

FRACTURE ANALYSIS OF INTERLEAVED COMPOSITES WITH Self-Same RESIN-RICH LAYERS USING MECHANICAL TESTING, ACOUSTIC EMISSION, AND MICRO-VISUALIZATION

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Abstract

The interlaminar toughness of composite materials remains a critical challenge, limiting their broader application in industries requiring high load-bearing capacity and resistance to delamination. This study investigates innovative interleaving methods to enhance the fracture performance of thermoplastic composites. Using self-same resin-rich layers (RRL), fracture tests under Mode-I loading conditions quantify strain energy release rates for crack initiation and propagation. Digital Image Correlation (DIC) was employed to monitor crack tip propagation, while acoustic emission data was recorded to analyze strain energy release during crack growth. The fracture performance improvement of interleaved components was limited. The results further underscore the critical role of strong fiber-matrix bonding and the predominance of fiber bridging during fracture in reference samples under separation forces (Mode I). These factors contribute to their high G_{Ic} values, to the extent that the introduction of an interlaminar resin layer—designed to promote increased matrix plastic deformation at the crack tip—fails to yield significant improvement in G_{Ic} values.

Keywords: Interlaminar Toughness; Resin-Rich Layers; Fracture Toughness; Mode-I

1 Introduction

Poor interlaminar strength remains a major limitation of CFRP laminated composites, often leading to delamination and structural failure [1-5]. In order to improve delamination resistance, interleaving methods have been widely investigated by introducing films, particles, or micro/nano-fibers into the interlaminar region [1, 2, 7]. The effectiveness of these toughening strategies is commonly evaluated using standardized fracture tests, particularly the double cantilever beam (DCB) test under Mode-I loading, which is one of the most established methods for characterizing interlaminar fracture toughness in interleaved composite laminates. In addition to mechanical testing, scanning electron microscopy (SEM) is widely employed after fracture tests to examine fracture surfaces, identify crack propagation paths, and reveal the toughening mechanisms responsible for improved delamination resistance [8].

Previous studies have shown that interleaving can significantly influence Mode-I fracture behavior. Resin-rich layers can accommodate larger plastic deformation zones near the crack

tip, increasing energy dissipation and improving interlaminar toughness [2, 9]. Discontinuous resin-rich regions may also contribute to crack arrest during delamination. Thermoplastic interleaves have similarly demonstrated improved Mode-I fracture toughness, with enhancements attributed to mechanisms such as plastic deformation, crack deflection, fiber bridging, and fiber breakage [10, 11]. Thermoplastic particles and nanoparticles have also been reported to improve DCB fracture toughness by forcing the crack to follow a more tortuous path. Hybrid films and electro spun nanofiber interleaves have further improved crack propagation resistance by promoting crack-tip plasticity, veil bridging, and increased energy dissipation [12]. Therefore, combining DCB testing with SEM fractography provides a reliable framework for evaluating both the quantitative fracture toughness and the underlying failure mechanisms of interleaved composite laminates.

In this study, the manufacturing process of CF/PEKK thermoplastic composites was modified to create interlaminar self-same resin-rich layers (RRLs) by promoting the flow of the primary matrix resin into the interlaminar region during consolidation. This approach preserves the overall fiber volume fraction, thereby minimizing potential reductions in in-plane properties and avoiding the weight penalty associated with conventional resin-film interleaving. The RRLs were produced in two configurations, namely pocket-type and continuous formats. Cross-sectional microscopy was used to assess RRL morphology and laminate void content, while DCB tests were conducted to evaluate Mode-I interlaminar fracture toughness. SEM was further employed for post-fracture surface analysis. The performance of the modified laminates was benchmarked against reference specimens.

2 Material

The thermoplastic plates utilized in this study was Toray Cetex® prepreg with the manufacturing designation TC1320 CF/PEKK (polyetherketoneketone). The plates exhibit a fiber volume fraction of 66 wt%, a glass transition temperature of 160°C, a melting temperature of 337°C and nominal thickness of 0.15 mm [13].

3 Laminate Manufacturing with self-same resin rich layer

Two CF/PEKK plies were first surface-patterned and placed at the laminate midplane to form a self-same interlaminar resin-rich layer during consolidation. A total of 26 plies were hot-pressed at 380 °C and 0.42 MPa to obtain laminates suitable for DCB testing. By varying the surface-patterning process, two RRL morphologies were produced: pocket-type (RRL2) and continuous wavy (RRL3). Both surface patterning and final consolidation were performed in a hot press, with maximum processing conditions of 400 °C and 0.54 MPa. Each batch produced 300 × 300 mm laminates with a thickness of 4.0 ± 0.4 mm. The specimen number and dimensions for the DCB tests were selected based on the procedure adopted in a previous study [7].

Figure 1 shows the cross-sections of the fabricated samples: Ref., without an interlaminar RRL; RRL2, with a resin-pocket structure; and RRL3, with a continuous wavy RRL. No autoclave post-processing was applied, as extended hot-press consolidation was sufficient to eliminate interlaminar voids.

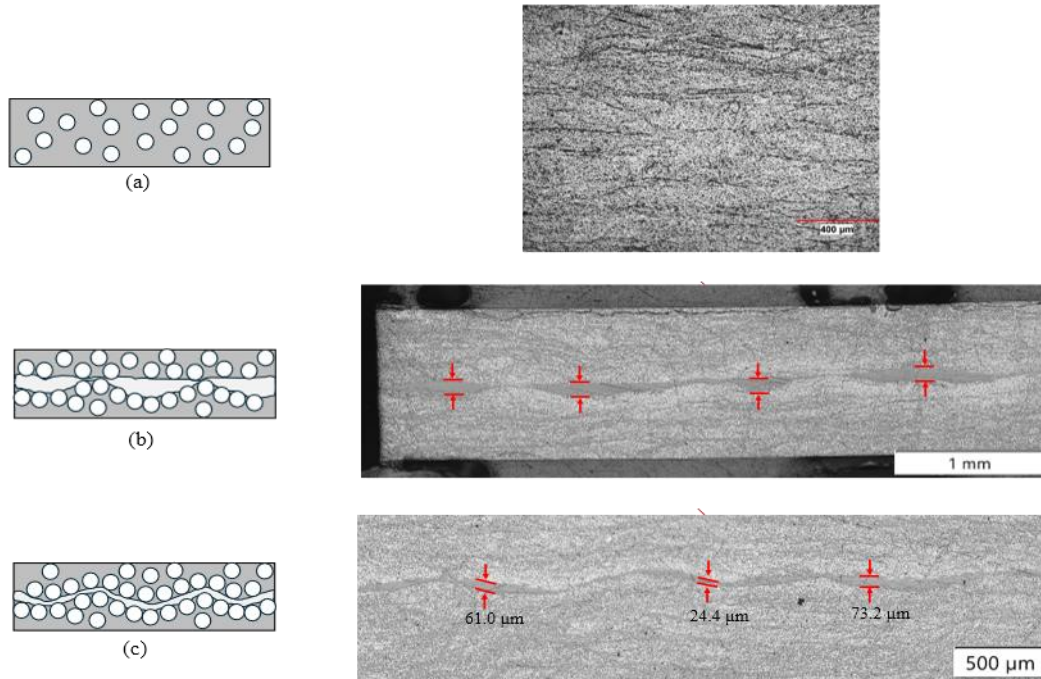


Figure 1. Cross section of (a) reference (Ref.), (b) Pocket format RRL (RRL2), (c) wavy continuous RRL (RRL3) samples.

4 Mode-I fracture Test

The fracture toughness of the three sample groups was evaluated under Mode-I loading using DCB tests in accordance with ASTM D5528 [14]. Crack-tip propagation was monitored by DIC, while AE signals were recorded during fracture using an attached sensor. Figure 2 shows the DCB test setup with the AE sensor and DIC camera.

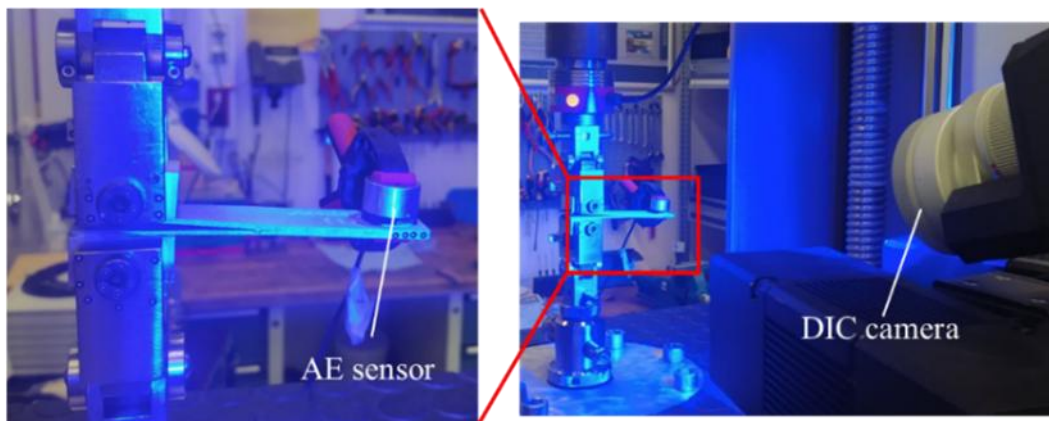


Figure 2. DCB test and integrated AE sensor and DIC systems.

The tests were performed at a constant displacement rate of 2 mm/min. Crack propagation distance, applied load, and crack opening displacement were continuously recorded using DIC system integrated in an Instron test machine. Load–crack opening displacement and strain energy release rate–crack propagation curves were then generated. The Mode-I critical strain

energy release rate G_{Ic} was calculated using the modified beam theory defined in ASTM D5528 (Equation 1).

$$G_{Ic} = 3P\delta / (2b(a + |\Delta|)) \quad (1)$$

where P [N] is the applied load, δ [mm] is the load point displacement, b [mm] is the sample width, a [mm] is the crack length, from loading point to crack tip, and Δ [mm] is the correction factor. Based on ASTM D5528 two methods were used for G_{Ic} initiation calculation, nonlinear visual crack initiation and the 5% offset/maximum load method. Since the latter overestimated the initiation values, the visual nonlinear crack initiation criterion was adopted for G_{Ic} calculation criteria.

Figure 3 compares the Mode-I interlaminar fracture toughness of the reference laminate and the modified laminates containing self-same resin-rich layers. Both crack initiation and propagation values are presented, allowing the influence of resin-rich layer morphology to be evaluated.

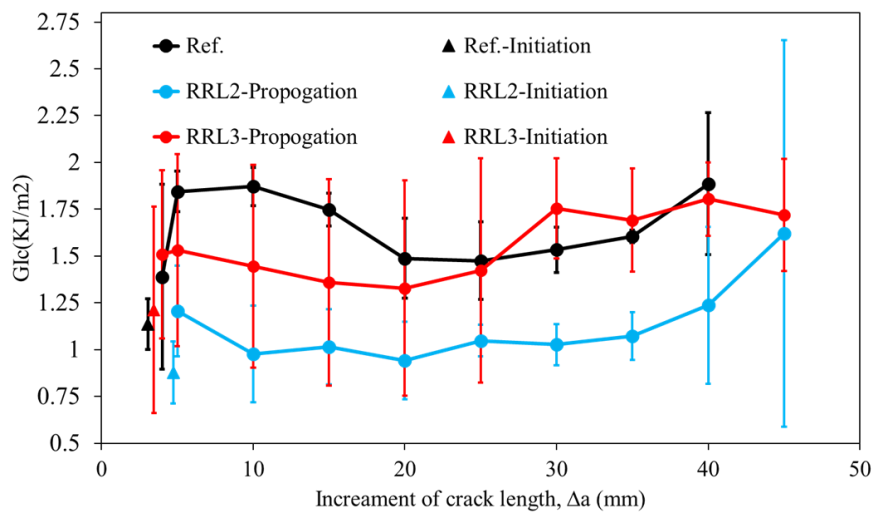


Figure 3. Mode-I strain energy release rate (G_{Ic}) versus crack length increment for the tested samples.

The reference sample shows relatively high G_{Ic} values immediately after crack initiation, followed by a reduction in fracture toughness during subsequent crack propagation. Variations in the fiber-bridging effect at different stages of crack growth can be potential reason behind this behavior.

The RRL2 sample, containing resin-rich pockets, shows the lowest G_{Ic} values during crack propagation. This indicates that the discontinuous pocket-type resin-rich regions do not provide an effective toughening mechanism under Mode-I loading. Instead of promoting a stable tortuous crack path, the pocket morphology may create local resin-rich zones and interfacial irregularities that facilitate easier crack propagation or localized damage. The relatively large error bars also suggest scatter in the fracture response, likely due to the nonuniform interaction between the propagating crack and the discrete resin pockets.

In contrast, the RRL3 sample, containing a continuous wavy resin-rich layer, exhibits the highest overall propagation resistance. Although its G_{Ic} values are initially lower than those of the reference sample, they gradually increase and eventually exceed the reference values. This suggests that the continuous wavy morphology becomes increasingly effective during stable crack growth by increasing crack-path tortuosity and promoting greater matrix-dominated energy dissipation.

A similar trend is also observed for the crack initiation values. The RRL3 sample shows a higher initiation G_{Ic} than the reference laminate, whereas the RRL2 sample exhibits lower initiation toughness than both the reference and RRL3 samples.

Overall, the results indicate that the effectiveness of self-same resin-rich layers strongly depends on their morphology. The pocket-type RRL2 configuration reduces Mode-I fracture toughness, while the continuous wavy RRL3 configuration gradually improves crack growth resistance and provides the most effective toughening response.

5 Fractography

The fracture surfaces of the DCB specimens were examined using scanning electron microscopy (SEM). Figure 4 presents the SEM images of the Ref. sample (a–d), the RRL2 sample (e–g), and the RRL3 sample (h–j).

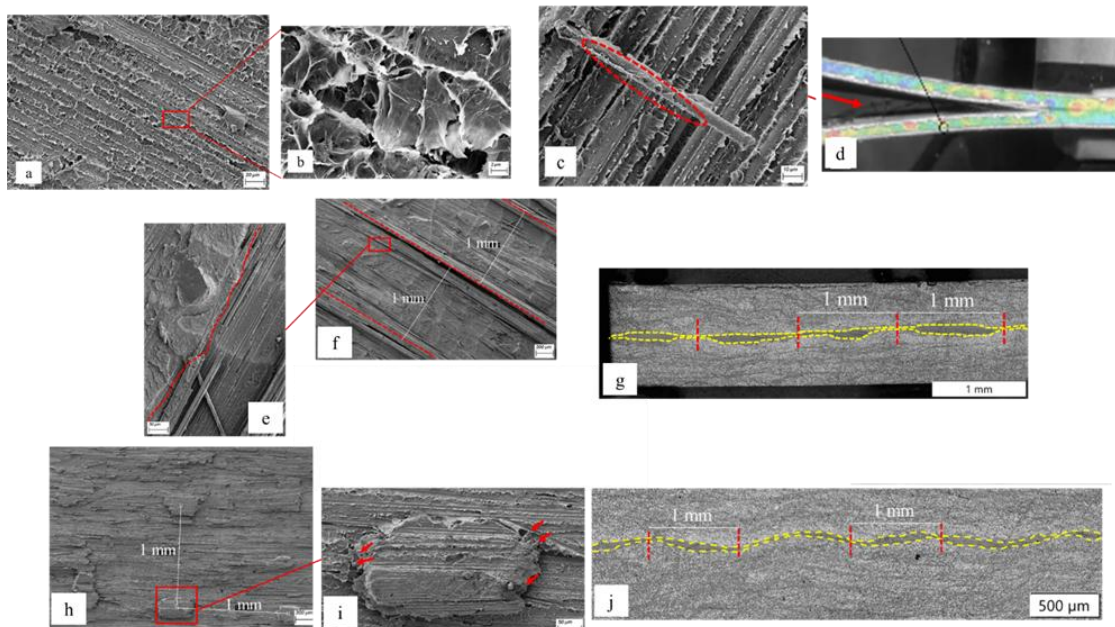


Figure 4. SEM images of fracture surfaces of Ref. sample (a–d), the RRL2 sample (e–g), and the RRL3 sample (h–j).

SEM observations reveal distinct fracture mechanisms depending on the interlaminar morphology. The Ref. sample exhibits a relatively uniform fracture surface characterized by fiber–matrix debonding, resin plastic deformation, broken fibers, and resin residues on the fiber surfaces. These features indicate effective fiber bridging and strong fiber–matrix adhesion, which contribute to the higher G_{Ic} values of the reference laminate. DIC observations during Mode-I fracture further confirm the presence of extensive fiber bridging.

For the RRL2 sample, which contains pocket-type resin-rich layers, the fracture surface shows periodically spaced fiber-debonding lines. Along these lines, fiber breakage and debonding dominate, whereas the regions between them are mainly governed by resin plastic deformation. Comparison with cross-sectional microscopy confirms that these debonding lines correspond to the thinnest regions of the resin pockets, with an approximate spacing of 1 mm.

The RRL3 sample, containing a continuous wavy resin-rich layer, exhibits a fracture surface dominated by resin-rich islands and broken fiber edges. The spacing between these resin islands corresponds closely to the wavelength of the wavy RRL, confirming that the fracture morphology

is governed by the imposed interlaminar pattern. Continuous wavy RRLs promote stronger mechanical interlocking and a more tortuous crack path compared with pocket-type RRLs. Overall, the reference laminate shows more uniform plastic deformation and dominant fiber-bridging mechanisms, leading to more stable crack propagation. In contrast, RRL2 and RRL3 specimens introduce localized toughening features and tortuous crack paths; however, the resin-rich layers partially suppress fiber bridging, which remains the dominant toughening mechanism under Mode-I loading. Thus, while self-same RRLs significantly alter the fracture mechanism, the more uniform fracture behavior of the reference laminate is associated with superior Mode-I fracture performance.

6 Conclusion

A novel manufacturing approach was proposed to create resin-rich layers in thermoplastic composites without using separate resin films. The method promotes primary matrix flow into the interlaminar region, forming self-same resin-rich layers while maintaining the overall fiber volume fraction and in-plane properties. Although the overall fracture toughness improvement was limited, the continuous wavy RRLs enhanced fracture resistance during initiation and progressive crack propagation.

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References

1. Chen SF, Jang BZ. Fracture behaviour of interleaved fiber-resin composites. *Compos Sci Technol.* 1991;41(1):77–97. doi:10.1016/0266-3538(91)90054-S
2. Hojo M, Ando T, Tanaka M, Adachi T, Ochiai S, Endo Y. Modes I and II interlaminar fracture toughness and fatigue delamination of CF/epoxy laminates with self-same epoxy interleaf. *Int J Fatigue.* 2006;28(10):1154–1165. doi:10.1016/j.ijfatigue.2006.02.004
3. Barzegar A, Karimi S, Sukur EF, Sas HS, Yildiz M. Effect of fiber orientation on temperature history during laser-assisted thermoplastic fiber placement. *J Reinf Plast Compos.* 2022. doi:10.1177/07316844221143448
4. Karimi S, Louyeh PY, Barzegar A, Yildiz M, Sas HS. A study of post-processing for AFP-fabricated thermoplastic composites with void content and air permeability characterization. *J Reinf Plast Compos.* 2023. doi:10.1177/07316844231166078
5. Barzegar A, Karimi S, Sas HS, Yildiz M. Enhancing interlaminar bonding quality estimation in laser-assisted fiber placement of CF/PEKK composites: a correction factor approach for improved prediction of intimate contact. *Compos Part A Appl Sci Manuf.* 2024;181:108095. doi:10.1016/j.compositesa.2024.108095
6. Sasidharan S, Anand A. Interleaving in composites for high-performance structural applications. *Ind Eng Chem Res.* 2023;62(1):16–39. doi:10.1021/acs.iecr.2c03061

7. Karimi S, Barzegar A, Al-Nadhari A, et al. Performance and failure mechanisms of interleaved E-glass/epoxy composites using knitted fiber-reinforced adhesive films. *Polym Compos.* 2025;46(16):15255–15273. doi:10.1002/pc.30127
8. Mohammadi R, Saeedifar M, Toudeshky HH, Najafabadi MA, Fotouhi M. Prediction of delamination growth in carbon/epoxy composites using a novel acoustic emission-based approach. *J Reinf Plast Compos.* 2015;34(11):868–878. doi:10.1177/0731684415583166
9. Sacchetti F, Grouve WJB, Warnet LL, Villegas IF. Effect of resin-rich bond line thickness and fibre migration on the toughness of unidirectional carbon/PEEK joints. *Compos Part A Appl Sci Manuf.* 2018;109:197–206. doi:10.1016/j.compositesa.2018.02.035
10. Thorn TDS, Wang Y, Guo H, Lu L, Liu Y, Bilotti E, et al. Discontinuous interleaving strategies for toughening, damage sensing and repair in multifunctional carbon fibre/epoxy composites. *Compos Part A Appl Sci Manuf.* 2024;185:108320. doi:10.1016/j.compositesa.2024.108320
11. Too DK, Kumar S, Shin DH, Kim YH. Synergistic toughening of CFRP laminates using melt-infused thermoplastic interleaves and toughened epoxy. *Compos Commun.* 2025;53:102192. doi:10.1016/j.coco.2024.102192
12. Zhang H, Bharti A, Li Z, Du S, Bilotti E, Peijs T. Localized toughening of carbon/epoxy laminates using dissolvable thermoplastic interleaves and electrospun fibres. *Compos Part A Appl Sci Manuf.* 2015;79:116–126. doi:10.1016/j.compositesa.2015.09.024
13. Toray. Toray Cetex® TC1320 PEKK product data sheet. 2019.
14. ASTM International. ASTM D5528-01: Standard test method for mode I interlaminar fracture toughness of unidirectional fiber-reinforced polymer matrix composites. West Conshohocken, PA: ASTM International; 2014.