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Quasi-static mechanics of reactively formed epoxy–thermoplastic interphases in CFRP laminates

Matthias Taubert^a , Andreas Erich Scherer^b , Bruno Musil^a and Philipp Höfer^a

^aInstitute of Lightweight Engineering, University of the Bundeswehr Munich, Munich, Germany; ^bInstitute of Aeronautical Engineering, University of the Bundeswehr Munich, Munich, Germany

ABSTRACT

This study compares amorphous PEI, PES, PSU, and PC films co-cured as functional interlayers in aerospace-grade CFRP laminates based on HexPly® M18/1. A unified quasi-static test matrix comprising single lap shear, interlaminar shear strength, and Mode I and Mode II fracture toughness was applied to establish a directly comparable mechanical ranking. All interlayer systems preserved structural integrity, increasing single-lap shear strength by 6.2% to 13.3% to about 16 to 18 MPa. Interlaminar shear strength remained in a range of 48 to 59 MPa, but depended on interlayer placement. Fracture testing revealed two distinct response types. PEI and PES showed the most effective toughening, with PEI reaching the highest Mode I toughness at 590 J/m² (+59.0%) and PES the highest Mode II toughness at 6180 J/m² (+518.6%). In contrast, PSU and PC showed poor opening-mode resistance. Overall, the results demonstrate that interlayer selection governs both the fracture mechanism and achievable damage tolerance.

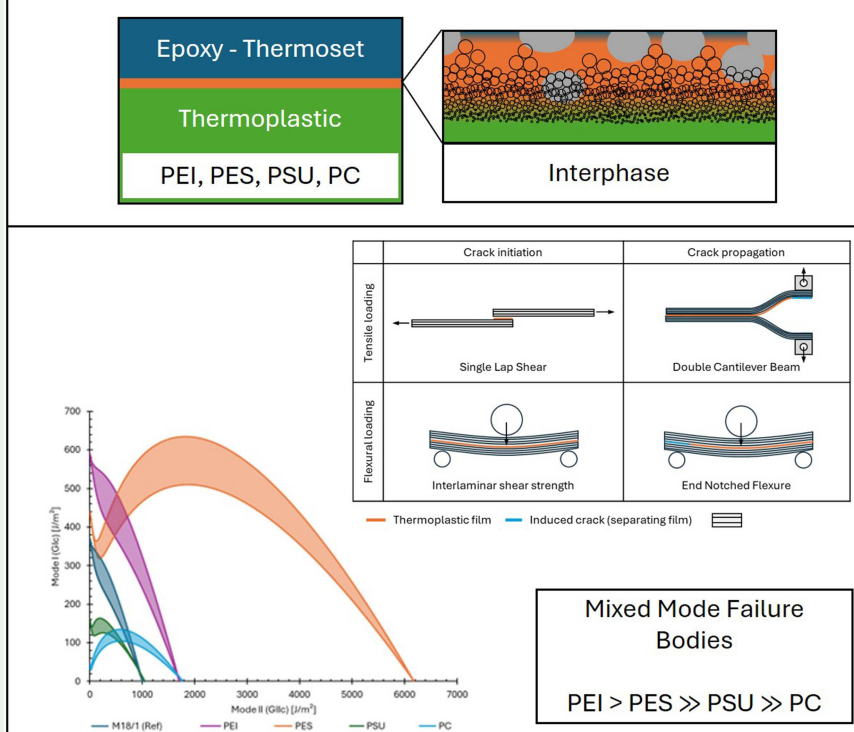
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Quasi-Static Mechanics of Reactively Formed Epoxy–Thermoplastic Interphases in CFRP Laminates



CONTACT Matthias Taubert matthias.taubert@unibw.de; matthias.taubert@gmail.com Institute of Lightweight Engineering, University of the Bundeswehr Munich, Munich, Germany

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1. Introduction

Fiber-reinforced polymer composites are increasingly recognized as a platform for functional integration. The use of thermoplastic components facilitates shaping [1–3], toughening [4–15], repair [16–22], welding [3,23–33], and retrofitting [34,35]. Thermosetting epoxy systems provide high stiffness, strength, and thermal resistance [1,2,36]. Combining both material classes enables capabilities that neither class can achieve on its own [1–3,36,37].

A pivotal element of such hybridization is the interphase. During epoxy cure, the thermoplastic partially dissolves in the reactive resin and diffuses into it. As curing proceeds, reaction-induced phase separation occurs. The resulting boundary is not a sharp interface but a chemically and morphologically graded zone that enables cohesive load transfer and structurally bonded joints. This mechanism is well documented in the literature and forms the base for a wide range of functional integrations [3,16,38–55].

A broad evidence base exists for PEI (polyetherimide) and PES (polyether sulfone) regarding interphase formation with aerospace epoxies, including HexPly M18/1 and RTM epoxy systems [3,35,38,39,56–60]. Imaging and in situ studies demonstrate that both thermoplastics initially dissolve in the reactive epoxy and undergo reaction-induced phase separation during curing. Instead of a sharp interface, a graded interphase forms [38,39,57,60]. For PES, high initial miscibility with epoxy precursors and specific interactions with glycidylamine-based epoxies have been documented [3,7,27,35,58,60]. Pronounced interdiffusion across the interphase has also been reported. For PEI, significant interdiffusion exceeding 50 μm has been observed, even in the presence of carbon fibers and morphologies ranging from particulate to co-continuous can develop depending on the cure process [25,56–58,60]. These findings confirm that PEI and PES interphases are driven by reactions and diffusion, forming a graded, load-carrying interphase rather than an abrupt interface [25,38,40,57,60,61]. Regarding mechanical performance, laminate level data are available for PEI and PES interphases, including single lap shear as well as Mode I and Mode II fracture characterization in thermoplastic modified aerospace epoxies. These studies provide a benchmark for the present comparative evaluation [3–5,7,9–12,35,62,63].

While the fundamental physicochemical coupling of PSU (polysulfone) to epoxy, including interdiffusion and toughening mechanisms, is well described [3,40,58,60,64], the literature exhibits a strong bias towards PEI and PES [3]. Historically, once the basic reactivity of these thermoplastics was proven, most laminate-level research quickly pivoted to PEI and PES, driven by their superior intrinsic mechanical properties and established usage in aerospace [3,37,58,64]. Consequently, comparative mechanical data for PSU and PC (polycarbonate) under identical laminate conditions remain sparse. However, this premature narrowing of focus overlooked the unique potential of these alternatives for functional integration. Specifically for concepts such as thermoplastic-assisted bonding or welding [3,23,25–33,49], the lower processing and melting temperatures of PSU and PC offer distinct manufacturing advantages over PEI and PES that warrant a detailed mechanical reassessment [65]. For PC, evidence of epoxy compatibility is largely limited to blend studies, and laminate-integrated interphases have only sporadically been addressed in older work [60,66,67].

Recent microsection studies of CFRP laminates with M18/1 show uniform PEI and PES interphases extending into the first fiber layer. For PSU, however, the interphases are more irregular and extend deeper into the fiber matrix. For PC, a blurred interphase with resin washout and spherulite formation has been observed. These findings demonstrate that interphase morphology is material-specific and that the canonical picture does not apply to PSU and PC. However, without mechanical testing, they do not permit conclusions about laminate load-bearing performance [60].

Meta-analyses confirm that PEI and PES are more present in the literature [3], PSU is more often characterized optically or chemically than mechanically at the laminate level [58,64], and PC appears in only a few studies [3,37,60]. Consequently, a rigorous, equivalent comparison of relevant mechanical parameters for PSU and PC versus PEI and PES under identical conditions is lacking. Open questions concern quasi-static mechanics in particular. It is also unclear whether the mechanical performance is primarily governed by the intrinsic properties of the thermoplastics, such as stiffness, toughness, and glass transition temperature, or by the interphase as an independent, graded structure. Therefore, this work deliberately frames the comparison against an M18/1 baseline and evaluates PEI, PES, PSU, and PC as reactively formed interphases within the same resin and laminate environment.

This paper provides a comprehensive, quasi-static mechanical characterization of PEI, PSU, PES, and PC as reactively formed interphases in carbon fiber reinforced polymer (CFRP) laminates of an aerospace epoxy. The materials and processing conditions were selected to reproduce the microsection conditions reported in the literature [60]. This approach enables the evaluation of the documented structural states, including matrix displacement in PSU and PC [60]. Core tests include single-lap shear, Mode I and Mode II fracture using DCB (Double cantilever beam) and ENF (End-notched flexure), as well as ILSS (Interlaminar shear strength). Morphological evidence from the literature is used to provide context, not as a direct predictor of strength [3,38–40,56,57,60].

Interphases are considered a basis for welded joints, repair strategies, and subsequent functionalization between thermoplastics and epoxies [3,23–33,37,49]. A robust, laminate-level comparison of PEI, PES, PSU, and PC provides an empirical foundation to delineate the application scope of these materials. The analysis explicitly considers use cases with constrained thermal parameters and scenarios in which the processing, cost, or recyclability of thermoplastics are relevant considerations. The results clarify the extent to which load-bearing capacity is governed by the base thermoplastic versus the interphase. They also explain how different thermoplastics act mechanically through the interphase and provide a reliable basis for selecting materials for functional integration concepts in aerospace epoxy CFRP.

2. Materials and methods

2.1. Materials

2.1.1. Thermoset matrix system

HexPly® M18/1, supplied by Hexcel Corporation, was selected as the baseline material system for investigating thermoplastic interphases [68]. This system is a high-performance, aerospace-grade carbon-fiber-reinforced toughened epoxy prepreg that is widely applied in primary aircraft structures and has been utilized extensively in recent studies. The utilized woven fabric prepreg features a nominal resin content of 43% by weight, which yields a nominal fiber volume fraction of 55% in the cured laminates. In prior work, the hybridization of M18/1-type epoxy systems with high-performance thermoplastics, such as PEI and PES, has been reported to be feasible [35,39,56,57,59,60]. In addition, studies indicated potential interphase formation for PSU and PC under suitable processing conditions [60].

2.1.2. Thermoplastic interlayers

The investigated thermoplastics were PEI, PES, PSU and PC (Supplier: LiteFilms—LITE GmbH; Products: Lite I, Lite S, Lite U, Lite PC). PEI and PES have been assessed in several studies under quasi-static loading and were therefore treated as established reference materials for benchmarking PSU and PC [35,39,56–59]. PSU and PC are of particular interest from a functionalization perspective because their lower melting and processing temperatures, compared with PEI and PES, make them more attractive candidates for melt-based processes such as fused filament fabrication, overmolding, and welding [65]. For all configurations, thermoplastic films with a thickness of 0.1 mm were used.

2.1.3. Interphases

To establish the morphological baseline of the consolidated composite interfaces prior to mechanical testing, cross-sectional scanning electron microscopy (SEM) was utilized. Figure 1 illustrates the distinct interphase morphologies resulting from the integration of the four thermoplastic films (PEI, PES, PSU, and PC).

The representative micrographs shown in Figure 1 are sourced from Taubert et al. It is imperative to note that the specimens depicted in these references were manufactured using the identical raw material batches (carbon fiber, matrix resin, and thermoplastic films) and consolidated under the exact same autoclave processing parameters as the specimens subjected to the mechanical testing in this study. Therefore, these morphologies serve as a direct, representative visual basis for the structural integrity of the interfaces.

As observed in the micrographs, PEI and PES develop pronounced and visually distinct interphases with clear phase boundaries. In contrast, the PSU interphase is significantly more dispersed within the

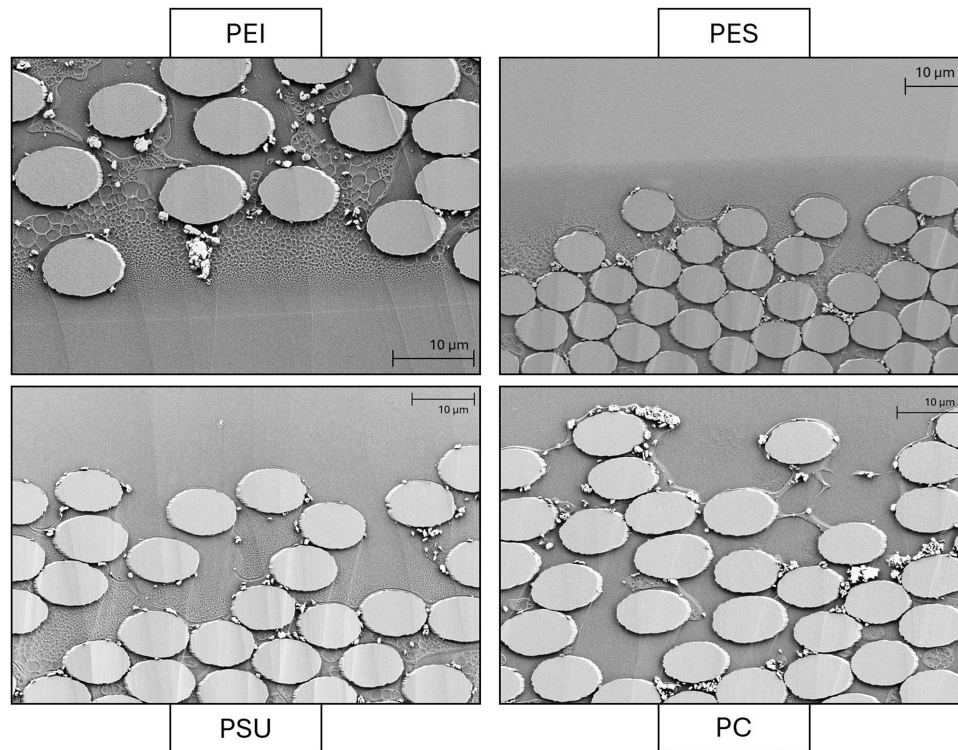


Figure 1. Representative SEM cross-sections of the consolidated laminates showing the distinct interphase morphologies for PEI, PES, PSU, and PC interlayers adapted from [60].

fiber bed, resulting in a diffuse transition zone lacking a sharply defined phase boundary. For the PC configuration, a classical continuous interphase layer is not visible; instead, the thermoplastic polymer extensively infiltrates the fiber-rich thermoset matrix, forming a complex morphology characterized predominantly by mechanical interlocking.

2.2. Methods

2.2.1. Autoclave manufacturing

All specimens were manufactured in an autoclave process. PEI and PES laminates were cured at a maximum temperature of 180 °C with a dwell time of 2 h. PC and PSU laminates were cured at 160 °C with a dwell time of 3 h, following the processing approach reported in the study [60]. As demonstrated in this prior work, these adapted temperature profiles optimize the interdiffusion behavior of the respective thermoplastic films. This tailored approach promotes a more favorable interphase formation for each specific polymer without compromising the ultimate curing quality or the baseline properties of the M18/1 epoxy system. Both cure cycles were performed at 2 bar autoclave pressure with a heating rate of 3 K/min.

2.2.2. Overall test design

The test program was designed to classify the mechanical performance of interphase-based connections across complementary loading modes. Joint level tests commonly used for bonded structures, such as single lap shear, were combined with fracture mechanics-based characterization to distinguish crack initiation from crack propagation. This approach enables a differentiated assessment of whether a given interphase configuration provides primarily strength, damage tolerance, or stable crack growth resistance. In addition, the framework allows discussing potential links between the observed mechanical response and the interphase morphologies described in [60]. Figure 2 summarizes the applied test matrix by contrasting crack initiation versus crack propagation (horizontal axis) and end (axial) loading versus flexural loading (vertical axis). This structure enables a consistent comparison across joint level shear

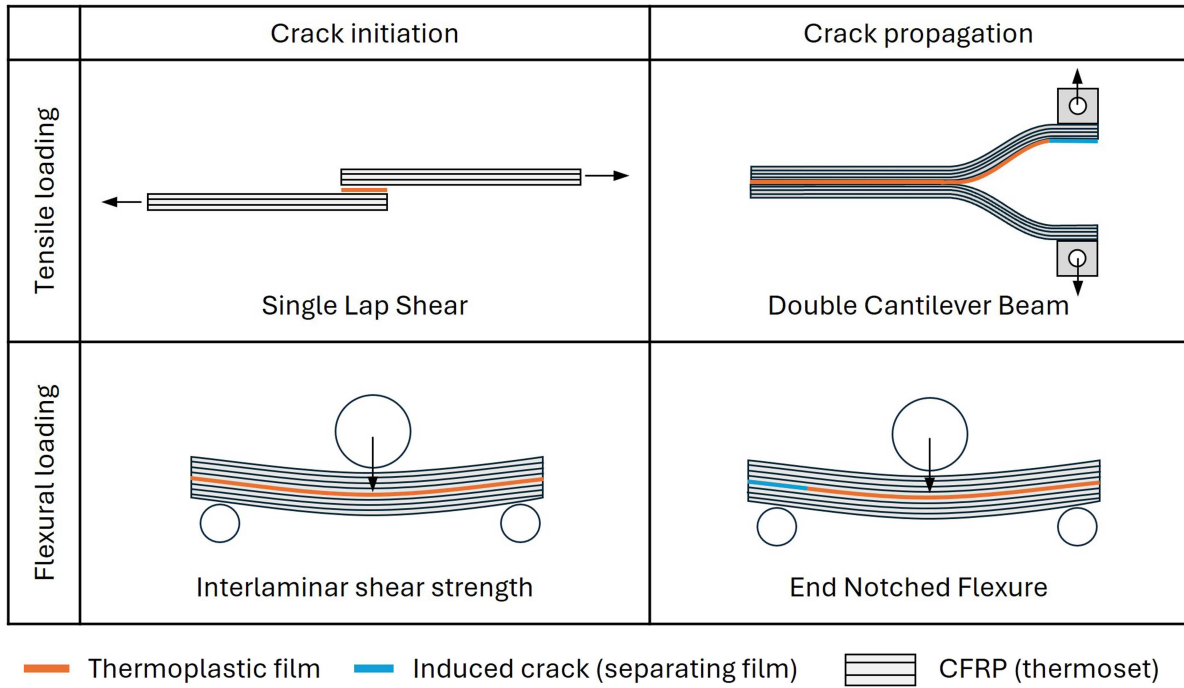


Figure 2. Test design for the evaluation of bonding properties of different interphase systems.

transfer (single lap shear), opening-mode fracture behavior (double cantilever beam), interlaminar shear performance under bending (ILSS), and shear-dominated fracture behavior (Mode II) under flexural loading (End-notched flexure).

2.2.3. Single lap shear (SLS)

A key indicator for the mechanical performance of the interphases is the single lap shear (SLS) test [29,35]. In particular, it enables an initial classification of joint load capacity in comparison with established adhesive joint concepts [19,35,69]. The specimens were manufactured according to DIN EN 1465 and tested in tension on a universal testing machine (ZwickRoell Z150). In this study, SLS specimens are primarily used to assess the joining quality and in plane shear load transfer across the overlap [3,29,35]. The entire specimen, including the end tabs, was completely co-cured in the autoclave process without the use of secondary adhesives.

To evaluate the mechanical stability of the bonded joints, the apparent lap shear strength τ was calculated. The following formula was used to perform the data reduction:

$$\tau = \frac{F_{\max}}{b \cdot l_0},$$

where:

- $F_{\max}$ is the maximum sustained load at failure (N).
- b is the width of the bonded overlap area (mm). DIN EN 1465: 25 mm (± 0.25 mm)
- l_0 is the initial length of the bonded overlap (mm). DIN EN 1465: 12, 5 mm (± 0.25 mm)

2.2.4. Interlaminar shear strength (ILSS)

Interlaminar shear strength testing was used to assess how the thermoplastic interlayers affect delamination tendency under flexural loading and transverse shear, which are relevant for bending-dominated applications [70–72]. In this context, ILSS serves as an indicator of laminate-level integrity when interphases are introduced as functional interlayers. Specimens were manufactured according to ISO

14130 and tested on a ZwickRoell universal testing machine (Z150). Two layups were investigated for each material combination (see Figure 3): a configuration with a single thermoplastic film placed at the midplane, and a configuration with the film placed at the midplane and additionally symmetrically distributed interlayers on both laminate halves. This design isolates the influence of interlayer placement under lateral force – induced interlaminar shear and supports a more mechanistic interpretation of the laminate response. A monolithic CFRP laminate served as the reference.

To evaluate the interlaminar adhesion of the composite laminates, the apparent interlaminar shear strength τ was calculated. The following formula was used to perform the data reduction in accordance with the ISO 14130 standard:

$$\tau = \frac{3F}{4bh},$$

where:

- F is the maximum sustained load at failure (N).
- b is the width of the specimen (mm), ISO 14130: 10 mm (± 0.2 mm)
- h is the thickness of the specimen (mm), ISO 14130: 2 mm (± 0.2 mm)

2.2.5. Mode I fracture toughness (G_{Ic})

To evaluate the integrity of an interface in a bond-like configuration, peel-dominated loading is commonly used because it is sensitive to adhesion and interfacial weakness [15,35,73,74]. In this work, Mode I fracture toughness testing was performed to quantify the energy required for crack propagation along the interphase region, i.e. to assess adhesion on both sides of the thermoplastic-rich zone. Double cantilever beam (DCB) specimens with a predefined starter crack were manufactured and tested according to ASTM D5528/D5528M. During testing, the specimen arms were displaced in tension while crack length and the corresponding energy release rate were measured. The resulting G_{Ic} values therefore serve as an indicator of potential adhesion limitations, effectively probing the interphase as a boundary layer between thermoset and thermoplastic. For DCB testing, the characteristic fracture toughness values were determined using the load displacement state at which the crack front, tracked visually during the test, reached a predefined marker position. The reported values, therefore represent a standardized visually defined early crack extension criterion.

To evaluate the Mode I interlaminar fracture toughness G_{Ic} of the composite joints, data reduction was performed in accordance with the ASTM D5528 standard. The calculation of the critical strain energy release rate was automated utilizing the integrated material testing software (testXpert, ZwickRoell). Utilizing the software's automated routine ensures the continuous, synchronized processing of load, crosshead displacement, and crack propagation data, thereby minimizing subjective user bias and guaranteeing high reproducibility of the data reduction process.

Specifically, the software computes the G_{Ic} values based on the Modified Beam Theory (MBT) using the following formula:

$$G_{Ic} = \frac{3P\delta}{2b(a + \Delta)},$$

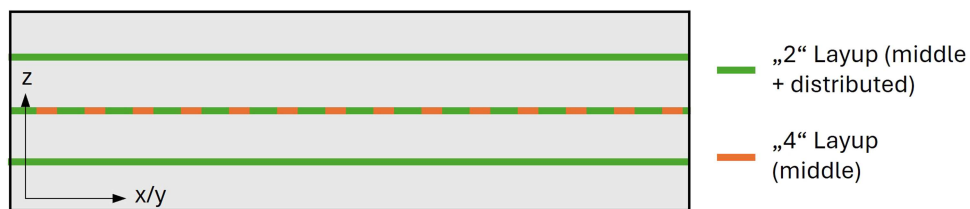


Figure 3. ILSS Test Layup laminates with PEI, PES, PSU and PC as interlayers in the middle (every four layers) and distributed symmetrically around the middle (every two layers).

where:

- P is the applied load (N).
- δ is the load point displacement (mm).
- b is the measured width of the specimen (mm). with $b = 25$ mm
- a is the measured delamination length (mm).
- Δ is the correction term from the compliance calibration implemented in the test evaluation routine of the testXpert Software (ZwickRoell)

2.2.6. Mode II fracture toughness (G_{IIc})

To quantify the energy absorption capacity for shear driven crack growth along the interphase plane, Mode II fracture toughness testing was conducted [15,35,73,74]. End-notched flexure (ENF) specimens with a defined starter crack were tested according to DIN EN 6034. The pre-cracked specimens were loaded in three-point bending, and crack propagation was correlated with the energy release rate to determine G_{IIc} . This test targets damage mechanisms relevant for hybrid laminates under bending-induced shear, where a thermoplastic interlayer may promote interlaminar sliding or, conversely, provide shear resistance. In this sense, G_{IIc} characterizes the tendency of laminate plies to slide relative to each other along the interphase plane under shear-dominated conditions. The reported G_{IIc} values correspond to the fracture toughness calculated from the load displacement state at the first visually detected crack propagation event in the ENF test, in accordance with DIN EN 6034.

To evaluate the Mode II interlaminar fracture toughness G_{IIc} of the composite joints, data reduction was performed in accordance with the DIN EN 6034 standard utilizing an End-Notched Flexure (ENF) test configuration. The critical strain energy release rate was calculated analytically post-test, utilizing the synchronized load and crosshead displacement data recorded during the testing procedure.

Specifically, the G_{IIc} values were determined based on the load-displacement behavior using the following standard equation:

$$G_{IIc} = \frac{9Pa^2\delta}{2b\left(\frac{L^3}{4} + 3a^3\right)},$$

where:

- P is the load recorded at the first visually detected crack propagation event (N).
- δ is the corresponding crosshead displacement at this visually detected crack propagation event (mm).
- a is the measured initial crack length (mm). DIN EN 6034: 35 mm (± 1 mm)
- b is the measured width of the individual specimen (mm), DIN EN 6034: 25 mm (± 0.2 mm)
- L is the total span length between the lower supports (mm), DIN EN 6034: 100 mm

3. Results

3.1. Single lap shear (SLS)

The Single lap shear (SLS) results, see Figure 4 and Table 1, show that the in-plane load transfer across the joint works well and is improved by the thermoplastic interlayers. The measured shear strengths range from about 16 to 18 MPa and are consistently higher than the M18/1 baseline without an interlayer (15.08 MPa). When normalized to this reference, the interlayer configurations yield increases of +6.2% (PC), +13.3% (PEI), +9.1% (PES), and +10.8% (PSU). These values fall into the medium strength range reported for relevant bonded joints [2] and indicate that the interphase does not limit in-plane shear transfer in the overlap region. Instead, it contributes to a more effective shear carrying capability of the joint.

Macroscopic inspection of the fracture surfaces (see Figure 5) does not show purely cohesive failure within the thermoplastic film. Rather, failure is predominantly interfacial with fiber exposure (light fiber

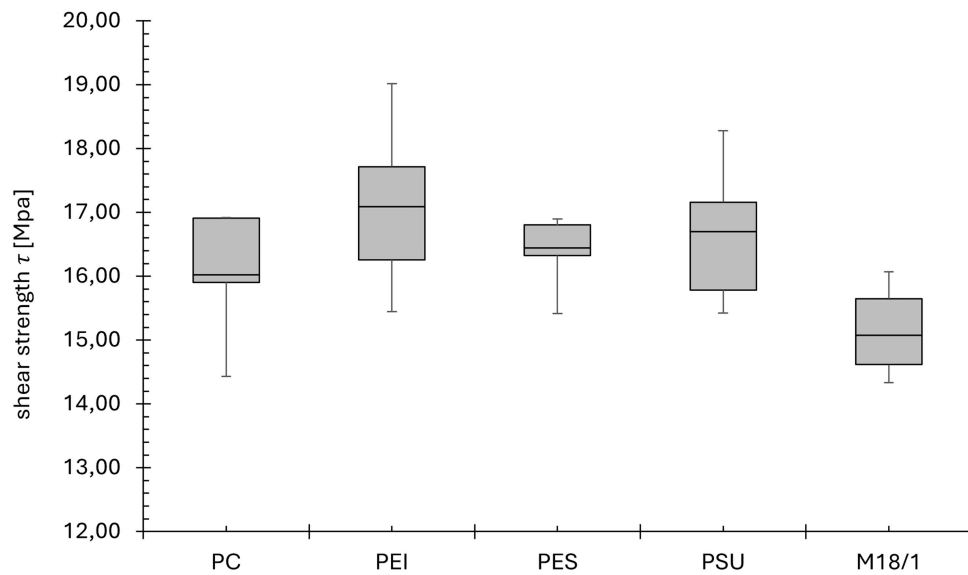


Figure 4. Test results – single lap shear specimens with PEI, PES, PSU, and PC interlayers ($n = 5$; box plots showing median, IQR, and min/max).

Table 1. Apparent single lap shear strength τ in MPa for the reference and hybridized configurations. Data shown as mean MPa standard deviation ($n = 5$).

	PC	PEI	PES	PSU	M18/1
τ in MPa	16.02 (± 1.02)	17.09 (± 1.17)	16.45 (± 0.55)	16.70 (± 1.02)	15.08 (± 0.66)

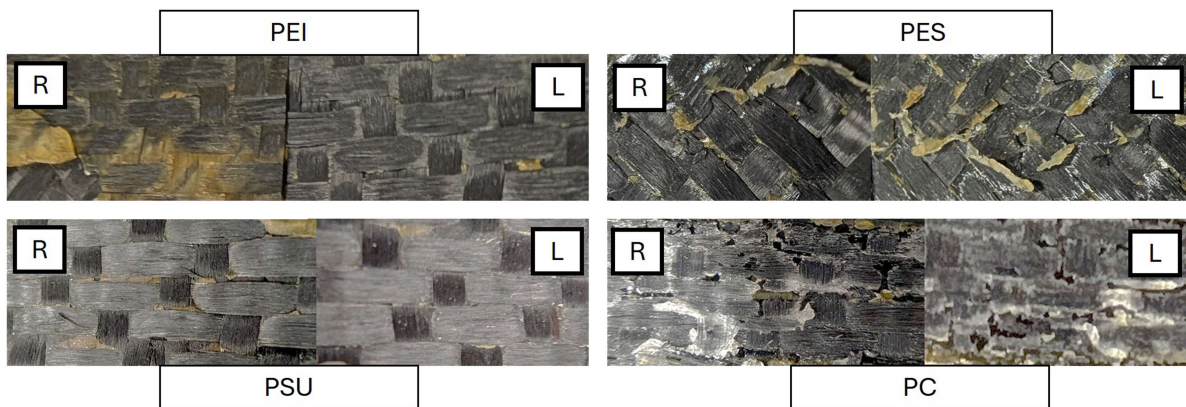


Figure 5. Macroscopic fracture surfaces of single lap shear specimens with PEI, PES, PSU, and PC interlayers after failure.

tear). This is important for interpretation: even though the overall SLS strength increases compared to the baseline, the critical region remains the coupling between the thermoplastic modified matrix and the top fiber ply, including the fiber sizing. A high local thermoplastic content near the surface can alter the fiber matrix adhesion, so that the interfacial mechanism governs the observed failure mode. From an application perspective, the interphase appears robust under in-plane shear, but the near fiber interface should be treated as a design relevant risk.

3.2. Interlaminar shear strength (ILSS)

The ILSS results reveal a configuration-dependent response picture that is strongly governed by the interlayer configuration. While the average levels across all specimens appear relatively stable, separating

the data by layup highlights pronounced differences between the investigated stacking concepts (see Figure 6 and Table 2).

Notably, the established high-performance thermoplastics PES and PEI exhibit the lowest ILSS levels when implemented as a single midplane interlayer (Layup '4'), showing reductions of -12.9% and -8.7% relative to the reference. In the distributed configuration (Layup '2'), both move back toward the reference level, reaching -1.2% (PES) and $+2.7\%$ (PEI). PSU shows the opposite trend: it reaches the highest value in Layup '4' ($+7.1\%$) and decreases to $+1.0\%$ in Layup '2'. PC remains close to the reference in Layup '4' ($+0.3\%$) but decreases to -4.6% when distributed (Layup '2').

However, interpreting ILSS solely based on peak strength is incomplete. The load–displacement responses in Figure 7 show that several configurations retain substantial residual load after the first damage event, and that the post-peak response differs systematically between material systems and stacking concepts.

For PEI and PES, the peak is followed by a stepwise decay rather than an immediate collapse, indicating progressive damage development with sustained load transfer. This stable post-peak regime is particularly pronounced for PEI in the distributed configuration, whereas PES maintains a similarly stable curve shape across both layups despite the peak-level differences reported above. In contrast, PC shows a distinctly flatter peak region and an extended high-load plateau, together with an increased deformation range before the later drop. PSU does not show the same plateau signature as PC, but it exhibits a clear post-peak load-carrying regime in both layups, with the distributed configuration showing an earlier first drop than the midplane configuration.

The combined evidence from Figures 6 and 7 suggests that interlayer placement not only shifts the peak level but also alters how damage evolves after the first event. For PEI and PES, the stable post-peak behavior and the recovery of peak performance when the interlayers are distributed are consistent with the concept that multiple shear-active planes reduce stress localization and enable load redistribution through the thickness (Figure 8). In line with the microsection-based observations reported in [60], PC and PSU may benefit more strongly from a mechanical interlocking (where the thermoplastic penetrates deeper into the adjacent fiber layers, creating physical anchors rather than relying purely on chemical bonds), which can elevate the peak response. At the same time, the earlier first drop observed when distributing PC and PSU indicates that adding multiple interphase planes can introduce additional initiation sites for early damage, such that the potential benefit of multiple shear-active planes can be partially offset even when residual capacity remains substantial.

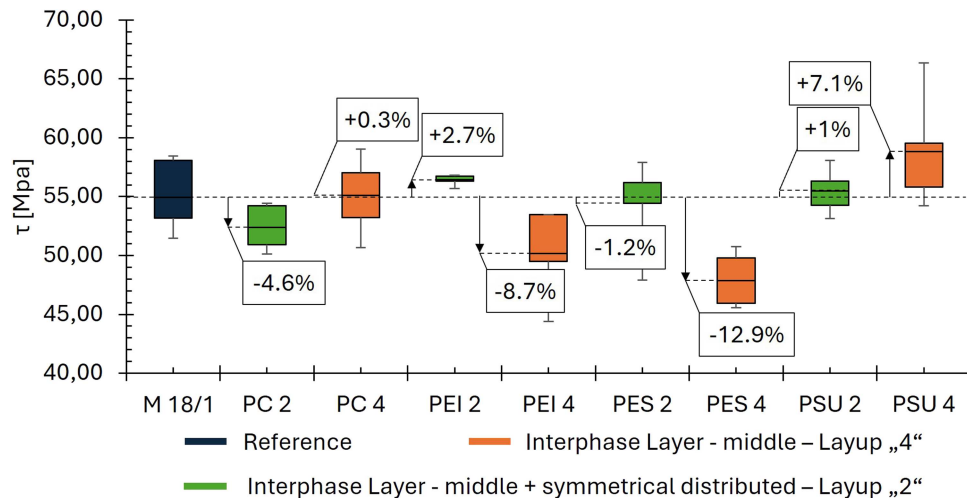


Figure 6. ILSS test results with PEI, PES, PSU and PC as interlayers in the middle (every four layers) and distributed symmetrically around the midplane (every two layers) ($n = 5$; box plots showing median, IQR, and min/max).

Table 2. Apparent ILSS τ in MPa for the reference and hybridized configurations. Data shown as mean MPa standard deviation ($n = 5$).

	M18/1	PC 2	PC 4	PEI 2	PEI 4	PES 2	PES 4	PSU 2	PSU 4
τ in MPa	54.95 (± 2.79)	52.40 (± 1.72)	55.12 (± 2.92)	56.43 (± 0.41)	50.17 (± 3.33)	54.27 (± 3.36)	47.88 (± 2.06)	55.49 (± 1.70)	58.85 (± 4.18)

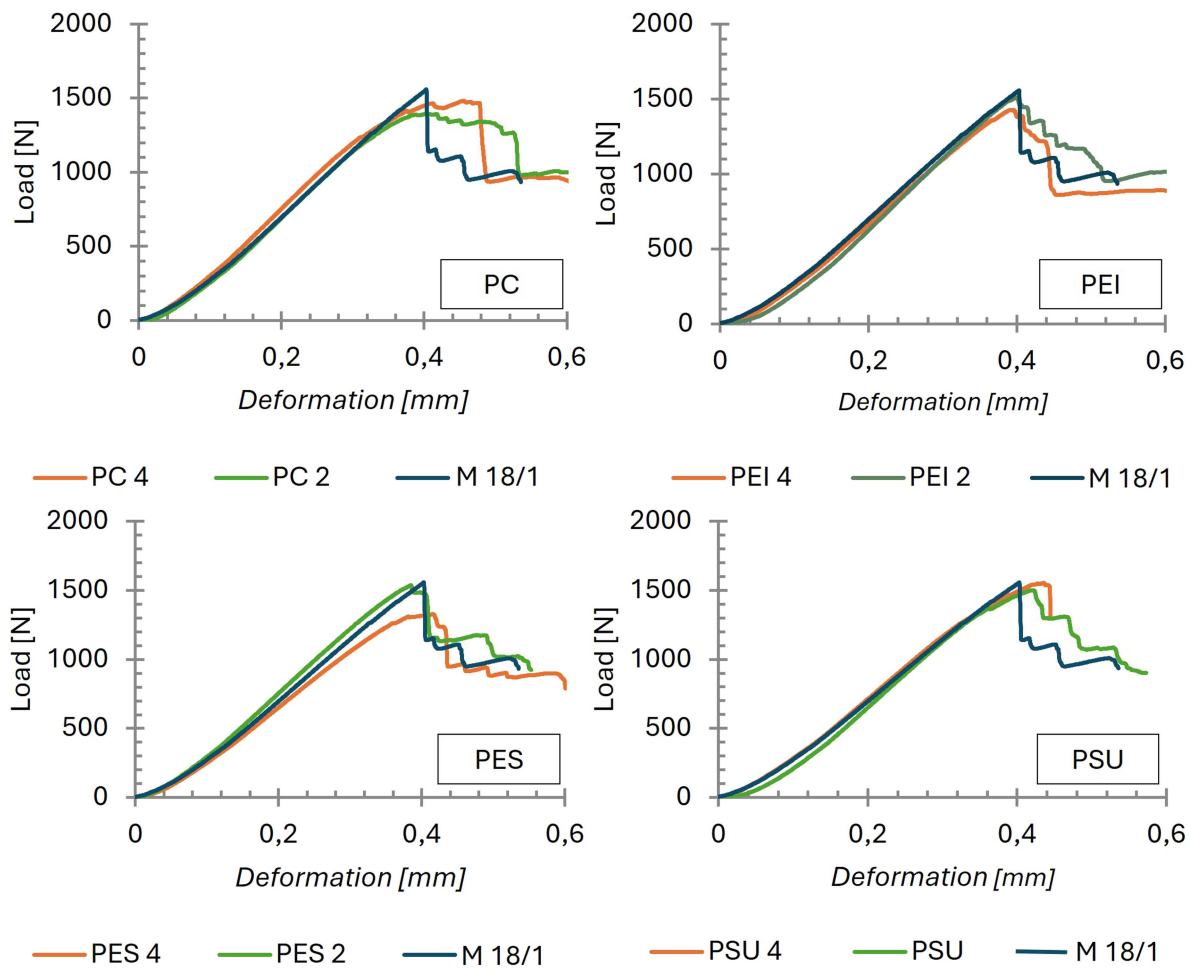


Figure 7. Load–displacement curves for ILSS tests of PEI, PES, PSU and PC as interlayers in a thermoset CFRP laminate ($n = 5$).

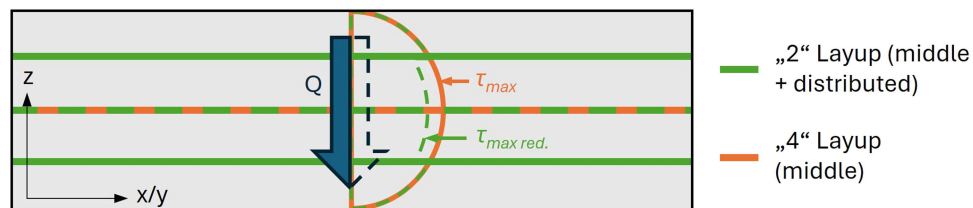


Figure 8. Schematic comparison of shear stress profiles τ_z for Layup '4' and Layup '2' with bending load Q .

3.3. Mode I fracture toughness (G_{Ic})

The results of the DCB tests are summarized in Figure 9. The analysis of the critical strain energy release rates demonstrates that the selection of the interphase material plays a decisive role in determining Mode I fracture toughness, resulting in substantial performance disparities between the tested groups.

In the following analysis, the PEI interphase serves as the benchmark for high toughness, exhibiting the maximum values within this experimental series with a mean G_{Ic} of 590 J/m².

In direct comparison, the chemically related PES interphase (440 J/m²) retains approximately 75% of the PEI performance level. While the reduction in fracture toughness by roughly one-quarter is a substantial decrease, both systems remain in the high-toughness category relative to the other materials tested.

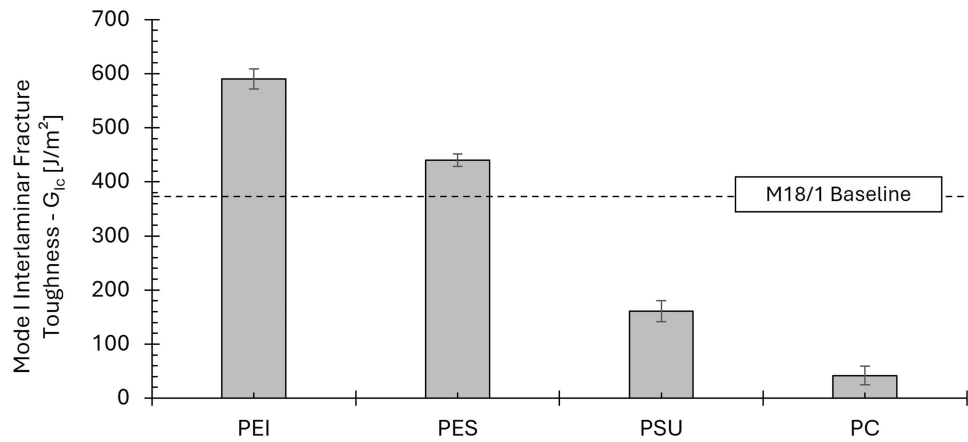


Figure 9. Interlaminar fracture toughness G_{Ic} of the hybrid laminates modified with PEI, PES, PSU, and PC interlayers ($n = 5$).

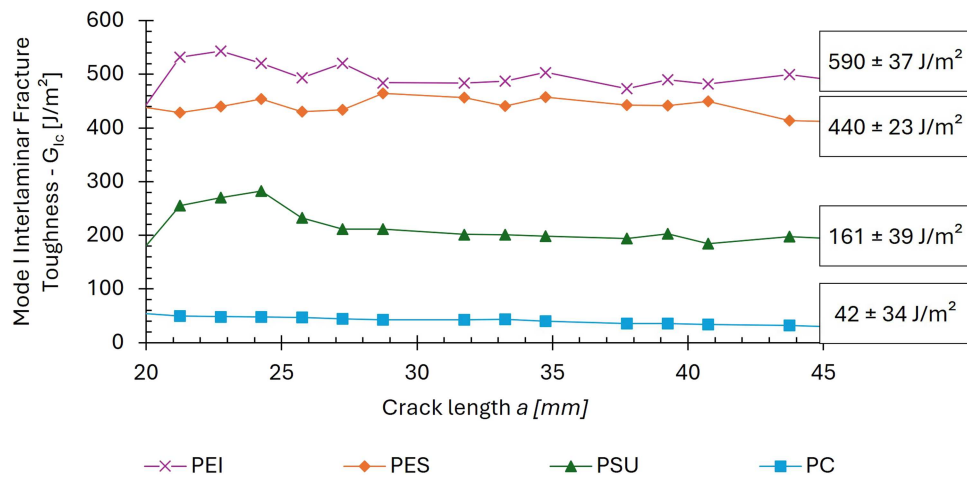


Figure 10. Development of the critical strain energy release rate G_{Ic} as a function of crack extension (R-curves) of an exemplary test set of hybrid laminates modified with PEI, PES, PSU, and PC interlayers ($n = 5$).

A drastic drop in performance is observed for the remaining thermoplastics. The PSU specimens achieved, on average, only 27% of the fracture toughness recorded for PEI. With a mean G_{Ic} of 161 J/m^2 , the energy absorption capability of the PSU interphase is lower by a factor of more than 3.5 compared to the peak values of PEI.

The lower end of the spectrum is defined by the PC interphase. Averaging merely 42 J/m^2 , these samples reached only 7% of the PEI reference value. This massive reduction of more than one order of magnitude suggests that polycarbonate provides only limited resistance under Mode I loading conditions in this specific matrix system.

The resistance curves (R-curves) presented in Figure 10 reflect this hierarchy:

- PEI and PES: both materials display distinct R-curve behavior. Following crack initiation, the resistance increases steeply before settling into a stable, elevated propagation plateau. The relative scatter of the plateau values (coefficient of variation) for both systems is low (6%–8%), suggesting homogeneous interphase formation and consistent failure mechanisms.
- PSU: the PSU samples also exhibit a stable propagation plateau, albeit at the substantially reduced energy level described above.

- PC: in contrast, the PC curves demonstrate unstable crack growth behavior. Beyond the extremely low mean value, the data is characterized by an enormous relative scatter of over 80% (relative to the mean). This suggests highly localized variations in failure mechanisms, where regions of near-total adhesion failure alternate with zones of minor adhesion.

The observed ranking in Mode I fracture toughness can thus be summarized as follows:

PEI > PES $\gg$ PSU $\gg$ PC

Placing these results in the context of the base material properties reveals the fundamental impact of the interphase modification. According to BRUCKBAUER, the typical strain energy release rate G_{Ic} for a M18/1 laminate is stated as 371 J/m² [35].

Comparing this reference value with the experimental data highlights the different Mode I responses of the tested thermoplastics. PEI and PES act as effective toughening agents, increasing the fracture toughness above the reference level by approximately +59.0% and +18.6%, respectively. By contrast, PSU and PC show markedly lower Mode I fracture resistance, remaining about -56.6% and -88.7% below the reference, respectively. These results indicate that PSU and especially PC do not provide effective toughening under opening-mode loading, but instead are associated with early interfacial failure that limits the achievable laminate toughness.

Macroscopic inspection of the Mode I fracture surfaces (see Figure 11) visually corroborates the distinct failure mechanisms implied by the R-curves. Both PEI and PES exhibit a highly uniform and continuous fracture morphology. This even surface texture indicates an excellent structural integration of the thermoplastic into the epoxy matrix, leading to stable, cohesive failure propagation within the interphase and consistent load transfer.

In contrast, the PSU interphase displays a notably heterogeneous pattern. The fracture surface is characterized by localized ‘islands’ of varying failure modes, alternating between areas of exposed fibers and patches of cohesively torn thermoplastic. This uneven phase integration reflects a mixed interfacial-cohesive failure, which correlates directly with the substantial drop in Mode I energy absorption.

This island-like phase separation is most severe in the PC specimens. The failure surface reveals extensive areas of relatively unperturbed thermoset and bare fibers, interspersed with rough, isolated patches of residual polycarbonate. The visual evidence strongly suggests a purely adhesion-limited interfacial debonding, where the PC film essentially peeled away from the underlying epoxy network with negligible synergistic interaction. This macroscopic lack of chemical bonding is consistent with the near-zero baseline performance and the enormous scatter observed in the G_{Ic} results.

3.4. Mode II fracture toughness (G_{IIc})

The results for the Mode II critical strain energy release rate G_{IIc} , summarized in Figure 12, reveal a distinctly different ranking compared to the Mode I results. While PES stands out as the superior material,

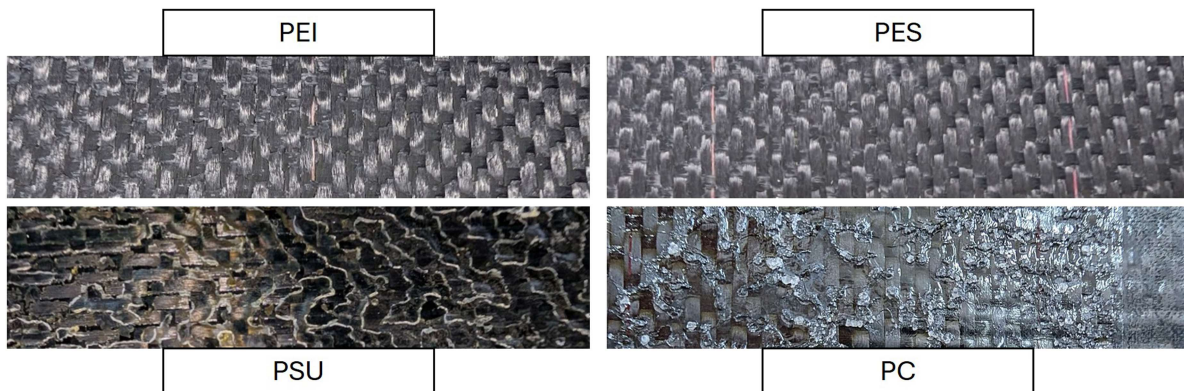


Figure 11. Macroscopic fracture surfaces of Double Cantilever Beam (DCB) specimens with PEI, PES, PSU, and PC interlayers after Mode I failure.

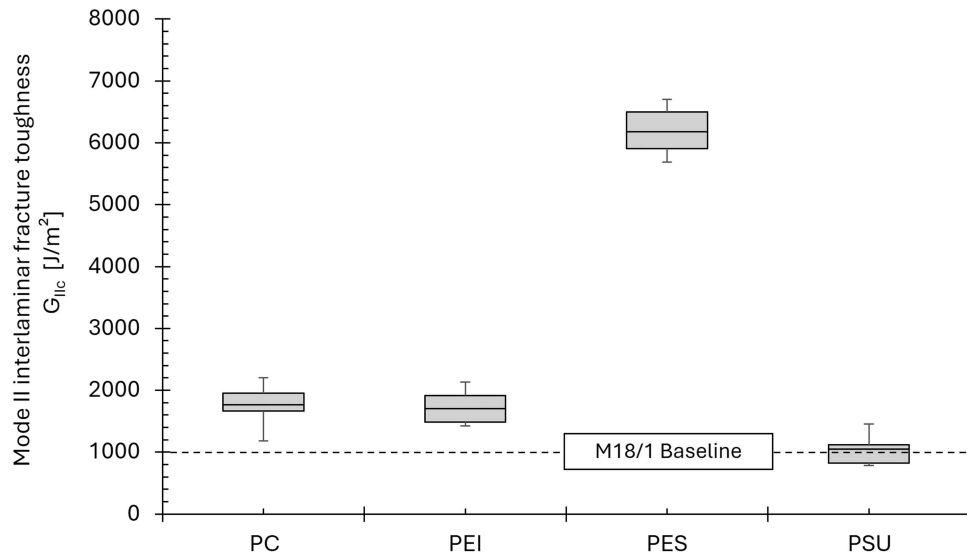


Figure 12. Mode II interlaminar fracture toughness G_{IIc} of the hybrid laminates modified with PC, PEI, PES, and PSU interlayers ($n = 5$).

PC and PEI perform at a comparable level, whereas PSU exhibits the lowest resistance to shear fracture, but still slightly above the baseline M18/1 Material. The measured levels span from $\sim 1050 \text{ J/m}^2$ (PSU) to $\sim 6180 \text{ J/m}^2$ (PES), with PC and PEI around ~ 1700 to 1800 J/m^2 .

The exceptionally high shear toughness of PES aligns with established literature data, such as the findings in [35]. A plausible interpretation is the formation of a tough, co-continuous interphase morphology that facilitates extensive plastic deformation under shear stress. The material is capable of undergoing significant ductile yielding ('ductile drawing') within the fracture process zone, thereby dissipating large amounts of energy before separation occurs.

A notable observation is the performance of PC, which achieves G_{IIc} values on par with the benchmark material PEI, despite having shown negligible chemical adhesion in Mode I (G_{Ic}) tests. This apparent contradiction is likely attributable to the mechanical interlocking effect described in [60]. As shown in the microstructural analysis, the low viscosity and 'wash-out' behavior of PC lead to a deep infiltration of the thermoplastic into the fiber bed, see Chapter 2.1. Under Mode II shear loading, which involves sliding rather than opening, this macroscopic mechanical interlocking acts as a frictional brake. The roughness and deep physical anchoring compensate for the lack of chemical bonding, providing resistance against crack propagation that was not available under the tensile opening loads of Mode I.

In contrast, PSU shows the least pronounced toughening response within the present test series. It appears to occupy an unfavorable intermediate position, lacking both the strong chemically supported damage tolerance observed for the PEI and PES systems and the pronounced mechanically dominated shear response associated with PC. As a result, PSU does not develop a comparably robust fracture resistance profile and remains clearly below the best-performing interlayer systems, even though its Mode II toughness is still close to the reference level.

Beyond the fracture energy values, the load-displacement curves (see Figure 13) highlight a unique functional characteristic of the PC interlayers. While the absolute peak force for PC is not superior to the other candidates, the material exhibits a remarkably high deflection prior to failure. This behavior correlates with the high intrinsic ductility and elongation at break of polycarbonate as a bulk material. This observation underscores an important design implication: In specific load cases where high compliance or deformation tolerance is required, the intrinsic mechanical properties of the thermoplastic film itself play a governing role.

Furthermore, the force-displacement response of PES and PEI reveals a remarkable behavior beyond the standardized crack initiation point (where G_{IIc} is typically determined). Contrary to the expected load drop associated with brittle fracture propagation, both materials, and PES in particular, exhibit a continued increase in force even after the initial crack onset. This suggests that the crack does not propagate

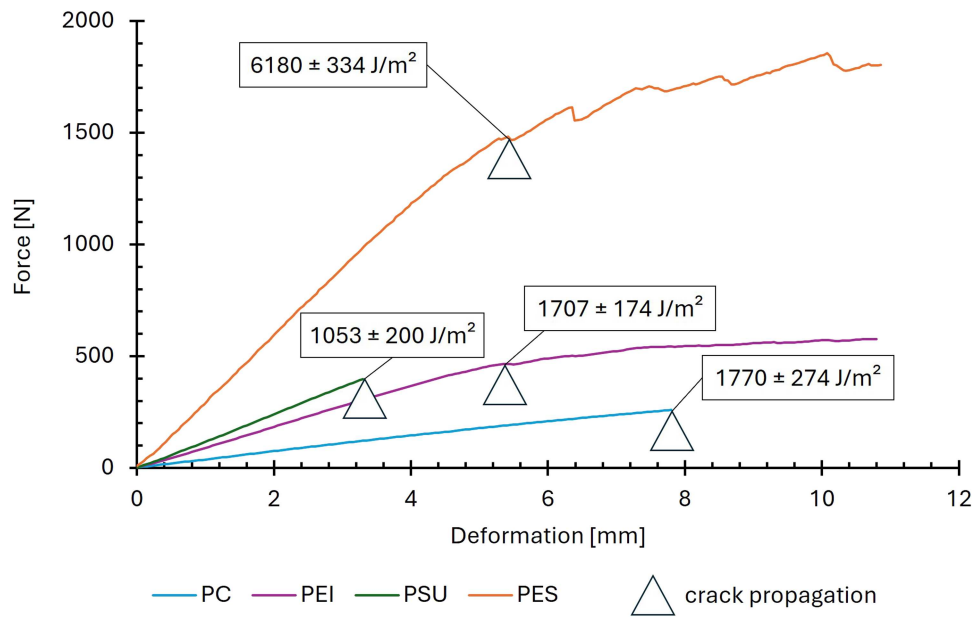


Figure 13. Representative force-deformation curves from End-Notched Flexure (ENF) testing for PC, PEI, PSU, and PES interlayers.

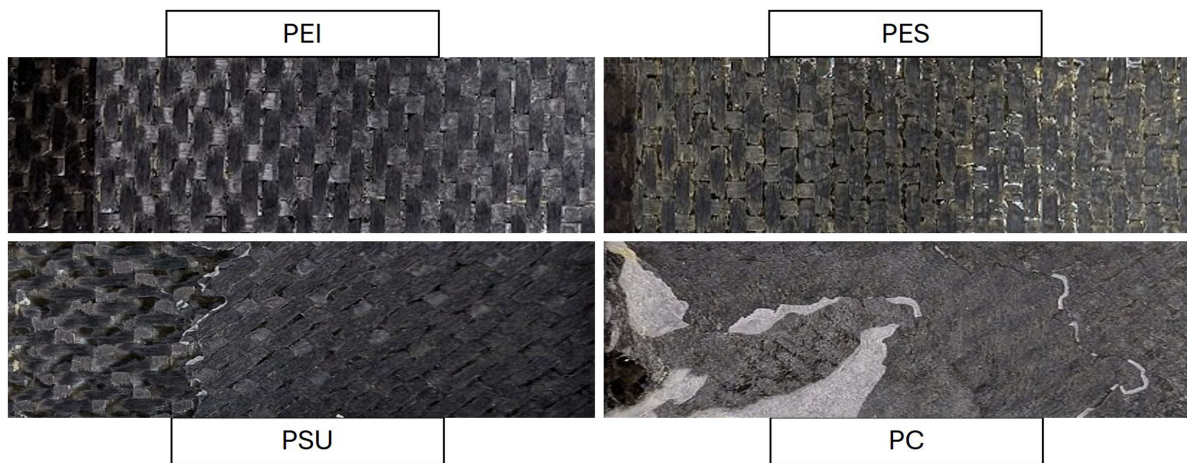


Figure 14. Macroscopic fracture surfaces of End-Notched Flexure (ENF) specimens with PEI, PES, PSU, and PC interlayers after Mode II failure.

catastrophically but is effectively arrested by the tough interphase. This ability to re-stabilize and sustain increasing loads despite local damage serves as compelling evidence for the superior chemical adhesion and the robust integration of the thermoplastic within the composite. It indicates that the interphase actively resists crack growth through extensive plastic deformation and bridging mechanisms, providing a safety margin that extends well beyond the critical failure threshold.

To assess the toughening efficiency, the results were compared with the literature baseline for unmodified M18/1 laminates [35], using a reference value of $G_{IIc} \approx 999 \text{ J/m}^2$. PES shows by far the strongest effect, reaching approximately 6180 J/m^2 and exceeding the reference by about +518.6%, consistent with its pronounced ductility and exceptional shear-dominated energy absorption. PEI and PC also provide substantial toughening, reaching approximately 1707 J/m^2 and 1770 J/m^2 , corresponding to increases of about +70.9% and +77.2%, respectively. By contrast, PSU reaches approximately 1053 J/m^2 , which corresponds to only a slight increase of about +5.4% relative to the reference and therefore indicates only limited toughening in Mode II.

Macroscopic evaluation of the Mode II fracture surfaces (see Figure 14) further illustrates the mechanisms driving the shear toughness. The PEI and PES specimens exhibit relatively similar, continuous

failure patterns. However, the PES fracture surface appears slightly more textured and irregular compared to PEI. This pronounced surface roughness indicates a higher degree of localized plastic yielding and mechanical interlocking during shear deformation, which might account for the exceptionally high energy dissipation and superior G_{IIc} values recorded for PES.

In contrast, the PSU interphase displays a highly regular and smooth fracture morphology without noticeable abnormalities or significant deformation features.

The PC configuration presents the most distinct behavior. In stark contrast to the interfacial debonding observed in Mode I, the Mode II fracture surface reveals that the polycarbonate film itself has undergone severe tearing and plastic rupture. This confirms that under shear-dominated sliding loads, the structural coupling is significantly more effective. The cohesive failure of the film acts as a substantial energy sink, corroborating the unexpectedly competitive G_{IIc} performance despite the material's poor opening-mode adhesion.

4. Discussion

4.1. Overall mechanical performance across test modes

Table 3 summarizes the quasi-static response as a multi-metric profile rather than a single ranking. Two main observations can be made. First, the interlayers are mechanically compatible with joint-level in-plane load transfer. SLS increases by +6.2% to +13.3% (reaching 16–18 MPa) for all materials compared with the M18/1 baseline. This quantitatively substantiates the concepts reviewed by Deng et al., proving that the reactively formed interphases do not act as weak separation planes, but rather effectively transfer in-plane shear loads for structural joining [3]. Second, the interlayers primarily affect fracture-controlled properties, while shear strength metrics remain comparatively stable. ILSS shifts are moderate and placement-dependent, which indicates that interphase position is a structural design variable rather than a processing detail.

The fracture response separates the systems not only by absolute magnitude but also by failure character, while remaining in good agreement with existing literature. PEI and PES show the most robust damage tolerance signatures, frequently exhibiting residual load-carrying capacity after crack initiation and a tendency to stabilize crack advance rather than transitioning immediately into unstable growth. Based on the revised reference values of 371 J/m² for Mode I and 999 J/m² for Mode II, PEI reached 590 J/m² in Mode I, corresponding to a relative increase of +59.0% [35]. This result is in very good agreement with the M18/1-based findings reported by BRUCKBAUER, who likewise observed an increase of about 59% for PEI interphases in prepreg laminates [35]. Compared with the stronger G_{Ic} improvements of about +170% to +190% reported by SCHÜTT for PEI-modified epoxy systems, the present data confirm the same toughening trend, albeit at a clearly lower magnitude [64]. SCHÜTT's comparative results nevertheless rank the Mode I toughening efficiency of PEI clearly above that of PSU, which supports the same performance hierarchy observed in the present unified test set.

For PES, the current study revealed an extraordinary shear-dominated energy absorption of 6180 J/m², corresponding to a relative increase of +518.6% in Mode II. This strong enhancement is in very good alignment with the prepreg-based results reported by BRUCKBAUER, who measured an average G_{IIc} value of 6228 J/m² for M18/1 laminates with PES interlayers. It is also consistent with the broader literature trend that PES interlayers can produce pronounced gains under shear-dominated fracture loading. Together,

Table 3. Overview of G_{Ic} , G_{IIc} , ILSS (Layups 2 and 4), and SLS results, normalized to the M18/1 baseline.

Property/test mode	Reference baseline (M18/1)	Ø PEI (vs. ref)	Ø PES (vs. ref)	Ø PSU (vs. ref)	Ø PC (vs. ref)
Fracture toughness Mode I G_{Ic}	371 J/m ²	+59%	+18.6%	−56.6%	−88.7%
Fracture toughness Mode II G_{IIc}	~999 J/m ²	+70.9%	+518.6%	+5.4%	+77.2%
ILSS Layup '2' (distributed)	~55 MPa	+2.7%	−1.2%	+1.0%	−4.6%
ILSS Layup '4' (midplane)	~55 MPa	−8.7%	−12.9%	+7.1%	+0.3%
Single lap shear (SLS)	~15 MPa	+13.3%	+9.1%	+10.8%	+6.2%
Functional focus		Balanced toughening	Shear-dominated toughening	Limited toughening efficiency	Compliance, Mode I critical

these comparisons support the reliability of the unified test set, while also highlighting that quantitative effect sizes remain system-dependent and should not be transferred across epoxy formulations without caution.

In contrast to the established toughening systems, PC and PSU more often exhibit abrupt load drops once a critical condition is reached, indicating a more localized and less tolerant crack growth process. For PC, the pronounced asymmetry between very low Mode I resistance at 42 J/m², corresponding to −88.7% relative to the reference, and clearly elevated Mode II performance at 1770 J/m², corresponding to +77.2%, supports a fundamentally different governing mechanism than classical chemically dominated toughening. The data are consistent with a mechanically dominated interphase contribution under shear, such as interlocking and frictional resistance, whereas the opening-mode response remains adhesion-limited.

Failure observations support this interpretation without repeating the detailed fractographic results. The macroscopic fracture surfaces of the DCB and ENF specimens visually confirm the pronounced mode asymmetry: under Mode I opening loads, PEI and PES exhibit uniform, cohesive failure patterns indicative of excellent structural integration, whereas PC demonstrates severe adhesion-limited interfacial peeling. Conversely, under Mode II shear loading, the PC film undergoes significant cohesive tearing and plastic deformation. This proves that its competitive shear resistance relies heavily on mechanical interlocking and film rupture rather than chemical adhesion. Similarly, the visible surface roughness of PES under shear corroborates its massive capability for plastic yielding. Complementing these fracture mechanics observations, the SLS fracture surfaces are dominated by interfacial failure with light fiber tear rather than cohesive film failure. This indicates that the overall limiting region is often the fiber-adjacent interface, including sizing compatibility with the thermoplastic-modified matrix.

While overall ILSS performance remains robust across all tested materials, the placement-driven trends reveal distinct material-specific behaviors. Stable interphases like PEI and PES effectively redistribute stress and improve load-bearing capacity when distributed across multiple active shear planes. In contrast, increasing the number of PC and PSU layers reduces their interlaminar shear performance, as multiple weak interfaces likely introduce additional sites for early damage initiation. Ultimately, these findings demonstrate that the targeted placement of thermoplastic layers within a composite is a highly effective design tool to deliberately tailor a laminate's compliance and shear properties.

4.2. Mixed mode framework via Benzeggagh–Kenane

Structural delamination in composite applications rarely occurs under pure Mode I (opening) or pure Mode II (shear) loading conditions. Particularly in aerospace structures subjected to bending, out-of-plane forces, or impact events, crack propagation is almost entirely driven by complex, mixed-mode stress states. For the thermoplastic interphases investigated in this study, evaluating this mixed-mode behavior is especially critical. As demonstrated by the pure mode results, the interlayers exhibit severe performance asymmetries, most notably PC, which showed near-zero Mode I adhesion but highly competitive Mode II resistance, or PES, which dominated the shear regime while remaining moderate under opening loads. Evaluating these materials solely on isolated Mode I or Mode II values would therefore yield an incomplete and potentially misleading picture of their true structural viability.

To evaluate crack growth resistance under such realistic conditions and to create comparable, continuous failure envelopes, the semi-empirical Benzeggagh–Kenane (B–K) criterion is widely utilized. The B–K criterion expresses the critical mixed-mode strain energy release rate G_C as a function of the pure mode endpoints and the mode mixity [73]:

$$G_C = G_{IC} + (G_{IIC} - G_{IC}) \cdot \left(\frac{G_{II}}{G_{II} + G_I} \right)^\eta.$$

The exponent η dictates the curvature of the failure envelope. While η is conventionally determined through extensive experimental testing across various modal ratios using a Mixed-Mode Bending (MMB) apparatus, it can also be estimated to evaluate theoretical material performance. According to Benzeggagh and Kenane, this interaction parameter is fundamentally linked to the matrix's fracture behavior; literature

values typically range from $\eta \sim 1.5$ to 2.0 for relatively brittle epoxy systems, while highly ductile matrices exhibit higher values of $\eta \sim 3.0$ or more.

To assess the potential failure boundaries of the investigated interphases without conducting full MMB test campaigns, η values were approximated based on these literature bounds and the fractographic observations of ductility, see Table 4.

According to the semi-empirical criterion established by Benzeggagh and Kenane, the exponent η governs the shape of the mixed-mode fracture envelope, defining how a material's fracture toughness responds to increasing shear mode mixity. Although the exact empirical derivation of η necessitates dedicated Mixed-Mode Bending (MMB) test campaigns, approximating these parameters within physically justified bandwidths already provides highly valuable mechanistic insights into the mixed-mode failure behavior. Therefore, specific parametric ranges for η were estimated based on the measured G_{Ic} to G_{IIc} asymmetry and the fractographically observed failure modes:

M18/1 (Ref.) ($\eta \sim 1.5$ – 2.5): the baseline epoxy exhibits a G_{IIc}/G_{Ic} ratio of approximately 2.7 (999/371 J/m²), consistent with a relatively brittle matrix response. This ratio aligns with the lower range of literature-reported η values for aerospace-grade epoxy systems, indicating a progressive but moderately curved transition between Mode I and Mode II toughness without pronounced shear-dominated plasticity.

PEI ($\eta \sim 1.5$ – 2.5): with a G_{IIc}/G_{Ic} ratio of approximately 2.9 (1707/590 J/m²), PEI exhibits the most balanced toughness increase across both fracture modes. This relatively low ratio, comparable to the baseline epoxy, reflects a proportional elevation of both opening and shear resistance driven by cohesive interphase integration and stable crack bridging, without fundamentally shifting the mixed-mode interaction toward shear dominance. The fractographically observed uniform, cohesive failure morphology under both Mode I and Mode II loading corroborates this interpretation, supporting an η estimate that remains within the baseline epoxy range.

PES ($\eta \sim 3.5$ – 4.5): with a G_{IIc}/G_{Ic} ratio of approximately 14 (6180/440 J/m²), PES exhibits by far the most pronounced toughness asymmetry among the investigated systems. This extreme ratio directly motivates a significantly elevated η estimate, as the B–K exponent must capture a failure envelope that remains relatively flat under opening-dominated conditions before rising steeply toward the Mode II endpoint. The fractographic evidence supports this interpretation: the highly textured, plastically deformed fracture surface under Mode II loading confirms extensive ductile drawing within the fracture process zone, while the Mode I surface remains comparatively uniform, indicating that the energy dissipation mechanism is strongly shear-activated. The chosen η range is therefore physically anchored in both the quantitative mode asymmetry and the observed failure morphology.

PSU ($\eta \sim 2.5$ – 3.5): with a G_{IIc}/G_{Ic} ratio of approximately 6.5 (1053/161 J/m²), PSU occupies an intermediate position between the balanced PEI response and the strongly shear-dominated PES behavior. This moderately elevated ratio motivates a correspondingly moderate η range, reflecting a failure envelope that begins to shift toward shear dominance but without the pronounced plasticity observed in PES. The fractographic evidence supports this interpretation: the comparatively smooth and featureless Mode II fracture surface indicates limited ductile energy dissipation, suggesting that the moderate G_{IIc}/G_{Ic} asymmetry is driven less by plastic yielding and more by the irregular, diffuse interphase morphology reported in [60]. Within the B–K framework, the assigned η range captures this intermediate behavior, the toughness increase under mixed-mode loading is delayed toward higher shear fractions, but the transition remains more gradual than for PES or PC.

PC ($\eta \sim 4.5$ – 6.0): with a G_{IIc}/G_{Ic} ratio of approximately 42 (1770/42 J/m²), PC exhibits by far the most extreme mode asymmetry of all investigated systems, nearly two orders of magnitude separate its opening

Table 4. Benzeggagh–Kenane (B–K) parameters ($G_{Ic}|G_{IIc}|\eta$) describing the mixed mode critical energy release rate G_c

Material	G_{Ic} (Mode I) [J/m ²]	G_{IIc} (Mode II) [J/m ²]	B–K exponent η (approx.)
M18/1 (Ref.) [35]	~ 371	999	~ 1.5 – 2.5
PEI	590 ± 37	1707 ± 174	~ 1.5 – 2.5
PES	440 ± 23	6180 ± 334	~ 3.5 – 4.5
PSU	161 ± 39	1053 ± 200	~ 2.5 – 3.5
PC	42 ± 34	1770 ± 274	~ 4.5 – 6

and shear resistance. This exceptional ratio directly motivates the highest η range assigned in this study. Within the B–K framework, such an extreme exponent describes a failure envelope that remains essentially flat and near-zero under opening-dominated conditions, with a sharp and abrupt transition toward competitive toughness values only at almost pure Mode II mixity. This behavior is physically consistent with the governing mechanism identified for PC: the weak chemical adhesion renders the interphase virtually ineffective under peel-dominated loading, while the deep mechanical interlocking resulting from thermoplastic infiltration into the fiber bed provides substantial frictional and geometric resistance exclusively under shear-dominated sliding. The fractographic evidence corroborates this directly. The Mode I surface shows adhesion-limited interfacial peeling with negligible thermoplastic involvement, whereas the Mode II surface reveals severe cohesive tearing and plastic rupture of the PC film itself, confirming that the energy dissipation mechanism is uniquely and exclusively shear-activated. The upper bound of the η range reflects the possibility that the transition toward competitive toughness occurs even closer to pure Mode II than the mean experimental data suggest, consistent with the relatively high scatter observed in the G_{Ic} results.

Based on these estimated parameters, theoretical failure envelopes were calculated (Figure 15) to conceptualize how the materials might perform under varying mixed-mode loading. These calculated curves highlight the distinct technical potential of each interphase concept:

PEI elevates the overall damage resistance uniformly, implying a balanced toughening effect that is highly suitable for multi-axial loading scenarios. In contrast, the calculated envelope of PES expands massively towards the Mode II domain. This suggests it is highly effective for bending-dominated applications, provided that pure peel stress is structurally mitigated. For PC, the theoretical envelope reveals that despite its competitive shear resistance, its operational window is severely constrained by its near-zero Mode I adhesion intercept. Finally, PSU shows an overall restricted envelope, consistent with its reduced fracture metrics across both primary modes. However, it must be reiterated that without dedicated

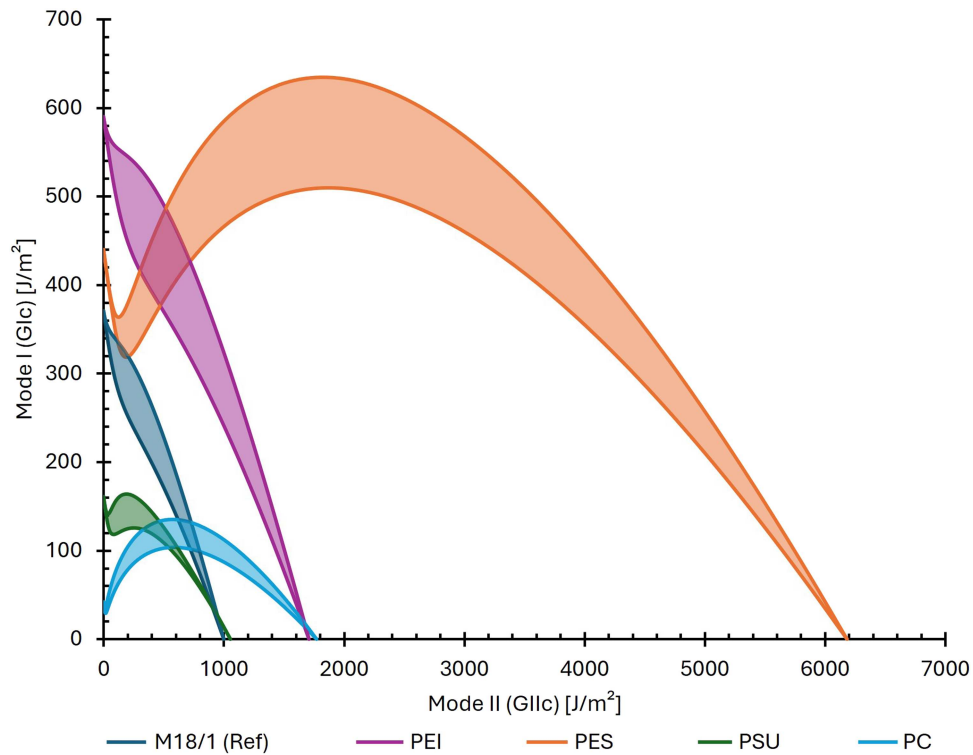


Figure 15. Critical fracture energy envelopes (G_I vs. G_{II}) derived from the experimental G_{Ic} and G_{IIc} data using the B–K semi-empirical law.

mixed-mode testing, these calculated boundaries serve as qualitative estimations of the failure spaces rather than exact quantitative predictions.

4.3. Tailoring compliance and failure behavior by interphase type and placement

The placement sensitivity in ILSS and the asymmetric fracture profiles imply that interphases can tailor laminate behavior beyond peak strength. Distributed interphases can support stress redistribution across multiple interfaces, while a single midplane interphase can concentrate deformation in a defined layer. This enables two design routes: increasing delamination tolerance while maintaining acceptable strength levels, or deliberately introducing a compliant layer to control where and how damage evolves.

PC is the clearest candidate for compliance-oriented concepts because it combines comparatively high Mode II resistance with large deformation capability, but its Mode I weakness implies that such functionality requires explicit control of opening stresses and interface conditions. PEI is better suited for balanced toughening when both opening and shear contributions matter. PES is best positioned for shear-dominated energy absorption, but should be paired with design measures against peel-dominated loading. PSU remains viable where joint shear transfer and ILSS performance are sufficient, but its reduced fracture toughness suggests limited benefit for crack growth resistance. In this work, η is treated as a pragmatic estimate tied to this dataset, rather than as a universal material constant.

5. Conclusion & Outlook

5.1. Conclusion

Reactive thermoplastic interlayers enable the application-driven tailoring of CFRP laminates by shifting key performance aspects into an interphase-governed design space, rather than relying solely on a fixed baseline laminate response. Based on the comparative quasi-static evaluation of these reactively formed interphases, it is evident that quasi-static shear transfer remains feasible at both the joint and laminate levels across all investigated systems. However, this requires that interphase placement is actively treated as a structural design variable rather than a purely processing-related choice.

The core advantage of the interphase concept lies in controlling delamination resistance and mixed-mode behavior. Depending on the thermoplastic system, the mechanical response can be deliberately tuned from more balanced toughening to strongly shear-dominated energy absorption. In this context, the decisive structural limitation is typically not the intrinsic strength of the thermoplastic film itself, but the condition of the fiber-adjacent interface. This specific boundary region governs whether damage evolves in a tolerant, progressive manner or transitions into a more unstable, localized fracture process.

With established systems like PEI and PES, interlaminar functionalization is already feasible for tailoring damage tolerance and compliance. Extending this functionalization to broader processing windows makes lower-temperature candidates such as PC highly attractive, provided their specific interface limitations can be overcome. Ultimately, systems that demonstrate a mechanically dominated contribution under shear, most notably PC, indicate substantial headroom for improvement. Targeted interface engineering and activation strategies aimed at raising opening-mode performance and reducing variability would significantly expand the usable functional design space. Overcoming these interfacial limits will further strengthen the case for application-driven laminate tuning using thermoplastic interphases.

While the unified test set provides robust comparative insights into the distinct failure behaviors of these thermoplastic systems, certain limitations of the current study must be acknowledged. The experimental framework relies on a single specific baseline epoxy system (M18/1). Because thermodynamic compatibility, interdiffusion gradients, and fiber-sizing interactions are highly dependent on the host matrix chemistry, the absolute quantitative fracture values and the exact dimensions of the interphase morphologies are specific to this formulation. Nevertheless, the fundamental mechanistic findings derived here, such as the massive shear-yielding capacity of PES or the shear-activated mechanical interlocking of PC, are expected to translate phenomenologically to other epoxy composite systems, even if the absolute performance scaling varies with different baseline matrices.

5.2. Outlook

The present work motivates a focused set of follow-up steps to strengthen the evidence base and expand applicability. First, the fiber-adjacent interface should be addressed explicitly, since fracture observations indicate that failure can remain localized near the first ply rather than within the thermoplastic film itself. Targeted interface engineering approaches, such as thin neat-resin tie layers, controlled surface activation, or sizing-compatible resin modifications, should be screened to determine their effect on both peak performance and post-peak stability.

Second, the microstructure–property link should be verified directly within the tested specimens. While the present interpretation draws on reported morphologies, follow-up work should correlate microsections and fracture surface features with the mechanical metrics to distinguish morphological effects (penetration, phase-separated topology) from interfacial chemistry effects (adhesion, sizing interaction), particularly for systems showing strong mode asymmetry and high scatter.

Third, placement-driven design rules should be quantified beyond the two configurations investigated here. Systematic stacking studies varying the number, spacing, and distance to the neutral plane of interphase layers are required to map when added shear-active planes enable stress redistribution versus when they introduce additional initiation sites. This should be evaluated under mixed-mode relevant loading states, using the B–K framework as a consistent comparison basis.

Finally, the quasi-static trends should be tested for durability under service-relevant conditions, including fatigue, impact, and environmental exposure. Such campaigns are required to assess variability, damage accumulation, and potential interphase degradation mechanisms that cannot be inferred from quasi-static data alone.

Author contributions

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

Disclosure statement

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ORCID

Matthias Taubert  0009-0005-6819-6796
Andreas Erich Scherer  0009-0000-3081-5719
Bruno Musil  0000-0003-3383-8881
Philipp Höfer  0000-0001-9090-1643

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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