area attenuation. Interestingly, increasing the density to ‘2’ does not reduce the damage size but tends to consolidate the large footprint, supporting the finding that PSU layers act as structural weak points.

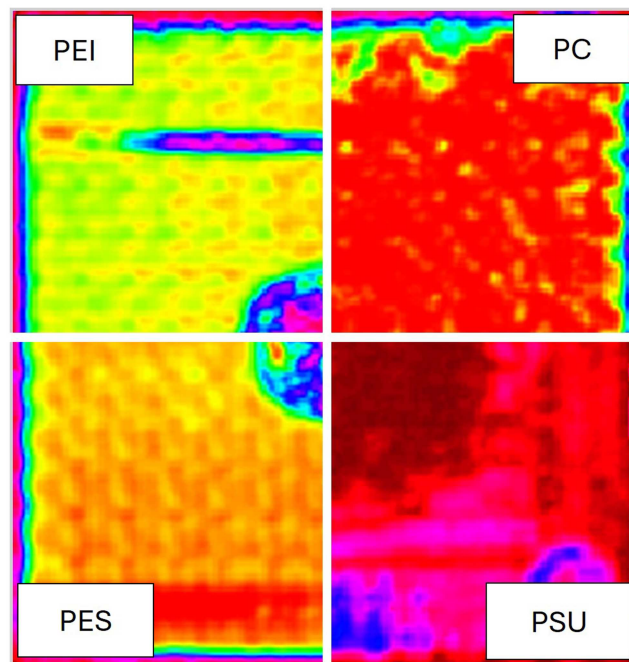


Figure 9. C-Scan of impacted CFRP/thermoplastic hybrids with thermoplastic layers every 4 plies.

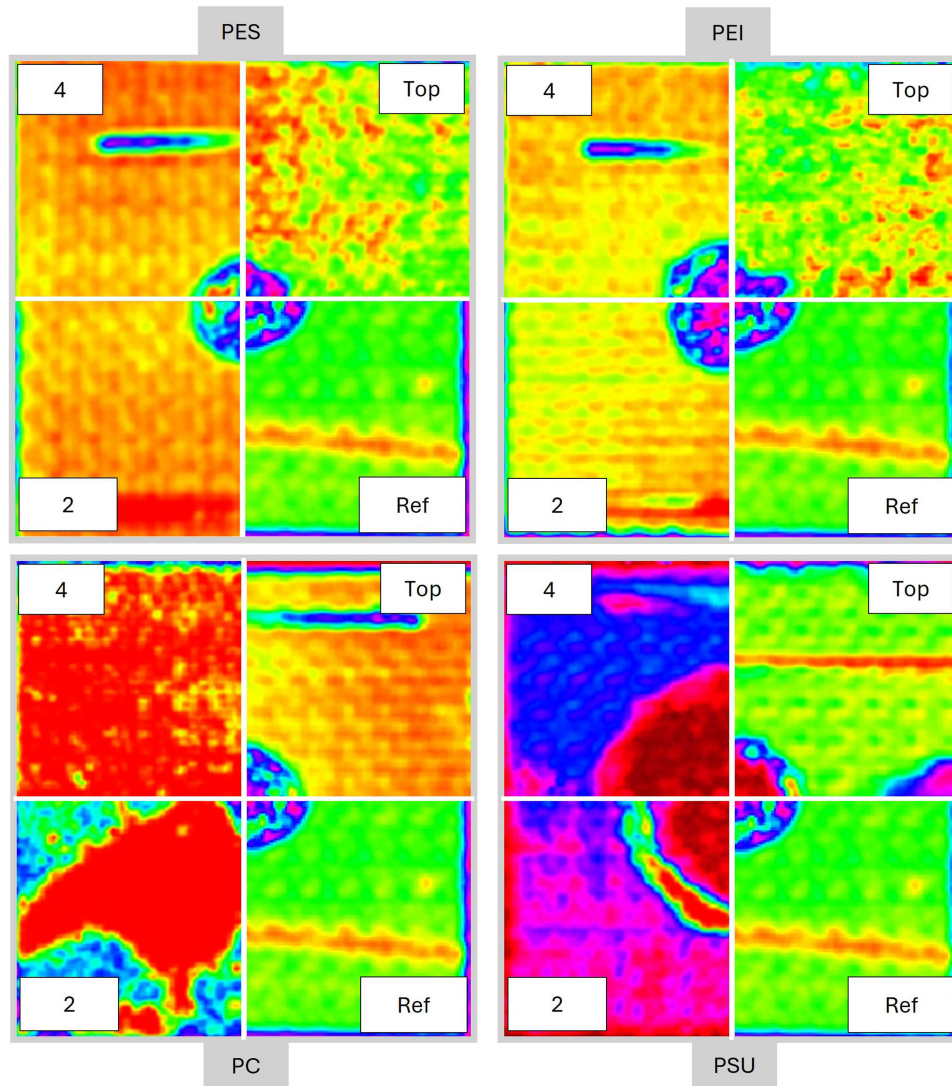


Figure 10. C-scan of impacted CFRP thermoplastic hybrids with thermoplastic layers every 2 plies ('2'), every 4 plies('4'), as top layers('top'), and with no thermoplastic layers ('Ref').

3.7. Microstructural analysis—cohesive behavior

Microscopy of the compatible systems (PEI and PES) provides the physical explanation for their high damage initiation thresholds and force values. Figure 11 depicts a representative cross-section for the PES configuration; however, it must be emphasized that PEI exhibits a highly comparable microstructural damage morphology. For both materials, the microscopic analysis reveals that the interphase integration remains intact even under peak impact loads.

Instead of a continuous adhesive failure along the thermoplastic-thermoset boundary, the damage manifests in two primary cohesive modes:

1. **Transverse** micro-cracking: when a primary flexural or shear crack reaches the interphase, it is effectively arrested or deflected. Fine cracks propagate transversely through the thermoplastic layer (visible in Figure 11, dashed blue box), indicating that the failure stress exceeded the cohesive tensile strength of the film before the interfacial bond could fail.
2. **Matrix cohesive failure:** delaminations are confined to the fiber-reinforced epoxy regions (visible in Figure 9, dashed green box) and do not coalesce along the interphase boundary.

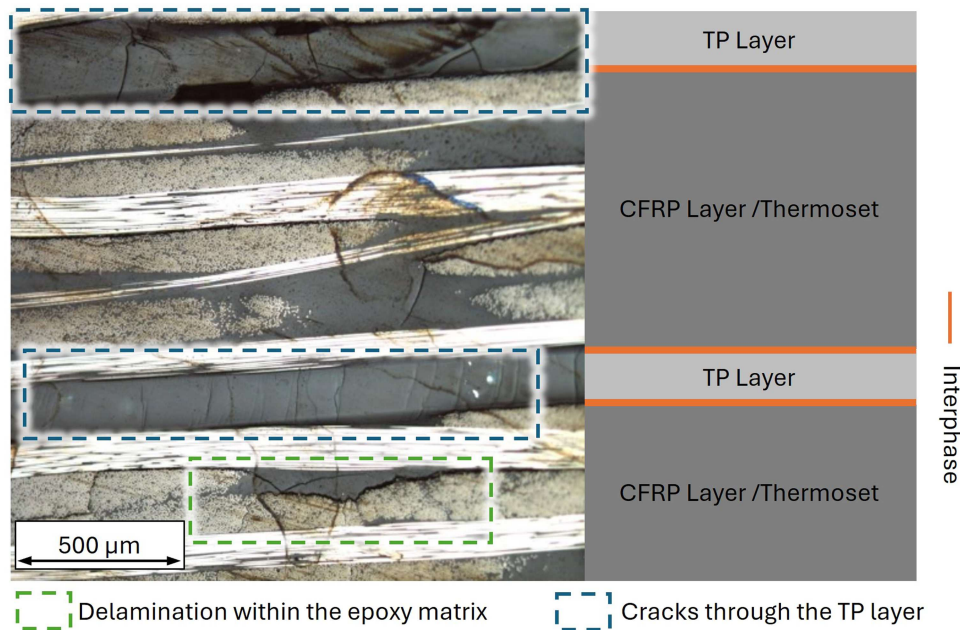


Figure 11. CFRP/PES hybrid material with PES every four CFRP plies – microsection center.

The observation that cracks run through the thermoplastic layer or branch into the composite matrix, rather than delaminating the interface, demonstrates that the reactive interphases formed by both PES and PEI provide superior adhesion. This complex, tortuous crack path proves that the interfacial adhesion strength locally exceeds the cohesive strength of the bulk materials. This forces the damage path into the constituent materials, thereby activating energy dissipation mechanisms within the thermoplastic film and preventing the ‘clean’ separation often observed in weakly bonded insertions.

3.8. Microstructural analysis—stress-dependent mixed failure modes (PSU)

In contrast to the highly stable PES and PEI configurations, the PSU hybrid reveals a distinct shift in the dominant failure mode. This exposes a hybrid behavior that is highly sensitive to the local stress intensity. This spatial dependence becomes evident when analyzing the damage morphology at different distances from the impact center.

High-stress zone (impact center)

Figure 12 illustrates the microstructural damage in the direct center of the impact area, where maximum transverse shear and out-of-plane tensile loads occur.

The micrograph highlights two competing damage mechanisms:

1. **Interfacial delamination:** a large portion of the crack path propagates directly along the boundary between the thermoplastic film and the thermoset matrix. This indicates the onset of adhesive failure, directly explaining the premature first force drop observed in the macroscopic impact curves (~6500 N).
2. **Localized cohesive failure:** simultaneously, isolated regions exhibit cracks running through the thermoplastic layer or within the epoxy matrix.

Transition to cohesive failure (low-stress periphery)

Moving away from the center into regions of lower stress intensity, the failure mechanism shifts fundamentally. Figure 13 shows the cross-section of the same PSU configuration at a distance of 10 mm from the impact center. In this peripheral zone, the local stress levels fall below the critical adhesive strength threshold of the PSU-epoxy bond. Consequently, the cracks no longer propagate along the interface. Instead, they are confined to the thermoplastic film (transverse micro-cracking) or the

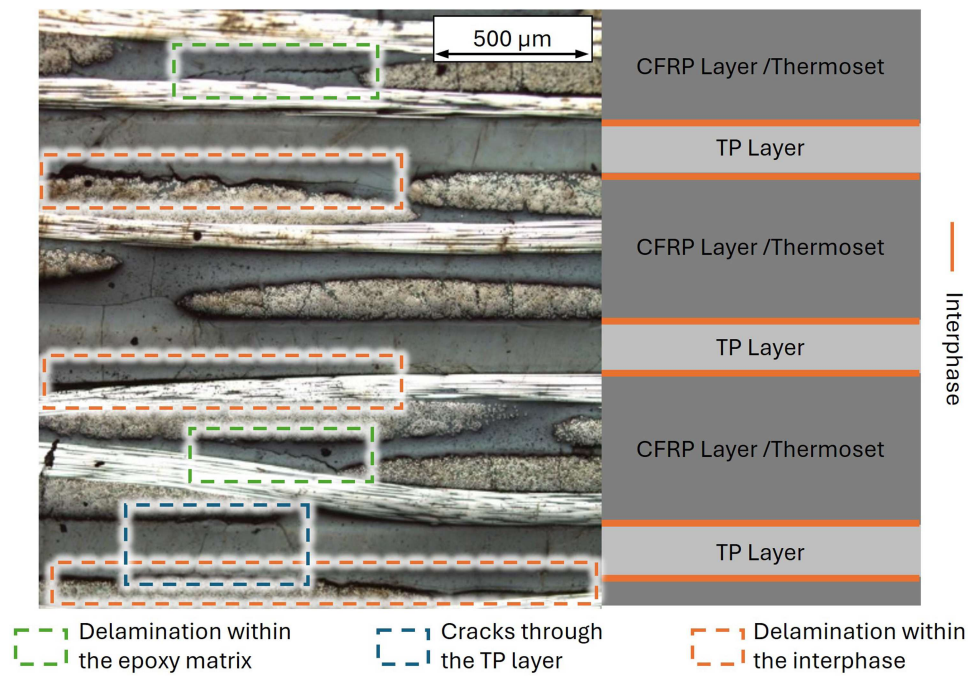


Figure 12. CFRP/PSU hybrid material with PSU every two CFRP plies – microsection—center.

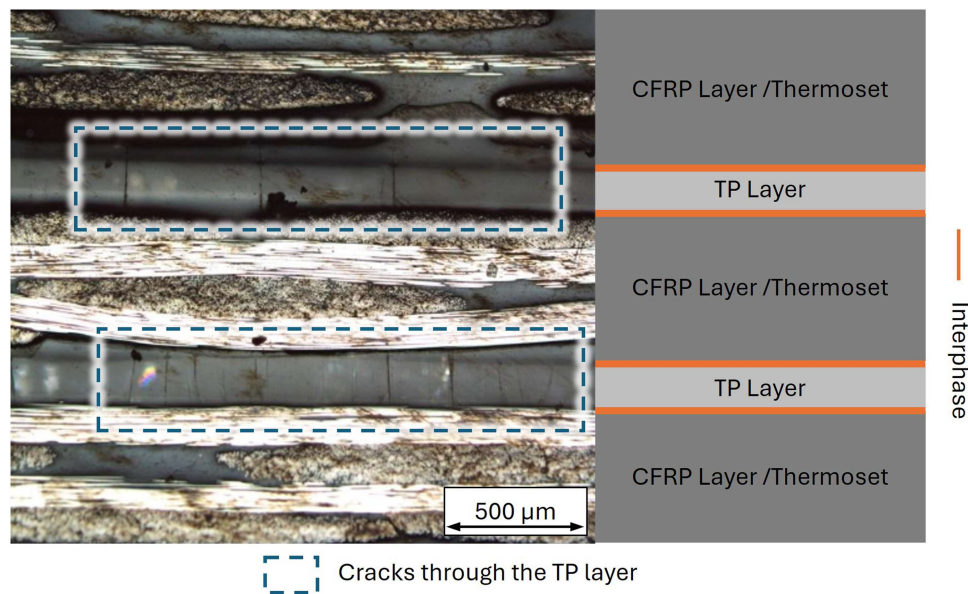


Figure 13. CFRP/PSU hybrid material with PSU every two CFRP plies microsection – 10 mm off-center.

surrounding epoxy matrix, effectively mirroring the cohesive failure mode observed in the highly compatible reference systems (PEI/PES).

Synthesis of PSU performance

This spatial transition between the center (Figure 12) and the periphery (Figure 13) implies that the PSU interphase is not globally non-functional (unlike the complete decoupling observed for PC). Instead, it is strictly stress-limited.

It provides sufficient interfacial adhesion to remain locally connected under moderate loading, but it decisively debonds under peak impact stresses in the center. This spatially heterogeneous failure is crucial:

unlike PC, PSU does not undergo immediate global decoupling of the laminate. As a result, no continuous interlaminar sliding path is created across the damaged zone. The remaining bonded and cohesively cracked peripheral regions interrupt large-scale relative ply movement and thereby suppress the development of a PC-like friction-dominated plateau. Consequently, PSU combines premature local adhesive failure with insufficient global sliding, which explains why its intrinsic film ductility is not converted into an equally effective macroscopic energy dissipation mechanism. This microstructural evidence directly supports the classification of PSU as an inefficient structural compromise.

3.9. Global decoupling (PC)

For PC, no stable microsections could be prepared as the specimens disintegrated upon cutting. This physical instability serves as a direct validation of the C-scan results, confirming a total loss of interfacial integrity (global adhesive failure) throughout the impacted region.

3.10. Strain rate sensitivity of neat films

To assess the intrinsic contribution of the thermoplastic films under dynamic loading, tensile impact tests were conducted on the neat films. The resulting values are summarized in Table 2. Overall, the data indicate a clear ranking in rate-dependent deformation capacity. PC showed the highest displacement at maximum force and the highest nominal strain rate, indicating the most compliant and dynamically deformable response among the investigated films. PES exhibited the lowest values, consistent with the stiffest response, while PEI and PSU ranked intermediately.

This ranking is important because it demonstrates that the neat films do not fail in the same order as the corresponding hybrid laminates under impact. In other words, the laminate response cannot be explained by intrinsic film deformability alone. If that were the case, the dynamically most compliant film should also provide the structurally most effective toughening. This is not observed. PC indeed exhibits the highest dynamic deformability as a neat film, but in the laminate it simultaneously shows the strongest loss of structural integrity due to immediate interfacial decoupling. Conversely, PEI and PES maintain the highest structural stability under impact despite their lower intrinsic dynamic deformability.

The comparison therefore indicates that strain rate effects in the film act only as a secondary amplifier once the interface remains intact. For compatible systems such as PEI and PES, the interphase is strong enough to transfer the impact induced stresses into the surrounding material, so that cohesive cracking and stable damage containment dominate. For weakly bonded systems such as PC, the interface opens before the intrinsic plastic reserve of the film can be converted into structural toughening. In that case, the film ductility contributes only after debonding, mainly through large scale sliding and frictional dissipation, but no longer through load carrying interlaminar reinforcement. PSU occupies an intermediate regime, where the bond is too weak to preserve structural continuity under peak loading, yet too strong to enable the fully developed sliding mechanism observed for PC.

These results support the conclusion that the impact response of the hybrid laminates is governed primarily by interphase integrity rather than by the intrinsic rate dependent deformability of the neat thermoplastic films alone.

Table 2. Strain rate measurement results of PC, PSU, PEI and PES.

	t [mm]	$S_{F_{\max}}$ [mm]	$t_{F_{\max}}$ [ms]	$\dot{\epsilon}$ [s ⁻¹]
PC	0.125	13.2	2.2	60.0
PC	0.175	13.8	2.1	65.7
PEI	0.125	11.4	2.3	49.6
PEI	0.175	11.9	2.2	54.1
PES	0.125	9.5	2.6	36.5
PES	0.175	10.1	2.5	40.4
PSU	0.125	10.7	2.4	44.6
PSU	0.175	11.2	2.3	48.7

4. Discussion

4.1. Transferability of quasi-static toughness to impact performance

A central objective of this study was to evaluate whether quasi-static fracture toughness (G_{Ic} , G_{IIc}) serves as a reliable predictor for low-velocity impact resistance in hybrid laminates. While Chou et al. postulated that Mode II propagation toughness (G_{IIc}) is the governing factor for impact resistance in conventional laminates, our results indicate that for interphase-modified systems, Mode I initiation toughness (G_{Ic}) plays the decisive role. Correlating the baseline characterization (Table 1) with the obtained impact results reveals three distinct regimes supporting this hypothesis.

The tensile impact results of the neat thermoplastic films further clarify why this transferability remains limited. The dynamic ranking of the films, with PC showing the highest deformation capacity and PES the lowest, does not match the structural ranking observed in the laminate impact tests. This mismatch demonstrates that the strain-rate sensitivity of the neat film is not the primary controlling parameter for laminate impact performance. Instead, the rate-dependent plasticity of the film can only contribute beneficially if the interphase remains intact during the early opening-dominated stage of the impact event. Once the interface fails prematurely, the intrinsic dynamic ductility of the film is no longer translated into structural reinforcement, but only into post-debonding deformation or frictional dissipation.

In this context, Lai et al. showed for CNT (carbon nanotube)-toughened CFRP laminates that improved interlaminar fracture toughness can translate into higher peak force, lower displacement, and smaller impact damage, provided that stress transfer across the interlaminar region is maintained [38]. Liu et al. likewise emphasized that the interphase governs stress transfer and strongly affects the overall macroscopic behavior of CFRP systems. The present results show that this transferability breaks down once the interphase integrity becomes the limiting factor [37].

This is also consistent with broader interlaminar toughening studies. Wang et al. reported for modified CFRP laminates that the increase in fracture toughness can be more pronounced in Mode II than in Mode I. In other words, shear-dominated crack propagation often benefits more strongly from interlaminar modification than opening-dominated crack initiation. The present results show, however, that this benefit cannot be fully exploited if the interphase fails already at low Mode I loading [68].

The compatible regime (PEI, PES)—diminishing returns of G_{IIc}

For the reference materials PEI and PES, a strong correlation between high static Mode I toughness and impact containment was observed. Both materials exhibited high G_{Ic} initiation values (590 J/m² and 440 J/m², respectively). Under impact, this translated into confined damage footprints. However, a critical deviation from a purely G_{IIc} -driven model is observed here. Although PES exhibited a vastly superior propagation toughness ($G_{IIc} \approx 6180$ J/m²) compared to PEI ($G_{IIc} \approx 1707$ J/m²) in the static characterization, this 3.6-fold increase did not translate into a proportional gain in impact energy absorption. Both materials performed on a similar level (+4%–5%). This suggests a law of diminishing returns: Once the interfacial toughness exceeds a certain threshold, the failure mode shifts away from the toughened interlaminar region into the intralaminar fiber/matrix interface or results in fiber breakage. Consequently, the higher G_{IIc} of PES provides no additional benefit. Instead, the high G_{Ic} of PEI appears to be sufficiently effective in suppressing the initial delamination opening during the early contact stage of impact, confirming that Mode I adhesion acts as the ‘gatekeeper’ at damage initiation. Once this early opening is suppressed, the laminate can maintain structural continuity into the subsequent flexural and shear dominated stage, where additional energy dissipation mechanisms can be activated. In this sense, G_{Ic} is decisive primarily because it prevents early shock induced decoupling and thereby preserves the structural conditions required for later stage toughening.

The mechanism shift (PC)—adhesion as a precondition

For Polycarbonate (PC), the dominance of G_{Ic} becomes even more apparent. Despite the PC film possessing the highest intrinsic ductility and high static Mode II toughness (G_{IIc}), the laminate impact performance was dominated by its low static Mode I adhesion ($G_{Ic} \approx 42$ J/m²). The impact response shifted from a ‘toughness-controlled’ mode to an ‘interface-controlled’ mode. This refines the postulate by Chou et al. [69]: High intrinsic toughness and G_{IIc} are rendered irrelevant if the G_{Ic} initiation threshold is too low.

The interface fails in Mode I opening already during the early contact stage, before the laminate can enter a structurally continuous late-stage response. As a result, the Mode II sliding mechanisms and the film's plasticity are not activated as structural toughening mechanisms, but only after global decoupling has already occurred. This strongly supports the findings of de Souza et al. [18], emphasizing that chemical compatibility is the prerequisite for any toughening mechanism.

The inefficient performance compromise (PSU)

The results for Polysulfone (PSU) further validate the criticality of G_{Ic} . PSU sits in a suboptimal intermediate regime. Its static values suggested moderate adhesion, yet under impact, it exhibited a conflicting property profile:

- **Insufficient mode I adhesion:** unlike PEI/PES, the PSU interface is not strong enough to transmit vertical tensile loads during the contact phase, leading to an early force drop (−17%) and adhesive failure in the center.
- **Insufficient slip:** unlike PC, the adhesion is not *low enough* to allow for immediate, global decoupling.

Thus, PSU combines the structural weakness of an incompatible system without gaining the full benefits of a sacrificial sliding layer. In contrast to PC, whose very low G_{Ic} enables immediate large-scale decoupling and frictional sliding, the PSU interface fails only locally in the high-stress center while remaining partly bonded in the periphery. This stress-limited transition prevents the formation of a continuous sliding interface and therefore suppresses the activation of a PC-like frictional dissipation mechanism. At the same time, the local adhesive failure is already sufficient to reduce the load-bearing capacity of the laminate. PSU therefore represents a true intermediate regime: too weak to maintain structural continuity under peak loading, yet too strong to enable fully developed pseudoplastic decoupling.

4.2. Failure mode framework

To classify these divergent behaviors, the observed damage patterns were reduced to two recurring failure classes, schematically summarized in Figure 14.

Class A: adhesive/interphase-controlled failure (left in Figure 14)

Here, the crack propagates mainly along the thermoplastic-epoxy boundary. The interphase acts as a predetermined breaking point. This was observed for PSU in the high-stress damage center and for PC across the entire panel. This mode validates the warning by de Souza et al. that without sufficient chemical bonding, the interleaf acts as a structural defect.

Class B: cohesive/intralaminar failure (right in Figure 14)

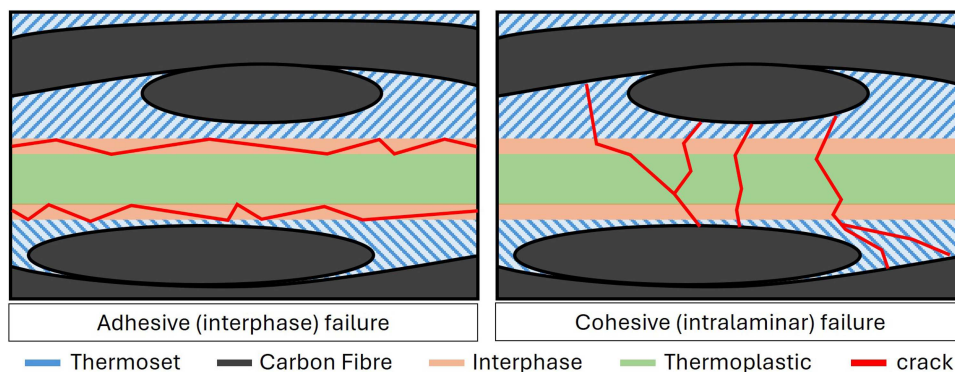


Figure 14. Schematic failure classes under impact: adhesive delamination along the interphase (left) versus cohesive intralaminar cracking through film and matrix (right).

Here, the interphase remains intact. Damage is forced into the thermoplastic ply (transverse cracking) and into the epoxy-rich regions. This mode, observed for PEI and PES, is the prerequisite for effective toughening, as it engages the bulk material properties of the film.

4.3. The special case of pc: pseudoplasticity via decoupling

While PC fits the ‘Non-Compatible’ classification due to its adhesive failure, its macroscopic response differs from the brittle failure often associated with weak bonds. The force-displacement curves showed a long, low-slope plateau.

This suggests that the PC mechanism is not simple brittle delamination, but pseudoplasticity driven by friction. Upon impact, the weak interface fails immediately (shock-induced delamination). However, the tough PC film remains physically present between the plies. As the laminate deforms, the separated layers slide against the ductile film. This large-scale frictional sliding, combined with the unrestricted deformation of the decoupled film, dissipates significant energy (+10% vs. reference). Thus, PC acts as a ‘sacrificial layer’: It sacrifices structural integrity (stiffness) to gain energy absorption through global decoupling.

The tensile impact data of the neat PC film show that the material itself possesses a high dynamic deformation capacity. However, this intrinsic rate-dependent ductility is activated only after interfacial failure has already occurred. Therefore, the strain-rate sensitivity of PC does not stabilize the laminate against damage initiation, but instead amplifies the post-debonding sliding response of the already decoupled structure. Thus, the decisive role of G_{Ic} in the PC system is not to govern the amount of late-stage dissipation, but to determine whether the laminate reaches that stage in an intact load carrying state at all.

This interpretation is consistent with the general interphase framework described by Liu et al., who highlighted that the quality of the interphase governs whether local material properties can be transferred into global composite performance [37]. In the present case, the high intrinsic deformability of PC remains structurally ineffective during damage initiation because the weak interface fails first.

From an engineering perspective, this failure mode must be regarded as critical for the residual post-impact load-bearing capability of the laminate. The observed global decoupling implies that the sub-laminates can no longer effectively stabilize each other against local buckling under subsequent compressive loading. In that sense, the high energy absorption of PC is achieved at the expense of a severe loss of structural integrity. This interpretation is consistent with the general relationship reported by Lai et al., who showed that laminates with lower impact damage maintain higher residual compressive strength after impact [38]. By contrast, the near-complete loss of interfacial continuity observed here for PC would be expected to cause a pronounced penalty in compression-after-impact performance. Accordingly, the present PC concept is more appropriately interpreted as a sacrificial crash mechanism than as a damage-tolerant solution for primary load-bearing laminate structures.

4.4. Scalability: rejection of the additive hypothesis

The comparison between the ‘4’ (midplane only) and ‘2’ (distributed) configurations aimed to test whether damage tolerance scales additively with the number of interleaves. The data suggests that saturation or even negative effects occur.

The PES results indicate that this decline is not primarily caused by a loss of interfacial adhesion. The compact C-scan damage footprint of the denser ‘2’ configuration argues against a fundamentally weakened interphase. Instead, the data suggest an architectural penalty: additional compliant interlaminar zones locally reduce the bending stiffness of the laminate and alter stress redistribution during impact. As a result, the toughening potential of the PES interphase is no longer converted proportionally into higher structural force levels. A similar configuration sensitivity was reported by Zhang et al., who showed that the toughening efficiency of hard-soft layered laminates depends strongly on the selected architecture and crack path rather than on the mere presence of a compliant interlaminar phase [70].

For the compatible system PES, increasing the density from ‘4’ to ‘2’ resulted in a slight drop in peak force. For the incompatible system PSU, the denser ‘2’ layout consolidated the large damage footprint.

This supports the ‘defect density’ theory proposed by Fotouhi et al. [17]. Every interleaf introduced into the laminate is a potential discontinuity in the fiber architecture. For adhesion-limited materials, adding more layers simply increases the probability of introducing a critical flaw in the high-stress zone. Therefore, the hypothesis that ‘more is better’ must be rejected. Targeted placement (e.g. protecting the neutral axis or surface) appears to be a more efficient strategy than global volume modification.

5. Conclusion & outlook

5.1. Conclusion

This study correlates the low-velocity impact response of thermoplastic-interleaved M18/1 laminates with their fracture mechanics properties. The results reveal a clear functional distinction between compatible and incompatible interphases:

Significant suppression of damage initiation (PEI/PES)

The chemically compatible systems (PEI/PES) proved highly effective in reinforcing the laminate against initial failure. By bridging the stiffness mismatch, PEI increased the First Failure Load by up to +23% compared to the reference. This demonstrates that cohesive interphases are a valuable tool for enhancing damage resistance, effectively extending the operational range of the composite before the onset of delamination.

The energy dissipation limit

While damage initiation was successfully delayed, the total energy absorption of compatible systems plateaued (+4%–5%). This indicates a physical limit: Once the interphase toughness exceeds the matrix toughness, failure shifts to the intralaminar epoxy, preventing further gains. Consequently, massive increases in energy dissipation (>10%) are only achievable through a mechanism shift to pseudoplasticity (as seen in PC), which inevitably comes at the expense of structural integrity and would be expected to severely impair residual post-impact compressive performance.

Inefficient compromise of intermediate systems

Polysulfone (PSU) represents an inefficient structural compromise. It suffered from premature adhesive failure (reducing load capacity) without achieving the full decoupling benefits of PC. This confirms that intermediate compatibility results in a ‘worst of both worlds’ scenario, lacking both the stiffness of PEI and the ductility activation of PC.

Dominance of mode I adhesion (G_{Ic})

Contrary to models prioritizing Mode II propagation (G_{IIc}), Mode I initiation toughness was identified as the ‘gatekeeper.’ Its decisive role lies primarily in suppressing the early opening driven decoupling of the laminate during impact initiation and thereby preserving structural continuity during the subsequent response. However, this does not mean that high G_{IIc} automatically translates into proportionally higher impact performance. As shown by PES, high G_{IIc} values can yield only diminishing returns once the response becomes limited by matrix dominated damage processes. In contrast, low G_{Ic} in PC led to immediate decoupling before these later stage mechanisms could contribute to structural toughening. Thus, a high Mode I threshold is the prerequisite for structural toughening.

Rejection of the additive hypothesis

Doubling the interleave density did not scale damage tolerance linearly. Instead, saturation effects occurred for compatible systems, while defect density increased for adhesion-limited systems. Thus, targeted placement is superior to global placement.

5.2. Outlook

Based on the identification of two distinct failure modes—cohesive reinforcement versus adhesive decoupling—future development should pursue two specific routes:

Route 1: design for controlled pseudoplasticity

Turn the ‘failure’ observed in PC into a functional crash concept. Research should focus on defining a precise separation threshold (via film thickness and placement) to activate frictional sliding reliably under high-energy impact, while verifying that quasi-static load capacity requirements are met.

Route 2: design for impact stable interphases

For high-potential films like PC and PSU, the focus must shift to interface engineering. Processing optimization and surface activation (e.g. plasma treatment) are required to raise the Mode I initiation threshold, ensuring that the intrinsic film toughness contributes to energy dissipation before debonding dominates.

Author contributions

CRediT: **Matthias Taubert**: Conceptualization, Data curation, Formal analysis, Investigation, Methodology, Project administration, Resources, Validation, Visualization, Writing – original draft, Writing – review & editing; **Luis Lippert**: Data curation, Formal analysis, Investigation, Visualization; **Bruno Musil**: Conceptualization, Funding acquisition, Supervision; **Philipp Höfer**: Conceptualization, Funding acquisition, Supervision.

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Data availability statement

Raw data were generated at Institute of Lightweight Engineering—University of the Bundeswehr Munich. Derived data supporting the findings of this study are available from the corresponding author [Matthias Taubert] on request.

Declaration of AI use

In accordance with the Taylor & Francis AI Policy, the author(s) declare the use of Gemini (Version 3.1 Pro), ChatGPT (Version 5.4 Pro), Claude (Version Sonnet 4.6) and Perplexity (based on GPT 5.1), during the drafting process. The tool was exclusively used to suggest improvements for sentence structure, grammar, and academic phrasing to ensure clear communication of the research. The author(s) reviewed all AI-generated suggestions and hold final responsibility for the content and integrity of this paper.

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