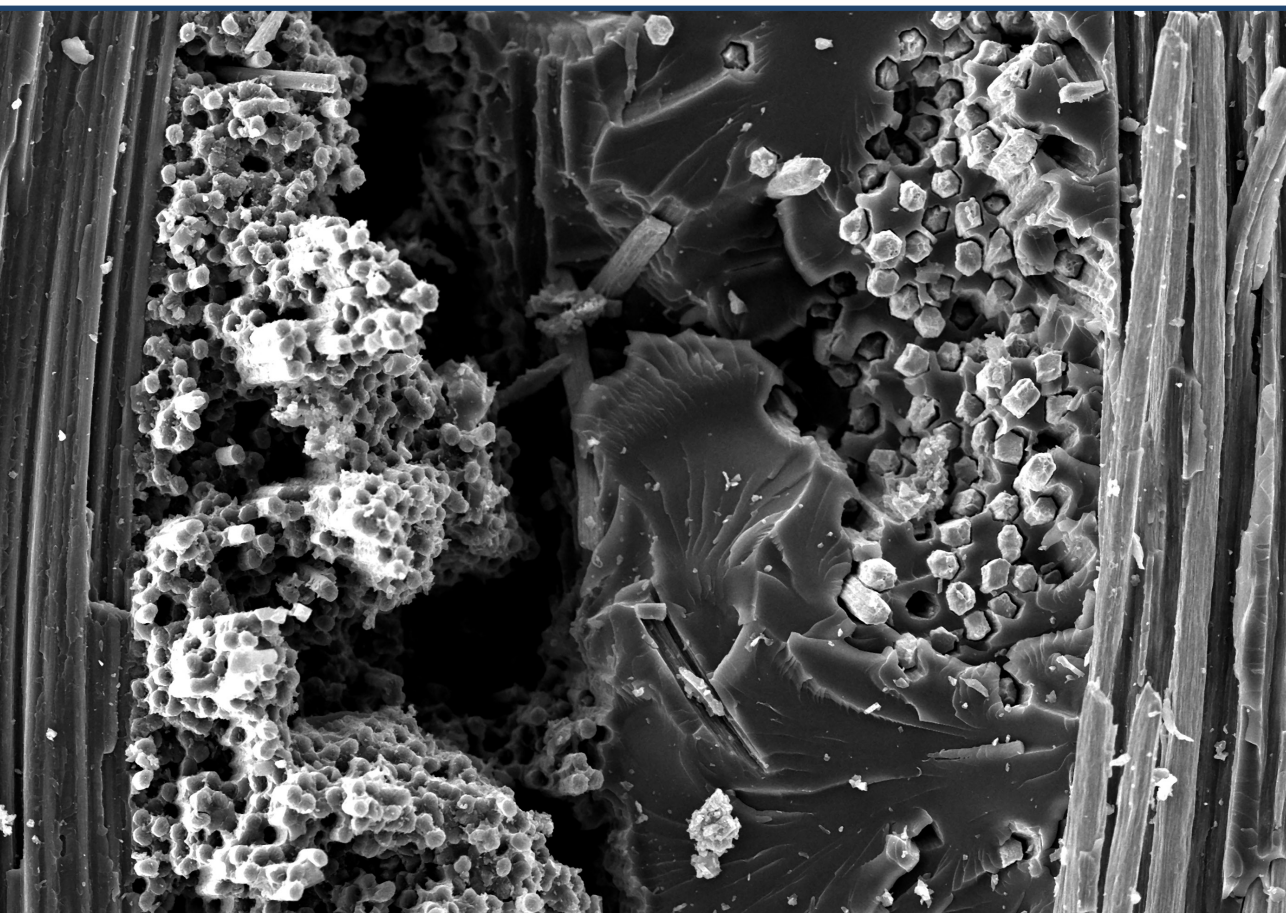


Anish Niranjana Kulkarni

**MICRO-DAMAGE EVOLUTION IN
NOVEL POLYMER COMPOSITE MATERIALS**

Doctoral Thesis



RIGA TECHNICAL UNIVERSITY

Faculty of Civil and Mechanical Engineering
Institute of High-Performance Materials and Structures

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Doctoral Student of the Study Programme “Mechanical Engineering and Mechanics (Applied Mechanics)”

**MICRO-DAMAGE EVOLUTION IN NOVEL
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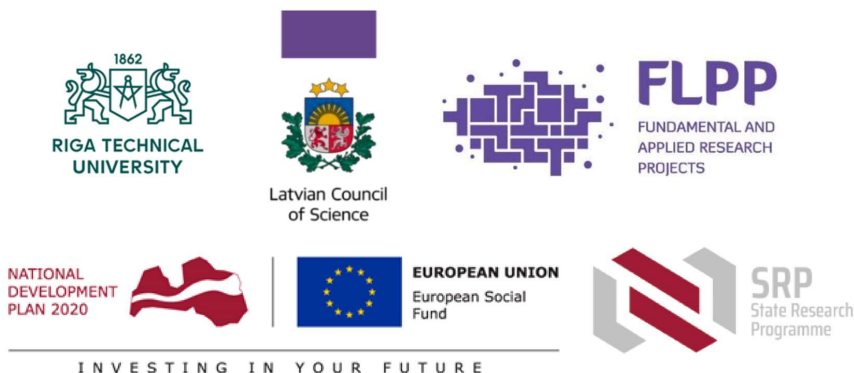
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PREFACE

I cannot begin without saying a huge thanks to Andrejs, Liva and Janis for being the best advisors, guardians, mentors and friends that I can ever get. Thanks for making me part of the family (or families, be it the Pupurs, IHPMS, BMF, LTU, ...), and for introducing me to potatoes and dill – they do go well together (sometimes :)

The thesis is dedicated to Aai and Baba, who have supported me tirelessly throughout this journey. This was a hard grind, and being homesick was but natural. Thanks to them for always being there, for believing in my dreams, for cheering me up when I was down, and for making sure that I sail through to the other side.

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Coming back to the most important person in this journey, thanks Andrejs for believing in me. Hopefully one day, I'll be as good a person as you are. Thanks for everything.

Anish Niranjana Kulkarni
Riga, 06.03.2026

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DECLARATION OF ACADEMIC INTEGRITY

I hereby declare that the Doctoral Thesis submitted for the review to Riga Technical University for the promotion to the scientific degree of Doctor of Science (Ph. D.) is my own. I confirm that this Doctoral Thesis had not been submitted to any other university for the promotion to a scientific degree.

Anish Niranjana Kulkarni (signature)

Date: 06.03.2026

The Doctoral Thesis has been written in English. It consists of a summary of 40 pages, including 20 figures, 2 tables, and 54 titles in the bibliography. The summary contains an introduction to the thesis topic, the main results of the thesis, and 6 appended papers. The papers are written in English, and their total length is 90 pages.

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ABBREVIATIONS

FEM	Finite element method
CF	Carbon fiber
UD	Unidirectional
ERR	Energy release rate
GF/EP	Glass-fiber/epoxy composite
CF/EP	Carbon-fiber/epoxy composite
ρ	Transverse crack density
ρ_n	Normalized crack density
σ_T	Transverse stress
σ_{Tn}	Normalized transverse stress
d	Half-thickness of 90-layer in [0,90]s cross-ply laminate (Section 2.1)
t_0	Thickness of 0-layer in [0,90]s cross-ply laminate
t_{90}	Thickness of 90-layer in [0,90]s cross-ply laminate
G	Energy release rate
G_T	Total energy release rate
a_N	Transverse crack length (Section 2.1)
E_L^0	Longitudinal elastic modulus of 0-layer in [0,90]s cross-ply laminate
G_{LT}^0	In-plane shear modulus of 0-layer in [0,90]s cross-ply laminate
E_T^{90}	Transverse elastic modulus of 90-layer in [0,90]s cross-ply laminate
l_d	Delamination length (Section 2.1)
l_c	Half-distance between two adjacent transverse cracks (Section 2.1)
E_{x0}	Initial longitudinal elastic modulus of composite (Section 2.2)
ROM	Rule of mixtures
SEM	Scanning electron microscopy
RVE	Representative volume element
3D	Three-dimensional
D_L	Longitudinal damage parameter of composite RVE (Section 2.2)
X_T	Longitudinal tensile strength of composite RVE (Section 2.2)
Y_T	Transverse tensile strength of composite RVE (Section 2.2)
S_L	Longitudinal shear strength of composite RVE (Section 2.2)
DSC	Differential scanning calorimetry
T_g	Glass transition temperature
E	Elastic modulus of composite (Section 2.3)

E_n	Normalized elastic modulus of composite (Section 2.3)
σ_{max}	Ultimate tensile strength of composite (Section 2.3)
ε_{max}	Strain-to-failure of composite (Section 2.3)

GENERAL OVERVIEW OF THE THESIS

This is a summary of six papers to be defended for the doctoral thesis titled ‘Micro-damage evolution in novel polymer composite materials’ in the doctoral study programme ‘Mechanical Engineering and Mechanics (Applied Mechanics)’. The main objective of the doctoral thesis is to study the micro-damage evolution in polymer composite materials with i) high crack density, ii) novel hybrid textile meso-structure, and iii) novel manufacturing method of electromagnetic induction heating.

Paper I and **Paper II** address the micro-damage in the form of transverse (intralaminar) cracks and interlaminar delamination cracks in [0,90]_s cross-ply laminates. Under uniaxial quasi-static tensile loading, the propagation of transverse cracks and delaminations has been numerically studied using FEM by analyzing changes in the strain energy release rate available for crack growth. The novelty of this study lies in i) analyzing growth of a new transverse crack growing in the regions of high crack density with prominent transverse stress gradients along the 90-ply thickness direction, and ii) conducting a detailed analysis of parameters (including geometrical parameters, material constants and transverse crack density) responsible for the symmetrical or non-symmetrical growth of local delaminations initiated at the tips of transverse cracks.

Papers III, IV, and V address the initiation and propagation of micro-damage in hybrid carbon/steel fiber textile composites. The hybrid composites are manufactured with plain woven or plain weft knitted fabric reinforcement and an epoxy matrix. Under uniaxial quasi-static tensile loading along the warp yarn direction, these composite laminates exhibit micro-damage in the form of weft yarn/matrix-interface delaminations and fiber breaks in the warp yarns. The numerical study of the mechanical behavior of hybrid carbon/steel fiber textile composites using FEM shows that the fiber breaks in the carbon fiber warp yarns strongly affect the composite’s macro-mechanical behavior. The novelty in this study lies in the manufacturing of hybrid carbon/steel fiber textile composites and the application of the elastoplastic damage model developed by C  zar et al. [1,2] to simulate the mechanical behavior of the hybrid textile composite materials.

Paper VI addresses micro-damage evolution in carbon fiber (CF)- reinforced thermoset composites manufactured using an electromagnetic induction heating system with a moving-coil design. Induction heating of a CF-reinforced thermoset composite with a stationary coil results in the formation of a cold spot. Although a moving-coil design mitigates the formation of a single cold spot, it creates a time-varying temperature gradient in the composite, leading to the entrapment of air bubbles formed during curing. Thus, large void content can be observed in the induction-cured composites that affects the initiation of micro-damage under mechanical loading. The novelty in this study lies in the manufacturing of induction-cured CF-reinforced thermoset composites and the comparison of the mechanical behavior of induction-cured and reference oven-cured composites under uniaxial quasi-static tensile loading.

Objective and tasks

The main objective of the present thesis is to study the evolution of micro-damage mechanisms such as transverse cracks and delaminations in composite materials with

- i. a damage state not studied in the literature, i.e., high crack density
- ii. a novel meso-structure, i.e., hybrid carbon/steel fiber textile meso-structure
- iii. a novel electromagnetic induction heating manufacturing process.

Scientific novelty and main results

The scientific novelty of the present thesis lies in the study of micro-damage mechanisms in: i) polymer composites with high crack density; ii) hybrid carbon/steel fiber composites with textile meso-structure; iii) polymer composites manufactured by means of electromagnetic induction heating. The main results obtained in the thesis regarding the observed micro-damage mechanisms are as follows –

1. In the high crack density regions in 90-layers in cross-ply laminates under uniaxial tensile loading, high gradients in transverse stress are generated along the thickness direction. Such gradients in stress strongly affect the strain energy release rate available for the growth of a new transverse crack.
2. Due to high stress concentrations at the transverse crack tips, local delaminations are generated in cross-ply laminates under uniaxial tensile loading. The delaminations grow symmetrically or non-symmetrically with respect to the existing transverse crack and the laminate midplane, forming I-shaped, T-shaped, C-shaped, or S-shaped cracks. The growth of these delaminations depends on the transverse crack density, 0/90-layer thickness ratio, and 0/90-layer stiffness ratios. A non-symmetrical I-shaped delamination tends to grow into a symmetrical I-shaped delamination before growing further.
3. In hybrid carbon/steel fiber woven composites under uniaxial quasi-static tensile loading along the warp yarn direction, micro-damage initiates in the form of weft yarn/matrix-interface delaminations. Upon further loading, fiber breaks are observed in the warp yarns that strongly affect the composite's macro-mechanical behavior. A cluster of fiber breaks results in warp yarn failure, and is accompanied by the ultimate failure of the hybrid composite.
4. During the manufacturing of carbon fiber reinforced epoxy composites using electromagnetic induction heating, a moving-coil design mitigates the problem of the formation of a cold spot. But it allows the entrapment of air bubbles generated during curing due to a time-varying temperature gradient in the composite material. This results in a large void content in the induction-cured composite specimens. Under uniaxial tensile loading, the voids help in the micro-damage initiation in induction-cured composites at a lower applied strain level than the reference oven-cured composites.

Structure of the Thesis

The summary of the doctoral thesis contains the introduction to the thesis topic, and the main results of the thesis divided into three sub-sections – i) Micro-damage evolution in cross-ply laminates with high crack density, ii) Micro-damage evolution in hybrid carbon/steel fiber textile composites, and iii) Micro-damage evolution in induction-cured thermoset composites. It also contains six appended papers which are listed in the sub-section ‘Publications and approbation of the Thesis’.

The Doctoral Thesis has been written in English. It consists of a summary of 40 pages, including 20 figures, 2 tables, and 54 titles in the bibliography. The summary contains an introduction to the thesis topic, the main results of the thesis, and 6 appended papers. The papers are written in English, and their total length is 90 pages.

Publications and approbation of the Thesis

Paper I

Kulkarni AN, Pupurs A, Varna J. Analysis of initiation and growth of new transverse cracks in high crack density regions in cross-ply laminates. *J Compos Mater* 2025;59:539–50. <https://doi.org/10.1177/00219983241295811>.

Paper II

Kulkarni AN, Pupurs A, Varna J. Parameters affecting growth of local delaminations at transverse crack tips in [0m,90n]s cross-ply laminates. *Eng Fract Mech* 2025;328:111549. <https://doi.org/10.1016/J.ENGFRACMECH.2025.111549>.

Paper III

Kulkarni AN, Pupurs A, Šņepsts A, Irbe M. MICRO-DAMAGE EVOLUTION IN HYBRID CARBON/STEEL KNITTED AND WOVEN FABRIC COMPOSITES. In: Falzon B, McCarthy C, editors. *Proceedings of the 2023 International Conference on Composite Materials*, Belfast: Queen’s University Belfast; 2023.

Paper IV

Kulkarni AN, Kononova O, Jevstignejevs V, Pupurs A. FE MODELING OF DAMAGE DEVELOPMENT IN HYBRID CARBON/STEEL FIBER TEXTILE COMPOSITES USING FRACTURE MECHANICS AND COHESIVE ZONE APPROACH. In: Binetruy C, Jacquemin F, editors. *Proceedings of the 21st European Conference on Composite Materials*, Nantes: The European Society for Composite Materials (ESCM) and the Ecole Centrale de Nantes; 2024, p. 1362–68.

Paper V

Kulkarni AN, Pupurs A, Kononova O, Jevstignejevs V, Maimi P, Subramani A. Parametric study of micro-damage evolution in woven hybrid carbon/stainless steel fiber composites. Under Preparation.

Paper VI

Pupurs A, Lācis V, Šņepsts A, **Kulkarni AN**, Irbe M. ANALYSIS OF MICRO-DAMAGE EVOLUTION IN INDUCTION-CURED THERMOSET COMPOSITES. In: Binetruy C, Jacquemin F, editors. Proceedings of the 21st European Conference on Composite Materials, Nantes: The European Society for Composite Materials (ESCM) and the Ecole Centrale de Nantes; 2024, p. 1529–34.

Results of the Thesis were presented at the following international conferences:

1. 20th European Conference on Composite Materials (ECCM20), June 26 – 30, 2022, Lausanne, Switzerland
2. 63rd International Scientific Conference of Riga Technical University, Section ‘Engineering, Mechanics and Mechanical Engineering’, October 13, 2022, Riga, Latvia
3. 81st International Scientific Conference of the University of Latvia, Section ‘Advanced Composite and Applications’, February 16, 2023, Riga, Latvia
4. 23rd International Conference on Composite Materials (ICCM23), July 31 – August 4, 2023, Belfast, UK
5. 64th International Scientific Conference of Riga Technical University, Section ‘Engineering, Mechanics and Mechanical Engineering’, October 11, 2023, Riga, Latvia
6. 82nd International Scientific Conference of the University of Latvia, Section ‘Advanced Composite and Applications’, February 13, 2024, Riga, Latvia
7. 19th European Mechanics of Materials Conference (EMMC19), May 29 – 31, 2024, Madrid, Spain
8. 21st European Conference on Composite Materials (ECCM21), July 2 – 5, 2024, Nantes, France
9. TexComp-15 Conference, September 11 – 13, 2024, Leuven, Belgium
10. 65th International Scientific Conference of Riga Technical University, Section ‘Civil and Mechanical Engineering’, October 17, 2024, Riga, Latvia
11. 83rd International Scientific Conference of the University of Latvia, Section ‘Advanced Composite and Applications’, February 11, 2025, Riga, Latvia
12. 12th International Conference on Composite Testing and Model Identification (CompTest 2025), May 21 – 23, 2025, Riga, Latvia
13. 24th International Conference on Composite Materials (ICCM24), August 4 – 8, 2025, Baltimore, USA
14. 66th International Scientific Conference of Riga Technical University, Section ‘Civil and Mechanical Engineering’, October 16, 2025, Riga, Latvia

1. INTRODUCTION

Polymer composite materials consist of a thermosetting or thermoplastic polymer matrix combined with reinforcements such as short discontinuous or long continuous fibers. The fibers can be randomly dispersed in the matrix or aligned in a specific meso-structure. In a fiber-reinforced composite material, fibers act as the primary load-bearing component. Thus, the alignment of fibers with respect to the loading direction is critical in governing the mechanical performance of a composite material. In addition to their lower density compared to metallic materials, polymer composites are well known for providing exceptional stiffness and strength, making them highly suitable for lightweight applications. Today, the use of polymer composite materials has widened in various engineering sectors such as transport (aircraft fuselages and wings, aero-turbine blades, helicopter wings, high-speed trains, military vehicles), energy (wind turbines, hydrogen storage tanks, solar power systems), and infrastructure (pipes, bridges, roofs), refer to Fig. 1.1.

In the case of composite materials with long fiber reinforcement, the fibers can be i) randomly distributed, ii) aligned parallel to each other, i.e., unidirectional (UD) reinforcement, or iii) in the form of a textile fabric, with warp and weft yarns. Note that in a woven fabric, the warp yarns are placed lengthwise, whereas the weft or fill yarns are later run through the warp yarns widthwise (refer to Fig. 1.2). According to the service requirements, the individual composite plies can be aligned in different orientations and stacked in a laminate with a specific layup. For example, in **Papers I, II and VI**, a composite laminate with a cross-ply layup is considered, where UD plies are either stacked along the loading direction (0° ply, or 0-ply) or perpendicular to it (90° ply, or 90-ply). Upon uniaxial tensile loading, damage initiates in the 90-plies in cross-ply laminates in the form of transversal debonds at the fiber/matrix-interface [3].

In the composite plies with fibers oriented perpendicular to the loading direction, the damage initiation takes place due to stress concentration at the fiber/matrix interface [4]. The stress concentration arises from the difference in transverse stiffness between the fiber and the matrix. A localized fiber-matrix debond is created, and upon increase in loading, the adjacent debonds coalesce to form a matrix crack (refer to Fig. 1.3). With a further increase in loading, in the case of a cross-ply laminate, the matrix crack in the 90-layer grows along the ply thickness direction as a ‘transverse crack’ until it reaches the stiffer 0-layer [5,6]. In the commonly used analytical models studying the stress state in damaged 90-plies, it is assumed that a transverse crack grows instantaneously along the fiber direction (i.e., along the laminate width direction) until a free surface is reached, creating a tunnelling crack [7–11]. This assumption may not be valid for laminates with thin plies, where transverse cracking gets partially or totally suppressed [5].

In textile composites loaded along the warp yarn direction, micro-damage initiates as a fiber/matrix-interface debond at the weft yarn/matrix boundary. The coalescence of debonds takes place either to form a transverse crack inside the weft yarn or a delamination at the weft yarn/matrix-interface [12]. Fig. 1.4 shows examples of i) a transverse crack in $[0,90]_s$ cross-ply laminate, ii) a weft yarn/matrix-delamination, and iii) a transverse crack inside a weft yarn in a

textile composite laminate with plain woven meso-structure. The damage evolution in the textile composites takes place in the form of coalescence of yarn/matrix-interface delaminations and transverse cracks as shown in Fig. 1.5 [13,14].

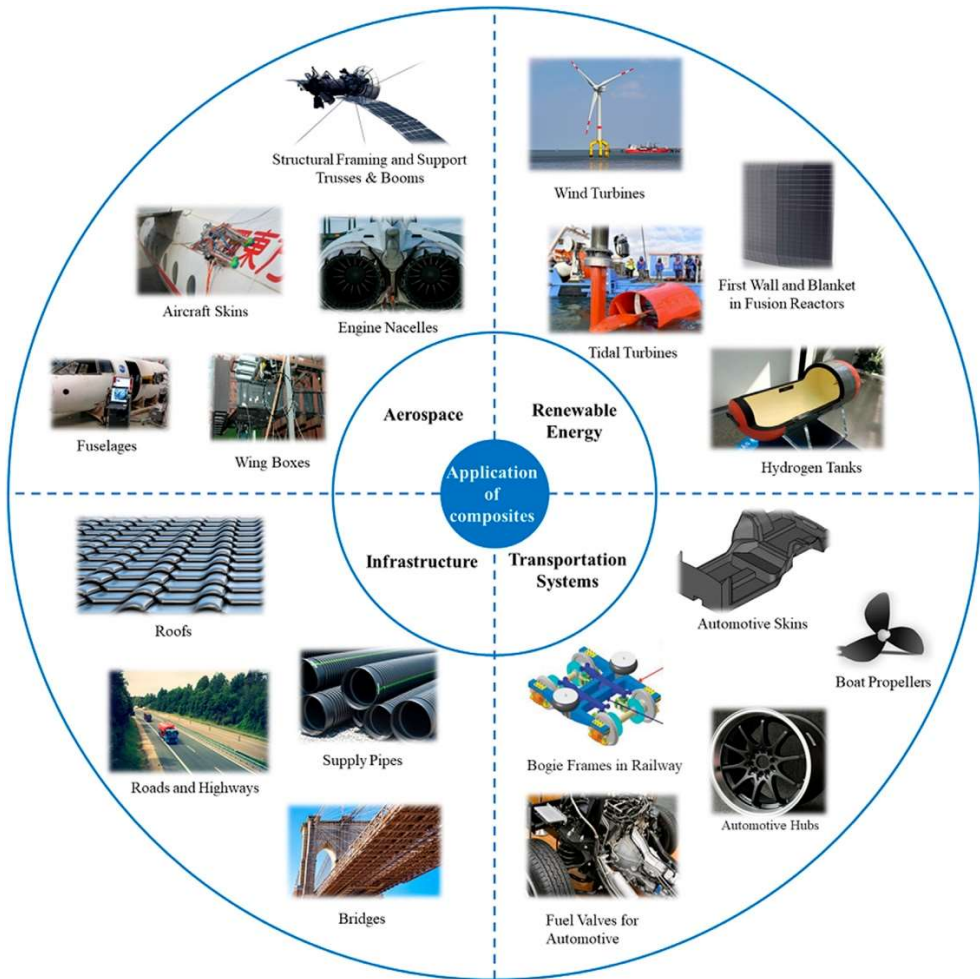


Figure 1.1. Applications of polymeric composite materials. Reproduced with permission from the publisher and the corresponding author of [15].

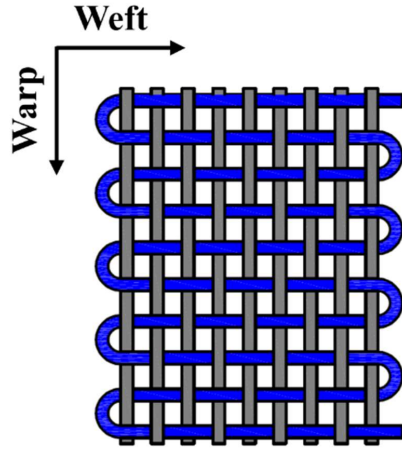


Figure 1.2. A plain woven textile fabric with warp and weft yarns.

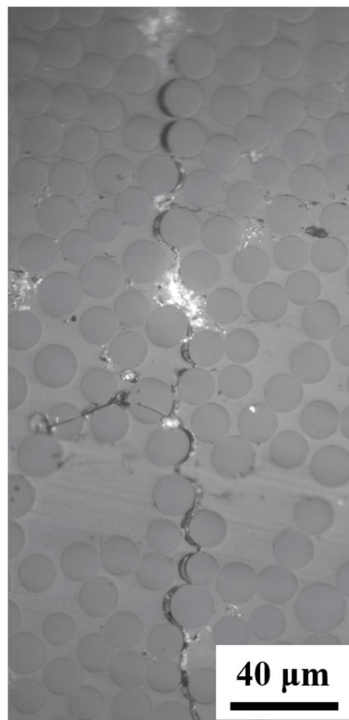
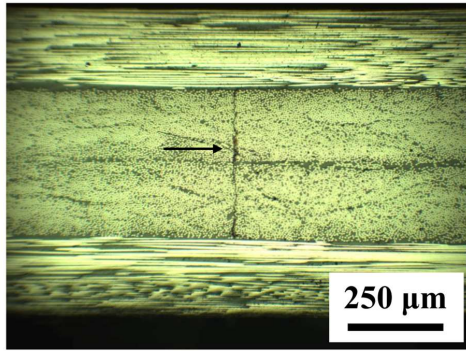
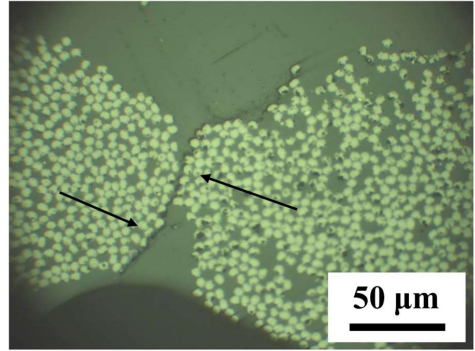


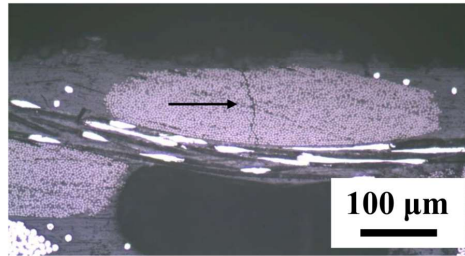
Figure 1.3. Coalescence of transversal fiber-matrix debonds to form a matrix crack in glass-fiber/epoxy composite laminate [16].



(a)



(b)



(c)

Figure 1.4. Examples of matrix cracks in carbon-fiber/epoxy composite laminates under uniaxial tensile loading, (a) transverse crack in $[0,90]_s$ cross-ply laminate, (b) weft-yarn/matrix delamination, and (c) transverse crack inside a weft yarn in plane woven composite laminate.

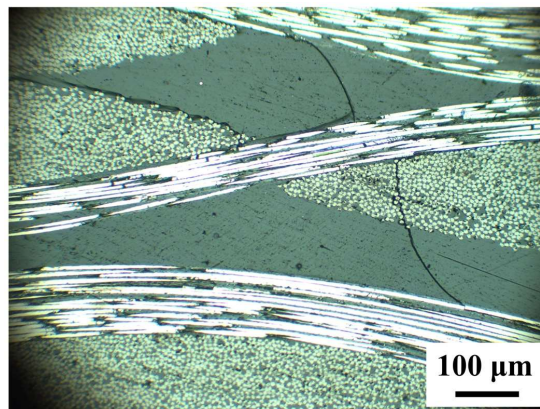


Figure 1.5. Coalescence of yarn/matrix-interface delaminations and transverse cracks in carbon-fiber/epoxy plane woven composite laminates under uniaxial tensile loading.

In a cross-ply laminate with a transverse crack in the 90-layer, the damaged 90-layer does not carry any load at the location of the transverse crack. At some distance away from the transverse crack, load is transferred back to the 90-layer with the help of out-of-plane shear [11]. Upon further loading, multiple transverse cracking takes place in the 90-layer, increasing the transverse crack density [17]. As the crack density exceeds a threshold, adjacent transverse cracks begin to interact, creating stress perturbations. This reduces the available strain energy release rate (ERR) for the growth of a new transverse crack [18]. Eventually, the transverse crack density reaches saturation [17,19].

At the transverse crack tip, the generation of very high stresses along the transverse and out-of-plane shear directions causes the formation of a new damage mode, i.e., a local delamination at the tip of an existing transverse crack [6,20–22]. The initiated delamination propagates along the 0/90-layer interface, and it also reduces the available ERR for the growth of a new transverse crack in its vicinity [18]. The growth of a local delamination depends on various parameters, such as 0/90-layer thickness ratio, 0/90-layer stiffness ratios, and transverse crack density [23]. Based on the local stress state, local delaminations can grow into various shapes, such as I-shaped, T-shaped, S-shaped, and C-shaped cracks. Fig. 1.6 shows examples of I-shaped and S-shaped delamination cracks in [0,90]_s cross-ply laminates.

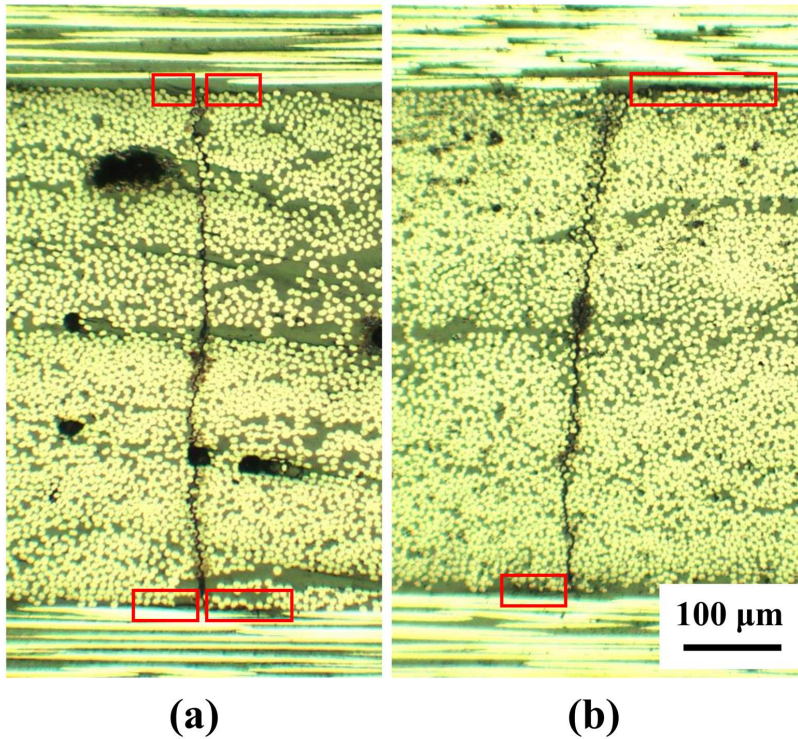


Figure 1.6. Examples of (a) I-shaped and (b) S-shaped delamination cracks (shown using red rectangles) in carbon-fiber/epoxy [0,90]_s cross-ply laminates under uniaxial tensile loading.

The ultimate failure of a thermoset composite laminate often occurs in the form of fiber breakage in the loading direction catastrophically in a brittle manner [24]. But some composites can exhibit considerable plasticity before final failure, depending on the specific damage evolution behavior. Epoxy-based composites with ductile steel fiber UD reinforcement exhibit significant plastic behavior under uniaxial tensile loading. They can reach failure strains up to 7%, which can be further increased using a matrix with higher ductility [25,26]. UD glass-fiber/epoxy composites hybridized with thin carbon-fiber/epoxy plies can also exhibit ‘pseudo-ductility’ through distributed thin-ply cracking and controlled delamination propagation [27].

In addition to improving ductility and toughness, fiber hybridization can enhance the sustainability of thermoset composites. There are numerous examples of successful industrial applications where glass-fiber/epoxy or carbon-fiber/epoxy composite parts have been hybridized with natural fibers such as flax fibers without a significant loss of the necessary mechanical performance [28]. Recently, there has been an increased interest in some more ‘green’ initiatives in the field of thermoset composites, such as i) improving the stiffness and strength of recycled carbon fibers by optimizing the pyrolysis treatment [29], and ii) reducing the energy consumption in the process of curing. This could be achieved, for example, by using the recently emerged volumetric curing processes such as induction heating, microwave heating, or direct electric heating over oven curing [30]. It has been shown that the induction-cured thermoset composites exhibit mechanical properties such as elastic modulus, ultimate tensile strength, and strain-to-failure on par with the oven-cured composites, although the amount of porosity in the induction-cured composites was much higher than the oven-cured composites [31].

2. MAIN RESULTS OF THE THESIS

2.1. Micro-damage evolution in cross-ply laminates with high crack density

The 90-layers in cross-ply laminates undergo transverse cracking under uniaxial tensile loading. The transverse cracks generate stress perturbations around them, which can be characterized using analytical models available in the literature [7–10,17]. Regarding the distribution of transverse stress in the damaged 90-layer, these models either assume a uniform stress state along the ply thickness direction, or they use simple shape functions to model the stress distribution.

In regions with high crack density, the stress perturbations from transverse cracks begin to interact. Thus, a non-uniform stress state is generated along the ply thickness direction, which can only be characterized using semi-analytical models [32–35] or using FEM. The FEM numerical analysis in **Paper I** shows significant gradients in the transverse stress along the 90-ply thickness direction for glass-fiber epoxy (GF/EP) and carbon-fiber/epoxy (CF/EP) cross-ply laminates with different values of normalized crack densities (ρ_n), see Fig. 2.1. Note that the transverse crack density, ρ , is normalized with respect to the 90-layer thickness, $2 \cdot d$, such that $\rho_n = \rho \cdot 2 \cdot d$. It can be seen that the stress gradients increase with transverse crack density. These gradients in transverse stress strongly affect the available ERR for the growth of a new transverse crack, see Fig. 2.2. At very high crack density ($\rho_n = 2.0$), compressive stresses are generated near the laminate midplane (see Fig. 2.1), which reduce to zero the ERR available for the growth of a new transverse crack (see Fig. 2.2), completely stopping transverse crack growth.

The ERR for the growth of a new transverse crack is also reduced due to the local delaminations initiated at the tips of the existing transverse cracks. Such local delaminations are caused by high normal and shear stresses at the transverse crack tips. The growth of local delaminations can be studied using compliance-based analytical models, which assume self-similar growth characteristics independent of the delamination length [20,36]. But FEM numerical analysis in **Paper II** shows that the ERR available for the growth of local delaminations in cross-ply laminates strongly depends on delamination length and transverse crack density, as shown in Fig. 2.3.

A detailed parametric analysis given in **Paper II** concludes that, along with delamination length and transverse crack density, the ERR available for the growth of a local delamination in [0,90]s cross-ply laminates depends on the layer thickness ratio, t_0/t_{90} , and some specific ratios of elastic constants - E_L^0/E_T^{90} and G_{LT}^0/E_T^{90} . For small delaminations, an increase in these ratios reduces the ERR available for delamination growth, whereas for long delaminations, the effect is opposite.

Depending on the local stress state and local variations in fracture toughness, local delaminations can grow into various shapes such as I-shape, T-shape, S-shape, and C-shape. The ERR available for the delamination growth also depends on the delamination shape, as

shown in Fig. 2.4. The ERR analysis in **Paper II** also shows that non-symmetrical I-shaped delaminations tend to grow into symmetrical delaminations before growing further.

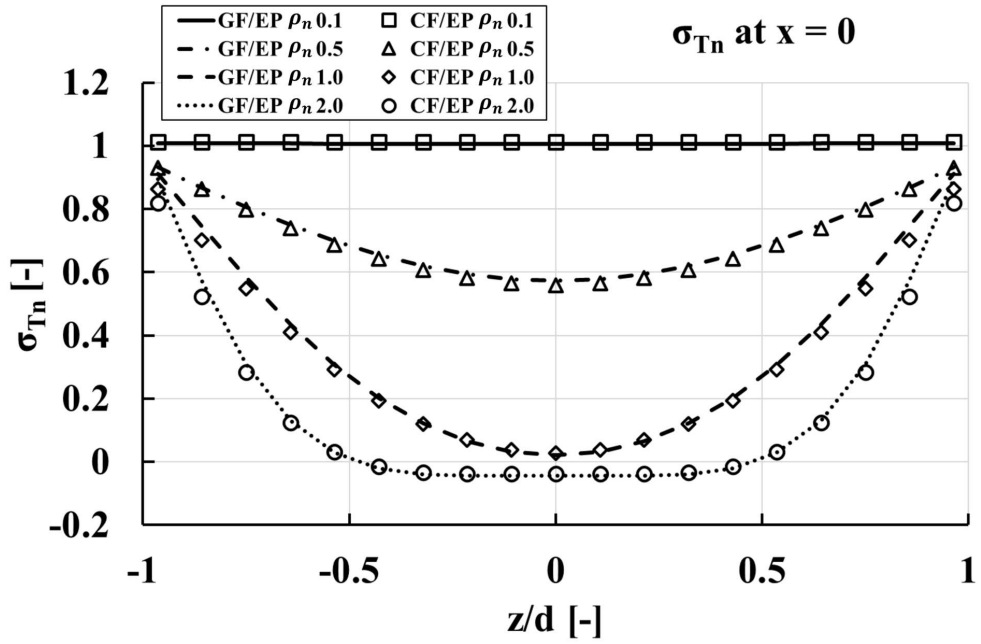


Figure 2.1. Distribution of normalized transverse stress (σ_{Tn}) in the damaged 90-layer in GF/EP and CF/EP [0,90]_s cross-ply laminates along the ply thickness (z-axis) direction. Stress values are normalized with respect to the transverse stress in the undamaged 90-layer. d is half-thickness of the 90-layer. Transverse crack density (ρ) is normalized with respect to $2 \cdot d$. Reprinted from **Paper I**.

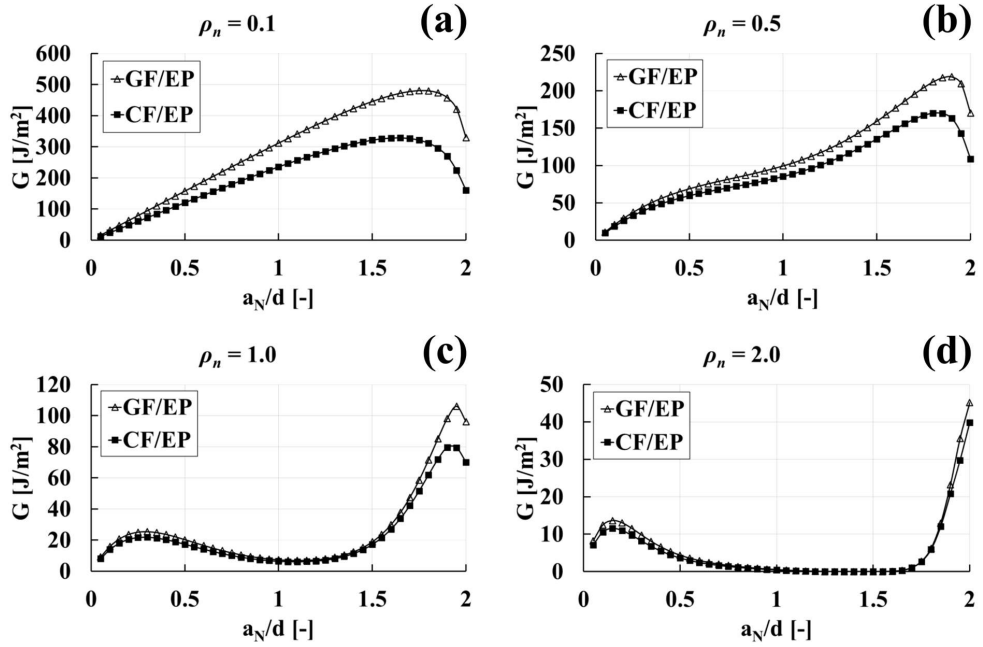


Figure 2.2. Trends in available energy release rate for the growth of a new transverse crack in the 90-layer in GF/EP and CF/EP [0,90]s cross-ply laminates with normalized crack density (a) 0.1, (b) 0.5, (c) 1.0, and (d) 2.0. a_N is the transverse crack length. Reprinted from **Paper I**.

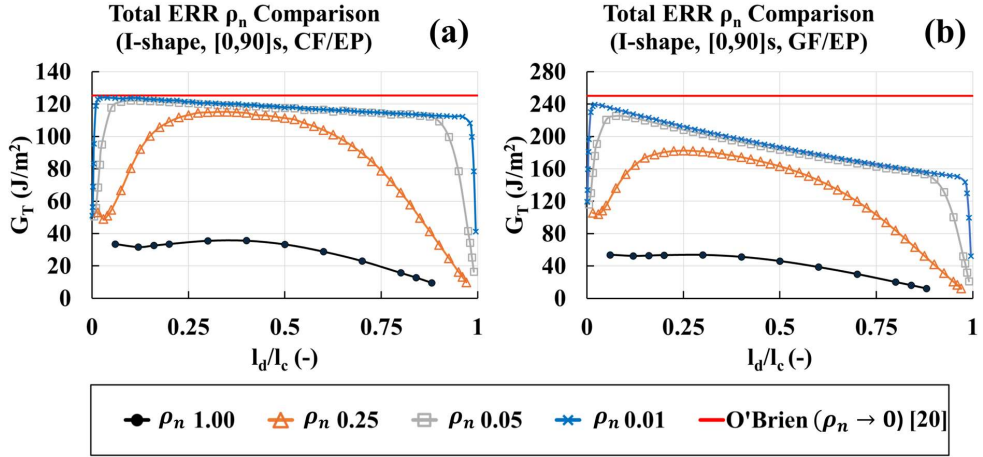


Figure 2.3. Trends in total available energy release rate for the growth of a local delamination in (a) CF/EP and (b) GF/EP [0,90]s cross-ply laminates for normalized crack densities 1.00, 0.25, 0.05, and 0.01 calculated using FEM. The FEM ERR trends are compared with the constant ERR value given by the compliance-based analytical model by O'Brien [20].

Reprinted from **Paper II**.

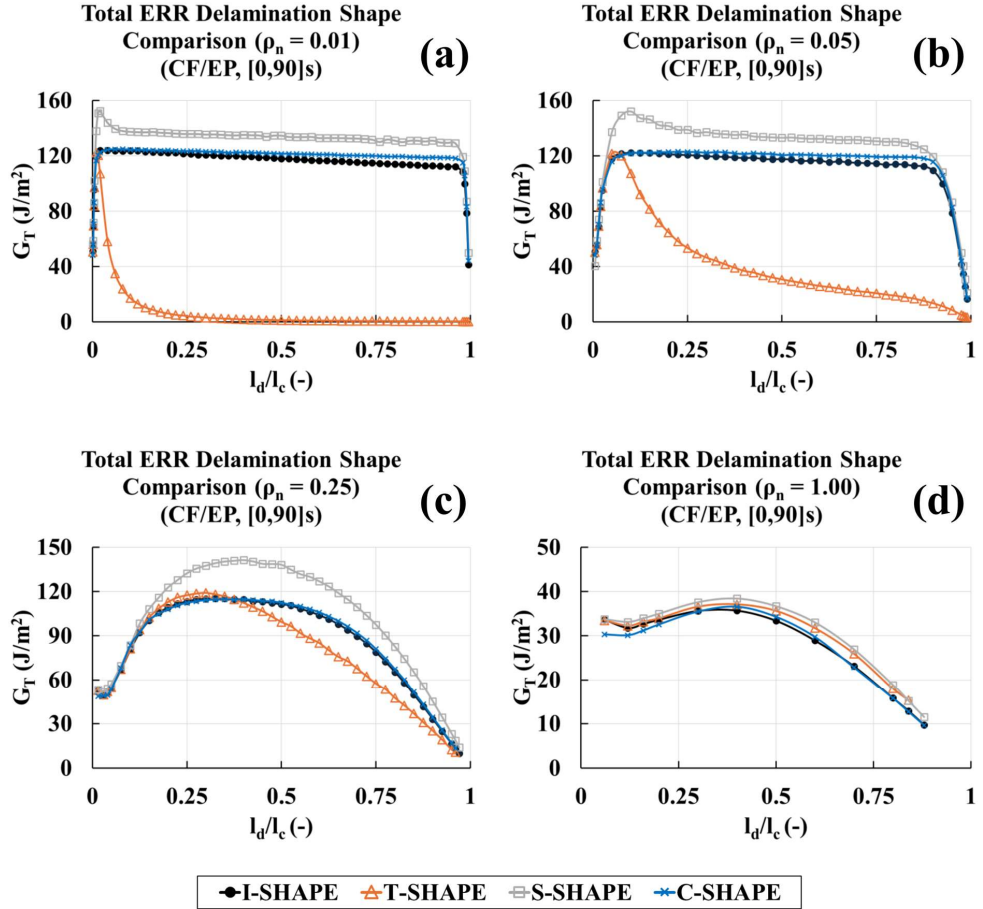


Figure 2.4. Trends in total available energy release rate for the growth of a local delamination with I-shape, T-shape, S-shape, and C-shape for normalized crack densities (a) 0.01, (b) 0.05, (c) 0.25, and (d) 1.00. Reprinted from **Paper II**.

2.2. Micro-damage evolution in hybrid carbon/steel fiber textile composites

The technique of hybridization with ductile steel fibers has been explored in UD composite laminates to improve mechanical properties (ductility, in particular) and multifunctionality of the composites [37–41]. Rehra et al. [38,39], in their studies on the hybridization of UD carbon-fiber/epoxy laminates with ductile steel fibers, concluded that the relative volume fraction of carbon and steel fibers, and steel fiber/matrix adhesion, were important factors that determined the increase in ductility due to hybridization. McBride and O'Brien [40,41] reported a significant improvement in toughness of hybrid glass/steel fiber UD composites under quasi-static tensile and cyclic loading. Nevertheless, hybridization with ductile steel fibers in composites with a textile meso-structure has not been explored in the literature to date. Thus, **Papers III, IV, and V** present the techniques of manufacturing, mechanical testing, and micro-damage analysis of composites with hybrid carbon/steel fiber textile reinforcement.

Table 2.1

Hybrid and non-hybrid textile composite configurations.

Type		Notation	Lay-up
Knitted	Reference Non-hybrid	C1	[(T300) ₈]*
		M1	[(SY 11/1) ₈]*
		M2	[(SY 11/1 3filaments) ₈]
	Hybrid	H1	[(Hybrid SY11/1+T300) ₈]
		H2	[(T300/SY11/1)] _{2s}
Woven	Reference Non-hybrid	C4	[T300 1x] ₄
		M4	[SY 11/1 2x] ₃
	Hybrid	H5	[T300 1x + SY 11/1 2x] ₄ [#]
		H6	[SY 11/1 2x + T300 1x] ₄
		H8	[(SY 11/1 2x + T300 1x)(T300 1x + SY 11/1 2x)] _{2.5}
		H9	[T300 1x / SY 11/1 2x] _{2.5}

*T300 is carbon-fiber yarn, SY 11/1 is steel-fiber yarn

[#]Warp yarn + Weft yarn

Hybrid carbon/steel fiber composites with plain woven and plain weft knitted meso-structures were manufactured using the vacuum infusion method [42]. Composite specimens with hybrid and non-hybrid configurations (as shown in Table 2.1) were tested under uniaxial quasi-static tensile loading along the warp-yarn direction at a 2 mm/min, i.e., 1%/min displacement rate. A comparison of initial elastic modulus values of the composite specimens is given in Fig. 2.5. It was found that the yarn-level hybridization (refer configuration H1) yielded composites with higher longitudinal stiffness than the layer-level hybridization (refer configuration H2) in the case of knitted meso-structure. For woven composites, the longitudinal stiffness values were mainly governed by the yarns along the warp direction.

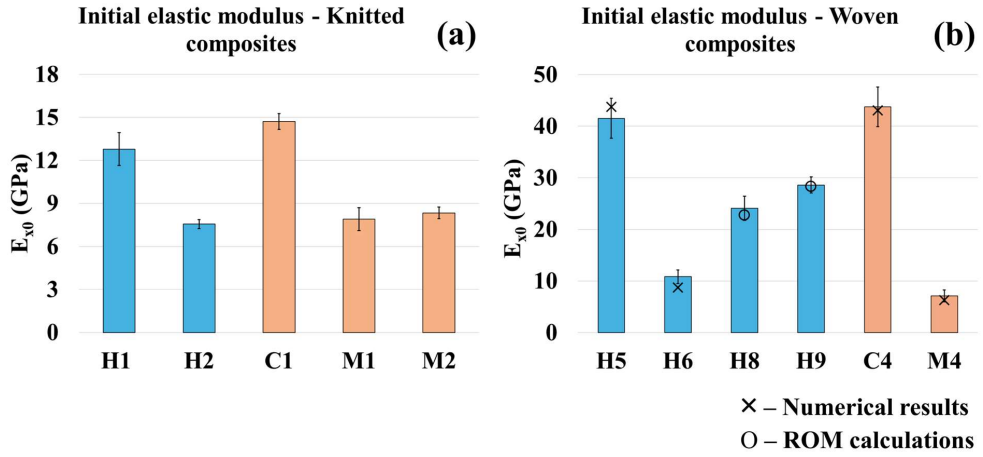


Figure 2.5. Comparison of initial longitudinal elastic modulus (E_{x0}) values in knitted and woven composite configurations. The cross-marks indicate FEM numerical results. The modulus values for H8 and H9 are calculated using the rule of mixtures (ROM), based on the numerical results of H5, H6, C4 and M4. Adapted from **Papers III** and **V**.

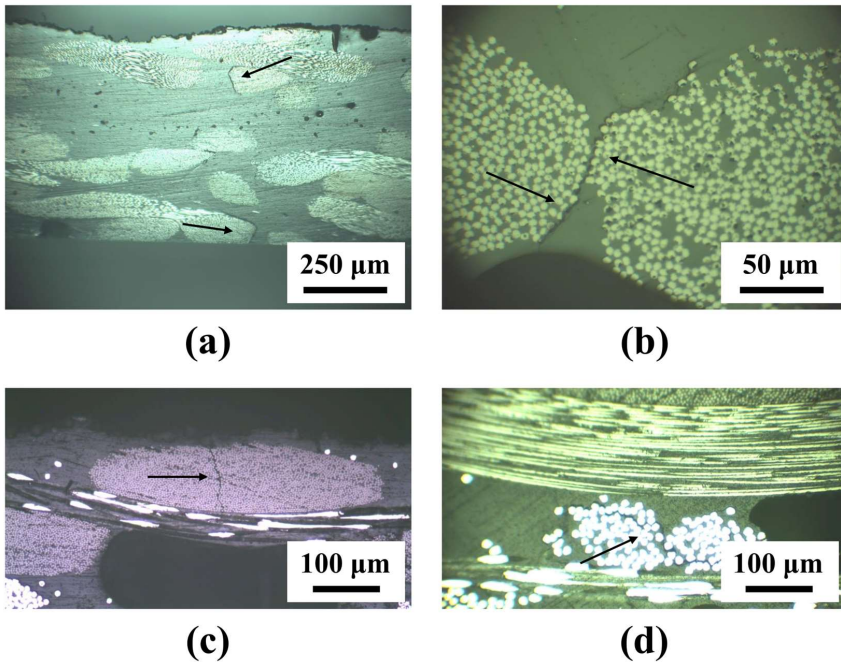


Figure 2.6. Damage initiation in knitted and woven composites under uniaxial quasi-static tensile loading along the warp yarn direction, (a) yarn/matrix-interface delaminations in C1 at 0.6% applied strain, (b) carbon-fiber weft-yarn/matrix delamination in H8 at 0.8% applied strain, (c) transverse crack through carbon-fiber weft-yarn in H8 at 0.8% applied strain, and (d) shear matrix crack through steel-fiber weft-yarn in H9 at 0.8% applied strain.

At the applied strain value of 0.6%, micro-damage initiation was observed in knitted composites along the yarn/matrix interfaces (see Fig. 2.6a). In knitted hybrid composites, the interface delaminations were present only along the carbon fiber yarn/matrix boundaries, and not along the steel fiber yarn/matrix boundaries. In the case of woven composites, micro-damage initiated around the applied strain values of 0.6 – 0.8% in the forms of carbon fiber weft yarn/matrix delaminations and transverse cracks through carbon fiber weft yarns (see Fig. 2.6b,c). Shear matrix cracks through steel fiber weft yarns were also observed (see Fig. 2.6d). The damage evolution took place in the form of interconnection of yarn/matrix-interface delaminations in knitted composites and coalescence of transverse cracks and yarn/matrix-interface delaminations in woven composites (see Fig. 2.7).

Comparing the final failure behavior of different composite configurations, the trend in tensile strength values was similar to the longitudinal stiffness values. The knitted composites with yarn-level hybridization exhibited higher longitudinal tensile strength than those with layer-level hybridization. In the case of woven composites, the longitudinal tensile strength was mainly governed by the yarns along the warp direction. The SEM analysis of fracture surfaces of the woven composite specimens showed large longitudinal debonds along the carbon fiber warp yarn/matrix-interface. In woven composite specimens with steel fiber warp yarns, small debonds were observed at the warp yarn/matrix interface, with microscale necking in the steel fibers prior to failure.

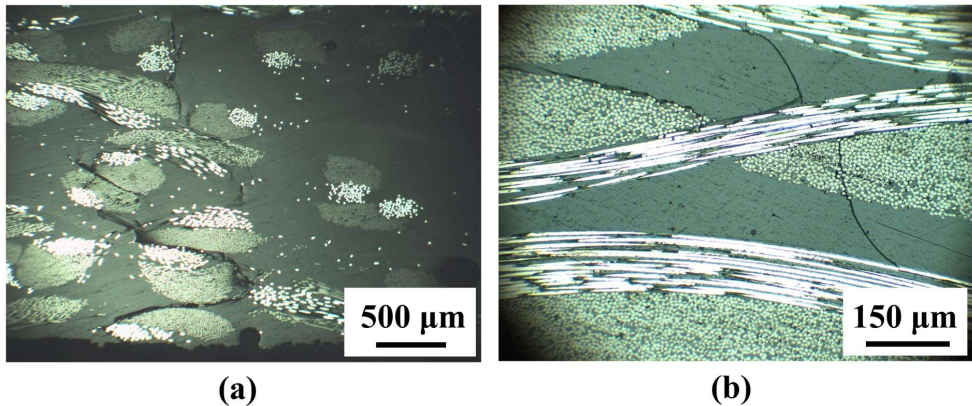


Figure 2.7. Damage evolution in knitted and woven composites under uniaxial quasi-static tensile loading along the warp yarn direction, (a) interconnection of yarn/matrix-delaminations in H1 at 1.2% applied strain, and (b) coalescence of delaminations and transverse cracks in C4 at 0.8% applied strain. Adapted from **Paper III**.

The mechanical behavior of hybrid and non-hybrid textile composites was studied numerically using FEM in **Papers IV and V**. 3D representative volume elements (RVEs) were generated for the composite configurations H5, H6, C4, and M4. The resin-impregnated fiber yarns were treated as homogeneous composite materials, and their orthotropic material properties were calculated using a micromechanics-based concentric cylinder assembly model.

The yarns had an elliptical cross-section. The warp yarns had sinewave curvature, whereas the weft yarns were assumed straight. The yarns were embedded in a cuboidal matrix block with isotropic material properties. The geometrical specifications of the yarns and the matrix block were determined from optical microscopy images of the composite specimens. The RVEs were further calibrated by comparing their elastic behavior with the experimental data. The assembly of yarns and matrix block was meshed with 3D solid elements in ABAQUS software. To study their behavior under quasi-static tensile loading, the elastoplastic damage model for fiber-reinforced composite materials developed by C  zar et al. [1,2] was applied to the textile yarns, and the concrete damaged plasticity model [43] available in ABAQUS software was applied to the matrix block.

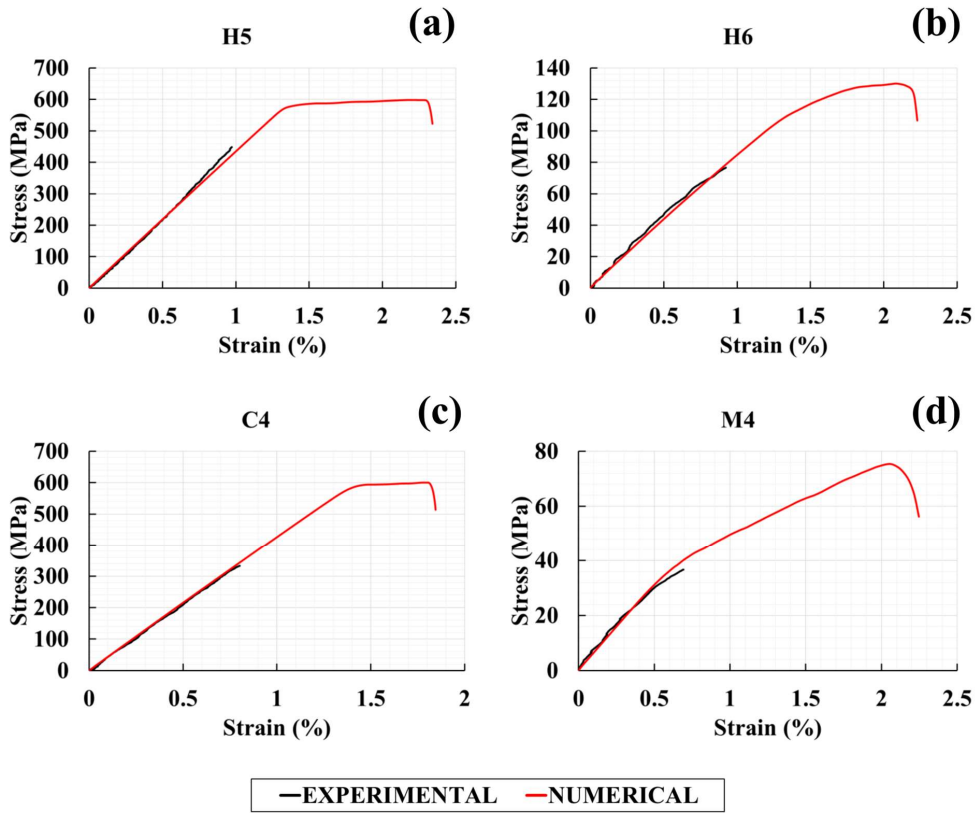


Figure 2.8. Comparison of the full stress-strain behavior under uniaxial quasi-static tensile loading along the warp yarn direction calculated using experimental and numerical techniques for the composite configurations (a) H5, (b) H6, (c) C4, and (d) M4. Adapted from **Paper V**.

A comparison of the full stress-strain behavior of the composite configurations calculated using experimental and numerical techniques is given in Fig. 2.8. The experimental stress-strain curves match closely with the numerical curves in the linear elastic stage. As shown in Fig. 2.8, the numerical curves provide further insight into the non-linear plastic and damage evolution

behavior of the composite configurations. In the hybrid configuration H5, the numerical stress-strain curve shows linear elastic behavior until micro-damage initiates in the form of matrix cracking at the weft yarn/matrix interface at 1.2% applied strain. This damage event at the steel fiber weft yarns does not change the slope of the stress-strain curve, as the macro-mechanical behavior of H5 is mainly governed by the carbon fiber warp yarns. Refer to Fig. 2.9, a knee point is observed in the stress-strain curve at 1.37% applied strain, which is caused by the initiation of fiber failure (denoted by damage parameter D_L) in the warp yarn. Along the plateau region seen in the stress-strain curve, the fiber damage distributes across the warp yarn, creating multiple clusters of fiber damage shown by red color in the damage contour plots. Finally, at an applied strain of 2.3%, the warp yarn fails at one of the fiber damage clusters, and the composite undergoes ultimate failure.

Paper V presents a parametric analysis of the full stress-strain behavior of the hybrid composite configuration H5, calculated using FEM. As the behavior of a textile composite under tensile loading along the warp yarn direction is mainly governed by the warp yarn characteristics, the warp yarn strength properties X_T (longitudinal tensile strength), Y_T (transverse tensile strength) and S_L (longitudinal shear strength) were chosen for the parametric analysis. As shown in Fig. 2.10a, it was observed that X_T affects the knee point in the stress-strain curve, i.e., the initiation of fiber damage in the warp yarn. A lower value of Y_T , as shown in Fig. 2.10b, results in a reduction in the ultimate tensile strength due to changes in overall damage evolution characteristics. Finally, the change in S_L , see Fig. 2.10c, causes a change in the distribution of damage in the warp yarn after the initiation of fiber damage, slightly altering the strain-to-failure.

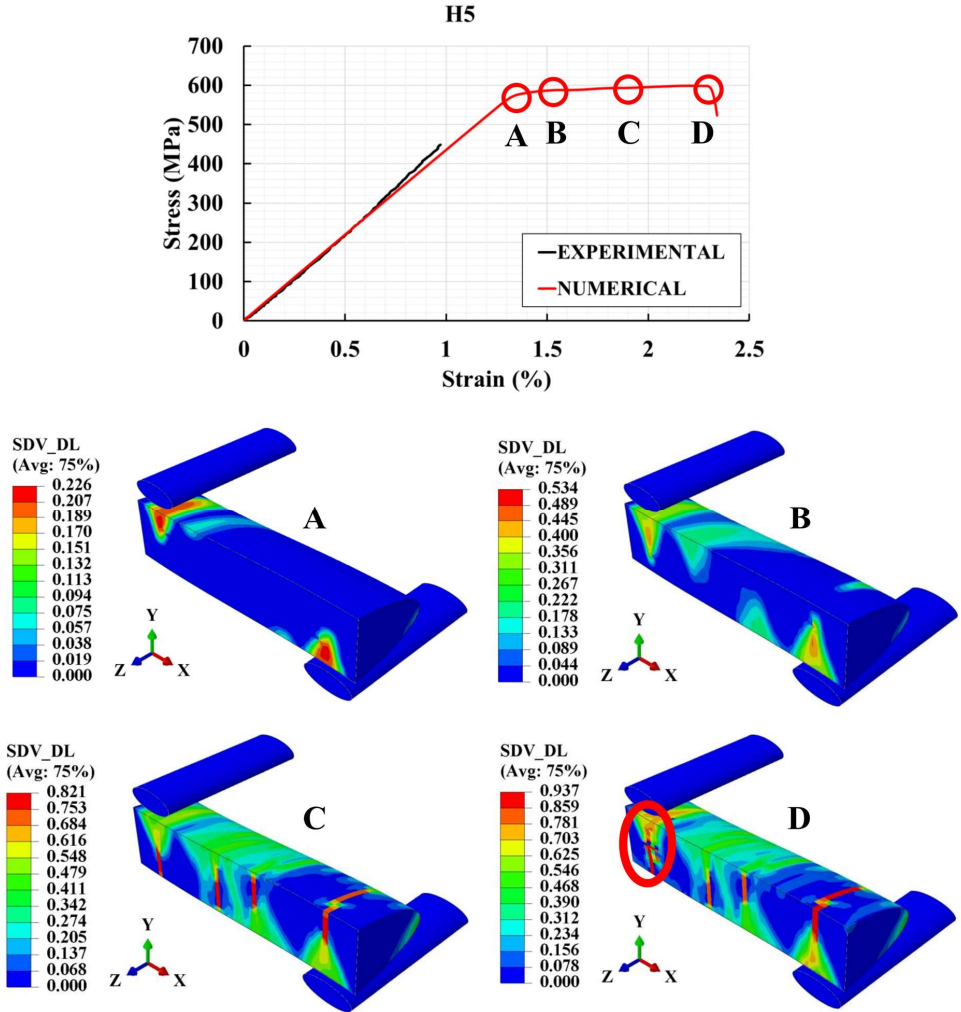


Figure 2.9. Evolution of longitudinal damage (D_L), i.e., fiber breakage, in the warp yarn of the hybrid composite configuration H5 under quasi-static tensile loading along the warp yarn direction. Adapted from **Paper V**.

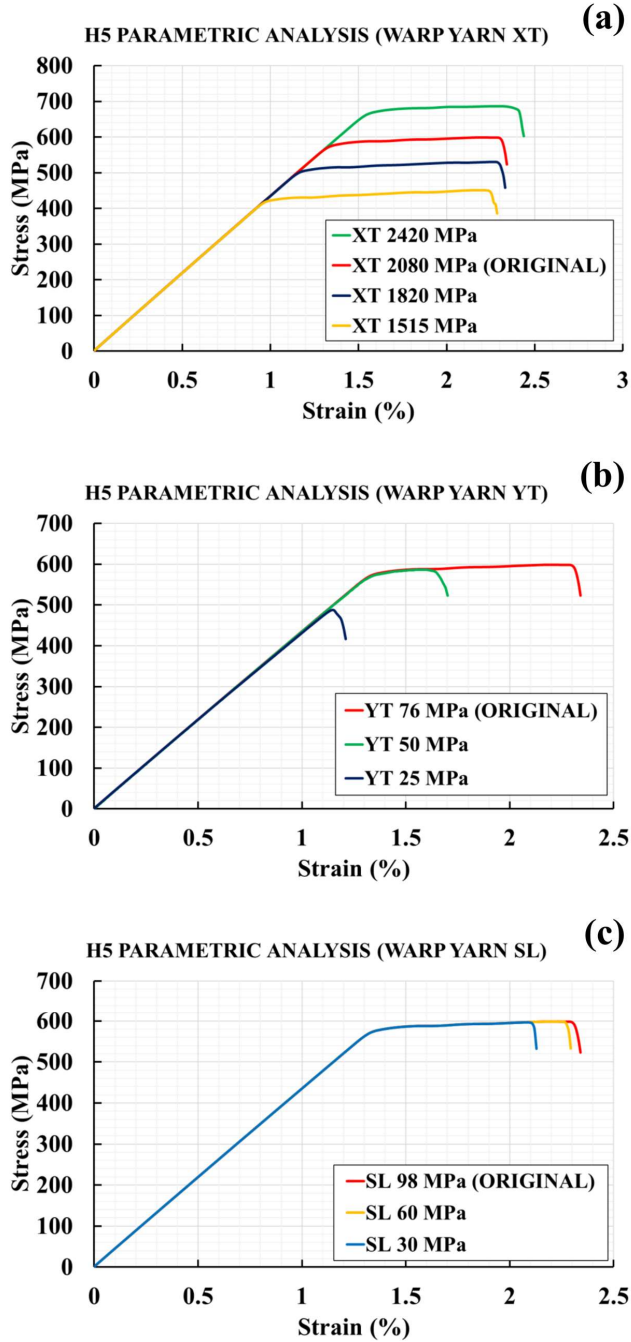


Figure 2.10. Parametric analysis of the full stress-strain behavior of the hybrid composite configuration H5 under uniaxial tensile loading along the warp yarn direction, by varying (a) warp yarn longitudinal strength X_T , (b) warp yarn transverse strength Y_T , and (c) warp yarn longitudinal shear strength S_L . Reprinted from **Paper V**.

2.3. Micro-damage evolution in induction-cured thermoset composites

Despite their excellent mechanical properties, the use of carbon-fiber-reinforced thermoset composites is limited to high-value engineering sectors due to the high cost of raw materials and manufacturing. The costs in manufacturing can be reduced by choosing a less energy-intensive alternative to the conventional autoclave method for the curing of thermoset resin [30]. Volumetric heating methods, such as microwave heating and electromagnetic induction heating, can be such energy-efficient alternatives for thermoset curing, where heat is generated from inside the composite part [44,45].

Induction heating of carbon-fiber-reinforced thermoset composites requires a closed loop of carbon fiber reinforcement, which can be achieved using multi-layered composite laminates with meso-structures such as textile and cross-ply [46,47]. An induction coil supplied with an alternating current induces current inside the closed-loop of carbon fibers with the help of an oscillating magnetic field, and the carbon fiber reinforcement gets heated due to different mechanisms, such as Joule losses, dielectric hysteresis, and contact resistance heating [48–50]. This, in turn, heats the thermoset matrix around the current-carrying carbon fibers, which can ultimately lead to the curing of the thermoset resin.

Induction heating of CF-reinforced thermoset composites is very sensitive to the distance between the induction coil and the workpiece (here, the composite part), as the maximum temperature reached in the workpiece has an inverse-square relation with the distance from the induction coil [51]. Also, it has been observed that the heating pattern on a textile or cross-ply composite resembles the coil shape [52]. In the case of a commonly used circular induction coil, it would make a near-circular heating pattern on the composite workpiece, creating a cold spot in the middle [53]. The generation of cold spots in the composite workpiece prevents uniform curing of the thermoset resin.

Using a novel moving-coil induction-heating setup described in **Paper VI**, a near-uniform cure of the epoxy resin in CF-reinforced [0,90]_s cross-ply composites was attempted. According to the recommended curing cycles for the material system used in the present study, i.e., XC130-C12UD-300(300) high-strength UD carbon fiber epoxy prepreg [54], the curing cycle of 4 hours at 100 °C was chosen based on the capabilities of the induction heating system. Uniformity of curing was assessed using a degree-of-cure analysis by differential scanning calorimetry (DSC) during the pilot test. Figure 2.11 shows the glass transition temperature (T_g) values for induction-cured and reference oven-cured composite specimens that is directly proportional to the degree of cure. A significant variation, up to 20%, is observed in the T_g values of induction-cured composite specimens, with the average value of glass transition temperature lower in the induction-cured than in the oven-cured composite specimens.

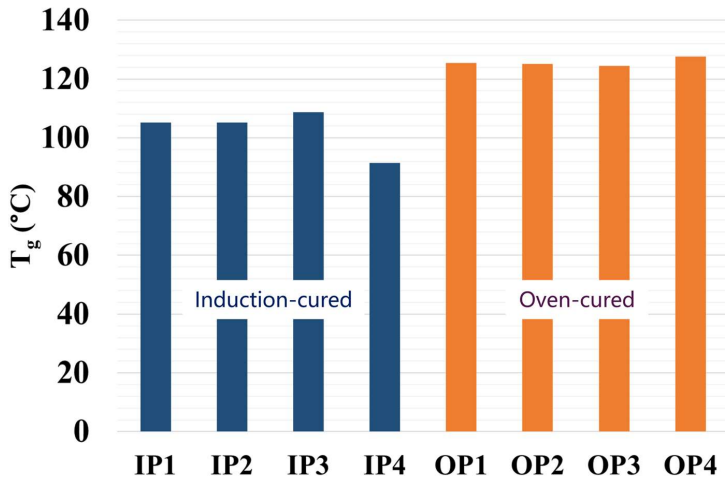


Figure 2.11. Glass transition temperature (T_g) values in induction-cured (IP1 – IP4) and oven cured (OP1 – OP4) specimens indicating the degree-of-cure. Adapted from **Paper VI**.

Following the pilot test, the predetermined path for the induction coil movement was improved to manufacture a new 300mm x 300mm composite plate with a [0,90]_s cross-ply layup using induction heating. Rectangular specimens were cut from different locations on the cured composite plate, and the variation in mechanical properties at these locations was studied using quasi-static tensile testing. Analysis of the mechanical properties of the new induction-cured composite specimens is currently ongoing.

Refer to Fig. 2.12 and Table 2.2, although the mechanical properties varied considerably, up to 9%, across different locations on the induction-cured composite plate, the average values of elastic modulus, ultimate tensile strength, and strain-to-failure of the induction-cured specimens are still within 6% difference as compared to the reference oven-cured specimens.

Induction heating using a moving-coil setup generates a transient temperature gradient in the workpiece. This forms soft and hard regions in a composite plate cured using the moving-coil setup. The air trapped inside the laminate can escape easily through the soft regions, but not through the hard regions. As the soft regions are always surrounded by the cold, hard regions, air remains trapped during the moving-coil induction curing process. Thus, it was observed that the induction-cured composite plates had a large number of voids. These voids served as the micro-damage initiation sites in the composite specimens. Fig. 2.13 shows a transverse crack in an induction-cured composite specimen initiated near a large void. Due to the high void content, all induction-cured composite specimens showed damage initiation and evolution at lower applied strain levels than the reference oven-cured specimens, as shown in Fig. 2.14.

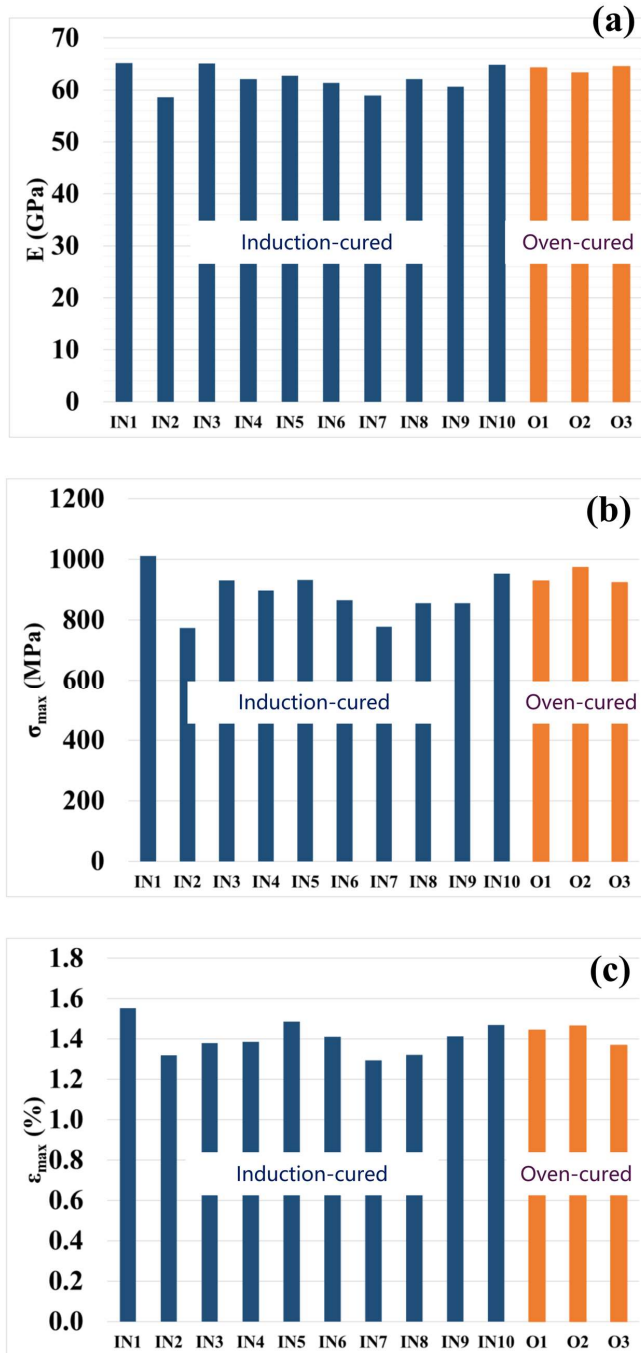


Figure 2.12. Comparison of (a) longitudinal elastic modulus, (b) ultimate tensile strength, and (c) strain-to-failure values in induction-cured and oven-cured [0,90]_s cross-ply laminates under uniaxial quasi-static tensile loading.

Table 2.2

Comparison of average mechanical properties between induction-cured and oven-cured composite specimens.

	Induction-cured	Oven-cured	Difference in %
Elastic modulus (GPa)	62.14 ± 2.40	64.04 ± 0.64	2.95
Tensile strength (MPa)	885.53 ± 75.37	941.96 ± 27.54	5.99
Strain-to-failure (%)	1.40 ± 0.08	1.43 ± 0.05	1.58

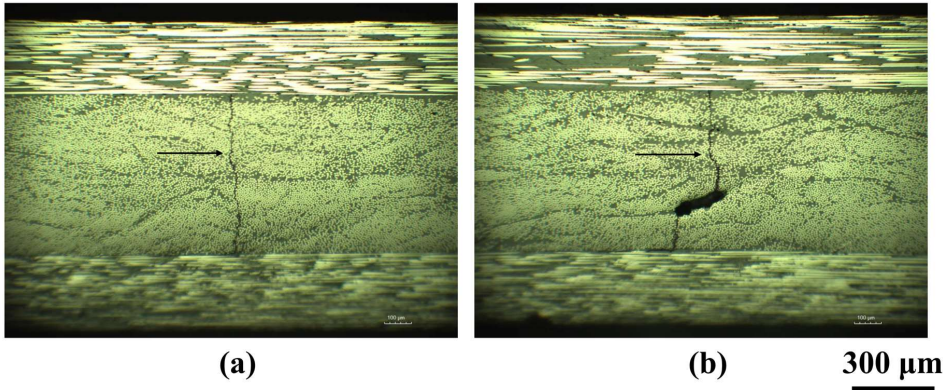
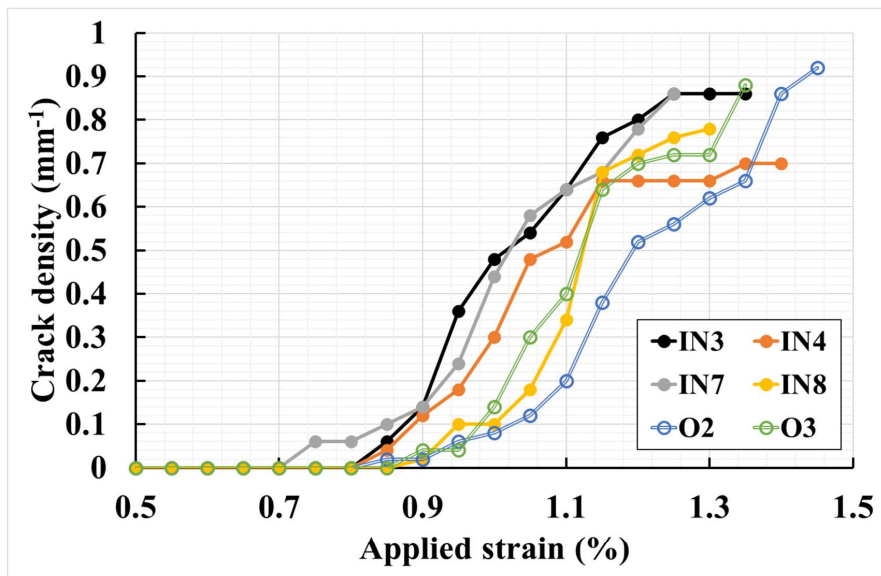
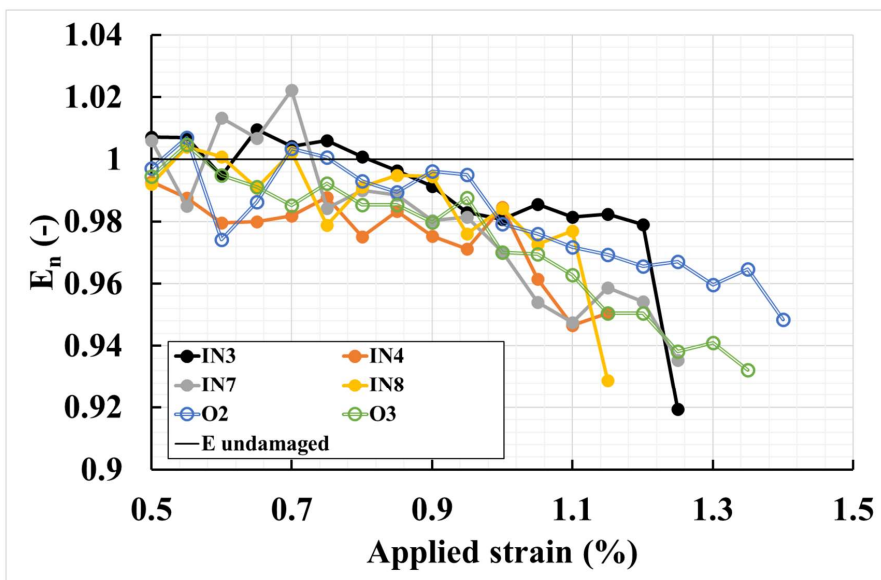


Figure 2.13. Transverse cracks generated in induction-cured [0,90]_s cross-ply composite specimens under uniaxial quasi-static tensile loading, (a) at a random location in the 90-layer, and (b) near a large void.



(a)



(b)

Figure 2.14. (a) Evolution of transverse crack density in induction-cured and oven-cured $[0,90]_s$ cross-ply laminates with increasing applied strain level, and (b) corresponding reduction in the longitudinal stiffness value of the composite specimens.

3. CONCLUSIONS

In the present thesis, micro-damage evolution in (a) cross-ply laminates with high crack density, (b) hybrid carbon/steel fiber textile composites, and (c) induction-cured thermoset composites has been studied. The main conclusions of the work are listed below:

1. High transverse stress gradients are created in the regions with high crack density in 90-layers in cross-ply laminates under uniaxial tensile loading. The stress gradients strongly affect the strain energy release rate available for the growth of a new transverse crack. In the regions with ultra-high crack density (crack density multiplied by 90-layer thickness > 2.0), the compressive stresses generated near the laminate midplane due to high stress gradients completely stop the growth of a new transverse crack.
2. At the tips of the existing transverse cracks in cross-ply laminates under uniaxial tensile loading, the generation of local delaminations takes place due to high stress concentration. The local delamination growth is strongly affected by the transverse crack density, 0/90-layer thickness ratio, and the specific ratios of elastic moduli of 0- and 90-layers, such as E_L^0/E_T^{90} and G_{LT}^0/E_T^{90} . The non-symmetrical I-shaped delaminations tend to grow into symmetrical I-shaped delaminations before growing further.
3. Damage initiation takes place in the form of weft yarn/matrix-interface delaminations in hybrid carbon/steel fiber woven composites under uniaxial tensile loading along the warp yarn direction. This damage event at the weft yarns does not affect the slope of the stress-strain curve of a hybrid composite with carbon fiber warp yarns, as in this case, the composite's macro-mechanical behavior is mainly governed by the warp yarn properties.
4. A knee point is observed in the stress-strain curve of a hybrid carbon/steel fiber woven composite with carbon fiber warp yarns when fiber damage is initiated in the warp yarns. Upon further loading, damage evolution takes place with the distribution of fiber damage and the formation of multiple clusters of fiber breaks. The failure of warp yarn takes place at a fiber-break cluster, which in turn results in the ultimate failure of the hybrid composite.
5. In the composites manufactured using a moving-coil induction heating setup, large void formation due to entrapment of air bubbles results in the initiation and evolution of micro-damage at a lower applied strain level than the oven-cured composites.

BIBLIOGRAPHY

- [1] Cózar IR, Otero F, Maimí P, González E V., Miot S, Turon A, et al. A three-dimensional plastic-damage model for polymer composite materials. *Compos Part A Appl Sci Manuf* 2022;163. <https://doi.org/10.1016/j.compositesa.2022.107198>.
- [2] Cózar IR, Otero F, Maimí P, González E V., Turon A, Camanho PP. An enhanced constitutive model to predict plastic deformation and multiple failure mechanisms in fibre-reinforced polymer composite materials. *Compos Struct* 2024;330. <https://doi.org/10.1016/j.compstruct.2023.117696>.
- [3] París F, Correa E, Mantič V. Kinking of transversal interface cracks between fiber and matrix. *Journal of Applied Mechanics, Transactions ASME* 2007;74. <https://doi.org/10.1115/1.2711220>.
- [4] Gorbatikh L, Lomov S V. Damage accumulation in textile composites. *Modeling Damage, Fatigue and Failure of Composite Materials*, 2016. <https://doi.org/10.1016/B978-1-78242-286-0.00003-0>.
- [5] Nairn JA. Matrix Microcracking in Composites. *Comprehensive Composite Materials*, 2000. <https://doi.org/10.1016/b0-08-042993-9/00069-3>.
- [6] Blázquez A, Mantič V, París F, McCartney LN. Stress state characterization of delamination cracks in [0/90] symmetric laminates by BEM. *Int J Solids Struct* 2008;45:1632–62. <https://doi.org/10.1016/J.IJSOLSTR.2007.10.013>.
- [7] Lim SG, Hong CS. Prediction of Transverse Cracking and Stiffness Reduction in Cross-Ply Laminated Composites. *J Compos Mater* 1989;23:695–713. <https://doi.org/10.1177/002199838902300704>.
- [8] Hashin Z. Analysis of cracked laminates: a variational approach. *Mechanics of Materials* 1985;4. [https://doi.org/10.1016/0167-6636\(85\)90011-0](https://doi.org/10.1016/0167-6636(85)90011-0).
- [9] Varna J, Berglund L. Multiple transverse cracking and stiffness reduction in cross-ply laminates. *Journal of Composites Technology and Research* 1991;13. <https://doi.org/10.1520/ctr10213j>.
- [10] Varna J, Berglund LA. Thermo-elastic properties of composite laminates with transverse cracks. *Journal of Composites Technology and Research* 1994;16. <https://doi.org/10.1520/ctr10397j>.
- [11] Berthelot JM. Transverse cracking and delamination in cross-ply glass-fiber and carbon-fiber reinforced plastic laminates: Static and fatigue loading. *Appl Mech Rev* 2003;56. <https://doi.org/10.1115/1.1519557>.
- [12] Lomov S V., Ivanov DS, Truong TC, Verpoest I, Baudry F, Vanden Bosche K, et al. Experimental methodology of study of damage initiation and development in textile composites in uniaxial tensile test. *Compos Sci Technol* 2008;68. <https://doi.org/10.1016/j.compscitech.2007.07.005>.
- [13] Kulkarni AN, Pupurs A, Šņepsts A, Irbe M. MICRO-DAMAGE EVOLUTION IN HYBRID CARBON/STEEL KNITTED AND WOVEN FABRIC COMPOSITES. In: Falzon B, McCarthy C, editors. *Proceedings of the 2023 International Conference on Composite Materials*, Belfast: Queen's University Belfast; 2023.

- [14] Kulkarni AN, Pupurs A, Kononova O, Jevstignejevs V, Maimi P, Subramani A. Hybrid carbon/steel fiber textile composites for ductility improvements. n.d.
- [15] Chen J, Yu Z, Jin H. Nondestructive testing and evaluation techniques of defects in fiber-reinforced polymer composites: A review. *Front Mater* 2022;9. <https://doi.org/10.3389/fmats.2022.986645>.
- [16] Kulkarni AN. Energy and Strength-based Criteria for Intralaminar Crack Growth in Regions with High Stress Gradients. Independent thesis Advanced level (degree of Master (Two Years)). Luleå University of Technology, 2021.
- [17] Garrett KW, Bailey JE. Multiple transverse fracture in 90° cross-ply laminates of a glass fibre-reinforced polyester. *J Mater Sci* 1977;12:157–68. <https://doi.org/10.1007/BF00738481>.
- [18] Kulkarni AN, Pupurs A, Varna J. Analysis of initiation and growth of new transverse cracks in high crack density regions in cross-ply laminates. *J Compos Mater* 2025;59:539–50. <https://doi.org/10.1177/00219983241295811>.
- [19] Berthelot JM, Le Corre JF. Statistical analysis of the progression of transverse cracking and delamination in cross-ply laminates. *Compos Sci Technol* 2000;60:2659–69. [https://doi.org/10.1016/S0266-3538\(00\)00140-8](https://doi.org/10.1016/S0266-3538(00)00140-8).
- [20] O'Brien TK. ANALYSIS OF LOCAL DELAMINATIONS AND THEIR INFLUENCE ON COMPOSITE LAMINATE BEHAVIOR. ASTM Special Technical Publication, 1985. <https://doi.org/10.1520/stp36310s>.
- [21] Wang ASD, Kishore NN, Li CA. Crack development in graphite-epoxy cross-ply laminates under uniaxial tension. *Compos Sci Technol* 1985;24. [https://doi.org/10.1016/0266-3538\(85\)90058-2](https://doi.org/10.1016/0266-3538(85)90058-2).
- [22] Kim TW, Kim HJ, Im S. Delamination crack originating from transverse cracking in cross-ply composite laminates under extension. *Int J Solids Struct* 1991;27:1925–41. [https://doi.org/10.1016/0020-7683\(91\)90186-J](https://doi.org/10.1016/0020-7683(91)90186-J).
- [23] Kulkarni AN, Pupurs A, Varna J. Parameters affecting growth of local delaminations at transverse crack tips in [0m,90n]s cross-ply laminates. *Eng Fract Mech* 2025;328:111549. <https://doi.org/10.1016/J.ENGFRACMECH.2025.111549>.
- [24] Wang L, Wu J, Chen C, Zheng C, Li B, Joshi SC, et al. Progressive failure analysis of 2D woven composites at the meso-micro scale. *Compos Struct* 2017;178. <https://doi.org/10.1016/j.compstruct.2017.07.023>.
- [25] Callens MG, Gorbatiikh L, Verpoest I. Ductile steel fibre composites with brittle and ductile matrices. *Compos Part A Appl Sci Manuf* 2014;61:235–44. <https://doi.org/10.1016/J.COMPOSITESA.2014.02.006>.
- [26] Allaer K, De Baere I, Lava P, Van Paepegem W, Degrieck J. On the in-plane mechanical properties of stainless steel fibre reinforced ductile composites. *Compos Sci Technol* 2014;100. <https://doi.org/10.1016/j.compscitech.2014.05.009>.
- [27] Czél G, Wisnom MR. Demonstration of pseudo-ductility in high performance glass/epoxy composites by hybridisation with thin-ply carbon prepreg. *Compos Part A Appl Sci Manuf* 2013;52. <https://doi.org/10.1016/j.compositesa.2013.04.006>.

- [28] Swolfs Y, Verpoest I, Gorbatikh L. Recent advances in fibre-hybrid composites: materials selection, opportunities and applications. *International Materials Reviews* 2019;64. <https://doi.org/10.1080/09506608.2018.1467365>.
- [29] Corvo Alguacil M, Umeki K, Gaidukovs S, Barkāne A, You S, Joffe R. The impact of thermal treatment parameters on the preservation of carbon fiber mechanical properties after reclamation. *Current Research in Green and Sustainable Chemistry* 2024;9. <https://doi.org/10.1016/j.crgsc.2024.100431>.
- [30] Collinson MG, Bower MP, Swait TJ, Atkins CP, Hayes SA, Nuhiji B. Novel composite curing methods for sustainable manufacture: A review. *Composites Part C: Open Access* 2022;9. <https://doi.org/10.1016/j.jcomc.2022.100293>.
- [31] Pupurs A, Lācis V, Šņepsts A, Kulkarni AN, Irbe M. ANALYSIS OF MICRO-DAMAGE EVOLUTION IN INDUCTION-CURED THERMOSET COMPOSITES. In: Binetruy C, Jacquemin F, editors. *Proceedings of the 21st European Conference on Composite Materials, Nantes: The European Society for Composite Materials (ESCM) and the Ecole Centrale de Nantes*; 2024, p. 1529–34.
- [32] McCartney LN, Schoeppner GA. Predicting the effect of non-uniform ply cracking on the thermoelastic properties of cross-ply laminates. *Compos Sci Technol* 2002;62. [https://doi.org/10.1016/S0266-3538\(02\)00091-X](https://doi.org/10.1016/S0266-3538(02)00091-X).
- [33] Hajikazemi M, Sadr MH. A variational model for stress analysis in cracked laminates with arbitrary symmetric lay-up under general in-plane loading. *Int J Solids Struct* 2014;51. <https://doi.org/10.1016/j.ijsolstr.2013.10.024>.
- [34] Hajikazemi M, Sadr MH, Hosseini-Toudeshky H, Mohammadi B. Thermo-elastic constants of cracked symmetric laminates: A refined variational approach. *Int J Mech Sci* 2014;89. <https://doi.org/10.1016/j.ijmecsci.2014.08.008>.
- [35] Hajikazemi M, McCartney LN. Comparison of Variational and Generalized Plane Strain approaches for matrix cracking in general symmetric laminates. *International Journal of Damage Mechanics* 2018;27. <https://doi.org/10.1177/1056789516685381>.
- [36] Rahmani M, Farrokhbadi A. Prediction of induced delamination development in [0/90]_s composite laminates using a computational analytical approach. *Compos Struct* 2018;203. <https://doi.org/10.1016/j.compstruct.2018.07.018>.
- [37] Hannemann B, Backe S, Schmeer S, Balle F, Breuer UP. New multifunctional hybrid polymer composites reinforced by carbon and steel fibers. *ICCM International Conferences on Composite Materials*, vol. 2015- July, 2015.
- [38] Rehra J, Hannemann B, Schmeer S, Hausmann J, Breuer UP. Approach for an Analytical Description of the Failure Evolution of Continuous Steel and Carbon Fiber Hybrid Composites. *Adv Eng Mater* 2019;21. <https://doi.org/10.1002/adem.201800565>.
- [39] Rehra J, Jungbluth J, Katri B, Schmeer S, Gurka M, Balle F, et al. Damage and failure mechanisms of hybrid carbon fiber and steel fiber reinforced polymer composites. *Compos Part A Appl Sci Manuf* 2024;185:108366. <https://doi.org/10.1016/j.compositesa.2024.108366>.

- [40] McBride AK, Turek SL, Zaghi AE, Burke KA. mechanical behavior of hybrid glass/steel fiber reinforced epoxy composites. *Polymers (Basel)* 2017;9. <https://doi.org/10.3390/polym9040151>.
- [41] O'Brien C, Zaghi AE. Mechanical characteristics of hybrid composites with $\pm 45^\circ$ glass and $0^\circ/90^\circ$ stainless steel fibers. *Materials* 2018;11. <https://doi.org/10.3390/ma11081355>.
- [42] Andrejs P, Inita B, Inga Ļ, Māris M, Mārtiņš I, Alens Š. MANUFACTURING AND EXPERIMENTAL TESTING OF COMPOSITE/METAL JOINTS WITH TEXTILE MESOSTRUCTURE. *ECCM 2022 - Proceedings of the 20th European Conference on Composite Materials: Composites Meet Sustainability*, vol. 2, 2022.
- [43] Concrete Damaged Plasticity. *ABAQUS Analysis User's Guide*, vol. Volume III Materials, 2016.
- [44] Thostenson ET, Chou TW. Microwave and conventional curing of thick-section thermoset composite laminates: Experiment and simulation. *Polym Compos* 2001;22:197–212. <https://doi.org/10.1002/PC.10531>.
- [45] Bayerl T, Duhovic M, Mitschang P, Bhattacharyya D. The heating of polymer composites by electromagnetic induction - A review. *Compos Part A Appl Sci Manuf* 2014;57. <https://doi.org/10.1016/j.compositesa.2013.10.024>.
- [46] Becker S, Mitschang P. Influence of thread count of carbon twill textile-reinforced polyamide 66 laminates on the inductive heating behavior. *Journal of Thermoplastic Composite Materials* 2021;34. <https://doi.org/10.1177/0892705719854493>.
- [47] Kidangan RT, Unnikrishnakurup S, Krishnamurthy CV, Balasubramaniam K. The influence of interlaminar microstructure on the induction heating patterns of CFRP laminates. *Mater Today Commun* 2022;33. <https://doi.org/10.1016/j.mtcomm.2022.104338>.
- [48] Miller AK, Chang C, Payne A, Gur M, Menzel E, Peled A. The nature of induction heating in graphite-fiber, polymer-matrix composite material. *SAMPE Journal* 1990;26.
- [49] Fink BK, McCullough RL, Gillespie JW. A local theory of heating in cross-ply carbon fiber thermoplastic composites by magnetic induction. *Polym Eng Sci* 1992;32. <https://doi.org/10.1002/pen.760320509>.
- [50] Kim H, Yarlagadda S, Gillespie JW, Shevchenko NB, Fink BK. A study on the induction heating of carbon fiber reinforced thermoplastic composites. *Advanced Composite Materials* 2002;11. <https://doi.org/10.1163/156855102753613309>.
- [51] Rudolf R, Mitschang P, Neitzel M. Induction heating of continuous carbon-fibre-reinforced thermoplastics. *Compos Part A Appl Sci Manuf* 2000;31. [https://doi.org/10.1016/S1359-835X\(00\)00094-4](https://doi.org/10.1016/S1359-835X(00)00094-4).
- [52] Uzzell JPN, Pickard LR, Hamerton I, Ivanov DS. Novel cellular coil design for improved temperature uniformity in inductive heating of carbon fibre composites. *Mater Des* 2024;237. <https://doi.org/10.1016/j.matdes.2023.112551>.
- [53] Lundström F, Frogner K, Andersson M. Analysis of the temperature distribution in weave-based CFRP during induction heating using a simplified numerical model with a

cross-ply representation. Compos B Eng 2021;223.

<https://doi.org/10.1016/j.compositesb.2021.109153>.

[54] XC130 Autoclave Cure Component Prepreg - Technical Datasheet - 10/11/2022. n.d.

Analysis of initiation and growth of new transverse cracks in high crack density regions in cross-ply laminates

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Abstract

In cross-ply laminates under loading, multiple intralaminar (transverse) cracking in the 90-layers causes laminate stiffness reduction which can be predicted by analytical models studying average of the stress perturbation between transverse cracks. The commonly used analytical models such as shear-lag type or variational-type models assume that stress in damaged 90-layers does not depend on the laminate thickness coordinate. But with increase in crack density, the stress perturbations created by transverse cracks start to interact, generating high stress gradients in through-the-thickness direction in the 90-layers. Location of a new crack and features of its propagation in a high stress gradient region between two cracks in a cross-ply laminate are analyzed in this paper. FEM is used to calculate stress distribution between two cracks. The location of the next crack and its initial orientation is found using principal stresses and orientation of principal axes. Most often these cracks are initiated at ply interface in the middle between already existing cracks. Their propagation across the layer thickness is analyzed using energy release rate based criterion. It is shown that at very high crack density the crack propagation across the layer thickness is unstable in the beginning, but it is stopped when approaching the middle of the layer where the crack opening becomes zero. Presence of local delaminations at the tips of existing cracks changes the energy release rate for propagation of new transverse cracks.

Keywords

Multiple transverse cracking, high crack density, stress gradient, energy release rate, local delaminations

Introduction

Under loading, some layers in laminates are subjected to combination of tensile transverse and in-plane shear stresses causing intralaminar (transverse) cracks. These cracks run parallel to fibers, the crack plane is transverse to the laminate midplane^{1–4} and they often span the whole thickness of the layer. A single transverse crack does not result in ultimate failure of the laminate, as after transverse cracking, load is shared by remaining layers.^{2,4–6} With increasing loading, at some distance from the created crack, new cracks are formed because the stress transfer through the layer-interface (shear lag mechanism^{5,6}) ensures recovery of the stress in the damaged layer. At higher crack densities, transverse cracks also lead to other types of failures such as delaminations^{1,7–9} at the transverse crack tips or localized fiber fracture in the neighboring layer.

In uniaxially loaded cross-ply laminates, as the applied tensile load increases, multiple transverse cracks form in the 90-layers eventually reaching saturation.¹⁰ Alternatively, fiber breakage occurs in the 0-layers and the subsequent

laminate failure prevents transverse crack density in the 90-layers from reaching saturation.¹¹ Competing damage mechanisms such as delaminations may also lead to ultimate failure of the laminate before the transverse crack density could reach to a very high value.¹² Various experimental methods can be used to measure the transverse crack density such as optical measurements,¹³ X-ray imaging³ or acoustic emission analysis.¹⁴

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Data Availability Statement included at the end of the article

An immediate consequence of transverse cracking is degradation of laminate stiffness.^{2,6,11,15-18} In order to accurately predict the stiffness reduction, characterization of stress state in the damaged 90-layers is necessary. In literature, various analytical models of different complexity are available to calculate the local stress state around a transverse crack. It was shown¹⁹ that the effect of the transverse crack on laminate stiffness depends on the difference between the far field transverse stress in the layer and the volume averaged transverse stress in the layer with cracks. In most realizations of the approach, sometimes referred as “micromechanics modelling”, an approximate analytical expression for stress distribution is obtained and used, see, for example,¹ to derive expression for the damaged laminate axial modulus and sometimes also for the Poisson’s ratio. Most of the used models assume the shape of the dependency of the transverse stress on thickness coordinate, when calculating the local stress state in the repeating element between cracks. The simplest models use the shear lag concept or variational principles and assume that the transverse stress in the layer does not depend on the thickness coordinate^{4,15,18,20-23} or the dependency is defined by a simple shape function with coefficients found in result of solution.^{17,18} Usually these solutions are applicable to cross-ply type of laminates with cracks in 90-layers only. More accurate models lead to very complex analysis and numerical procedures²⁴⁻²⁸. In the referred papers, the ply with cracks is divided in many sub-layers, keeping simple stress approximation in each of them. The accuracy of these semi-analytical models is comparable with FEM modelling.

From the above discussion follows that only the complex semi-analytical models²⁵⁻²⁸ would be suitable for analysis of initiation and growth of new transverse cracks in high crack density region where all stress components in the layer (including the out-of-plane) are significant and the dependency on the coordinate in the laminate thickness direction is strong. However, the corresponding codes are available only to authors of.²⁵⁻²⁸ In this paper, the FEM software ANSYS²⁹ is used for calculation of the stress-state.

The need for numerical methods is even more obvious in analyzing growth of an individual crack.

As a criterion for the initiation of a crack, either stress-based, or energy-based criteria have been used. Both approaches use the calculated stress state and obtain parameters used to predict onset and multiplication of transverse cracking in the 90-layers. Strength-based approach, developed by Manders et al. (1983)³⁰ and Peters and Chou (1987)⁵ considers Weibull distribution of transverse failure stress in 90-layers. Considering the 90-layer as consisting of a chain of small elements, probability of failure of an element is calculated based on its proximity to an existing transverse crack and using Weibull distribution of failure stress. Energy-based approach, on the other hand, uses fracture mechanics parameters such as critical energy release rate (ERR).

The energy-based approach for laminates was pioneered by Wang (1984).³¹ This approach for prediction of propagation of a transverse crack uses Griffith criterion, which states that at the instance of crack propagation in solids, the loss of stored potential energy U is entirely converted to the surface energy S of the newly created crack. The balance of energy is given in equation (1),

$$-\frac{\partial U}{\partial a} = \frac{\partial S}{\partial a} \quad (1)$$

where a is the crack length. The right-hand side of equation (1) is often replaced by critical energy release rate $G_c = 2\gamma$, where γ is the free surface energy density of the material. A crack would extend when the available ERR (left-hand side in equation (1)) at the crack tip reaches its critical value. The value of ERR under a given stress state for a given crack length can be found by FEM using for example virtual crack closure technique or J -integral method.³² Such energy-based approach to predict the load required for crack propagation was combined with the strength-based approach to predict the location of crack initiation by Dvorak and Laws (1987)³³ to study multiple transverse cracking in 90-layers.

At high crack density perturbations in the stress state caused by transverse cracks start to interact with each other. This gives rise to high stress gradients in through-the-thickness direction in the 90-layer.¹⁸

The present work focuses on initiation and propagation in ply thickness direction of new cracks between two closely located pre-existing cracks in $[(0_2/90_4/0_2)_n]_s$ cross-ply glass-fiber epoxy composite (denoted as GF/EP) and carbon-fiber epoxy composite (CF/EP) laminates using FEM. The particular symmetric cross-ply laminate lay-up $[(0_2/90_4/0_2)_n]_s$ is selected so that a parametric study of stress distribution at high crack density can be conducted where symmetric cross-ply laminate lay-ups (i) with the same thickness of supporting 0-layers but with periodicity of the lay-up in the thickness direction leading to non-zero out-of-plane normal stress on the surface of the periodic element $[(0/90_4/0)_n]_s$, and (ii) with traction-free top- and bottom-faces $[0_2/90_4/0_2]$, have been considered.

The main objective of this work is to study transverse cracking in high crack density regions. The study is performed by analyzing potential locations for new transverse crack initiation and by investigating ERR trends for crack propagation along the ply thickness direction. The novelty of this work lies in considering very high crack densities, where stress fields become highly non-uniform including the thickness direction. Such non-uniform stress fields are often overlooked in models predicting crack evolution which assume uniform stress distribution along the thickness of the 90-layer. The stress state in the 90-layer in high crack density region is studied and is compared with the

stress state in the 90-layer in low crack density region. Location for initiation and orientation of two new types of cracks (delta crack close to the tip of an existing intralaminar crack and transverse crack close to the 0/90 layer-interface in the middle between existing intralaminar cracks) is found by calculating the main principal stress. Growth of the new transverse crack, in the region with very high transverse stress dependence on thickness coordinate, between two close intralaminar cracks that may have or may not have delaminations at the crack tips is analyzed by means of the available ERR. It has been observed that the trends in ERR are closely linked to local transverse stress state in the 90-layer.

Model description

Representative volume element of a cross-ply laminate with $[(0_2/90_4/0_2)_n]_s$ layup is a cross-ply laminate $[0_2/90_4/0_2]$ in a stack of n laminates placed on top of each other. With proper boundary conditions generated in ANSYS FEM software,²⁹ stress analysis and ERR calculations for this laminate and for stand-alone $[0_2/90_4/0_2]$ laminate with cracks in 90-layer can be performed. A schematic of a representative unit depicting the left-half of the region (in deformed state) between two adjacent intralaminar cracks in a damaged 90-layer is shown in Figure 1. The schematic depicts the 2D FEM model that was used for stress analysis between two fully formed through-the-thickness intralaminar cracks (referred hereafter as model 1). The distance between intralaminar cracks is denoted as ' $2a_0$ ', the 0-layer thickness is denoted as ' b ', and the 90-layer thickness is denoted as ' $2d$ '. The ply thickness $t = 0.072$ mm, and for $[(0_2/90_4/0_2)_n]_s$ and $[0_2/90_4/0_2]$ laminate layups, 0-layer thickness $b = 2t$. For $[(0/90_4/0)_n]_s$ laminate layup with relatively thinner 0-layers, 0-layer thickness $b = t$ and for all the laminate layups, 90-layer thickness $2d = 4t$. In model 1, displacement (U_x) which corresponds to the average strain of 1% in 0-layer was applied along z -axis at $x = 0$ and symmetry boundary conditions were applied on 0-layers at $x = -a_0$. To model a $[(0_2/90_4/0_2)_n]_s$ laminate, the bottom-face of model 1 was constrained with zero displacement ($u_z = 0$) along x -axis and the top-face was constrained with uniform displacement (u_z coupling) along x -axis. For the sake of comparison, calculations were performed also for $[(0/90_4/0)_n]_s$ laminate with relatively thinner 0-layers and for $[0_2/90_4/0_2]$ laminate with traction-free top- and bottom-faces.

Schematics shown in Figures 2 and 3 depict the FEM models for repeating elements between two intralaminar cracks (without and with delaminations respectively) containing also a small transverse crack initiated at the upper 0/90 layer-interface in the middle between the two intralaminar cracks. Thus, the new crack (denoted as N-crack in Figures 2 and 3) with initial length $a_N = 0.05d$ is

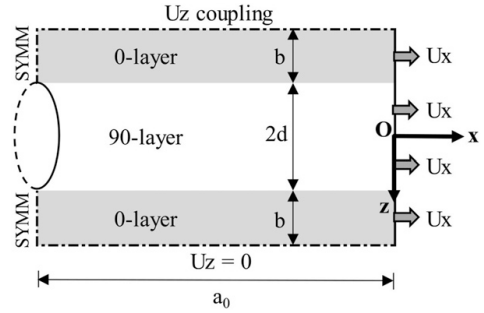


Figure 1. Left half of the deformed representative volume element of $[(0_2/90_4/0_2)_n]_s$ cross-ply laminate for stress analysis (model 1).

introduced in the 90-layer originating at the 0/90 layer-interface at $x = 0, z = -d$. The origin for initiation and the direction for subsequent growth of the N-crack inside 90-layer was assumed based on stress analysis results described in the sections "Stress distribution between cracks" and "Principal stresses close to the existing crack".

Growth of N-crack in the 90-layer through-the-thickness direction (along positive z direction) towards the lower interface is modelled and ERR for propagation at the tip of N-crack is calculated using ANSYS²⁹ built-in J -integral calculation routine. The boundary conditions used in models 2 and 3 are the same as the ones used in model 1. In model 3, growth of N-crack is modelled in the presence of local delaminations originating at the tips of existing intralaminar cracks (denoted as D in Figure 3) with a fixed length of $a_d = 0.5a_0$.

For an adequate representation of the effect of intralaminar cracking, the intralaminar crack density ρ (number of intralaminar cracks in a layer per unit length) was normalized with respect to the thickness of the 90-layer, $2d$:

$$\rho_n = \frac{1}{2a_0} 2d \quad (2)$$

2D 8-node elements PLANE183²⁹ with plane strain behavior were used in all FEM models. Plane strain elements were selected because the Poisson's ratio of cross-ply laminates is very small, and the contraction strain of the laminate is much smaller than the applied strain. Since, for relatively high crack density cases ($\rho_n = 1.0, 2.0$) the transverse stress $\sigma_x^{90} = \sigma_T(x, z)$ may be compressive locally, contact elements CONTA172 and TARGE169 were used in ANSYS²⁹ at N-crack faces and delamination faces (for $\rho_n = 0.5, 1.0$ in model 3) to avoid element interpenetration.

In all the models and calculations, a uniform temperature change $\Delta T = -100^\circ\text{C}$ was applied simultaneously with strain ϵ_{appl} to simulate thermal stresses which arise in laminates by cooling from the manufacturing temperature

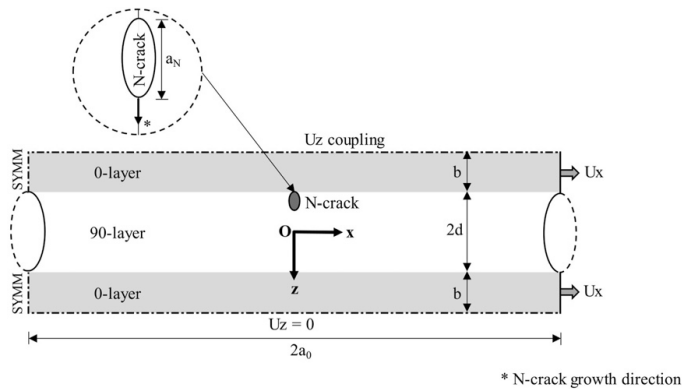


Figure 2. The deformed representative volume element of $[(0_2/90_4/0_2)_n]_s$ cross-ply laminate for ERR calculations (model 2). New transverse crack 'N-crack' is introduced at the 0/90 layer-interface in the 90-layer.

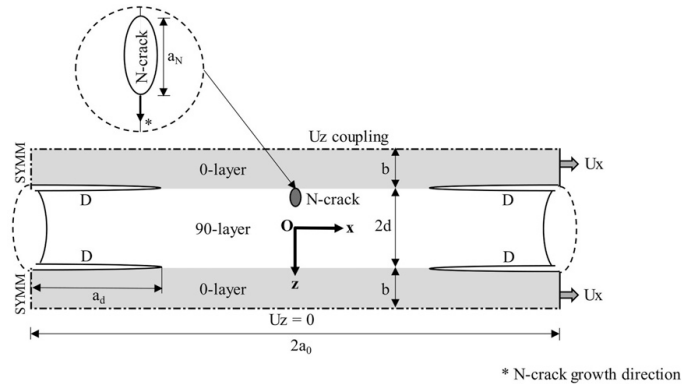


Figure 3. The deformed representative volume element of $[(0_2/90_4/0_2)_n]_s$ cross-ply laminate for ERR calculations (model 3) of the N-crack, with local delaminations 'D' with a fixed length $a_d = 0.5a_0$ originating at the tips of existing intralaminar cracks.

(120°C) to the room temperature (20°C). In experiments, first the temperature difference is applied leading to free shrinkage of the laminate (ϵ_{th}) and after that the mechanical strain ϵ_{mech} . Therefore, performing FEM with simultaneous mechanical and thermal "load" the value of applied strain must be modified according to equation (3) such that,

$$\epsilon_{appl} = \epsilon_{mech} + \epsilon_{th} \quad (3)$$

where ϵ_{appl} is applied strain ($= U_x/a_0$ for model 1, $= U_x/2a_0$ for models 2 and 3), ϵ_{mech} is mechanical strain ($= 1\%$) and ϵ_{th} is free thermal shrinkage strain (negative).

It should be noted that a recent study in the literature³⁴ has reported that a significant reduction in thermal residual stresses takes place due to relaxation after curing of epoxy-based composite materials. Although this may be important for

accurate prediction of load levels for new transverse cracks, the main task of the present study is to perform parametric analysis revealing how different parameters, such as ply properties, ply thickness and crack density affect the ERR associated with growth of a new transverse crack. The results of the study and the observed trends are not expected to be much different if relaxation of thermal stresses would be taken into account. Hence, a conservative approach has been taken in this study by assuming uniform temperature change causing thermal stresses with no relaxation.

Table 1 lists the unidirectional (UD) ply material properties of GF/EP and CF/EP composites which were used as inputs in FEM models for stress analysis and ERR calculations. Material properties for UD GF/EP plies are taken from¹⁰ and material properties for UD CF/EP plies are taken from.³⁵

Results and discussion

Stress distributions between cracks

Prior to studying intralaminar crack growth using fracture mechanics (ERR analysis in the section “Energy release rate analysis for cracks initiated at 0/90 layer-interface”), the stress state in damaged 90-layers in regions around existing intralaminar cracks was studied for $[(0_2/90_4/0_2)_n]_s$ laminate using model 1 (Figure 1). Figures 4 and 5 show the

distribution of axial (transverse) stress in the 90-layer ($\sigma_x^0 = \sigma_T(x, z)$) along x - and z -axis respectively, and Figure 7 shows the distribution of major principal stress (σ_I) along x -direction close to the ply interface ($z/d = 0.95$) and the orientations in the maximum point of the potentially introduced crack planes with respect to the laminate x -axis (for transverse cracks orientation is 90°). In these plots, stress values on the vertical axis are normalized with respect to far-field transverse stress in the 90-layer. Far-field stress is

Table 1. Unidirectional ply material properties of GF/EP and CF/EP composites. CTE denotes the coefficient of thermal expansion.

Material property	Notation	Unit	GF/EP	CF/EP
Longitudinal modulus	E_L	GPa	44.73	138.00
Transverse modulus	E_T	GPa	12.76	8.10
Longitudinal Poisson's ratio	ν_{LT}	-	0.30	0.25
Transverse Poisson's ratio	ν_{T3}	-	0.42	0.45
In-plane shear modulus	G_{LT}	GPa	5.80	5.00
Out-of-plane shear modulus	G_{T3}	GPa	4.49	2.79
Longitudinal CTE	α_L	$1/^\circ\text{C}$	8.60×10^{-6}	0
Transverse CTE	α_T	$1/^\circ\text{C}$	22.10×10^{-6}	27.00×10^{-6}
Ply thickness	t	mm	0.072	0.072

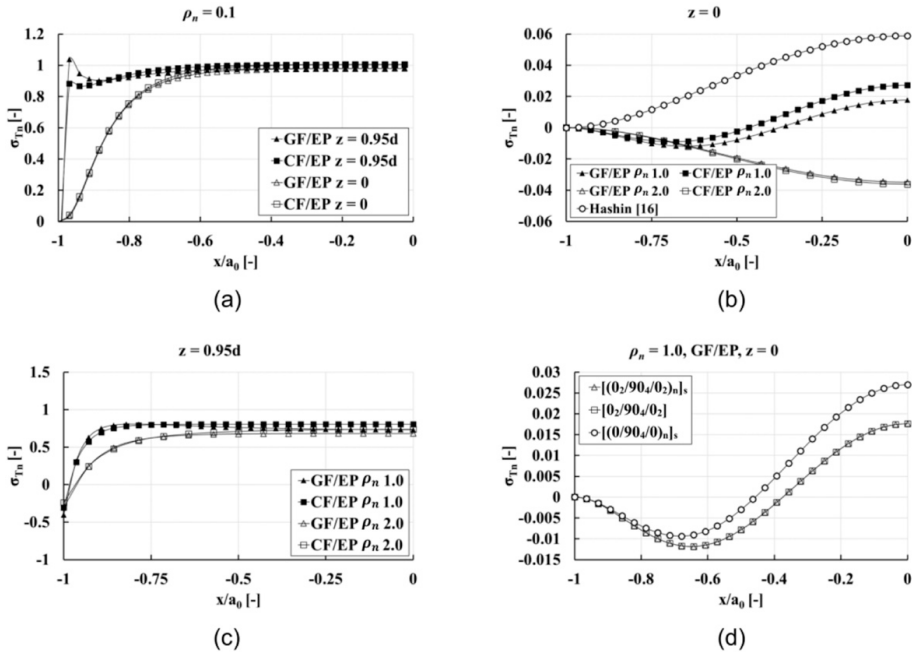


Figure 4. σ_{Tn} distribution along x -axis in damaged 90-layers in cross-ply laminates with different crack densities for GF/EP and CF/EP composites. Plots in (a)-(c) are obtained for $[(0_2/90_4/0_2)_n]_s$ laminate, plots in (d) are obtained for $[(0_2/90_4/0_2)_n]_s$, $[0_2/90_4/0_2]_s$ and $[(0/90_4/0)_n]_s$ laminates. Plot according to Hashin's model¹⁶ in (b) is obtained for GF/EP $[(0_2/90_4/0_2)_n]_s$ laminate with $\rho_n = 1.0$.

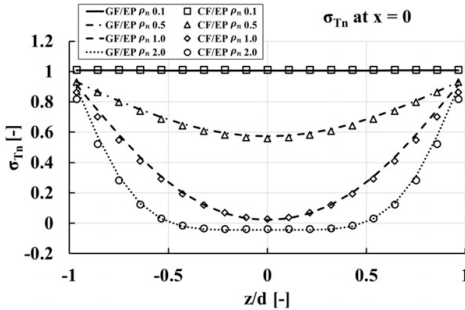


Figure 5. σ_{Tn} distribution in through-the-thickness direction (along z -axis) at $x = 0$ in damaged 90-layers in GF/EP and CF/EP $[(0_2/90_4/0_2)_n]_s$ laminates with different crack densities.

the thermo-mechanical transverse stress σ_{T0} in an undamaged 90-layer, which is calculated by applying 1% strain to an undamaged laminate. The normalized transverse stress σ_{Tn} in a damaged laminate (see Figures 4 and 5) was calculated as,

$$\sigma_{Tn} = \frac{\sigma_T}{\sigma_{T0}} \quad (4)$$

Transverse stress values in the damaged 90-layer plotted along x -axis from $x = -a_0$ to $x = 0$ are shown in Figures 4(a)–(d). At low crack density ($\rho_n = 0.1$), see Figure 4(a), σ_T in the 90-layer reaches the far-field stress value at some distance away from the intralaminar crack, but the x -dependency at midplane $z = 0$ differs from the trend near the 0/90 layer-interface at $z = 0.95d$. Near the intralaminar crack ($x = -a_0$), the x -dependency also differs with materials, σ_{Tn} near interface in GF/EP attaining a higher peak value than in CF/EP. At high crack density ($\rho_n = 1.0, 2.0$), see Figure 4(b), local compressive stresses can be observed at the midplane $z = 0$. σ_{Tn} along the midplane reaches a minimum at some distance away from the intralaminar crack, whereas σ_{Tn} values near the 0/90 layer-interface $z = 0.95d$, see Figure 4(c), increase with distance from the intralaminar crack before reaching a plateau value.

In Figure 4(d), the axial stress distribution at midplane $z = 0$ is compared for GF/EP cross-ply laminates with different boundary conditions at $\rho_n = 1.0$. In this plot, a laminate with layup $[(0_2/90_4/0_2)_n]_s$ is compared with a laminate with layup $[0_2/90_4/0_2]$ and a laminate with layup $[(0/90_4/0)_n]_s$. The stress distribution in the laminate with layup $[(0_2/90_4/0_2)_n]_s$ and the laminate with layup $[0_2/90_4/0_2]$ is almost identical. Although numerical values differ for the laminate with layup $[(0/90_4/0)_n]_s$ compared to the laminate with layup $[(0_2/90_4/0_2)_n]_s$, the stress distribution trend remains the same. In comparison, the transverse stress at the midplane is higher in the $[(0/90_4/0)_n]_s$ laminate with

relatively thinner 0-layer than in the $[(0_2/90_4/0_2)_n]_s$ laminate or stand-alone $[0_2/90_4/0_2]$ laminate.

Distribution of transverse stress in through-the-thickness direction (along z -axis) in the damaged 90-layer of $[(0_2/90_4/0_2)_n]_s$ laminate in the middle between two cracks ($x = 0$) is shown in Figure 5. Rather uniform $\sigma_T(z)$ distribution is observed at low crack density ($\rho_n = 0.1$) in GF/EP and in CF/EP laminates. As the crack density increases, σ_T decreases with the distance from the 0/90 layer-interface reaching minimum on the middle line $z = 0$. The nonuniformity increases with increase in crack density. Similar to Figure 4(b), local compressive stresses at the midplane $z = 0$ can also be observed in Figure 5 at $\rho_n = 2.0$ in GF/EP and in CF/EP, the curves are almost coinciding. Note that at $\rho_n = 2.0$, σ_T is compressive in the region between $z/d \approx -0.5$ and 0.5 , with a plateau region in between.

Figures 4(a)–(d) and Figure 5 prove that at high crack density, the transverse stress σ_T strongly depends on z coordinate and, therefore, the average of the transverse stress over the 90-layer thickness would have rather questionable applicability in predicting the position for next crack. The crack can be initiated at $z/d = 0.95$ and $x = 0$ but is it going to propagate across the 90-layer thickness? Hence, stress distributions obtained using models assuming independence of transverse stress on thickness coordinate (Hashin's model,¹⁶ shear lag models^{5,6}) or averaging over the thickness the transverse stress calculated with FEM cannot be used to find the x -position of the next crack in the so called “strength approaches”.^{5,30} The highest transverse stress is close to the interface (see Figure 5) and there the coalescence of debonds to initiate a crack can be expected.

At low crack densities, the distribution of transverse tensile stresses between two intralaminar cracks in a 90-layer is uniform in through-the-thickness direction. Initiation of a new transverse crack in such situation depends on several parameters at a micro-scale level such as non-uniform fiber distribution,³⁶ defects,³⁷ etc. Previous studies in the literature have shown that the initiation site may also depend on micromechanical parameters such as the fiber-matrix interface strength,³⁸ fiber alignment^{39,40} and the hydrostatic stress in the matrix.⁴¹ At high crack densities, however, stress analysis results of the present study show significantly magnified transverse tensile stresses near the 0/90 layer-interface (see Figure 5) and although micro-scale stress distributions may still be important, the stress analysis allows us to assume with rather high certainty that the initiation site for new transverse crack growth will be near the 0/90 layer-interface. The exact x -position along the interface is impossible to predict because the resistance to failure in different positions is different due to non-uniform fiber distribution. However, based on the performed stress

analysis, two x -positions for a new crack initiation close to the interface can be suggested: (a) close to the intralaminar crack tip where not only the transverse stress is high but also the rest of stress components; (b) in the middle between the two existing intralaminar cracks (because of symmetry a pure Mode I growth of the N-crack, see Figures 2 and 3, is expected in this case and the analysis is simplified). The former case is analyzed calculating principal stresses in the section “Principal stresses close to the existing crack”, the latter case is analyzed calculating ERR in the section “Energy release rate analysis for cracks initiated at 0/90 layer-interface”.

Principal stresses close to the existing crack

The transverse stress in the ply interface region does not have a distinct maximum exactly in the middle between existing cracks and the position of a new initiated crack is not necessarily in that position.

At high crack density and especially in cross-ply laminates with low b/d ratio, a different type of cracks has been observed in the literature close to the tip of the intralaminar crack; delta cracks (Δ)^{42,43} and curved cracks (C)⁴⁴ shown in Figure 6 as an illustration.

Such cracks have been studied, for example, in⁴³ and the location of their initiation and orientation were estimated based on principal stress analysis in the 90-layer. To investigate favorable conditions for position of initiation of these cracks near the intralaminar crack tip and their orientation, principal stress distribution was studied in the present work close to the 0/90 layer-interface, at $z = 0.95d$, and at various intralaminar crack densities using model 1 (Figure 1). The direction perpendicular to the direction of the main principal stress is defined as direction of the potential crack to be initiated and the crack is characterized by the angle between this direction and the laminate x -axis direction (shown as θ in Figure 6, for an ideal transverse crack this angle is 90°). Figure 7 shows normalized values of σ_I (major principal stress) plotted along x -axis at $z = 0.95d$ for GF/EP $[(0_2/90_4/0_2)_n]_s$ and $[(0/90_4/0)_n]_s$ laminates. The origin is located at the laminate midplane at the position of the intralaminar crack (see Figure 6). Maximum values in each of the plots ($\max. \sigma_I$) are highlighted with a dot marker. The location of $\max. \sigma_I$ is in a close vicinity to the existing intralaminar crack, as was observed in.⁴⁴ If we assume that the largest principal stress defines damage initiation, a new delta crack would preferably initiate at the location of $\max. \sigma_I$ and it would be perpendicular to the corresponding principal plane at that location. The inclination angles of the potential cracks to be initiated (θ) are indicated near the corresponding plots in Figure 7. With increase in crack density, the inclination angle slightly increases, whereas with reduction in constraining layer (0-layer)

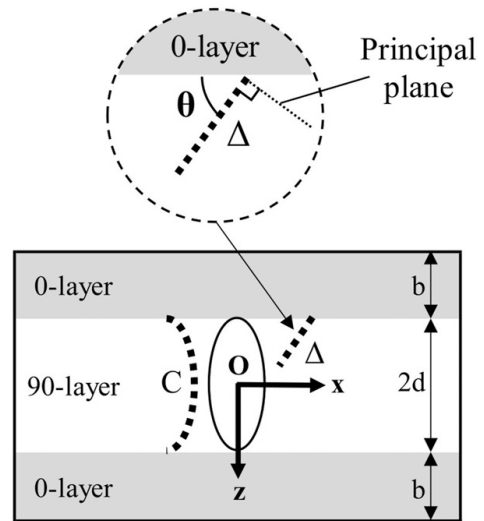


Figure 6. Different types of cracks observed close to the tip of the intralaminar crack – Delta crack (Δ) and curved crack (C). Note that the above damage pattern is shown solely for the purpose of illustration.

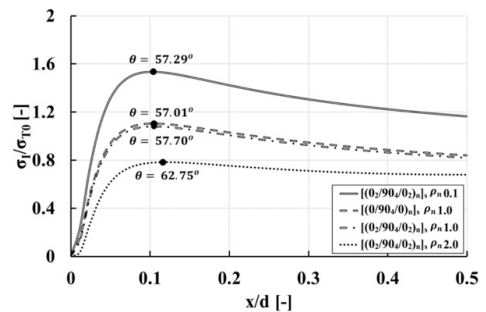


Figure 7. Major principal stress (σ_I) distribution along x -axis at $z = 0.95d$ in damaged 90-layers in GF/EP $[(0_2/90_4/0_2)_n]_s$ and $[(0/90_4/0)_n]_s$ laminates with different crack densities. Origin of the x -axis is at the position of the existing intralaminar crack.

thickness, the inclination angle decreases. It should also be noted that analysis of principal stress distribution further away from the intralaminar crack tip along the 0/90 layer-interface showed that the value of θ reaches 90° at the middle between adjacent intralaminar cracks.

The results presented in Figure 7 should be analyzed with caution. The apparent result that at $\rho_n = 0.1$ the principal stress magnification is higher than at $\rho_n = 1.0$ and the conclusion that, therefore, the delta cracks are more probable at low intralaminar crack density is not correct and

the concept of statistical failure properties distribution along the transverse direction of the layer has to be used. The principal stress in Figure 7 is normalized with respect to the far-field stress in undamaged laminate corresponding to 1% applied strain. In a real test, the crack density increases with the applied load and in the process of reaching $\rho_n = 1.0$, very high load would be applied (first cracks appear in weakest positions). This means that σ_T to reach high crack density is much higher than for reaching $\rho_n = 0.1$. Hence, the “non-normalized” principal stress when $\rho_n = 1.0$ is reached is much higher than for reaching the relatively low crack density $\rho_n = 0.1$. That means higher probability of delta cracks close to existing intralaminar cracks at higher crack density.

The transverse stress at the 0/90 layer-interface further away from the crack (e.g., in the middle between them) is also relatively high at high crack density and so is the probability to initiate a new transverse crack there. However, it is not enough only to initiate a crack - conditions for its growth into through-the-thickness intralaminar crack have to be satisfied as well. This question is addressed in the section “Energy release rate analysis for cracks initiated at 0/90 layer-interface”.

Energy release rate analysis for cracks initiated at 0/90 layer-interface

As shown in the section “Principal stresses close to the existing crack”, the delta cracks emerging from the intralaminar crack tip should have a nonsignificant preference for initiation comparing to positions close to the 0/90 layer-interface but further away from the intralaminar crack tip. Indeed, Figure 7 shows that the principal stress at the interface away from the intralaminar crack tip is only slightly lower than the maximum near the crack tip region and crack initiation is possible even in the middle of the repeating element close to the interface. In that position, the out-of-plane shear stress is zero and $\sigma_T > \sigma_z$. Hence, $\sigma_I = \sigma_T$ and θ is 90° with respect to the midplane. Thus, the introduced N-crack (see Figures 2 and 3) is at 90° with respect to the midplane (the N-crack is a small transverse crack). The remaining questions: is it going to grow and how is it going to grow?

ERR for a N-crack, initiated at the upper 0/90 layer-interface $z = -d$ in the middle between two existing intralaminar cracks, see model 2 in Figure 2, growing in through-the-thickness direction (along the z -axis) in the 90-layer towards the lower 0/90 layer-interface $z = d$ is shown in Figure 8 for the case of $[(0_2/90_4/0_2)_n]_s$ laminate. Displacement U_x leading to 1% of average tensile strain and a uniform temperature change of $\Delta T = -100^\circ\text{C}$ were applied as the loading. The ERR values reported in the sections “Energy release rate analysis for cracks initiated at 0/90

layer-interface” and “Effect of local delaminations on ERR for new crack growth” were calculated using ANSYS²⁹ software built-in J -integral calculation routine, which in plane strain conditions gives total mode-I + mode-II ERR values. However, prior to performing J -integral calculations, mode-mixity was analyzed for all the studied cases using ANSYS²⁹ built-in virtual crack closure technique (VCCT) calculation routine and it was observed that mode-I crack growth was dominating over mode-II for all the cases reported in these sections.

In the low crack density region ($\rho_n = 0.1$), see Figure 8(a), the ERR is proportional to the N-crack length a_N over major part of the 90-layer thickness, similarly as it is for a single crack in infinite medium. At N-crack length of about 75%–80% of the 90-layer thickness, the ERR increase slows down, reaches maximum and then decreases approaching finite value at the lower 0/90 layer-interface. This phenomenon, which is caused by the constraint effect on crack opening from the stiffer adjacent 0-layer has been studied by Blázquez et al.⁴⁵ The ERR values for the CF/EP laminate are consistently lower than for GF/EP. That can be explained by a lower transverse modulus leading to the lower nominal transverse stress in the 90-layer at the applied 1% strain. From Figure 8(a) follows that, if the ERR for a short N-crack has been sufficient for starting crack growth, its growth will be unstable almost over the whole ply thickness. The initial N-crack length was assumed to be 0.05 times the half 90-layer thickness that for $d = 0.25 - 0.5$ mm is about 1–4 fiber diameters (for typical size of carbon and glass fibers). The interface will be reached if the final value of ERR when the N-crack tip is close to the lower interface is higher than ERR for the N-crack length when the crack started to grow due to ERR reaching the critical value. If not, the N-crack will be “arrested” close to the interface, and it can approach the interface only with increasing applied strain. Similar behavior can be observed in Figure 8(b), for $\rho_n = 0.5$. The initial values of ERR in this case are lower but rather similar as for $\rho_n = 0.1$ and the initial increase of ERR with respect to crack length is linear. However, the behavior changes when N-crack length reaches $a_N > 0.25d$ because at $\rho_n \geq 0.5$, the stress perturbations from the existing intralaminar cracks make the stress distribution in the middle between cracks z -dependent (as shown in Figure 5). The transverse stress slightly reduces approaching middle line $z = 0$, slowing down the ERR increase. After crossing the middle line $z = 0$, the ERR increases faster because the N-crack tip is back in the region with high transverse stress (as shown in Figure 5). The final stage is as in the $\rho_n = 0.1$ scenario: after reaching maximum the ERR decreases due to constraint from the 0-layer.

Considering the case with $\rho_n = 1.0$, see Figure 8(c), the effect on ERR of the transverse stress in the middle part of the 90-layer being much lower than at the interface, see

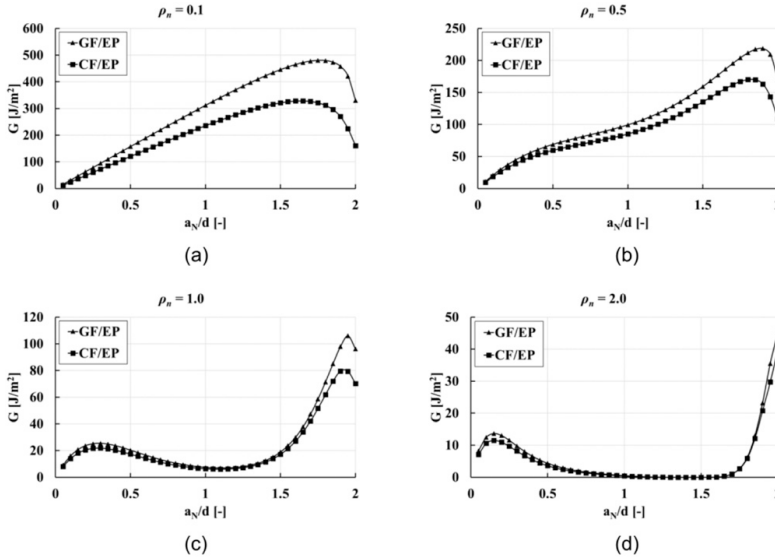


Figure 8. ERR plots for new transverse crack growth in GF/EP and CF/EP $[(0_2/90_4/0_2)_n]_s$ laminates with (a) $\rho_n = 0.1$, (b) $\rho_n = 0.5$, (c) $\rho_n = 1.0$ and (d) $\rho_n = 2.0$.

Figure 5, is much more pronounced. The increase of the ERR stops at N-crack length $a_N \approx 0.2d$ and ERR reaches a minimum at $a_N \approx d$. After that, the N-crack tip reaches a higher stress region resulting in an increase in ERR. The N-crack would start to grow unstably if ERR is above the critical value, but the growth would stop at some distance from the middle line $z = 0$ where the ERR goes below the critical value. Hence, additional strain should be applied to extend the crack further. A peculiarity arises for the case with $\rho_n = 2.0$, see Figure 8(d), as the transverse stress is compressive close to the middle line $z = 0$, see Figure 5, and the ERR goes to zero at $a_N = d$. At this location, N-crack growth would completely stop, and propagation is impossible even with increasing applied strain. This phenomenon is the reason for sometimes observed partial cracks. Due to this, it is not clear if these cracks would grow as tunnel-cracks through the width of the specimen. The ERR for propagation along fibers depends on the dimension of the crack in the thickness direction.

Effect of local delaminations on ERR for new crack growth

In this section, we investigate the effect of local delaminations at the existing intralaminar crack tips on ERR for the N-crack propagation in the middle between them as shown in model 3 in Figure 3. It has been stated in the literature⁴⁶ that formation

of local delaminations at the intralaminar crack tips is possible for glass-fiber epoxy laminates with the 90-layer thickness which is used in this FEM study that is, 0.288 mm. The parametric study reported in this section is conducted primarily to show that potential occurrence of local delaminations at existing intralaminar crack tips significantly changes ERR available for the growth of a new transverse crack.

GF/EP and CF/EP $[(0_2/90_4/0_2)_n]_s$ laminates with $\rho_n = 0.5$ and 1.0, and with fixed delamination length $a_{dl} = \frac{2a_d}{2a_0} = 0.5$ are analyzed. This means that 50% of the 0/90 layer-interface is delaminated. The ERR dependence on crack length for the cases with and without delaminations is shown in Figure 9, showing very strong effect of the introduced relatively long delaminations. Delaminations significantly reduce the available ERR for new crack growth. The trends are similar for medium crack density ($\rho_n = 0.5$) and high crack density ($\rho_n = 1.0$), although the effect of delaminations on ERR values is reduced for the case with high crack density.

In^{8,47}, analyzing the stress state with analytical models in a case with local delaminations, assumption was made that the stress in the delaminated part of the 90-layer ($-a_0 < x < -a_0 + a_d$ and $a_0 - a_d < x < a_0$, see Figure 3) is zero. Then, only the region $-a_0 + a_d < x < a_0 - a_d$ has to be analyzed, with zero stress boundary conditions applied at $x = \pm(a_0 - a_d)$ in the 90-layer. In the analyzed case, the problem becomes identical to a problem without delaminations with “effective” normalized crack density ρ_{neff} . Thus, for the case with $\rho_n = 0.5$,

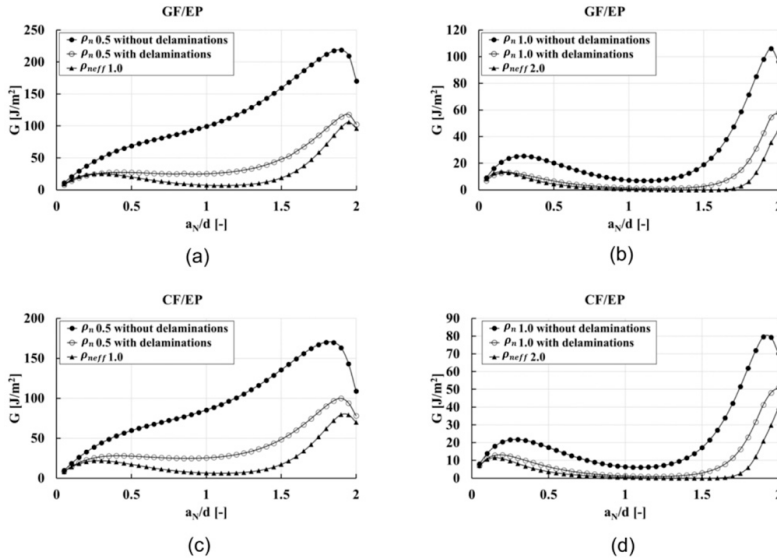


Figure 9. ERR trend comparison for N-crack growth in (a) GF/EP at $\rho_n = 0.5$, (b) GF/EP at $\rho_n = 1.0$, (c) CF/EP at $\rho_n = 0.5$ and (d) CF/EP at $\rho_n = 1.0$ for $[(0_2/90_4/0_2)_n]_s$ laminates with or without delaminations. $\rho_{neff} = 1.0$ and 2.0 for models without delaminations and with effective length between existing intralaminar cracks equal to the bonded length.

$$\rho_{neff} = \frac{d}{a_0 - a_d} = \frac{d}{(1 - 0.5)a_0} = \frac{0.5}{0.5} = 1.0 \quad (5)$$

ERR calculated under this assumption is also shown in Figure 9. Comparing ERR trends for $\rho_n = 0.5$ or 1.0 with delaminations and ERR calculations using $\rho_{neff} = 1.0$ or 2.0, it can be observed that using the simple model without delaminations and with effective length equal to the bonded length is not accurate. For both materials, ERR is too low using the model with “effective” distance between pre-existing cracks and zero delaminations.

Conclusions

In the present work, initiation and growth of new cracks through the thickness of highly damaged 90-layers of glass-fiber epoxy and carbon-fiber epoxy cross-ply composites is studied using finite element modeling. Transverse stress and principal stresses in the vicinity of existing intralaminar cracks are studied. For the cases with low crack density, there is a large plateau region where transverse stress is uniform along the 90-layer thickness. On the other hand, at high crack density, large stress gradient is observed in through-the-thickness direction and even local compressive stresses are seen at $[(0_2/90_4/0_2)_n]_s$ cross-ply laminate midplane. The observed gradient in transverse stress affects growth of a new transverse crack

(N-crack), initiated at the 0/90 layer-interface and growing inside the 90-layer in through-the-thickness direction. For a short N-crack compared to the 90-layer thickness, available energy release rate (ERR) values calculated at the crack tip are proportional to the crack length. At high crack density, ERR values decrease near the $[(0_2/90_4/0_2)_n]_s$ laminate midplane and in the presence of local compressive stresses, ERR becomes equal to zero, stopping the crack growth completely.

ERR values at the tip of the N-crack are lower in the presence of local delaminations initiated at the tips of existing intralaminar cracks. However, the reduction in ERR values is smaller than for the commonly used models assuming zero stress in the delaminated parts, effectively analyzing a repeating element with effective length equal to the bonded length.

Possibility of initiation of additional (competing) damage mode in the form of delta cracks close to the intralaminar crack tip is analyzed, and the potential initiation points and orientation of these cracks are found using major principal stress and orientation of principal axes.

Author contributions

Anish Niranjana Kulkarni: Investigation, Writing – Original draft, Data curation. Andrejs Pupurs: Writing – Review & Editing, Supervision, Methodology. Janis Varna: Conceptualization, Methodology, Writing – Review & Editing.

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Data Availability Statement

The datasets generated during and/or analyzed during the current study are available from the corresponding author on reasonable request.

References

- Nairn J and Hu S. Matrix microcracking. In: Pipes RB and Talreja R (eds) *Damage Mechanics of Composite Materials. Composite Materials series*. Amsterdam: Elsevier, 1994, pp. 187–243.
- Nairn JA. Matrix microcracking in composites. In: Kelly A and Zweben C (eds) *Comprehensive Composite Materials*. Amsterdam: Elsevier, 2000. doi:10.1016/b0-08-042993-9/00069-3.
- Jamison RD, Schulte K, Reifsnider KL, et al. Characterization and analysis of damage mechanisms in tension-tension fatigue of graphite/epoxy laminates. In: Wilkins DJ (ed) *Effects of Defects in Composite Materials (ASTM Special Technical Publication 836)*. Philadelphia: American Society for Testing and Materials, 1984.
- Garrett KW and Bailey JE. Multiple transverse fracture in 90° cross-ply laminates of a glass fibre-reinforced polyester. *J Mater Sci* 1977; 12: 157–168.
- Peters PWM and Chou TW. On cross-ply cracking in glass and carbon fibre-reinforced epoxy laminates. *Composites*. 1987; 18(1): 40–46. doi:10.1016/0010-4361(87)90006-1.
- Lim SG and Hong CS. Prediction of transverse cracking and stiffness reduction in cross-ply laminated composites. *J Compos Mater* 1989; 23: 695–713.
- Kashtalyan M and Soutis C. Analysis of composite laminates with intra- and interlaminar damage. *Prog Aero Sci*. 2005; 41(2): 152–173. doi:10.1016/j.paerosci.2005.03.004.
- Takeda N and Ogihara S. Initiation and growth of delamination from the tips of transverse cracks in CFRP cross-ply laminates. *Compos Sci Technol* 1994; 52: 309–318.
- Takeda N, Ogihara S and Kobayashi A. Microscopic fatigue damage progress in CFRP cross-ply laminates. *Composites*. 1995; 26(12): 859–897. doi:10.1016/0010-4361(95)90879-5.
- Pakkam Gabriel VR, Loukil MS and Varna J. Analysis of intralaminar cracking in 90-ply of GF/EP laminates with distributed ply strength. *J Compos Mater*. 2021; 55(26): 002199832110273. doi:10.1177/00219983211027346.
- Berthelot JM. Transverse cracking and delamination in cross-ply glass-fiber and carbon-fiber reinforced plastic laminates: static and fatigue loading. *Appl Mech Rev*. 2003; 56(1): 111–147. doi:10.1115/1.1519557.
- Altus E and Ishai O. Transverse cracking and delamination interaction in the failure process of composite laminates. *Compos Sci Technol* 1986; 26: 59–77.
- Varna J, and Asp L. Microdamage in composite laminates: experiments and observation. *AMM*. 2014; 518:84–89, 2014. doi:10.4028/www.scientific.net/AMM.518.84.
- Joffe R, Varna J and Berglund LA. Acoustic emission signals from composite laminates with well known fatigue damage characteristics. *AECM-5: 5th International Symposium on Acoustic Emission From Composite Material, Sundsvall*. Columbus: American Society for Nondestructive Testing, 1995, pp. 179–186.
- Zhang J, Fan J and Soutis C. Analysis of multiple matrix cracking in [± 0 m/90n]s composite laminates. Part 1: in-plane stiffness properties. *Composites*. 1992; 23(5): 291–298. doi:10.1016/0010-4361(92)90327-Q.
- Hashin Z. Analysis of cracked laminates: a variational approach. *Mech Mater Compos*. 1985; 4(2): 121–136. doi:10.1016/0167-6636(85)90011-0.
- Varna J and Berglund L. Multiple transverse cracking and stiffness reduction in cross-ply laminates. *J Compos Technol Res*. 1991; 13(2): 97–106. doi:10.1520/ctr10213j.
- Varna J and Berglund LA. Thermo-elastic properties of composite laminates with transverse cracks. *J Compos Technol Res*. 1994; 16(1): 11. doi:10.1520/ctr10397j.
- Varna J. Strategies for stiffness analysis of laminates with microdamage: combining average stress and crack face displacement based methods. *ZAMM Zeitschrift für Angewandte Mathematik und Mechanik*. 2015; 95(10): 1081–1097. doi:10.1002/zamm.201400296.
- Smith PA and Wood JR. Poisson's ratio as a damage parameter in the static tensile loading of simple crossply laminates. *Compos Sci Technol*. 1990; 38(1): 85–93. doi:10.1016/0266-3538(90)90073-E.
- Tsai CL and Daniel IM. The behavior of cracked cross-ply composite laminates under shear loading. *Int J Solid Struct*. 1992; 29(24): 3251–3267. doi:10.1016/0020-7683(92)90039-V.
- Tan SC and Nuismer RJ. A theory for progressive matrix cracking in composite laminates. *J Compos Mater*. 1989; 23(10): 1029–1047. doi:10.1177/002199838902301006.
- Vinogradov V and Hashin Z. Variational analysis of cracked angle-ply laminates. *Compos Sci Technol*. 2010; 70(4): 638–646. doi:10.1016/j.compscitech.2009.12.018.
- McCartney LN, Schoeppner GA and Becker W. Comparison of models for transverse ply cracks in composite laminates. *Compos Sci Technol*. 2000; 60(12–13): 2347–2359. doi:10.1016/S0266-3538(00)00030-0.

25. McCartney LN and Schoeppner GA. Predicting the effect of non-uniform ply cracking on the thermoelastic properties of cross-ply laminates. *Compos Sci Technol*. 2002; 62(14): 1841–1856. doi:[10.1016/S0266-3538\(02\)00091-X](https://doi.org/10.1016/S0266-3538(02)00091-X).
26. Hajikazemi M and McCartney LN. Comparison of Variational and Generalized Plane Strain approaches for matrix cracking in general symmetric laminates. *Int J Damage Mech*. 2018; 27(4): 507–540. doi:[10.1177/1056789516685381](https://doi.org/10.1177/1056789516685381).
27. Hajikazemi M and Sadr MH. A variational model for stress analysis in cracked laminates with arbitrary symmetric lay-up under general in-plane loading. *Int J Solid Struct*. 2014; 51(2): 516–529. doi:[10.1016/j.ijsolstr.2013.10.024](https://doi.org/10.1016/j.ijsolstr.2013.10.024).
28. Hajikazemi M, Sadr MH, Hosseini-Toudeshky H, et al. Thermo-elastic constants of cracked symmetric laminates: a refined variational approach. *Int J Mech Sci*. 2014; 89: 47–57. doi:[10.1016/j.ijsmecsci.2014.08.008](https://doi.org/10.1016/j.ijsmecsci.2014.08.008).
29. ANSYS Release 22.0. Canonsburg, PA, ANSYS Academic Research, ANSYS Inc. 2022.
30. Manders PW, Chou TW, Jones FR, et al. Statistical analysis of multiple fracture in 0°/90°/0° glass fibre/epoxy resin laminates. *J Mater Sci* 1983; 18: 2876–2889.
31. Wang ASD. Fracture mechanics of sublaminar cracks in composite materials. *Compos Technol Rev*. 1984; 6(2): 18. doi:[10.1520/ctr10817j](https://doi.org/10.1520/ctr10817j).
32. Zhu XK and Joyce JA. Review of fracture toughness (G, K, J, CTOD, CTOA) testing and standardization. *Eng Fract Mech*. 2012; 85: 1–46. doi:[10.1016/j.engfractmech.2012.02.001](https://doi.org/10.1016/j.engfractmech.2012.02.001).
33. Dvorak GJ and Laws N. Analysis of progressive matrix cracking in composite laminates II. First ply failure. *J Compos Mater*. 1987; 21(4): 002199838702100402. doi:[10.1177/002199838702100402](https://doi.org/10.1177/002199838702100402).
34. Bao R, Liu J, Xiao Z, et al. Thermo-chemo-mechanical modeling of residual stress in unidirectional carbon fiber-reinforced polymers during manufacture. *Materials*. 2024; 17(12): 3040. doi:[10.3390/ma17123040](https://doi.org/10.3390/ma17123040).
35. Pupurs A, Loukil MS, Marklund E, et al. Transverse crack initiation in thin-ply laminates subjected to tensile loading at low and cryogenic temperatures. *Mech Compos Mater*. 2024; 59(6): 1–16. doi:[10.1007/s1029-023-10156-0](https://doi.org/10.1007/s1029-023-10156-0).
36. Ben Kahla H, Ayadi Z, Edgren F, et al. Statistical model for initiation governed intralaminar cracking in composite laminates during tensile quasi-static and cyclic tests. *Int J Fatig* 2018; 116: 1–12.
37. Berthelot JM and Le Corre JF. Statistical analysis of the progression of transverse cracking and delamination in cross-ply laminates. *Compos Sci Technol* 2000; 60: 2659–2669.
38. Gamstedt EK and Sjögren BA. Micromechanisms in tension-compression fatigue of composite laminates containing transverse plies. *Compos Sci Technol* 1999; 59: 167–178.
39. Hojo M, Mizuno M, Hobbiebrunken T, et al. Effect of fiber array irregularities on microscopic interfacial normal stress states of transversely loaded UD-CFRP from viewpoint of failure initiation. *Compos Sci Technol* 2009; 69: 1726–1734.
40. Carraro PA and Quaresimin M. A damage based model for crack initiation in unidirectional composites under multiaxial cyclic loading. *Compos Sci Technol* 2014; 99: 154–163.
41. Asp LE, Berglund LA and Talreja R. Prediction of matrix-initiated transverse failure in polymer composites. *Compos Sci Technol* 1996; 56: 1089–1097.
42. Parvizi A and Bailey JE. On multiple transverse cracking in glass fibre epoxy cross-ply laminates. *J Mater Sci*. 1978; 13: 2131–2136. doi:[10.1007/BF00541666](https://doi.org/10.1007/BF00541666).
43. Groves SE, Harris CE, Highsmith AL, et al. An experimental and analytical treatment of matrix cracking in cross-ply laminates. *Exp Mech*. 1987; 27: 73–79. doi:[10.1007/BF02318867](https://doi.org/10.1007/BF02318867).
44. Hu S, Bark JS and Nairn JA. On the phenomenon of curved microcracks in [(S)/90n]s laminates: their shapes, initiation angles and locations. *Compos Sci Technol* 1993; 47: 321–329.
45. Blázquez A, Mantić V, París F, et al. Stress state characterization of delamination cracks in [0/90] symmetric laminates by BEM. *Int J Solid Struct* 2008; 45: 1632–1662.
46. Joffe R, Krasnikovs A and Varna J. COD-based simulation of transverse cracking and stiffness reduction in [S/90n]s laminates. *Compos Sci Technol* 2001; 61: 637–656.
47. O'Brien TK. Analysis of local delaminations and their influence on composite laminate behavior. In: Johnson WS (ed) *Delamination and Debonding of Materials (ASTM Special Technical Publication 876)*. Philadelphia: American Society for Testing and Materials, 1985.



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journal homepage: www.elsevier.com/locate/engfracmechParameters affecting growth of local delaminations at transverse crack tips in $[0_m, 90_n]_s$ cross-ply laminatesAnish Niranjana Kulkarni^{a,*}, Andrejs Pupurs^{a,*}, Janis Varna^{a,b}^a Institute of High-Performance Materials and Structures, Riga Technical University, LV – 1048 Riga, Latvia^b Department of Engineering Sciences and Mathematics, Luleå University of Technology, SE – 97187 Luleå, Sweden

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ABSTRACT

Transverse cracks in the 90-layers in cross-ply laminates have singular stress state at the crack tips. This causes formation of fiber/matrix debonds which coalesce into a local delamination along the 0/90-layer interface. Various studies in the literature have predicted onset strains for local delaminations at transverse crack tips using energy-based criteria such as critical strain energy release rate (ERR) and critical generalized stress intensity factors. Although similar ERR-based analyses have been carried out to predict the delamination growth as well, a systematic parametric analysis is lacking. Such systematic analysis of parameters that can affect the growth of local delaminations including geometrical parameters, elastic constants and transverse crack density is necessary to predict delamination growth under complex thermo-mechanical loading conditions. In the present work, FEM is used to carry out ERR-based analysis of the growth of local delaminations with different shapes in carbon-fiber epoxy and glass-fiber epoxy $[0_m, 90_n]_s$ cross-ply laminates with the help of virtual crack closure technique and J-integral method. Firstly, the ratios of elastic constants and geometrical parameters that can prominently affect the ERR values are identified by a simple analytical routine. Then, the analytical predictions are verified using FEM for local delaminations growing symmetrically (I-shaped and C-shaped) or non-symmetrically (T-shaped and S-shaped) with respect to the laminate midplane. It is shown that a non-symmetrical I-shaped crack would always transition into a symmetrical I-shaped crack before any further delamination growth, if energetically viable. Finally, a simplified strategy to calculate ERR values for local delaminations growing under combined thermo-mechanical loading is presented.

1. Introduction

Under loading, multiple damage mechanisms can be observed in $[0_m, 90_n]_s$ cross-ply laminates such as i) transverse cracking in 90-layers, ii) ply splitting in 0-layers, iii) edge delaminations along 0/90-layer interfaces, iv) local delaminations along 0/90-layer interfaces starting from tips of 90-layer transverse cracks or 0-layer splits, and v) fiber breakage in 0-layers. Among the different damage mechanisms listed here, transverse cracking in 90-layers is usually the first to appear. At the location of transverse crack, 90-layer does not carry any load, but load is transferred back to the 90-layer over the interface by out-of-plane shear stresses [1]. Thus, new transverse cracks may appear and the crack density in the 90-layer increases. Many studies in the literature have calculated stress state

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Nomenclature

Symbol

$[0_m, 90_n]_s$	Symmetric $[0, 90]$ cross-ply laminate with m 0-layers and n 90-layers
C	Intralaminar (transverse) crack
DELAM i	i^{th} Delamination crack ($i = 1$ or 2)
G_c	Critical energy release rate
G_T	Total energy release rate
G_{Tn}	Normalized total energy release rate
G_0	Normalization parameter for total energy release rate
G_I	Mode-I energy release rate
G_{II}	Mode-II energy release rate
t_0	0-layer thickness
t_{90}	90-layer thickness
l_c	Half distance between two intralaminar cracks
l_d	Length of delamination crack
l_{di}	Length of i^{th} delamination crack ($i = 1$ or 2)
l_{dn}	Normalized delamination crack length
ρ	Transverse crack density
ρ_n	Normalized transverse crack density
U_x	Applied axial displacement along x-direction
U_z	Axial displacement along z-direction
ε_x	Average applied mechanical strain along x-direction
ε_{ij}	All strain components in the 0- and in the 90-layer
σ_{ij}	All stress components in the 0- and in the 90-layer
σ_x^{90}	Axial (transverse) stress in damaged 90-layer
$\sigma_x^{90(CLT)}$	Axial (transverse) stress in undamaged 90-layer
$\sigma_{xmech}^{90(CLT)}$	Axial (transverse) stress in undamaged 90-layer under mechanical loading
$\sigma_{xth}^{90(CLT)}$	Axial (transverse) stress in undamaged 90-layer under thermal loading
E_L	Longitudinal modulus
E_T	Transverse modulus
E_i	All elastic moduli in the 0- and in the 90-layer ($i = 1, 2$ or 3 , where 1 corresponds to x-direction i.e., along fiber axis, 2 corresponds to y-direction i.e., transverse to fiber axis and 3 corresponds to z-direction i.e., along ply thickness)
E_i^0	0-layer elastic modulus
E_i^{90}	90-layer elastic modulus
E_L^0	0-layer longitudinal modulus
E_T^0	0-layer transverse modulus
E_L^{90}	90-layer longitudinal modulus
E_T^{90}	90-layer transverse modulus
Q_{22}	Component of stiffness matrix $[Q]$
G_{LT}	In-plane shear modulus
G_{T3}	Out-of-plane shear modulus
G_{ij}^0	0-layer shear modulus ($i, j = 1, 2$ or 3)
G_{ij}^{90}	90-layer shear modulus
G_{LT}^0	0-layer in-plane shear modulus
G_{T3}^0	0-layer out-of-plane shear modulus
G_{LT}^{90}	90-layer in-plane shear modulus
G_{T3}^{90}	90-layer out-of-plane shear modulus
ν_{LT}	Longitudinal Poisson's ratio
ν_{TL}	Longitudinal Poisson's ratio $\left(\frac{\nu_{LT}}{E_L} = \frac{\nu_{TL}}{E_T}\right)$
ν_{T3}	Transverse Poisson's ratio
ν_{ij}^0	0-layer Poisson's ratio ($i, j = 1, 2$ or 3)
ν_{ij}^{90}	90-layer Poisson's ratio
α_L	Longitudinal coefficient of thermal expansion
α_T	Transverse coefficient of thermal expansion
$f(x, y, z)$	Stress perturbation function

g_{ij}, φ, ϕ	Dimensionless functions
ϕ_{mech}	Dimensionless function under mechanical loading
ϕ_{th}	Dimensionless function under thermal loading
k_1	Laminate constant
ΔT	Change in temperature
$\bar{x}, \bar{y}, \bar{z}$	x, y, z coordinates normalized with respect to t_{90}
BEM	Boundary Element Method
σ_{TBEM}^{90}	90-layer transverse stress calculated using BEM
$\sigma_{Tpl.str}^{90}$	90-layer transverse stress under plane strain material behavior
ERR_{BEM}	ERR value calculated using BEM
$ERR_{pl.str}$	ERR value under plane strain material behavior

in the damaged 90-layers to predict i) further cracking in the same or adjacent layers of the laminate, and ii) laminate stiffness reduction [2–16]. Berthelot has reviewed different approaches to calculate stress distribution and crack evolution in cross-ply laminates under static and fatigue loading [1]. Unless any other critical damage mechanisms appear in the laminate, transverse crack density progressively reaches a saturation level [1,8,17,18].

Local stress concentration at the tips of transverse cracks can induce delaminations along 0/90-layer interfaces. Unlike delaminations at free edges which appear due to mismatch in Poisson contraction of adjacent layers, local delaminations in the bulk region appear due to stress concentration at the transverse crack tips [19]. In linear elasticity, transverse crack approaching 0/90-layer interface has a singular stress state at the crack tip [20,21]. Blázquez and co-workers [18,20,22] used boundary element method (BEM) to characterize the singular stress state at transverse and delamination crack tips. Two scenarios for onset of delaminations were compared: i) transverse crack has not yet reached the interface (Cook-Gordon mechanism [23]) and ii) transverse crack has reached the interface. It was found that the ERR values for the first case are much lower than for the second case, thus even if delamination-like micro-damage appears at the interface before transverse crack reaches the interface [22], a further increase of load is required for the delamination growth [18]. Kim et al. [24] used Stroh formulation for anisotropic elastic materials and method of eigenfunction expansion to calculate the singular stress state in [0,90]_s and [90,0]_s cross-ply laminates with cracks in 90-layers and delaminations. They predicted Mode-II growth and calculated the stress intensity factors at delamination crack tip depending on the delamination length. Since in the limit, stress does not have a square root singularity, they concluded that the ERR for initiation is zero. This obvious contradiction with test data was resolved by Carraro et al. [21] who calculated local singular stresses and introduced generalized stress intensity factors (GSIF). The stress calculated using their approach were validated with FEM. It was found that a) larger crack density leads to lower GSIF; b) Larger 90-layer/0-layer thickness ratio leads to larger normalized GSIF; c) GSIF in Mode-I (the most important) is larger for glass-fiber (GF) than for carbon-fiber (CF) composites.

Analytical solutions for delamination onset strain are based on ERR calculated using approximate stress models. Because of its simplicity and wide application as a first estimate, the O'Brien's [19] ERR expression for initiation of delaminations and growth must be mentioned. It is based on the change in laminate compliance. The ERR in [19] does not depend on the delamination length and does not account for Mode mixity change. Nevertheless, it has proven useful and, among others, the edge delaminations [25], and the onset strains for local delaminations in [$\pm 25,90$]_s laminates agree well with experimental data [26]. This approach was generalized to include a general in-plane loading using generalized plane strain conditions and analyzing unbalanced laminates in [27] assuming that delaminations are growing in a self-similar manner. As validated with FEM, this assumption is rather accurate for low crack density and long delaminations where the transverse crack and the delamination crack tips are not interacting. Further generalization for unsymmetric laminates including the bending stiffness and the coupling stiffness matrices in the analysis was presented in [28] where also the effective stiffness concept for damaged layers was used to analyze multiple delaminations. Finally, the differences in ERR for delamination of surface and internal layers was demonstrated in [29] showing that the delamination growth for surface layers with staggered cracks in cross-ply laminates happens at significantly lower axial load.

A different definition of the onset of local delaminations was introduced in [30–32]. According to their approximate stress models,

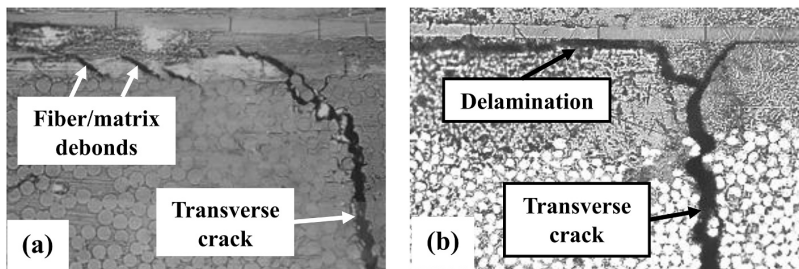


Fig. 1. Evolution of delaminations: (a) initiation of cracks in the resin caused by normal and shear stress combination; (b) well-defined delamination to be treated by LEFM [35].

the ERR for delamination initiation is rapidly decreasing if the distance between cracks increases, whereas the ERR for transverse cracking increases. The crossing point of both ERR curves determines the crack density when the transverse cracking stops, and the delamination growth becomes the only mechanism. Using this approach, Nairn and Hu [30] extended Hashin's variational analysis to include local delaminations and predicted values of transverse crack density at which onset of local delamination would occur. Vilanueva and Kassapoglou [31] studied onset strains for different damage patterns using the elastic compliance change method given in [33] and found that for same transverse crack density, I-shaped damage pattern (refer to Fig. 2) has the lowest delamination onset strain, while C-shaped damage pattern (Fig. 3) has the highest delamination onset strain. The value of onset strain for S-shaped damage pattern (Fig. 3) is only slightly higher than that of the I-shaped damage pattern. The same approach based on competition between the two damage mechanisms was used by Lim and Li [32] who showed that there is no crossing point using their complementary energy-based stress analysis resulting in delamination length independent ERR. They also pointed that the analysis solely based on total ERR values is erroneous because the mode-mixity is different in transverse cracking and in delamination leading to different G_c .

Wang et al. [34] defined the delamination onset strain from a rather voluntary assumption that the initiation is finished when the delamination has reached length equal to the ply thickness. The corresponding ERR value in their 3D FEM model is used to predict the onset strain using known G_c . The onset strain for local delaminations at the tips of transverse cracks starting from i) laminate free edges or ii) crossing points of 0-layer splits and transverse cracks based on predictions was in good agreement with the experiments on cross-ply laminates with different 90-layer thicknesses.

All the above studies have contributed to a better understanding of parameters affecting the initiation of delamination. However, the reality is more complex than the models assuming short delaminations of ideal geometry: Paris and co-authors describe in [18,20,22] fiber/matrix debonds in 90- and 0-layers before a well-defined delamination is formed whose propagation can be analyzed using ERR. As observed by a co-author of the present work, the coalescence of debonds and/or cracks in the resin shown in Fig. 1a finally leads to the initial size of damage that can be considered as a delamination crack, Fig. 1b. Thus, a proper description of the damage initiation most likely has to be done on the fiber/matrix scale, and additional sophisticated observations are necessary. Analysis of the initiation is out of the scope of the current paper.

Many authors have studied the growth of delaminations using ERR but with different methods for stress determination. Takeda and Ogihara [36] and Zhang et al. [37] used shear-lag approach. The former [36] compared their results with experiments on delamination growth, whereas the latter [37] compared their results with FEM virtual crack closure technique (VCCT). Both achieved a good agreement. Analyzing the ERR dependence on transverse crack density in $[0,\theta]_s$ angle-ply laminates, Kashtalyan and Soutis [38–40] used the equivalent constraint model (considering crack with delamination in one layer and the rest of damaged layers are replaced by equivalent homogeneous layers with reduced stiffness properties). In the studies [39–41], effects of different ply angles in angle-ply laminates on delamination onset and growth was studied but they are all restricted for a specific material (either carbon-fiber epoxy composite, or glass-fiber epoxy composite).

Another numerical approach includes extended FEM (XFEM), which was used by Zhao et al. [42] for the first time to simulate growth of delaminations in composite laminates and later used by Curiel-Sosa [43] to calculate ERR for local delaminations and compare with the BEM method. Results calculated with the two approaches were in good agreement.

In most of the literature cited here including [44], an equal length of all delaminations is assumed in ERR calculations (so called I-shaped damage pattern). Non-symmetrical configurations of delaminations with different shapes (Figs. 2–3) were analyzed in Noh and Whitcomb [41]. Different shapes have been experimentally observed by Carraro et al. [44] in symmetric cross-ply laminate layups under fatigue loading. Noh and Whitcomb [41] compared stiffness reduction in damaged plies with different damage patterns finding that, in cross-ply laminates, I-shaped damage pattern results in the highest crack opening among all the damage patterns which directly relates to larger stiffness degradation in the laminate. Carraro and co-authors [44,45] modelled growth of I-shaped delaminations at various crack densities under fatigue loading using FEM.

It appears that conclusions regarding delamination ERR based on approximate models are non-decisive whereas the number of numerical studies is rather limited and a systematic parametric analysis is lacking. Summarizing,

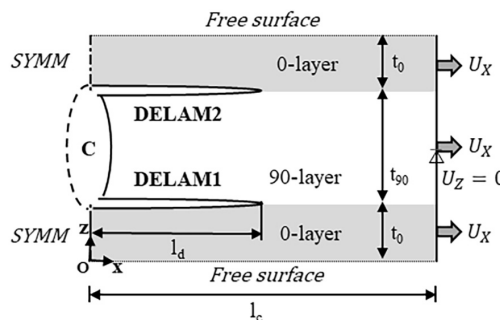


Fig. 2. Representative element of a $[0_m, 90_n]_s$ cross-ply laminate with intralaminar crack (C) and delamination cracks (DELAM1 and DELAM2) at the intralaminar crack tips (deformed shape). With symmetric delamination growth with respect to the intralaminar crack plane and with respect to the laminate midplane, the crack forms I-shaped damage pattern.

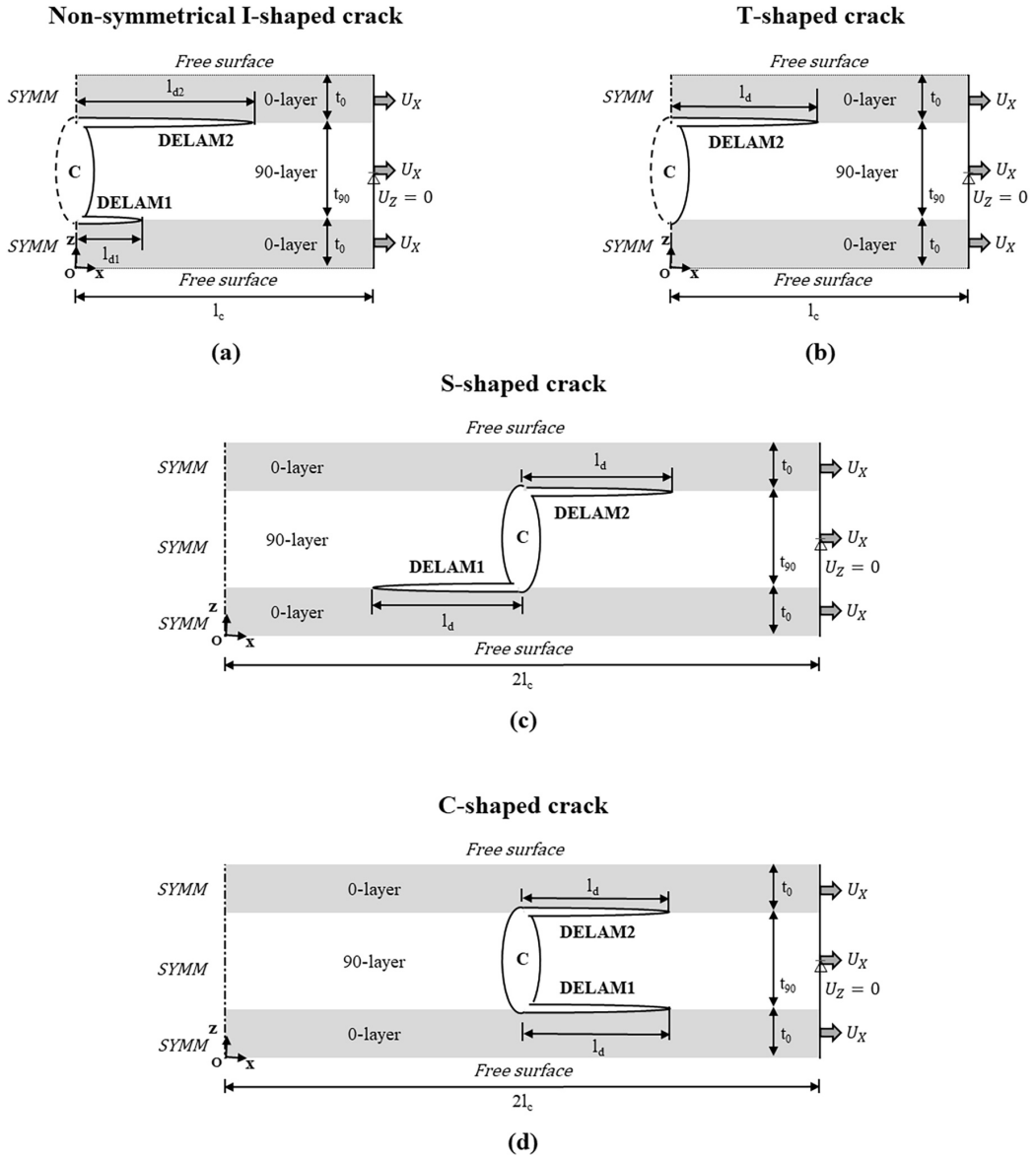


Fig. 3. Representative element of a $[0_m, 90_n]_s$ cross-ply laminate with intralaminar crack (C) and delamination cracks (DELAM1 and DELAM2) at the intralaminar crack tips (deformed shape). With non-symmetric delamination growth with respect to the laminate midplane or the intralaminar crack plane, the cracks may have a non-symmetrical I-shaped (I_{NS}-shaped), T-shaped, S-shaped or C-shaped damage pattern.

- Influence of the damaged ply thickness on ERR is reported in [19,21,24,31,34,36,41] but most of the papers fail to distinguish between the two parts of the effect: a) the scale effect on ERR, meaning that at fixed layer thickness ratio the ERR is proportional to the layer thickness and b) the effect of the layer thickness ratio.
- Most calculations are for glass-fiber epoxy or carbon-fiber epoxy composites without an in-depth analysis of which exactly out of the elastic constants and their ratios are responsible for much larger ERR in glass-fiber epoxy case. The constraining role of the 0-layer and how it affects the ERR in different delamination length regions is not sufficiently elucidated.
- The ERR of non-symmetrical delamination shapes are not sufficiently studied by numerical methods and the observation in [44,45] that equal length of delaminations from both crack tips is most common requires a theoretical explanation.

- Simple ERR recalculation routines for thermo-mechanical loading using ERR from mechanical loading is a question of practical significance due to increasing applications at cryogenic temperatures, and this problem needs more attention. Authors have found only two partially contradicting approaches in [22,34].

In the present work, all the above listed shortcomings are addressed and the dependences on material and geometrical parameters are analyzed. The ERR for the growth of local delaminations in $[0_m, 90_n]_s$ cross-ply laminates is studied by means of VCCT and the J-integral [46] using FEM with contact elements. It is assumed that the intralaminar cracks have fully grown along the width forming tunnel-shaped cracks. Thus, a two-dimensional FE model is considered to study growth of local delaminations only along the axial direction. First, I-shaped damage pattern is considered and the mode-mixity as well as the total ERR is analyzed using the 90-layer thickness and the ply thickness ratio as parameters. Similarly, the effect of the constraint layer's elastic properties such as longitudinal modulus and shear modulus is studied establishing trends. Then, the different non-symmetrical damage patterns shown in Fig. 3 are considered, and the differences in ERR are analyzed. The validity of the ERR trends identified for symmetric I-shape delaminations is verified. Finally, ERR trends for the thermo-mechanical loading case are compared with the mechanical loading case for various damage patterns.

2. Model description and geometrical configurations of local delaminations

Fig. 2 shows a schematic FEM model of the representative element of a $[0_m, 90_n]_s$ cross-ply laminate with a transverse crack (hereafter intralaminar crack, denoted as C in Figs. 2 and 3) and delamination cracks evolving from both tips of the intralaminar crack. The thickness of the 0-layer is t_0 ($= 0.29$ mm) and the thickness of the 90-layer is t_{90} ($= 0.58$ mm). The distance between two adjacent intralaminar cracks is $2 \bullet l_c$, which is parametrically varied to simulate the change in crack density. It is assumed that the intralaminar crack has completely grown along the laminate width forming a tunnel-shaped crack. A two-dimensional finite element model was built in ANSYS software version 2023 R1 [47], and the model was meshed using PLANE182 [48] 4-node solid elements with plane strain behavior. The assumed plane strain behavior is representative because the Poisson's ratio for cross-ply laminates is very small, and the lateral contraction strain of the laminate is much smaller than the applied strain. A refined mesh was used at the delamination crack tip with the smallest element size approximately equal to $1.1 \mu\text{m}$. The element size in the bulk region was approximately $25 \mu\text{m}$. The element size, specifically in the region around the delamination crack tip, was chosen after a systematic mesh sensitivity analysis of the ERR results. Axial displacement U_x corresponding to average mechanical strain of 1 % ($\epsilon_x = 0.01$) was applied at the right side-end ($x = l_c$) of the model. Constraint $U_z = 0$ was also applied at the right side-end at $z = t_0 + \frac{t_{90}}{2}$ for stability of the model. Top- and bottom faces of the model were kept traction free to allow for local curvatures due to intralaminar and interlaminar cracking. The pair-based contact definition was used at the intralaminar and delamination crack faces using element types CONTA172 and TARGE169 [48] with penetration tolerance factor of 0.001. For the contact algorithm, Lagrange multiplier method was used along the contact normal and penalty method was used along the tangent [49]. Note that friction between delamination crack faces is not considered, as it is confirmed in [20] that the presence of friction does not have a significant effect on the stress state around delamination crack tip and effectively on the delamination growth. Prior to performing ERR calculations and the parametric analysis, the FE model was carefully calibrated with respect to i) mesh refinement, ii) boundary conditions, iii) element types, iv) element behavior and v) presence of contact elements. The present FE model was also validated by comparing with literature data (Table 2). The model in Fig. 2 is symmetric with respect to the crack plane ($x = 0$) and with respect to the midplane ($z = t_0 + \frac{t_{90}}{2}$), because the delamination cracks DELAM1 and DELAM2 (Fig. 2) have identical lengths $l_{d2} = l_{d1} = l_d$. Henceforth, this configuration is referred to as I-shape.

The symmetric I-shape model is used as the reference case to study the effect of geometrical and thermo-elastic parameters on ERR. The results are used to verify that the observed trends are the same in cases with non-symmetrical delamination growth with respect to the plane of intralaminar crack forming S-shaped and C-shaped damage pattern, see Fig. 3. In the case of the S-crack, the symmetry with respect to the midplane is lost. The symmetry with respect to the midplane is lost also in the case of the non-symmetrical I-shaped (I_{NS}) crack in Fig. 3 where DELAM1 and DELAM2 have different lengths. The case with delamination growing at only one of the intralaminar crack tips ($l_{d1} = 0$), is called T-shaped damage pattern, and it is an extreme case of the I_{NS}-crack.

Table 1 shows thermo-elastic constants for carbon-fiber epoxy (CF/EP) and glass-fiber epoxy (GF/EP) unidirectional plies which were used in this study. The data were taken from literature.

Table 1
Thermo-elastic constants for CF/EP and GF/EP unidirectional plies.

Mechanical Property	Notation	Units	CF/EP [50]	GF/EP [51]
Longitudinal Modulus	E_L	GPa	138.00	44.73
Transverse Modulus	E_T	GPa	8.10	12.76
Longitudinal Poisson's Ratio	ν_{LT}	—	0.25	0.30
Transverse Poisson's Ratio	ν_{T3}	—	0.45	0.42
In-plane shear modulus	G_{LT}	GPa	5.00	5.80
Out-of-plane shear modulus	G_{T3}	GPa	2.79	4.49
Longitudinal CTE	α_L	1/°C	0	8.6×10^{-6}
Transverse CTE	α_T	1/°C	27.00×10^{-6}	22.1×10^{-6}

3. Results for symmetric I-shaped cracks

3.1. General features of the ERR

This section proposes a normalized nondimensional form of the ERR related to delamination growth based on the assumption that the major part of the strain energy release in a displacement-controlled test can be attributed to the strain energy reduction in the 90-layer with cracks. Hence, normalization is performed with respect to two parameters of the 90-layer: its thickness t_{90} and transverse modulus E_T .

Thermo-elastic constants of undamaged laminates do not depend on the thickness of layers; they depend on the volume fraction of a layer with a given orientation, which in laminates can be expressed as the layer thickness ratio. It is similar in damaged laminates: if the ratio of all geometrical parameters is the same, the elastic properties are the same, meaning that a proper term to characterize the transverse cracking state is not the distance between cracks or the crack density ρ that is defined as the number of cracks per unit length in the layer, but the distance between cracks that is normalized with t_{90} . Even the coordinates must be normalized with respect to the t_{90} , $\bar{x} = x/t_{90}$, $\bar{y} = y/t_{90}$, $\bar{z} = z/t_{90}$. Hence, the normalized crack density ρ_n and the normalized delamination length l_{dn} are introduced as

$$\rho_n = \frac{t_{90}}{2l_c} = \rho t_{90}, l_{dn} = \frac{l_d}{t_{90}} \quad (1)$$

For example, this means that thermo-elastic constants are the same for $[0,90]_s$ and $[0_4,90_4]_s$ laminates made of the same composite and with the same ρ_n and l_{dn} .

Similarly, the stress state in layers of the two considered damaged cross-ply laminates with the same ply thickness ratio, subjected to the same average strain in the x-direction ε_x , is the same. For both laminates, the stress distribution between two cracks in the 90-layer is identical and can be formally written as

$$\sigma_x^{90} = \sigma_x^{90(CL T)} [1 - f(\bar{x}, \bar{y}, \bar{z})] \quad (2)$$

Here, $f(\bar{x}, \bar{y}, \bar{z})$ is an unknown stress perturbation function due to cracks and delaminations, and $\sigma_x^{90(CL T)}$ is the axial (transverse) stress in the undamaged 90-layer corresponding to ε_x applied to the laminate. Then, from the stress equilibrium equations follows that all stress components in layers are proportional to $\sigma_x^{90(CL T)}$. The same applies to all strain components at an arbitrary point because of the linear stress-strain relationships in elasticity: they are proportional to $\sigma_x^{90(CL T)}$.

In Hooke's law, each strain component is inversely proportional to the elastic modulus. Taking E_T^{90} in the 90-layer as a normalizing constant, the expressions for strain (functions g_{ij} in (3)) are dimensionless.

$$\varepsilon_{ij} = \frac{\sigma_x^{90(CL T)}}{E_T^{90}} g_{ij} \left(\bar{x}, \bar{y}, \bar{z}, \rho_n, l_{dn}, \frac{t_0}{t_{90}}, \frac{E_i}{E_T^{90}} \right) \quad (3)$$

In (3), E_i is a symbolic representation of all elastic moduli of the 0- and the 90-layer.

Hence, the strain energy density $\sigma_{ij}\varepsilon_{ij}$ (double indices mean summation) at an arbitrary point is the same in both laminates, and it can be written as,

$$\sigma_{ij}\varepsilon_{ij} = \frac{[\sigma_x^{90(CL T)}]^2}{E_T^{90}} \varphi \left(\bar{x}, \bar{y}, \bar{z}, \rho_n, l_{dn}, \frac{t_0}{t_{90}}, \frac{E_i}{E_T^{90}} \right) \quad (4)$$

The strain energy is obtained by integrating over the volume (it is an area in the used plane strain model), which means that the strain energy becomes proportional to t_{90}^2 . Finally, the ERR is calculated by dividing the strain energy change by the increase in the delamination length, which for a fixed increase in l_{dn} is proportional to t_{90} .

$$G_T = \frac{[\sigma_x^{90(CL T)}]^2 t_{90}}{E_T^{90}} \phi \left(\rho_n, l_{dn}, \frac{t_0}{t_{90}}, \frac{E_L^0}{E_T^{90}}, \frac{E_T^0}{E_T^{90}}, \frac{G_{LT}^0}{E_T^{90}}, \frac{G_{T3}^0}{E_T^{90}}, \frac{E_L^{90}}{E_T^{90}}, \frac{G_{LT}^{90}}{E_T^{90}} \right) \quad (5)$$

Function ϕ in (5) is the normalized dimensionless ERR that depends on ratios of geometrical and elastic parameters. It will be shown in the following sections that the normalizing factor

$$G_0 = \frac{[\sigma_x^{90(CL T)}]^2 t_{90}}{E_T^{90}} \quad (6)$$

has a major effect on the ERR, and the dimensionless part

$$G_{Tn} = \frac{G_T}{G_0} = \phi \left(\rho_n, l_{dn}, \frac{t_0}{t_{90}}, \frac{E_L^0}{E_T^{90}}, \frac{E_T^0}{E_T^{90}}, \frac{G_{LT}^0}{E_T^{90}}, \frac{G_{T3}^0}{E_T^{90}}, \frac{E_L^{90}}{E_T^{90}}, \frac{G_{LT}^{90}}{E_T^{90}} \right) \quad (7)$$

gives fewer corrections, caused by different geometrical and elastic constraints expressed in terms of parameter ratios.

From (5) follows that the ERR is proportional to the 90-ply thickness t_{90} and there is no need to perform FEM calculations for laminates with different t_{90} if the ply thickness ratio is the same. This rather obvious conclusion was confirmed in FEM calculations. Thus, the only geometrical parameter of interest in analyzing the ERR dependence on normalized crack density and normalized delamination length is the effect of the ply thickness ratio t_0/t_{90} . A similar conclusion is valid for E_T^{90} : its effect is included in G_0 (6)). FEM is required to analyze the effect of the modulus ratios on ERR.

In fact, Poisson's ratios should also be included in ϕ (5)). However, parametric analysis presented in Section 3.3 shows that their effect on ERR is minor and can be neglected in practical applications. The ratio G_{T3}^{90}/E_T^{90} is not explicitly included in (5) because the ratio depends only on the transverse Poisson's ratio.

In the performed FEM calculations, the applied mechanical strain $\epsilon_x = 0.01$.

3.2. Examples of material, crack density and ply thickness ratio effect on ERR of I-shaped cracks

The paper focuses on features of the total strain energy release rate G_T and not on simulation of the delamination growth. However, any criterion for the delamination crack growth must include data regarding the mode-mixity in ERR values. The dependence of the total ERR (G_T), Mode I ERR (G_I) and Mode II ERR (G_{II}) on the normalized delamination length for I-shaped damage pattern in CF/EP laminate is shown in Fig. 4a. At normalized crack density $\rho_n = 0.25$ in Fig. 4a, the stress perturbations from transverse cracks are slightly interacting even in the absence of delaminations. For very short delamination length, the crack growth is Mode I dominated. But starting from $l_{dn} \approx 0.25$, Mode I ERR values become insignificant, and thereafter the growth is entirely in Mode II. The total ERR values increase until $l_{dn} \approx 0.5$. Thereafter, the G_T decreases reaching the value for the start of the Mode II region at $l_{dn} \approx 1.6$, and the unstable propagation would most likely stop. The applied load must increase to reach complete delamination i.e., $l_{dn} = 2$. The analysis of growth of large delaminations could be particularly relevant for laminates subjected to cryogenic temperatures. In such conditions, certain carbon/epoxy cross-ply laminates have exhibited growth of large delaminations which hinder the saturation of transverse matrix cracks [50]. The total ERR trend is like the one initially reported in [24], and the mode-mixity has been earlier reported in a work by some of the co-authors [52]. The values change, but the trends are similar for different ply thickness ratios, crack density and for different composites. Fig. 4b compares the methods for the calculation of total ERR values. As it can be seen that the total ERR values calculated using J-integral and VCCT methods almost coincide, with a difference of i) less than 5 % for extremely short or extremely long delaminations, and ii) less than 1.5 % for delaminations with normalized length l_d/t_{90} between 0.1 and 1.9, only J-integral ERR values are reported in the following analysis in the present work.

The layer thickness ratio effect on ERR when keeping the thickness of the 90-layer constant is shown in Fig. 5 for CF/EP and GF/EP cross-ply laminates. The differences between [0,90₄]s and [0₄,90₄]s laminates of the same material are minor; slightly larger ERR for short delaminations and slightly lower for large delaminations if the 0-layer is relatively thinner. The huge material effect on values is explained by a much higher transverse modulus of the GF/EP (12.76 GPa) comparing to 8.1 GPa for CF/EP. Therefore, the transverse stress in the GF/EP layer is higher and so is the ERR (see (5)).

Fig. 6 shows the dependence of the total ERR on delamination length for a) CF/EP and b) GF/EP [0,90]s laminates with normalized crack densities of 0.01, 0.05, 0.25 and 1.00. Because of the higher transverse modulus, the ERR is much larger in GF/EP laminates. In the case of the lowest crack density i.e., $\rho_n = 0.01$, the two neighboring intralaminar cracks with short delaminations are not interacting and this geometry is suitable for studies of parameters affecting the ERR during the initial phase of the delamination growth. After rapidly increasing, the total ERR curve has a linear part between crack lengths $0.1 \leq l_d \leq 0.9$. The slope of the linear part is steeper for GF/EP than CF/EP. As shown later in Section 3.3, the slope is governed more by the E_L^0/E_T^{90} ratio than by the actual value of the transverse modulus of the 90-layer.

The ERR curves for $\rho_n = 0.25$ and $\rho_n = 1.00$ do not show the linear region because of the interaction of stress fields at the tips of two approaching delaminations. As expected, the maximum values of ERR decrease with increasing ρ_n which is also a result of crack

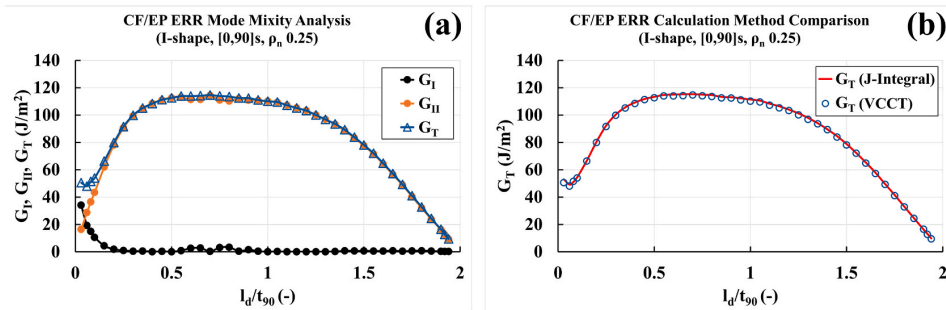


Fig. 4. (a) Mode I, Mode II and total ERR values calculated by VCCT method for the growth of I-shaped delaminations in CF/EP [0,90]s cross-ply laminate, (b) Comparison of G_T values calculated by the methods of J-Integral and VCCT. $\rho_n = 0.25$, $t_{90} = 0.58$ mm, $\epsilon_x = 0.01$.

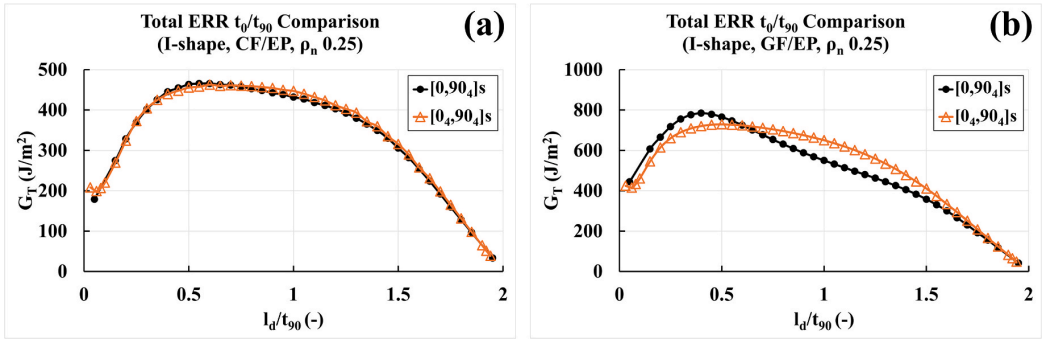


Fig. 5. Total ERR dependence on I-shaped delamination length in (a) CF/EP and (b) GF/EP $[0_m,90_n]s$ laminates with different layer thickness ratios. $\rho_n = 0.25$, $t_{90} = 2.32$ mm, $\epsilon_x = 0.01$.

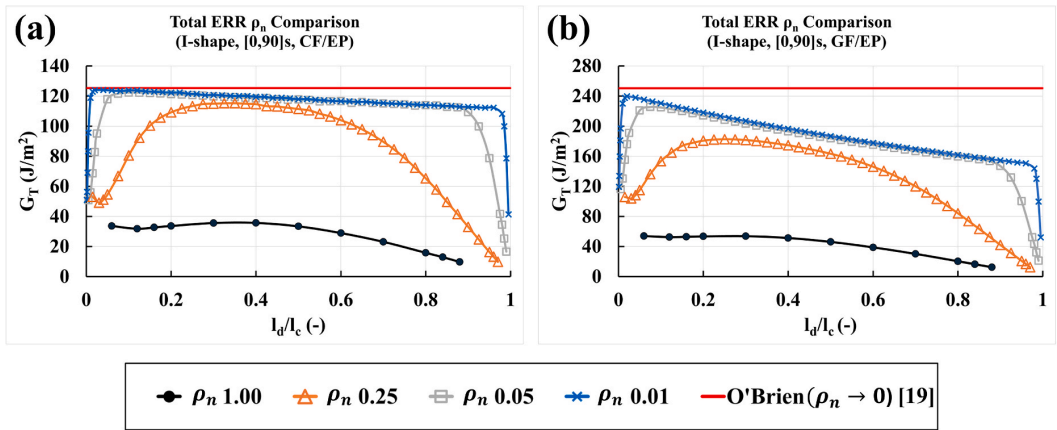


Fig. 6. Total ERR dependence on I-shaped delamination length in (a) CF/EP and (b) GF/EP $[0,90]s$ cross-ply laminates with different normalized crack densities, $t_{90} = 0.58$ mm, $\epsilon_x = 0.01$. For comparison, a plot depicting the constant ERR value according to O'Brien's analytical model [19] is also added.

interaction and lower strain energy between cracks. Hence, the delaminations are more prone to propagate from isolated intralaminar cracks than from cracks in high normalized crack density region. The ERR curves for all crack densities approach zero value at complete delamination length i.e., $\frac{l_d}{l_c} = 1$ making complete delamination improbable in this model. A similar conclusion was obtained in [18].

In [19], O'Brien considered an angle-ply laminate with delaminations originating at the tips of a single transverse crack. In his analytical model, the laminate length, $L \rightarrow \infty$, i.e., the normalized crack density, $\rho_n \rightarrow 0$. Thus, the ERR for the growth of local delaminations becomes independent of the delamination length. For comparison, a plot depicting the constant ERR value according to O'Brien's analytical model [19] is also given in Fig. 6.

The present FE model uses plane strain element behavior. As explained in Section 2, such behavior is representative because the Poisson's ratio for cross-ply laminates is very small, and the lateral contraction strain of the laminate is much smaller than the applied strain. For further validation of the plane strain ERR results in the present work, they were compared with the results presented in [53] by McCartney et al. (refer to Table 2). In their publication, McCartney et al. [53] numerically calculated ERR with the help of the

Table 2

Comparison of the ERR values calculated in the present work and in [53] by McCartney et al. for CF/EP $[0_m,90_n]s$ cross-ply laminates with delamination length $\frac{l_d}{l_c} = 0.5$.

	Present work ($\rho_n = 0.01$)	McCartney BEM [53] ($\rho_n = 0.01$)	Present work ($\rho_n = 0.05$)	McCartney BEM [53] ($\rho_n = 0.05$)
G_T (J/m ²)	117.88	115.65	117.70	115.22

boundary element method (BEM). For a [0,90]_s cross-ply laminate, they applied a small non-zero transverse compressive stress, resulting in a uniform non-zero transverse compressive strain of 0.00038. Similar to the present work, a uniform longitudinal tensile strain of 0.01 was also applied in [53]. The ratio of transverse stresses generated in the 90-layers (σ_T^{90}) in the two cases (BEM [53] and plane strain) would be,

$$\frac{\sigma_{TBEM}^{90}}{\sigma_{Tpl.str}^{90}} = \frac{Q_{22}(0.01 - 0.3 \cdot 0.00038)}{Q_{22} \cdot 0.01} = 0.9886 \quad (8)$$

According to (6), the ERR change would be proportional to the square of stress ratios,

$$\frac{ERR_{pl.str}}{ERR_{BEM}} = \frac{1}{0.9886^2} \quad (9)$$

For $\rho_n = 0.05$ (refer to Table 2), $ERR_{BEM} = 115.22$. Thus, according to (9), $ERR_{pl.str}$ would be

$$ERR_{pl.str} = \frac{115.22}{0.9886^2} = 117.89 \quad (10)$$

Referring to Table 2 and Fig. 6, the $ERR_{pl.str}$ value calculated in the present work is 117.70, with a difference of less than 0.2 % as compared to the $ERR_{pl.str}$ value calculated in (10).

3.3. Parametric analysis of the ERR

In the performed FEM plane strain calculations ($\varepsilon_y = 0$) and according to CLT, the transverse stress in the 90-layer with applied ε_x is

$$\sigma_x^{90(CLT)} = Q_{22}\varepsilon_x, \quad Q_{22} = \frac{E_T^{90}}{1 - \nu_{LT}^{90}\nu_{TL}^{90}} \quad (11)$$

Q_{22} is a component of the stiffness matrix $[Q]$ of the 90-layer. From (6) follows that G_0 used to normalize G_T is calculated as

$$G_0 = \frac{[Q_{22}]^2 t_{90} \varepsilon_x^2}{E_T^{90}} \quad (12)$$

The role of different elastic constants of layers in a plane strain solution must be assessed before analyzing the effect on ERR of the thickness and elastic ratios listed in (5) and (7). The role of elastic constants is different for the 90- and the 0-layer. To reveal more details, each layer is considered as an orthotropic material in calculations. CF/EP [0,90]_s laminate with $\rho_n = 0.25$ was used as a reference, varying l_{dn} and all relevant elastic constants. The conclusions are

1. In the 90-layer, the shear moduli G_{12}^{90} and G_{13}^{90} do not affect the plane strain stress state and the ERR.
2. Reducing the ν_{13}^{90} by 50 % leads to less than 0.02 % increase in ERR. The same reduction of the ν_{12}^{90} leads to less than 0.4 % decrease in ERR. The latter can be explained by reduced Q_{22} leading to reduced transverse stress in the 90-layer in the imposed plane strain conditions, see (11), consequently reducing the ERR.
3. The effect of the longitudinal modulus E_1^{90} on ERR is of similar nature and magnitude, because it affects the transverse stress via Poisson's ratio ν_{21}^{90} which becomes smaller with increasing E_1^{90} , and Q_{22} is reduced.
4. The G_{23}^{90} is proportional to E_2^{90} and it slightly depends on the out-of-plane Poisson's ratio ν_{23}^{90} which usually is in the range between 0.4 and 0.5. An extreme reduction of ν_{23}^{90} by 50 % leads to a 0.5 % increase in ERR for short delaminations and 4 % decrease for long delaminations.

These effects on ERR are much smaller than the ERR differences shown in Section 3.2 and we conclude that the only parameter of importance in the 90-layer is E_2^{90} .

In the 0-layer, the shear moduli G_{12}^0 and G_{23}^0 do not enter the plane strain solution. The Poisson's ratio ν_{12}^0 , ν_{13}^0 and ν_{23}^0 reduction by 50 % leads to ERR reduction by less than 0.05 %, 0.1 % and 0.1 % respectively. The effect of E_2^0 is of the same nature as the effect of E_1^{90} described above. Increasing G_{13}^0 ten times leads to an increase in ERR by approximately 5 %, which is not large but not negligible.

Thus, the elastic constants relevant for ERR are $E_1^0 = E_T^0$ and $G_{13}^0 = G_{LT}^0$ in the 0-layer and E_2^{90} in the 90-layer. Since $E_2^{90} = E_T^{90}$ and it is explicitly extracted in front of (5), the normalized ERR depends on following ratios of elastic constants,

$$G_{Tn} = \frac{G_T}{G_0} = \phi \left(\rho_n, l_{dn}, \frac{t_0}{t_{90}}, \frac{E_L^0}{E_T^{90}}, \frac{G_{LT}^0}{E_T^{90}} \right) \text{ with } G_0 = \frac{E_T^{90} t_{90}}{(1 - \nu_{12}^{90} \nu_{21}^{90})^2 \varepsilon_x^2} \quad (13)$$

Expression (13) shows that the suggested methodology and analysis of parameters is valid also for hybrid laminates where the 0-layer and the 90-layer are made of different materials.

In Fig. 7a–d, the dependence of the normalized ERR on delamination length in CF/EP and GF/EP [0,90]_s laminates with different

crack densities is compared. The differences in ERR for both materials are much smaller after normalization showing that differences in G_0 reflect a major part of ERR differences. “In average”, the G_{Tn} curves are rather similar, but the maximum value is slightly higher for GF/EP, except for high crack densities where due to crack interaction this trend is less pronounced. The decrease in ERR after reaching the maximum is faster in GF/EP and the G_{Tn} values for large delaminations are systematically lower than for CF/EP.

According to (13), the differences in G_{Tn} between both materials in Fig. 7 depend on $\frac{E_L^0}{E_T^0}$ and $\frac{G_{LT}^0}{E_T^0}$ (the layer thickness ratio was kept constant). To simplify the analysis by avoiding complexity due to crack interaction, an extremely low crack density, $\rho_n = 0.01$, was used to identify the trends. Then, the results for higher crack densities were used to confirm them.

For example, the E_L^0/E_T^0 ratio is 17 for CF/EP and 3.5 for GF/EP, and it should affect the G_{Tn} curves. To test this hypothesis, parametric analysis was done for a [0,90]s laminate with 90-layer made of GF/EP material. All the elastic constants of the 0-layer were also as for the GF/EP except for the longitudinal modulus, using E_L^0/E_T^0 as the varying parameter. In this study, the normalization parameter G_0 is the same for all combinations of elastic constants used. Hence, the normalized and the non-normalized ERR curves have the same relative shapes and normalization does not give any additional insight. Thus, normalization was not performed.

The maximum value of the non-normalized ERR curves in Fig. 8a decreases with an increasing E_L^0/E_T^0 ratio. This feature is valid also for higher crack densities, see Fig. 8b and c, but the maximum becomes less distinct at $\rho_n \geq 0.25$ because the intralaminar cracks start interacting at this crack density and separation of different effects is complicated. The maximum ERR dependence on E_L^0/E_T^0 is shown in Fig. 9 for $\rho_n = 0.01$ and another rather low crack density $\rho_n = 0.05$. The dependence which is smooth and with asymptotic behavior is similar for both ρ_n but the values are lower for $\rho_n = 0.05$.

It can also be observed that all ERR curves with different E_L^0/E_T^0 ratios for a fixed normalized crack density intersect at the same value of delamination crack length. At this delamination length, the E_L^0/E_T^0 ratio of the [0,90]s laminate has no effect on the growth of delamination cracks at 0/90-layer interfaces. For larger crack densities, the curves intersect at shorter crack lengths. The reason for this common intersection point and its dependence on crack density are not clear at present.

The slope of the decrease in ERR in the linear region increases with lower E_L^0/E_T^0 .

As the delamination growth is mode-II dominated for most of the range, the effect of the ratio G_{LT}^0/E_T^0 on ERR values is also studied. Fig. 10a and b and Fig. 11a show that for low crack densities ($\rho_n = 0.01$ and $\rho_n = 0.05$), the G_{LT}^0/E_T^0 ratio does not affect the ERR in the large linear region. The G_{LT}^0/E_T^0 ratio affects the delamination growth only in the regions where (i) the delamination crack tip is close to the intralaminar crack or (ii) for severely delaminated interface where the tips of the two delamination cracks approaching from

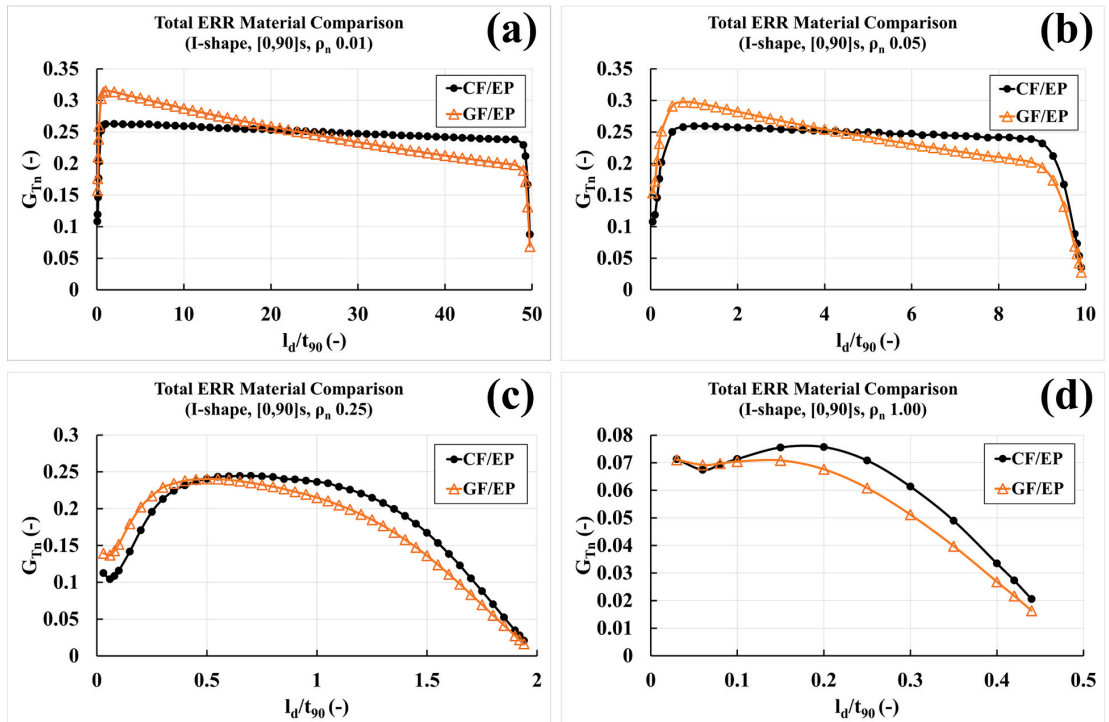


Fig. 7. Normalized ERR of I-shaped delaminations for CF/EP and GF/EP [0,90]s laminates with normalized crack density of (a) 0.01, (b) 0.05, (c) 0.25, and (d) 1.00.

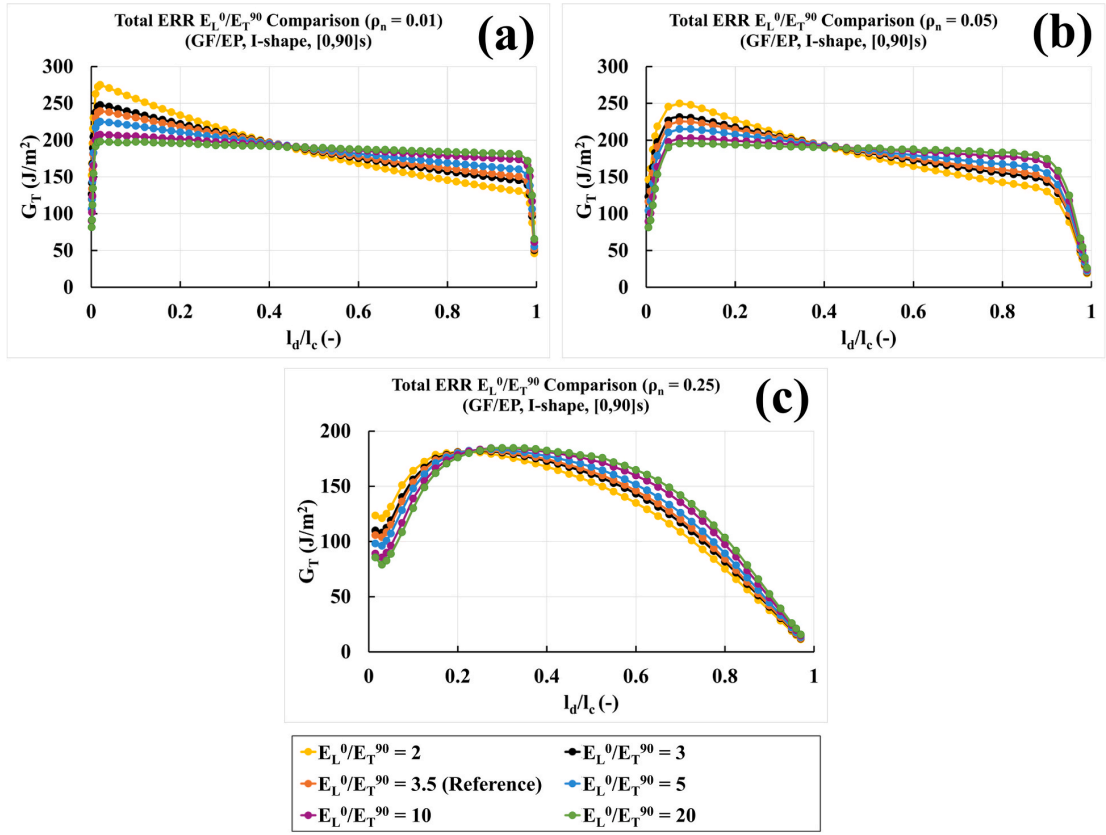


Fig. 8. Dependence of the total ERR on delamination length of I-shaped cracks for [0,90]s laminates with GF/EP 90-layer and a 0-layer which has all the elastic constants same as GF/EP except for E_T^0 which is used as a changing parameter. Normalized crack densities are (a) 0.01, (b) 0.05, and (c) 0.25. $E_L^0/E_T^{90} = 3.5$ is taken as a reference (see Table 1 for GF/EP). $t_{90} = 0.58$ mm.

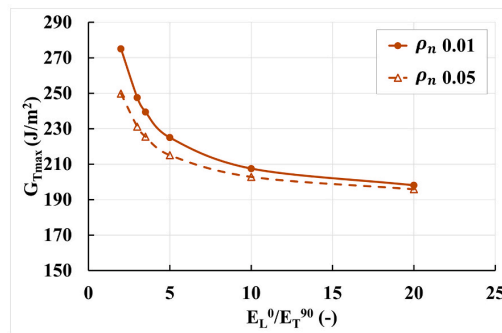


Fig. 9. Trends of maximum values of ERR for I-shaped delaminations for [0,90]s laminates with GF/EP 90-layer and modified GF/EP 0-layer where E_L^0/E_T^{90} is used as a parameter. Results for $\rho_n = 0.01$ and $\rho_n = 0.05$ are presented. $t_{90} = 0.58$ mm.

different sides strongly interact. The ERR values in Fig. 10b are higher if G_{LT}^0/E_T^{90} is lower and all curves converge when delamination length $l_d \approx t_{90}$. The trend is opposite in the region where fraction of the delaminated region is large, see Fig. 11: lower ratio leads to lower ERR. For high crack density ($\rho_n = 0.25$), the whole region between two adjacent intralaminar cracks in Fig. 11b can be

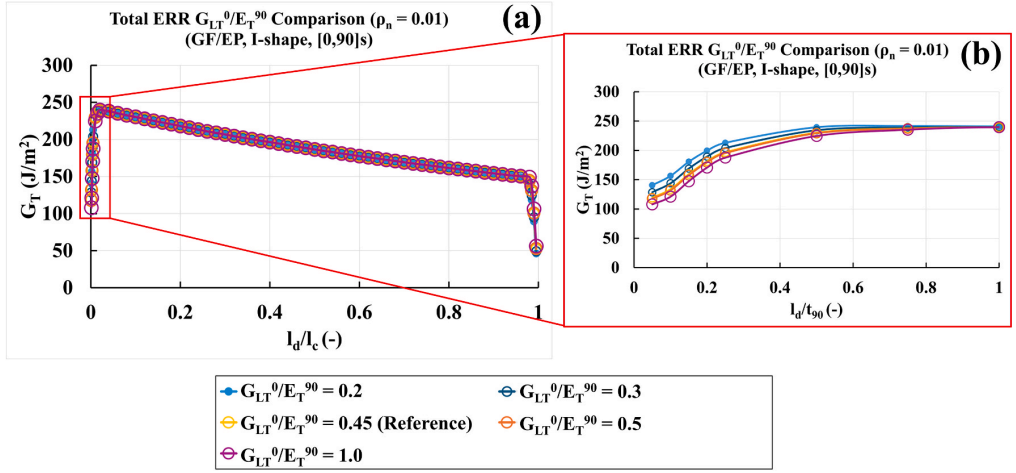


Fig. 10. Dependence of the total ERR on delamination length of I-shaped cracks for [0,90]s laminates with GF/EP 90-layer and a 0-layer which has all the elastic constants same as GF/EP except for G_{LT}^0 which is used as a changing parameter. Note that (a) depicts the ERR curves for the whole delamination length range, whereas (b) is a detail for the short region close to the intralaminar crack. $G_{LT}^0/E_T^{90} = 0.45$ is taken as a reference (see Table 1 for GF/EP). $\rho_n = 0.01$, $t_{90} = 0.58$ mm.

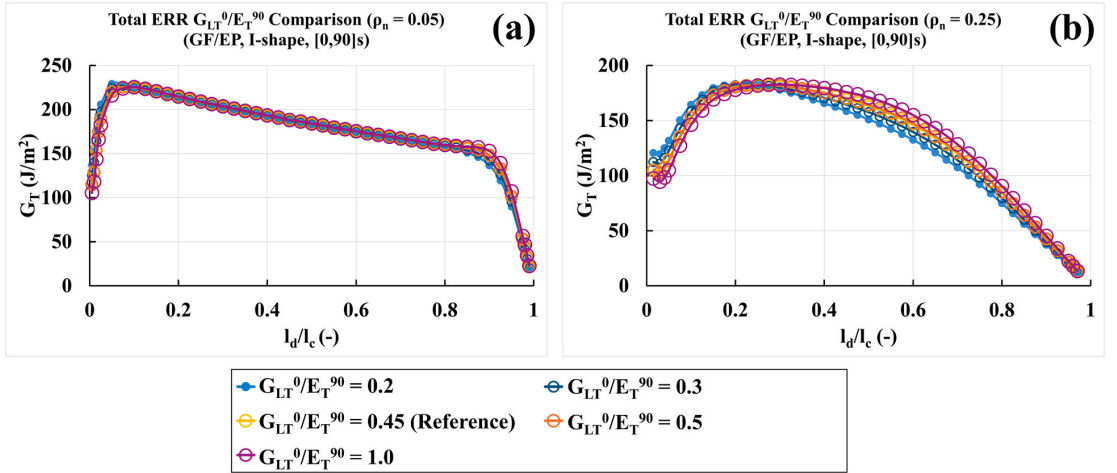


Fig. 11. Dependence of the total ERR on delamination length of I-shaped cracks for [0,90]s laminates with GF/EP 90-layer and a 0-layer which has all the elastic constants same as GF/EP except for G_{LT}^0 which is used as a changing parameter. $G_{LT}^0/E_T^{90} = 0.45$ is taken as a reference (see Table 1 for GF/EP). $\rho_n = 0.05$ in (a) and $\rho_n = 0.25$ in (b). $t_{90} = 0.58$ mm.

considered as the region of interaction. It can be seen comparing Fig. 8c and Fig. 11b, that the effects of E_L^0/E_T^{90} and G_{LT}^0/E_T^{90} ratios on ERR trends are similar for high crack density.

It is evident from Figs. 8, 10–11 that the ERR dependence on the ratio E_L^0/E_T^{90} is much greater than that on the ratio G_{LT}^0/E_T^{90} . Quantitatively, with the change in E_L^0/E_T^{90} ratio from 2 to 20, the G_{Tmax} value can change up to 28 % for small crack densities ($\rho_n = 0.01, 0.05$). For the large normalized crack density of 0.25, the change in G_{Tmax} with the variations in E_L^0/E_T^{90} and G_{LT}^0/E_T^{90} ratios is merely between 0.2 % and 2 %.

As shown in Figs. 8 and 9, the maximum in ERR is higher and the slope of the decrease in the linear region is steeper with lower E_L^0/E_T^{90} . Therefore, the features of delamination growth in GF/EP laminates may be different than in CF/EP laminates: in GF/EP laminates the expected propagation is stable whereas in CF/EP laminates the ERR is almost independent on the delamination length and propagation is rather unstable and more influenced by variation of G_{IIc} along the interface. In addition, due to larger G_0 , the

propagation in GF/EP takes place at lower applied strains than in CF/EP laminates. The G_{IT}^0/E_T^0 ratio shown in Fig. 10 affects the ERR only in a short delamination region and the effect is small compared with expected statistical variation of interface fracture toughness and and, therefore, it can be neglected in practical design. In addition, the applicability of the solution based on homogenized 90-layer for delamination length $0.05t_{90}$ that is comparable with fiber diameter is questionable.

4. Unsymmetric shapes of cracks with delaminations

The delamination onset strain dependence on the crack shape was analyzed in [31]. In the present work, the effect on ERR of the damage pattern (I-shaped, non-symmetrical I-shaped, T-shaped, S-shaped or C-shaped delaminations, see Fig. 2 and Fig. 3) is analyzed. Fig. 12 shows that the shape prominently affects the total ERR available for delamination growth. At low crack density ($\rho_n = 0.01$ and $\rho_n = 0.05$), the behavior of ERR for I- and C-shape of delaminations, of which both are symmetric with respect to the midplane, is rather similar with the slope of ERR decrease steeper for the I-shape. For S-shaped delaminations, the ERR is about 15 % larger than for the I-shape and with more distinct maximum for short delaminations, but in general the dependence on l_d is the same.

The T-shaped damage pattern clearly stands out from the rest, with total ERR values reducing drastically after reaching maximum at extremely short delamination lengths. Thereafter, total ERR reaches zero stopping the growth completely. This can be explained by the principle of Saint-Venant: the effect of the perturbation reduces with the distance from the source of perturbation. For example, in the case of an intralaminar crack, the stress state in the 90-ply recovers the CLT value at some distance away from the intralaminar crack. It would be similar for a delamination developing from the laminate edge along one of the interfaces in the laminate.

In the case of high crack densities ($\rho_n = 0.25$ and $\rho_n = 1.00$), all the delamination shapes have similar ERR trends, with S-shaped delaminations still achieving highest ERR values for all delamination lengths and T-shaped delaminations achieving the lowest ERR values for intermediate and large lengths.

The T-shaped delamination damage pattern is an extreme case of non-symmetrical I-shaped delamination with $l_{d1} \neq l_{d2}$ (see Fig. 3). Hence, it is expected that for non-symmetrical I-shaped delamination, the ERR available for delamination growth is intermediate between symmetrical I-shaped and T-shaped delaminations.

In Fig. 13, the length of the DELAM1 is kept constant, i.e., $l_{d1} = \text{const}$ while varying l_{d2} for the case of $\rho_n = 0.01$. First, we notice that the ERR for DELAM2 growth is several times larger in short DELAM2 region if l_{d1} is longer. The maximum in the ERR curves increases with increasing l_{d1} , which is an indication that DELAM2 is no longer affected by the presence of the DELAM1 tip. The dependence of the

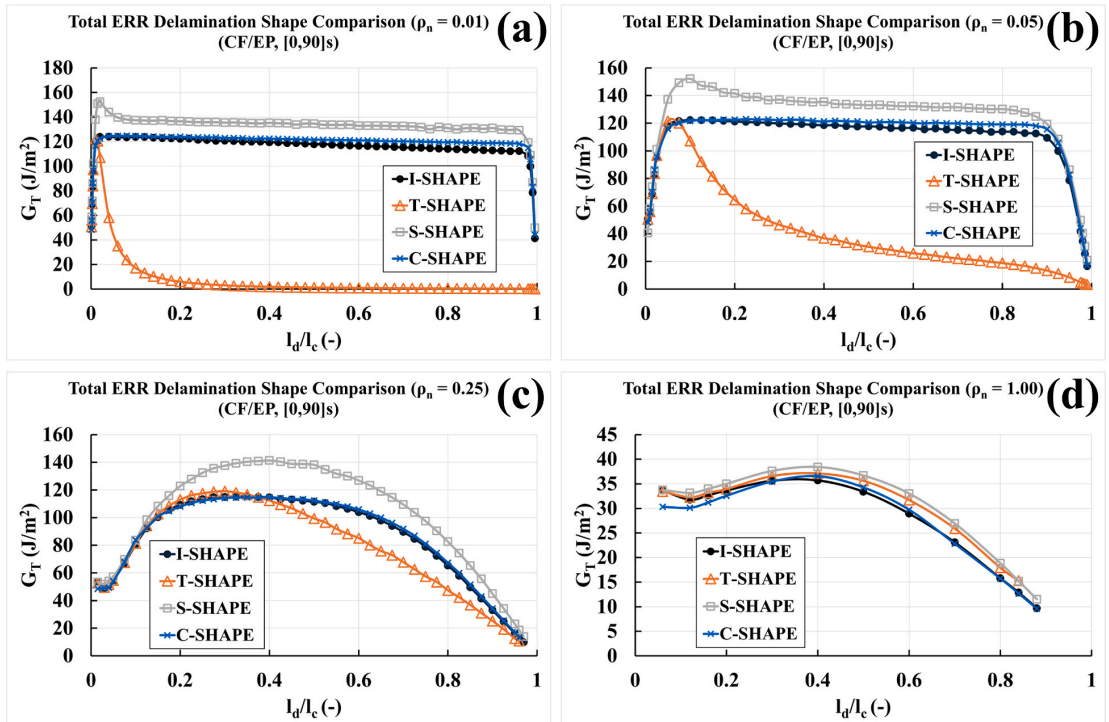


Fig. 12. Total ERR trends for the growth of I-shaped, T-shaped, S-shaped and C-shaped cracks for CF/EP [0,90]s cross-ply laminates with normalized crack densities of (a) 0.01, (b) 0.05, (c) 0.25 and (d) 1.00. $t_{90} = 0.58$ mm.

ERR maximum on l_{d1} appears to be asymptotic.

When l_{d2} approaches l_{d1} , the general monotonously decreasing trend locally changes. In the close presence of the DELAM1, the ERR for DELAM2 starts to increase. The ERR keeps increasing even when $l_{d2} > l_{d1}$, before it monotonously decreases again with the same shape but with slightly higher ERR values. Interestingly, the ERR value when $l_{d2} = l_{d1}$ is almost independent on the DELAM1 length. In fact, according to Fig. 12a (I-shaped crack, $\rho_n = 0.01$), this value should be slightly decreasing with increase in l_{d1} . But for CF/EP, the decrease is slow.

Analyzing the delamination process, a question arises – What happens with a very short DELAM2 in the non-symmetrical case with existing larger DELAM1. Does it grow faster, trying to catch up with DELAM1? Results in Fig. 13 show that it does try to catch up with DELAM1. The ERR for very short DELAM2 is much higher than for DELAM1, which behaves similarly to a T-shaped crack when $l_{d2} \ll l_{d1}$. When $l_{d2} > l_{d1}$, i.e., after the inflection point in Fig. 13, the ERR becomes smaller than for delaminations of equal length and then, DELAM1 has better conditions for growth. Conclusion: ‘Delaminations having initially different lengths are prone to develop into symmetric I-shaped delaminations’. After that, the growth could be more like in a small jump on one and then on the other interface than an equal extension of both delaminations. Of course, variation in fracture toughness along the interface would introduce additional irregularity.

In Figs. 14–16 trends with respect to changes in elastic constants and in ply thickness ratio are presented for four types of crack shapes to show that the trends found in Section 3 for I-shape cracks are valid also for other crack geometries, using as an example CF/EP composite. In Fig. 14 the effect of the 0-layer longitudinal modulus on ERR is presented using $E_L^0/E_T^0 = 17$ as the reference value and $E_L^0/E_T^0 = 8.5$ for comparison. For all damage shapes, the trend is the same as for I-shaped cracks: decrease of E_L^0 leads to higher ERR. In accordance with Fig. 8 in Section 3, the effect is small for I-shaped cracks and also for C-shaped cracks. The effect is significant for S-shaped cracks, probably because in this unsymmetric case local bending of the 0-layer is involved and the reduced bending stiffness leads to larger ERR.

The effect of the G_{LT}^0 on ERR is shown in Fig. 15a and b where two values of the G_{LT}^0/E_T^0 ratio are used. The conclusion for all crack shapes is the same as obtained in Section 3 for I-cracks: in the linear region far away from the intralaminar crack, the ERR is not affected, see Fig. 15a. For short delaminations, see Fig. 15b, there is a small difference: with larger G_{LT}^0/E_T^0 the ERR becomes smaller. The effect was larger for GF/EP laminate in Fig. 10.

Finally, the effect of the ply thickness ratio is shown in Fig. 16. For all crack shapes the ERR becomes larger with decreasing 0-layer thickness. The effect is larger for the unsymmetric S-shape crack because the bending stiffness of the 0-layer is much lower. It appears a general trend for all unsymmetric shapes, see the curves for T-shaped cracks. For symmetric I-shaped and C-shaped cracks the results are rather similar.

Results in Fig. 14 show that for very short delamination the ERR is rather similar for all delamination shapes, meaning that all shapes are equally possible and a realization of a particular shape depends on local fiber microstructure in the transverse crack tip region (resin rich regions, nonuniform fiber distribution etc.). For longer delaminations, the ERR for S-shaped cracks is noticeably larger than for other shapes. An interesting feature for S-cracks is a much stronger dependence of the ERR on the E_L^0/E_T^0 than for other shapes: in contrast to I-shaped cracks, the ERR for smaller ratio is larger even for large delamination. Similarly, ERR for S-shaped cracks is much more sensitive with respect to the ply thickness ratio than other crack shapes, see Fig. 16, with the ERR much larger if the t_0/t_{90} is smaller. These features make the S-shaped cracks more probable for GF/EP with relatively thick 90-layers.

The present work analyzes the growth of local delaminations in a purely qualitative way using trends in the total ERR values. A detailed quantitative analysis of the delamination growth by the comparison of calculated ERR values with the fracture toughness value of a given material system is out of scope of the current study. At the moment, the exact fracture toughness values of the material systems used in the present study are not known to the authors. Nevertheless, an estimate has been made regarding the local delamination growth based on G_{IIc} values of similar material systems available in literature.

From literature, it can be estimated that the Mode II fracture toughness (G_{IIc}) of CF/EP used in the present study would lie between 240 – 290 J/m² [26,54]. Among all cases presented in figures above, the delamination growth is possible only in Fig. 5 and Fig. 13. In

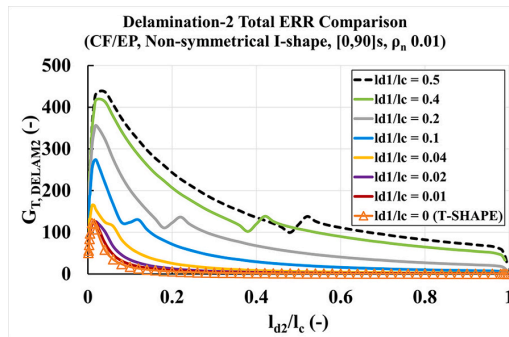


Fig. 13. Comparison for total ERR trends for the growth of non-symmetrical I-shaped cracks (refer Fig. 3), with constant l_{d1} and changing l_{d2} , for CF/EP [0,90]s cross-ply laminates with normalized crack density of 0.01. $t_{90} = 0.58$ mm.

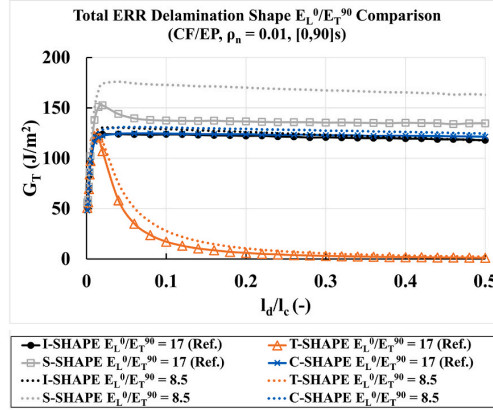


Fig. 14. Influence of the ratio of elastic constants E_L^0/E_T^{90} , see (13), on the trends of total ERR as a function of delamination length for I-shaped, T-shaped, C-shaped and S-shaped cracks in CF/EP [0,90]_s cross-ply laminates with normalized crack density of 0.01. $t_{90} = 0.58$ mm.

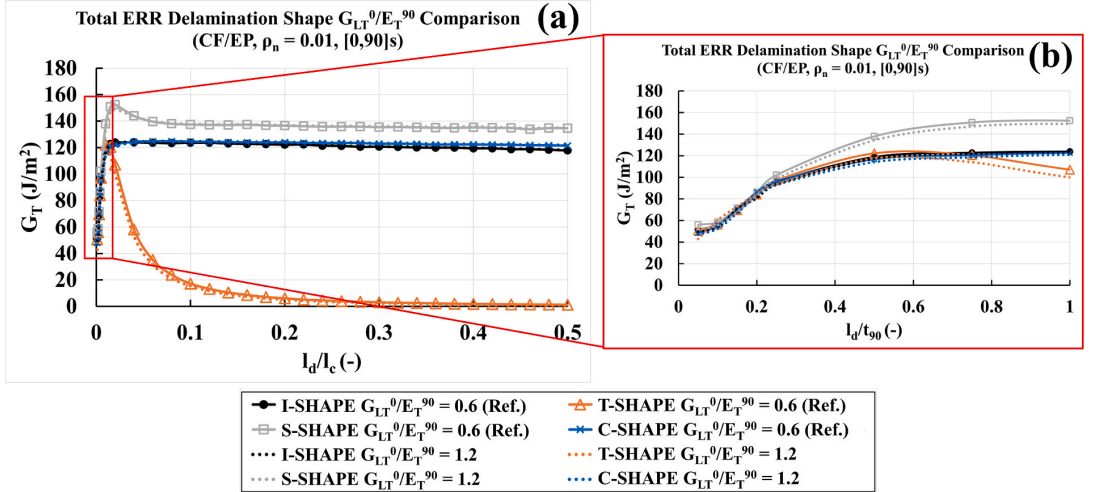


Fig. 15. Influence of the ratio of elastic constants G_{LT}^0/E_T^{90} , see (13), on the trends of total ERR as a function of delamination length for I-shaped, T-shaped, C-shaped and S-shaped cracks in CF/EP [0,90]_s cross-ply laminates with normalized crack density of 0.01. Note that (a) depicts the ERR curves for the whole delamination lengths range, whereas (b) is a detail for the short region close to the intralaminar crack. $t_{90} = 0.58$ mm.

Fig. 5, it is mainly due to large thickness of the 90-layer. In Fig. 13, representing a non-symmetrical I-shaped delamination, the growth of the shorter delamination is possible only if the delamination on the other interface is rather long. The growth of the shorter delamination would stop when approaching the length of the longer delamination.

As previously mentioned, a 2D-FEM model is used in the present work to study the growth of local delaminations only along the axial direction i.e., 1D delamination growth. Analysis involving 2D delamination growth would require knowledge of failure criteria and mode-mixity. For the analysis of delamination growth along the laminate width, the reader is directed to publications [55] and [56] concerning 3D simulation of the delamination growth in composite laminates.

5. Effect of thermal stress on ERR

The difference between the curing temperature and the service temperature, and the mismatch in thermal expansion coefficients lead to thermal stresses. In linear elasticity, these stresses are linearly superimposed to the stresses caused by mechanical loading. Since thermal loading leads to a different stress state in the damaged laminate than the mechanical, the question is how to use the gained

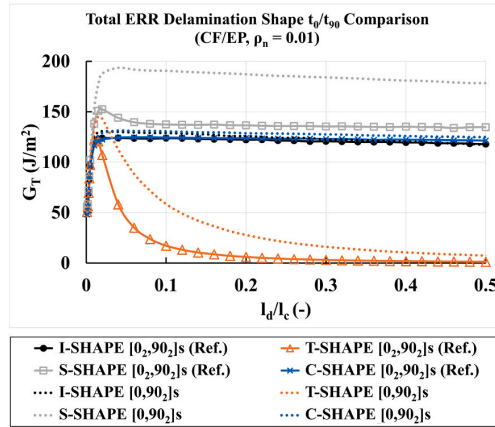


Fig. 16. Influence of the ratio of geometrical parameters t_0/t_{90} , see (13), on the trends of total ERR as a function of delamination length for I-shaped, T-shaped, C-shaped and S-shaped cracks in CF/EP $[0_m, 90_n]_s$ cross-ply laminates with normalized crack density of 0.01. $t_{90} = 0.58$ mm.

information on the ERR in mechanical loading to calculate the ERR in thermo-mechanical loading at given temperature without doing a new set of FEM calculations. This question is of great practical significance because of an increasing application of composites at cryogenic temperatures.

The problem was analyzed in [57]. Using our notation, the conclusion was that the function ϕ in (5) is different in thermal and mechanical loading, $\phi_{mech} \neq \phi_{th}$ and the ERR in thermomechanical loading can be written as

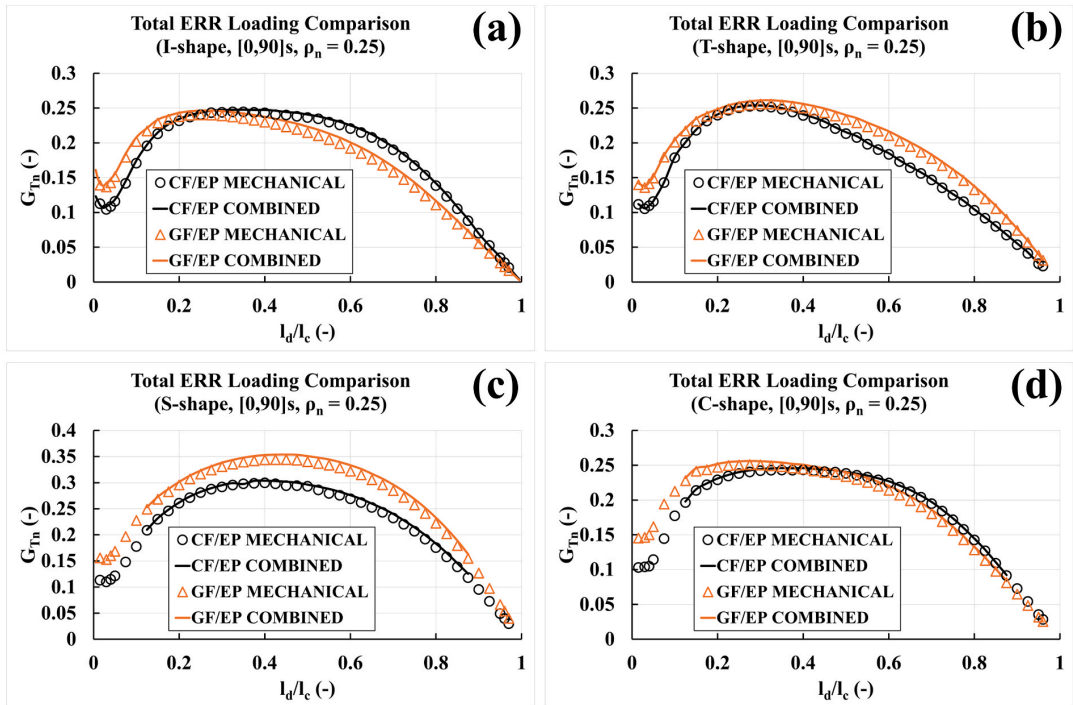


Fig. 17. Normalized ERR available for the growth of (a) I-shaped, (b) T-shaped, (c) S-shaped and (d) C-shaped delaminations in CF/EP and GF/EP $[0, 90]_s$ cross-ply laminates with normalized crack density of 0.25 under mechanical ($\epsilon_x = 0.01$) and combined mechanical + thermal ($\Delta T = -100^\circ\text{C}$) loading.

$$G_T = \left[\sigma_{xmech}^{90(CL.T)} \sqrt{\phi_{mech}} + \sigma_{xth}^{90(CL.T)} \sqrt{\phi_{th}} \right]^2 \frac{t_{90}}{E_T^{90}} \quad (14)$$

The differences and similarities in ϕ_{mech} and ϕ_{th} were analyzed numerically in [34] showing that the main trends are the same, but the shape is slightly different.

Paris et al. assumed in [22] that ϕ_{mech} and ϕ_{th} are proportional leading to

$$G_T = \left[\sigma_{xmech}^{90(CL.T)} + k_1 \sigma_{xth}^{90(CL.T)} \right]^2 \phi_{mech} \frac{t_{90}}{E_T^{90}} \quad (15)$$

where k_1 is a laminate constant.

The hypothesis to be verified in this section is that $k_1 = 1$ and a sufficient accuracy can be achieved in a combined thermo-mechanical loading just by using the normalized ERR in mechanical loading and multiplying it with the summary thermo-mechanical transverse stress in the 90-layer calculated using CLT. In other words, the calculated normalized ERR in mechanical and in thermo-mechanical loading is the same.

This statement is not entirely true: a) $\phi_{mech} \neq \phi_{th}$, but since the axial stress is dominant the difference may be small; b) the type of the boundary conditions could be slightly different in thermal and in mechanical problems (size of the contact zone etc.) violating the superposition of effects. Nevertheless, if proven sufficiently accurate, this gives a very simple ERR recalculation tool for any service temperature.

The accuracy of the hypothesis is demonstrated here for $[0_m, 90_n]_s$ laminate with rather large crack density $\rho_n = 0.25$ presenting data for CF/EP and GF/EP laminates at temperature that is 100°C below the manufacturing temperature, i.e., $\Delta T = -100^\circ\text{C}$.

Because of the problems outlined above, the normalized ERR values in mechanical and in combined thermo-mechanical loading cannot exactly coincide. However, the results presented in Fig. 17 show that for all shapes of delaminations the normalized ERR curves are very close, including for CF/EP where the thermal stress at $\varepsilon_x = 0.01$ is relatively larger than for GF/EP laminate. For all the normalized ERR curves shown in Fig. 17, the maximum difference between the values for mechanical and combined thermo-mechanical loading cases lie between 1.3 % (for CF/EP C-shaped cracks) and 6.2 % (for CF/EP S-shaped cracks) proving the applicability of (15) with $k_1 = 1$.

It must be noted that the thermo-elastic constants of a composite depend on temperature, the effect that should be accounted for when analyzing delaminations at cryogenic temperatures. For most composites, the transverse and shear moduli are higher at low temperatures, and the thermal expansion coefficients are slightly lower, affecting the value of thermal stress in the 90-layer and the G_0 term in the ERR expression (refer (5), (6), (14)). These in-situ values should be used to analyze delamination growth at low temperatures.

6. Conclusions

In this work, the strain energy release rate (ERR) for growth of local delaminations initiating at the tips of intralaminar cracks in $[0_m, 90_n]_s$ cross-ply laminates is studied using FEM with the help of virtual crack closure technique (VCCT) and the J-integral. It was assumed that the intralaminar cracks have fully grown along the laminate width forming tunnel shaped cracks. Thus, a 2D-FEM model was built to study the growth of local delaminations only along the axial direction. It was found that, except for short delaminations, the delamination growth is in Mode-II. Four different damage geometries with I-shaped, T-shaped, C-shaped and S-shaped delaminations are analyzed.

The main geometrical and elastic parameters affecting the ERR are identified through theoretical analysis, and plane strain FEM parametric analysis is carried out presenting the ERR versus the normalized delamination length for a fixed values of the normalized crack density. In all cases, i) the ERR rapidly increases with increasing delamination length, ii) slowly (linearly for low crack densities) decreases afterwards and iii) has a fast decrease region when the delaminations from two sides approach each other.

The parameters which have a major effect on ERR are the thickness of the 90-layer (linear dependence) and the transverse stress in the 90-layer (quadratic). These parameters are used to normalize the ERR, and in the normalized form the ERR values for CF/EP and GF/EP are rather similar.

The ply thickness ratio has a smaller effect, with ERR increasing for short delaminations if the 0-layer is relatively thinner, this effect being more pronounced in GF/EP laminate where the E_L/E_T is smaller than in CF/EP laminates. Elastic parameters affecting the values and the shape of the ERR are E_L/E_T and G_{LT}/E_T . With a decrease in these ratios, the maximum ERR value increases. The shear modulus affects the ERR curve only for very short delaminations lengths. These trends are common for all four damage shapes.

The S-shaped delamination has the highest ERR among the four geometrical configurations. The ERR for I-shaped and C-shaped cracks is rather similar, whereas for T-shaped cracks the ERR is the lowest and it rapidly decays with the delamination length.

For ERR in thermo-mechanical loading cases, a simple and rather accurate recalculation procedure is presented and validated showing that the normalized ERR with and without temperature effect is almost same if in normalization the total thermo-mechanical stress in the 90-layer is used.

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CRediT authorship contribution statement

Anish Niranjana Kulkarni: Writing – review & editing, Writing – original draft, Validation, Software, Investigation, Formal analysis. **Andrejs Pupurs:** Writing – review & editing, Validation, Supervision, Software, Funding acquisition, Data curation, Conceptualization. **Janis Varna:** Writing – review & editing, Writing – original draft, Supervision, Methodology, Conceptualization.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

Appendix A. Supplementary data

Supplementary data to this article can be found online at <https://doi.org/10.1016/j.engfracmech.2025.111549>.

Data availability

Data will be made available on request.

References

- [1] Berthelot JM. Transverse cracking and delamination in cross-ply glass-fiber and carbon-fiber reinforced plastic laminates: Static and fatigue loading. *Appl Mech Rev* 2003;56. <https://doi.org/10.1115/1.1519557>.
- [2] Parvizi A, Bailey JE. On multiple transverse cracking in glass fibre epoxy cross-ply laminates. *J Mater Sci* 1978;13. <https://doi.org/10.1007/BF00541666>.
- [3] Dvorak GJ, Laws N, Hejazi M. Analysis of Progressive Matrix Cracking in Composite Laminates I. Thermoelastic Properties of a Ply with Cracks. *J Compos Mater* 1985;19. <https://doi.org/10.1177/002199838501900302>.
- [4] Hashin Z. Analysis of cracked laminates: a variational approach. *Mech Mater* 1985;4. [https://doi.org/10.1016/0167-6636\(85\)90011-0](https://doi.org/10.1016/0167-6636(85)90011-0).
- [5] Talreja R. CONTINUUM MECHANICS CHARACTERIZATION OF DAMAGE IN COMPOSITE MATERIALS. *Proc R Soc Lond A* 1985;399. <https://doi.org/10.1098/rspa.1985.0055>.
- [6] Dvorak GJ, Laws N. Analysis of Progressive Matrix Cracking In Composite Laminates II. first Ply failure. *J Compos Mater* 1987;21. <https://doi.org/10.1177/002199838702100402>.
- [7] Lim SG, Hong CS. Prediction of Transverse Cracking and Stiffness Reduction in Cross-Ply Laminated Composites. *J Compos Mater* 1989;23:695–713. <https://doi.org/10.1177/002199838902300704>.
- [8] Varna J, Berglund L. Multiple transverse cracking and stiffness reduction in cross-ply laminates. *J Compos Tech Res* 1991;13. <https://doi.org/10.1520/ctr10213j>.
- [9] Zhang J, Fan J, Soutis C. Analysis of multiple matrix cracking in $[\pm 0\text{m}/90\text{s}]_s$ composite laminates. Part 1: In-plane stiffness properties. *Composites* 1992;23. [https://doi.org/10.1016/0010-4361\(92\)90327-Q](https://doi.org/10.1016/0010-4361(92)90327-Q).
- [10] Zhang J, Fan J, Soutis C. Analysis of multiple matrix cracking in $[\pm 0\text{m}/90\text{s}]_s$ composite laminates. Part 2: Development of transverse ply cracks. *Composites* 1992;23. [https://doi.org/10.1016/0010-4361\(92\)90328-R](https://doi.org/10.1016/0010-4361(92)90328-R).
- [11] Fan J, Zhang J. In-situ damage evolution and micro/macro transition for laminated composites. *Compos Sci Technol* 1993;47. [https://doi.org/10.1016/0266-3538\(93\)90040-N](https://doi.org/10.1016/0266-3538(93)90040-N).
- [12] Varna J, Berglund L. Two-dimensional transverse cracking in $[0\text{m}/90\text{s}]_s$ cross-ply laminates. *Eur J Mech A Solids* 1993;12:699–723.
- [13] Varna J, Berglund LA. Thermo-elastic properties of composite laminates with transverse cracks. *J Compos Tech Res* 1994;16. <https://doi.org/10.1520/ctr10397j>.
- [14] Joffe R, Varna J. Analytical modeling of stiffness reduction in symmetric and balanced laminates due to cracks in 90° layers. *Compos Sci Technol* 1999;59. [https://doi.org/10.1016/S0266-3538\(99\)00025-1](https://doi.org/10.1016/S0266-3538(99)00025-1).
- [15] Lundmark P, Varna J. Stiffness Reduction in Laminates at High Intralaminar Crack Density: effect of Crack Interaction. *Int J Damage Mech* 2011;20. <https://doi.org/10.1177/1056789509351840>.
- [16] Loukil MS, Hussain W, Kirti A, Pupurs A, Varna J. Thermoelastic constants of symmetric laminates with cracks in 90° -layer: Application of simple models. *Plast, Rubber Compos* 2013;42. <https://doi.org/10.1179/1743289811Y.0000000064>.
- [17] Nairn JA. Matrix Microcracking in Composites. *Comprehensive Composite Materials* 2000. <https://doi.org/10.1016/B0-08-042993-9/00069-3>.
- [18] París F, Blázquez A, McCartney LN, Mantić V. Characterization and evolution of matrix and interface related damage in $[0/90]_S$ laminates under tension. Part I: Numerical predictions. *Compos Sci Technol* 2010;70. <https://doi.org/10.1016/j.compscitech.2010.03.004>.
- [19] O'Brien TK. ANALYSIS OF LOCAL DELAMINATIONS AND THEIR INFLUENCE ON COMPOSITE LAMINATE BEHAVIOR. *ASTM Spec Tech Publ* 1985. <https://doi.org/10.1520/stp36310s>.
- [20] Blázquez A, Mantić V, París F, McCartney LN. Stress state characterization of delamination cracks in $[0/90]_S$ symmetric laminates by BEM. *Int J Solids Struct* 2008;45:1632–62. <https://doi.org/10.1016/J.IJSOLSTR.2007.10.013>.
- [21] Carraro PA, Novello E, Quaresimin M, Zappalorto M. Delamination onset in symmetric cross-ply laminates under static loads: Theory, numerics and experiments. *Compos Struct* 2017;176. <https://doi.org/10.1016/j.compstruct.2017.05.030>.
- [22] París F, Blázquez A, McCartney LN, Barroso A. Characterization and evolution of matrix and interface related damage in $[0/90]_S$ laminates under tension. Part II: Experimental evidence. *Compos Sci Technol* 2010;70. <https://doi.org/10.1016/j.compscitech.2010.03.006>.
- [23] Cook J, Gordon JE, Evans CC, Marsh DM. A mechanism for the control of crack propagation in all-brittle systems. *Proc R Soc Lond A* 1964;282. <https://doi.org/10.1098/rspa.1964.0248>.
- [24] Kim TW, Kim HJ, Im S. Delamination crack originating from transverse cracking in cross-ply composite laminates under extension. *Int J Solids Struct* 1991;27:1925–41. [https://doi.org/10.1016/0020-7683\(91\)90186-J](https://doi.org/10.1016/0020-7683(91)90186-J).
- [25] O'Brien TK. Characterization of Delamination Onset and Growth in a Composite Laminate. *ASTM Special Technical Publication*, vol. STP 775, 1982. <https://doi.org/10.1520/STP34325S>.
- [26] Grossman FW, Wang ASD. The Dependence of Transverse Cracking and Delamination on Ply Thickness in Graphite/Epoxy Laminates. *ASTM Special Technical Publication*, vol. STP 775, 1982. <https://doi.org/10.1520/STP34324S>.

- [27] Rahmani M, Farrokhabadi A. Prediction of induced delamination development in [0/90]s composite laminates using a computational analytical approach. *Compos Struct* 2018;203. <https://doi.org/10.1016/j.compstruct.2018.07.018>.
- [28] Delbariani-Nejad A, Farrokhabadi A. A failure criterion to predict the onset of matrix cracking induced delamination in general composite laminates. *Compos Struct* 2020;235. <https://doi.org/10.1016/j.compstruct.2019.111564>.
- [29] Rezaei Jafari S, Farrokhabadi A, Montesano J. The effect of staggered matrix crack induced delamination growth on the mechanical properties of cross-ply laminates. *Compos Struct* 2021;272. <https://doi.org/10.1016/j.compstruct.2021.114196>.
- [30] Nairn JA, Hu S. The initiation and growth of delaminations induced by matrix microcracks in laminated composites. *Int J Fract* 1992;57. <https://doi.org/10.1007/BF00013005>.
- [31] Romaris Villanueva J, Kassapoglou C. Damage history in composite laminates: Matrix cracks leading to delaminations. *J Compos Mater* 2024;58. <https://doi.org/10.1177/00219983241232983>.
- [32] Lim S-H, Li S. Energy release rates for transverse cracking and delaminations induced by transverse cracks in laminated composites. *Compos A Appl Sci Manuf* 2005;36:1467–76. <https://doi.org/10.1016/j.compositesa.2005.03.015>.
- [33] Socc CA, Kassapoglou C. Prediction of matrix crack initiation and evolution and their effect on the stiffness of laminates with off-axis plies under in-plane loading. *Compos Sci Technol* 2020;200. <https://doi.org/10.1016/j.compscitech.2020.108427>.
- [34] Wang ASD, Kishore NN, Li CA. Crack development in graphite-epoxy cross-ply laminates under uniaxial tension. *Compos Sci Technol* 1985;24. [https://doi.org/10.1016/0266-3538\(85\)90058-2](https://doi.org/10.1016/0266-3538(85)90058-2).
- [35] Varna J, Freire RTS, Loukil MS, Al-Ramahi NJ. COD and CSD based model for in-plane stiffness of symmetric laminates with cracks in plies and local delaminations: Analysis of crack face sliding. *Compos A Appl Sci Manuf* 2025;189:108594. <https://doi.org/10.1016/j.compositesa.2024.108594>.
- [36] Takeda N, Ogihara S. Initiation and growth of delamination from the tips of transverse cracks in CFRP cross-ply laminates. *Compos Sci Technol* 1994;52:309–18. [https://doi.org/10.1016/0266-3538\(94\)90166-X](https://doi.org/10.1016/0266-3538(94)90166-X).
- [37] Zhang J, Soutis C, Fan J. Strain energy release rate associated with local delamination in cracked composite laminates. *Composites* 1994;25. [https://doi.org/10.1016/0010-4361\(94\)90026-4](https://doi.org/10.1016/0010-4361(94)90026-4).
- [38] Kashtalyan M, Soutis C. A study of matrix crack tip delaminations and their influence on composite laminate stiffness. *Adv Compos Lett* 1999;8. <https://doi.org/10.1177/096369359900800401>.
- [39] Kashtalyan M, Soutis C. Analysis of local delaminations in composite laminates with angle-ply matrix cracks. *Int J Solids Struct* 2002;39. [https://doi.org/10.1016/S0020-7683\(02\)00007-0](https://doi.org/10.1016/S0020-7683(02)00007-0).
- [40] Kashtalyan M, Soutis C. Analysis of composite laminates with intra- and interlaminar damage. *Prog Aerosp Sci* 2005;41. <https://doi.org/10.1016/j.paerosci.2005.03.004>.
- [41] Noh J, Whitcomb J. Effect of delaminations on opening of transverse matrix cracks. *J Compos Mater* 2005;39. <https://doi.org/10.1177/0021998305050427>.
- [42] Zhao L, Zhi J, Zhang J, Liu Z, Hu N. XFEM simulation of delamination in composite laminates. *Compos A Appl Sci Manuf* 2016;80. <https://doi.org/10.1016/j.compositesa.2015.10.007>.
- [43] Curiel-Sosa JL, Tafazzolimoghaddam B, Zhang C. Modelling fracture and delamination in composite laminates: Energy release rate and interfacial stress. *Compos Struct* 2018;189. <https://doi.org/10.1016/j.compstruct.2018.02.006>.
- [44] Carraro PA, Maragoni L, Quaresimin M. Characterisation and analysis of transverse crack-induced delamination in cross-ply composite laminates under fatigue loadings. *Int J Fatigue* 2019;129. <https://doi.org/10.1016/j.ijfatigue.2019.105217>.
- [45] Maragoni L, Carraro PA, Simonetto M, Quaresimin M. A novel method to include crack-induced delamination in a fatigue damage predictive procedure for composite laminates. *Compos Sci Technol* 2023;238. <https://doi.org/10.1016/j.compscitech.2023.110011>.
- [46] Zhu XK, Joyce JA. Review of fracture toughness (G, K, J, CTOD, CTOA) testing and standardization. *Eng Fract Mech* 2012;85. <https://doi.org/10.1016/j.engfractmech.2012.02.001>.
- [47] Ansys Mechanical Enterprise Academic Research (©2023 Ansys, Inc.) 2023.
- [48] Reference E. Release 2022 R1. Canonsburg, PA: ANSYS, Inc.; 2022.
- [49] Contact Technology Guide. Release 2022 R1. Canonsburg, PA: ANSYS, Inc.; 2022.
- [50] Pupurs A, Loukil MS, Marklund E, Varna J, Mattsson D. Transverse Crack Initiation in Thin-Ply Laminates Subjected to Tensile Loading at Low and Cryogenic Temperatures. *Mech Compos Mater* 2024;59. <https://doi.org/10.1007/s11029-023-10156-0>.
- [51] Pakkam Gabriel VR, Loukil MS, Varna J. Analysis of intralaminar cracking in 90-ply of GF/EP laminates with distributed ply strength. *J Compos Mater* 2021; 55. <https://doi.org/10.1177/00219983211027346>.
- [52] Varna J, Loukil MS, Pupurs A, Joffe R. Damage and failure Analysis for Composites. *Comprehensive Structural Integrity* 2023. <https://doi.org/10.1016/B978-0-12-822944-6.00020-7>.
- [53] McCartney LN, Blazquez A, Paris F. Energy-based delamination theory for biaxial loading in the presence of thermal stresses. *Compos Sci Technol* 2012;72. <https://doi.org/10.1016/j.compscitech.2012.06.015>.
- [54] Zubillaga L, Turon A, Renart J, Costa J, Linde P. An experimental study on matrix crack induced delamination in composite laminates. *Compos Struct* 2015;127. <https://doi.org/10.1016/j.compstruct.2015.02.077>.
- [55] Carreras L, Bak BLV, Turon A, Renart J, Lindgaard E. Point-wise evaluation of the growth driving direction for arbitrarily shaped delamination fronts using cohesive elements. *Eur J Mech A Solids A/Solids* 2018;;72. <https://doi.org/10.1016/j.euromechsol.2018.05.006>.
- [56] Carreras L, Renart J, Turon A, Costa J, Bak BLV, Lindgaard E, et al. A benchmark test for validating 3D simulation methods for delamination growth under quasi-static and fatigue loading. *Compos Struct* 2019;210. <https://doi.org/10.1016/j.compstruct.2018.12.008>.
- [57] Wang ASD. FRACTURE MECHANICS OF SUBLAMINATE CRACKS IN COMPOSITE MATERIALS. *Composites. Technol Rev* 1984;6. <https://doi.org/10.1520/ctr10817j>.

MICRO-DAMAGE EVOLUTION IN HYBRID CARBON/STEEL KNITTED AND WOVEN FABRIC COMPOSITES

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ABSTRACT

Composites with fabric reinforcements offer excellent energy dissipation and high damage tolerance as compared to long fiber unidirectional composites. Their mechanical behavior depends on local stress fields generated due to complex fabric meso-structure. Under loading, microdamage can initiate at multiple locations in this meso-structure. Thus, the design of fabric reinforcements should be carefully considered to achieve desired mechanical properties. In this work, hybrid fabric reinforcements are designed with carbon- and stainless steel fibers to be used at composite-metal joints in structural applications to achieve a smooth property gradient from the composite-side to the metal-side in such joints. Hybrid composite test specimens are subjected to quasi-static tensile loading and their properties are compared with non-hybrid carbon- and steel fiber reference materials. Their polished edges are inspected under optical microscope to detect sites of micro-damage initiation in the hybrid meso-structure. Damage propagation and corresponding stiffness reduction is also studied with increase in applied strain levels.

1 INTRODUCTION

Since the last few decades, there has been an increase in the use of knitted and woven fabric composites in aerospace and automotive industries due to their benefits over metals and UD composite tapes. Fabric composites offer ease in handling and greater flexibility in processing than UD tapes [1-8]. They also offer excellent damage tolerance at moderate costs as compared to UD composites and great weight savings at the same performance as compared to metals [2, 4-9]. However, due to significant crimp in the yarns [3, 7], modulus and strength are lower for fabric composites than UD composites.

Material properties of fabric composites are highly dependent on their textile meso-structure [1, 7-8]. It has been shown that meso-structure of fibrous yarns is the main deciding factor for nature of micro-damage than can initiate in these materials upon mechanical loading [7]. Between knitted and woven fabric composites, predicting mechanical behavior of knitted composites is more difficult due to large complexity in their meso-structure. Micro-damage initiation is governed by local stress state during loading. Knitted composites have a high degree of yarn curvature which generates a complex stress state [10]. This makes it difficult to predict the location of micro-damage initiation with the help of analytical models. Thus, conducting physical experiments and characterizing micro-damage becomes a necessity to study mechanical behavior of such materials in detail.

Previously, researchers have used various methods to detect microdamage in fabric composites [9], among which the prominent ones are described here briefly. The method of acoustic emission has been used to identify changes in damage modes by detecting change in the energy released while cracking. Digital image correlation has been used for strain mapping by detecting changes in speckle dot patterns on the surface of test specimens. The main drawback of these methods is they must be followed by X-ray or ultrasonic scanning for an accurate identification of damage mode. In the present work, we have used optical microscopy as the tool to study micro-damage by observing polished edges of test

specimens under optical microscope and to study nature of fiber-matrix interface by observing fracture surfaces under high-resolution scanning electron microscope.

Incorporation of metals into fiber composites started with fiber metal laminates (FMLs) which were prepared by interlacing metal layers with fiber reinforced resins or adhesives [11]. Such hybrids were designed to combine benefits of both the material classes and thereby having mechanical properties superior to composites and metals. But the weak interfacial bond between polymeric composites and metals would cause delaminations in these hybrid materials under fatigue loading. Various efforts have been made to increase the bond strength between composites and metals including adhesive bonding, bolting, and welding. Among these efforts, bolting creates stress concentrations and can also cause fiber breakage initiating further damage [12, 13]. Adhesive bonding requires higher costs and additional surface treatments for better lap shear strength [14]. Welding can be effective, but it becomes costly and time consuming with complex processing demands [13].

To counter these issues, as reported in our previous work [15], hybrid composites are developed with stainless steel fibers and carbon fibers gradually inter-woven or -knitted to form a complex fabric meso-structure reinforcing epoxy resin. Such materials, with mechanical properties intermediate between carbon-fiber epoxy (CF/EP) textile composite and steel-fiber epoxy (SF/EP) textile composite, would offer a smooth property gradient when used at a metal-composite joint. The reader is directed to our previous publication [15] to know details about manufacturing of fabrics and hybrid composites and pilot test results. In the current work, mechanical properties of these hybrids under quasi-static tensile loading and study of initiation and propagation of micro-damage are reported.

2 MATERIALS AND METHODS

In the present study, fabric reinforcements in composite materials were either knitted or woven. Plain weft knitting was performed on manual knitting machines by Brother (Japan). Weaving was performed on manual weaving machines in plain weave pattern. Yarns used were carbon fiber yarn T300B-3K-50B by Toray (France) and steel fiber yarn SY 11/1 by Fujibo (Japan) / Bekaert (Belgium). In the carbon fiber yarns, individual fiber diameter was 7 μm . In the case of steel fiber yarns, individual fiber diameter was 12 μm . CF/EP and SF/EP reference composite plates and hybrid CF/EP-SF/EP composite plates were manufactured by vacuum infusion of epoxy resin into the fabric reinforcements. The resin used was epoxy LY1564 with XB 3404-1 hardener by Huntsman (USA). The types of composite plates along with their notations and layouts are listed in Tables 1-2.

Notation	Type	Lay-up
C1	Reference – CF/EP	[(T300) ₈]
M1	Reference – SF/EP	[(SY 11/1) ₈]
M2	Reference – SF/EP	[(SY 11/1 3filaments) ₈]
H1	Hybrid	[(Hybrid SY11/1+T300) ₈]
H2	Hybrid	[(T300/SY11/1)] _{2s}

Table 1: Knitted composites

Notation	Type	Lay-up
C4	Reference – CF/EP	[T300 FR 1x] ₄
M4	Reference – SF/EP	[SY 11/1 2x] ₃
M5	Reference – SF/EP	[SY 11/1 2x] ₄
H5	Hybrid	[T300 FR 1x + SY 11/1 2x] ₄

H6	Hybrid	[SY 11/1 2x + T300 FR 1x] ₄
H7	Hybrid	[T300 FR 1x + SY 11/1 2x] ₅ + [SY 11/1 2x + T300 FR 1x] ₅
H8	Hybrid	[(SY 11/1 2x + T300 FR 1x)(T300 FR 1x + SY 11/1 2x)] _{2.5}
H9	Hybrid	[T300 FR 1x / SY 11/1 2x] _{2.5}

Table 2: Woven composites

Square-sized specimens with dimensions 25 mm x 25 mm (length and width) were cut from each plate for fiber volume fraction studies. Fiber weight fraction and volume fraction were measured by burning off matrix in a closed box-type furnace at 400°C. In this process, composite specimens were kept inside the furnace for 180 minutes, and their weights before and after the process were recorded. Weight fraction was converted to volume fraction with the help of individual yarn densities of carbon- and steel fiber yarn. The resultant fiber volume fractions for each composite type are shown in Fig. 1.

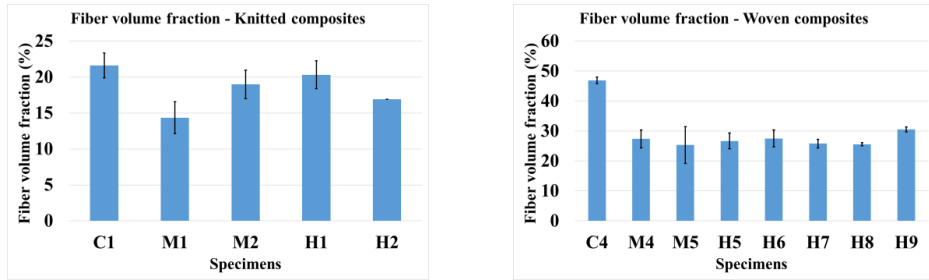
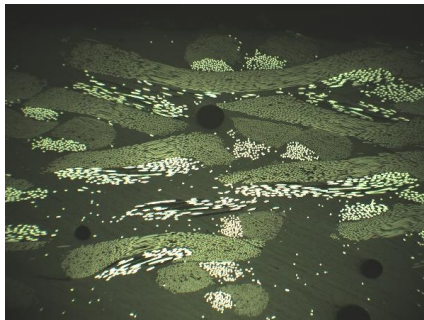
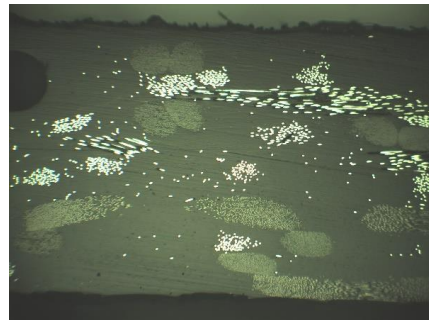


Figure 1: Fiber volume fraction – Knitted and woven composites

Rectangular specimens of 25 mm width and 200 mm length were cut from the composite plates for tensile testing and their edges were polished using manual polishing machine QATM Qpol 300 M2. The polished edges were observed under Motic 310Met optical microscope at 50x magnification to identify meso-structure of the fabric reinforcements. The meso-structures of hybrid composites are shown in Fig. 2-3.



(a)

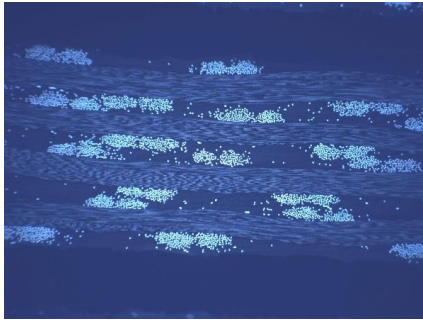


(b)

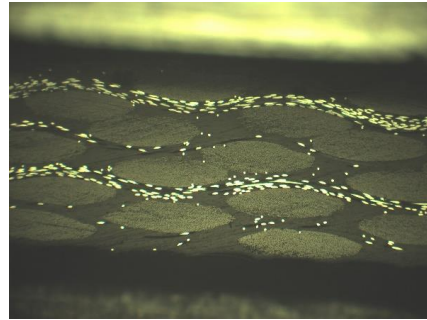
Figure 2: Meso-structures of knitted composites, (a) **H1** [8 layers, with each layer of hybrid yarns] and (b) **H2** [with 4 layers CF and 4 layers SF]. Note that fibers with darker shade are carbon fibers and fibers with lighter shade are steel fibers.

Reinforcement tabs were glued on both ends of rectangular test specimens using a two-component aerospace grade adhesive from Huntsman (USA). The specimens were tested in Zwick/Roell Z150 universal testing machine under quasi-static tensile loading. Non-contact-type video extensometer MESSPHYSIK ME 46-NG and contact-type extensometer HBM Strain Link DD1 were used for strain measurements.

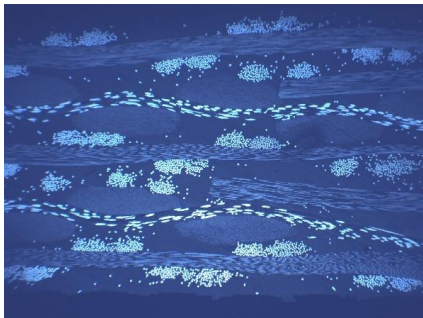
During quasi-static tensile loading, all the specimens were initially loaded up to 0.4% strain with 1.5 mm/min strain rate and then unloaded to zero strain with 2 mm/min strain rate. During this first loading-unloading cycle, the undamaged material modulus was measured. Following the undamaged modulus measurements, specimens were subjected to repetitive loading-unloading cycles with an increase of 0.2% in the maximum applied strain during each cycle. Between each test cycle, the polished edges were inspected in Motic 310Met optical microscope to detect presence of microdamage due to tensile loading. Stiffness was measured for each specimen during the unloading interval of each cumulative test cycle. The reduction in stiffness with increase in maximum applied strain was correlated with microdamage accumulation inside composite specimens. After final failure, fracture surfaces of failed specimens were observed under JEOL JSM IT-300 Environmental SEM to study similarities and differences in fiber fracture and nature of fiber-matrix interface for carbon- and steel fibers.



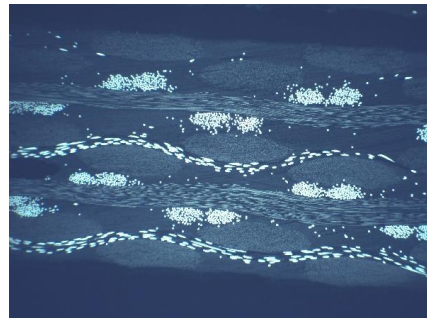
(a)



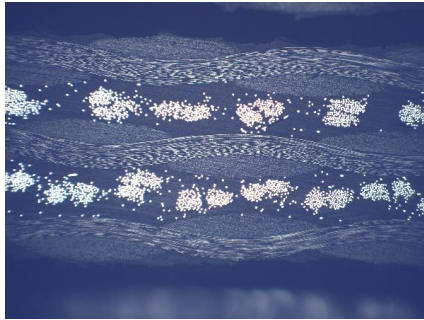
(b)



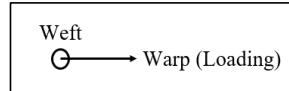
(c)



(d)



For all micrographs,



(e)

Figure 3: Meso-structures of woven composites, (a) **H5** [4 layers, with each layer of CF warp + SF weft], (b) **H6** [4 layers, with each layer of SF warp + CF weft], (c) **H7** [with 3 layers H5 & 2 layers H6], (d) **H8** [with 3 layers H6 & 2 layers H5] and (e) **H9** [with 3 layers CF & 2 layers SF]. Note that fibers with darker shade are carbon fibers and fibers with lighter shade are steel fibers.

3 RESULTS AND DISCUSSIONS

3.1 Elastic modulus measurements

The comparison of elastic moduli along the loading direction measured during the first loading-unloading cycle in quasi-static tensile testing is given in Fig. 4. In the case of knitted composites, H1 with hybrid CF-SF yarn has modulus intermediate to CF/EP reference materials C1 and SF/EP reference materials M1 and M2. H2 with alternate CF & SF layers has modulus equivalent to SF/EP reference materials. Comparing modulus values of H1 and H2, yarn hybridization gives better results than layer hybridization in the case of knitted composites.

In the case of woven composites, H5 with CF yarns in warp direction has modulus similar to CF/EP reference material C4. A similar relation can be found out with H6 and SF/EP reference materials, where H6 has SF yarns in warp direction. This clearly indicates that for hybrid woven composites, materials used in warp/weft direction almost entirely affect the material modulus in the respective directions. In the case of hybrid compositions H7-H9, both CF and SF were used in warp yarns. Thus, their moduli are intermediate to CF/EP and SF/EP reference materials. Comparing moduli of H7 and H8, the latter has two layers with CF warp in comparison with H7 having three layers of CF warp. This reflects in the results, with H8 having lower modulus in warp direction than H7.

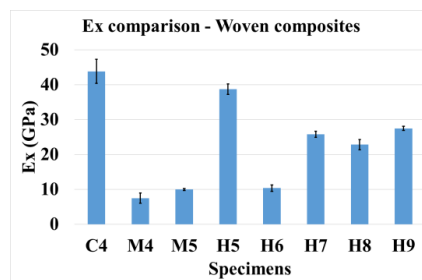
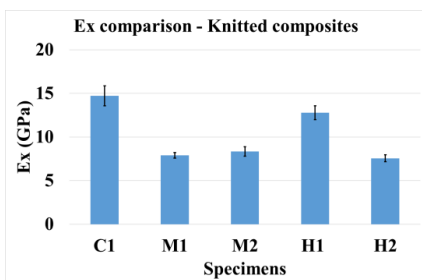


Figure 4: E (undamaged) along the loading direction – Knitted and woven composites

3.2 Micro-damage initiation and propagation – Knitted composites

Modulus measurements for undamaged materials were followed by further loading-unloading cycles with increasing levels of maximum applied strains. In the case of knitted composites, debondings in fiber-matrix interface started appearing at 0.6% maximum applied strain. Damage initiation in CF/EP reference materials as well as hybrid composites was characterized by transverse and shear cracks along weft yarn boundaries. This is shown in Fig. 5. Note that in yarn-hybrid H1, damage starts at CF yarn boundaries which then propagates along SF yarns at higher applied strain values (refer Fig. 6). Knitted hybrids show interconnecting cracks (refer Fig. 6) depicting damage propagation with increasing applied strain values.

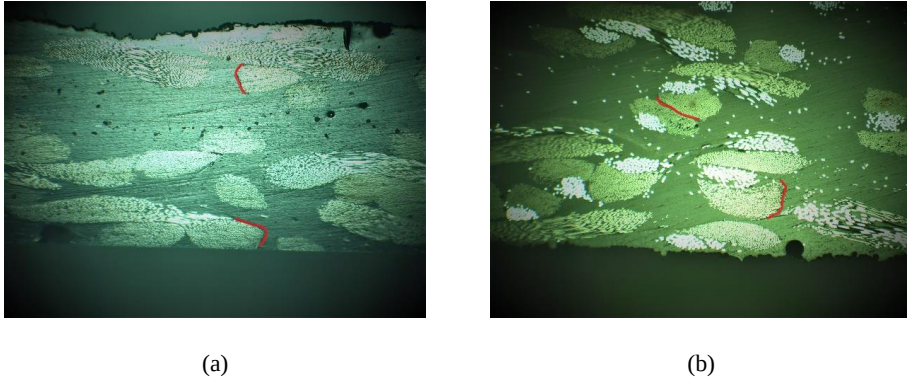


Figure 5: Damage initiation in knitted composites – (a) C1 (0.6% applied strain, 50x) and (b) H1(0.6% applied strain, 50x). Transverse and shear cracks are highlighted with red.

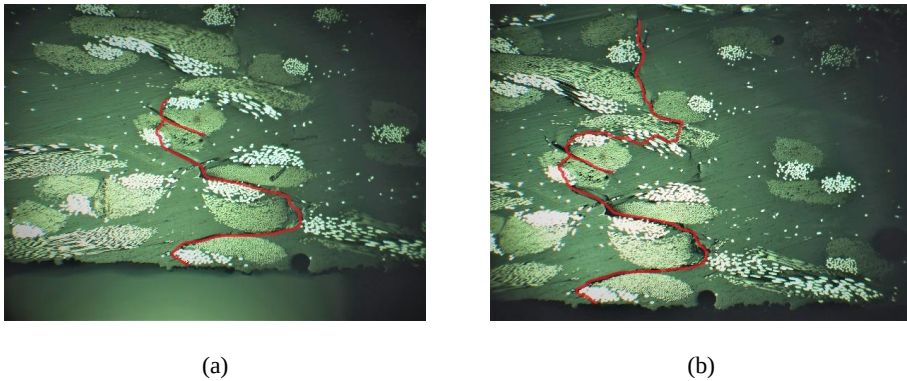


Figure 6: Damage propagation in knitted composites – (a) H1 (1.0% applied strain, 50x) and (b) H1 (1.2% applied strain, 50x). Transverse and shear cracks are highlighted in red.

3.3 Micro-damage initiation and propagation – Woven composites

In the case of woven composites, damage initiation was observed at 0.8% applied strain. In CF/EP

reference materials and in hybrid composites, transverse cracks appeared at CF weft yarn boundaries and shear cracks appeared through SF weft yarns. This is shown in Fig. 7. Fig. 8 shows damage propagation in CF/EP woven reference materials. An interesting observation is shown in Fig. 8(b), where local delaminations are connected by fiber breakage of the warp yarn.

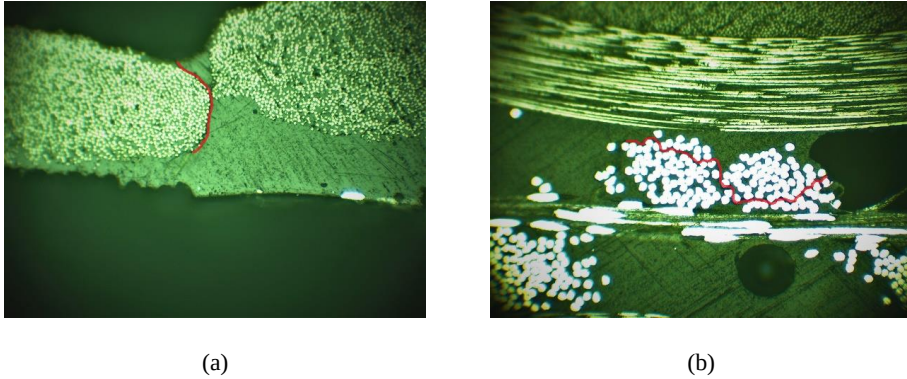


Figure 7: Damage initiation in woven composites – (a) H8 (0.8% applied strain, 200x) (b) H9 (0.8% applied strain, 200x). Transverse and shear cracks are highlighted in red.

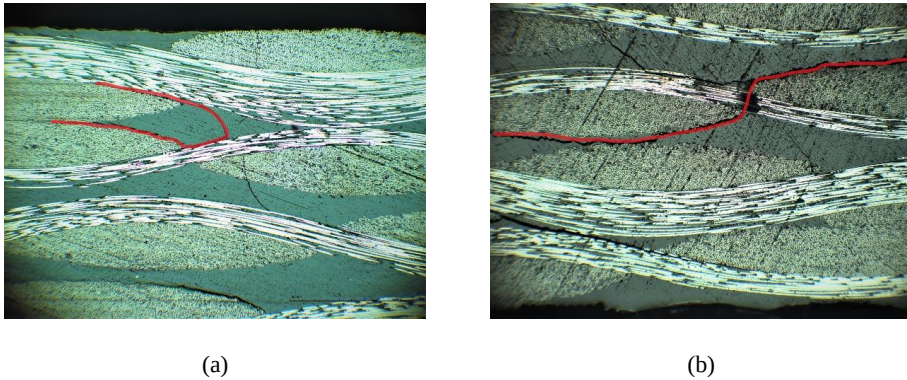


Figure 8: Damage propagation in woven composites – (a) C4 (local delamination growth constrained by warp yarn, 0.8% applied strain, 100x) and (b) C4 (delaminations connecting after fiber breakage, 0.8% applied strain, 100x). Delaminations are marked in red.

3.4 Stiffness reduction

One of the major consequences of micro-damage accumulation in textile composites is reduction in stiffness along the loading direction. Fig. 9 shows trends in E_x with increase in applied strain levels for knitted and woven composites. The values on the vertical axis are normalized according to the respective undamaged material modulus for each material (refer Fig. 4), so that trends in different materials can be easily compared. Stiffness reduction is not seen in CF/EP reference material C1. A non-noticeable stiffness reduction in CF fabric reinforced composite is also reported in [7] due to low mismatch in transverse stiffness values between polymer matrix and carbon fibers. A prominent decrease in stiffness is observed for SF/EP reference material M1, knitted hybrids (H1, H2) and woven hybrids (H5, H6).

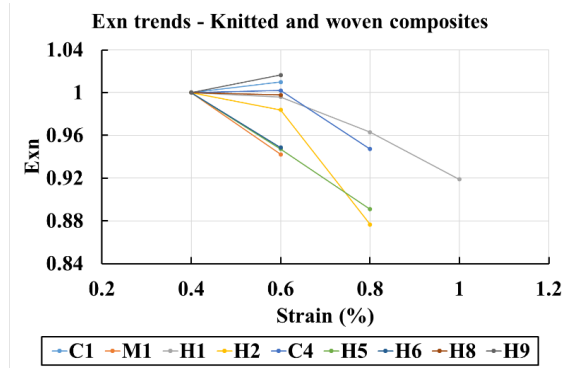
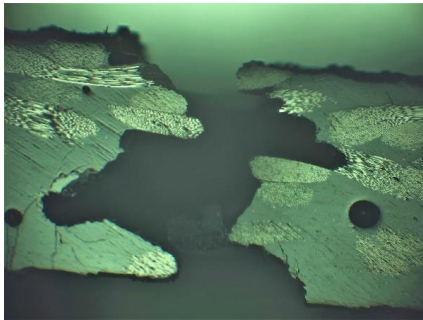


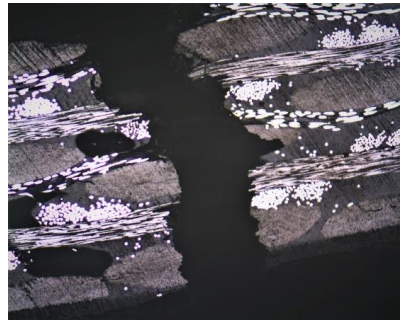
Figure 9: Ex trends for knitted and woven composites

3.5 Ultimate failure

Ultimate failure of knitted and woven specimens was characterized by multiple delaminations, and transverse cracks connected across layers through fiber breakage in warp yarns. This is depicted in Fig. 10, which shows fractured edges of C1 (knitted CF/EP reference material) and H8 (woven hybrid material). Fracture surfaces were also observed under scanning electron microscope. Fig. 11 shows SEM image of H9, which is a woven hybrid with alternate layers of CF and SF yarns. In Fig. 11, carbon- and steel fibers can be distinguished by their different sizes and shapes. Carbon fibers are smaller and round-shaped, whereas steel fibers are bigger, and polygon-shaped. The difference in the nature of fiber-matrix interfaces between carbon fibers and steel fibers is also noticeable.



(a)



(b)

Figure 10: Fractured edges after ultimate failure – (a) C1 (1.0% strain-to-failure, 50x) and (b) H8 (1.0% strain-to-failure, 50x)



Figure 11: SEM image of fracture surface of H9 (1.0% strain-to-failure)

3.6 Strength and strain-to-failure measurements

After ultimate failure of each specimen, its strength ($\sigma_{\max}$) along loading direction and strain-to-failure ($\epsilon_{\max}$) were measured. Fig. 12-13 show comparison of $\sigma_{\max}$ and $\epsilon_{\max}$ among different types of composites. $\sigma_{\max}$ comparison (refer Fig. 12) for knitted and woven composites shows a very similar trend to Ex comparison (refer Fig. 4). In the case of knitted composites, yarn-hybrid H1 has strength similar to CF/EP reference material C1 and layer-hybrid composite H2 has strength similar to SF/EP reference materials M1 and M2. In the case of woven composites, H5 with CF warp yarns has strength equivalent to CF/EP reference material C4, whereas H6 with SF warp yarns has strength equivalent to SF/EP reference materials M4 and M5.

$\epsilon_{\max}$ comparison (refer Fig. 13) shows that yarn-hybrid H1 has the highest strain-to-failure among all knitted and woven configurations. Such behavior signifies enhanced damage tolerance due to CF-SF yarn hybridisation. In woven configurations, all composite types have failure strains between 0.8% and 1.0% except SF/EP reference materials M4 and M5 and hybrid material H7. The cause of low failure strains in SF/EP reference materials can be the weak fiber-matrix interface which was observed in SEM images of fracture surfaces (refer Figure 11). Anomalous behavior of H7 was caused by macro-scale imperfections which arose during its manufacturing process. Efforts will be made to minimise such imperfections in the future work.

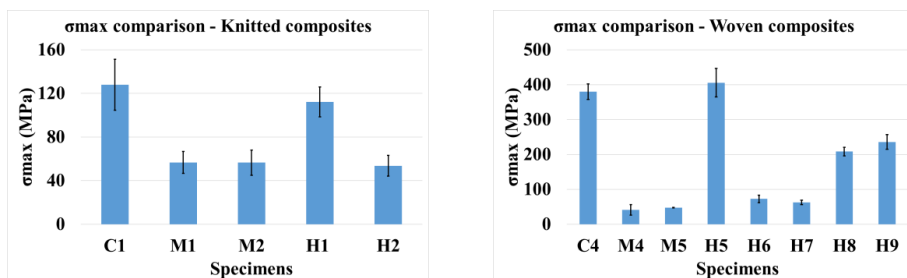
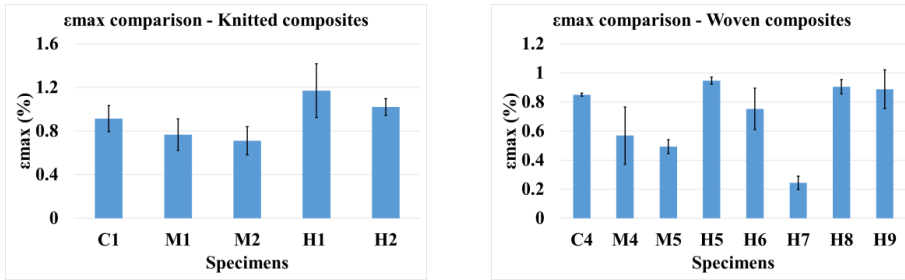


Figure 12: $\sigma_{\max}$ comparison – Knitted and woven composites

Figure 13: $\varepsilon_{\max}$ comparison – Knitted and woven composites

4 CONCLUSIONS

Multiple configurations of hybrid carbon/steel fabric reinforced epoxy composites were tested under quasi-static tensile loading and their mechanical properties were compared with non-hybrid reference materials. The tests were performed in repetitive loading-unloading cycles with increase in maximum applied strain during each cycle. Stiffness along loading direction was measured for undamaged materials and the reduction in stiffness was observed as a result of damage accumulation. Between each test cycle, polished edges of test specimens were observed under optical microscope to study microdamage initiation and propagation.

For knitted and woven composites, microdamage initiated in the form of transverse cracks at the yarn boundaries of carbon fiber yarns and in the form of shear matrix cracks through steel fiber yarns. Propagation of delaminations along weft yarn boundaries in woven composites was arrested by warp yarns. At macro-level, knitted composites with hybrid yarns showed promising results in terms of strength and stiffness when compared with knitted composites with hybrid layers. Woven hybrid composites with carbon fibers as warp yarns had better properties than hybrids with steel fibers as warp yarns.

ACKNOWLEDGEMENTS

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REFERENCES

- [1] U. Leischner and A. F. Johnson, Micromechanics analysis of hybrid woven fabric composites under tensile and compression load, *WIT Transactions on Engineering Sciences*, **4**, 1970, pp. 397-405 (doi: [10.2495/CP940451](https://doi.org/10.2495/CP940451)).
- [2] Ph. Vandeurzen, J. Ivens and I. Verpoest, A three-dimensional micromechanical analysis of woven-fabric composites: I. Geometric analysis, *Composites Science and Technology*, **56**(11), 1996, pp. 1303-1315 (doi: [10.1016/S0266-3538\(96\)00092-9](https://doi.org/10.1016/S0266-3538(96)00092-9)).

- [3] Ph. Vandeurzen, J. Ivens and I. Verpoest, A three-dimensional micromechanical analysis of woven-fabric composites: II. Elastic analysis, *Composites Science and Technology*, **56(11)**, 1996, pp. 1317-1327 (doi: [10.1016/S0266-3538\(96\)00091-7](https://doi.org/10.1016/S0266-3538(96)00091-7)).
- [4] B.H. Le Page, C.I.C. Manger, F.J. Guild, S.L. Ogin and P.A. Smith, Modelling effect of layer shift on properties of woven fabric composites, *Plastics, Rubber and Composites*, **31(9)**, 2002, pp. 385-391 (doi: [10.1179/146580102225006413](https://doi.org/10.1179/146580102225006413)).
- [5] B.H. Le Page, F.J. Guild, S.L. Ogin and P.A. Smith, Finite element simulation of woven fabric composites, *Composites Part A: Applied Science and Manufacturing*, **35(7-8)**, 2004, pp. 861-872 (doi: [10.1016/j.compositesa.2004.01.017](https://doi.org/10.1016/j.compositesa.2004.01.017)).
- [6] K. Kim, J.I. Curiskis, L. Ye, S. Fu, Mode-I interlaminar fracture behaviour of weft-knitted fabric reinforced composites, *Composites Part A: Applied Science and Manufacturing*, **36(7)**, 2005, pp. 954-964.
- [7] L. Gorbatikh and S.V. Lomov, *Damage accumulation in textile composites*, Modeling Damage, Fatigue and Failure of Composite Materials, In Woodhead Publishing Series in Composites Science and Engineering, Woodhead Publishing, Cambridge, 2016.
- [8] L. Wang, J. Wu, C. Chen, C. Zheng, B. Li, S.C. Joshi and K. Zhou, Progressive failure analysis of 2D woven composites at the meso-micro scale, *Composite Structures*, **178**, 2017, pp. 395-405 (doi: [10.1016/j.compstruct.2017.07.023](https://doi.org/10.1016/j.compstruct.2017.07.023)).
- [9] S.V. Lomov, D.S. Ivanov, T.C. Truong, I. Verpoest, F. Baudry, K. Vanden Bosche and H. Xie, Experimental methodology of study of damage initiation and development in textile composites in uniaxial tensile test, *Composites Science and Technology*, **68(12)**, 2008, pp. 2340-2349 (doi: [10.1016/j.compscitech.2007.07.005](https://doi.org/10.1016/j.compscitech.2007.07.005)).
- [10] M. Ravandi, A. Moradi, S. Ahlquist and M. Banu, Numerical Simulation of the Mechanical Behavior of a Weft-Knitted Carbon Fiber Composite under Tensile Loading, *Polymers*, **14(3)**, 2022, 451
- [11] Q. Liu, J. Ma, L. Kang, G. Sun and Q. Li, An experimental study on fatigue characteristics of CFRP-steel hybrid laminates, *Materials & Design*, **88**, 2015, pp. 643-650, (doi: [10.1016/j.matdes.2015.09.024](https://doi.org/10.1016/j.matdes.2015.09.024)).
- [12] J. Jahn, M. Weeber, J. Boehner and R. Steinhilper, Assessment Strategies for Composite-metal Joining Technologies – A Review, *Proceedings of the 26th CIRP Design Conference 2016 Vol. 50* (Eds. L. Wang, T. Kjellberg), Stockholm, Sweden, June 15-17, 2016, Elsevier B.V., 2016, pp. 689-694.
- [13] A.M. Joesbury, P.A. Colegrove, P.V. Rymenant, D.S. Ayre, S. Ganguly and S. Williams, Weld-bonded stainless steel to carbon fibre-reinforced plastic joints, *Journal of Materials Processing Technology*, **251**, 2018, pp. 241-250 (doi: [10.1016/j.jmatprotec.2017.08.023](https://doi.org/10.1016/j.jmatprotec.2017.08.023)).
- [14] M. Pawlik, L.Y.Y. Cheah, U. Gunpath, H. Le, P. Wood and Y. Lu, Surface engineering of carbon fibre/epoxy composites with woven steel mesh for adhesion strength enhancement, *International Journal of Adhesion and Adhesives*, **114**, 2022, 103105 (doi: [10.1016/j.ijadhadh.2022.103105](https://doi.org/10.1016/j.ijadhadh.2022.103105)).
- [15] A. Pupurs, I. Bramane, I. Łaśenko, M. Maniņš, M. Irbe and A. Šņepts, Manufacturing and Experimental Testing of Composite/Metal Joints with Textile Mesostructure, *ECCM 2022 - Proceedings of the 20th European Conference on Composite Materials: Composites Meet Sustainability Vol. 2* (Eds. A.P. Vassilopoulos, V. Michaud), Lausanne, Switzerland, June 26-30, 2022, Ecole Polytechnique Fédérale de Lausanne (EPFL), Lausanne, 2022, pp. 540-546.

FE MODELING OF DAMAGE DEVELOPMENT IN HYBRID CARBON/STEEL FIBER TEXTILE COMPOSITES USING FRACTURE MECHANICS AND COHESIVE ZONE APPROACH

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Keywords: Metal-composite joints, Textile composites, Mechanical testing, Finite element modeling, Damage development

Abstract

Joining of dissimilar materials such as metals and polymeric composites using bolting, riveting or adhesive joining results in abrupt change in mechanical properties at the joints causing stress concentration. Structural health of a hybrid metal-composite structure can be compromised by factors such as environmental degradation of adhesive or galvanic corrosion at the metal-composite joints. To solve these issues, metal-composite joints with textile mesostructure have been developed with hybrid carbon/steel fiber textile composites. The hybrid composites are tested under quasi-static tensile loading to study macro-mechanical properties and evolution of microdamage. Two configurations of woven hybrid composites are chosen for FE modeling of damage development. 3D representative volume elements are built such that resin-impregnated yarns are modelled as homogeneous orthotropic materials, and yarn/matrix-interface behavior is characterized by cohesive zone material modeling. During mechanical test simulation, stress-strain curves are plotted and damage schematics inside yarns and matrix are obtained. The FEM predictions match the experimental results of microdamage evolution, but they underestimate the experimental tensile strength values for hybrid composite configuration with stainless steel-fiber yarn in the warp-direction and carbon-fiber yarn in the weft-direction.

1. Introduction

Recent efforts to combine the use of metals and polymer-matrix composites in aerospace structures have led to the necessity of suitable metal-composite joining techniques. Jahn et al. [1] and Pramanik et al. [2] have reviewed the existing metal-composite joining techniques, among which the most common techniques are adhesive bonding and mechanical joining (bolting, riveting etc.). Although these techniques are effective in joining metal- and composite-parts, they have some serious drawbacks in terms of mechanical properties of joints. Adhesive metal-composite bonding results in an abrupt change of mechanical properties from the metal-side to the composite-side and the progressive environmental degradation of adhesive joints over time can be an issue [2]. In case of mechanical metal-composite joining using bolts or rivets, in addition to the abrupt change of mechanical properties and consequently high stress concentration at the joints, very poor notched strength of composite laminates as compared to their unnotched strength poses a serious threat to structural integrity of hybrid metal-composite structures [3].

While joining two dissimilar materials, their galvanic corrosion potential is an important factor which needs to be taken into consideration. This is critical while designing a metal-composite joint. In the case

of joining Al-alloys with carbon-fiber reinforced polymers (CFRP), Al-alloys must be coated at the joints, as the two materials have high galvanic corrosion potential [2,4,5]. On the contrary, alloys such as stainless steel are compatible with CFRP [5], and they can be joined together without any need of surface coating. In literature, there are multiple studies regarding CFRP-stainless steel hybrid stack-up or sandwich structures [6-8], where excellent joining is achieved using methods such as co-curing and weld-bonding.

In a similar study, Pupurs et al. [9] have recently developed CFRP-stainless steel hybrid joints with textile meso-structure. It is shown that epoxy-based hybrid composites with interwoven carbon-fiber yarns and stainless steel-fiber yarns possess mechanical properties intermediate to non-hybrid epoxy-based carbon-fiber and steel-fiber composites. The assembly of hybrid carbon/steel fiber textile composite with non-hybrid composites on either sides can act as a suitable metal-composite joint with gradual property change from the metal-side to the composite side.

In a follow-up study, Kulkarni et al. [10] have analyzed micro-damage evolution in hybrid carbon/steel fiber textile composites using experimental approach. A detailed description of possible damage modes in non-hybrid textile composites can be found in [11]. Similar damage modes have been reported in [10] for hybrid textile composites. Numerical study of mechanical behavior of hybrid carbon/steel fiber textile composites has been conducted in [12], where finite element models of the hybrid composites are developed, and their elastic properties are compared with the experimental values reported in [9].

The present work focuses on finite element modeling of microdamage in hybrid carbon/steel fiber textile composites. To evaluate damage inside fiber yarns and matrix, strength-based criteria are used. The yarn/matrix-interface is modelled with contact elements and yarn-matrix debonding is studied using cohesive zone material (CZM) modeling.

2. Materials and methods

Fabric reinforcements for the hybrid carbon/steel fiber textile composites were manufactured using T300 carbon-fiber yarns from Toray (Japan) and SY 11/1 stainless steel-yarns from Fujibo (Japan) and Bekaert (Belgium). The carbon-fiber yarns had a tow size of 1K with 7 μm individual filament diameter. The stainless steel-fiber yarns were 91 tex with 12 μm individual filament diameter. Two types of fabrics were manufactured - i) woven fabrics with plain weave pattern using manual weaving loom, and ii) knitted fabrics with plain weft knit pattern using manual knitting machine. Composite manufacturing was carried out using vacuum infusion of epoxy LY1564 resin (with XB 3404-1 hardener) into stacked dry fabric layers. The composite plates were oven-cured at 80 $^{\circ}\text{C}$ for 8 hours. Multiple configurations of knitted and woven composites with different meso-structures were manufactured. Composite specimens were tested under quasi-static tensile loading according to ASTM D3039 standard, and properties such as elastic modulus, tensile strength and strain-to-failure were obtained for each hybrid configuration. Additionally, micro-damage evolution with increasing level of applied strain was also studied. A detailed list of woven and knitted hybrid composite configurations, along with their mechanical properties obtained using experimental methods, can be found in [10].

For numerical analysis of mechanical properties and micro-damage evolution, two configurations among woven hybrid composites were chosen, namely, i) H5, with steel-fiber yarn in weft-direction and carbon-fiber yarn in warp-direction, and ii) H6, with carbon-fiber yarn in weft-direction and steel-fiber yarn in warp-direction. A schematic representation of the meso-structures of the two configurations is given in Fig. 1.

For FEM analysis, 3D representative volume elements (RVEs) for H5 and H6 were constructed in SolidWorks software and were imported in ANSYS Mechanical FEM software to generate element mesh and to apply boundary conditions. The curvature of warp yarns was modelled using a simple sine-wave function, whereas weft yarns were assumed to be straight. The yarns were embedded in a cuboidal matrix block, on which the relevant boundary conditions were applied. As shown Fig. 2, uniform displacement U_x along warp yarn-direction (X -axis) was applied on the right-face of the matrix block.

Symmetry boundary conditions were applied on the top-, bottom-, front- and left-faces, and U_z displacement coupling was applied at the back-face of the matrix block, allowing free contraction according to Poisson's effect along weft yarn-direction (Z -axis). Warp yarn-curvature and yarn cross-sections were calibrated according to optical microscopy images from the experimental results. Additional calibrations were performed by comparing elastic properties of the RVEs with the elastic properties obtained from the experimental results. The resin-impregnated carbon-fiber and stainless steel-fiber yarns were considered as homogeneous orthotropic materials with properties given in Table 1. Epoxy matrix was modelled as an isotropic material with Young's modulus $E = 3 \text{ GPa}$ and Poisson's ratio $\nu = 0.29$.

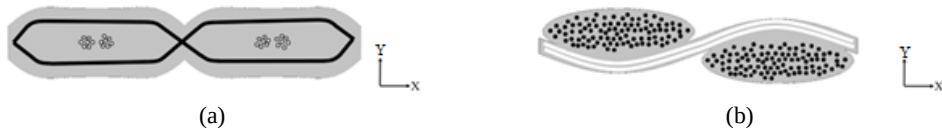


Figure 1. Schematic representation of (a) H5 and (b) H6. Carbon-fiber yarns are shown in black, and steel-fiber yarns are shown in white. Yarns along X -axis are warp yarns and yarns perpendicular to X -axis (along Z -axis, not shown in the figure) are weft yarns.

Table 1. Material properties of resin-impregnated fiber yarns.

Material property	Notation	Unit	Carbon-fiber yarn	Stainless steel-fiber yarn
Longitudinal modulus	E_L	GPa	139.21	105.00
Transverse modulus	E_T	GPa	7.52	105.00
Longitudinal Poisson's ratio	ν_{LT}	-	0.22	0.30
Transverse Poisson's ratio	ν_{T3}	-	0.40	0.30
In-plane shear modulus	G_{LT}	GPa	3.43	40.38
Out-of-plane shear modulus	G_{T3}	GPa	2.69	40.38

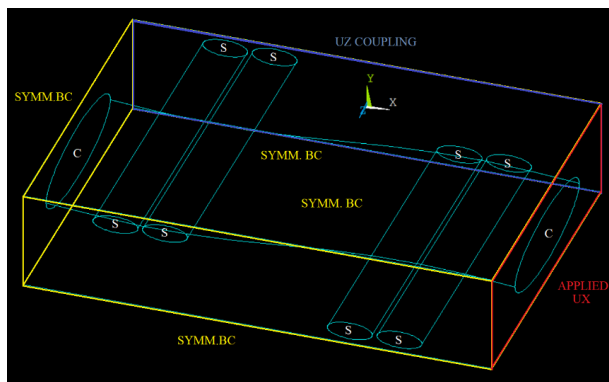


Figure 2. Representative Volume Element (RVE) of composite configuration H5. 'S' denotes steel-fiber yarn and 'C' denotes carbon-fiber yarn. Similar boundary conditions were applied for RVE of composite configuration H6.

During FE simulation, displacement U_x corresponding to 0.01% strain was applied at first time step. To study the quasi-static tensile test response of the FE model, value of applied strain was incremented by 0.01% at each subsequent time step until the model underwent final failure. To identify damage initiation sites inside fiber yarns and matrix, material strength limits were defined, and maximum stress criterion was used. To simulate progressive damage evolution, instant material stiffness reduction method was used such that material stiffness was reduced by 5% at every damage initiation instance. For analysis of yarn/matrix-interface debonding, contact elements were applied at the interface and CZM modeling was used. Bilinear material behavior with linear softening in the cohesive zone was defined by the values of maximum traction and critical energy release rates for normal separation and tangential slipping. Using the FEM solutions, stress-strain curves for H5 and H6 were plotted, and the schematics of i) damage status inside yarns and matrix and ii) contact status at yarn/matrix-interface were obtained.

3. Results and discussion

3.1. Stress-strain curves

To study mechanical behavior of FE models at the macro-scale, stress-strain curves were plotted for the two chosen composite configurations, and they were compared with the experimental results. Fig. 3 shows the comparison between experimental- and FEM stress-strain curves for H5 and H6.

In the case of composite configuration H5, FEM predictions for tensile strength and strain-to-failure in the warp direction are higher than the experimental values. The FEM model of H5 shows lower stiffness than experimental observation, but the difference in modulus values is small compared to the difference in FEM- and experimental modulus values for H6. For the FEM model of H6, the discrepancy in the use of input material properties for warp yarn (in this case, resin-impregnated stainless steel-fiber yarn) might have caused such a large difference in FEM- and experimental elastic modulus and tensile strength values along the warp-direction. Here, factors such as yarn twist, irregular fiber shapes and non-uniform fiber volume fraction should be considered, which cannot be replicated in the FEM analysis.

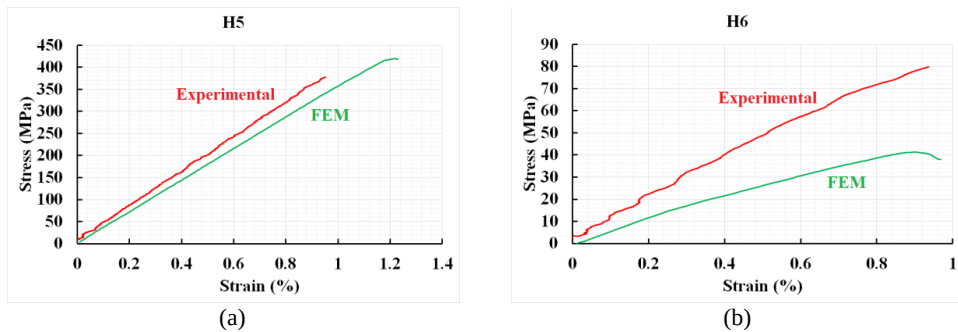
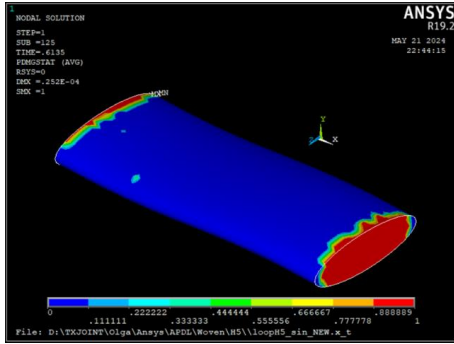


Figure 3. Stress-strain curves for composite configurations (a) H5 and (b) H6.

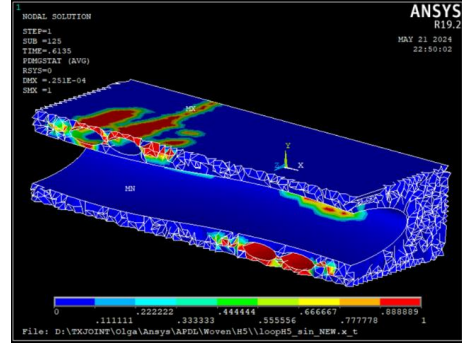
3.2. Microdamage studies

During quasi-static tensile loading simulation, microdamage inside yarns and matrix and debonding at the yarn/matrix-interface was monitored at each time step until final failure. Figs. 4 and 5 compare FEM damage status for yarn and matrix with experimental observations at the instance of final failure for H5 and H6. In composite configuration H5, the irregular damage of carbon-fiber warp yarns can be seen in Figs. 4(a) and 4(c). The matrix damage around weft yarns in FE model (refer Fig. 4(b)) is comparable with the matrix damage in optical microscopy image shown in Fig. 4(d). In composite configuration H6, partially damaged warp yarn/matrix-interface seen in Fig. 5(a) can be correlated with the yarn/matrix-interface in SEM fracture surface image in Fig. 5(c). The matrix damage surrounding warp yarns i.e.,

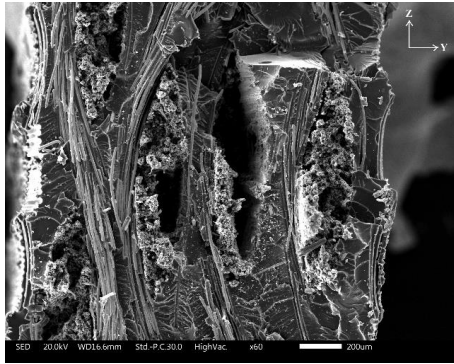
delaminations, can be seen in FE model (refer Fig. 5(b)) as well as in optical microscopy image (refer Fig. 5(d)). In terms of yarn/matrix-interface strengths, Figs. 4(c) and 5(c) depict that carbon-fiber yarn/epoxy-matrix interface is much weaker than the stainless-steel fiber yarn/epoxy-matrix interface.



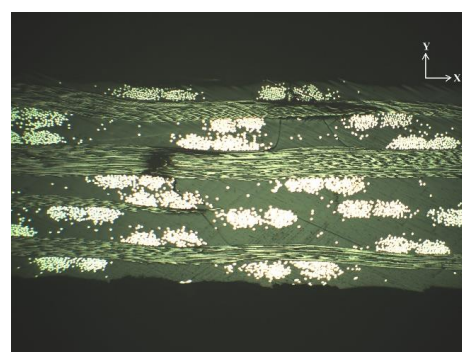
(a)



(b)



(c)



(d)

Figure 4. Schematics of microdamage (at the instance of final failure) for composite configuration H5, (a) FEM warp-yarn damage, (b) FEM matrix damage, (c) SEM image of fracture surface, and (d) Optical microscope image of polished edge (For FEM damage schematics, 0 = No damage, 1 = Full damage).

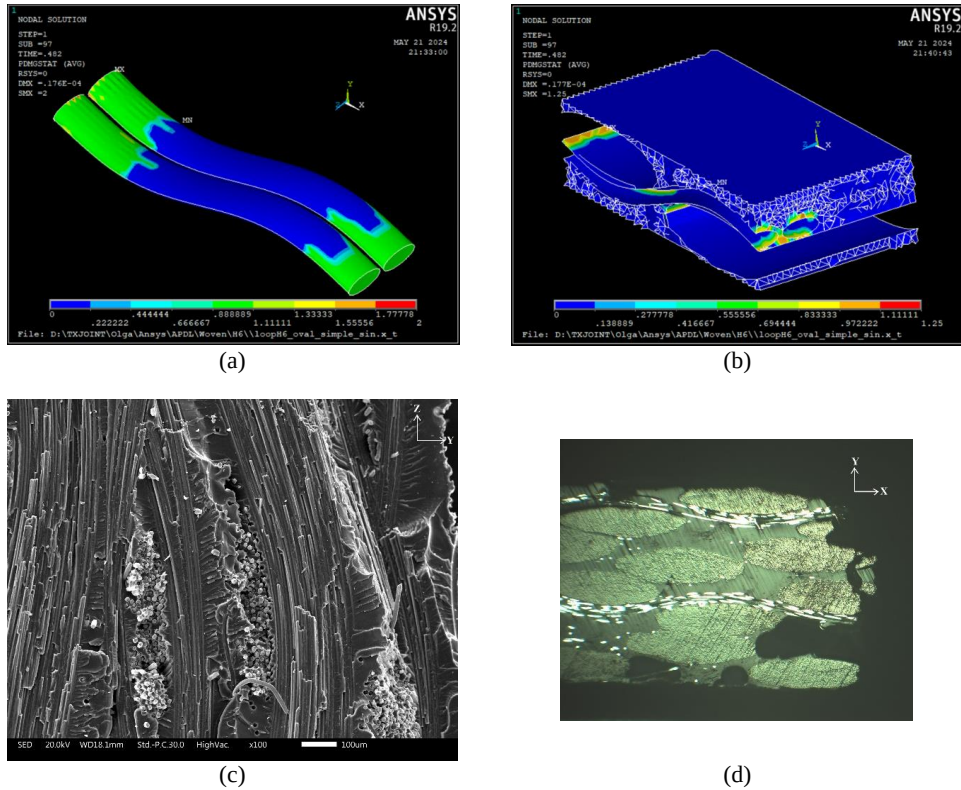


Figure 5. Schematics of microdamage (at the instance of final failure) for composite configuration H6, (a) FEM warp-yarn damage, (b) FEM matrix damage, (c) SEM image of fracture surface, and (d) Optical microscope image of polished edge (For FEM damage schematics, 0 = No damage, 1 = Partial damage, 2 = Full damage).

4. Conclusions

Damage development in hybrid carbon/steel fiber textile composites was studied using FE modeling. Representative volume elements were constructed for two configurations of hybrid woven composites and their mechanical response under quasi-static tensile loading was studied. To identify damage initiation sites inside fiber yarns and matrix, strength limits were defined, and maximum stress criterion was applied. Yarn/matrix-interface debonding was simulated by defining contact elements at the interface and by applying cohesive zone properties with bilinear material law. The macro-mechanical response of the hybrid composite FE models was captured by plotting stress-strain curves. For hybrid composite with stainless-steel fiber yarn in warp direction and carbon-fiber yarn in weft direction, FE model underestimates the experimental stiffness and strength values along warp direction by 50%. In terms of micro-damage evolution, FE models give accurate prediction of damage initiation sites and damage in fiber yarns and matrix at the instance of final failure.

Acknowledgments

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strategic fields of specialization” of the Specific Objective 8.2.2 “To Strengthen Academic Staff of Higher Education Institutions in Strategic Specialization Areas” of the Operational Programme “Growth and Employment”.

References

- [1] J. Jahn, M. Weber, J. Boehner and R. Steinhilper. Assessment Strategies for Composite-metal Joining Technologies - A Review. In *Procedia CIRP*, volume 50, pages 689-694, 2016.
- [2] A. Pramanik, A. Basak, Y. Dong, P. Sarker, M. Uddin, G. Littlefair, A. Dixit and S. Chattopadhyaya. Joining of carbon fibre reinforced polymer (CFRP) composites and aluminium alloys – A review. *Composites Part A: Applied Science and Manufacturing*, 101:1-29, 2017.
- [3] M. Saleh, Y. Wang, A. Yudhanto, A. Joesbury, P. Potluri, G. Lubineau and C. Soutis. Investigating the Potential of Using Off-Axis 3D Woven Composites in Composite Joints’ Applications. *Applied Composite Materials*, 24:377–396, 2017.
- [4] A.M. Joesbury. *New Approaches to Composite Metal Joining (Doctoral Thesis)*. Cranfield University, Cranfield, 2015.
- [5] B. Hannemann, S. Backe, S. Schmeer, F. Balle and U.P. Breuer. New Multifunctional Hybrid Polymer Composites Reinforced by Carbon and Steel Fibers. In *Proceedings of the 20th International Conference on Composite Materials*, 2015.
- [6] Q. Liu, J. Ma, L. Kang, G. Sun and Q. Li. An experimental study on fatigue characteristics of CFRP-steel hybrid laminates. *Materials and Design*, 88:643-650, 2015.
- [7] A.M. Joesbury, P.A. Colegrove, P.V. Rymenant, D.S. Ayre, S. Ganguly and S. Williams. Weld-bonded stainless steel to carbon fibre-reinforced plastic joints. *Journal of Materials Processing Technology*, 251:241-250, 2018.
- [8] Y. Shibuya, J. Zhang, Y. Sato and J. Yanagimoto. Evaluation of the mechanical properties and deformability of metal-based composite sheets made of thin stainless-steel sheets and carbon fiber reinforced plastics. *International Journal of Material Forming*, 15: 47, 2022.
- [9] A. Pupurs, I. Bramane, I. Ļašenko, M. Maniņš, M. Irbe and A. Šņepts. Manufacturing and Experimental Testing of Composite/Metal Joints with Textile Mesostructure. In *ECCM 2022 - Proceedings of the 20th European Conference on Composite Materials: Composites Meet Sustainability*, volume 2, pages 540-546, 2022.
- [10] A.N. Kulkarni, A. Pupurs, A. Šņepts and M. Irbe. Micro-damage Evolution in Hybrid Carbon/Steel Knitted and Woven Fabric Composites. In *Proceedings of the 23rd International Conference on Composite Materials*, 2023.
- [11] L. Gorbatikh and S. Lomov. Damage accumulation in textile composites. In *Modeling Damage, Fatigue and Failure of Composite Materials*. Woodhead Publishing, pages 41-59, 2016.
- [12] O. Kononova, V. Jevstignejevs, A. Pupurs and A.N. Kulkarni. Modeling Micro-damage Development in Functionally Graded Knitted and Woven Carbon/Steel Fibre Composites. In *ECCM 2022 - Proceedings of the 20th European Conference on Composite Materials: Composites Meet Sustainability*, volume 4, pages 298-305, 2022.

Parametric study of micro-damage evolution in woven hybrid carbon/stainless steel fiber composites

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Abstract

Under tensile loading, thermoset composites with brittle fibers, such as carbon or glass fibers, fail in a brittle manner, exhibiting little to no plastic deformation before final failure. In such cases, hybridization with ductile fibers, such as stainless-steel fibers, can be beneficial to achieve a significant amount of ductility. In engineering applications where metallic components and polymer composite material components are used together, composite-metal joints can exhibit high stress concentrations due to the abrupt property change. A hybrid material with mechanical properties intermediate between those of metals and polymer composites can provide a gradual property change at the joints. In the present work, hybrid carbon/steel fiber textile composites with various configurations were manufactured, and their mechanical stiffness and damage evolution properties under uniaxial quasi-static tensile loading have been investigated. Micro-damage evolution in hybrid composites was experimentally studied with optical microscopy. Fracture surfaces were analyzed with scanning electron microscopy showing a brittle failure of carbon fiber warp yarns with fiber cluster breaks, while for steel fiber warp yarns local plastic yielding was found.

Experimental study was supported and expanded by a thorough parametric analysis of competing damage mechanisms in various configurations of composites using 3D finite element modeling of representative volume elements. Continuum damage mechanics-based constitutive models were implemented in Abaqus software and significance of the main constituent damage parameters was investigated.

The study demonstrates how different hybridization strategies combining carbon and stainless-steel fibers can lead to different sequences of micro-damage mechanisms which in turn can lead to more ductile or brittle behavior of the composite.

Keywords

Hybrid textile composites, stainless steel fibers, composite-metal joints, micro-damage evolution, FEM multi-scale modeling.

1. Introduction

Carbon fiber-reinforced polymer (CFRP) composites are renowned for their exceptional stiffness and strength, combined with a low density which makes them highly suitable for lightweight applications such as aerospace and automotive. Although they exhibit extremely high tensile strength along the fiber direction, they typically fail in a brittle manner at low failure strain with little or no indication of stiffness reduction in the stress-strain response. Thus, engineering applications with CFRPs require use of high safety factors in their design [1].

There are multiple ways to impart ductility in polymeric composites. Fuller and Wisnom [2] achieved significant non-linearity in the tensile response of thin-ply $\pm 45^\circ$ angle-ply CFRP laminates with strain-to-failure up to 20%. The pseudo-ductility was achieved with the help of fiber reorientation under tensile loading. The authors also showed in [3] that the extent of ductility increases significantly in the thin-ply angle-ply CFRPs with the increase in initial fiber orientation angle with respect to the loading direction. Another way to increase ductility in fiber-reinforced polymer composites (FRPs) is to use ductile fibers as reinforcements instead of brittle carbon or glass fibers. Callens et al. [4] used unidirectional (UD) ductile steel fiber reinforcement in composites with brittle epoxy matrix, and they achieved a significant ductility with failure strain values up to 7%. Allaer et al. [5] used the same steel fibers as in [4], but in composites with ductile epoxy matrix, to increase the tensile failure strain values even further up to 19%. Polymer fibers, such as polypropylene fibers, and natural fibers, such as flax or regenerated cellulose fibers, are also commonly used to manufacture composites with notable ductility [6].

It has been shown that the method of fiber hybridization can also be used to increase ductility in polymer composite materials without a significant reduction in their stiffness and strength properties [7].

Hayashi reported the “hybrid effect” for the first time in 1972, when he showed a 40% increase in the failure strain of carbon fiber layers in carbon/glass fiber hybrid composites as compared to the carbon fiber non-hybrid composites [8,9]. In the following years, a similar hybrid effect was observed by several researchers for carbon/glass and carbon/Kevlar fiber hybrid composites [10–12]. In 1977, Zweben defined the term “hybrid effect” as the ratio between failure strains of the fiber-hybrid composite with low elongation (LE) and high elongation (HE) fibers and the non-hybrid composite with LE fibers [11]. The studies in [10–12] achieved a hybrid effect of up to 150% (strain to failure increases by 50%) according to Zweben’s definition. In the following year, a new definition of the hybrid effect was coined, referring to the ‘positive’ or ‘negative’ deviation in the mechanical properties of hybrid composites from the rule of mixtures consideration [13,14].

As reported in [6,15], the major reason for a positive hybrid effect is the series of non-catastrophic subcritical failures of LE fibers before the ultimate failure in a LE/HE hybrid composite.

Some common examples of fiber hybridization in polymer composite materials for the improvement of specific mechanical properties include i) hybridizing woven glass fiber composite skins of windsurf boards with carbon fibers to improve stiffness and strength, ii) hybridizing woven carbon fiber composite F1 car frame with polypropylene fibers to improve the toughness and crash resistance [16], iii) hybridizing glass fiber composite wind turbine blades with carbon fibers to increase weight savings [17], and iv) using flax fibers in a carbon fiber composite bicycle frame to improve the vibration damping behavior [18].

Considering the applications of CFRPs in the aerospace industry, there are multiple advantages of using hybrid or non-hybrid carbon fiber composites in the aerostructures, the primary advantage being their weight-saving capability. It has been found that CFRP wings are between 34 – 40% lighter than the equivalent metallic wings, resulting in up to 8% savings in take-off gross-weight and total fuel burn for a commercial passenger aircraft [19]. Also, the CFRP components comprise of significantly smaller number of parts than the metallic components, which results in the reduction of the required number of structure maintenance tasks [20]. In addition, the reinforcement meso-structures in CFRP components can be tailored specifically to increase the microwave absorbent characteristics, which are necessary in the stealth aircraft [21].

In large aerostructures comprising composite parts and metallic parts, the choice of an effective strategy for composite-metal joining is critical for ensuring the structural integrity of an aircraft during service. Messler [22], Jahn [23], and Pramanik [24] have extensively reviewed different strategies for composite-metal joining, which can be broadly classified as mechanical joining and adhesive joining. Although mechanically interlocked joints can sustain significantly higher loads than adhesive joints, they have limitations, including higher stress concentrations around holes or notches and susceptibility to galvanic corrosion from direct contact between two materials with large electrical potential differences [23,25]. Adhesive joints are also susceptible to environmental degradation and require the selection of adhesives that are compatible with dissimilar materials [26]. To avoid these limitations, methods such as “hybrid” joining, including co-curing [27] and weld-bonding [28], have been explored, where a combination of mechanical interlocking and adhesive bonding techniques is used.

Camanho et al. have demonstrated the use of metal/CFRP layered hybrid composites to enhance the efficiency of bolted CFRP joints [29,30]. They showed that hybrid composites with alternate Ti/CFRP layers in the bolt-bearing regions produce joints that are stiffer and stronger than those made with non-hybrid composites. In a similar way, the current work aims to introduce fiber-hybrid textile composites that can improve the efficiency of co-cured composite-metal joints (see Fig.1). With i) a sufficient ductility, ii) the stiffness and strength properties intermediate to the adjacent non-hybrid CFRP and metallic parts, and iii) good electrochemical compatibility with non-hybrid

CFRP, hybrid textile composites with gradual co-weaving of carbon fibers and stainless-steel fibers would provide a smooth property gradient at composite-metal joints, reducing the high stress concentrations and thereby improving the joint efficiency.

Mechanical properties of fiber-hybrid composites with UD stainless steel fiber reinforcement have been studied in the literature [31–37]. However, literature lacks studies on incorporating steel fibers with carbon fibers in form of textile fabrics. Hybrid carbon/steel fiber fabrics would allow for designing gradual property transition from carbon fiber composites to metallic parts. Such hybrid fabrics would be highly valuable for application in composite/metal joints where abrupt changes of properties of the joined components typically results in large stress concentrations. Application of hybrid fabrics would also serve as a more cost-efficient solution compared with UD composite meso-structures. The present work investigates carbon/stainless steel hybrid woven composites addressing aspects of manufacturing, mechanical behavior and micro-damage evolution under mechanical loading. A detailed parametric study is performed to reveal significance of yarn properties and meso-structure on the mechanical properties of hybrid carbon/steel fiber textile composites.

In the current work, Section 2 describes the composite manufacturing and experimental testing. Section 3 presents the multi-scale modeling of hybrid and non-hybrid composite representative volume elements (RVEs). The RVEs consist of resin-impregnated homogeneous yarns, and their behavior under quasi-static tensile loading is modeled using the elastoplastic damage model developed by C  zar et al. [38,39] for UD composites. Additional novelty of the current work is demonstrating the applicability of the material model [38,39] to predict damage evolution and ultimate failure of woven hybrid composites. The results of the experimental and numerical studies are discussed in Section 4, followed by the conclusions in Section 5.

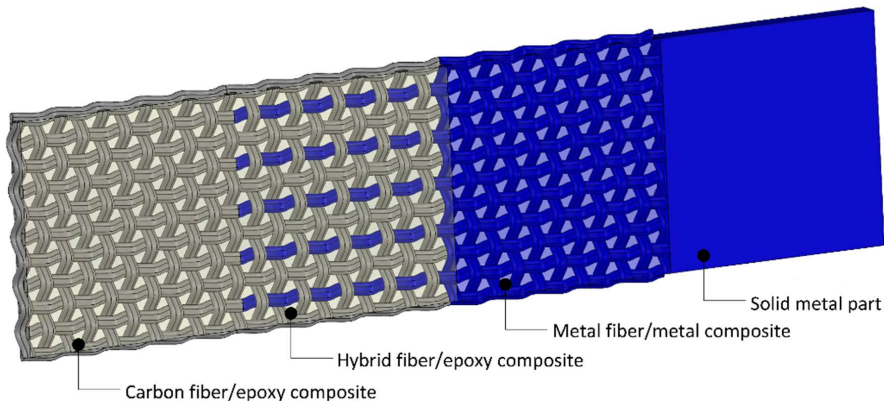


Figure 1. Hybrid carbon/steel fiber textile composite in the assembly of a co-cured composite-metal joint.

2. Experimental work

The hybrid woven composites were manufactured from carbon fiber (CF) yarns and stainless-steel fiber (SF) yarns and epoxy (EP) matrix. T300B-3000-50B carbon fiber yarns were obtained from Toray (France). Yarns have a tow size of 3000 and are made from untwisted surface-treated carbon fibers with an individual filament diameter of 7 μm and a density of 1.76 g/cm^3 [40]. SY 11-1 stainless steel fiber yarns were obtained from Bekaert (Belgium). These yarns consist of discontinuous stainless-steel fibers (manufactured at Fujibo, Japan) twisted into long continuous yarns. The average diameter of a single stainless steel fiber was 12 μm . As the stainless steel fibers were manufactured using a bundle drawing technique, the fibers had an irregular polygonal cross-section [41]. The linear density, i.e., weight (g) for a 1000 m of yarn of stainless-steel yarns was 91 tex. For the matrix material, Araldite® LY 1564 epoxy resin with XB 3404-1 hardener from Huntsman (USA), was used.

2.1. Woven fabric manufacturing

The woven fabrics were manufactured with plain weave architecture on a custom manual weaving loom shown in Fig. 2. In the weaving procedure, warp yarns approximately 2 m in length were first aligned and mounted on a support frame 300 mm in width. The ends of the warp yarns were fixed and during the weaving process the weft yarns were manually woven and compacted transversally to the warp yarns gradually moving the loom along the warp yarn length and producing a plane weave fabric.

Two types of hybrid woven fabrics were manufactured: i) Hybrid fabric denoted as type I - SF weft yarn and CF warp yarn, and ii) Hybrid fabric type II - CF weft yarn and SF warp yarn (see Fig. 3). Reference non-hybrid woven fabrics with only carbon or stainless-steel fiber yarns respectively were also manufactured (refer to Table 1).

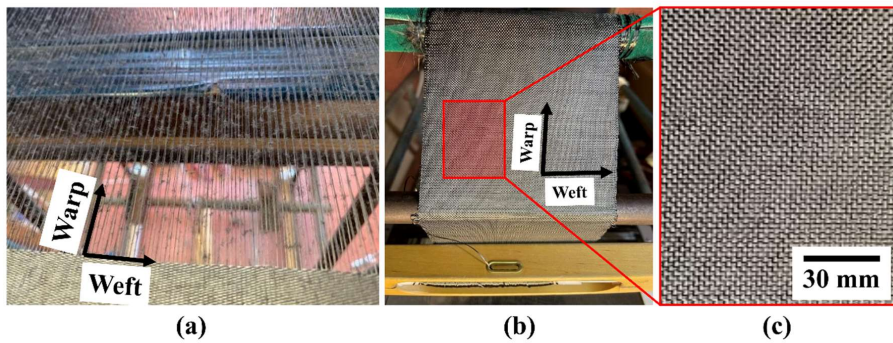


Figure 2. Manufacturing of non-hybrid plain woven fabric with carbon fiber yarns on a manual weaving loom.

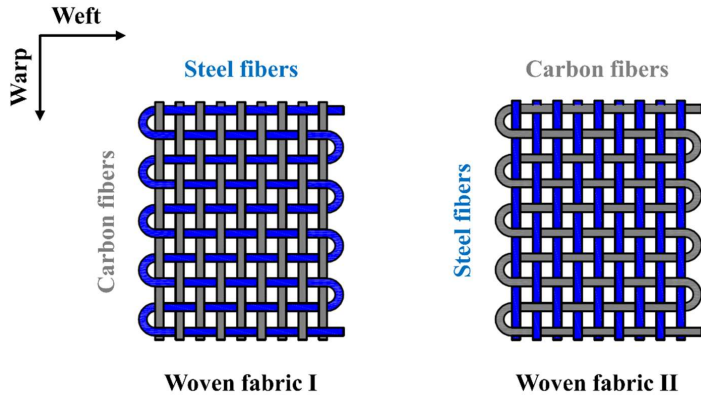


Figure 3. Schematic image of manufactured plain-woven hybrid fabrics.

Table 1. Non-hybrid and hybrid plain woven fabrics

Fabric notation	Fabric type	Warp yarn	Weft yarn
I	Non-hybrid	Carbon fiber (CF)	Carbon fiber (CF)
II	Non-hybrid	Stainless steel fiber (SF)	Stainless steel fiber (SF)
III	Hybrid	Carbon fiber (CF)	Stainless steel fiber (SF)
IV	Hybrid	Stainless steel fiber (SF)	Carbon fiber (CF)

2.2. Composites manufacturing

Four types of woven hybrid and two types of non-hybrid woven composite laminates were manufactured using the method of stacking dry fabric layers and vacuum infusion of resin. Details of the hybrid and non-hybrid composite configurations are given in Table 2. The notations of the composite configurations follow the numbering of multiple batches manufactured within research project which are not all presented in this paper. The full batch of materials included hybrid composites H1 – H9 with carbon fiber yarns and stainless-steel fiber yarns, non-hybrid composites with carbon fiber yarns C1 – C4, and non-hybrid composites with stainless steel fiber yarns M1 – M4. After the vacuum infusion process, the composite plates were cured for 8 hours at 80 °C temperature. The cured plates were cut into rectangular specimens with a length of 200 mm and a width of 20 mm. The specimen thickness varied with the configuration type, and the average thickness values for each configuration are listed in Table 2.

The edges of the rectangular specimens were thoroughly polished so that the meso-structure of the produced textile composites could be experimentally measured by optical microscopy. Optical microscopy measurements of material meso-structure allowed building representative volume element models for numerical analysis. Likewise, the polished specimen edges allowed to observe micro-damage evolution during mechanical testing. Fig. 4 shows the optical microscopy images of the polished specimen edges before mechanical testing. Table 2 presents the values of global fiber volume fractions for the composite configurations, which were measured using the ASTM D2584-18 standard method of burning off the matrix [42]. The average values of local fiber volume fractions in resin-impregnated yarns were measured using image analysis of the optical micrographs of specimens with H5 and H6 composite configurations, and they were 68.6% for CF yarn and 54.3% for SF yarn.

Table 2. Non-hybrid and hybrid composite configurations

Composite notation	Lay-up (fabrics)	Average global fiber volume fraction, V_{fg} (%)	Average specimen thickness, t (mm)
C4	[I] ₄	46.9	1.49
M4	[II] ₃	27.4	1.37
H5	[III] ₄	26.7	1.68
H6	[IV] ₄	27.5	1.54
H8	[IV, III, IV, III, IV]	25.6	2.06
H9	[I, II, I, II, I]	30.5	1.94

2.3. Quasi-static tensile testing

The composite specimens were subjected to quasi-static tensile testing in accordance with the ASTM D3039 standard [43] at a strain rate of 1 %/min. The initial loading–unloading step, with a maximum applied strain of up to 0.4%, was used to measure the elastic moduli of the undamaged composite specimens. Loading-unloading cycles were continued with maximum applied strain level increased by 0.2% in each consecutive step. After each loading-unloading cycle, the polished specimen edges were studied with optical microscopy to observe the micro-damage evolution. Elastic modulus values were measured at each loading step to follow the material stiffness reduction due to micro-damage evolution. Loading-unloading cycles were continued until reaching the ultimate failure of composites. The failure surfaces of material specimens were studied under scanning electron microscope (SEM) to analyze the failure mechanisms in reference non-hybrid and hybrid composite materials.

Tensile testing was performed using a Zwick/Roell Z150 universal testing machine. The strain measurements were conducted using a non-contact-type video extensometer, Messphysik ME 46-NG, and a contact-type extensometer, HBM Strain Link DD1. The micro-damage observations

were carried out using a Motic BA 310Met optical microscope, and the fracture surface observations were carried out using a JEOL JSM IT-300 environmental SEM.

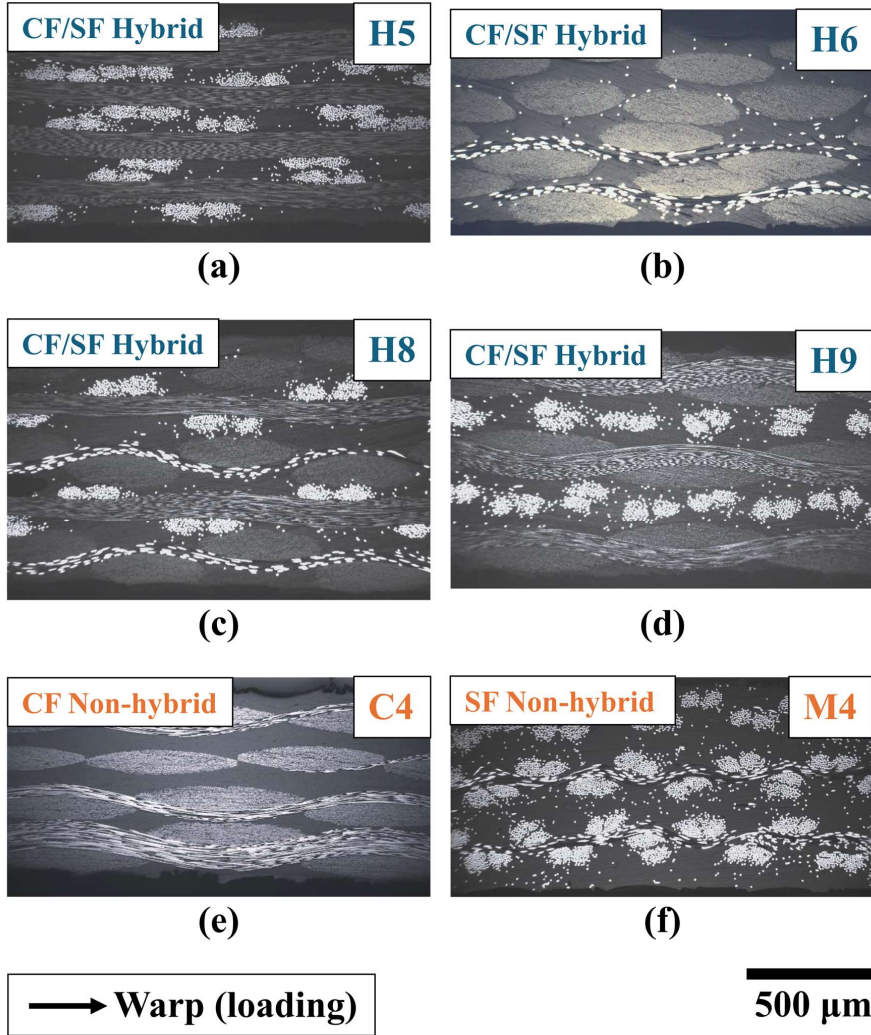


Figure 4. Optical microscopy images of polished specimen edges depicting meso-structures of various hybrid – (a) H5, (b) H6, (c) H8, (d) H9, and non-hybrid– (e) C4, and (f) M4 composite configurations. CF denotes carbon fibers; SF denotes stainless steel fibers.

3. Modeling mechanical behavior

Manufacturing of hybrid fabrics in the present work was highly resource demanding requiring several person-months to produce limited amount of materials. Furthermore, specific structure of the used SY 11-1 stainless-steel yarn materials (twisted yarn of discontinuous fibers) resulted in physical limitations not reaching the full potential of the provided ductility effect in hybrid fiber composites. Hence, for a thorough analysis of initial mechanical behavior, damage evolution, and ultimate failure, numerical models of representative volume elements (RVE) of non-hybrid and hybrid composites were created in Abaqus CAE software [44]. A detailed analysis of parameters affecting the sequence of damage mechanisms leading to ultimate failure was performed. 3D FEM models were generated representing the textile composites by homogenized yarns embedded in epoxy resin block. The RVEs were built according to the actual meso-structure, yarn alignment (amplitude, wavelength) and global and local fiber volume fractions that were determined experimentally. The homogenized yarn properties were determined using micro-mechanics analytical models as described in the following.

3.1. Micromechanics-based analytical modeling of elastic properties

The elastic properties of the homogenized yarns were calculated using the micromechanics-based concentric cylinder assembly (CCA) model developed by Hashin and Rosen [45], Hashin [46] and Christensen and Lo [47] for transversally isotropic materials. The model considers an infinitely long cylindrical fiber embedded in a concentric cylindrical matrix block. The relative radii of the concentric cylinders are calculated based on the fiber volume fraction, V_f , such that,

$$V_f = \frac{r_1^2}{r_2^2} \quad (1)$$

In Eq. (1), r_1 is the fiber cylinder radius and r_2 is the matrix cylinder radius. Table 3 lists the input properties of the transversally isotropic carbon and stainless-steel fibers required for the CCA model to calculate the elastic moduli of the CF/EP and SF/EP homogenized composite yarns, respectively. The cylindrical epoxy block is considered an isotropic material with the elastic modulus of 3 GPa and the Poisson's ratio of 0.29. The calculated elastic properties for the homogenized yarns, as determined by the CCA model, are listed in Table 4.

Table 3. Elastic constants of constituents used as input parameters for the CCA micro-mechanical model.

Input parameter	Symbol	Unit	Carbon fiber [40]	Stainless steel fiber [4]	Epoxy
Longitudinal modulus	E_{11}	GPa	230.00	193.00	3.00
Transverse modulus	E_{22}	GPa	20.00	193.00	3.00
In-plane Poisson's ratio	ν_{12}	-	0.18	0.30	0.29
Out-of-plane Poisson's ratio	ν_{23}	-	0.42	0.30	0.29
In-plane shear modulus	G_{12}	GPa	12.00	74.23	1.16

Table 4. Elastic properties of homogenized composite yarns calculated using the CCA model (CF/EP: Carbon fiber/Epoxy yarn, SF/EP: Steel fiber/Epoxy yarn).

Input parameter	Symbol	Unit	CF/EP	SF/EP
Longitudinal modulus	E_{11}	GPa	158.62	106.21
Transverse modulus	E_{22}	GPa	8.81	8.79
In-plane Poisson's ratio	ν_{12}	-	0.21	0.30
Out-of-plane Poisson's ratio	ν_{23}	-	0.40	0.38
In-plane shear modulus	G_{12}	GPa	4.18	3.75

3.2. Representative volume elements

For construction of RVEs of different woven composite configurations, the homogenized composite yarns were modelled with elliptical cross-sections. The dimensions of the respective ellipses for the carbon and stainless-steel fiber yarns were determined from optical microscopy images. Geometry of RVEs was generated in SolidWorks CAD software [48]. For each composite configuration, the warp yarn curvature and the relative placement of the warp and weft yarns in the composite RVEs were determined from optical microscopy images. For simplicity, the weft yarns (aligned transversally with respect to loading) were assumed to be aligned in straight lines. Parametric analysis showed that the weft yarn waviness has a small influence on elastic properties and stress-strain curve. The warp yarns (aligned in the loading direction) followed a sinewave shape according to the wavelength and amplitude experimentally determined from optical microscopy. The assembly of the warp and weft yarns was embedded in a cuboidal matrix block. The RVE length was equal to the single weft yarn spacing, width was equal to warp yarn spacing, and the height was equal to the average thickness of a layer. For brevity and simplicity, numerical models of RVEs were built for selected composite configurations: the non-hybrid composites M4, C4; and hybrid-fiber composites H5 and H6. Table 5 lists the respective RVE dimensions for the hybrid and

non-hybrid composite configurations. Fig. 5 shows the assembly of the warp and weft yarns in the RVEs of the hybrid and non-hybrid composite configurations selected for FEM numerical modeling.

Unlike for M4, C4 and H5 and H6 composites for which the thickness of RVEs is equal to thickness of one ply, the RVEs for hybrid composite configurations H8 and H9 would consist of multiple layers according to their respective lay-ups (see Table 2). Specifically, the RVE of H8 would comprise a layer of H5 and a layer of H6, and the RVE of H9 would comprise a layer of C4 and a layer of M4. Thus, while detailed FEM simulations of mechanical properties were performed for C4, M4, H5, H6, mechanical properties and behavior of H8 and H9 could be derived indirectly using simple rule-of-mixtures considerations.

Table 5. Dimensions of representative volume elements (RVEs) of the hybrid and non-hybrid composite configurations for the finite element analysis.

RVE	Epoxy matrix block			Elliptical yarn dimensions (major diameter $\times$ minor diameter)		Warp yarn sinewave amplitude, A (mm)
	Length, l (mm)	Width, w (mm)	Height, h (mm)	Weft yarn, $a \times b$ (mm $\times$ mm)	Warp yarn, $a \times b$ (mm $\times$ mm)	
H5	1.04	1.46	0.44	0.20×0.08	0.86×0.25	0.03
H6	0.90	1.45	0.54	0.86×0.25	0.20×0.08	0.11
C4	1.44	1.21	0.52	1.42×0.14	1.18×0.18	0.10
M4	0.40	0.50	0.70	0.19×0.08	0.20×0.08	0.08

3.3. Constitutive models for yarns and matrix

Numerical simulation of the behavior of the textile composite RVEs under uniaxial quasi-static tensile loading was performed in Abaqus CAE software [44]. The yarns (impregnated fiber bundles) were modeled as homogenized transversally isotropic materials with the elastic properties previously calculated using micro-mechanical CCA models (see Table 4). The elastoplastic damage behavior was defined according to the constitutive model developed by C  zar et al. [38,39]. This model combines a plasticity model based on a non-associative flow rule [49], with the 3D damage model developed by Quintanas-Corominas et al. in [50]. While this constitutive model was originally developed for UD fiber-reinforced composite materials [38,39], the present work explores the capability of this model to predict the mechanical and damage evolution behavior of woven textile composites.

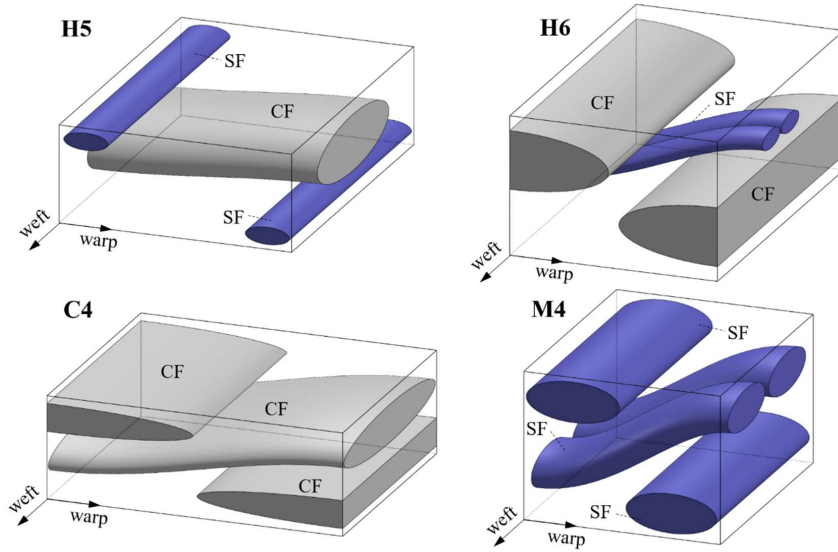


Figure 5. RVEs of the hybrid composites H5, H6, and the non-hybrid composites C4, M4. CF – carbon fiber yarn, SF – stainless steel fiber yarn.

According to the model by C  zar et al. [38,39], along the fiber direction, the composite yarns behave as purely elastic until the onset of damage. Response in transverse direction includes elastoplastic part, whereas in longitudinal direction response is purely linear-elastic until reaching damage. The plastic response is defined using a non-associative flow rule, where the direction of plastic flow is determined by a plastic potential function. Beyond the onset of damage, the behavior along the principal directions is defined using softening laws. The main material strength parameters used in the elastoplastic damage model [38,39] for the homogenized CF/EP and SF/EP yarns are listed in Table 6. For a full list of input material parameters for the elastoplastic damage model for the yarns, refer to Appendix A1 Table A1.1.

Table 6. Main material strength parameters used in the elastoplastic damage model [38,39] for the homogenized CF/EP and SF/EP yarns.

Symbol	Description	Unit	CF/EP yarn	SF/EP yarn
X_T	Longitudinal tensile strength	MPa	2080.00 [40]	260.00 [4]
Y_T	Transverse tensile strength	MPa	76.00	40.00
S_L	Longitudinal shear strength	MPa	98.00	50.00

The epoxy matrix was modelled as an isotropic material with elastic properties given in Table 3. The damage behavior of composites was defined according to the concrete damaged plasticity

(CDP) model [51] available in the Abaqus CAE software [44]. The model is based on Lubliner et al. [52] and Lee and Fenves [53], and it is capable of simulating the mechanical behavior of epoxy resins [54–56].

According to the CDP model [51], epoxy resin exhibits purely elastic behavior under uniaxial tensile loading and elastoplastic behavior under uniaxial compressive loading until the onset of damage. In the present work, although the RVEs are loaded in pure tension, local compressive stresses are expected in the yarns and matrix due to the waviness of the warp yarn. The CDP model for the epoxy matrix uses softening laws to define the evolution of damage under tensile and compressive loading. For a full list of CDP model parameters [51,54,55,57,58], refer to Appendix A1 Table A1.2.

3.4. Calculation of macro-mechanical and damage evolution behavior

FEM models according to RVEs for non-hybrid C4, M4 and hybrid H5, H6 composites (see Fig. 5) were generated in Abaqus CAE software [44]. The RVE models were meshed with the 8-node linear brick C3D8R solid elements with reduced integration and default hourglass control. An element size of 0.02 mm was pre-defined as a guideline for automatically generated mesh. Local material coordinate systems were generated in each element in the warp and the weft yarns according to the yarn orientation with respect to the global coordinate system. Fig. 6 schematically shows the boundary conditions applied on an RVE model. The node in the origin of the global coordinate system ($x, y, z = 0$) was constrained against all displacements, symmetry boundary conditions were applied on the faces of the matrix block passing through the origin. Displacements in z axis direction were coupled for all nodes on the face $z = -w$ and displacements in y axis direction were coupled for all nodes on the face $y = h$. Uniform uniaxial displacement in the x -axis direction was applied at the matrix block face at $x = l$.

Initially, the FEM simulations were performed on multiple CPUs using a parallel computing method. However, it was observed that when the domain-level parallelization method was chosen for parallel computing, the solution of the non-linear analysis depended significantly on the number of CPUs used. Thus, the method of loop-level parallelization was used in calculations to obtain CPU core-independent results. The technique of variable mass scaling was employed to obtain a convergent solution, and the target time increment for the mass scaling routine was determined by checking the energy balance.

During the post-processing of the results, stress-strain curves were plotted to study the composite macro-mechanical behavior, and they were compared with the experimental stress-strain curves of the respective composite configurations. The initiation and propagation of micro-damage was studied based on the evolution of the damage variables – i) d_l corresponding to the longitudinal failure in tension and compression, ii) d_t corresponding to the mode-I matrix cracking, iii) d_{st} corresponding to the mode-II matrix cracking, and iv) d_{sl} corresponding to the longitudinal tensile

and matrix damage mechanisms associated with the longitudinal shear stress invariant. A parametric analysis was also performed to understand the effect of the warp yarn material properties, such as longitudinal strength (X_T), transverse strength (Y_T), and longitudinal shear strength (S_L) on the macro-mechanical and damage evolution behavior of a hybrid composite material.

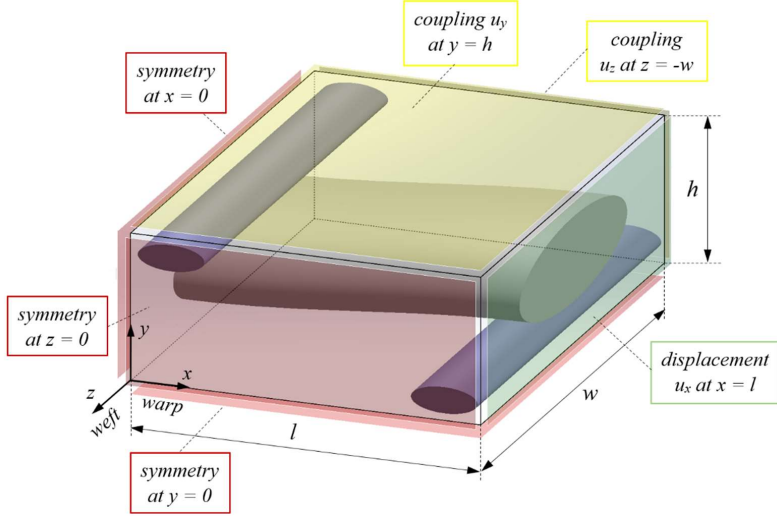


Figure 6. Schematic illustration of boundary conditions applied on an RVE.

4. Results

4.1. Elastic modulus

The experimentally determined values of elastic modulus, ultimate tensile strength and strain-to-failure for the tested composite configurations are presented in Table 7.

Table 7. Experimental values of elastic modulus, ultimate tensile strength and strain-to-failure for the tested composite configurations

Composite notation	Elastic modulus, E_{x0} (GPa)	Ultimate tensile strength, σ_{max} (MPa)	Strain-to-failure, ϵ_{max} (%)
H5	41.53	405.61	0.95
H6	10.84	72.06	0.74
H8	24.08	205.84	0.90
H9	28.60	235.77	0.89
C4	43.77	380.10	0.85
M4	7.09	40.92	0.60

Fig. 7 gives the comparison of experimentally determined elastic modulus values and the numerical results. As previously mentioned, FEM simulations were performed for C4, M4, H5 and H6 configurations, whereas the mechanical properties of H8 and H9 were derived indirectly using rule-of-mixtures considerations. The elastic modulus values were mainly governed by the warp yarn materials. Refer to Fig. 7, the experimentally determined elastic modulus value of H5 composite with carbon fiber warp yarns is close to the elastic modulus value of non-hybrid composite with carbon fiber yarns C4. Similarly, the elastic modulus value of H6 composite with stainless steel fiber warp yarns is close to the elastic modulus value of non-hybrid composite with stainless steel fiber yarns M4. A good agreement can be seen between the experimentally determined elastic modulus values and the numerical results for all composite configurations.

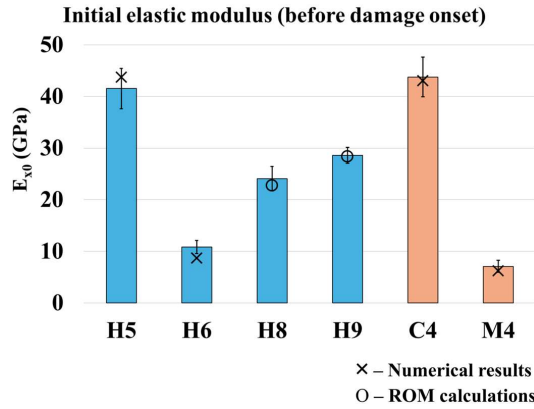


Figure 7. Comparison of elastic modulus values determined using the experimental and numerical analyses.

4.2. Damage initiation, evolution and ultimate failure

Micro-damage appearance was observed in the composite specimens at applied strain levels of 0.6–0.8%. Among all the composite configurations studied, an extensive damage evolution was experimentally observed only in carbon fiber non-hybrid base material composite C4 (see Fig. 8a–c). In the case of hybrid composite specimens, the damage initiation was quickly followed by the ultimate failure. The reduction in longitudinal stiffness values, as a consequence of micro-damage evolution until the ultimate failure, is shown in Fig. 9. The ratio of the load taken by the weft yarns versus the load taken by the warp yarns under loading along the warp yarn direction is higher in steel fibers than carbon fibers due to a higher value of E_{22}/E_{11} (refer to Table 3). As the micro-damage evolution in the woven composites loaded along the warp yarn direction mainly comprises matrix cracking around the weft yarns, the reduction in stiffness due to micro-damage evolution is higher in the composites with steel fiber weft yarns than in the ones with carbon fiber weft yarns. This phenomenon is evident in Fig. 9, where composite configurations with steel fiber weft yarns,

H5, H8, and H9, exhibit a higher reduction in longitudinal stiffness compared to the composite configurations with carbon fiber weft yarns, C4 and H6, before the ultimate failure.

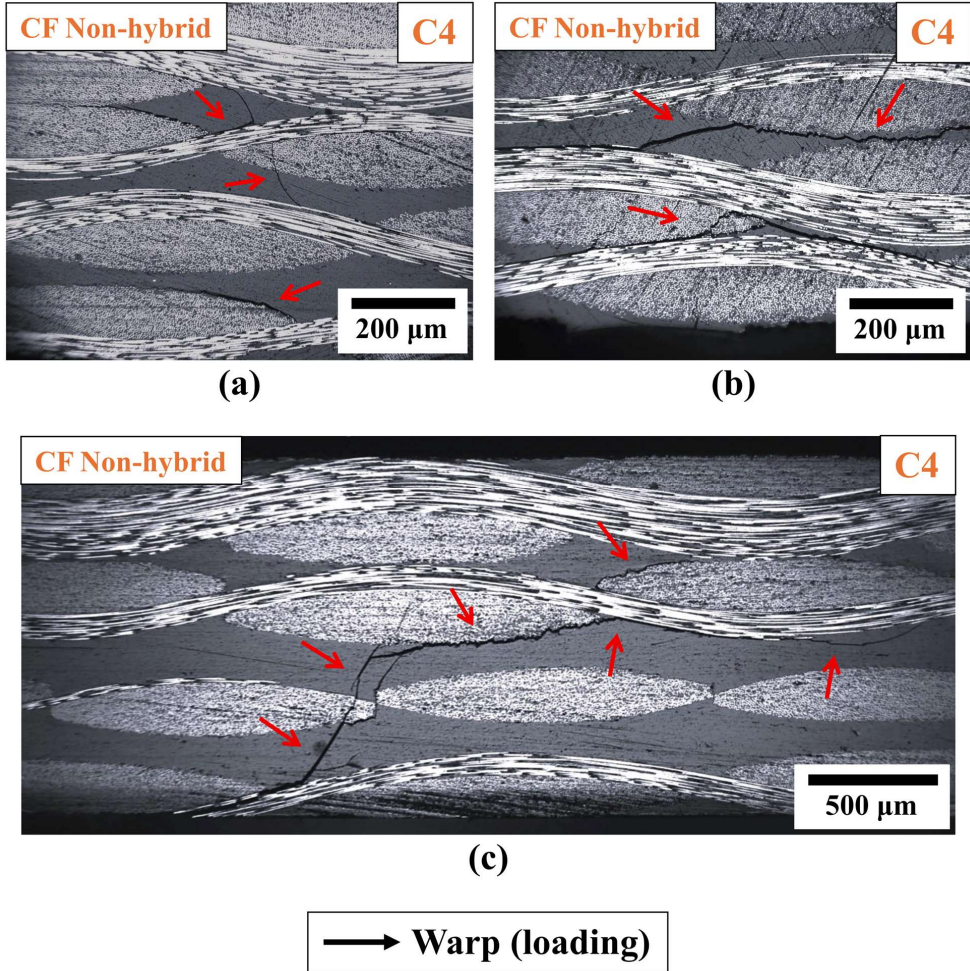


Figure 8. (a – c) Micro-damage evolution in woven composite specimens under uniaxial quasi-static tensile loading in the form of interconnection of yarn/matrix-interface delaminations and transverse cracks.

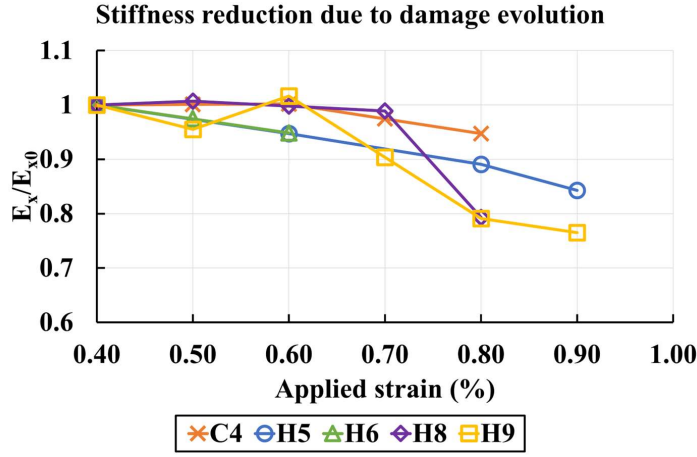


Figure 9. Reduction of longitudinal modulus in the composite configurations due to damage evolution in the form of weft yarn matrix cracking.

The failure of warp yarns resulted in the ultimate failure of the woven composite specimens. Thus, the values of tensile strength and strain-to-failure for the various composite configurations are governed by on the respective characteristics of the warp yarns. The hybrid configuration H5 and the non-hybrid configuration C4, both with carbon fiber warp yarns, exhibited similar tensile strength values (see Fig. 10a). Correspondingly, the same trend was observed between H6 and M4 composites both with stainless-steel fiber warp yarns. The tensile strength values shown by the composite specimens with complex hybrid configurations H8 and H9 followed well the rule of mixtures calculation, i.e., they exhibited a neutral hybrid effect in terms of the tensile strength, as explained by Swolfs et al. in [7]. The experimental values of failure strain for various composite configurations, however, showed a scattered trend which is therefore not shown in Fig. 10b. The experimental values showed a positive hybrid effect in H8 and H9 in terms of the strain-to-failure. Numerical simulations showed that non-hybrid composites with steel fiber yarns had the highest failure strain among all composite configurations. This value was significantly reduced when steel fibers were replaced with carbon fibers in either the warp or weft yarns. Compared with non-hybrid composites with carbon fiber yarns, the carbon/steel fiber hybrid composites exhibited up to 20% higher FEM strain-to-failure values. This highlighted the major advantage of hybridizing woven carbon fiber reinforced composites with ductile steel fibers. However, the anticipated ductile behavior was not observed in the non-hybrid composite specimens with ductile steel fibers (as evidenced by the low failure strain values observed during the experimental tests by M4 in Fig. 10b), which could have resulted from a) premature necking of steel fibers that prevents progression of matrix damage (also seen in [35]) and b) a strong adhesive force at the steel fiber/epoxy interface due to steel fiber surface treatment and high yarn twist [59–61].

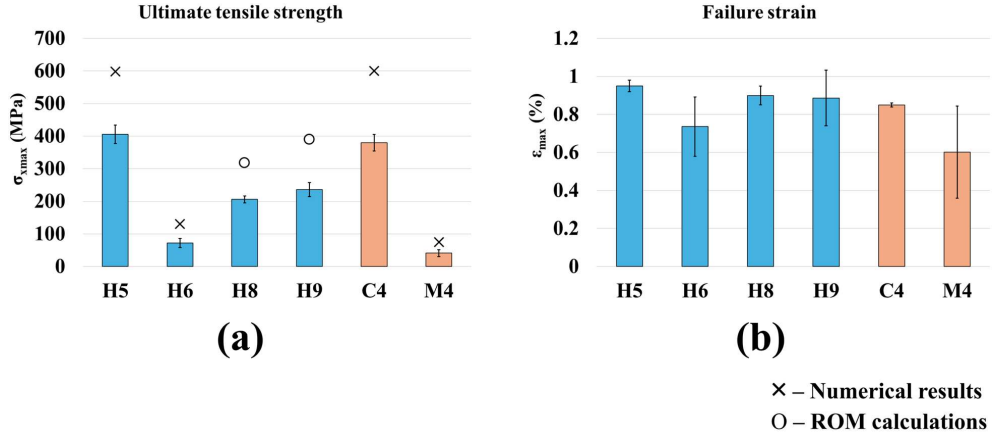


Figure 10. (a) Ultimate tensile strength and (b) failure strain values in the longitudinal direction for different composite configurations.

The fracture behavior and strength of resin-impregnated yarns are influenced by various parameters. Compared to the carbon fiber/epoxy interface, the steel fiber/epoxy interface exhibits higher stress concentrations due to a greater mismatch in the transverse stiffness of the fiber and matrix [62]. However, in terms of fiber-matrix debonding behavior, the effect of fiber/matrix interface adhesion strength is greater than that of stress concentrations. With low interface adhesion strength and high stress concentrations, a larger debond occurs at an untreated steel fiber/epoxy interface than at a carbon fiber/epoxy interface at the same applied stress level. For a treated steel fiber/epoxy interface, however, the debond length is smaller than at a carbon fiber/epoxy interface [63]. Yarn twist also has a considerable effect on the fracture behavior and strength of a yarn. It increases inter-fiber cohesion and fiber obliquity, which have contradictory effects on yarn strength [64]. Thus, a compromise has to be made to optimize the yarn twist angle [60,65]. Beyond the optimum yarn twist angle (15° for carbon fiber/epoxy yarn [60]), the yarn strength reduces significantly.

The fracture surfaces of various woven composite configurations were observed under SEM, as shown in Figs. 11-12. For composite specimens with carbon fiber warp yarns (C4 in Fig. 11a, H5 in Fig. 11b), a significant amount of adhesive failure can be observed at the fiber/matrix interface. Large debonds are seen that may have formed at the locations of fiber-break clusters. In contrast, for specimens with steel fiber warp yarns (H6 in Fig. 11c, M4 in Fig. 11d), cohesive failure occurs within the fibers, accompanied by small debonds at the fiber/matrix interface due to localized necking (refer to Fig. 11e-f). A few instances of fiber pull-out can also be observed in Fig. 11e. The ultra-high-magnification SEM image of a single broken steel fiber in M4, shown in Fig. 11f, depicts

the steel fiber surface texture with parallel bands, necking at the fracture location, and the formation of internal voids in the fiber before final failure.

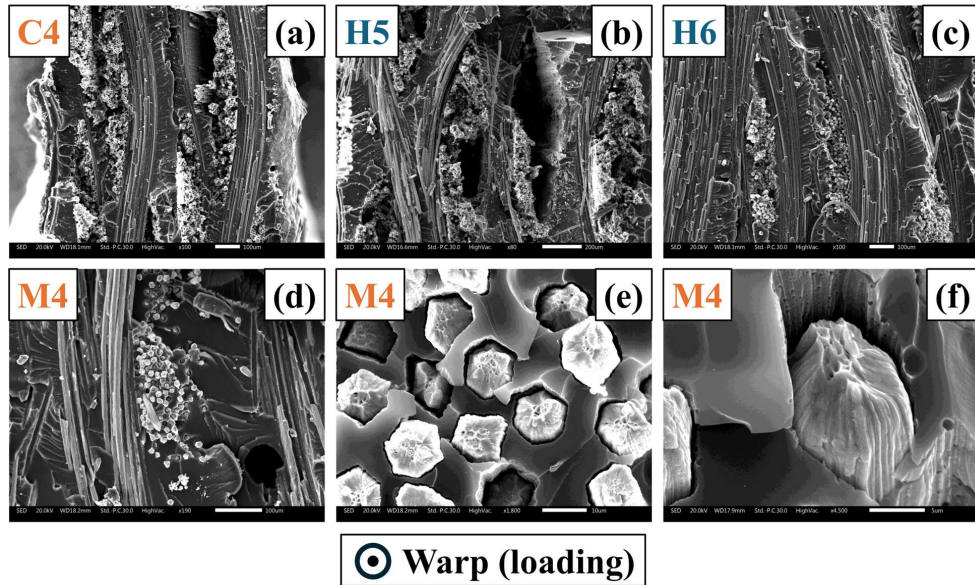


Figure 11. SEM images of fracture surfaces of woven composite specimens with configurations (a) C4, (b) H5, (c) H6, and (d-f) M4.

The fracture surface SEM image of the composite configuration H8, shown in Fig. 12, highlights the differences in failure between steel and carbon fiber yarns. With large fiber/matrix interface debonds in the warp yarn (see the green ellipse in Fig. 12) and low matrix residue on the fibers in the weft yarn (green rectangle in Fig. 12), the carbon fiber/epoxy interface in H8 is weaker than the steel fiber/epoxy interface, which has small interface debonds in the warp yarn (red ellipse in Fig. 12) and high matrix residue on the fibers in the weft yarn (red rectangle in Fig. 12).

4.3. FEM Numerical analysis

The FEM analysis of the elastoplastic damage behavior of the hybrid and non-hybrid textile composites was carried out using Abaqus CAE software [44]. For the analysis, the respective RVEs were subjected to quasi-static uniaxial tensile displacement in the warp yarn direction until reaching the ultimate failure. Fig. 13 compares the stress-strain behavior calculated using FEM with the experimental data. It can be seen that a good agreement was achieved between the two methods for the elastic behavior of all composite configurations. During experiments, the specimens with all composite configurations except M4 failed in a brittle manner, showing no plasticity before ultimate failure. The numerical solutions, on the other hand, allow exploring the

quasi-ductile behavior in all composite configurations using the selected combinations of material strength parameters.

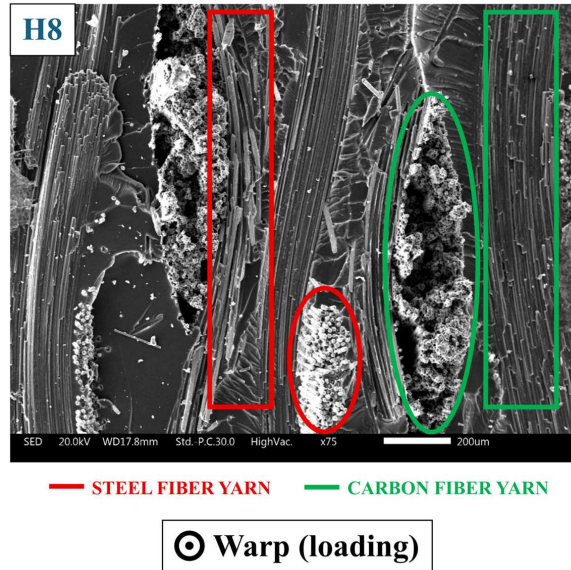


Figure 12. Fracture surface SEM image of woven composite configuration H8 showing the differences in failure between steel and carbon fiber yarns.

Considering the FEM stress-strain behavior of the hybrid composite configuration H5 with default failure properties of the yarns and matrix (refer to Appendix A1 Table A1.1) under quasi-static tensile loading (Fig. 13a), a linear elastic behavior is observed until micro-damage initiates in the form of weft yarn/matrix-interface debonds at 1.2% applied strain. At 1.37% applied strain, a knee point can be seen in the stress-strain curve, which is characterized by the initiation of fiber damage in the warp yarn. As explained in Fig. 14, in the plateau region following the knee point, the fiber damage (denoted by the damage parameter d_L) spreads across the warp yarn such that regions with high fiber break density are created, see damage parameter d_L at stage B in Fig. 14. At an applied strain of 2.2%, the maximum stress in the RVE is reached. The ultimate failure in the composite is caused by warp yarn failure at one of the regions with high fiber break density. At this applied strain level, all the other damage parameters, namely, longitudinal shear damage in the fibers and matrix (d_{SL}), Mode-I matrix cracking (d_T), and Mode-II matrix cracking (d_{ST}) also achieve the highest value ($= 1$) at the location of ultimate failure. The failed warp yarn causes matrix damage in the adjoining region, and the composite RVE ultimately fails forming an inclined fracture surface as shown in Fig. 16.

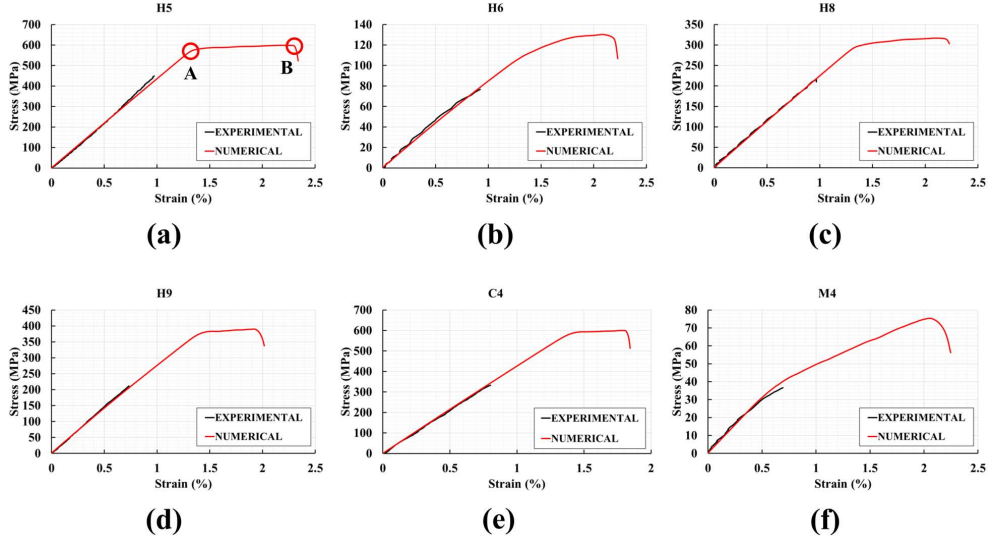


Figure 13. Comparison of full stress-strain behavior of composite configurations, (a) H5, (b) H6, (c) C4, and (d) M4.

Apart from M4, micro-damage initiation occurs at the weft yarn/matrix interface in all composite configurations. M4 shows damage initiation along the warp yarn/matrix interface. The sequence of damage evolution is similar across all configurations, with yarn/matrix interface delaminations and transverse cracks in weft yarns leading to regions of high stress concentration in the adjoining warp yarns, resulting in multiple fiber breakages in the warp yarns that ultimately lead to the composite's ultimate failure. Although a detailed sequence of damage evolution was not observed during the experimental analysis, Fig. 15 shows an occurrence of a transverse crack in carbon fiber weft yarn observed during the quasi-static tensile loading of a composite specimen with the configuration H8, which is similar to the transverse cracks observed in H6 RVE during the FEM analysis. Fig. 16 shows the nature of fully damaged RVEs using the contour plots of tensile damage parameter in the epoxy matrix. A characteristic shear cracking of the matrix block, which is consistent with the concrete-damaged plasticity constitutive model, is clearly visible in all composite configurations.

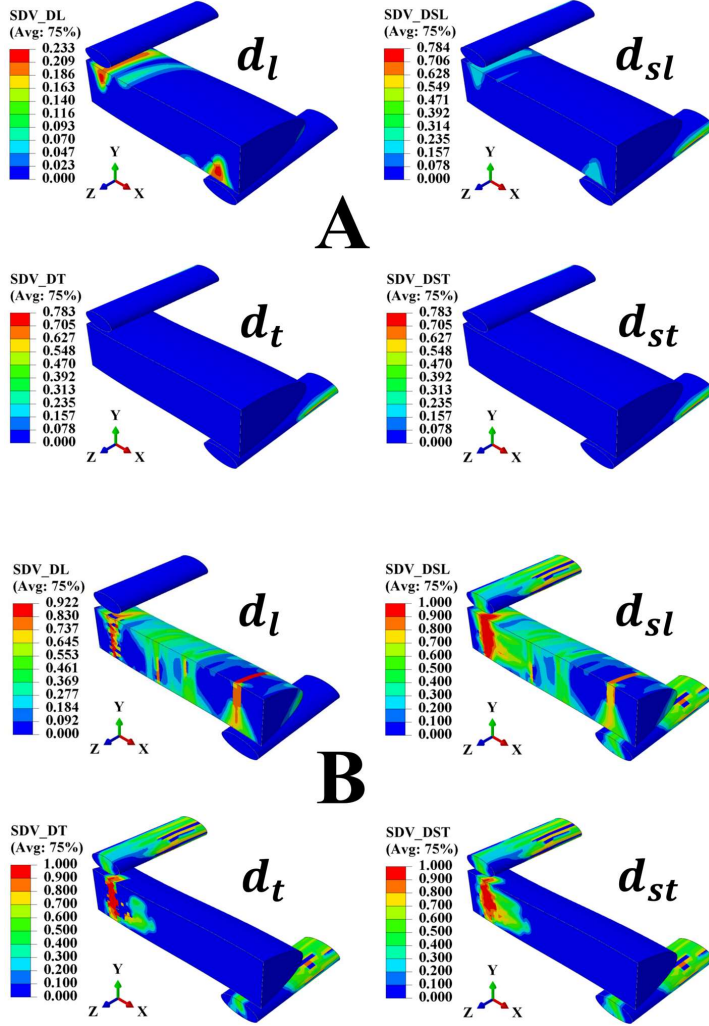


Figure 14. A comparison of stages A and B (with respect to stress-strain curve shown in Fig. 13a) in the evolution of the damage parameters in hybrid composite configuration H5.

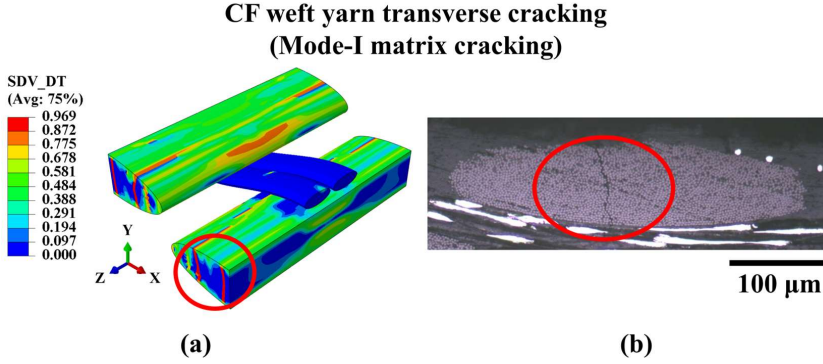


Figure 15. The occurrence of transverse cracking inside the carbon fiber weft yarns in (a) H6 and (b) H8 composite configurations observed during the FEM and experimental analyses.

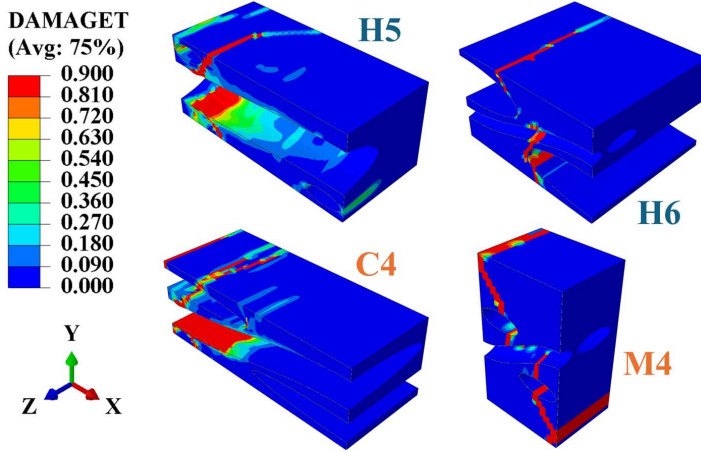


Figure 16. Epoxy matrix tensile damage parameter contour plots showing fully damaged composite RVEs.

Referring to Fig. 7 and Fig. 10a, it can be concluded that the stiffness and strength of a woven composite under quasi-static tensile loading along the warp yarn direction are mainly determined by the warp yarn characteristics. Thus, during the FEM numerical analysis, a parametric analysis was performed on the composite configuration H5 by varying warp yarn material parameters and studying their effects on the macro-mechanical and damage evolution behavior of the woven hybrid composite. Fig. 17a shows the effects of warp yarn longitudinal tensile strength (X_T) on the composite macro-mechanical behavior. It can be seen that the variation in X_T mainly affects the

knee point on the stress-strain curve, which is caused by the activation of longitudinal damage parameter d_L in the warp yarn. Thus, a higher value of warp yarn X_T delays the activation of d_L i.e., fiber breakage in the warp yarn. The effect of Y_T i.e., warp yarn transverse tensile strength, on the H5 stress-strain behavior is shown in Fig. 18b. The warp yarn transverse strength mainly affects the activation of transverse damage mode d_T in the warp yarn. The damage evolution study shows that a lower warp yarn Y_T results in the formation of delamination on the warp yarn/matrix interface. In the case of $Y_T = 50 \text{ MPa}$, the presence of delamination still allows the composite to reach warp yarn longitudinal tensile strength, causing the knee point in the stress-strain curve i.e., activation of warp yarn fiber breakage. At $Y_T = 25 \text{ MPa}$, the early activation of warp yarn/matrix-interface delamination at 0.6% applied strain does not allow the composite to reach the warp yarn longitudinal tensile strength before the ultimate failure. Finally, the warp yarn longitudinal shear strength, S_L , mainly influences the micro-damage evolution after the warp yarn longitudinal tensile strength is reached, refer to Fig. 17c. A lower warp yarn S_L value results in the activation of longitudinal shear damage D_{SL} in a wider region of the warp yarn, causing a change in the matrix damage distribution along the warp yarn/matrix-interface. This causes a slight change in the post-warp-yarn-failure behavior of the composite RVE.

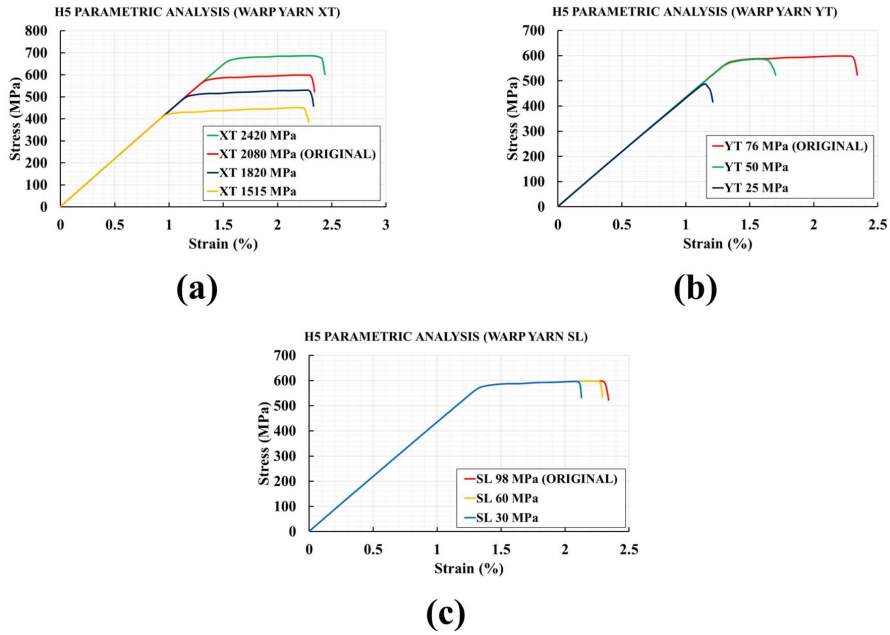


Figure 17. The effects of warp yarn (a) longitudinal tensile strength (X_T), (b) transverse tensile strength (Y_T), and (c) longitudinal shear strength (S_L) on the macro-mechanical behavior of the composite configuration H5.

Conclusions

The present paper studied micro- and macro-scale mechanical behavior of novel hybrid textile composites consisting of carbon and stainless-steel fibers with high potential for ductility improvements in composites. Various configurations of hybrid and reference composites with carbon and stainless-steel warp and weft yarns were manufactured and experimentally tested under quasi-static mechanical loading. The occurred damage mechanisms in hybrid textile composites were thoroughly inspected using optical and scanning electron microscopy. Modeling methodology was proposed for prediction of elastic properties of hybrid textile composites using micro-mechanical models for estimation of homogenized yarn composite properties and building RVE finite element models for textile composites. Using the actual warp and weft yarn geometrical parameters experimentally measured from optical microscopy images, in combination with micro-mechanical models, an excellent agreement with experimental values was achieved for elastic modulus for all studied hybrid composites. This highlights suitability of the proposed methodology as a cost-efficient alternative to recently emerged automated RVE generation techniques using micro computed tomography (μ -CT).

Detailed analysis of damage mechanisms in hybrid composites that lead to the ultimate tensile failure was performed extending the RVE finite element models calibrated at the elastic behavior stage. Damage behavior was analyzed implementing the elastoplastic damage constitutive model previously developed by C  zar [38,39] for unidirectional composites. C  zar's model [38,39] was applied to model damage evolution in composite yarns, whereas for matrix, concrete damaged plasticity model was used. A parametric study was performed revealing significance of composite yarn strength parameters. In particular, it was found that although yarn/matrix delaminations and transverse cracking of yarns are typically the first mode of damage, the stress-strain behavior of hybrid textile composites is mostly affected by formation of multiple warp yarn breaks which cause a plateau in the stress-strain curve and lead to the ultimate failure. Parametric analysis also demonstrated that particularly low transverse strength values for the yarns and the associated localization of damage may lead to premature failure of hybrid textile composites without utilizing the ductility potential of the warp yarns.

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References

- [1] Czél G, Rev T, Jalalvand M, Fotouhi M, Longana ML, Nixon-Pearson OJ, et al. Pseudo-ductility and reduced notch sensitivity in multi-directional all-carbon/epoxy thin-ply hybrid composites. *Compos Part A Appl Sci Manuf* 2018;104. <https://doi.org/10.1016/j.compositesa.2017.10.028>.
- [2] Fuller JD, Wisnom MR. Pseudo-ductility and damage suppression in thin ply CFRP angle-ply laminates. *Compos Part A Appl Sci Manuf* 2015;69. <https://doi.org/10.1016/j.compositesa.2014.11.004>.
- [3] Fuller JD, Wisnom MR. Exploration of the potential for pseudo-ductility in thin ply CFRP angle-ply laminates via an analytical method. *Compos Sci Technol* 2015;112. <https://doi.org/10.1016/j.compscitech.2015.02.019>.
- [4] Callens MG, Gorbatiikh L, Verpoest I. Ductile steel fibre composites with brittle and ductile matrices. *Compos Part A Appl Sci Manuf* 2014;61:235–44. <https://doi.org/10.1016/J.COMPOSITESA.2014.02.006>.
- [5] Allaer K, De Baere I, Lava P, Van Paepegem W, Degrieck J. On the in-plane mechanical properties of stainless steel fibre reinforced ductile composites. *Compos Sci Technol* 2014;100. <https://doi.org/10.1016/j.compscitech.2014.05.009>.
- [6] Wisnom MR, Pimenta S, Shaffer MSP, Robinson P, Potter KD, Hamerton I, et al. High performance ductile and pseudo-ductile polymer matrix composites: A review. *Compos Part A Appl Sci Manuf* 2024;181. <https://doi.org/10.1016/j.compositesa.2024.108029>.
- [7] Swolfs Y, Gorbatiikh L, Verpoest I. Fibre hybridisation in polymer composites: A review. *Compos Part A Appl Sci Manuf* 2014;67. <https://doi.org/10.1016/j.compositesa.2014.08.027>.
- [8] Hayashi T. Development of new material properties by hybrid composition. *Fukugo Zairyo (Composite Materials)* 1972:18–20.
- [9] Hayashi T, Koyama K., Yamazaki A., Kihira M. Development of new material properties by hybrid composition. 2nd report. *Fukugo Zairyo (Composite Materials)* 1972:21–5.
- [10] Bunsell AR, Harris B. Hybrid carbon and glass fibre composites. *Composites* 1974;5. [https://doi.org/10.1016/0010-4361\(74\)90107-4](https://doi.org/10.1016/0010-4361(74)90107-4).
- [11] Zweben C. Tensile strength of hybrid composites. *J Mater Sci* 1977;12. <https://doi.org/10.1007/BF00540846>.

- [12] Manders PW, Bader MG. The strength of hybrid glass/carbon fibre composites - Part 1 Failure strain enhancement and failure mode. *J Mater Sci* 1981;16.
<https://doi.org/10.1007/BF00542386>.
- [13] Marom G, Fischer S, Tuler FR, Wagner HD. Hybrid effects in composites: conditions for positive or negative effects versus rule-of-mixtures behaviour. *J Mater Sci* 1978;13:1419–26. <https://doi.org/10.1007/BF00553194/METRICS>.
- [14] Summerscales J, Short D. Carbon fibre and glass fibre hybrid reinforced plastics. *Composites* 1978;9:157–66. [https://doi.org/10.1016/0010-4361\(78\)90341-5](https://doi.org/10.1016/0010-4361(78)90341-5).
- [15] Jalalvand M, Czél G, Wisnom MR. Damage analysis of pseudo-ductile thin-ply UD hybrid composites – A new analytical method. *Compos Part A Appl Sci Manuf* 2015;69:83–93.
<https://doi.org/10.1016/J.COMPOSITESA.2014.11.006>.
- [16] Swolfs Y, Verpoest I, Gorbatiikh L. Recent advances in fibre-hybrid composites: materials selection, opportunities and applications. *International Materials Reviews* 2019;64.
<https://doi.org/10.1080/09506608.2018.1467365>.
- [17] Swolfs Y. Perspective for Fibre-Hybrid Composites in Wind Energy Applications. *Materials* 2017, Vol 10, Page 1281 2017;10:1281. <https://doi.org/10.3390/MA10111281>.
- [18] Le Guen MJ, Newman RH, Fernyhough A, Emms GW, Staiger MP. The damping–modulus relationship in flax–carbon fibre hybrid composites. *Compos B Eng* 2016;89:27–33. <https://doi.org/10.1016/J.COMPOSITESB.2015.10.046>.
- [19] Kennedy GJ, Martinsy JRRA. A comparison of metallic and composite aircraft wings using aerostructural design optimization. 12th AIAA Aviation Technology, Integration and Operations (ATIO) Conference and 14th AIAA/ISSMO Multidisciplinary Analysis and Optimization Conference 2012. <https://doi.org/10.2514/6.2012-5475>.
- [20] Irving P, Soutis C. Polymer composites in the aerospace industry. 2019.
<https://doi.org/10.1016/C2017-0-03502-4>.
- [21] Li Z, Haigh A, Soutis C, Gibson A. X-band microwave characterisation and analysis of carbon fibre-reinforced polymer composites. *Compos Struct* 2019;208:224–32.
<https://doi.org/10.1016/J.COMPSTRUCT.2018.09.099>.
- [22] Messler RW. Joining Dissimilar Material Combinations. *Joining of Materials and Structures* 2004:697–741. <https://doi.org/10.1016/B978-075067757-8/50015-4>.
- [23] Jahn J, Weber M, Bochner J, Steinhilper R. Assessment Strategies for Composite-metal Joining Technologies-A Review. *Procedia CIRP*, vol. 50, 2016.
<https://doi.org/10.1016/j.procir.2016.05.034>.

- [24] Pramanik A, Basak AK, Dong Y, Sarker PK, Uddin MS, Littlefair G, et al. Joining of carbon fibre reinforced polymer (CFRP) composites and aluminium alloys – A review. *Compos Part A Appl Sci Manuf* 2017;101. <https://doi.org/10.1016/j.compositesa.2017.06.007>.
- [25] Joesbury A. *New Approaches to Composite Metal Joining*. Cranfield University, 2015.
- [26] Messler RW. Joining composite materials and structures: Some thought-provoking possibilities. *Journal of Thermoplastic Composite Materials* 2004;17:51–75. <https://doi.org/10.1177/0892705704033336>;PAGE:STRING:ARTICLE/CHAPTER.
- [27] Campbell FC. Adhesive Bonding and Integrally Cocured Structure. *Manufacturing Technology for Aerospace Structural Materials* 2006:369–418. <https://doi.org/10.1016/B978-185617495-4/50008-1>.
- [28] Joesbury AM, Colegrove PA, Van Rymenant P, Ayre DS, Ganguly S, Williams S. Weld-bonded stainless steel to carbon fibre-reinforced plastic joints. *J Mater Process Technol* 2018;251. <https://doi.org/10.1016/j.jmatprotec.2017.08.023>.
- [29] Camanho PP, Fink A, Obst A, Pimenta S. Hybrid titanium–CFRP laminates for high-performance bolted joints. *Compos Part A Appl Sci Manuf* 2009;40:1826–37. <https://doi.org/10.1016/J.COMPOSITESA.2009.02.010>.
- [30] Fink A, Camanho PP, Andrés JM, Pfeiffer E, Obst A. Hybrid CFRP/titanium bolted joints: Performance assessment and application to a spacecraft payload adaptor. *Compos Sci Technol* 2010;70:305–17. <https://doi.org/10.1016/J.COMPSCITECH.2009.11.002>.
- [31] Hannemann B, Backe S, Schmeer S, Balle F, Breuer UP. New multifunctional hybrid polymer composites reinforced by carbon and steel fibers. *ICCM International Conferences on Composite Materials*, vol. 2015- July, 2015.
- [32] Rehra J, Hannemann B, Schmeer S, Hausmann J, Breuer UP. Approach for an Analytical Description of the Failure Evolution of Continuous Steel and Carbon Fiber Hybrid Composites. *Adv Eng Mater* 2019;21. <https://doi.org/10.1002/adem.201800565>.
- [33] Rehra J, Andriß C, Schmeer S, Breuer UP. Describing the Material Behavior of Steel and Carbon Fiber Reinforced Composites Using a Combined Damage-Plasticity Approach. *Journal of Composites Science* 2022;6. <https://doi.org/10.3390/jcs6080235>.
- [34] Rehra J, Jungbluth J, Katri B, Schmeer S, Gurka M, Balle F, et al. Damage and failure mechanisms of hybrid carbon fiber and steel fiber reinforced polymer composites. *Compos Part A Appl Sci Manuf* 2024;185:108366. <https://doi.org/10.1016/j.compositesa.2024.108366>.

- [35] McBride AK, Turek SL, Zaghi AE, Burke KA. mechanical behavior of hybrid glass/steel fiber reinforced epoxy composites. *Polymers (Basel)* 2017;9. <https://doi.org/10.3390/polym9040151>.
- [36] O'Brien C, McBride A, Zaghi AE, Burke KA, Hill A. Mechanical behavior of stainless steel fiber-reinforced composites exposed to accelerated corrosion. *Materials* 2017;10. <https://doi.org/10.3390/ma10070772>.
- [37] O'Brien CN, Zaghi AE. Modelling the nonlinear shear stress-strain response of composites with metal and non-metal reinforcement. *Compos B Eng* 2021;221. <https://doi.org/10.1016/j.compositesb.2021.109009>.
- [38] Cózar IR, Otero F, Maimí P, González E V., Miot S, Turon A, et al. A three-dimensional plastic-damage model for polymer composite materials. *Compos Part A Appl Sci Manuf* 2022;163. <https://doi.org/10.1016/j.compositesa.2022.107198>.
- [39] Cózar IR, Otero F, Maimí P, González E V., Turon A, Camanho PP. An enhanced constitutive model to predict plastic deformation and multiple failure mechanisms in fibre-reinforced polymer composite materials. *Compos Struct* 2024;330. <https://doi.org/10.1016/j.compstruct.2023.117696>.
- [40] TORAYCA® Technical Manual. 2020.
- [41] Callens MG. DEVELOPMENT OF DUCTILE STAINLESS STEEL FIBRE COMPOSITES. KU Leuven, 2014.
- [42] ASTM D2584: Test Method for Ignition Loss of Cured Reinforced Resins. West Conshohocken, PA: 2018. <https://doi.org/10.1520/D2584-18>.
- [43] ASTM D3039: Test Method for Tensile Properties of Polymer Matrix Composite Materials. West Conshohocken, PA: 2025. https://doi.org/10.1520/D3039_D3039M-17R25.
- [44] Abaqus/Explicit 2022.
- [45] Hashin Z, Walter Rosen B. The Elastic Moduli of Fiber-Reinforced Materials. *J Appl Mech* 1964;31:223–32. <https://doi.org/10.1115/1.3629590>.
- [46] Hashin Z. Analysis of Composite Materials—A Survey. *J Appl Mech* 1983;50:481–505. <https://doi.org/10.1115/1.3167081>.
- [47] Christensen RM, Lo KH. Solutions for effective shear properties in three phase sphere and cylinder models. *J Mech Phys Solids* 1979;27:315–30. [https://doi.org/10.1016/0022-5096\(79\)90032-2](https://doi.org/10.1016/0022-5096(79)90032-2).
- [48] SOLIDWORKS Premium 2025 SP5.0 n.d.

- [49] Vogler M, Rolfes R, Camanho PP. Modeling the inelastic deformation and fracture of polymer composites – Part I: Plasticity model. *Mechanics of Materials* 2013;59:50–64. <https://doi.org/10.1016/j.mechmat.2012.12.002>.
- [50] Quintanas-Corominas A, Maimí P, Casoni E, Turon A, Mayugo JA, Guillaumet G, et al. A 3D transversally isotropic constitutive model for advanced composites implemented in a high performance computing code. *European Journal of Mechanics, A/Solids* 2018;71. <https://doi.org/10.1016/j.euromechsol.2018.03.021>.
- [51] Concrete Damaged Plasticity. ABAQUS Analysis User's Guide, vol. Volume III Materials, 2016.
- [52] Lubliner J, Oliver J, Oller S, Oñate E. A plastic-damage model for concrete. *Int J Solids Struct* 1989;25:299–326. [https://doi.org/10.1016/0020-7683\(89\)90050-4](https://doi.org/10.1016/0020-7683(89)90050-4).
- [53] Lee J, Fenves GL. Plastic-Damage Model for Cyclic Loading of Concrete Structures. *J Eng Mech* 1998;124:892–900. [https://doi.org/10.1061/\(ASCE\)0733-9399\(1998\)124:8\(892\)](https://doi.org/10.1061/(ASCE)0733-9399(1998)124:8(892)).
- [54] Naya Montáns F. Prediction of mechanical properties of unidirectional FRP plies at different environmental conditions by means of computational micromechanics. Universidad Politécnica de Madrid, 2017. <https://doi.org/10.20868/UPM.thesis.46239>.
- [55] Herráez M, Bergan AC, Lopes CS, González C. Computational micromechanics model for the analysis of fiber kinking in unidirectional fiber-reinforced polymers. *Mechanics of Materials* 2020;142. <https://doi.org/10.1016/j.mechmat.2019.103299>.
- [56] Costa S, Zrida H, Olsson R, Herráez M, Östlund R. A unified physically-based finite deformation model for damage growth in composites. *Compos Part A Appl Sci Manuf* 2022;161. <https://doi.org/10.1016/j.compositesa.2022.107103>.
- [57] Araldite LY 1564 / Hardeners XB 3403 / XB 3404-1 / Aradur 3405 - Material data sheet, Huntsman Advanced Materials (Switzerland) GmbH. 2008.
- [58] Hsieh TH, Kinloch AJ, Masania K, Taylor AC, Sprenger S. The mechanisms and mechanics of the toughening of epoxy polymers modified with silica nanoparticles. *Polymer (Guildf)* 2010;51. <https://doi.org/10.1016/j.polymer.2010.10.048>.
- [59] Faes JC, Rezaei A, Van Paepegem W, Degrieck J. Influence of matrix toughness and interfacial strength on the toughness of epoxy composites with ductile steel fabric reinforcement. *Compos Interfaces* 2015;22. <https://doi.org/10.1080/09276440.2015.1076639>.
- [60] He ZC, Qin CF, Wang YB, Li SH, Li E, Li RQ. Tensile properties of helical carbon fiber tows. *Int J Mech Sci* 2024;280. <https://doi.org/10.1016/j.ijmecsci.2024.109521>.

- [61] Yang Z, Xu Y, Zhang B, Li L, Huan H. Fracture behavior of twisted fiber bundle under tensile load. *Compos B Eng* 2026;311:113275.
<https://doi.org/10.1016/J.COMPOSITESB.2025.113275>.
- [62] Gorbatikh L, Lomov S V. Damage accumulation in textile composites. *Modeling Damage, Fatigue and Failure of Composite Materials*, 2016. <https://doi.org/10.1016/B978-1-78242-286-0.00003-0>.
- [63] Sabuncuoglu B, Lomov S V. Micro-scale numerical study of fiber/matrix debonding in steel fiber composites. *J Eng Fiber Fabr* 2020;15.
<https://doi.org/10.1177/1558925020910726>.
- [64] Shah DU, Schubel PJ, Clifford MJ. Modelling the effect of yarn twist on the tensile strength of unidirectional plant fibre yarn composites. *J Compos Mater* 2013;47.
<https://doi.org/10.1177/0021998312440737>.
- [65] Zaidi BM, Zhang J, Magniez K, Gu H, Miao M. Optimizing twisted yarn structure for natural fiber-reinforced polymeric composites. *J Compos Mater* 2018;52:373–81.
<https://doi.org/10.1177/0021998317707333>;WEBSITE:WEBSITE:SAGE;REQUESTEDJOURNAL:JOURNAL:JCMA;WGROU:STRING:PUBLICATION.

Appendix A1

Table A1.1. Input parameters for the elastoplastic damage model for transversally isotropic fiber-reinforced composite materials by C  zar et al. [38,39]

Behavior	Symbol	Description	Unit	CF/EP yarn	SF/EP yarn
Elastic (Calculated using CCA model)	E_{11}	Young's Modulus in the longitudinal direction	MPa	158618.00	106208.00
	E_{22}	Young's Modulus in the transverse direction	MPa	8809.00	8787.00
	G_{12}	Young's Modulus in the longitudinal shear direction	MPa	4176.00	3748.00
	ν_{12}	In-plane Poisson's ratio	-	0.21	0.30
	ν_{23}	Out-of-plane Poisson's ratio	-	0.40	0.38
Plastic [49]	$\hat{Y}_{TP}$	Transverse tensile potential yield stress	MPa	40.56	40.56
	$\hat{Y}_{CP}$	Transverse compressive potential yield stress	MPa	102.87	102.87
	$\hat{S}_{LP}$	Longitudinal shear potential yield stress	MPa	28.97	28.97
	$\hat{\mu}_{lp}$	Plastic potential curvature parameter plane $\tilde{\sigma}_{22} - \tilde{\sigma}_{12}$	-	0.00	0.00
	$\hat{\mu}_{tp}$	Plastic potential curvature parameter plane $\tilde{\sigma}_{22} - \tilde{\sigma}_{22}$	-	0.00	0.00
Damage [38]	X_C	Longitudinal compressive strength	MPa	1470.00	120.00
	f_{X_C}	Portion of longitudinal compressive strength	-	0.20	0.20
	$f_{X_C} * G_{X_C}$	Longitudinal compressive fracture toughness of the first segment	N/mm	26.58	26.58
	X_T	Longitudinal tensile strength	MPa	2080.00 [40]	260.00 [4]
	f_{X_T}	Portion of longitudinal tensile strength	-	0.20	0.20
	G_{X_T}	Longitudinal tensile fracture toughness	N/mm	100.00	100.00
	$f_{G_{X_T}}$	Portion of longitudinal tensile fracture toughness	-	0.50	0.50
	Y_C	Transverse compressive strength	MPa	140.00	120.00

G_{Y_C}	Transverse compressive fracture toughness	N/mm	2.80	2.80
Y_T	Transverse tensile strength	MPa	76.00	40.00
G_{Y_T}	Transverse tensile fracture toughness	N/mm	0.40	0.40
S_L	Longitudinal shear strength	MPa	98.00	50.00
G_{S_L}	Longitudinal shear fracture toughness	N/mm	0.80	0.80
μ_{sl}	Damage curvature parameter plane $\tilde{\sigma}_{22} - \tilde{\sigma}_{12}$	-	1.00	1.00
μ_t	Damage curvature parameter plane $\tilde{\sigma}_{22} - \tilde{\sigma}_{22}$	-	0.90	0.90
η_t	Damage curvature parameter plane $\tilde{\sigma}_{11} - \tilde{\sigma}_{22}$ (linear coefficient)	-	12.00	12.00
η_t^q	Damage curvature parameter plane $\tilde{\sigma}_{11} - \tilde{\sigma}_{22}$ (quadratic coefficient)	-	350.00	350.00
η_{sl}	Damage curvature parameter plane $\tilde{\sigma}_{11} - \tilde{\sigma}_{12}$ (linear coefficient)	-	9.50	9.50
η_{sl}^q	Damage curvature parameter plane $\tilde{\sigma}_{11} - \tilde{\sigma}_{12}$ (quadratic coefficient)	-	0.00	0.00

Table A1.2. Input parameters for the concrete damaged plasticity model for the epoxy matrix [51].

Behavior	Symbol	Description	Unit	Epoxy matrix
Elastic	E	Elastic modulus	MPa	3130 [57]
	ν	Poisson's ratio	-	0.35 [57]
Plastic	ψ	Dilation angle	°	29 [54]
	ε	Eccentricity	-	0.1 [51]
	$\frac{\sigma_{b0}}{\sigma_{c0}}$	Initial equibiaxial compressive yield stress / initial uniaxial compressive yield stress	-	1.32 [54]
	K	Ratio of second stress invariant on tensile meridian to that on compressive meridian	-	1.0 [54]
	μ	Viscosity parameter (representing relaxation time of viscoplastic system)	-	0.0001 [54]
Compression hardening	σ_c	Yield stress	MPa	109, 111 [55]
	ε_c^{in}	Inelastic strain	mm/mm	0, 0.32 [54]
Compression damage	d_c	Compressive damage variable	-	0, 0, 0.8
	ε_c^{in}	Inelastic strain	mm/mm	0, 0.32, 0.64 [54]
Tension stiffening	σ_t	Yield stress	MPa	75 [55]
	G_f	Fracture energy	N/mm	0.184 [58]
Tension damage	d_t	Tensile damage variable	-	0, 0.9
	u_t	Cracking displacement	mm	0, 0.0049 [54]

ANALYSIS OF MICRO-DAMAGE EVOLUTION IN INDUCTION-CURED THERMOSET COMPOSITES

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Keywords: induction heating, carbon fiber composites, epoxy, elastic properties, micro-damage

Abstract

In the present study a novel induction heating system for curing of carbon/epoxy composites has been developed and tested. Direct induction heating approach with carbon fibers acting as susceptors for heat generation was used. The novelty of the approach lies in the application of electromagnetic induction which is performed using a moving coil able to perform translation movement in x, y, z directions with respect to the plane of the manufactured plate. Intention of such design is to increase uniformity of induced temperature, which is known to be highly non-uniform as a result of material anisotropy which is in turn strongly related to laminate lay-up.

Carbon/epoxy cross-ply laminates made from unidirectional prepreg layers were manufactured using induction heating and with a pre-defined path of coil movement set up using Linux CNC software. Temperature distribution was monitored using a high resolution thermal imaging camera A6752sc from FLIR. Specimens from the induction cured laminates were cut and tested in uniaxial tensile loading to measure their elastic and micro-cracking properties. In addition DSC tests were performed to estimate the degree of cure of the specimens. The results were also compared with reference specimens, which were manufactured using conventional oven curing at a constant temperature.

1. Introduction

High performance lightweight composites, such as carbon fiber reinforced polymers are increasingly used in various engineering applications. Nevertheless, conventional manufacturing methods for production of lightweight composites are highly energy-consuming. For example, oven curing has been estimated to consume approximately 20 MJ/kg largely due to excess heating of chamber volumes and molds [1]. To reduce the high energy consumption in the composites manufacturing process alternative curing methods that can deliver heat directly to the composite part must be exploited. A highly promising direct curing method which stands out with low power consumption, ability to provide high heating rates, flexibility to deliver contactless energy to a targeted location in complex components and low health and safety requirements is electromagnetic induction heating [1-3].

Induction curing of thermoset composites holds a strong promise for fast, controlled and highly adjustable internal curing of high performance carbon fiber composite parts. Although carbon fibers have favourable electric conductivity and electromagnetic properties, the anisotropic nature of composite layer properties in fiber and transverse directions leads to significant difficulties to obtain uniform heating and a uniform degree of cure [4].

As part of a larger scale research project the non-uniformity of temperature distribution is solved by designing and applying in manufacturing a frame of a prototype with movable induction coil in 3 dimensions that uses feedback from thermal imaging camera to ensure even volumetric temperature distribution during the curing.

In the present study, the focus is on elastic and micro-cracking properties of carbon/epoxy UD, cross-ply composites manufactured with the above described induction heating rig with movable coil. Cross-ply composite plates of 300 x 300 mm were manufactured from UD carbon/epoxy prepregs. Tensile tests were performed on specimens from the manufactured plates to determine elastic properties. Loading-unloading tensile tests with increasing maximum strain levels were performed to study initiation and evolution of transverse matrix cracks. Additional characterization of the glass transition temperature and degree of cure were performed using DSC analysis. Results were compared with corresponding reference materials manufactured from the same raw materials but instead using conventional oven curing.

2. Manufacturing

2.1. Material system

Composite laminates were made from UD carbon/epoxy prepregs XC130 from Xpreg. Carbon fibers were from Pyrofil high strength carbon fiber grade and the recommended pre-preg curing cycles range from minimum of 80°C to maximum of 120°C. The curing times accordingly correspond to a range between 1 to 16h depending on the temperature. In the present work, taking into consideration the capabilities and limitations of the manufacturing rig a curing cycle of 4 hours at 100°C temperature was selected and applied. A higher temperature would be preferable, however, it was not achievable with the present set-up consisting of a power supply, a coil and an electronic controller unit. One of the challenges was also the thickness of a non-inductive glass/epoxy base plate on which carbon pre-preg layers were placed and which acts as a spacer between the coil and carbon/epoxy laminate. It was found that 2 mm thickness suits well for effective application of heat to the carbon/epoxy laminate plate via induction.

Laminates with lay-up [0/90]s were manufactured. Each laminate consists of 4 UD layers each with approximate weight of 150 g/m². The plate size was 300 x 300 mm.

2.2. Induction heating system

In the present work, a custom made induction heating rig was built. The frame design is shown in Figure 1. The system consists of the main frame, thermal imaging camera and a control/inspection unit as shown in the figure. The frame was built in similarity to 3D printers and CNC machines capable of performing independent translation movements in x, y and z axis directions controlled by Linux CNC software.

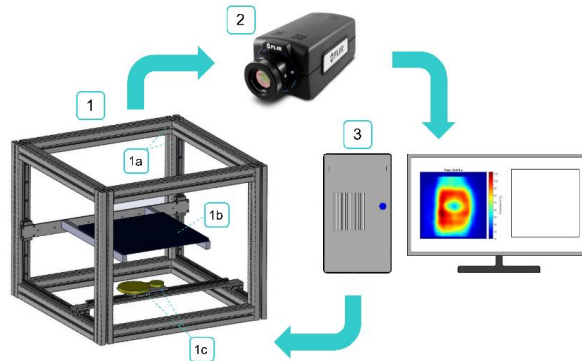


Figure 1. Drawing of an induction heating manufacturing rig: 1) main frame; 2) high resolution thermal imaging camera; 3) control and inspection unit

The size of the frame was 160 x 90 cm, and the height is approximately 60 cm. The diameter of induction coil was 150 mm. A control unit was built providing a constant frequency of electromagnetic field. The coil was connected to an adjustable power supply to control the output current of the induction coil.

2.3. Manufacturing process

As found in previous trials and in the literature [4,5], high non-uniformity of temperature is expected when the induction coil is static (no movement). The novelty of the present approach and test set-up is movable coil which can be pre-programmed to perform various heating paths in order to mitigate the non-uniformity. A trial circular heating path approximately following the inscribed circle inside the square plate of 300x300 mm was first applied leading to conclude that large cold spots near the plate corners are present irrespective of the coil output and speed of the movement. For better results of heating and curing uniformity, the heating path was applied in two consecutive rectangular paths each covering approximately one half of the composite plate, i.e., 150 x 300 mm. This approach according to observations on FLIR A6752sc thermal imaging camera during the induction heating led to the most uniform temperature distribution, however, with a drawback of increased time consumption to complete the curing of the whole 300 x 300 mm plate. The carbon/epoxy composite plate was placed on a non-conductive glass fiber base plate. Vacuum bag was applied for compacting the plate, however, only the vacuum film was put on top without any plate typically used to apply mechanical pressure as typically done in conventional hot-press manufacturing. This was done to enable a clear temperature distribution inspection with the FLIR A6752sc thermal imaging camera as shown in Figure 2.

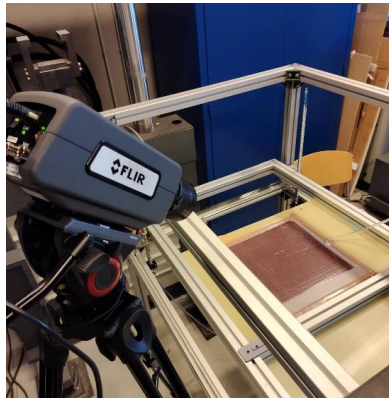


Figure 2. Overview of induction heating process: carbon/epoxy plate placed on a base plate and sealed in vacuum bag with thermal imaging camera placed to record temperature distribution from top view.

A constant output from the induction coil was applied to generate heat. Maximum temperature of 100°C was reached in a region approximately corresponding to the size of the coil. When the coil moved away from the heated part, a slight decrease of temperature was observed. An example temperature distribution is demonstrated in Figure 3.

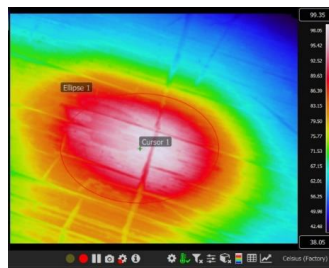


Figure 3. An instant temperature distribution during induction heating. Diameter of the highlighted region approximately corresponds to 150 mm.

As mentioned above, induction coil was following a programmed path of a rectangle corresponding to a half of the plate. In total, induction heating was applied for 4 hours on each half of the plate following the manufacturer's recommended curing cycle. It was clearly observed that the corners of the plate have significantly lower temperature, which was expected as it is typically difficult to induce the Eddy currents in the corners. For manufacturing of reference plates, a cross-ply plate with size of 150 x 300 mm was placed on a glass fiber/epoxy base plate and packed in a vacuum bag with no additional mechanical pressure applied to match conditions used in induction curing. Reference plates were cured in an oven at 100°C for 4 hours.

3. Experimental testing

3.1. Specimen preparation

After curing, plates were removed from the vacuum bag and specimens cut according to Figure 4. As shown in the figure, specimens were cut from regions which were in the central part of the plate where temperature distribution was the most uniform during the induction heating. In total 9 specimens from induction heated plate were obtained with size of 20 x 100 mm. Specimens were cut and their ends reinforced with tabs prior to testing. From the reference plate specimens with dimensions 20 x 150 mm were cut.

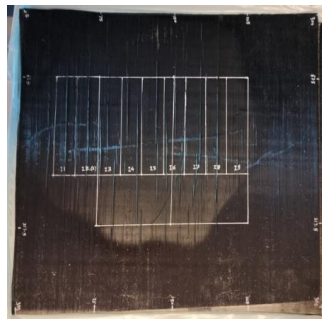


Figure 4. Specimens cut from the induction heated cross-ply laminate.

3.2. Testing procedure

Tensile tests were performed on tensile testing machine Zwick/Roell Z150 using 1% per minute tensile strain rate. At first, elastic modulus was measured, followed by cyclic loading/unloading cycles with increasing maximum strain level. Specimens were removed for optical microscopy inspection and reloaded in the testing machine to apply higher strain levels in the next step. The initiation of transverse cracking was monitored in optical microscopy by quantifying the amount of transverse (intralaminar) cracks in the 90° layers of the cross-ply laminate specimens. Additionally DSC scans were performed on both induction heated and reference plate specimens to analyze the degree of cure in both composites. DSC tests were performed on specimens cut from area in vicinity to the ends of the tensile test specimens.

3.3. Results and discussion

The elastic modulus results are summarized in Figure 5. Specimen numbering corresponds to induction cured (I1 to I4) and oven cured (O1 to O4). In general, approximately 10% higher tensile elastic modulus was found for the oven cured (reference) materials as shown in Figure 5. These results also correlate well with the nominal glass transition temperature (T_g) obtained from DSC tests which are also plotted in the results summary in Figure 5. For the oven cured specimens, the nominal glass transition temperature T_g is slightly above 120°C while for the induction cured specimens the nominal T_g is approximately equal to 100°C.

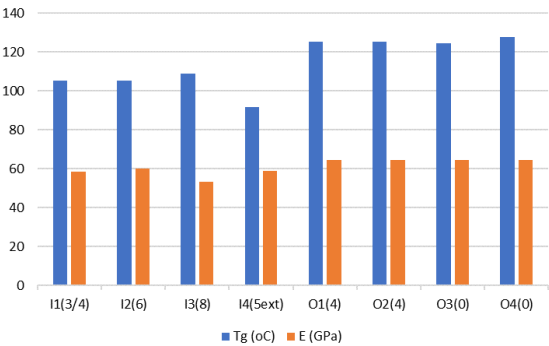


Figure 5. Summary of tensile elastic modulus and glass transition temperature from DSC tests.

Initiation and evolution of micro-damage was studied with optical microscopy and the results are summarized below in Table 1. Notation of specimens follows induction curing (I1, I2, I3, I4, I8) and oven curing (O1). According to the data summary in Table 1, micro-cracks (in form of transverse matrix cracks in the 90° layers) initiate at around 0.75% tensile strain for the oven cured (reference) specimens. For induction cured specimens, on the other hand, crack initiation takes place at significantly lower strain of 0.50% for several specimens. A significant increase of the amount of transverse cracks was observed for induction cured specimens at 0.8% strain. Such significant differences in micro-cracking were found to be related to significantly higher amount of voids in the induction cured specimens as shown in Figure 6, despite using the same vacuum pressure and curing temperature and time. Although at present only two plates have been tested, this observation deserves a more thorough study in the future work.

Table 1. Write your table caption here.

ε [%]	O1	I1	I2	I4	I3	I8
0.50	0	-	0	0	3	1
0.55	0	-	0	1	-	2
0.60	0	-	-	-	-	2
0.70	0	-	-	-	-	3
0.75	4	-	-	-	-	3
0.80	4	17	-	-	-	-

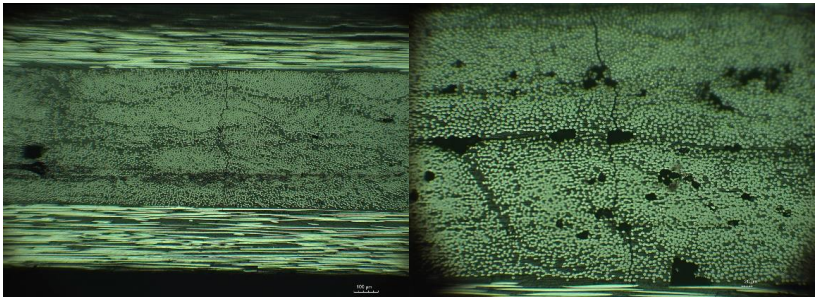


Figure 6. Comparison of micro-structure between oven cured (left) and induction cured (right) laminates showing significantly more voids in the latter.

4. Conclusions

The present study investigates elastic and micro-damage properties of induction heated cross-ply laminates and the results are compared with reference material manufactured using oven curing. In general, a higher elastic modulus was found for the oven cured material than for induction cured material. The found differences were also confirmed by DSC tests, which showed a higher nominal T_g, and thus presumably a higher degree of cure for the oven cured specimens than for induction cured specimens. Significant differences were also found in the microstructure in both materials. While oven cured specimens appeared to be in typical pre-preg quality with a few voids, the induction cured specimens had a significant amount of voids which contribute to an earlier micro-crack formation. The future work will address improvement of the induction curing process to obtain a fully-cured condition and better uniformity of the temperature during induction heating. In addition, the observed high amount of voids in induction heated specimens needs a further research.

Acknowledgments

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References

- [1] M.G. Collinson, M.P. Bower, T.J. Swait, C.P. Atkins, S.A. Hayes and B. Nuhiji. Novel composite curing methods for sustainable manufacture: A review. *Composites Part C: Open Access*, 9:art. no. 100293, 2022.
- [2] T. Bayerl, M. Duhovic, P. Mitschang and D. Bhattacharyya. The heating of polymer composites by electromagnetic induction - A review. *Composites Part A: Applied Science and Manufacturing*, 57: 27-40, 2014.
- [3] T.J. Ahmed, D. Stavrov, H.E.N. Bersee and A. Beukers. Induction welding of thermoplastic composites-an overview. *Composites Part A: Applied Science and Manufacturing*, 37(10):1638-1651, 2006.
- [4] J.B. Khan, A.C. Smith, P.M. Tuohy, M. Gresil and C. Soutis. Three-dimensional finite-element analysis multiphysics modelling of electromagnetic Joule heating in carbon fibre composites. *IET Electric Power Applications*, 14 (10):1966-1973, 2020.
- [5] F. Lundstrom, K.Frogner, M.Andersson. Analysis of the temperature distribution in weave-based CFRP during induction heating using a simplified numerical model with a cross-ply representation. *Composites Part B*, 223:art.no.109153, 2021.



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