

Optimizing layering configurations for enhanced flexural and impact performance in perforated Al foil/glass fiber/epoxy hybrid laminates

Harini Sosiati^{*1)}, Arya Setyaki Abdillah¹⁾, Anugrah Cahya Farisa¹⁾, Ankas Pamasti¹⁾, Pinta Astuti²⁾, Fetra Venny Riza³⁾ and Satoshi Hata⁴⁾

¹⁾Department of Mechanical Engineering, Faculty of Engineering, Universitas Muhammadiyah Yogyakarta, Yogyakarta 55183, Indonesia

²⁾Department of Civil Engineering, Faculty of Engineering, Universitas Muhammadiyah Yogyakarta, Yogyakarta 55183, Indonesia

³⁾Department of Civil Engineering, Faculty of Engineering, Universitas Muhammadiyah Sumatra Utara, Medan 20238, Indonesia

⁴⁾Department of Advanced Materials Science and Engineering, Faculty of Engineering Sciences, Kyushu University, Fukuoka 8168580, Japan

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Abstract

Lightweight panels for aircraft, vehicles, and protective structures need to absorb damage without a large weight penalty. In practice, many fiber metal laminates (FMLs) still struggle with shaping or curvature and can delaminate early, largely because the metal plies are relatively thick and the metal/polymer interface is prone to separation. An alternative hybrid laminate was developed using thin perforated Al foil (A) (~0.1 mm; perforations ~150 μm) combined with woven glass fiber (G)/epoxy, and the influence of stacking sequence on flexural and impact behavior was examined. Five configurations were manufactured by hand lay-up and evaluated using three-point bending (ASTM D790) and unnotched Charpy impact (ASTM D6110). Fracture surfaces and interfacial features were examined by optical microscopy and SEM. The laminates showed a clear, design-driven performance balance between bending strength and impact resistance. The symmetric, staggered lay-up V4 (GGGAAAGAAAGGG) delivered the highest flexural strength (229.74 MPa), whereas the metal-rich core V3 (GGGAAAAAAGGGG) produced the highest impact resistance (123 kJ/m²). Microstructural characterization indicates that resin flow through the perforations forms mechanical “anchors” at the Al foil/epoxy interface, limiting adhesive-type separation and promoting more progressive damage through cohesive cracking and controlled delamination. The results suggest that perforated Al foil/glass/epoxy hybrids offer a practical route to tune stiffness and toughness for lightweight skins/panels and protective components requiring high bending performance with moderate-to-high impact resistance.

Keywords: Perforated Al foil, Glass fiber, Epoxy, Hybrid laminated composite, Mechanical properties

1. Introduction

The demand for lightweight, damage-tolerant structural materials continues to increase in aerospace, automotive, and protective systems. Similar requirements are now also relevant for external medical support devices where stiffness, durability, and resistance to accidental impact are desirable. Conventional fiber-reinforced polymers (FRPs) provide excellent specific strength and stiffness, but their tolerance to impact-induced damage and their residual load-carrying capability after damage can be limited. Metals, by contrast, offer ductility and progressive load redistribution, yet they typically add weight. Hybrid concepts have therefore gained attention, particularly fiber metal laminates (FMLs), which combine metallic layers and FRP plies to leverage the advantages of both classes of materials [1, 2]. A widely cited example is GLARE (glass fiber/epoxy with aluminum sheets), which has been adopted in aircraft fuselage structures because of its high damage tolerance and resistance to fatigue-crack growth [3, 4].

Evidence across the literature indicates that FMLs can deliver attractive mechanical performance, but the benefits are not “automatic”, they depend strongly on architecture and interface quality. For example, Kadhum et al. [1] proposed epoxy-novolac-based laminates for aircraft wing concepts and reported gains in stiffness and durability relative to conventional FRP baselines. Khan et al. [5] discussed carbon fiber/epoxy laminate concepts for unmanned aerial vehicle (UAV) components, highlighting the need to balance lightweighting targets with structural reliability when compared with metallic aerospace materials. In a different material system, Khalili et al. [6] showed that basalt fiber metal laminates (BFMLs) can absorb more energy than basalt fiber-reinforced epoxy (BFRE) in tensile and bending loading. Overall, these studies support the versatility of FMLs, while repeatedly pointing to the same practical message: laminate sequence and interfacial integrity often govern the final response as much as (or more than) the constituent choice itself.

Stacking configuration is particularly influential under flexural and impact loading because it controls through-thickness stress gradients, shear transfer, and the location where damage initiates. Zareei et al. [7] demonstrated that changing the configuration in Al sheet-hybrid basalt/jute composites lead to pronounced differences in bending and impact performance. At the same time, the

*Corresponding author.

Email address: hsosiati@ft.umsy.ac.id

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metal/polymer interface is widely recognized as a dominant factor for durability and damage tolerance in layered hybrid systems. Dhanaraj et al. [8] reported that perforated Al sheets can enhance adhesion with epoxy and improve impact-related performance, while other studies examined the influence of fiber orientation ($0^\circ/90^\circ$) within Al/FRP/Al systems [9, 10]. More broadly, several studies have emphasized that an explicit integrity assessment of metal/polymer/metal layered composites is essential for structural reliability, and that the choice of test method is critical for judging interface quality [11]. Consistent with this, metal–polymer sandwich systems incorporating a perforated metallic interlayer indicate that perforation-assisted interlocking can materially change the mechanical response and interface reliability [12]. These findings support the rationale that perforations can promote resin anchoring and reduce the likelihood of premature delamination.

In addition to aerospace and transportation, Al-based hybrid laminates have also been explored for orthoses and external-support components, highlighting the relevance of lightweight hybrid architectures beyond purely structural sectors [13]. One practical limitation in many conventional FML designs is the use of metallic layers that are relatively thick, often on the order of a few tenths of a millimeter (commonly reported around 0.3–0.5 mm in many systems), which can reduce formability and restrict suitability for thin, contoured components such as UAV fuselage skins and tail planes [14]. Interface vulnerability under dynamic loading further complicates this problem. El-Sagheer et al. [10] reported that fiber/epoxy layers placed between Al plates can exhibit partial delamination at 0° and more pronounced delamination at 90° after tensile-impact loading, underscoring the sensitivity of metal/polymer interfaces to loading direction and mixed deformation modes. Complementary fracture observations on Al/GFRP/Al sandwich panels under quasi-static and impact tensile loading also highlight that damage evolution is governed by competition between intralaminar cracking and interlaminar debonding at the Al/FRP interface, and that the response is highly dependent on laminate architecture [15]. Overall, these results reinforce that achieving thin, formable hybrid laminates with improved damage tolerance requires not only thinning the metal layer, but also explicitly designing the interface and the stacking strategy to control the damage path.

To improve conformability while retaining the benefits of a metallic layer, thin aluminum foils (≤ 0.1 mm) have increasingly been considered as alternative metal plies capable of following tighter curvatures. In this context, perforation becomes especially relevant because it can convert a primarily adhesive interface into a combined adhesive–mechanical interlock, potentially stabilizing load transfer even when bending and impact drive through-thickness separation tendencies. Rajamurugan et al. [16] incorporated perforated aluminum foils into natural fiber laminates and associated improvements in drilling and impact-related performance with mechanical interlocking, suggesting that perforation can be a scalable route to strengthen interfacial bonding in hybrid laminates. Recent advances in laminate architecture including fatigue-relevant designs such as NiTi sheet/CF/PEKK FMLs [17] and aluminum sandwich structures with woven-ply natural fiber/glass hybrid cores [18], further emphasize that micro-architectural design and interface tailoring are central to achieving simultaneous improvements in stiffness, energy absorption, and damage stability.

Despite these developments, studies that combine thin perforated aluminum foil with woven glass fiber/epoxy laminates remain limited, particularly when the goal is to connect macroscopic performance to the underlying fracture mechanisms. Our prior work suggested that symmetric stacking can improve flexural resistance by reducing shear concentration and stabilizing load transfer [19]. Similarly, Karthick et al. [20] reported that ordered and symmetric configurations can reduce the propensity for delamination under impact loading. However, systematic investigations that explicitly map how perforated Al foil placement and woven glass fiber stacking sequence jointly control flexural stiffness, impact energy absorption, and damage evolution supported by microstructural evidence are still scarce.

Therefore, this study systematically investigates the effect of stacking configuration on the mechanical response of thin, perforated Al foil/woven glass fiber/epoxy hybrid laminates. Three-point flexural testing (ASTM D790) and unnotched Charpy impact testing (ASTM D6110) were conducted, and optical/SEM observations were used to examine fracture mechanisms and interfacial bonding characteristics. The novelty of this work is the controlled combination of thin perforated Al foil plies with deliberate stacking variations to clarify the interplay between flexural stiffness, impact energy absorption, and interface-driven damage development. The outcomes provide practical guidance for lightweight skins and panels in UAV and automotive structures and may be extended to impact-resistant, non-implantable external medical support devices, subject to appropriate material selection and biocompatibility validation.

2. Materials and methods

2.1 Raw materials

The hybrid laminates were produced from three main constituents: (1) a commercial aluminum foil (A) with an average thickness of ~ 0.1 mm; (2) a woven E-glass fabric (G) used as the reinforcing layer without additional surface treatment; and (3) a room-temperature-curing epoxy system used as the polymer matrix. To promote interfacial bonding, the aluminum foil was mechanically perforated using a derma-roller, creating irregular circular openings with an average diameter of ~ 150 μm (Figure 1). Epoxy and its corresponding hardener were mixed at a 1:1 ratio prior to lamination. The E-glass fabric and epoxy system were obtained from local suppliers.

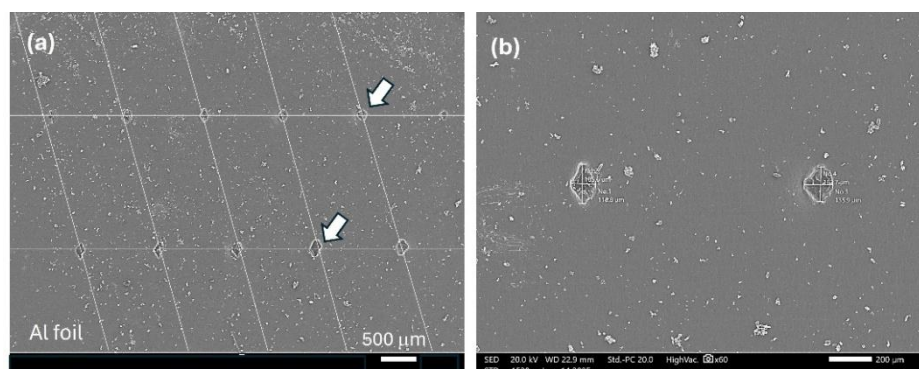


Figure 1 (a) Al foil with irregular perforated holes (white arrows), and (b) hole size measure.

2.2 Laminate configuration

Five stacking configurations were designed to investigate the influence of A and G arrangements (Figure 2). Each laminate consisted of six layers of A and seven layers of G, with a consistent overall thickness across samples. The stacking sequences were: **V1**: GAGAGAGAGAG (alternating sequence), **V2**: GGAAAAAAGGGGG (clustered Al foil at core, symmetric), **V3**: GGGAAAAAGGGG (Al-rich central region, symmetric), **V4**: GGGAAAGAAAGGG (balanced symmetric with staggered Al distribution), and **V5**: AAAGGGGGGAAA (symmetric with Al skins). All laminates used identical numbers of A and G plies and were fabricated to a comparable overall thickness to ensure fair comparison.

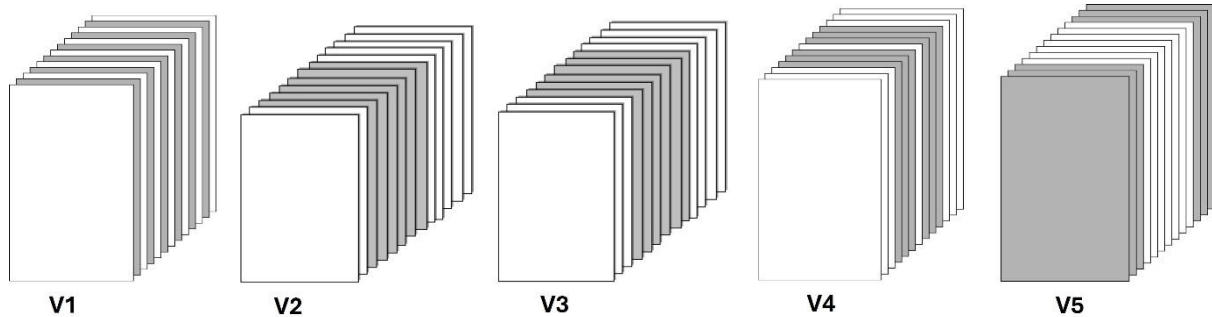


Figure 2 Layering configuration of G-A/epoxy composite specimens.

2.3 Fabrication of composites

Figure 3 shows a route process of laminated hybrid composites that was conducted by hand lay-up using a vacuum hot press molding at a hydraulic pressure of 500 psi at 100°C for 60 minutes. A and G were cleaned by spraying ethanol (96%) on their surfaces and drying with a hand dust blower to avoid the dirty surfaces (Figure 3a and 3b). The A-surface was then perforated using a derma-roller as a mechanical treatment (Figure 3c), but nothing was conducted on the G-surface. Epoxy and its corresponding hardener were manually mixed (Figure 3d) and continued with an electrical mixer at 80 rpm for 5 minutes to reach a homogeneous epoxy (Figure 3e). Subsequently, Figure 3f to 3i illustrated the arrangement of A, G, and epoxy in the molding for each layering configuration; then, the molding was closed (Figure 3j) and finally inserted into the hot press (Figure 3k). During lay-up, the epoxy penetrated the micrometer-scale perforations in the Al foil and, after curing under hot-press conditions, solidified into through-thickness anchors (“epoxy keys”). This perforation-induced anchoring provides mechanical interlocking across the Al/epoxy interface, complementing interfacial adhesion and improving resistance to interfacial debonding.

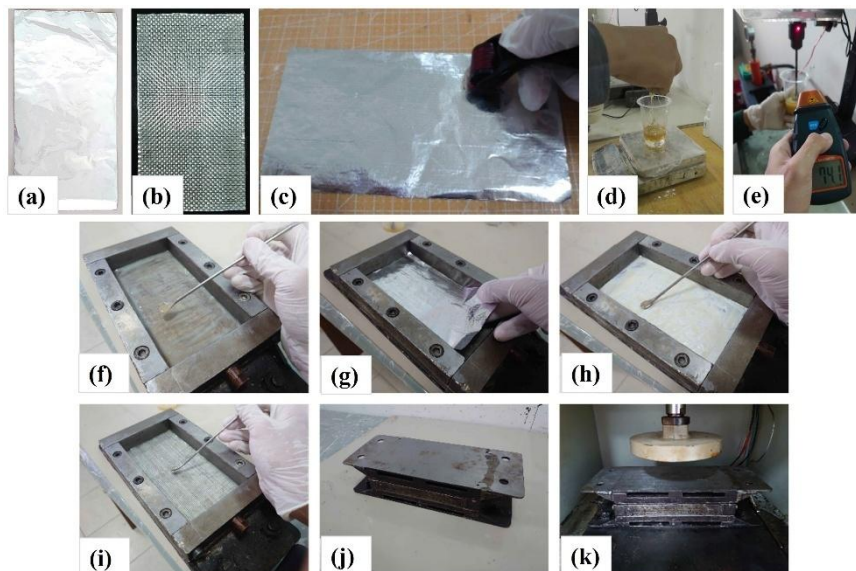


Figure 3 A route fabrication process of laminated hybrid composites. (a) and (b) Al foil and woven E-glass, respectively, (c) perforation of Al foil surface, (d) and (e) mixing epoxy and hardener, (f) to (k) arrangement of A, G, and epoxy to fabricate the laminated composites.

2.4 Mechanical testing and characterization

A three-point bending test was carried out using a universal test machine (UTM, Zwick/Roell Z020) in accordance with the ASTM D790 standard. The specimen was sized 127 mm × 12.7 mm × 3.2 mm and was subjected to a preload of 0.2 N at a crosshead speed of 2 mm per minute. The span size employed was 80 mm. The same UTM was used for the Charpy impact test (ASTM D6110), using the identical specimen measurements of 127 mm × 12.7 mm × 3.2 mm and leaving the flatwise orientation test unnotched. For the impact test, the absorbed energy was recorded, and the impact strength (kJ/m²) was calculated by normalizing the absorbed energy by

the specimen cross-sectional area, following ASTM D6110. In addition, load–deflection curves were reconstructed from the machine-exported flexural stress–strain data using the ASTM D790 relations ($L = 80$ mm), and for each configuration a single representative curve from the five replicates (closest to the mean response) is presented.

For each configuration, five samples were tested to ensure repeatability. After testing, each fractured specimen was observed with an optical microscope (Olympus SZ61) to evaluate crack propagation and delamination. Scanning electron microscopy (SEM) observations were then performed on representative specimens (V3 and V4) (JSM-IT210) to examine interfacial bonding and the local failure morphology.

3. Results and discussion

3.1 Flexural properties and stacking symmetry

The flexural performance of the five hybrid laminate configurations (V1–V5) is presented in Figure 4. The results show a clear influence of stacking sequence on flexural strength. Among the tested laminates, V4 (GGGAAAGAAAGGG) had the highest average flexural strength at 229.74 ± 14.23 MPa. V3 (154.33 ± 12.04 MPa) and V2 (138.67 ± 5.79 MPa) came in second and third. V1 exhibited the lowest flexural strength (91.50 ± 4.37 MPa), whereas V5 showed a moderate-to-high value (190.03 ± 7.29 MPa), which remained below V4 (230.06 ± 10.69 MPa).

The V4 layout performs best for a straightforward reason: it is symmetric and it does not “pile up” the Al in one region. During bending, the outer G/epoxy plies take most of the tensile/compressive stress, and the foil layers in the interior help keep the section stiff and stable. Similar benefits of symmetric, ordered stacking, especially in terms of reduced shear concentration and more stable load transfer have been reported in related hybrid laminates [19, 20].

By contrast, the lowest-performing configuration (V1) is associated with early interfacial opening and rapid delamination growth, which limits the ability of the laminate to sustain increasing bending load. Similar sensitivities to stacking imbalance have been reported in aluminum–hybrid natural-fiber systems, where non-symmetric arrangements promote interfacial instability and reduce flexural strength [7]. While V5 still reached relatively high flexural strength, its performance remained below V4, suggesting that bending-optimized designs benefit not only from stiffness, but also from damage stability: i.e., suppressing early interface opening. Overall, these results emphasize that symmetric/balanced stacking is the most reliable route for flexural-dominated applications, such as lightweight skins and panels. Notably, V4 also exceeds flexural strengths reported for some perforated-sheet or natural-fiber-based hybrid laminates (often below ~ 100 MPa) [8, 16], although direct numerical comparison is limited by differences in laminate thickness, constituent materials, and test conditions.

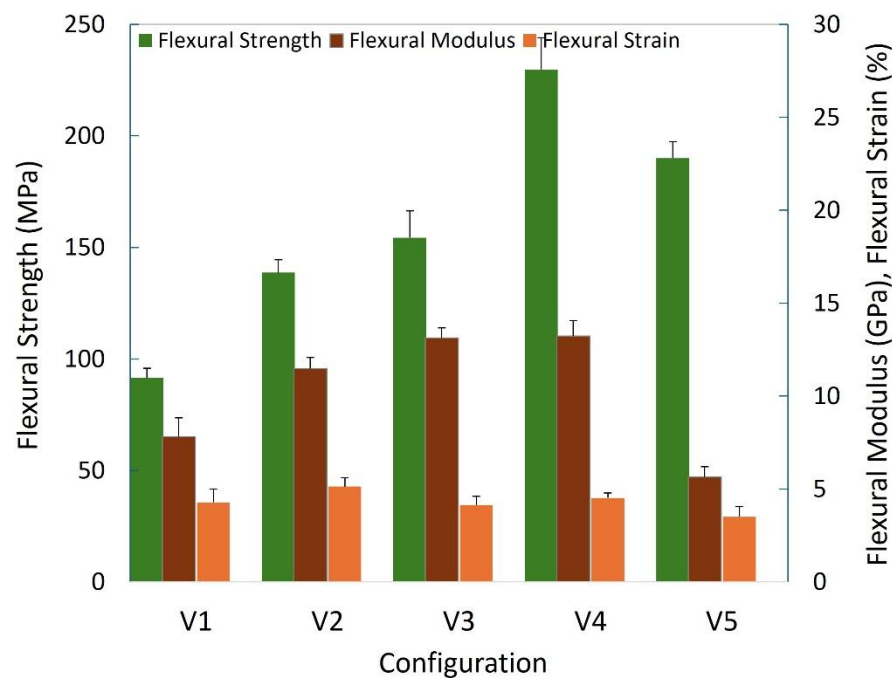


Figure 4 Flexural properties of V1–V5 hybrid laminate configurations. Error bars indicate ± 1 standard deviation ($n = 5$).

Additionally, Figure 5 shows representative load–deflection curves for V1–V5 under three-point bending ($L = 80$ mm). The curves follow a similar linear trend at small deflections, but they separate once damage initiates, as reflected by differences in the slope, the peak load level, and the way the response softens after the peak.

V4 stands out by reaching the highest peak load (≈ 210 – 220 N) and by developing load quickly in the early stage. This steep rise is in line with its higher bending stiffness and strength. After the peak, however, the load drops more sharply than in the other lay-ups, which suggests that once a critical damage mode is triggered, such as local interface separation and/or compressive-side cracking, the remaining intact region is less able to sustain the same load.

A different behavior is seen in V5. Although its peak is lower (≈ 155 – 165 N), the curve extends to larger deflections, and the post-peak reduction is more gradual. In practical terms, V5 does not maximize peak capacity, but it maintains a meaningful load level as bending continues, which is typical of a more progressive damage development and greater energy dissipation during the later stage of deformation.

The responses of V2 and V3 fall between these two patterns. V3 reaches a moderate peak ($\approx 145\text{--}150\text{ N}$) and then decreases smoothly, indicating a stable transition into softening rather than a sudden loss of capacity. V2 is notable for its long post-peak region, maintaining load up to $\sim 17\text{--}18\text{ mm}$ deflection. Across the five replicates, this extended tail repeatedly appeared for V2, supporting its description as a more deformation-tolerant configuration, although its initial stiffness is lower than that of V3–V4. Finally, V1 shows the lowest peak ($\approx 80\text{--}90\text{ N}$) and the weakest post-peak retention, which is consistent with early damage initiation and limited ability to redistribute bending stresses in the alternating lay-up.

Overall, these curves provide a useful complement to the summary metrics in Figure 4 by illustrating *how* each configuration loses load-carrying capacity. V4 prioritizes stiffness and peak load, whereas V5 (and, to a lesser extent, V2) sustains load over larger deflections. This trend aligns with the interfacial observations in Section 3.5, where perforation-assisted anchoring and stacking arrangement influence whether bending damage proceeds abruptly or evolves in a more gradual, energy-dissipating manner.

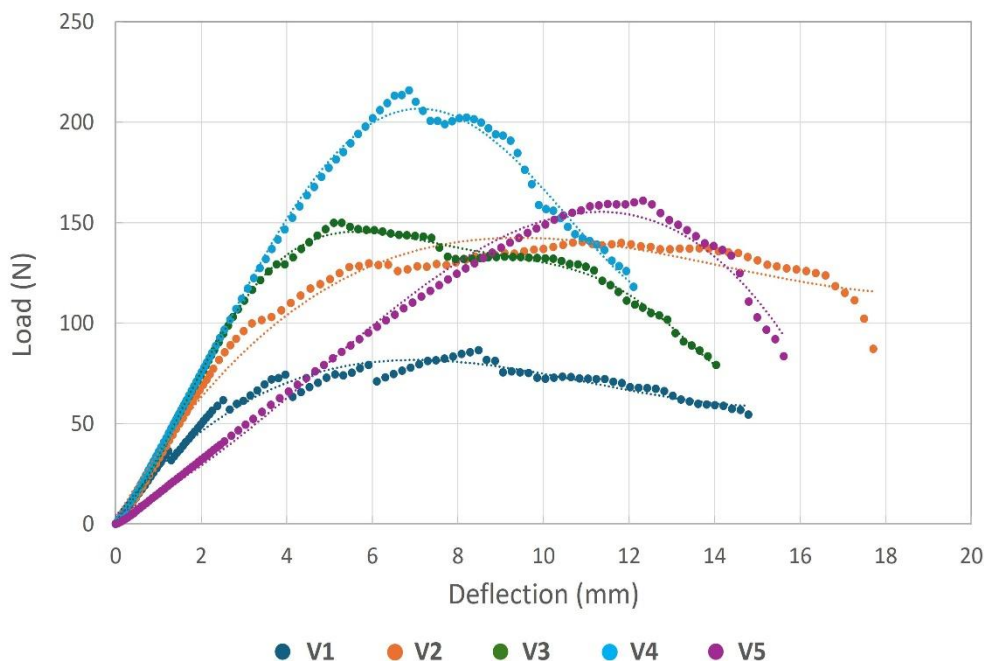


Figure 5 Representative load–deflection curves of V1–V5 hybrid laminates under three-point bending (ASTM D790, $L = 80\text{ mm}$).

3.2 Impact properties and core aluminum distribution

The Charpy impact results are presented in Figure 6. Stacking configuration had a clear effect on the energy-absorption capability of the laminates. The V3 configuration (GGGAAAAAGGGG) delivered the highest impact strength, reaching $123.0 \pm 4.79\text{ kJ/m}^2$, while V4 showed a slightly lower value of $103.1 \pm 3.36\text{ kJ/m}^2$. The remaining configurations, especially V1 and V5 exhibited markedly lower impact strengths, in the range of $93\text{--}95\text{ kJ/m}^2$. Configuration V2 also performed well ($119.0 \pm 12.71\text{ kJ/m}^2$), although it remained slightly below V3. A detailed summary of the flexural and impact properties is provided in Table 1. Impact strength (kJ/m^2) was obtained by normalizing the absorbed energy by the specimen cross-sectional area (ASTM D6110). The higher variability observed for V2 likely reflects specimen-to-specimen differences associated with hand lay-up (e.g., local resin-rich regions/voids and slight thickness variations), which can noticeably influence the Charpy response.

Among all lay-ups, V3 absorbed the most impact energy. Placing six aluminum layers in the laminate core created a metal-rich “buffer zone” that can dissipate impact energy through local plastic deformation and progressive damage within the metallic sub-laminate. This trend agrees with the general concept that metal-rich cores improve crashworthiness and impact tolerance in hybrid laminates, as discussed for perforated Al sheet/glass systems in the literature [8].

In absolute terms, the energy absorbed by V3 (4.71 J) is slightly lower than the $\sim 5.3\text{ J}$ reported for the G/Al/Al/G configuration in [8]. This difference is plausible given the differences in specimen architecture, perforation density, and reinforcement type, which affect both deformation modes and failure progression. In the present laminate, the perforated aluminum foil also supports stronger interfacial bonding. The $\sim 150\text{ }\mu\text{m}$ perforations allow resin to penetrate and solidify within the holes, forming mechanical interlocks that supplement adhesion. Very small holes tend to hinder resin penetration, whereas overly large holes can promote local resin pooling and leave regions with limited metal–resin contact. In Dhanaraj et al. [8], the perforations were much larger ($\approx 3\text{ mm}$) and more widely spaced, which reduces the number of anchoring sites per unit area. In this sense, the perforation scale used here provides a practical balance between wetting and interlocking, consistent with the cohesive lamination observed after testing.

Although V4 achieved the highest flexural strength, its impact strength was lower than V3, highlighting the expected trade-off between stiffness-driven and energy-absorption-driven designs. Many prior Al-based FML studies place aluminum sheets on the outer surfaces and vary FRP fiber orientation (e.g., $0^\circ/90^\circ$) to tune response, as reported for Al/glass fiber/epoxy laminates [10, 21]. Direct numerical comparison with the present Charpy results is not appropriate because those studies employed different metal thicknesses and impact methods (e.g., low-velocity impact or tensile-impact setups). For that reason, the discussion here emphasizes interlayer adhesion and cross-sectional damage features observed by microscopy to clarify how stacking configuration influences mechanical response. Overall, the findings reinforce that stacking design should be selected based on the dominant loading conditions expected in service.

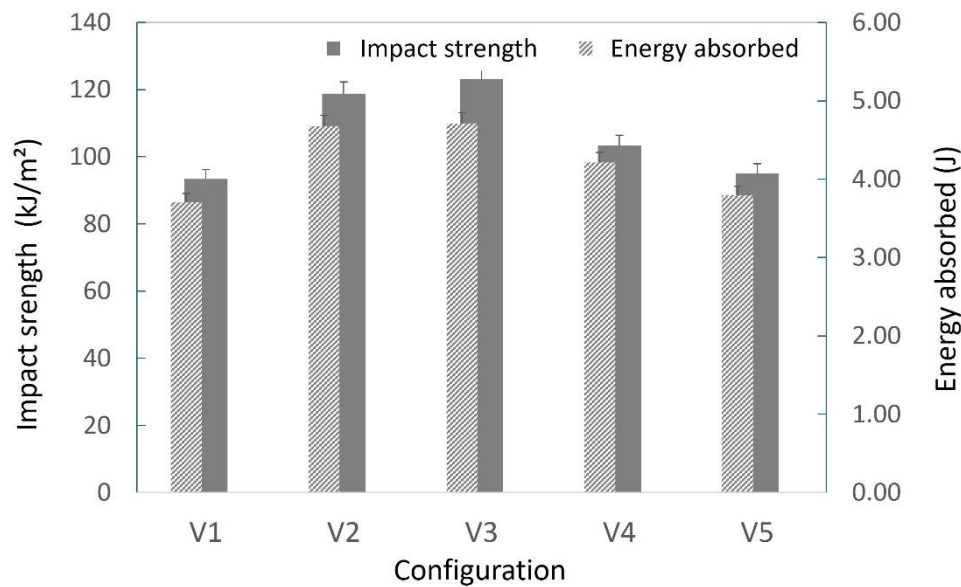


Figure 6 Impact strength of V1–V5 hybrid laminate configurations. Error bars indicate ± 1 standard deviation ($n = 5$).

Table 1 Flexural and impact properties of hybrid laminate configurations.

Configuration	Stacking sequence	Flexural strength (MPa)	Flexural modulus (GPa)	Flexural strain	Impact strength (kJ/m²)
V1	GAGAGAGAGAGAG	91.50 ± 4.37	7.82 ± 1.01	4.28 ± 0.72	93.00 ± 6.77
V2	GGAAAAAAGGGGG	138.67 ± 5.79	11.48 ± 0.60	5.13 ± 0.46	119.00 ± 12.71
V3	GGAAAAAAGGGGG	154.33 ± 12.04	13.13 ± 0.54	4.13 ± 0.47	123.00 ± 4.79
V4	GGGAAAGAAAGGG	229.74 ± 14.23	13.24 ± 0.82	4.52 ± 0.27	103.30 ± 3.36
V5	AAAGGGGGGGAAA	109.03 ± 7.29	5.64 ± 0.57	3.51 ± 0.57	95.00 ± 0.91

Values are mean ± 1 standard deviation ($n = 5$)

3.3 Optical microscopy observations

Optical images after testing (Figure 7 and 8) show that the damage pattern depends strongly on the lay-up. In V1 and V2, cracking is dominated by long interfacial separations. The most obvious feature is the extended delamination running along the Al and epoxy boundary, particularly near the outer regions. Once the interface opens, the crack tends to propagate along the same plane, producing a wide damage zone rather than a localized fracture band. This behavior is consistent with a lay-up that places dissimilar layers next to each other in a way that amplifies local mismatch in bending response, which can promote early interface opening and unstable delamination growth.

V3 and V4 appear different. The fractured region is more compact, and the interface separation looks less continuous. Instead of one long delamination front, the damage is more segmented and distributed through the thickness. The crack path does not remain “locked” to a single interface for long distances, suggesting more stable load transfer between layers. In addition, the symmetric nature of these lay-ups likely reduces localized deformation and suppresses early peeling-type opening at the interfaces, which helps explain their better mechanical response. This qualitative trend agrees with prior observations that symmetric stacking generally improves interfacial stability and delays failure initiation [22]. V5, meanwhile, shows a less regular crack pattern, with a broader fracture zone near the outer Al-rich region and a more uneven crack trajectory.

3.4 Scanning electron microscopy observations

SEM images (Figure 9) were examined to assess how the interfaces failed at the microscale. For V3 and V4, the Al/epoxy and glass/epoxy regions show limited evidence of large, clean interfacial gaps. The fracture surface is instead characterized by rougher features and matrix-dominated damage in many areas, with only small local separations. In practical terms, this appearance is more consistent with a “cohesive-dominant” fracture contribution than with a purely adhesive debonding along a smooth interface. These views also show limited visible voids, suggesting a reasonably good consolidation quality for the better-performing configurations.

It is important to note that the epoxy-filled features inside the perforations are not directly resolved in the available SEM views, which is not unusual for through-hole interlocking because the key features are three-dimensional and may not be captured by the selected fracture/section plane. Nevertheless, the combined evidence, namely (i) the reduced, less continuous interface opening seen in optical micrographs for V3 and V4 and (ii) the improved flexural/impact performance, supports the interpretation that perforation assists interfacial coupling. In V4, the placement of glass/epoxy plies toward the mid-plane provides a strong bending-resistant backbone, which is consistent with its superior flexural response. In V3, a metal-rich core is expected to contribute more strongly to impact energy dissipation, which aligns with its higher Charpy impact strength. A schematic mechanism (Figure 10) is therefore used to illustrate the expected load-transfer route and crack-path control enabled by the perforated foil, while keeping the microscopy discussion strictly tied to the observed features.

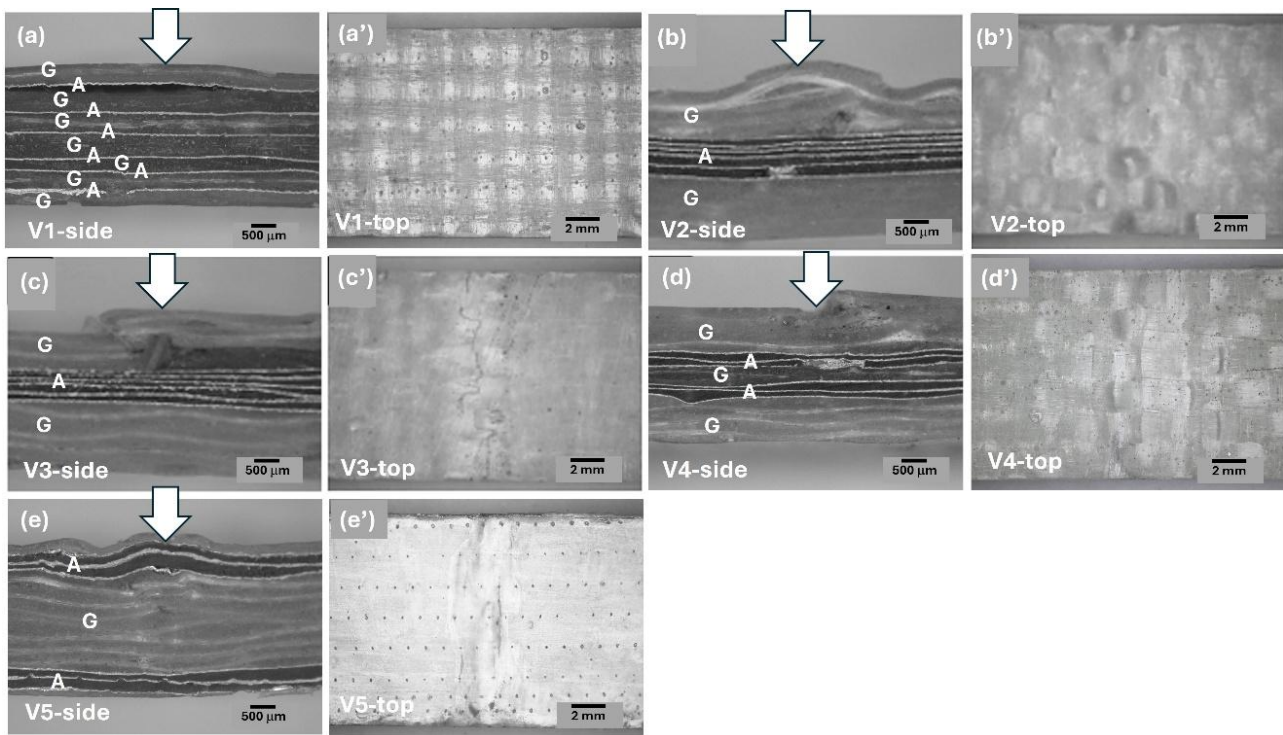


Figure 7 Optical micrographs of the composite specimens following the bending test. (a), (b), (c), and (d) are obtained from side views; (a'), (b'), (c'), and (d') are obtained from top views. Arrows show the direction of the bending load.

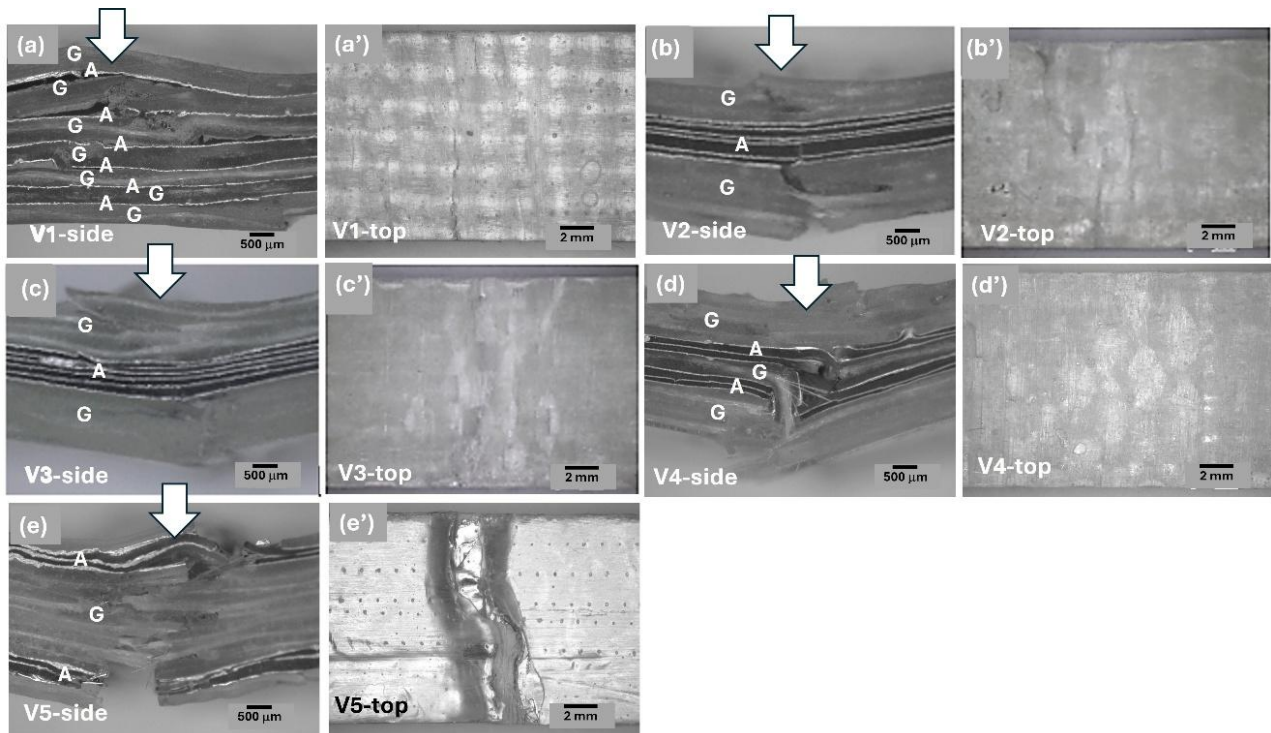


Figure 8 Optical micrographs of the composite specimens following the impact Charpy test. (a), (b), (c), and (d) are obtained from side views; (a'), (b'), (c'), and (d') are obtained from top views. Arrows show the hitting direction.

3.5 Microscopy and interfacial bonding

Optical microscopy and SEM were used to clarify how damage developed and how effectively load was transferred across the Al/epoxy interface. In V1 and V2, the Al/epoxy interface was dominated by long, continuous delamination traces. This morphology is consistent with poor interfacial stress transfer and explains the limited energy absorption observed in these lay-ups. In contrast, V3 and V4 exhibited more confined fracture regions and a larger proportion of cohesive damage within the matrix and fiber-rich zones. The crack path in these laminates was less “interface-driven,” indicating stronger interfacial cooperation between the foil and the G/epoxy layers.

SEM observations further show that the perforated aluminum foil (hole diameter $\sim 150\ \mu\text{m}$) enabled resin penetration through the openings and formed local mechanical interlocks, as schematically illustrated in Figure 10a–10b. The resin that solidified inside the perforations acted as anchoring features (“epoxy keys”), as depicted in Figure 10b, which reduced adhesive-type separation at the Al–epoxy interface and promoted cohesive cracking. Mechanistically, these anchors can deflect and branch cracks, forcing a more tortuous propagation route and activating additional energy-dissipation processes such as frictional sliding, fiber bridging/pull-out, and localized deformation of the remaining aluminum ligaments between perforations (Figure 10c). This interpretation is consistent with Rajamurugan et al. [16], who linked perforated aluminum foils to improved interlaminar performance in natural-fiber hybrid laminates, and with recent reports showing that surface texturing/perforation can enhance wetting and crack deflection in metal–polymer layered systems [23].

More generally, the effectiveness of a perforated interface is sensitive to geometric quality, such as edge integrity, taper, and the presence of damage or burrs around holes, which can govern crack initiation sites and subsequent propagation. The observed improvement in V3–V4 is therefore coherent with published geometry–damage correlations in fiber/epoxy and hybrid composite systems [24–27].

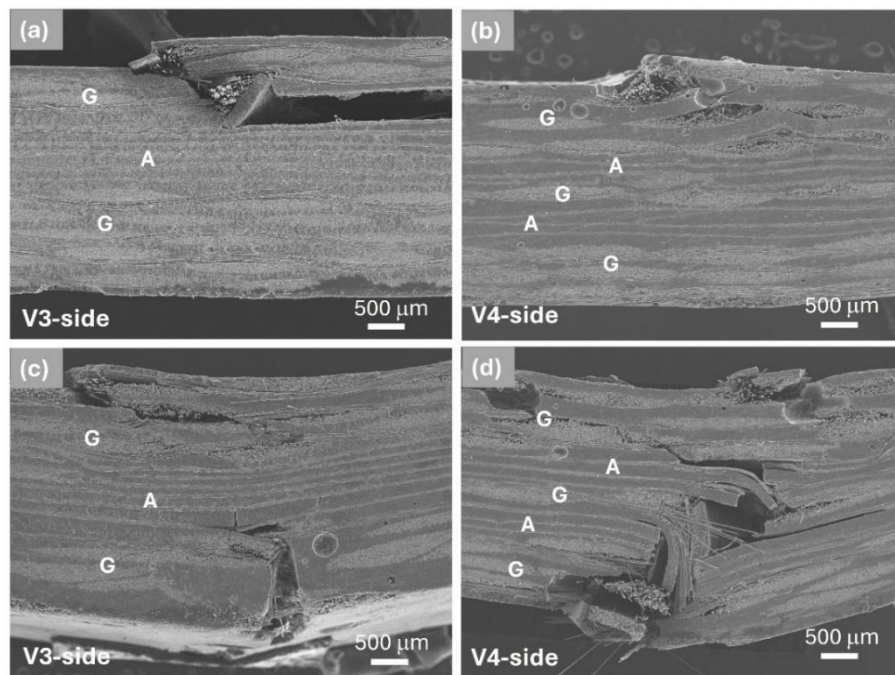


Figure 9 SEM images V3 and V4 composites following bending test (a and b) and Charpy impact test (c and d).

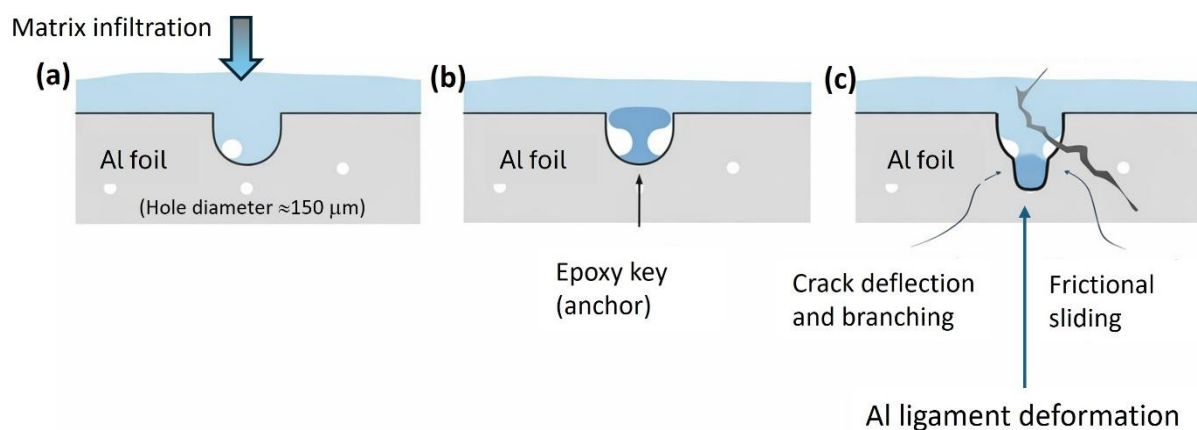


Figure 10 Schematic of perforation-induced interlocking at the Al/epoxy interface: (a) matrix infiltration through $\sim 150\ \mu\text{m}$ perforations during lay-up, (b) formation of epoxy “keys” after curing, and (c) crack deflection/branching and associated energy-dissipation mechanisms under loading.

3.6 Comparison of literature and novelty

Compared with perforated Al foil fiber metal laminates (FMLs) that employ an ultrathin foil ($\sim 24\ \mu\text{m}$) with circular perforations ($d = 5\ \text{mm}$) and relatively large hole pitches ($p \approx 20\text{--}60\ \text{mm}$, center-to-center) for which flexural strengths below 100 MPa are commonly reported [16], the symmetric V4 lay-up in the present study reached $\sim 230\ \text{MPa}$ in three-point bending (ASTM D790).

For impact loading, V3 absorbed 4.71 J in the unnotched Charpy test (flatwise). This value is slightly lower than the $\sim 5.3\ \text{J}$ reported by Dhanaraj et al. [8] for glass/Al perforated hybrids with larger perforations ($d \approx 3\ \text{mm}$; $p \approx 25\ \text{mm}$), yet it exceeds the $< 2\ \text{J}$ reported

in [16]. Because specimen geometry, thickness, support span, and impact configuration differ across studies, these numbers are best interpreted as indicative rather than directly comparable. Importantly, optical microscopy and SEM in the present work show that micrometer-scale perforations ($d \approx 150 \mu\text{m}$; $p \approx 1 \text{ mm}$) enable resin infiltration and dense micro-mechanical interlocking while preserving the continuity of the aluminum ligaments. This micro-anchoring mechanism helps limit adhesive-type interfacial separation and supports a favorable balance between bending stiffness and impact energy absorption.

The novelty of this work lies in explicitly linking perforation scale and stacking design and demonstrating their combined influence on flexural and impact responses. Prior studies have often examined foil thickness/perforation geometry or stacking sequence in isolation; the present results highlight that these variables interact, and that an optimized symmetric configuration can better exploit micro-perforation-enabled interlocking to enhance both flexural strength and impact performance. Overall, the findings support micrometer-perforated Al foil combined with an optimized symmetric stacking strategy as a practical route toward a more favorable stiffness–impact balance than previously reported perforated foil FML systems, while maintaining appropriate caution regarding cross-study comparisons.

Furthermore, by evaluating multiple stacking sequences at comparable constituent fractions, this study provides a practical design guideline, Al-rich cores for impact energy absorption (V3) and symmetric-staggered layouts for flexural strength (V4), that can be directly adopted for lightweight panels and skins requiring tailored stiffness–toughness trade-offs.

3.7 Implications and design considerations

The results show that stacking sequence provides a practical design lever to tailor laminate performance to specific service demands. For stiffness-driven structures, such as UAV skins and automotive panels, symmetric and staggered architectures (e.g., V4) are preferable because they better stabilize load transfer under bending and limit premature interface opening. In contrast, Al-rich core concepts (e.g., V3) favor impact-dominated applications, where energy absorption and damage containment are prioritized (e.g., protective housing, localized crash-relevant substructures, and protective gear). Overall, the findings highlight that micro-architectural design is central to the performance of thin-foil FMLs, and that micrometer-scale perforations can strengthen the interface through mechanical interlocking while preserving a useful stiffness–toughness balance.

Beyond aerospace and automotive components, the same stiffness–impact trade-off is relevant to external, non-implantable medical and assistive devices such as prosthetic sockets, orthopedic braces, and exoskeleton parts. In these products, low mass is important and accidental knocks are common during daily use. In the present laminates, the best flexural strength reached 229.74 MPa (V4) and the best impact strength reached 123.0 kJ/m² (V3). These levels exceed those typically reported for thermoplastics widely used in prosthetic/orthotic fabrication, particularly polypropylene and PETG, whose flexural strengths are commonly in the 30–70 MPa range and whose impact strengths are generally below $\sim 40 \text{ kJ/m}^2$ [28, 29].

Advanced composite systems (GFRP/CFRP) used in orthotics can achieve higher flexural strengths (often ~ 200 – 700 MPa), but their performance is sensitive to lay-up and processing and may raise cost and damage-tolerance concerns in certain designs [30]. In this context, thin-foil hybrid laminates may offer an intermediate pathway: higher stiffness than typical thermoplastics while maintaining a degree of damage stability through interface-controlled mechanisms.

For wearable and medical-adjacent applications, however, mechanical performance alone is not sufficient. Long-term use introduces additional requirements related to moisture/sweat exposure, cleaning/sterilization protocols, edge durability, and skin-contact safety. Practical deployment would therefore require appropriate surface finishing and protective coatings or barrier layers to improve corrosion resistance and reduce any risk associated with prolonged skin contact, along with validation of biocompatibility and hygiene compatibility under relevant standards. With these considerations addressed, the present results support optimized thin perforated Al foil/glass/epoxy laminates as a candidate material class for lightweight components requiring biomechanical support, energy absorption, and a tunable stiffness–toughness balance.

4. Conclusion

This study evaluated how stacking sequence governs the flexural and impact performance of hybrid laminates combining thin perforated aluminum foil, woven glass fiber, and epoxy. Optical microscopy and SEM indicate a robust Al/epoxy interface, attributed to resin penetration through $\sim 150 \mu\text{m}$ perforations that form mechanical interlocks and shifts fracture toward cohesive-dominant damage. Consistent with these observations, the mechanical results confirm that the laminate response can be tailored effectively through stacking design.

The results provide clear design guidance. Symmetric and balanced lay-ups are better suited to stiffness-driven applications (e.g., lightweight skins and panels), whereas Al-rich core architectures are more favorable when impact energy absorption is the primary requirement. Accordingly, perforation-assisted Al foil/GFRP hybrid laminates show promise for lightweight structural skins and panels (e.g., UAV structures and automotive interior/exterior panels), protective housings or covers exposed to accidental impact, and other thin-walled components where a controlled stiffness–toughness trade-off is needed at low weight.

This work has several limitations. The experimental program was restricted to one glass-fiber system, a single perforation size ($\sim 150 \mu\text{m}$), and quasi-static flexural testing with unnotched Charpy impact as the impact metric. Future work should expand to additional fiber systems and a broader range of perforation geometries (diameter, pitch, and pattern) and should assess durability under cyclic loading as well as dynamic/high strain-rate impact conditions. Further optimization of Al surface treatment and manufacturing scalability, together with practical quality-control metrics, will also be important to strengthen long-term reliability and industrial applicability.

Overall, these findings offer practical design guidance for thin perforated Al foil/glass fiber/epoxy laminates and support perforation-assisted interlocking as a viable route to address limitations frequently reported for conventional FML concepts by enabling a tunable balance between flexural performance and impact energy absorption.

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