



Multiscale modelling of matrix-crack percolation and gas leakage in fibre-reinforced composites at cryogenic temperatures

Wenkai Chang^a, L.R. Francis Rose^a, Garth M. Pearce^a, Luke P. Djukic^{a,b},
Anthony J. Kinloch^c, Chun-Hui Wang^{a,*}

^a School of Mechanical and Manufacturing Engineering, University of New South Wales, Sydney, Australia

^b Omni Tanker Pty Ltd, Smeaton Grange, NSW, Australia

^c Department of Mechanical Engineering, Imperial, South Kensington Campus, London, SW7 2AZ, UK

ARTICLE INFO

Keywords:

Extreme environments
Multiscale modelling
Matrix crack
Digital twin
Carbon fibre composites
Pressure vessels
Leak rate
Cryogenic engineering

ABSTRACT

Computational prediction of matrix-crack percolation and gas leakage in fibre-reinforced composites is critical for the design, certification, and safe operation of all-composite Type V pressure vessels under extreme temperatures. Existing approaches rely heavily on full-scale testing and empirical correlations, limiting certification by analysis. This study presents a novel multiscale modelling framework that delivers three key advances. First, a continuum damage mechanics (CDM) approach is enriched by a micromechanical damage model, enabling the determination of eleven ply-level strength and toughness parameters directly from microscale damage evolution within a representative volume element using a single set of matrix and fibre-matrix interface cohesive properties. Second, the derived CDM parameters are then used to predict the evolution of through-thickness matrix-crack density with applied load, eliminating the need for experimental measurement of crack density. Third, a computationally efficient method is developed by integrating the crack density with a doubly periodic unit-cell model incorporating cohesive surfaces to predict crack-intersection openings and the corresponding throat areas governing gas transport. Coupling these features with an orifice-flow conductance model enables the determination of through-thickness gas leakage under combined thermal and mechanical loading. The framework is validated against room-temperature experimental data and subsequently applied to predict the cryogenic leakage behaviour, thereby determining the threshold strains for matrix-crack percolation and allowable leak rates. For a representative cross-ply laminate, doubling the cryogenic matrix toughness increases the gas-percolation threshold strain at $-253\text{ }^{\circ}\text{C}$ from 0.14% to 0.36%, demonstrating that matrix toughening can substantially delay crack percolation and enhance leakage resistance at cryogenic temperatures.

1. Introduction

Accurately determining the leakage-threshold condition, i.e. the pressure at which carbon fibre reinforced polymer (CFRP) vessels begin to exceed allowable gas leak rates, remains a major challenge for the design, certification, and continuing operation of composite vessels for cryogenic fuel storage [1]. While polymer liners in Type IV vessels can provide an effective gas barrier for compressed-gas storage, cryogenic liquid-hydrogen service introduces liner-related challenges, such as thermal-contraction mismatch, embrittlement, permeation, and interfacial gas accumulation, causing liner delamination and wrinkling [2]. Linerless Type V vessels avoid these issues and require the CFRP wall to

serve simultaneously as the load-bearing structure and gas barrier. The key challenge is therefore to predict when damage within the composite wall evolves into connected leakage pathways that compromise the vessel's gas barrier function. Under increasing thermomechanical loading, matrix-dominated damage can evolve progressively from fibre-matrix debonding to discrete microcracking, then to ply-splitting matrix cracks (i.e., transverse cracks that extend the ply thickness and span the laminate width [3]), and ultimately to interconnected, multi-layer crack networks throughout the entire laminate thickness, referred to as percolation. This multiscale damage evolution spans approximately nine orders of magnitude (10^9) in length scale, ultimately giving rise to continuous pathways for gas leakage [4–6], as illustrated in Fig. 1.

This article is part of a special issue entitled: Extreme Environments published in Composites Part B.

* Corresponding author.

E-mail address: chun.h.wang@unsw.edu.au (C.-H. Wang).

<https://doi.org/10.1016/j.compositesb.2026.114036>

Received 25 April 2026; Received in revised form 28 June 2026; Accepted 26 July 2026

Available online 27 July 2026

1359-8368/© 2026 The Authors. Published by Elsevier Ltd. This is an open access article under the CC BY license (<http://creativecommons.org/licenses/by/4.0/>).

The vast scale separation and inherent complexity of the matrix cracking process pose significant challenges to existing analytical and computational approaches, which consequently tend to rely heavily on full-scale experimental testing.

The onset of percolation is typically marked by an abrupt increase in leak rate that is several orders of magnitude higher than diffusion through pristine laminates or laminates containing only isolated matrix cracks (typically $< 10^{-10} \text{ m}^3/\text{m}^2/\text{s}$ for He and H_2 at room temperature [7–9]). Existing qualification standards and design guidelines for CFRP hydrogen vessels [10–17] adopt a leak-before-burst (LBB) safety philosophy to prevent catastrophic rupture while enforcing explicit allowable leak-rate limits. For example, an H_2 allowable leak rate of $5.74 \times 10^{-6} \text{ m}^3/\text{m}^2/\text{s}$ is specified for cryogenic launch-vehicle tanks [17], with substantially stricter limits applied to on-road hydrogen storage systems in international standards [10–12].

While clear pass-or-fail certification criteria have thus been defined at a prescribed limit pressure [10–12,15,16], there remains a lack of physics-based tools to predict the leakage-threshold load or pressure at which these limits are exceeded. Consequently, current design practice for all-composite Type V pressure vessels remains largely empirical and *ad hoc*, relying on costly trial-and-error prototype and full-scale testing, which significantly increase development time and certification costs.

Kumazawa and co-workers [7,18,19] established systematic biaxial-tension experiments and a semi-analytical framework that links matrix-dominated damage in CFRP laminates to through-thickness gas leakage. In their approach, leakage was modelled as a network-flow problem in which gas transport through the thickness of a laminated composite was governed by the local conductance at intersections between matrix cracks in adjacent plies, commonly idealized as orifice-type throats within a multilayer conductance model [7,20], as shown in Fig. 2(a). Subsequent work [7,18–21] identified two key parameters for leakage rates: (a) the ply-wise matrix crack density, ρ , and (b) the throat areas, A , formed at the intersections of matrix cracks across the interface between adjacent plies, as depicted in Fig. 2. This throat area is basically the product of the two crack-tip opening displacements (CTODs or Δ) evaluated at the ply interface. The leak rate is conventionally related to the throat A through an empirically determined constant [7,18–21].

To address the empirical calibration of the conductance constant, Kumazawa and Whitcomb [22] combined finite element analysis (FEA) with computational fluid dynamics (CFD) to predict leak rates. In their approach, FEA was performed on a unit cell with periodic boundary conditions representing a doubly periodic array of matrix cracks.

However, the appropriate crack spacings corresponding to different applied load levels were taken directly from experimental measurements reported in Ref. [7], rather than predicted. The unit-cell model also incorporated interply delamination of varying lengths, with the appropriate length being determined via trial and error to match experimentally measured leak rates under purely thermal loading (i.e., in the absence of mechanical load) reported in Ref. [7]. The same calibrated delamination length was then used for predicting leak rates under combined loading conditions [22]. For the CFD analysis, simplified analytical profiles were assumed for crack opening displacements (CODs), being linear in the outer plies and quadratic in the inner ply whilst retaining the crack tip openings obtained from FEA but neglecting delamination. The predicted leak rates exceeded experimental values by about 70%, a discrepancy attributed to surface-roughness effects in real laminates, whereas the CFD model assumed idealized smooth crack surfaces [22].

While the studies discussed above have considerably advanced the understanding of gas leakage mechanisms in CFRPs resulting from the percolation of matrix cracks and provided useful empirical estimation tools, significant gaps remain. First, existing gas-leakage models rely heavily on destructive or post-mortem experimental data, which are often unavailable under relevant service conditions, particularly at cryogenic temperatures where matrix cracking profiles are difficult to obtain. Second, there is a lack of an integrated multiscale modelling framework capable of predicting the initiation, growth, coalescence, and percolation of matrix cracks that govern gas leakage.

Existing cohesive-zone modelling (CZM) provides a physically sound approach for calculating crack-tip opening displacements (and thus throat areas) [23]. However, its implementation requires predefining potential crack paths to insert cohesive elements [24], which becomes impractical when the crack locations and spacings are not known *a priori*. In contrast, continuum damage mechanics (CDM) enables the efficient simulation of progressive matrix-dominated failure in full-scale laminates without explicit crack discretization [25,26]. Nevertheless, conventional CDM faces two major limitations: (a) it requires a large set of ply-level strength and fracture-energy parameters, many of which are difficult to measure or lack standardized characterisation methods [25]; and (b) it treats matrix cracking as a smeared band of damaged elements [27–29], which does not directly provide the crack openings needed to quantify crack-intersection throat areas. Recent advances in micro-mechanical damage models (MDM) have addressed the first limitation by reducing the number of model parameters to a small set of constituent-level material properties of the matrix and the fibre–matrix

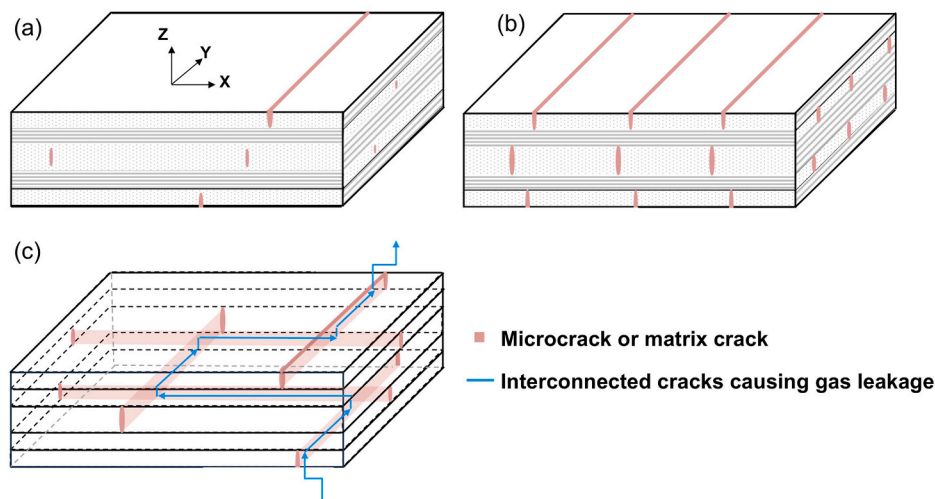


Fig. 1. Matrix-dominated damage progression in a cross-ply CFRP laminate under biaxial tension and its implication for leakage-controlled qualification. (a) Through-thickness growth of dispersed microcracks leading to matrix cracks that split plies. (b) Multiplication of matrix cracks with increasing load, forming a quasi-periodic crack array. (c) Linking of multilayer intersecting matrix cracks forms a through-thickness crack network causing gas-percolation leakage.

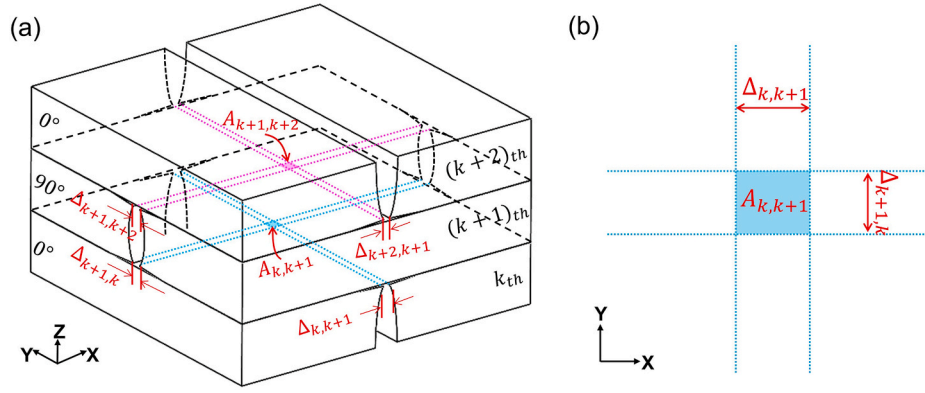


Fig. 2. Schematic of the multilayer conductance-based leakage model for cross-ply CFRP laminates. Ply indices are defined along the laminate thickness (Z-axis), increasing from bottom to top (i.e. in the positive Z-direction), such that the lower ply is denoted as k_{th} ply, the adjacent upper ply as $(k+1)_{th}$ ply. (a) Intersections of matrix cracks in adjacent plies, illustrating the formation of crack-intersection throats at ply interfaces. (b) Plan-view definition of the throat area at a crack intersection on the k_{th} – $(k+1)_{th}$ ply interface, denoted as $A_{k,k+1}$. This area is equal to the product of the CTODs of the intersecting matrix cracks in the k_{th} ply and the $(k+1)_{th}$ ply, denoted $\Delta_{k,k+1}$ and $\Delta_{k+1,k}$, respectively. Here, the first subscript identifies the ply containing the crack, while the second subscript indicates the adjacent ply with which the matrix crack intersects. The resulting throat area, $A_{k,k+1}$, governs the local conductance.

interface [25]. However, direct full-field application of MDM at a laminate scale remains computationally prohibitive due to the micron-scale resolution required to resolve fibres and interfaces [25,30].

The present work addresses these challenges by developing a multiscale framework to predict matrix-crack percolation and gas leak rate in carbon fibre composites. Central to this approach is a coupled MDM–CDM strategy [25] for predicting ply-wise crack density under biaxial loading. Under a given applied load, matrix-crack spacings within the plies throughout the laminate are obtained from the MDM–CDM model, while the corresponding crack-intersection throat areas are subsequently determined using a computationally efficient unit cell model. With the matrix crack densities and throat areas determined, the leak rate is computed following the network flow approach of [7], which requires a discharge coefficient. It is shown that a commonly employed value for micron-scale orifice flow [31–33] yields predictions in good agreement with the experimental data reported in Refs. [7,18,19], thereby providing a fully predictive, physics-based digital twin model for laminate leakage. Compared with existing approaches, the proposed framework offers several advantages: (a) it provides a multiscale, physics-based prediction of crack density as a function of applied load, eliminating reliance on experimental crack-density data that may be unavailable (e.g., under cryogenic conditions); and (b) it employs a unit-cell model in which matrix crack planes and ply interfaces are modelled using a cohesive zone model (CZM) [23,34], enabling physically consistent predictions of the throat openings at matrix-crack intersections under coupled thermal and mechanical load, rather than relying on empirically prescribed delamination lengths.

The remainder of this paper is structured as follows. Section 2 details the multiscale methodology for determining matrix crack densities and crack-intersection throat areas. Section 3 verifies several key aspects of the proposed modelling and experimental methods. Section 4 reports the predicted evolution of matrix crack density and compares these results with experiments. These predicted damage states are then employed to parameterize the cohesive-network unit cell model for predicting the leak rates at various load levels in Section 5. These predictions are compared with allowable leak-rate limits in existing standards to determine the leakage-threshold load at both room temperature (RT) and cryogenic temperatures (CTs), which are representative of extreme-environment service conditions. The merits and limitations of the present approach are briefly discussed in Section 6, followed by concluding remarks in Section 7.

2. Multiscale modelling framework

The aim of the proposed framework is to predict the two critical damage-state parameters required for through-thickness gas leakage modelling: (a) the matrix crack density in each ply, as a function of applied load or strain, ϵ_{app} , and (b) the CTODs at crack-intersection throats, as noted in Section 1 and in Fig. 2. For this purpose, the present modelling approach includes two novel features relative to prior leakage analyses [7,18–21]. First, the evolution of ply-wise crack density under biaxial loading is explicitly predicted using a multiscale MDM–CDM strategy [25] in Sections 2.1–2.2. Second, based on the crack-density state obtained from the MDM–CDM model, the corresponding crack-intersection throat areas are determined using a computationally efficient unit cell model in Section 2.3. The resulting density and throat areas are then employed to predict leak-rate evolution with applied load via a compressible-gas orifice model described in Section 2.4. The flowchart in Fig. S1 of Supplementary Material schematically describes the proposed modelling framework. To facilitate direct comparison of the present model with reported biaxial tension and gas leakage measurements in Refs. [7,18,19], predictions are obtained for the IM600/#101 carbon-fibre/epoxy system [7,18,19], employing properties for the IM600 fibres, the #101 epoxy, and the fibre-matrix interface [25,34], as summarized in Table S1 in Supplementary Material.

2.1. Micro-scale model for the ply-level traction-separation relationships

To significantly reduce the number of model parameters required by the continuum damage mechanics [25] to a small set of constituent properties for the matrix, fibre, and fibre-matrix interface, micro-mechanical modelling of a representative volume element (RVE) is employed, as shown in Fig. 3(a). The RVE dimensions, fibre distribution, and the minimum inter-fibre spacing are determined using a sensitivity method [25] to ensure statistically representative responses. Fibres are modelled as straight, transversely isotropic, linear-elastic solids. The matrix is described by a pressure-dependent elastic–plastic constitutive law with continuum damage and crack-band regularization [25], implemented in Abaqus/Explicit [35] via a user-defined VUMAT. The fibre–matrix interface is represented by zero-thickness cohesive elements with a coupled bilinear traction–separation law [25]. Interface damage initiates according to a stress criterion and evolves according to a mode-independent fracture energy G_{int} , consistent with experimental evidence that interface toughness is a fraction of the bulk epoxy's toughness, i.e., $G_{int} = \beta G_{c,ep}$ [25]. The peak interface traction, S_{int} , is

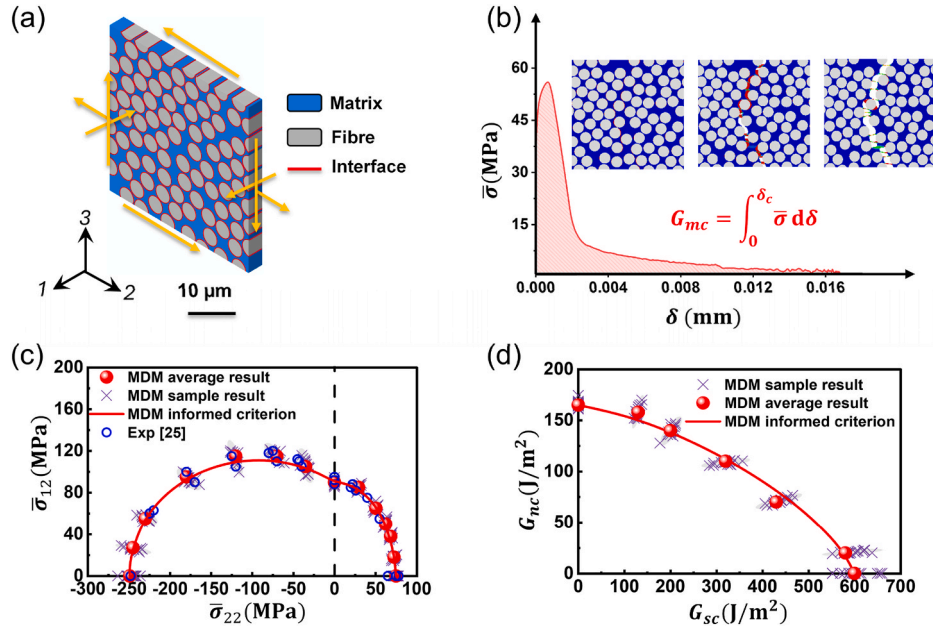


Fig. 3. MDM for matrix-dominated damage. (a) Micromechanical RVE of a CFRP ply under multiaxial loading states. (b) Computation of the traction-separation relationship of the matrix-dominated damage by fibre-matrix debonding and matrix degradation. (c) Strength envelope and (d) fracture energy envelope for plies obtained from the MDM model.

taken as the matrix strength, as discussed in Ref. [25]. The scaling factor $\beta = 0.1$ is determined using the inverse approach reported in Ref. [25], where the MDM-predicted transverse ply strengths are calibrated against experimental measurements of a $[90]_8$ carbon fibre/epoxy laminate initially at room temperature. The calibration results are summarized in Fig. S2(a and b) of the Supplementary Material. Unified periodic boundary conditions [25] are applied in all three principal directions to enable homogenization under general multi-axial loading. Thermal residual stresses due to the difference between curing temperature and operating temperature are included by an initial thermal step prior to mechanical loading.

Computational simulations are then performed by prescribing normal and shear separations, Δu_n and Δu_s , to the RVE along loading paths, each being characterised by a fixed mode-ratio $\lambda = \Delta u_s / \Delta u_n$. All analyses are performed under quasi-static conditions using Abaqus/Explicit [35] with energy balance checks to ensure negligible spurious kinetic or hourglass energy. The average stresses along the unit cell edges and separations across the opposing edges of the unit cell are then post-processed into size-independent traction-separation relations by decomposing the total separation into elastic and inelastic components; the latter defines the crack-band separations, as follows [25]:

$$\delta_n = \Delta u_n - \bar{\sigma}_n / K_n, \quad \delta_s = \Delta u_s - \bar{\sigma}_s / K_s, \quad (1)$$

where K_n and K_s denote the initial normal and shear stiffnesses of the RVE model, $\bar{\sigma}_n(\delta_n)$ and $\bar{\sigma}_s(\delta_s)$ represent the cohesive tractions. The fracture energy for a given mode is calculated as the area under the corresponding traction-separation curve. For example, the calculation of the matrix-cracking toughness G_{mc} under the pure normal loading is depicted in Fig. 3(b). For mixed-mode loading, equivalent separation and traction measures are introduced [36],

$$\delta_{eq} = \sqrt{\delta_n^2 + \delta_s^2}, \quad \bar{\sigma}_{eq} = \sqrt{\bar{\sigma}_n^2 + \bar{\sigma}_s^2}, \quad (2)$$

so that each MDM simulation with a fixed λ gives an effective traction-separation curve $\bar{\sigma}_{eq}(\delta_{eq}; \lambda)$. By varying the value of λ , the MDM generates strength and fracture-energy envelopes for matrix-dominated failure, as shown in Fig. 3(c and d), which are fitted with compact

power-law relations, i.e., $\left(\frac{G_n}{G_{nc}}\right)^m + \left(\frac{G_s}{G_{sc}}\right)^m = 1$, to determine the mode-mixity coefficient, m , where G_n and G_s denote the dissipated energy in the normal and shear directions; G_{nc} and G_{sc} the fracture energies for pure normal- and pure shear-mode failures, respectively. The complete traction-separation response is then described using a power-law softening law parameterized by the cohesive strengths and fracture energies, providing a unified description over the range of mode ratios. In this way, numerous ply-level strength/toughness quantities that would otherwise need to be determined experimentally or empirically for mesoscale ply-level modelling are instead derived directly from the MDM based on relatively few constituent properties as inputs.

2.2. Meso-scale modelling of matrix crack densities

At the mesoscale, each ply is modelled as a homogeneous orthotropic solid. Matrix-dominated intralaminar damage (manifesting as matrix cracking under transverse tension) is represented using a crack-band CDM formulation, as shown in Fig. 4(a). Damage initiation at an element is governed by the MDM-derived strength envelope, evaluated using the element stresses σ_{22} , σ_{23} , σ_{12} . Damage evolution follows the MDM-derived traction-separation law, mapped to an element-level equivalent separation, $\Delta u_{eq} = \sqrt{(L_e \epsilon_{22})^2 + (L_e \epsilon_{23})^2 + (L_e \epsilon_{12})^2}$, where L_e is the characteristic element length. The corresponding equivalent elemental stress is defined by $\sigma_{eq} = \sqrt{\sigma_{22}^2 + \sigma_{12}^2 + \sigma_{23}^2}$. A scalar damage variable D is introduced such that the degraded response satisfies $\sigma_{eq} = (1 - D)K \Delta u_{eq}$, and $D(\Delta u_{eq})$ is selected to reproduce the MDM traction law while preserving the fracture energy through crack-band regularization, as illustrated in Fig. 4(a). This gives [25]:

$$D = 1 - \frac{\Delta u_0}{\Delta u_{eq}} \left(1 - \frac{\Delta u_{eq} - \Delta u_0}{\Delta u_c - \Delta u_0} \right)^\eta \quad (3)$$

where Δu_0 denotes the equivalent separation at damage onset, Δu_c the critical separation corresponding to complete loss of load-carrying capacity, and η the softening exponent. The required CDM parameters (i.e., ply strengths, fracture energies, the mode-mixity coefficient, and soft-

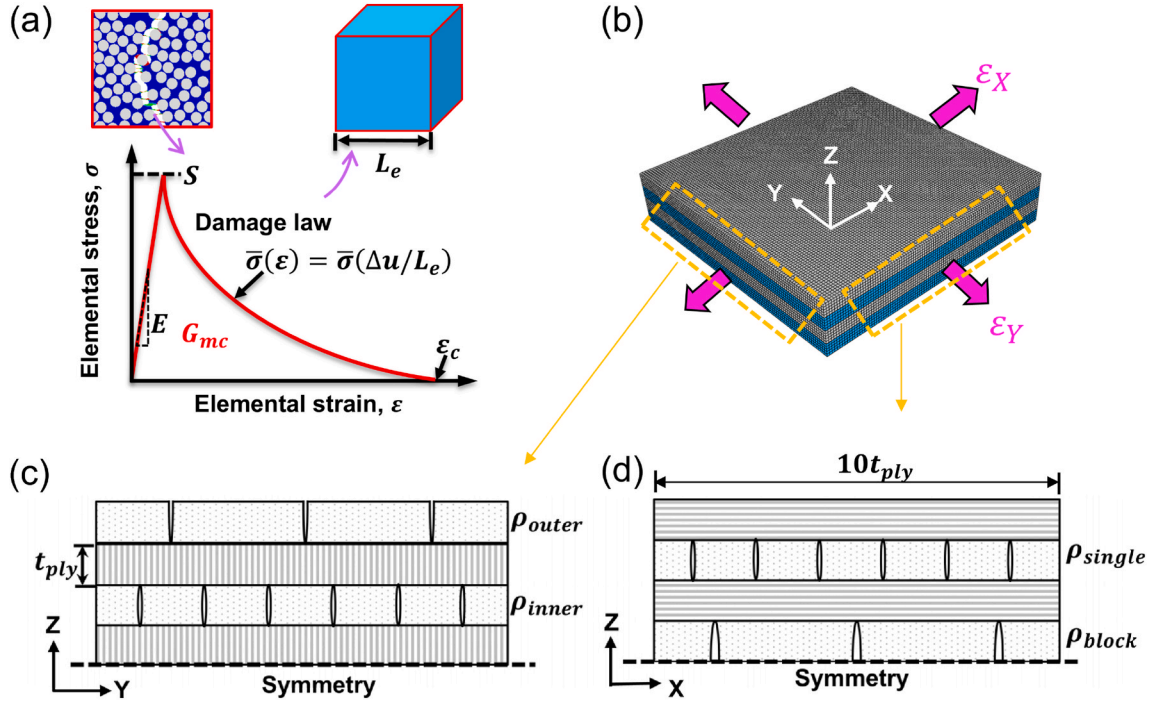


Fig. 4. Schematic of CDM and FE modelling of ply-wise matrix crack density under biaxial tension. (a) Crack-band approach used to embed the MDM-derived softening law into a CDM formulation at the element level. (b) 3D FE laminate CDM model subjected to biaxial loading to predict ply-wise matrix cracking evolution. (c,d) Schematic of Y-Z and X-Z plane views depicting the cross-ply layout and definitions of crack-density functions for four transverse-ply categories.

ening exponents) are obtained from the MDM outputs, as summarized in Table S2 in Supplementary Material, and implemented in Abaqus/Explicit [35] via a user-defined VUMAT. Other thermoelastic properties of the CFRP system are also derived from MDM and compared with experimentally measured values, as summarized in Table S3 in Supplementary Material.

The meso-scale CDM formulation is applied to a 3D FE model of a square laminate region representative of the interior region of a large symmetric cross-ply laminate $[0/90/0/90]_S$ with nominal ply thickness $t_{ply} = 0.125$ mm, consistent with experiments reported in Ref. [19]. Half the laminate is modelled using a symmetry boundary condition at the mid-plane, as shown in Fig. 4(b). To sufficiently capture the crack density evolution and minimize edge effects, the in-plane dimension of the region is taken as ten times the layer thickness, following the convergence study in Ref. [25], and periodic boundary conditions are applied on the four faces to enforce in-plane deformation. Each ply is discretized using 3D solid elements (C3D8R). Mesh sensitivity is assessed by varying the ratio of ply thickness to element size from 1 to 9 (i.e., each ply is discretized by between 1 and 9 rows of elements in the thickness). Plies are linearly elastic prior to damage initiation. Interlaminar delamination is captured by cohesive surface elements inserted at all the 0/90 interfaces, using the same MDM-informed traction–separation and fracture parameters such that matrix cracking and delamination are governed by a consistent matrix-dominated fracture description. Cure-induced residual stresses are included by applying an initial uniform thermal step of $\Delta T = -75$ °C (from 100 °C to 25 °C) prior to mechanical loading.

In the present study, a matrix crack is identified as a contiguous band (in the Z-axis direction) of fully damaged elements (with damage variable $D = 1.0$) spanning the full thickness of the ply or blocked-ply layer. Biaxial loading is achieved through periodic boundary conditions that apply prescribed uniform displacements to the reference nodes tied to the specimen faces, producing equi-biaxial strain $\epsilon_X = \epsilon_Y = \epsilon_{app}$ for the laminate.

The matrix crack density, ρ , is defined as the number of matrix cracks in a transverse ply per unit length from the observing view plane. Fig. 4

(c and d) distinguish four crack density functions analysed in this study: ρ_{block} and ρ_{single} for a block of two-ply layer and single ply in X-Z view plane; ρ_{inner} and ρ_{outer} for inner embedded and outer plies in Y-Z view plane. The resulting $\rho(\epsilon_{app})$ data are presented later in Section 4 to validate the equi-biaxial MDM–CDM predictions of matrix crack density evolution.

2.3. Meso-scale modelling of throat areas at crack intersections

The pattern of matrix cracking in a cross-ply laminate under equi-biaxial loading can be regarded as being periodic in two orthogonal directions (i.e. doubly periodic), as a reasonable approximation, as indicated in Fig. 5(a). The average crack spacing for a given applied load level is determined from the crack density derived through the MDM–CDM approach in Sections 2.1–2.2. Under the assumption of periodicity, a computationally efficient approach to determining throat areas at interply crack intersections is to employ a unit cell model with in-plane dimensions equal to the average crack spacing, as shown in Fig. 5(a). The approach reported in Ref. [22] relied on empirical calibration, whereby a range of arbitrary interfacial delamination lengths was introduced at the tips of through-thickness matrix cracks to generate finite CTODs at crack intersections; the delamination length that reproduced the experimentally measured leak rate was then selected. By contrast, the present study employs a CZM for the ply interfaces to directly predict the CTODs and consequently the associated throat areas under combined thermal and mechanical loading. As shown in Fig. 5(a), the unit-cell model represents a symmetric cross-ply laminate patch, constructed as a repeating unit cell consistent with the laminate architecture described in Section 2.2, i.e., $[0/90/0/90]_S$. Periodic boundary conditions are applied on the four faces. The composite plies are modelled as homogeneous orthotropic solids using 3D, 8-node, linear brick elements with reduced integration and hourglass control (i.e., C3D8R elements) and are assumed linearly elastic. Matrix cracks are assumed to occur in a uniformly distributed (i.e., periodically spaced) pattern within each layer. The unit-cell dimensions are then selected

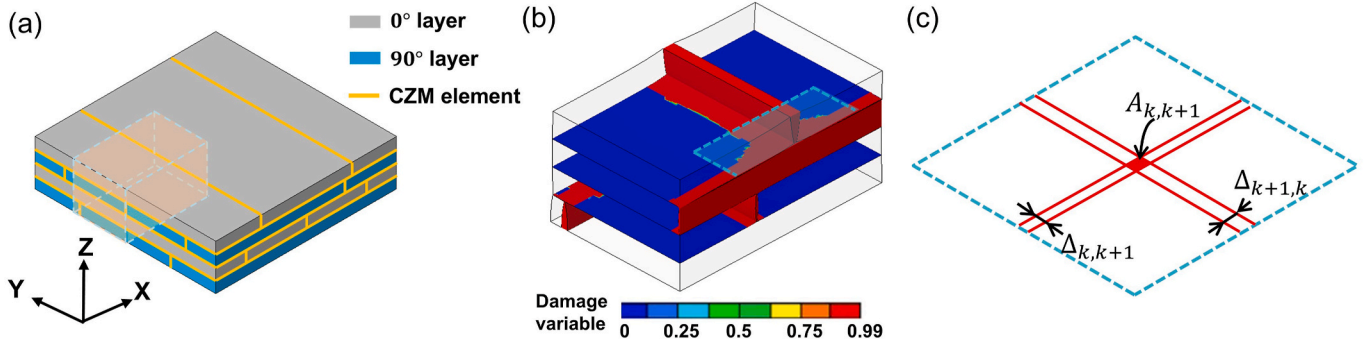


Fig. 5. Unit-cell model of doubly periodic matrix cracking for computation of crack-intersection CTOD and throat area. (a) Model configuration of the cohesive network. (b) An example damage contour plot showing intersecting matrix cracks and delamination around a crack intersection throat. (c) Plan-view definition of the crack-intersection throat geometry for the computation of throat area $A_{k,k+1}$ as the product of the CTODs, $\Delta_{k,k+1}$ and $\Delta_{k+1,k}$.

such that the model contains one representative repeat segment of the periodic crack pattern in each cracked layer, as depicted in Fig. 5(a). Zero-thickness cohesive elements available in Abaqus [35] are pre-inserted along the intralaminar planes where matrix cracking is assumed to occur as well as the ply interfaces for interlaminar delamination. The CZM parameters follow those established in Sections 2.1–2.2 and summarized in Table S2, ensuring a physically correct fracture description for both matrix cracking and interlaminar delamination. For each prescribed crack-density state, the unit cell is loaded in equi-biaxial in-plane tension up to the corresponding strain level, as determined in Section 2.2. A representative damage contour for cohesive elements at a crack intersection under $\varepsilon_{app} = 1\%$ is shown in Fig. 5(b). The CTOD is then taken as the crack-opening displacement evaluated at the ply interface. As shown in Fig. 5(c), the throat created by the intersection of matrix cracks in adjacent plies is a rectangular opening, whose side lengths are given by the CTODs of the two intersecting matrix cracks. Thus, for an individual throat formed by two intersecting matrix cracks in the k_{th} and $(k+1)_{th}$ plies at their interface, the associated throat area is $A_{k,k+1} = \Delta_{k,k+1} \Delta_{k+1,k}$, as shown in Figs. 2 and 5(c). Repeating this calculation for all intersections at ply interfaces within the unit cell yields the effective throat area (details are presented in Section 2.4, Eq. (6)) required by the compressible-gas orifice model described below.

2.4. Compressible-gas orifice modelling of leak rate in laminates with percolated matrix cracks

Having determined both the crack density in all plies and the crack-tip openings (hence throat areas) at the ply interfaces, one can next calculate the leak rate per unit area, Q , across a damaged laminate by adopting the approach originally proposed by Kumazawa et al. [7] in which Q is proportional to the pressure drop across the laminate, as follows:

$$Q = C \frac{(P_u - P_d) T_{ref}}{P_{ref} T}, \quad (4)$$

where C denotes the laminate conductance per unit area, P_u and P_d are the pressures on the two surfaces of the laminate, T is the gas temperature, and P_{ref} , T_{ref} denote a reference pressure and temperature; unless otherwise specified, they are set to 1 atm and 25 °C. The laminate conductance per unit area can be calculated using a compressible-gas orifice model [37]:

$$C = C_d A_{eff} \sqrt{\gamma R_s T} \left(\frac{2}{\gamma + 1} \right)^{\frac{\gamma+1}{2(\gamma-1)}}, \quad (5)$$

where C_d denotes the discharge coefficient, A_{eff} the effective throat area per unit laminate area (defined below). R_s is the specific gas constant of the gas, while γ , the heat-capacity ratio, is temperature dependent. For

the interface between the k_{th} and $(k+1)_{th}$ plies, the number of crack intersections per unit laminate area at a given load level is $\rho_k \rho_{k+1}$. For a general laminate, the fibre directions of two adjacent plies, namely the k_{th} ply and the $(k+1)_{th}$ ply, intersect at an oblique angle, $\theta_{k,k+1}$. Consequently, the throat area formed by the two intersecting cracks is $\Delta_{k,k+1} \Delta_{k+1,k} / \sin \theta_{k,k+1}$, following [8]. As indicated in the caption to Fig. 2, the first subscript (k or $k+1$) denotes the ply containing the split crack, while the second subscript ($k+1$ or k) denotes the adjacent ply that the split crack intersects. Considering the presence of multiple cracks in the two adjacent plies, the effective throat area fraction for this interface is $\rho_k \rho_{k+1} \Delta_{k,k+1} \Delta_{k+1,k} / \sin \theta_{k,k+1}$, where ρ_k and ρ_{k+1} denote the matrix crack densities for the two adjacent plies. For a laminate comprising multiple plies, the overall effective throat area can be determined using a reciprocal relation analogous to series conductance through successive throats across the thickness, given by

$$A_{eff} = \left(\sum_k \frac{\sin \theta_{k,k+1}}{\rho_k \rho_{k+1} \Delta_{k,k+1} \Delta_{k+1,k}} \right)^{-1}. \quad (6)$$

For the cross-ply laminates analysed in the sequel, $\theta_{k,k+1} = 90^\circ$. Combining Eqs. (4)–(6) yields the leak rate per unit area for a given load level.

For a specified gas and operating temperature, the values of γ and R_s can be obtained from standard references (e.g., Ref. [38]). In contrast, the discharge coefficient, C_d , should ideally be determined empirically for each laminate configuration and load level. Nevertheless, for regularly shaped sharp-edged orifices, C_d values in the range 0.60–0.63 are commonly reported in the literature [31–33,39] over a wide range of Reynolds numbers, Re . On this basis, a representative value of $C_d = 0.615$ is adopted in the present work. The implications and validity of this assumption are examined further in Section 6.

3. Verification of MDM–CDM–CZM model

3.1. Effect of mesh size on CDM predictions of matrix crack density

Mesh sensitivity of the MDM–CDM model was assessed under equi-biaxial tensile loading. Fig. 6(a) shows the predicted evolution of matrix crack density of the blocked ply layer, ρ_{block} , as a function of the applied strain, ε_{app} , for different element sizes, expressed as the ratio t_{ply}/L_e . The stress (or strain) for the first matrix-cracking event is essentially mesh-independent, with all curves in Fig. 6(a) starting to rise within a narrow strain band centred around $\varepsilon_{app} \approx 0.63\%$. This mesh-insensitive behaviour is attributed to the crack-band formulation, which enforces the correct fracture energy for a single localized damage band, making the onset of localization largely independent of element size. In contrast, the subsequent evolution of crack density is strongly mesh-dependent for coarse meshes. When only one or two elements are

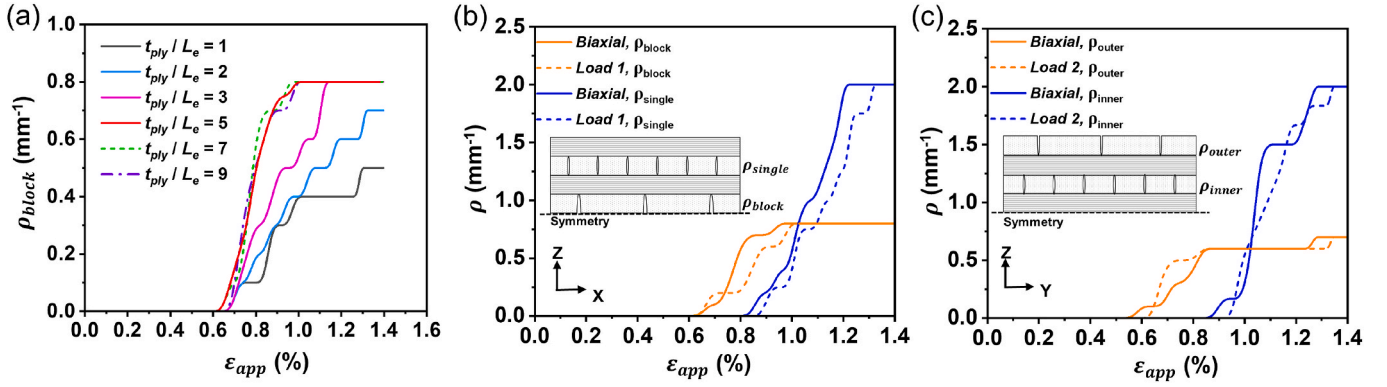


Fig. 6. Verification of the MDM-CDM modelling of matrix crack density. (a) Effect of FE mesh size L_e on the predicted ρ_{block} versus ϵ_{app} under equi-biaxial loading. Comparison of predicted crack density evolution, $\rho(\epsilon_{app})$ curves, for various transverse ply categories under equi-biaxial loading (solid lines) with sequential uniaxial "Load 1" in (b) and with "Load 2" in (c).

used in the thickness of a ply, each element row acts as a wide smeared crack band; once damage localizes, any further increase in thermo-mechanical loading will cause the damage band to increase rather than form multiple distinct crack-band zones. This leads to step-like $\rho - \epsilon_{app}$ curves. When the mesh is refined so that $t_{ply}/L_e \geq 5$, the $\rho - \epsilon_{app}$ curves nearly coincide, indicating convergence and a smeared-crack representation that more closely approximates a family of discrete matrix cracks. This convergence is also consistent with prior MDM-to-CDM comparisons [25], which showed that around five solid elements per ply are sufficient for damaged-element bands to approach discrete cracks without incurring excessive computational cost. This indicates that the converged element length is commensurate with the effective fracture-process zone size at a matrix crack tip near the interface [40]. Accordingly, $t_{ply}/L_e = 5$ (i.e., $L_e = 25 \mu\text{m}$) is adopted in the remainder of this paper for the CDM as well as the cohesive-network unit cell modelling.

3.2. Comparison between equi-biaxial and sequential uniaxial tension

While biaxial tension is the most representative loading condition for interrogating leakage-relevant matrix cracking in CFRPs, it is experimentally demanding, and challenges may be further exacerbated in extreme environments, such as at cryogenic temperatures. To assess whether biaxial loading can be adequately simulated by sequential uniaxial loading, the following loading path is also investigated computationally: (a) *Load 1*: uniaxial loading in X direction to a prescribed strain $\epsilon_X = \epsilon_{app}$ with the Y direction traction free, followed by complete unloading; and then (b) *Load 2*: uniaxial loading in Y to the same $\epsilon_Y = \epsilon_{app}$ with the X direction traction free. Predicted ply-wise matrix crack density evolutions for the four transverse-ply categories are compared in Fig. 6(b and c). It can be seen that matrix cracking initiates at a slightly lower ϵ_{app} under the equi-biaxial loading, but the difference in onset stress remains modest (i.e., within 13%). More importantly, the subsequent evolution of matrix crack densities is nearly identical under equi-biaxial and sequential uniaxial loading. The difference in crack densities at a given level of ϵ_{app} remains below about 10%. This indicates that, once matrix cracking has occurred, the progression and final extent of matrix-cracking density are largely insensitive to whether the cross-ply laminate is loaded simultaneously or sequentially in the two in-plane directions.

This similarity arises for cross-ply laminates because matrix cracking is matrix-dominated and only weakly perturbs the laminate in-plane stiffness; hence, matrix damage accumulated during Load 1 has little influence on the stress redistribution during Load 2. This weak coupling is consistent with classical laminate theory [41] for symmetric cross-ply laminates, where the coupling stiffness A_{12} (and thus the effective laminate Poisson's ratio) is small, so that sequential uniaxial loading

closely reproduces the biaxial damage state. These results indicate that sequential uniaxial testing represents a feasible and attractive experimental route for future validation of equi-biaxial matrix-cracking predictions for cross-ply laminates, particularly where true biaxial testing at cryogenic temperatures is not readily achievable.

3.3. Matrix crack COD profiles and CTODs at crack intersections

Fig. 7 benchmarks the proposed cohesive-network unit-cell model (Section 2.3) as a practical route to compute the CTOD at crack intersections, as required for leakage modelling. Fig. 7(a) compares the through-thickness COD profile, $COD(z)$, for a matrix crack in the blocked-ply layer under equi-biaxial tension at $\epsilon_{app} = 1\%$. Using the converged discretization $t_{ply}/L_e = 5$ as determined in Section 3.1, the cohesive-network prediction (red solid curve) is compared with a 2D sublamine MDM result (dashed curve), which exhibits small fluctuations due to the random fibre distribution and the resulting heterogeneous debonding/matrix cracking as shown in Fig. 7(a). In the CDM laminate model, the matrix crack is represented as a band of fully damaged elements (see Fig. 8) and therefore does not provide an explicit opening; an effective COD is instead reconstructed by interpreting this damaged band as an equivalent discontinuity and computing the increase in transverse nodal spacing across the band relative to the undeformed element length (purple solid curve). The three COD profiles show good overall agreement in both shape and magnitude, indicating that the cohesive-network unit cell captures the through-thickness displacement field and stress redistribution induced by matrix cracking. Fig. 7(b) further highlights its computational advantage. On a 56-core Intel Xeon (1.9 GHz), the cohesive-network analysis is approximately six times faster than the CDM laminate simulation and forty times faster than the 2D-MDM for obtaining the COD profiles shown in Fig. 7(a).

The key motivation for using the cohesive-network model, however, is to obtain a reliable local CTOD for calculating the crack-intersection throat area in a computationally efficient way, as discussed in Section 2. Although the overall COD profiles are comparable among different approaches, Fig. 7(c) shows that CTOD values determined using the CDM laminate model are strongly mesh dependent over the investigated range, even though the matrix crack density prediction has already converged. This highlights the major limitation of extracting the local crack-intersection opening from a smeared damage-band representation. By contrast, the cohesive-network CTOD converges for $t_{ply}/L_e \geq 5$, consistent with established CZM mesh-objectivity guidance [42], and the converged CTOD value is within 10% of the MDM reference value at a substantially lower computational cost. Accordingly, coupling the MDM-CDM's crack-density prediction with the cohesive-network unit-cell model provides an efficient and robust route to CTOD and throat

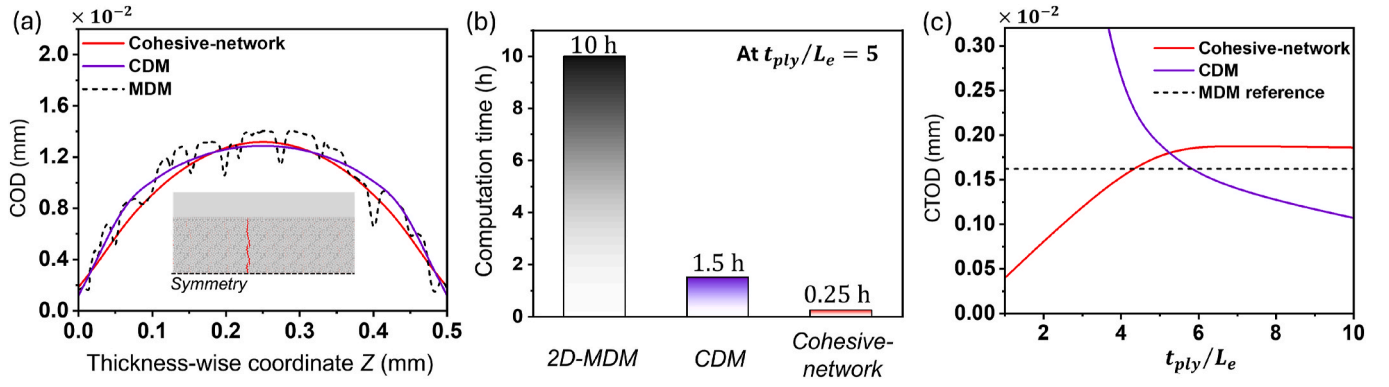


Fig. 7. Predictions of matrix crack opening and computational efficiency for the cohesive-network unit-cell model. (a) Through-thickness COD profiles by the cohesive-network unit-cell model, CDM laminate model, and 2D MDM sublaminar model (shown as inset) at $\epsilon_{app} = 1\%$. (b) Computation time for the three approaches. (c) Mesh sensitivity of CTOD predictions showing convergence of the cohesive-network prediction relative to the MDM reference and persistent mesh dependence for CDM.

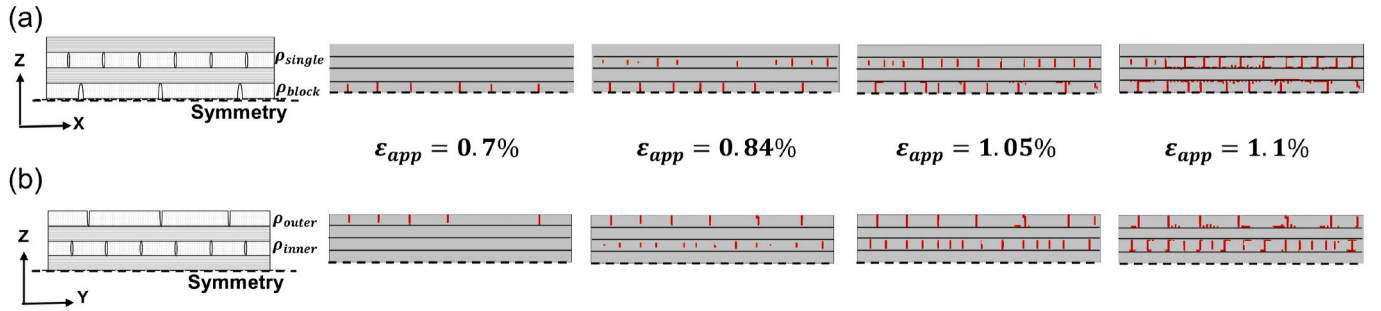


Fig. 8. Through-thickness matrix damage patterns predicted by the FE laminate model under equi-biaxial tension at various levels of ϵ_{app} . Cross-sectional views in (a) X-Z plane and (b) Y-Z plane.

area evaluation while retaining consistency with the laminate-scale damage response.

4. Predicted matrix cracking under equi-biaxial load

4.1. Matrix crack densities at room temperature and comparison with experiments

A comparison presented in Fig. 9 shows an excellent agreement between the MDM-CDM predictions of matrix crack density, ρ , and the equi-biaxial tension measurements reported by Kumazawa et al. [19] at room temperature (RT). The model predictions at RT are shown as solid

curves for the four transverse-ply categories defined in Section 2.2 (see insets of Fig. 9). Here, ρ is reported in mm^{-1} as a function of the applied laminate strain, ϵ_{app} . The intersection of each $\rho - \epsilon_{app}$ curve with $\rho = 0$ provides the predicted onset strain for matrix cracking in the corresponding ply. Symbols in Fig. 9 represent measured crack densities from C-scan radiography observations [19], which indicate that matrix cracks always span the specimen width, thereby resulting in ply splitting.

Overall, the MDM-CDM accurately captures both the onset strains and the subsequent crack-multiplication trends with increasing ϵ_{app} . The predicted sequence of matrix cracking is also consistent with observations: under equi-biaxial loading, matrix cracking initiates first in the outer ply, followed by the middle block layer. At higher applied strains,

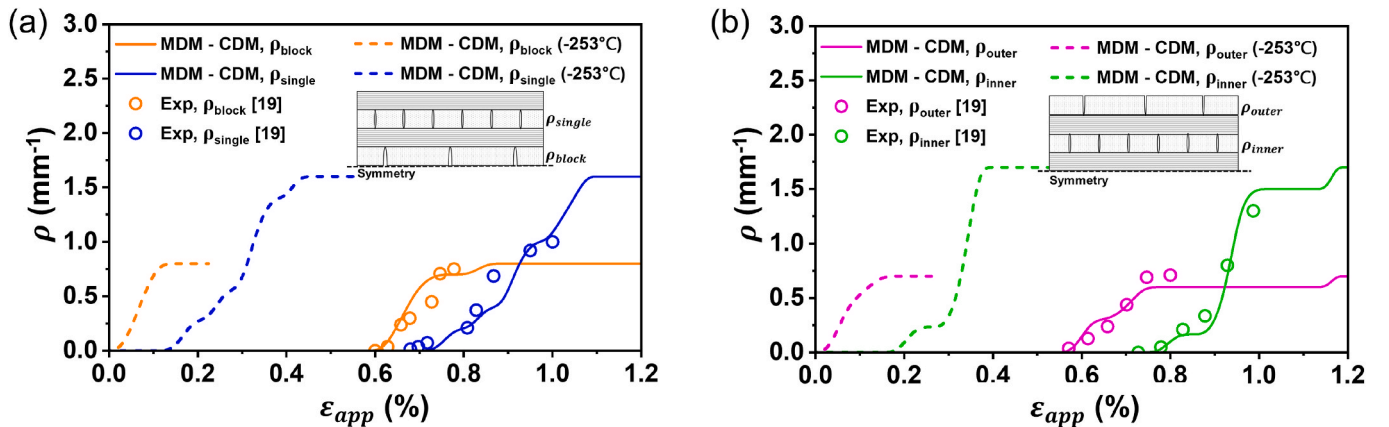


Fig. 9. Comparison of predicted $\rho - \epsilon_{app}$ results with experimental measurements [19] for (a) ρ_{block} and ρ_{single} ; (b) ρ_{inner} and ρ_{outer} . Predictions for cryogenic matrix cracking are shown by dashed curves, for which the matrix toughness, $G_{c,ep}$, is assumed to be 50% of that at room temperature.

however, the inner and single transverse plies develop systematically higher crack densities than the outer and block plies, typically by a factor of around 1.5, reflecting the layer-dependent stress redistribution and constraint within the cross-ply architecture. Collectively, these results validate the MDM-CDM framework for predicting the ply-wise matrix crack density in cross-ply laminates under biaxial loading at RT, providing the basis for extending the same modelling strategy to cryogenic temperatures in Section 4.2.

4.2. Matrix crack densities at cryogenic temperature

Also given in Fig. 9 are the resulting CT predictions, shown as the dashed $\rho - \epsilon_{app}$ curves. Cooling a CFRP laminate to cryogenic temperatures (CT) introduces relatively higher thermal residual stresses at two interconnected length scales [34]. At the intra-ply scale, residual stresses arise between the fibres and matrix because of their mismatched coefficients of thermal expansion (CTEs). At the inter-ply scale, residual stresses arise between plies of different orientations within a laminate, due to the variation in the homogenized CTE relative to the fibre orientation. To reflect a representative cryogenic tank scenario, the simulations in this section apply a temperature drop of $\Delta T = -353^\circ\text{C}$, corresponding to cooling from the curing, stress-free temperature of 100°C to liquid-hydrogen temperature of -253°C . This thermal step is applied consistently in both the MDM (in Section 2.1) and the laminate-scale CDM model (in Section 2.2), so that the subsequent equi-biaxial loading is superposed on the thermal residual stress field.

In addition to the residual stresses, temperature-dependent fracture properties, particularly the epoxy matrix fracture toughness, are expected to strongly influence matrix-dominated damage at cryogenic temperatures. In contrast, no statistically significant variation in epoxy matrix strength was observed in our prior experimental measurements [43] and this observation is also supported by other studies [44,45]. However, the epoxy modulus, E_{ep} , typically increases at cryogenic temperature, for example by 40% at CT relative to RT in Ref. [43]. Notwithstanding, this increase in the matrix modulus was found to have only a very minor influence on matrix-crack onset and crack-density evolution, as shown in Fig. S3(a) in the Supplementary Material. This weak sensitivity to the matrix modulus is consistent with the crack-band CDM formulation, in which the critical separation of an element corresponding to complete failure is governed primarily by the fracture energy and strength rather than by the stiffness [46]. Accordingly, the cryogenic simulations retain the RT matrix modulus and strength, while explicitly accounting for thermal residual stresses and reduced fracture toughness. The epoxy matrix's toughness is therefore assumed to reduce from its RT value of $G_{c,ep} = 300 \text{ J/m}^2$ to a CT value of $G_{c,ep} = 150 \text{ J/m}^2$, i.e. a 50% reduction consistent with our prior experimental measurements [45,47,48] and modelling investigations [49,50]. The fibre-matrix interface toughness is scaled consistently as $G_{int} = \beta G_{c,ep}$ at CT. Also, following the inverse approach in Section 2.1, in the present study the term β was calibrated against transverse ply-strength measurements at -196°C . As shown in Figs. S2 (c,d) in the Supplementary Material, considering the relatively large experimental scatter at CT, $\beta = 0.11 \pm 0.02$. Thus $\beta = 0.1$ remains a reasonable and consistent choice for the CT modelling. The assumption that the interface toughness remains a small fraction of matrix toughness at CT is further supported by microscopic evidence as indicated in Fig. S4 of Supplementary Material and by extensive observations in the literature [43,45,51,52], which all show that fibre-matrix interfacial debonding is the prevailing initial damage mechanism at CT.

The resulting CT predictions are shown as the dashed $\rho - \epsilon_{app}$ curves in Fig. 9. Compared to RT, the onset of matrix cracking occurs at a significantly lower applied strain, and crack densities grow earlier with increasing ϵ_{app} . This shift of the matrix cracking response directly reflects the fact that the laminate enters mechanical loading with substantial inter-ply residual strain, which is independent of internal

pressure and therefore represents an intrinsic challenge for cryogenic CFRP vessels.

5. Predicted through-thickness gas leak rate

5.1. Helium leak rate at RT and comparison with experiments

To validate the proposed through-thickness leakage modelling framework, it is benchmarked against the helium leak-rate measurements of Kumazawa et al. [7,18,19]. For direct comparison, the material system, laminate configuration, and biaxial loading conditions are set to follow Kumazawa et al. [7]. The pressure difference in Eq. (4) is set to $P_u - P_d = 1 \times 10^5 \text{ Pa}$ to match the experimental conditions in Refs. [7, 18,19]. Specifically, the laminate is an IM600/epoxy cross-ply $[0_2/90_2]_S$ composite, as shown in Fig. 10(a), and an initial thermal step from 100°C to 25°C prior to mechanical loading is set for consistency with the experiments conducted at RT. The present modelling replicates the same two-step "damage-then-leak" procedure as reported in Refs. [7,18, 19]: the laminate is first loaded under equi-biaxial tension to $\epsilon_X = \epsilon_Y = 0.61\%$ to generate matrix cracks (with $\rho = 0.1 \text{ mm}^{-1}$ determined from the MDM-CDM being subsequently employed in the unit-cell model shown in Fig. 10(a) to calculate throat areas), fully unloaded, and then reloaded to $\epsilon_{X,reload} = 0.25\%$ at biaxiality ratios $\epsilon_{Y,reload}/\epsilon_{X,reload} = -0.2, 0.31$ and 1 , during which the leak rate was measured in Refs. [7,18] and is modelled here for comparison. Because *in-situ* leak-rate measurements during progressive damage accumulation under true biaxial loading are experimentally challenging and remain lacking in the literature, available data are typically obtained either after unloading from a prescribed maximum load or during a low-level reloading stage where no additional damage develops. Accordingly, we compare the predicted leak rate during reloading directly with the available experimental measurements across various biaxiality ratios, aiming to provide a stringent validation of the present damage-to-leakage modelling approach. For additional benchmarking against existing leakage analyses, Fig. 10(d) also includes the conventional semi-empirical model reported in Ref. [7] for reloading biaxiality ratios of $\epsilon_{Y,reload}/\epsilon_{X,reload} = -0.2$ and 1 , shown as dotted curves.

The computed COD profiles across the thickness of the middle block of 90° plies are shown in Fig. 10(b) for the fully unloaded state and for the subsequent equi-biaxial reloading. The intersections of curves with the vertical axis indicate the value of the CTOD, $\Delta_{2,3}$, at the crack-intersection throat. Notably, without any applied load, i.e., $\epsilon_{X,reload} = 0$, the CTOD is non-zero when the effect of curing is ignored by setting $\Delta T = 0$ (as shown by the dashed curve in Fig. 10(b)), indicating that the residual crack opening can be attributed to thermal residual stresses. The resulting CTODs for the intersecting 90° and 0° plies are plotted in Fig. 10(c), showing approximately linear increases with the re-applied strain. Consequently, their product, representing the crack-intersection throat area, shows a quadratic dependence on strain. Using Eqs. (4)–(6) in Section 2.4, the predicted helium leak rate is presented in Fig. 10(d) for $\epsilon_{Y,reload}/\epsilon_{X,reload} = 1$, and for additional biaxiality ratios of 0.31 and -0.2 (the latter value corresponding to uniaxial tensile reloading). Across all biaxiality ratios, the model predictions are in excellent agreement with the measured helium leak rate through the matrix-cracked laminates. The present predictive multiscale model shows good overall agreement with both the experimental measurements and the conventional semi-empirical model [7] but now computing the crack density and throat-area inputs directly from the MDM-CDM and cohesive-network unit-cell models rather than by prescribing them empirically.

5.2. Hydrogen leak rate under biaxial load at room temperature and cryogenic temperatures

Building on the validation in Sections 4 and 5.1, the framework is

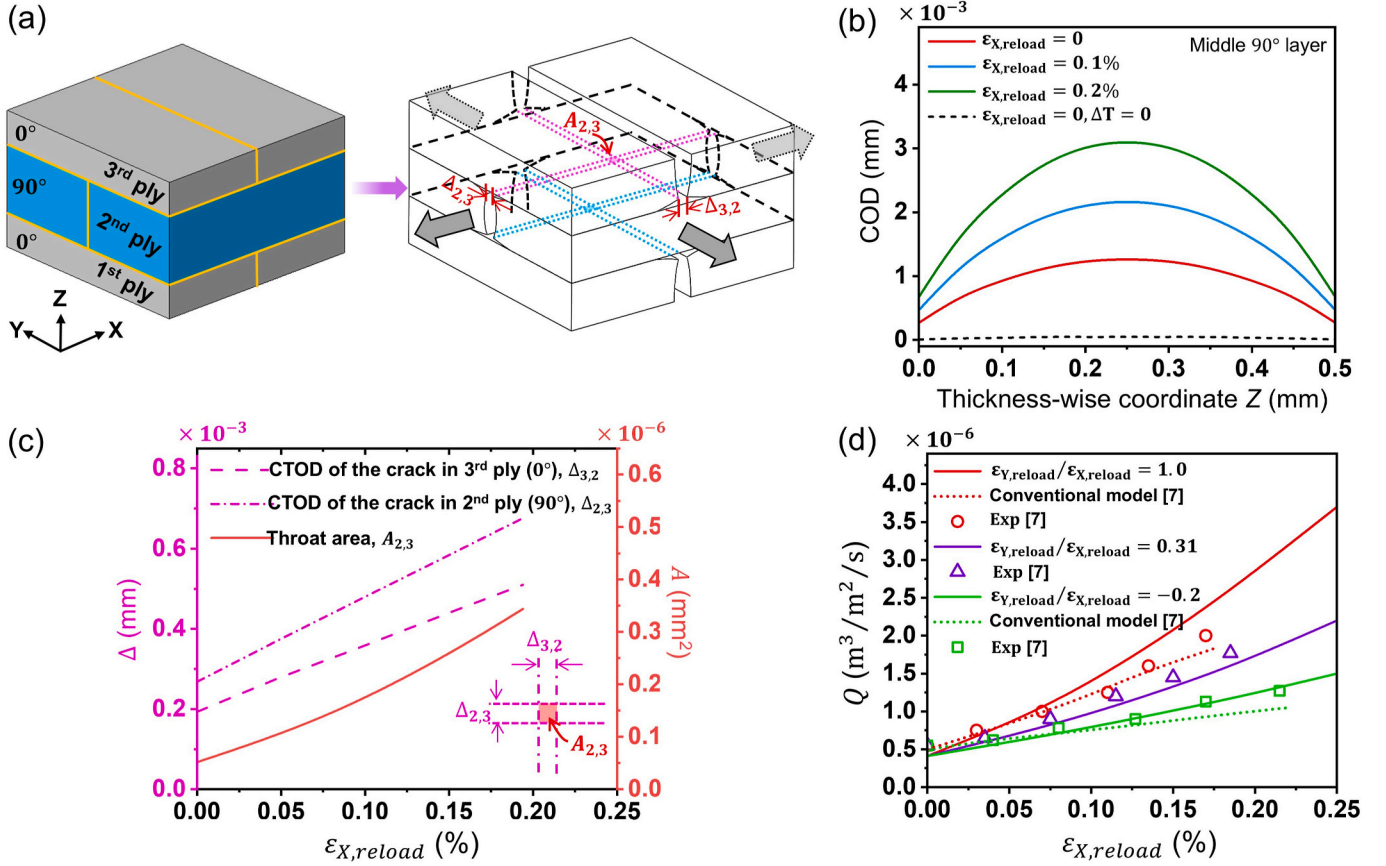


Fig. 10. Validation of the leak-rate modelling by comparing model predictions with helium leakage measurements for $[0_2/90_2]_S$ laminates reported in Ref. [7]. (a) Cross-ply laminate unit cell highlighting the cohesive-network modelling and illustration of the crack intersection. (b) Predicted COD profile by cohesive-network modelling, through the thickness of the middle 90° ply, at complete unload and during equi-biaxial reloading. (c) Predicted CTODs of the intersecting matrix cracks in the 3rd ply (0°) and 2nd ply (90°), $\Delta_{3,2}$ and $\Delta_{2,3}$, respectively, evaluated at the 2nd/3rd ply interface, and the resulting intersection throat area $A_{2,3}$ at the 2nd/3rd ply interface, as functions of re-applied strain. (d) Predicted helium leak rate for various reloading biaxiality ratios, compared with experimental measurements and the conventional semi-empirical model reported in Ref. [7].

next applied to predict the *in-situ* hydrogen leak rate under increasing load for the $[0/90/0/90]_S$ cross-ply laminate, for which experimental measurements are currently unavailable. While hydrogen can be stored in liquid form at cryogenic temperatures ($< -253^\circ\text{C}$) in the vessel, the state of the leaking fluid is governed by the local conditions within the

crack network. Since the leakage path is subjected to strong pressure reduction and thermal interaction with the surrounding structure, local flashing and gas-phase flow occur as the fluid traverses the crack path and approaches the outer surface [2,53]. Accordingly, following Pediraju et al. [53], the use of gaseous properties in Eq. (5) is adopted in

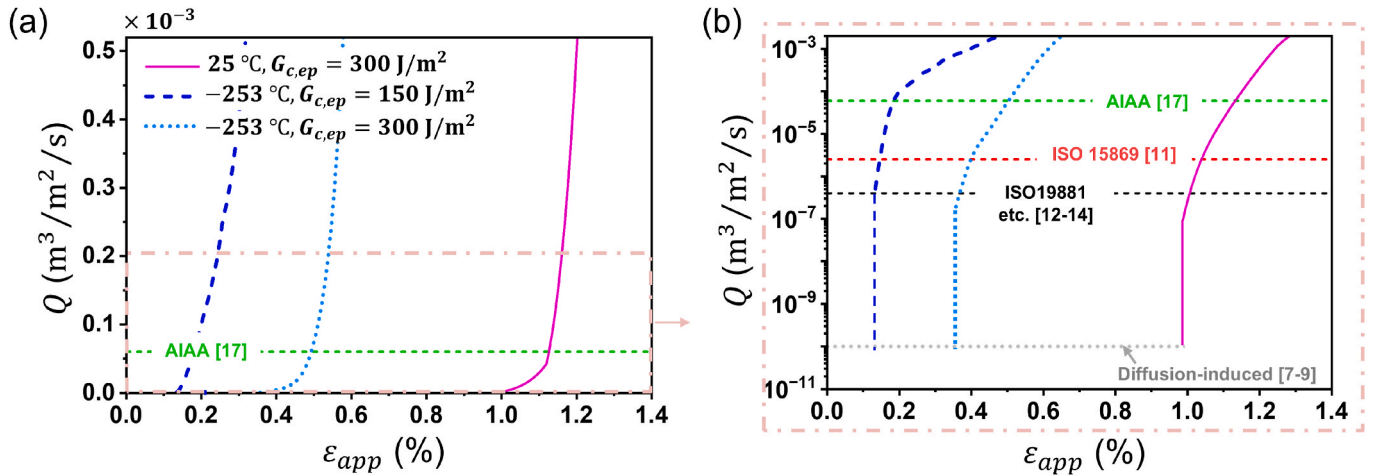


Fig. 11. Predicted *in-situ* hydrogen leak rate under equi-biaxial load for $[0/90/0/90]_S$ cross-ply laminates at RT and CT. (a) Linear scale Q - ϵ_{app} response compared with the allowable leak-rate limit from AIAA guideline [17]. (b) Zoomed plot on a logarithmic Q scale showing the diffusion-controlled leakage baseline measured in Refs. [7-9] and comparisons with representative allowable leak-rate limits from existing standards [11-14].

this study as a practical and physically reasonable first-order approximation for cryogenic leakage modelling, while detailed two-phase leakage analysis is beyond the scope of the present work. Fig. 11(a) plots the predicted $Q - \varepsilon_{app}$ response for three cases: (a) the RT baseline; (b) at -253°C with the matrix exhibiting a 50% reduced fracture energy from the RT value reported in Refs. [45,47,48], i.e., $G_{c,ep} = 150 \text{ J/m}^2$; and (c) at -253°C with a toughened matrix having $G_{c,ep} = 300 \text{ J/m}^2$. The remaining cryogenic modelling assumptions follow Section 4.2. In particular, the increased matrix modulus at CT is not included in the leakage predictions here, since the results in Fig. S3(b) of the Supplementary Material again show a negligible effect of matrix stiffening at CT on the predicted leak-rate response. The logarithmic plot in Fig. 11(b) also includes the diffusion-controlled permeation baseline, Q_{diff} , measured for pristine laminates and for laminates containing only isolated matrix cracks [7–9]. Once a through-thickness crack network forms, the leak rate increases abruptly by several orders of magnitude relative to the diffusion baseline, consistent with observations in the literature and marking the onset of gas percolation [7,8]. Such a step change in Q represents a transition in gas-transport mechanism from diffusion-controlled permeation to mass flow through an interconnected crack network. Accordingly, the gas-percolation threshold strain, ε_{cp} , is taken as when this abrupt rise above Q_{diff} occurs, giving $\varepsilon_{cp} = 0.98\%$ at RT, and a markedly lower value of $\varepsilon_{cp} = 0.14\%$ at -253°C for case (b). This shift at cryogenic temperature is consistent with the earlier onset and accelerated accumulation of matrix cracking predicted in Section 4.2 due to the combined effects of thermal residual stresses and reduced matrix toughness. The results for case (c) show that a tougher matrix with $G_{c,ep} = 300 \text{ J/m}^2$ yields a significantly higher gas percolation threshold of $\varepsilon_{cp} = 0.36\%$ at -253°C , indicating the potential benefits of matrix nano-toughening techniques for improving the gas percolation threshold at CT [45,47,48].

Also presented in Fig. 11 is a comparison between the leakage-threshold strain, ε_{lt} , defined by the intersection of the predicted $Q(\varepsilon_{app})$ curve, with the allowable leak-rate limit, Q_{lim} . Considering the AIAA guideline limit for launch-vehicle LH_2 tanks [17], shown in Fig. 11(a), the predicted thresholds are $\varepsilon_{lt} = 1.14\%$ at RT, decreasing to $\varepsilon_{lt} = 0.19\%$ at -253°C for a matrix with reduced toughness, but increasing the matrix's fracture energy raises the leakage threshold strain to $\varepsilon_{lt} = 0.51\%$. Meeting the more stringent on-road limits [12–14] by ISO15869 and ISO 19881 shown in Fig. 11(b) means that the composite vessels need to operate at lower loads.

To further examine the temperature dependence of leakage resis-

tance, the present modelling framework is used to predict the hydrogen $Q - \varepsilon_{app}$ response at a series of progressively lower temperatures down to -253°C . For simplicity, the epoxy matrix's fracture energy is assumed here to decrease linearly with temperature, with the value at -253°C taken as 50% of its RT value. The remaining cryogenic modelling assumptions follow Section 4.2. Fig. 12(a) shows the predicted $Q - \varepsilon_{app}$ response at 25, -50 , -196 , and -253°C . A clear progressive left-shift of the curves is observed as temperature decreases, indicating reduced leakage resistance at lower temperatures. Accordingly, both the gas-percolation threshold strain, ε_{cp} , and the leakage-threshold strain, ε_{lt} (with respect to the AIAA allowable leak-rate limit [17], for example), decrease monotonically with decreasing temperature. As shown in Fig. 12(b), this reduction becomes increasingly pronounced at lower temperatures, with both threshold strains dropping rapidly between -196 and -253°C , highlighting the strong sensitivity of the CFRPs' leakage performance in the cryogenic temperature regime (i.e. below -150°C [47]).

It is instructive to examine whether this simulated temperature dependence can be represented in a compact predictive form. An Arrhenius-type form [54] is adopted here as an empirical representation of the strong, nonlinear temperature sensitivity of the threshold strains, since such relations are widely used to describe thermally-activated changes in reactions and diffusion in polymer-based materials [1,54]. To this end, the threshold strain, ε_{th} , is fitted using the Arrhenius-type relation [54],

$$\varepsilon_{th} = \varepsilon_{th0} \cdot \exp\left(\frac{-E_a}{R_m T}\right), \quad (7)$$

where ε_{th} denotes either ε_{cp} or ε_{lt} , T the absolute temperature, R_m the universal gas constant, and ε_{th0} and E_a are fitting parameters. Although Eq. (7) is introduced here only as an empirical regression law rather than a mechanistic derivation, the results in Fig. 12(b) show that both threshold strains are captured well by this Arrhenius-type form. Given that direct leak testing at low and cryogenic temperatures remains technically difficult and rarely reported, this observation is valuable: once the threshold strains are validated at a limited number of benchmark temperatures, the present framework, together with the relation in Eq. (7), could provide a practical route for estimating leakage-threshold behaviour over a wider temperature range.

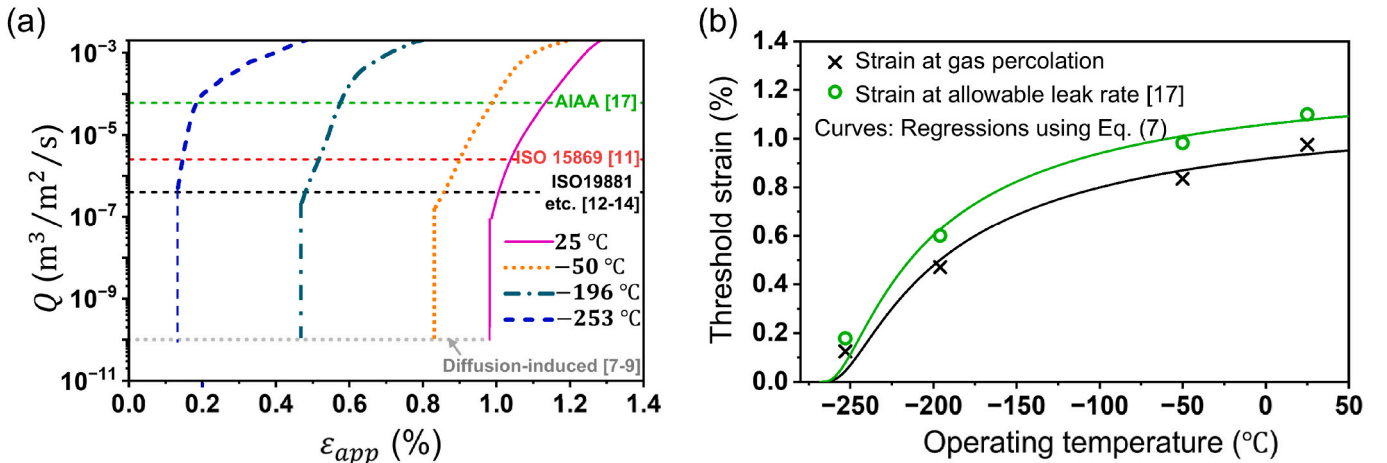


Fig. 12. Predicted temperature dependence of hydrogen leakage resistance for $[0/90/0/90]_S$ cross-ply laminates under equi-biaxial load. (a) Predicted *in-situ* hydrogen leak rate, Q , as a function of ε_{app} at various temperatures, compared with representative allowable leak-rate limits. (b) Predicted gas-percolation threshold strain, ε_{cp} , and leakage-threshold strain, ε_{lt} , with respect to the AIAA allowable limit, plotted as a function of temperature; solid curves show Arrhenius-type regressions using Eq. (7).

5.3. Predicted leakage behaviour for other cryogenic propellant storage conditions

To further illustrate the broader applicability of the present framework to cryogenic propellant tanks, Fig. 13 compares the predicted leakage response for two additional cryogens that are increasingly relevant to rocket propulsion, namely oxygen at -183°C and methane at -160°C [2], alongside hydrogen at -253°C discussed previously in Section 5.2. These temperatures correspond approximately to the boiling points of the three cryogens under practical storage conditions. In all three cases, the same damage-to-leakage modelling framework is employed, while the gas-specific parameters, R_s and γ , and temperature T in the leakage model, i.e., Eq. (5), are specialized to the corresponding gaseous cryogen, as noted in Section 5.2. The temperature-dependent matrix fracture energies are treated consistently with the assumptions introduced in Section 5.2.

As shown in Fig. 13, the predicted leakage responses for cryogenic oxygen- and methane-storage conditions are shifted to substantially higher applied strains than that for hydrogen, indicating higher leakage resistance and gas-percolation thresholds. This trend is attributed to the reduced thermal residual stresses and the higher assumed matrix fracture toughness at the less severe cryogenic temperatures associated with oxygen and methane storage. By contrast, the predicted curves for oxygen and methane are relatively close to each other, with oxygen showing a slightly lower rate of increase with applied load after percolation. This difference reflects the influence of the gas properties entering Eq. (5), given that the storage temperatures and thus the laminate damage states are comparable for cases of oxygen and methane. Thus, the onset of gas percolation is governed primarily by the fracture resistance and thermal stress state of the composite, whereas the choice of cryogen mainly affects the post-percolation leak-rate magnitude. These results further demonstrate the potential of the present modelling approach as a general virtual-testing tool for leakage-controlled design of cryogenic all-composite Type V vessels.

6. Discussion

6.1. Key modelling advances and predictive capability

The present work establishes a novel multiscale framework for simulating gas percolation through an interconnected network of matrix cracks in fibre-reinforced polymeric composites. It introduces several innovations that lead to a predictive, mechanics-based framework without recourse to empirical calibration. First, a multiscale micro-mechanical damage model (MDM) enriched continuum damage model

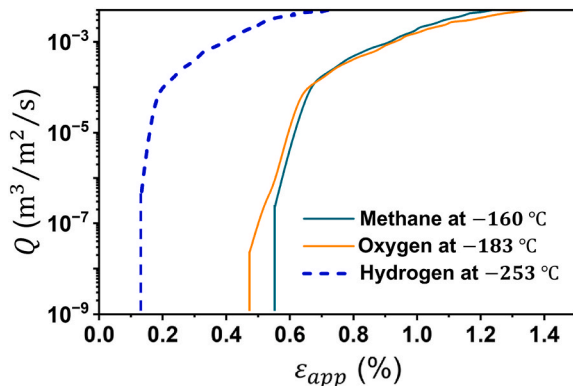


Fig. 13. Predicted leak rate response of $[0/90/0/90]_s$ cross-ply laminates under equi-biaxial loading for three cryogenic propellant storage conditions: (i) hydrogen at -253°C ($R_s = 4124 \text{ J}/(\text{kg}\cdot\text{K})$, $\gamma = 1.66$), (ii) oxygen at -183°C ($R_s = 260 \text{ J}/(\text{kg}\cdot\text{K})$, $\gamma = 1.40$), and (iii) methane at -160°C ($R_s = 518 \text{ J}/(\text{kg}\cdot\text{K})$, $\gamma = 1.33$) [38].

(CDM) is developed to determine matrix crack densities in individual plies as a function of the multiaxial load (see Fig. 9), rather than relying on post-mortem observations. Second, a scale-bridging strategy is introduced using a hierarchical homogenization strategy to quantify the crack-intersection CTODs and resulting throat geometry required by the multilayer orifice gas flow calculation. This is resolved directly using a doubly periodic cohesive-network unit cell (see Fig. 10), rather than relying on a trial-and-error approach for determining an appropriate delamination size based on only one load case (the zero applied load case). The employed MDM-CDM approach reduces reliance on many independently measured ply-level strength and fracture energy inputs, an issue that becomes particularly acute in extreme environments, such as at cryogenic temperatures, where diverse failure modes interact nonlinearly and standardized coupon testing methods are lacking [25]. In parallel, the mesh-refinement study (see Fig. 6) establishes a practical element-size guideline under which the CDM crack-band representation converges to a reliable surrogate for discrete matrix cracks, addressing a common concern with smeared continuum damage formulation in representing explicit cracks in CFRP laminates. Compared with discrete damage modelling (DDM) that typically requires prescribed crack paths and heavy enrichment of the FE mesh [28], the present MDM-CDM strategy provides a computationally efficient approach for calculating ply-wise crack density. This is achieved via a full-field laminate simulation, coupled with a computationally efficient cohesive-network unit cell model for calculating the throat areas at matrix crack intersections that are required for leakage rate prediction (see Fig. 7).

These improvements for calculating the throat areas are coupled in the present work with the discharge coefficient C_d in the throat conductance model originally proposed by Kumazawa et al. [7], given by Eq. (5), thereby providing a fully autonomous prediction of gas leak rates that does not require any empirical calibration. Thus, the present modelling approach can be considered to constitute a virtual testing capability that should prove valuable for reducing the design iteration and certification costs for all-composite Type V pressure vessels, particularly those that are required to operate at cryogenic temperatures [2]. Future work will extend the framework to more representative CFRP pressure-vessel layups and loading states, including the biaxial stress ratios characteristic of cylindrical vessels (e.g. with the axial:hoop ratio of 1:2), so that the methodology can be applied more directly to structural laminates used in practice.

6.2. Justification and limitation of the discharge coefficient in the orifice-flow modelling

The current choice of $C_d = 0.615$ falls within the range $C_d \approx 0.6 - 0.7$ that has been shown to predict measured micro-orifice flow rates within a few percent error over a wide range of Reynolds number, Re , including separated- and reattached-flow regimes [32,33]. In the specific context of cracked CFRP laminates, Kumazawa and Whitcomb [22] reported that CFD simulations assuming perfectly smooth crack faces overestimated experimental leak rates by 70%. Interpreted through the orifice framework, this discrepancy corresponds to an effective discharge coefficient of 0.59, which aligns closely with commonly reported engineering values (i.e., 0.60–0.63 [31,39,55]). Furthermore, varying C_d within 0.59–0.68 alters the prediction results by only about $\pm 6.5\%$ (see Eq. (5)). Thus, the adopted value $C_d = 0.615$ can be considered a pragmatic choice vindicated by the good agreement with measured helium leak rates under various biaxial loading ratios (see Fig. 10). Nevertheless, it is noted that Re is of the order of unity [53] in the present context, falling far below the range investigated in Ref. [32] (i.e., $Re = 1000$ to 7000), where a constant C_d of 0.68 was reported. Accordingly, further work is warranted to establish a rigorous theoretical basis for the adoption of $C_d = 0.615$ in predicting leak rate through extremely narrow cracks in composite laminates at low Reynolds numbers.

6.3. Significance of determined threshold strains

To illustrate the potential use of the current modelling capability, we have extended the validated room-temperature predictions of biaxial matrix-crack density and helium leakage (see Figs. 9 and 10) to hydrogen leakage at -253°C , explicitly accounting for multiscale thermal residual stresses and incorporating degraded matrix and interface fracture properties at this temperature. The framework enables the determination of two leakage-relevant threshold strains: (a) the gas-percolation threshold strain, ε_{cp} , and (b) the leakage-threshold strain, ε_{lt} . The significance of ε_{cp} is that it marks the transition from molecular diffusion or isolated-crack-controlled permeation ($Q < 10^{-10} \text{ m}^3/\text{m}^2/\text{s}$ [7–9]) to crack-network-dominated gas leakage ($Q > 10^{-7} \text{ m}^3/\text{m}^2/\text{s}$ [10–12,17]), thereby indicating when the CFRP ceases to function as an effective gas barrier. By contrast, ε_{lt} marks the strain at which the leak rate reaches a specified allowable leak rate limit, such as $Q = 5.74 \times 10^{-6} \text{ m}^3/\text{m}^2/\text{s}$ defined for cryogenic launch-vehicle tanks [17] and introduced in Section 1. It thus links the multiscale damage state, crack-intersection throat geometry, gas properties and qualification criterion into a practical design quantity, such as the leakage-critical pressure limit for Type V CFRP vessels.

The difference between these two thresholds, $\varepsilon_{lt} - \varepsilon_{cp}$, therefore provides a quantitative measure of the post-percolation leakage margin, i.e., the tolerance of the CFRP to connected matrix cracking before the leak rate becomes critical. In the limiting case, ε_{lt} may coincide with ε_{cp} , indicating that gas percolation immediately produces an unacceptable leak rate. Overall, these two determined thresholds will reduce the complex damage-to-leakage response into explicit design parameters that link mechanics, material design and practical qualification across temperatures and allowable leak-rate criteria.

6.4. Influence of temperature and cryogenic matrix toughness

Although systematic equi-biaxial datasets at -253°C are not yet available for direct validation [2], the modelling framework enables constituent-informed sensitivity studies that quantify how cryogenic matrix toughness governs gas-percolation onset and leakage-threshold strain. For example, doubling the cryogenic matrix toughness increases the gas-percolation threshold from $\varepsilon_{cp} = 0.14\%$ to $\varepsilon_{cp} = 0.36\%$ (see Fig. 11), indicating the scope for significantly improving the anti-leakage performance of CFRPs at cryogenic temperatures through matrix toughening. Notably, our previous nano-toughening studies [43, 45,47–49] have demonstrated that cryogenic epoxy fracture energy can be increased several-fold relative to an unmodified matrix, suggesting a practical route to extend the allowable operating window (i.e., the leakage-threshold pressure) of Type V CFRP LH₂ vessels. The framework is further extended to a range of lower temperatures down to -253°C (see Fig. 12), showing that the threshold strains decrease monotonically with decreasing temperature, with the reduction becoming increasingly pronounced in the cryogenic regime. It is particularly noteworthy that these threshold strains are well represented by an Arrhenius-type relation, suggesting a practical route for mapping leakage-relevant allowables continuously across the service-temperature range, rather than only at isolated temperatures. This is particularly relevant to the design and certification of Type V CFRP vessels for liquid-hydrogen, -oxygen, -methane and cryo-compressed storage, where leakage performance at cryogenic temperatures is a primary design constraint [2]. Furthermore, the present multiscale framework provides a pathway to future integration with molecular-dynamics-informed constituent and interface properties [56], moving toward increasingly first-principles predictions of structural integrity and gas leak rates for cryogenic pressure vessels.

7. Conclusion

The present study develops a multiscale modelling framework to

predict the percolation of matrix cracks and gas leak rate of carbon fibre reinforced composites under thermomechanical loading relevant to extreme environments, including biaxial tension at cryogenic temperatures where direct measurements are often very challenging. The key contributions include a multiscale model that uses constituent properties and a scale bridging strategy to determine key damage-state variables required for multilayer conductance leakage calculations. Specifically, we have demonstrated a new framework for determining (a) the ply-wise matrix crack density and (b) the crack-intersection opening and associated throat area that governs local conductance. By eliminating post-mortem or destructive experimental testing to quantify crack density and empirical calibration of an effective conductance used in prior analyses, the proposed framework provides a method to fully characterize the damage-to-leakage chain, enabling direct prediction of the leak-rate-load response and the leakage-threshold load for CFRP laminates at both room and cryogenic temperatures.

The matrix crack density evolution is predicted using a multiscale MDM-CDM approach. A micromechanical RVE derives matrix-dominated strengths, fracture energies, and mixed-mode softening laws that are directly upscaled to a 3D laminate CDM model under combined thermal and mechanical loading. This approach has been shown to reduce reliance on a large set of independently identified ply-level CDM parameters, many of which lack standardized test methods and become increasingly difficult to measure at cryogenic temperatures where multiple competing damage modes may interact in highly nonlinear ways. The resulting modelling capability is therefore well suited to quantify cryogenic effects, such as thermal residual stresses and temperature-dependent matrix toughness, on the onset and progression of matrix cracking. Furthermore, because true biaxial coupon testing, particularly at cryogenic temperatures, may not be readily accessible, the present simulations indicate that a more practical sequential uniaxial protocol can reproduce the key biaxial damage response of cross-ply laminates.

With the crack-density states predicted by the MDM-CDM at each load level, a computationally efficient and accurate doubly periodic unit-cell model with a prescribed cohesive network is then used to resolve local crack-intersection openings and the associated throat areas that govern gas conductance. These matrix cracking descriptors are then coupled to a compressible-gas orifice/multilayer conductance formulation to compute the through-thickness leak-rate response under combined thermal and mechanical loading. Excellent agreement with published helium leak-rate measurements obtained after biaxial loading at multiple biaxiality ratios demonstrates the predictive capability of the proposed framework. The validated model is subsequently applied to predict *in-situ* hydrogen leak rate under biaxial loading at both room temperature and -253°C , for which corresponding experimental data are not currently available. Leakage-threshold strains are identified by benchmarking against allowable limits in current qualification standards, and the cryogenic predictions further quantify the benefit of matrix toughening in increasing the threshold leakage strain. Extension of the framework across a wider low-temperature range further shows that both the gas-percolation and leakage-threshold strains decrease monotonically with decreasing temperature and can be represented by an Arrhenius-type relation, providing a practical route to determine leakage-relevant allowables for any given temperature within the service-temperature range. Overall, the present modelling framework offers a physics-based virtual testing capability for CFRP composites operating in extreme environments, enabling earlier screening of material design options and reducing iteration prior to full-scale vessel proof tests, while informing future analysis-supported leakage criteria for all-composite Type V vessel standards.

CRedit authorship contribution statement

Wenkai Chang: Conceptualization, Data curation, Formal analysis, Investigation, Methodology, Software, Validation, Visualization,

Writing – original draft. **L.R. Francis Rose**: Conceptualization, Methodology, Writing – review & editing. **Garth M. Pearce**: Conceptualization, Investigation, Methodology. **Luke P. Djukic**: Conceptualization, Investigation, Validation. **Anthony J. Kinloch**: Methodology, Writing – review & editing. **Chun-Hui Wang**: Conceptualization, Funding acquisition, Methodology, Project administration, Supervision, Writing – review & editing.

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.

Acknowledgements

This publication was made possible by FA2386-25-1-4004 from US Air Force Research Laboratory's Asian Office of Aerospace Research and Development Grant. Its contents are solely the responsibility of the authors and do not necessarily represent the official views of the Asian Office of Aerospace Research and Development. The authors are thankful for the financial support received from Australian Research Council's Discovery Grant Program (DP230101204). The authors acknowledge and thank Dr David Mollenhauer and Dr David Newell at AFRL for their insightful and constructive comments, which have significantly contributed to the improvement of this work.

Appendix A. Supplementary data

Supplementary data to this article can be found online at <https://doi.org/10.1016/j.compositesb.2026.114036>.

Data availability

Data will be made available on request.

References

- [1] Katsivalis I, Signorini V, Ohlsson F, Langhammer C, Minelli M, Asp LE. Hydrogen permeability of thin-ply composites after mechanical loading. *Compos Appl Sci Manuf* 2024;176:107867.
- [2] Bodaghi M, Nezhad HY, Nikzad M, Bayreuther CG, Nowakoski P, Koerdt M, Polle C, Krenz A, Maack B, May D. A review of technical challenges and research gaps to address permeability controls in cryogenic composite hydrogen tanks. *Compos B Eng* 2026;315:113427.
- [3] Fleck N. Compressive failure of fiber composites. *Adv Appl Mech* 1997;33:43–117.
- [4] Hohe J, Neubrand A, Fliegerer S, Beckmann C, Schober M, Weiss K-P, Appel S. Performance of fiber reinforced materials under cryogenic conditions—A review. *Compos Appl Sci Manuf* 2021;141:106226.
- [5] Berthelot J-M. 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):111–47.
- [6] Llobet J, Maimí P, Essa Y, de la Escalera FM. Progressive matrix cracking in carbon/epoxy cross-ply laminates under static and fatigue loading. *Int J Fatig* 2019;119:330–7.
- [7] Kumazawa H, Aoki T, Susuki I. Analysis and experiment of gas leakage through composite laminates for propellant tanks. *AIAA J* 2003;41:2037–44.
- [8] Yokozeki T, Ogasawara T, Aoki T, Ishikawa T. Experimental evaluation of gas permeability through damaged composite laminates for cryogenic tank. *Compos Sci Technol* 2009;69:1334–40.
- [9] Bae J, Jin J, Yun J, Park S, Lee J. Nano-engineered permeation barriers for composite laminates: polyamide nanofiber interlayers for cryogenic cycling durability and gas-barrier performance. *Compos B Eng* 2025;113219.
- [10] ISO 11119-3 Gas cylinders—Design, construction and testing of refillable composite gas cylinders and tubes. Switzerland: International Organization for Standardization; 2020.
- [11] ISO 15869 Gaseous hydrogen and hydrogen blends — land vehicle fuel tanks. London: International Organization for Standardization; 2009.
- [12] ISO 19881 Gaseous hydrogen - land vehicle fuel containers. Switzerland: International Organization for Standardization; 2025.
- [13] UN GTR No. 13 global technical regulation on hydrogen and fuel cell vehicles. United Nations; 2013.
- [14] SAE J2579 standard for fuel systems in fuel cell and other hydrogen vehicles. SAE International Technical Standard; 2023.
- [15] ANSI/AIAA S-081B-2018. Space systems—composite overwrapped pressure vessels. USA: American National Standard; 2018.
- [16] European Standard EN 12245 transportable gas cylinders-fully wrapped composite cylinders. 2022.
- [17] Robinson MJ. Determination of allowable hydrogen permeation rates for launch vehicle propellant tanks. *J Spacecraft Rockets* 2008;45:82–9.
- [18] Kumazawa H, Susuki I, Aoki T. Gas leakage evaluation of CFRP cross-ply laminates under biaxial loadings. *J Compos Mater* 2006;40:853–71.
- [19] Kumazawa H, Hayashi H, Susuki I, Utsunomiya T. Damage and permeability evolution in CFRP cross-ply laminates. *Compos Struct* 2006;76:73–81.
- [20] Roy S, Benjamin M. Modeling of permeation and damage in graphite/epoxy laminates for cryogenic fuel storage. *Compos Sci Technol* 2004;64:2051–65.
- [21] Hamori H, Kumazawa H, Higuchi R, Yokozeki T. Gas permeability of CFRP cross-ply laminates with thin-ply barrier layers under cryogenic and biaxial loading conditions. *Compos Struct* 2020;245:112326.
- [22] Kumazawa H, Whitcomb J. Numerical modeling of gas leakage through damaged composite laminates. *J Compos Mater* 2008;42:1619–38.
- [23] Elices M, Guinea G, Gomez J, Planas J. The cohesive zone model: advantages, limitations and challenges. *Eng Fract Mech* 2002;69:137–63.
- [24] Joosten M, Yang Q, Blacklock M, Wang C. A cohesive network approach for modelling fibre and matrix damage in composite laminates. *Compos Struct* 2018;206:658–67.
- [25] Chang W, Rose LF, Jiang B, Kinloch AJ, Wang CH. Can micromechanical modelling determine the continuum damage model inputs for matrix-dominated failure in fibre-reinforced polymer composites? *Compos B Eng* 2026;308:112978.
- [26] Maimí P, Camanho PP, Mayugo J, Dávila C. A continuum damage model for composite laminates: part II—Computational implementation and validation. *Mech Mater* 2007;39:909–19.
- [27] Liu H, Liu J, Ding Y, Hall ZE, Kong X, Zhou J, Blackman BR, Kinloch AJ, Dear JP. A three-dimensional elastic-plastic damage model for predicting the impact behaviour of fibre-reinforced polymer-matrix composites. *Compos B Eng* 2020;201:108389.
- [28] Shabani P, Li L, Laliberte J, Qi G, Rapping D, Mollenhauer D. High-fidelity simulation of low-velocity impact damage in fiber-reinforced composite laminates using integrated discrete and continuum damage models. *Compos Struct* 2023;313:116910.
- [29] Bazant ZP, Oh BH. Crack band theory for fracture of concrete. *Mater Construcción* 1983;16:155–77.
- [30] Arteiro A, Catalanotti G, Reinoso J, Linde P, Camanho PP. Simulation of the mechanical response of thin-ply composites: from computational micro-mechanics to structural analysis. *Arch Comput Methods Eng* 2019;26:1445–87.
- [31] Kurita CH. Critical flow restricting orifices. United States: U.S. Department of Energy Office of Scientific and Technical Information; 1988.
- [32] Tu X, Hrnjak P, Bullard C. Flow of R134a through micro-orifices. International refrigeration and air conditioning conference. Purdue University; 2002.
- [33] Hasegawa T, Suganuma M, Watanabe H. Anomaly of excess pressure drops of the flow through very small orifices. *Phys Fluids* 1997;9:1–3.
- [34] Chang W, Rose LRF, Wu S, Kinloch AJ, Wang CH. Increasing crack growth resistance for through-thickness matrix cracking and its role in suppressing ply cracking in thin-ply laminates. *Compos Appl Sci Manuf* 2022;163:107219.
- [35] ABAQUS 2024 documentation. Providence: Dassault Systèmes Simulia Corp; 2024.
- [36] Camanho PP, Davila CG, De Moura M. Numerical simulation of mixed-mode progressive delamination in composite materials. *J Compos Mater* 2003;37:1415–38.
- [37] ISO 5167-1. Measurement of fluid flow by means of pressure differential devices inserted in circular cross-section conduits running full - part 1: general principles and requirements. International Organization for Standardization; 2003. 2003.
- [38] Thermodynamics tables and charts, Wayne State University.
- [39] Bean HS. Fluid meters. sixth ed. The American Society of Mechanical Engineers (ASME); 1983.
- [40] Anderson TL. Fracture mechanics: fundamentals and applications. fourth ed. USA: CRC Press; 2017.
- [41] Clyne TW, Hull D. An introduction to composite materials. third ed. Cambridge University Press; 2019.
- [42] Turon A, Davila CG, Camanho PP, Costa J. An engineering solution for mesh size effects in the simulation of delamination using cohesive zone models. *Eng Fract Mech* 2007;74(10):1665–82.
- [43] Islam MS, Chang W, Sha Z, Wang J, Wu S, Rose LF, Kinloch AJ, Wang CH. Mitigating cryogenic microcracking in carbon-fibre reinforced polymer composites using negative thermal-expansion nanoparticles functionalized by a polydopamine coating. *Compos B Eng* 2023;257:110676.
- [44] Chen Z-K, Yang J-P, Ni Q-Q, Fu S-Y, Huang Y-G. Reinforcement of epoxy resins with multi-walled carbon nanotubes for enhancing cryogenic mechanical properties. *Polymer* 2009;50:4753–9.
- [45] Wang J, Chang W, Islam MS, Huang F, Wu S, Rose LF, Zhang J, Wang CH. Toughening epoxy by nano-structured block copolymer to mitigate matrix microcracking of carbon fibre composites at cryogenic temperatures. *Compos Sci Technol* 2024;251:110548.
- [46] Lapczyk I, Hurtado JA. Progressive damage modeling in fiber-reinforced materials. *Compos Appl Sci Manuf* 2007;38:2333–41.
- [47] Chang W, Rose LRF, Islam MS, Wu S, Peng S, Huang F, Kinloch AJ, Wang CH. Strengthening and toughening epoxy polymer at cryogenic temperature using cupric oxide nanorods. *Compos Sci Technol* 2021;208:108762.
- [48] Jiang B, Chang W, Wang X, Sha Z, Pearce G, Rose LF, Kinloch AJ, Wang CH. Increasing the flame retardancy and cryogenic toughness of an epoxy polymer using polydopamine nanoparticles. *Compos B Eng* 2025;304:112675.

- [49] Chang W, Rose LF, Sha Z, Huang F, Kinloch AJ, Wang CH. Multiscale modelling of nanoparticle toughening in epoxy: effects of particle-matrix interface, particle size, and volume fraction. *Compos Sci Technol* 2024;256:110788.
- [50] Chang W, Rose F, Wu S, Kinloch A, Wang C. Through-thickness crack growth resistance in fibre composites and its role in preventing ply cracking in cross-ply laminates. *Eng rXiv* 2022.
- [51] Fujishiro K, Ogasawara T, Mizutani K, Kumazawa H, Aoki T. Microscopic damage behavior in CFRP cross-ply laminates at cryogenic temperature. *Compos Sci Technol* 2024;256:110774.
- [52] Huang F, Chang W, Islam MS, Wang J, Jiang B, Tan Z, Wang CH. Microcracking resistance of 3D printed fibre composites at cryogenic temperatures. *Addit Manuf* 2024;89:104307.
- [53] Peddiraju P, Noh J, Whitcomb J, Lagoudas DC. Prediction of cryogen leak rate through damaged composite laminates. *J Compos Mater* 2007;41:41–71.
- [54] Humpenöder J. Gas permeation of fibre reinforced plastics. *Cryogenics* 1998;38: 143–7.
- [55] Miller RW. Flow measurement engineering handbook. New York: U.S. Department of Energy; 1996. McGraw-Hill.
- [56] Chowdhury SC, Gillespie Jr JW. Strain-rate dependent mixed-mode traction laws for glass fiber-epoxy interphase using molecular dynamics simulations. *Compos B Eng* 2024;275:111351.