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Analysis of Mechanical Behaviour and Damage of Carbon Fabric-reinforced Composites in Bending

by

Himayat Ullah

A doctoral thesis submitted in partial fulfilment of the requirements for the award of Doctor of Philosophy of Loughborough University

Wolfson School of Mechanical and Manufacturing Engineering

February 2013

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Certificate of Originality

This is to certify that I am responsible for the work submitted in this thesis, that the original work is my own except as specified in acknowledgments or in footnotes, and that neither the thesis nor the original work contained therein has been submitted to this or any other institution for a degree.

Signed......................................................................

Date...........................................................................
Dedicated to my family
Abstract

Carbon fabric-reinforced polymer (CFRP) composites are widely used in aerospace, automotive and construction structures thanks to their high specific strength and stiffness. They can also be used in various products in sports industry. Such products can be exposed to different in-service conditions such as large bending deformations caused by quasi-static and dynamic loading. Composite materials subjected to such bending loads can demonstrate various damage modes - matrix cracking, delamination and, ultimately, fabric fracture. Damage evolution in composites affects both their in-service properties and performance that can deteriorate with time. Such damage modes need adequate means of analysis and investigation, the major approaches being experimental characterisation and numerical simulations. This work deals with a deformation behaviour and damage in carbon fabric-reinforced polymer (CFRP) laminates caused by quasi-static and dynamic bending. Experimental tests are carried out first to characterise the behaviour of a CFRP material under tension, in-plane shear and large-deflection bending in quasi-static conditions. The dynamic behaviour of these materials under large-deflection bending is characterised by Izod-type impact tests employing a pendulum-type impactor. A series of impact tests is performed on the material at various impact energy levels up to its fracture, to obtain a transient response of the woven CFRP laminate. Microstructural examination of damage is carried out by optical microscopy and X-ray micro computed tomography (Micro-CT). The damage analysis revealed that through-thickness matrix cracking, inter-ply delaminations, intra-ply delamination such as tow debonding, and fabric fracture was the prominent damage modes.

These mechanical tests and microstructural studies are accompanied by advanced numerical models developed in a commercial code Abaqus. Among those models are (i) 2D FE models to simulate experimentally observed inter-ply delamination, intra-ply fabric fracture and their subsequent interaction under quasi-static bending conditions and (ii) 3D FE models based on multi-body dynamics used to analyse interacting damage mechanisms in CFRP under large-deflection dynamic bending conditions. In these models, multiple layers of bilinear cohesive-zone elements are placed at the damage locations identified in the Micro CT study. Initiation and progression of inter-laminar delamination and intra-laminar ply
fracture are studied by employing cohesive elements. Stress-based criteria are used for damage initiation while fracture-mechanics techniques are employed to capture its progression in composite laminates. The developed numerical models are capable to simulate the studied damage mechanisms as well as their subsequent interaction observed in the tests and microstructural damage analysis. In this study, a novel damage modelling technique based on the cohesive-zone method is proposed for analysis of interaction of various damage modes, which is more efficient than the continuum damage mechanics approach for coupling between failure modes. It was observed that the damage formation in the specimens was from the front to the back at the impact location in the large-deflection impact tests, unlike the back-to-front one in drop-weight tests. The obtained results of simulations showed a good agreement with experimental data, thus demonstrating that the proposed methodology can be used for simulations of discrete damage mechanisms and their interaction during the ultimate fracture of composites in bending.

The main outcome of this thesis is a comprehensive experimental and numerical analysis of the deformation and fracture behaviours of CFRP composites under large-deflection bending caused by quasi-static and dynamic loadings. Recommendations on further research developments are also suggested.

Keywords: CFRP; Large-deflection bending; Impact tests; Micro-CT; Delamination; fracture; Finite-element analysis; Cohesive-zone elements
Acknowledgment

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**Symbols**

$C_d$  
Wave speed of material

$D, d$  
Damage variable

$E$  
Elastic modulus of material

$E_{11}$  
Elastic modulus of material in longitudinal direction

$E_{22}$  
Elastic modulus of material in transverse direction

$E_{33}$  
Elastic modulus of material in thru-thickness direction

$F$  
Maximum tensile load

$F_m$  
Maximum flexural load

$G_{12}$  
In-plane shear modulus

$G_{13}, G_{23}$  
Shear modulus of material in 1-3 and 2-3 planes

$G_C$  
Fracture toughness

$G_{IC, IIc}$  
Fracture toughness in Mode I and II

$G_T$  
Total work done by interface tractions

$G_S$  
Work done by shear components of interface tractions

$h_d$  
Pendulum drop height,

$I$  
Internal forces on elements

$K$  
Interface element stiffness

$m$  
Mass

$P$  
External forces on elements

$P_{crit}$  
Delamination initiation critical load

$S$  
Span between the beam supports

$t, h$  
Specimen thickness

$w, b$  
Specimen width

$\ddot{u}$  
Acceleration

$\sigma_0$  
Interfacial strength
\( \sigma_{11,22} \) Stress component in the fibre and transvers directions, in the material coordinate system, respectively

\( \sigma_{I0}, \sigma_{III0}, \sigma_{III0} \) Cohesive interface strengths in opening Mode I and shear Modes II and III, respectively

\( \sigma_{xx}, \sigma_{yy} \) Axial and transverse components of the stress in the specimen coordinate system

\( \sigma_x \) Maximum flexural stress

\( \tau_{12} \) In-plane shear strength

\( \varepsilon \) Tensile strain

\( \gamma_{12} \) Shear strain in the material coordinate system

\( \delta \) Maximum flexural deflection

\( \delta(t) \) Dynamic deflection

\( v_i \) Velocity at impact

\( \Delta t \) Time increment size

\( \theta \) Angle

\( \nu_{12} \) In-plane Poisson’s ratio
# Acronyms

<table>
<thead>
<tr>
<th>Acronym</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>2D</td>
<td>Two-dimensional</td>
</tr>
<tr>
<td>3D</td>
<td>Three-dimensional</td>
</tr>
<tr>
<td>A-FEM</td>
<td>Augmented finite-element method</td>
</tr>
<tr>
<td>BCL</td>
<td>Bottom/back cohesive layer</td>
</tr>
<tr>
<td>CCL</td>
<td>Crack cohesive layer</td>
</tr>
<tr>
<td>CDM</td>
<td>Continuum damage mechanics</td>
</tr>
<tr>
<td>CFRP</td>
<td>Carbon fabric-reinforced polymer</td>
</tr>
<tr>
<td>CLT</td>
<td>Classical laminate theory</td>
</tr>
<tr>
<td>CZE</td>
<td>Cohesive-zone element</td>
</tr>
<tr>
<td>CZM</td>
<td>Cohesive-zone model</td>
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<tr>
<td>DCB</td>
<td>Double cantilever beam</td>
</tr>
<tr>
<td>DIC</td>
<td>Digital image correlation</td>
</tr>
<tr>
<td>DOF</td>
<td>Degree of freedom</td>
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<tr>
<td>ENF</td>
<td>End notched flexure</td>
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<tr>
<td>$F_{di}$</td>
<td>Damage initiation force</td>
</tr>
<tr>
<td>$F_{di}$-FE</td>
<td>Damage initiation force-finite-element model</td>
</tr>
<tr>
<td>$F_{di}$-Exp</td>
<td>Damage initiation force-experimental</td>
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<td>FCL</td>
<td>Front cohesive layer</td>
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<td>Finite-element</td>
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<td>Finite-element method</td>
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<td>FSDT</td>
<td>First-order shear deformation theory</td>
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<tr>
<td>GFRP</td>
<td>Glass fabric-reinforced polymer</td>
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<tr>
<td>HSDT</td>
<td>Higher-order shear deformation theory</td>
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<td>ILSS</td>
<td>Interlaminar shear strength</td>
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<td>MCL</td>
<td>Mid cohesive layer</td>
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<tr>
<td>MMB</td>
<td>Mixed-mode bending</td>
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<tr>
<td>Acronym</td>
<td>Full Form</td>
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<tr>
<td>Micro-CT</td>
<td>Micro computed tomography</td>
</tr>
<tr>
<td>NA</td>
<td>Neutral axis</td>
</tr>
<tr>
<td>NP</td>
<td>Neutral plane</td>
</tr>
<tr>
<td>RVE</td>
<td>Representative volume element</td>
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<td>TCL</td>
<td>Top cohesive layer</td>
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<tr>
<td>UD</td>
<td>Unidirectional</td>
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<tr>
<td>VCCT</td>
<td>Virtual crack closures technique</td>
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<tr>
<td>XFEM</td>
<td>Extended finite-element method</td>
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Chapter 1: Introduction

1.1 Background

Composite materials have found many applications in aerospace, automotive, medical and construction components and structures due to their better specific strength and stiffness, excellent fatigue strength, good corrosion behaviour and low thermal conductivity. There are more than 50 years of advanced research in characterising and modelling the underlying mechanical behaviour of composites and developing tools and methodologies for predicting their damage tolerance in various applications [1]. One specific class of composites - woven fabric-reinforced laminates - offer a number of attractive properties compared to their unidirectional-tape counterparts including lower production costs, better drapability, good resistance to fracture and transverse rupture due to weaving resistance, and high impact strength. These properties have attracted the sports industry to incorporate woven laminates in the design of sports products such as athletic footwear, skis, tennis and squash rackets, fishing poles, poles used in jumping and bicycle frames. Expanding the application away from the traditional aerospace structures results in new types of loading regimes such as large-deflection bending experienced in sports products, in service, which are not studied. Apart from the sports products, some mechanical and structural elements, such as helicopter blades, robot arms, transmission axles, turbine blades, may be modelled as components subjected to loading regimes with significant bending moments. These quasi-static and dynamic loads generate high local stresses and strains leading to complex damage modes due to heterogeneity and anisotropy of composite laminates. Evolution of these intralaminar and interlaminar damage mechanisms results in significant reduction of in-service mechanical properties and leads to a loss of structural integrity of the composite sports products with time. In contrast to traditional impact events in aerospace structures such as bird strike or tool drop, in sports products, kinematics, velocity and energy of dynamic event are significantly different. However, the design of sports products usually require good shock (and thus energy) absorptions, controlled level of stiffness and minimum mass, which are similar to those for aerospace structures. Additionally, due to heterogeneous
microstructure, these events are also characterised by realisation of multiple damage modes.

Quasi-static and dynamic loading of composite laminates result in complex damage mechanisms, in which matrix cracking, delaminations, tow debondings and fabric fracture are dominant ones. Though damage in composite laminates has been investigated in depth for decades, the major focus was on the case of unidirectional laminates under uniaxial tensile loads. As a result, research on initiation, progression and interaction of various damage modes in woven laminates subjected to large bending deformation is limited. Further, the behaviour of woven composites under transverse dynamic loading conditions is still to be fully exploited. Therefore, a further research work is needed to investigate and analyse these failure modes under large-deflection quasi-static and dynamic bending loads resulting in more rational, optimised and durable designs. Thus, the motivation of this study is to understand and quantify the mechanical behaviour and analyse damage mechanisms in woven composites under bending conditions using experimental material characterisation, microstructural damage evaluation and numerical simulations.

1.2 Aim and objectives

The aim of this study is to analyse the mechanical behaviour and damage modes in carbon fabric-reinforced polymer (CFRP) composites under bending loads using a combination of mechanical, microstructural and numerical studies. A robust experimental material data should be obtained to enable adequate numerical modelling of its deformation and damage behaviours observed in the microstructural examination of the composites. The obtained data will be used to develop models and simulation tools with the capability to investigate these behaviours under quasi-static and dynamic large-deflection loading.

In order to achieve the aim of this study, the following objectives are identified:

1. Characterisation of the mechanical behaviour of CFRP laminates in warp (0°), weft (90°) and in-plane shear (45°) directions under quasi-static tensile and flexural loading.
Chapter 1

2. Analysis of the material strength and energy absorption response in the warp and weft orientations under large-deflection dynamic bending conditions using Izod type impact tests.

3. Microstructural examination of the materials for damage identification using techniques such as optical microscopy and X-ray micro computed tomography (Micro-CT).

4. Developing 2D finite-element (FE) models based on plane-strain formulations to simulate the deformation behaviour and damage in CFRP laminates under quasi-static large-deflection bending.

5. Developing 3D multi-body dynamics FE models to study the deformation and damage in CFRP under dynamic bending conditions.

6. Formulating the interaction of various damage modes in the FE models using a novel discrete modelling approach of cohesive-zone method (CZM) rather than continuum damage mechanics (CDM).

1.3 Research methodology

The methodology adopted in this research is summarised schematically in Fig. 1.1. The thesis covers five main areas: introduction, literature review, experimentation, simulations, and conclusions and future work. Apart from introduction and conclusions and future work, each area is presented in more than one chapter; a brief description of the chapters will be given in the following section. The rest of this section is focused on the interaction between the elements of research methodology. It is comprised of two main parts: experimentation and simulations. Experimentation is carried out to characterise the material behaviour under quasi-static and dynamic conditions. Quasi-static tests are carried out to determine the material’s elastic constants and strengths using uniaxial tensile and bending tests on the material specimens in the warp (0°), and weft (90°) directions. Similarly, the in-plane shear properties of the material are determined by performing tensile tests on an off-axis ±45° laminate using a full-field strain-measurement digital image correlation (DIC) technique. The dynamic material behaviour is characterised by testing the material specimens in the same orientations using the Izod type pendulum hammer. All the specimens are tested under large-deflection dynamic bending conditions up to their ultimate fracture. Microstructural
examination of the virgin and tested materials is carried out using optical microscopy and X-ray Micro-CT techniques. The damage behaviour on surfaces of the tested specimens is observed by microscopy, whereas the Micro-CT provides the internal deformation and damage picture of the tested specimens. The experimental results are used in development and validation of numerical models.

In order to predict the deformation and damage behaviours of CFRP laminates, novel finite-element models are developed. At the beginning, the large-deflection bending behaviour of the on-axis laminates is studied by developing quasi-static
Chapter 1

2D FE models based on the plane-strain formulation. These models are then improved to incorporate multiple delaminations, fabric fracture and their interaction during the damage progression. The dynamic behaviour of the tested material is studied by developing 3D multi-body dynamics models. Here, too, cohesive-zone elements are defined at each ply interface as well as at the location of fabric fracture. Interaction of damage modes during loading is formulated by using a novel procedure based on cohesive-zone models (CZM) rather than traditional continuum damage mechanics (CDM) approach. The computational results are validated by comparison with experimental data.

1.4 Thesis outline

A brief description of the remaining chapters of this thesis is given below:

Chapter 2 Composites and their Mechanical Behaviour

A detailed introduction of composite materials especially fabric-reinforced laminates, their mechanical behaviour, damage mechanisms induced by various loads and the methods for the material's microstructural examination are presented in this chapter.

Chapter 3. Finite-element Modelling of Composites and their Damage

Modelling techniques based on the finite-element method for composite laminates and various damage modes are summarised in this chapter. A detailed description of the cohesive-zone modelling scheme for damage characterisation is also presented.

Chapter 4. Mechanical Behaviour of Fabric-reinforced Composites under Quasi-static Loading

The experimental methods used for acquiring the mechanical properties in quasi-static conditions are given in this chapter. These include uniaxial tension tests, in-plane shear tests and flexural tests. Furthermore, the experimental results and their discussion are also included. Methods to examine the microstructure and damage of the material such as optical microscopy and X-ray Micro-CT are also described.

Chapter 5. Mechanical Behaviour of Fabric-reinforced Composites under Large-deflection Dynamic Bending
This chapter describes the characterization of mechanical properties of woven CFRP laminates under large-deflection dynamic bending tests. Then, the experimental results of the dynamic tests and their discussions are provided. Damage analysis of the tested materials using Micro-CT technique is also presented.

Chapter 6. Finite-element Modelling of Damage and Fracture in Fabric-reinforced Composites under Quasi-static Bending

This chapter is divided into two parts. The first part highlights details of modelling and simulation of the material's deformation and damage under large-deflection quasi-static bending. The second part describes simulation of interaction of various damage modes observed in CFRP laminates during tests. The results are validated with experiments.

Chapter 7. Finite-element Modelling of Damage and Fracture in Fabric-reinforced Composites under Large-deflection Dynamic Bending

Details of multi-body dynamics FE models used to study deformation of, and damage, in the woven laminate under dynamic bending are given. Effects of the material's properties on the damage progression and results of simulations with their discussion are also presented in the chapter.

Chapter 8. Conclusions and Future Work

The outcomes and the conclusions drawn from the research are presented in this chapter. Suggestions and recommendations for the possible future research are introduced.
2 Chapter 2: Composites and their Mechanical Behaviour

2.1 Introduction

Fibre-reinforced composites have found a tremendous growth in industrial applications due to their outstanding mechanical properties, flexibility in design capabilities and ease of fabrication. As stated in the previous chapter, the aim of this study is to analyse the mechanical behaviour and damage modes of carbon fabric-reinforced polymer (CFRP) composites under bending loads using a combination of mechanical, microstructural and numerical studies. Before addressing this problem, it is necessary to provide an introduction to the laminated composite materials and their classification. This is followed by a review of the mechanical behaviour of woven composites under quasi-static and dynamic loading scenarios. A detailed account of various damage mechanisms ensued by static and dynamic loadings is presented. Microstructural damage characterisation techniques adopted in this study are also elaborated. Whilst an attempt is made to present an overview of these areas of research, the main focus is on the materials and methods used in the current work and justification of the employed methods. Finally, a summary of the key decisions made to formulate the research methodology, especially the experimental material characterisation, is given.

2.2 Composite materials

Quest for stronger, stiffer, durable, light-weight and tailor-made structures and components is ever increasing to meet various needs and necessities of human life. Composite structures have the potential to fulfil these requirements. A composite material is a combination of two or more materials, on macroscopic scale, with significantly different properties of its constituents that remain separate and distinct within the composite material. In many cases, composites are generally understood to mean a combination of high-strength but brittle fibres and a weaker but ductile matrix. The fibres are stronger and stiffer than the matrix, hence, defining the mechanical characteristics of the composite such as its strength and stiffness [2]. The engineering properties of composite materials are better than that of conventional materials such as metals. The properties that can
be improved by fabricating a composite material are stiffness, strength, weight, corrosion resistance, thermal properties, fatigue life and wear resistance.

Composite materials have a long history of application. Nature itself presents various forms of composite structures such as wood, cobweb and human body itself is a mixture of bones and tissues. In ancient times, man-made composites appeared in the form of mud bricks strengthened by straw; similarly medieval swords and armours were constructed of layers of different metals. Plywood was fabricated when it was realised that wood can be rearranged to achieve superior material properties. Recently, polymer-matrix composites that offer high strength-to-weight and stiffness-to-weight ratios have become important in the design of light-weight and strong structures such as aerospace, automotive, civil and sports products. Nowadays, in sports industry, application of polymer-matrix composites is evidenced in many sports activities and products, e.g. sports shoes, rackets, skis, fishing poles, surf boards, golf clubs, sails, roller skates etc.

There are three common types of composite materials [3]:

(a) particulate composites having particles in a matrix such as concrete, metal-matrix and ceramic-matrix composites;

(b) fibrous composites having fibres in a matrix; and

(c) laminated composites consisting of layers of various materials bonded together.

Presently, fibrous composites have found wide application in industry. The principal fibres used for engineering applications are glass, carbon (graphite), Kevlar and boron. The reinforcing fibres in the fibrous composites should have the following properties:

1. high modulus of elasticity;
2. high ultimate strength;
3. low variation of strength between individual fibres;
4. stability and retention of strength during handling and fabrication;
5. uniform fibre cross-section;

Fibres themselves are unable to withstand the loads in engineering applications unless bound by continuous medium called the ‘matrix’. The matrix is required to have the following characteristics:
1. support and bind the fibres together;
2. transfer the load to the fibres;
3. stop, to some extent, a crack propagation straight through a mass of fibres;
4. provide protection to the fibres from damage during handling and in-service, including environment;
5. chemically and thermally compatible with the fibres.

The most widely used matrix materials for fibrous composites are polymers called plastics. Hence, the term fibre-reinforced plastic/polymer (FRP) is used for such composites. A type of assembly of fibres to make forms for fabrication of composite materials may be uni-dimensional (UD) such as unidirectional tows or tapes; two-dimensional (2D) such as woven fabrics and three-dimensional (3D) such as fabrics with fibres oriented along many directions, e.g. 3D interlock weaves. Since this study is focused on laminates made of woven carbon fabric-reinforced plastic (CFRP), the next section describes the architecture of woven composites.

2.3 Woven fabric composites

Mechanics of woven structural composites can be best studied by taking into account their hierarchical organization. There are usually four important levels in a manufacturing process of woven composites:

FIBRE > YARN > FABRIC > COMPOSITE

A selection of fibres represents the first step in fabrication of woven composites. The fibres are grouped in a specific pattern and impregnated with resin to form yarns. The size of yarns is usually expressed as filament count which is the number of fibres (usually in thousands) in a single yarn. The yarns are then interlocked in specific patterns to form fabrics. Fabrics are classified as woven nonwoven, knitted, braided, 2D as well as 3D. 2D means that the woven fabric features only in-plane reinforcing properties [4]. Fabric preforms are stacked on top of each other in a specific stacking sequence to obtain thickness. The fabrics are impregnated with the matrix and cured forming a composite laminate. A laminate is thus a collection of laminae where each lamina is reinforced with a layer of fabric [4].
Chapter 2

The woven structure is characterized by the interlacing of two sets of yarns called the *warp* (0°) and *weft* (90°) *yarns* in a regular pattern or weave style. The fabric's integrity is maintained by the mechanical interlocking of the yarns. The weft tows run perpendicular to the direction of the warp tows. The woven fabrics are usually balanced and symmetric. A balanced fabric is one where the number and weight of fibres in yarns along the warp and weft directions are the same [2]. The material is therefore identical along these two directions. The warp and fill directions play equal roles affecting the thermomechanical properties of the composite laminate. An unbalanced fabric may be used to obtain different mechanical properties along the warp and weft directions. Weave patterns such as plain weave, twill weave, and satin weave are the common forms of woven architectures and are shown in Fig. 2.1. In plain weaves, each weft yarn goes over a warp yarn then under a warp yarn and so on. Due to the alternative interlacing, there is a high level of waviness or yarn crimp, which imparts relatively low mechanical properties such stiffness and strength of the composite compared to other weave styles [5]. Plain weaves are not very drapable. Twill weave, as shown in Fig. 2.1b, is formed by weaving one or more warp yarns over and under two or more weft yarns in a regular repeated manner. It is recognised by its characteristic diagonal ribs seen on the face. Hence, the yarn crimp is reduced by decreasing the number of exchanges. The resulting fabric has a smoother surface and slightly higher mechanical properties than plain weaves due to increased lengths of straight segments of yarn (float). Satin weave preforms are the modified twill weaves with few interactions of warp and weft tows. Satin weave has interlacing yarns arranged in such a way, that on one face of the fabric, most of the exposed yarns are weft, while on the
other face most of the exposed yarns are warp [4]. In a 5-harness satin weave, as shown in Fig. 2.1c, each weft yarn goes over 4 warp yarns before going under the fifth one. Here, low waviness of yarns results in good mechanical properties. The satin weave is asymmetric, where one side of the fabric is predominantly warp yarns and the other is weft. Exchange sites also break symmetry because they bend yarns in an asymmetric way. Bending and stretching in a satin weave ply are consequently coupled. There is also coupling between stretching and in-plane shear, because exchange locations are not symmetric about either in-plane axis (Fig. 2.1c). Coupling between bending and stretching tends to cause warping during cure because of thermal strains [5].

A selection of a weave for a specific application involves manufacturing considerations as well as final mechanical properties. The type of weave affects dimensional stability and conformability of the fabric over complex surfaces. Further, weaves with more yarn crimps such as plain weave offer low stiffness and strength. On the other hand, satin weaves have less crimp and good conformability, but their shear resistance is low due to the increased straight segments of yarns. Since the sports products are subjected to bending deformations causing shear damage of laminates in the form of delamination, thus, an optimum composite material is a polymer matrix reinforced with a symmetric balanced twill 2/2 woven fabric. The elastic properties of the fabrics can be estimated by considering these to consist of two UD plies crossing at $90^\circ$ angles with each other [2].

### 2.4 Mechanical behaviour of woven composites

Composite materials have many characteristics of mechanical behaviour that are different from those of conventional engineering materials. Unlike metals, composites are heterogeneous i.e. their properties depend on position in the material. Composites are not isotropic as the material properties are dependent on the orientation. They are called orthotropic when the material properties are different in three mutually perpendicular directions at a point in the material. The inherent anisotropy of composite materials leads to mechanical characteristics that are quite different from those of conventional isotropic materials [3]. This anisotropic nature of laminates needs determination of their mechanical parameters such as strength, stiffness and physical properties in each direction.
This aim is obtained through experimental characterisation of composites by determining the material behaviour and properties through tests conducted on suitable material specimens along different directions. Elastic constants and strengths are the basic mechanical properties of materials. The number of independent material constants such as stiffness and the Poisson’s ratio needed to describe their mechanical behaviour depends on the material’s planes of symmetry. For example, nine material properties are needed for an orthotropic material with three planes of symmetry and five materials constant are required for transversely isotropic (one plane of isotropy) laminate. Apart from the elastic material constants, the ultimate strengths in each direction should also be known to fully define the material’s failure envelope. The materials’ elastic constants and strength parameters are characterised through rigorous testing such as tensile, compressive, flexural and in-plane shear loading in various directions. Similarly, to study the damage and fracture behaviour of composites, their strengths and fracture toughness values should also be determined by means of mechanical tests such as double-cantilever beam (DCB), end-notched flexure (ENF) and mixed-mode bending tests. Some of mechanical tests such as tensile, flexural and in-plane shear are carried out to characterise the behaviour of woven composites, which are elaborated in Chapter 4. Since the present study deals with woven fabric-reinforced polymer composites, their mechanical behaviour is described in this section.

Woven-fabric structural composites offer a unique opportunity in that the microstructure of fibre preforms can be tailored to satisfy specific needs for mechanical performance as well as adequate structural integrity of composite structures [6]. The use of woven-fabric composites has many other advantages compared to unidirectional (UD) tape composites such as [7]:

- ease of handling for automation;
- ability to conform complex shapes and lower production costs;
- a possibility of applying a variety of processing techniques (e.g., hand lay-up, resin transfer moulding, resin film infusion, chemical vapour infiltration);
- improved impact resistance and damage tolerance;
- reduced notch sensitivity;
• improved out-of-plane mechanical properties (e.g., reduced delamination crack growth and higher peel strength);
• a possibility to choose among a large number of fabric architectures

Woven composites provide balanced in-plane mechanical properties in both orthogonal directions within a single ply. In woven laminates, thick interlaminar resin-rich regions are present, particularly at the inflection points of warp/weft yarns. These resin-rich pockets allow a significantly larger plastic yield zone to be developed ahead of the crack tip during delamination than that in non-woven UD laminates, thus increasing the interlaminar fracture toughness of woven composites [6]. These properties, especially a high work of fracture, have provided many opportunities for woven composites in applications demanding exceptional damage tolerance.

However, woven laminates exhibit a fibre breakage failure mode more commonly than UD laminates because fibres are interlaced with each other, resulting in a high degree of fibre crimp or waviness at cross-over points inducing stress concentrations. The stress concentrations may also result in early damage initiation in the form of fibre/matrix interface debonding or matrix cracking. Yarn waviness also promotes local fibre buckling under compression [6]. Besides, planar 2D woven laminates consist of continuous fibres that result in lower in-plane mechanical properties such as stiffness and in-plane strength due to fibre crimping [8]. Since the yarns are inevitably crimped along the whole length due to the woven nature of the fabric, this results in low in-plane shear resistance and reduced tensile and compressive strengths. These are some of the major disadvantages of woven fabrics over unidirectional or cross-ply laminates. Despite these deficiencies in the mechanical response, woven composites are the ideal candidate for sports applications subjected to quasi-static and dynamic bending deformations causing delamination damage. The dynamic behaviour of woven composites is described in the next section.

2.5 Dynamic behaviour of woven composites

Composite structures can be subjected to dynamic loads under service conditions. Aerospace structures, for example, can receive impacts during maintenance operations, or during service such as hailstones or other kind of debris. Sports
products experience dynamic loading during service, for example, sports trainers during running, tennis rackets and composite hockey sticks. A dynamic behaviour of composites is greatly influenced by the damage processes induced by impact loading. Thus, it is important to study the fracture process in woven composites to better understand their impact behaviour. Woven composites offer better resistance to impact damage such as delamination than UD laminates due to their interlacing (weaving) tow architecture limiting the damage growth between layers [9]. Furthermore, transverse tensile strength of woven composites is much higher than that of UD composites, which is one of possible reasons for superior impact-resistance characteristics of woven composites [10]. With a wide application of woven fabric-reinforced composites in structures and components subjected to dynamic loading scenarios, the understanding of their dynamic characteristics is critical to both the designers and end-users. This section highlights the types of impacts and impact testing for dynamic characterisation of composites.

Impact of two bodies occurs when their interaction occurs rapidly. One criterion is to compare the time of load application to the natural time period of the system. If the load application time is less than one-half of the natural time period, then an impact phenomenon occurs. If the load application time is more than three times the natural time period of the system, then the loading is quasi-static. In between there is a grey area [11]. Further, an important difference between static and impact loadings is that statically loaded components are designed to carry the load whereas components subjected to impacts are designed to absorb energy.

Generally, dynamic response of composite structures can be categorized into high-velocity, intermediate-velocity and low-velocity impact. In high-velocity impact (low mass), the impact event is very short and dominated by stress wave propagation through the thickness of the material. The structure does not have time to absorb substantial amounts of the impact energy, often leading to much localized damage in the form of penetration and perforation of the structure [12]. It is also called **ballistic** and **wave-controlled impact**, in which the effect of boundary conditions can be ignored because the impact event passes before the stress waves reach the boundary as shown in Fig. 2.2a. Here, deformation of the target structure is localized to the region around the impact point. High-velocity impacts usually range from 50 m/s to 100 m/s [12]. Ballistic impact is usually a result of
small arms fire or explosive warhead fragments. Intermediate-velocity impact events occur in the 10 m/s to 50 m/s range and are characteristic of both low and high velocity [12]. Intermediate-velocity impacts are usually caused by secondary blast debris, hurricane debris and runway debris or, for instance, a baseball striking a ball. The response to intermediate velocity impacts in Fig. 2.2b is also designated as wave-controlled impact. The gas gun apparatus, in which a small impactor is propelled at high speeds, used in testing ballistic impact is also used for intermediate-velocity testing. On the other hand, low-velocity (large mass) impact is associated with an impact event, which is long enough for the entire structure to respond to the impactor by absorbing energy elastically and, possibly, through initiation of damage. Deformation of the entire structure is established in low-velocity impact. Such an impact also known as *boundary-controlled impact* results from conditions arising from tool drops and athletic running. The impact duration is much longer than the time for the waves to reach the structure boundaries as shown in Fig. 2.2c. When the composite material is subjected to impact, it undergoes a large amount of strains depending on the magnitude of impact, temperature range and strain rates. The behaviour of composites at high strain rates is usually characterised with split Hopkinson bar pressure tests for the intended applications. Since the focus of this study is to investigate the composites’ behaviour in low-velocity dynamic event, described in the next section.

Fig. 2.2. Classification of impacts: (a) high-velocity impact, very short impact times with dilatational wave dominated response; (b) intermediate-velocity impact, short impact times with flexural and shear wave dominated response; (c) low-velocity impact, long impact times with quasi-static response [13]

2.5.1 Low-velocity impact

Response to low-velocity impacts can be treated as quasi-static because the deflection and load would have similar relation as in a static loading. The upper
velocity limit of these events can vary from 1 to 10 m/s, depending on the target’s stiffness, material properties and the impactor’s mass and stiffness. When the impact velocities are below 5 m/s, the response type is controlled by the impactor/target (laminate) mass ratio rather than the impact velocity [13]. In low-velocity impact, the dynamic structural response of the target is of utmost importance as the contact duration is long enough for the entire structure to respond to the impact and, in consequence, more energy is absorbed elastically and eventually in internal damage formation in composites. This non-visible impact damage is usually termed as barely visible impact damage (BVID). Cantwell and Morton [14] has classified low velocity as up to 10 m/s, by considering the test techniques, which are generally employed in simulating the impact events such as Charpy, Izod and drop-weight impact tests. A brief account of these low-velocity impact tests and their procedures is given in the following section.

2.5.2 Impact testing

Low-velocity impact testing of composite laminates can be performed using several types of equipment arrangements. However, the impact test fixture should be designed to replicate the loading conditions, to which a composite structure is subjected in service conditions, and then reproduce the failure modes and mechanisms likely to occur [14]. In this section the most commonly employed techniques and methodology is presented.

(a) Drop-weight impact testing

In this set-up, a known weight is allowed to drop from a pre-defined height to strike a test specimen placed on rigid supports in the horizontal plane. Here, usually a flat, rectangular composite plate is subjected to an out-of-plane, concentrated impact with a hemispherical impactor. The desired level of impact energy can be varied by changing the drop height or adding weights to the tup holder. Frequently, the impactor is instrumented, enabling the force-time history to be determined from the point of initial contact with the specimen and as the striker traverses through the thickness of the specimen. Energy is calculated from the force-time signal, and similarly, load-displacement parameters are also recorded. This test procedure is now increasingly used for studying the impact behaviour of composites. Damage resistance of the material is quantified in terms of the resulting size and type of
damage in the specimen. One of the advantages of this test is that a wider range of test geometries can be tested along with various shapes of the impactor. Drop-weight tests are used to simulate low-velocity impacts, typical of the tool-drop problem. These tests are usually performed according to ASTM D 7136 standard. In general, the impact event does not cause complete destruction of the test specimen; it rebounds, enabling a residual energy to be determined if required [14].

(b) Charpy impact testing

A very common way to evaluate impact properties is to determine the material toughness by measuring the energy required to break specimen of a particular geometry. The well-known Charpy and Izod tests that were developed for isotropic materials are used for this purpose. Many of the early impact studies on composite materials were undertaken using the Charpy test method [14]. In this test, a beam specimen is simply supported at the two ends and struck by a swinging pendulum. The test specimen is generally a thick beam, sometimes incorporating a notch at its mid-point as shown in Fig. 2.3a. The energy dissipated during impact usually is essentially the energy required for propagation of the material’s fracture. According to [14], the Charpy test is only suitable for ranking the impact performance of composites and as a first step in determining the dynamic toughness of these materials.

(c) Izod impact testing

In the Izod test, the specimen is fixed at one end in the vertical plane as a cantilever beam and impacted by a swinging pendulum at its free end. The specimen may be notched or unnotched. A notched specimen is used to determine the fracture toughness of the material whereas an unnotched specimen may be used for large-deflection dynamic bending tests. The Izod impact test for notched specimen is shown schematically in Fig. 2.3b. Izod tests for material’s fracture strength are performed on a notched specimen according to ASTM D256 and dynamic bending tests on an unnotched one are conducted according to ASTM D4812. The measuring method is based on determining the amount of energy needed to break a notched specimen under specific conditions, such as specimen location, notch shape and speed of the hammer. Although the notched specimen Izod test may not be adequate to represent a realistic impact condition of composite specimens, in which the fracture process is more complex [12], yet
the unnotched specimen Izod test can replicate a dynamic behaviour of composite structures subjected to cantilever-type loading. Therefore, the Izod tests with the unnotched specimens are conducted in this study to investigate the large-deflection dynamic behaviour of woven composites.

![Diagram of impact test](image)

Fig. 2.3. (a) Chapy impact test; (b) Izod impact test [14]

### 2.6 Damage of composite laminates

Unidirectional and woven-fabric composites are used to manufacture tailored laminated composite structures by deploying fibres in several layers, and these layers are stacked in angled orientations to achieve high stiffness and strength in the desired directions. Thus, composite laminates show higher in-plane strength and stiffness in the fibre direction than in the transverse direction. An arbitrary laminate subjected to an in-plane tensile load experiences not only extension along the load application axis and lateral contraction, but also bending, twisting, and in-plane shear deformation. Thus, balanced and symmetric laminates are usually designed to minimise these coupling effects. Despite their desirable mechanical characteristics such as high stiffness- and strength-to-weight ratios, good fatigue and corrosion resistance, laminated composites are rather fragile and susceptible to damage during service.
Fibre-reinforced composite structures must be capable to perform their intended design functions during their service life while being subjected to a series of events such as structural as well as environmental loading scenarios. Such events can cause structural degradation which subsequently can affect the ability of structures to perform their functions. Investigation of stiffness and strength degradation, initiation and growth of damages, and maximum load that a structure can withstand before failure is necessary for performance assessment of composite laminated structures and for their safe and reliable design. The structural degradation phenomenon in fibre-reinforced composite components is quite different from that of their metallic counterparts because the failure process is not uniquely defined in composite materials. Laminated composites develop multiple matrix and intralaminar cracks, fibre-matrix interface debonding, local delamination distributed in an interlaminar plane subjected to service loads. These defects degrade the material’s macroscopic strength and stiffness. Such a process of structural deterioration of material, which results from the creation, growth and coalescence of microscopic defects is called damage [15]. Damage is a collective reference to irreversible distributed changes brought about by energy-dissipating mechanisms such as breakage of atomic bonds. Examples of damage are multiple fibre-bridged matrix cracking, multiple intralaminar cracking in a laminate, local delamination distributed in an interlaminar plane, and fibre/matrix interfacial slip associated with multiple matrix cracking. The field of damage mechanics is concerned with conditions for initiation and progression of distributed changes as well as consequences of those changes for the response of a structural material subjected to external loading [16]. Subsequently, the degraded structure’s material experiences fracture under the applied service load making the structure incapable to perform its design function. Thus, the knowledge and analysis of damage and failure mechanisms of composite laminates play an important role in instructing safe, reliable and practical design of composite structures. Hence, damage of composite laminates is a complex phenomenon, and various mechanisms leading towards the catastrophic failure of composite structures need to be understood. The next section highlights various damage mechanisms encountered by laminated woven composites under service conditions.
2.7 Damage mechanisms in woven composite laminates

Composite laminated structures subjected to various static and dynamic loads develop complex damage mechanisms due to their extreme level of anisotropy and heterogeneity, which is rarely observed in their homogeneous metallic counterparts. Manufacture of fabric-reinforced composites and the resulting architecture of fibre reinforcements are quite different from those of UD composites. The manufacturing process of tows’ interlacing results in a complex 3D internal architecture of the fabric. Even if the fabric reinforcement features 2D in-plane reinforcing properties, it has a 3D internal architecture due to waviness induced by the weaving process. This 3D architecture results in complex stress and strain fields in the composite material, triggering new damage mechanisms and failure modes not seen in traditional unidirectional laminates. For example, specific failure mechanisms encountered in fabric-reinforced composites include: intra-ply delamination (i.e., delamination between overlapping tows of the same woven ply), transverse normal and shear failure modes under axial loading, and failure of pure matrix regions [7]. In textile composites, damage begins at micro-scale with matrix cracking, fibre-matrix debonding and fibre failure within the ply. This is followed by meso-scale damage such as intra-yarn cracking and inter-ply delaminations. On the macro-scale, composite failure is characterised by a strong interaction of intra-ply cracking and inter-ply delamination and ultimate fabric rupture [17]. As a consequence, the internal 3D structure of reinforcement translates into degradation of stiffness and strength properties of the composite material leading to a loss of structural integrity. Therefore, it is necessary to understand initiation and evolution of these damage mechanisms during laminate failure process. Hence, a short description of these damage mechanisms is given in this section.

2.7.1 Matrix cracking

Matrix cracks are the most common damage mode because the strength and stiffness of a matrix is considerably lower than that of the reinforcing fibres. Matrix cracks also referred as transverse cracks, intralaminar cracks and ply cracks appear first in the layers transverse to the loading direction, traversing through the ply thickness and running parallel to the fibres in that ply. Matrix cracks are an intralaminar form of damage, and involve cracks or voids between fibres within a
single composite lamina. The initiation and growth of matrix cracks is dependent on the loading, scheme and composite’s lay-up, and usually a single matrix crack may develop a series of cracks within a lamina at a characteristic spacing. In low-velocity impacts, matrix cracks in upper layers of composite laminate initiate at the contact edges of an impactor. An example of a matrix crack in the weft yarn of 5-harness satin woven laminate subjected to tension is shown in Fig. 2.4a. Initiation and development of transverse cracks and their effect on structural integrity and durability of laminated composites were extensively studied by many researchers [18]. A study by Parvizi et al. [19] is among the earliest researches; they carried out extensive experiments to observe transverse cracks. They observed that transverse-crack spacing showed a decreasing trend with increasing applied stress and an increasing trend with the increase in transverse ply thickness. They also found that cracks formed in a direction parallel to transverse reinforcement and thickness of the transverse layer influence initiation and propagation of the matrix cracking process. Silberschmidt [20] presented that a random spatial distribution of fibres results in variations of local properties of composite laminates. This non-uniformity not only affects the effective properties of composite materials but is also a crucial factor in initiation and evolution of damage and fracture processes that are also spatially random. Such randomness in microstructure and in damage evolution is responsible for non-uniform distributions of stresses in composite specimens even under externally uniform loading, resulting in a random distribution of matrix cracks in cross-ply laminates. Crack density, which is the number of cracks per unit length, increases abruptly with the applied load after initiation of cracking, until cracking comes to saturation state called a characteristic damage state. Matrix cracking gradually reduces stiffness and strength of the laminate and changes its coefficient of thermal expansion, moisture absorption and structure’s natural frequency [21].

2.7.2 Fibre-matrix debonding

The next intralaminar micro-level damage mode is fibre-matrix deboning, where a debond, parallel to the fibres’ direction, separates the constituents from each other; thus, the matrix support to the fibres in that region is eliminated. An interface between fibres and matrix resin plays a significant role in load transfer between fibres and matrix, with interface adhesive properties controlling the composite’s
performance. Debonding occurs at a weak interface between fibre and matrix. This damage mode initiates at the constituent level and gradually evolves into a macro level as the applied load increases, with all the layers in the laminate readjusting the load they carry in order to properly redistribute the stress released by damage. Although matrix cracking and fibre-matrix debonding do not cause ultimate structural failure of laminated composites, they can result in significant degradation of material’s stiffness and can also trigger more severe forms of damage such as delamination and fibre breakage [22]. Fibre-matrix debonding in the weft tow of a woven laminate under tensile load is shown in Fig. 2.4b.

![Image](image.png)

(a) (b)

Fig. 2.4. Damage mechanisms in woven laminate: (a) transverse matrix cracking; (b) fibre-matrix debonding in weft yarn under tension [17]

Interfacial sliding or fibre pull-out can occur when different displacement fields are imposed on different constituents [1]. One example of this is the case of thermal loading; due to different coefficients of thermal expansion, the fibres and the matrix will be under different loading conditions. Interfacial strength plays a very important role also in this case for the existence of this kind of damage mechanism. Sliding of fibres against the matrix can cause further damage due to frictional wear. Fibre pull-out can also occur when brittle or discontinuous fibres are embedded in a tough matrix.
2.7.3 Inter-ply and Intra-ply delamination

As discussed earlier, with the onset of damage, a layer weakens, and the adjacent layers take on additional loads and, subsequently, undergo damage initiation and evolution processes. First matrix cracks are formed randomly, and then they coalesce and lead to delaminations at interfaces between layers. Delaminations are separations between internal layers of a composite laminate due to the lack of reinforcement in the thickness direction. These delaminations are caused by high through-thickness shear and normal stresses, a mismatch in the Poisson’s ratio between the layers and a presence of geometric discontinuity between layers of laminated composites. Similarly edge delaminations initiate at load–free edges of the composite laminate due to a mismatch in the Poisson’s ratio and transverse shear stresses close to these edges. In the vicinity of free edges, stresses are three-axial consisting of in-plane and out-of-plane stresses; hence, the classical laminate theory (CLT) based on a plane-stress assumption becomes invalid. These out-of-plane stresses (also called interlaminar stresses) depend on a stacking sequence of plies and are primarily due to the mismatch of engineering properties such as the Poisson’s ratios and shear-coupling coefficients between adjacent plies. These interlaminar stresses produce matrix cracks at the free edges, from which they propagate into the laminate and initiate delaminations, leading to stiffness degradation, strength loss and failure of laminated structures [22]. Delamination can also occur near any stress risers in the laminate such as holes, cut-outs etc. in addition to free edge effects. Low-velocity impact loading is also a major cause of delamination failure in fibre-reinforced composites, resulting in reduction of compressive residual strength of laminates. In transverse dynamic loading, interlaminar shear and bending stresses are the major causes of delamination [12]. Once interlaminar damage is initiated, its propagation and residual strength of the structure depend on the toughness of interlaminar layers and the level of energy that is required to propagate the crack. In woven laminates, delamination may be interlaminar such as separation between two plies as well as intralaminar such as debonding between tows within a single ply as shown in Fig. 2.5. Since a matrix forms the weakest link in a laminate’s through-thickness direction which is responsible for delamination cracking. In general,
brittle resins favour delamination while tougher resins tend to induce kink-band formation and overall shear failure.

Fig. 2.5. Damage mechanisms such as delamination between load-aligned and transverse tows, cracking of transverse tows and kinking of load-aligned tows in a 5- harness satin weave CFRP laminate under compression [23].

Delamination has a more detrimental effect on structural integrity of composite laminates than transverse cracking. Delamination can render a structure incapable of load-carrying and trigger the structural failure if it grows in size under an increasing load, particularly if the laminate is subjected to compression. On the other hand, this is not generally the case with transverse cracking where neighbouring plies can constrain the effect of this damage and maintain residual load-carrying capability of the structure [24]. Delamination can be reduced by either improving the fracture toughness of the material or increasing interlaminar strength of the laminate by modifying the fibre orientation and ply stacking sequence as well as Z-pins stitching.

The problem of delamination has also been investigated widely, and many works have been published addressing this failure mode. The first analytical model to predict the energy release rate associated with the growth of delamination induced
by a transverse crack was developed by O'Brien [25]. Wang et al. [26] used a three-dimensional finite-element analysis to evaluate the energy released with the delamination growth, by considering delamination induced by transverse cracks and free edges of the laminate. Nairn and Hu [27] investigated the initiation and growth of crack-induced delaminations based on variational approach [28, 29] for transverse cracking in cross-ply laminates. The variational-mechanics analysis predicted that transverse cracking developed until the crack density reached some critical density for delamination. A substantial amount of research was carried out in the investigation of delaminations induced by various mechanisms within composite laminates and presented in review papers. Garg [30] reviewed the state of the art of delamination behaviour since 1970s to 1980s discussing some aspects such as causes of delamination and its effect on structural performance, analytical and experimental techniques to predict its behaviour and some of the preventive measures to delay the delamination so as to make the structure more damage-tolerant. Pagano and Schoeppner [31] conducted critical reviews of many selected papers, especially the pioneering works on delamination research. Tay [32] reviewed major developments in the analysis and characterization of buckling-driven delamination from 1990 to 2001. Brunner et al. [33] reviewed the developments leading towards new standardized test procedures for determination of delamination resistance or fracture toughness of fibre-reinforced polymer–matrix composites.

2.7.4 Fibre/tow fracture

The initiation of fibre fracture is often considered as the ultimate failure mode of the laminate, as a drastic reduction in the load-carrying capacity occurs at this stage because fibres act as the principal load-bearing constituent, and resist most types of the applied loads. As the applied load is increased, progressive matrix cracks lead to fibre-matrix debonding and delamination resulting in a complex stress state. Now, as the matrix is debonded and shattered, the only load-carrying members are the fibres, which start to fail when the laminate’s strain reaches the fibre’s fracture strain resulting in multiple fibre cracks. In low-velocity impact events, fibre failure occurs just below the impactor due to high local stresses and indentation effects; and on non-impacted face due to high bending tensile stresses [12]. Accumulation of individual fibre fractures within tows and, subsequently, plies
leads to ultimate laminate’s failure when there are not enough fibres remaining intact to carry the required load. Damage progression becomes catastrophic and the material’s ultimate strength is achieved.

2.7.5 Fibre/tow kinking

Fibre failure in compression occurs due to micro-buckling and formation of kink bands, when a unidirectional or orthogonal woven composite laminate is subjected to a compressive load [34]. When an on-axis composite laminate is loaded in longitudinal compression, there is a phenomenon of local instability caused by failure of the matrix supporting fibres leading to formation of a kink band. Hence, fibres start buckling due to the lack of lateral support. As the applied load is increased, the fibres/tows reach a critical buckling load and initial fibre/tow fracture occurs in the compression side. Finally, the kink band is fully formed when fibres/tows fail at the top side of the kink band as shown in Fig. 2.5. The critical buckling load is thus a function of the properties of fibres/tows, matrix and their adhesion.

The performance and development of the above mentioned damage mechanisms leading to failure of any composite structure depend on a range of parameters including the geometry, material, lay-up, loading conditions, load history and failure modes. Each damage mechanism has a different governing length scale and evolves differently when the applied load is increased. The damage scenario becomes more complicated when there is interaction between individual mechanisms. As the loading increases, load transfer takes place from high damage regions in the laminate to those with low damage, and the composite failure results from the criticality of the last load bearing element. These damage parameters are highlighted in a recent experimental study by De Carvalho et al. [23] of two different woven composites - twill and satin weaves - subjected to compression. They concluded that tows behaved as structural elements at the reinforcement level, damage morphology was affected by the weave architecture and geometry, and tows tended to fail at the crimp region. It was found that kink-band formation, matrix cracking and transverse tow cracking were the predominant damage propagation mechanisms in compression. Similarly, damage observed in woven laminates as a consequence of tensile loading was in the form of transverse matrix cracking and delamination at the crimp regions [35]. However,
studies of damage mechanisms induced by large-deflection bending in woven composites are very limited. Thus, variability of woven composites due to their reinforcement architecture promotes interaction between different micro-mechanical damage mechanisms, increasing the difficulty to study their failure. Also, as these damage mechanisms are often embedded within the plies of a composite laminate, they may easily escape detection. Therefore, an insightful experimental investigation is fundamental to support the development of tools capable to predict and model effectively failure of woven laminates. The focus of the next section is on damage characterisation techniques for composites at microstructural level.

2.8 Microstructural damage examination

As described in the previous section, damage and defects in composite materials affect their performance and structural integrity. It is of paramount importance to identify and investigate various damage mechanisms in composite structures for their safe design, analysis and performance. Microstructural damage mechanisms such as matrix cracking, delamination and fibre breaking induced by static and dynamic loading are barely visible and cannot be detected by simply examining the exposed surfaces of composite’s specimen. Many destructive and non-destructive techniques have been developed to determine the type, location and extent of these damage modes. The traditional damage assessment methods are visual inspection, optical or scanning electron microscopy, thermography and ultrasound. Although non-destructive techniques such as ultrasonic C-scan [36], X-ray radiography and thermography [37] are efficient for providing detailed information of damage area, they are limited in their resolution and the ability to track the interaction of various damage modes. Destructive techniques such as de-ply technique and serial sectioning of specimens followed by microscopy can provide very detailed characterisation of damage [38]. However, specimen preparation is time consuming, and there are doubts of introducing new damage to the specimens, affecting any residual stresses within the specimens during cutting. Further, all these techniques provide damage information of materials in 2D cross-sections, which are limited in characterising the inherent 3D internal damage processes within the composite laminate. This shortcoming can be overcome with the use of X-ray Micro-CT, which can provide 3D images of internal deformation.
mechanisms and their interaction in the composite with high resolution. In this study, optical microscopy and X-ray Micro-CT are employed for microstructural characterisation of woven composites; a brief account of these two techniques is given below.

2.8.1 Optical microscopy

Microscopic analysis of composite specimens is a suitable technique for evaluation of damage in woven-fabric laminates. Microscopic analysis allows to visualize the type and location of damage, to study the comprehensive damage behaviour of the composite specimen. Detailed maps of damage can be obtained by sectioning several strips of material at different locations and orientations throughout the specimen. After careful preparation, microscopic examinations of each section are used to construct detailed maps of various damage features observed on the laminate’s surfaces. Daggumati et al. [39] employed this technique for measuring the weft yarn dimensions and damage characterisation of 5-harness satin woven CFRP laminates under tension. Similarly Lomov et al. [40] studied the onset and progression of various damage mechanisms in woven and braided composites under tensile loads using optical microscopy along with x-ray and acoustic emission. However, the technique is time-consuming in sample preparation and new damage feature may be introduced in the specimen during cutting and polishing.

2.8.2 X-ray micro computed tomography

High resolution X-ray micro computed tomography (Micro-CT), initially developed for medical applications, is gaining a considerable interest as a powerful tool for studying internal deformation and mechanics of composite materials. Micro-CT is a non-destructive technique, which allows for continuous monitoring of external and internal damage and analysis of its characteristics in a 3D space at micron or even sub-micron level. Besides the principle virtues of being non-destructive and non-invasive, Micro-CT allows a large amount of the internal geometry to be determined with a single scan, enabling internal cross-sections to be visualised. Its application to study the composites has started only recently. The technique has been successfully employed to analyse microstructure and various damage modes and their interaction in woven composites. For example, Badel et al. [41] used this method to obtain experimental undeformed and deformed 3D geometries of the
textile reinforcements for development of a 3D FE model. Awaja et al. [42] employed the technique to examine the internal structural deformation and damage mechanisms in glass and carbon fabric-reinforced composites. Enfedaque et al. [43] studied deformation and fracture micromechanisms in woven carbon and carbon-glass hybrid composites under low-velocity impacts using X-ray microtomography. Seltzer et al. [44] studied the damage micromechanisms in 3D woven carbon, glass fibre and hybrid composites under low-velocity impact and compared their behaviour with 2D woven laminates. Scott et al. [45] used high resolution synchrotron radiation computed tomography to capture fibre damage progression in a carbon–epoxy notched [90/0]s laminate loaded to failure. They carried out direct in-situ measurement of accumulation of fibre fractures for a high performance material under structurally relevant load conditions. Similarly, the authors of [46] identified various damage mechanisms such as matrix cracking, delamination and tow debonding in woven glass composites under large-deflection bending.

X-ray micro computed tomography (Micro-CT) generates 3D images of materials through the process of collecting a series of 2D X-ray radiographs taken at incremental rotations of the object. Following acquisition, a software program builds a precise 3D volume of the sample from 2D radiographs by 'stacking' the individual slices one on top of the other; this process is known as reconstruction. In this technique, a contrast is based on differences in X-ray attenuation of materials, which is proportional to the sample’s density. Thus, it is a useful tool to identify in-homogeneities, voids, fractures, microcracks, and porous structures in polymer composites where there is a significant variation in density [42].

However, one key problem with Micro-CT is that to acquire high resolution images, the size and geometry of the object becomes a limiting factor. This is because the field of view of the detector limits the spatial dimensions of the volume, affecting the size of the sample that can be scanned. Hence, as the resolution is increased, the field of view of the sample is reduced. However, samples must remain within the field of view to obtain radiographs of the region of interest (ROI); thus, a trade-off is needed. Therefore, large real-size specimens are scanned at low resolution losing the detailed microstructural damage information. One way to carry out scanning large objects at high resolutions is to cut ROIs from the specimen into
more conveniently shaped coupons, in most cases ‘matchsticks’ with square cross-sections [47]. This makes the technique a destructive method that is potentially not favourable to damage assessments. Another limitation is a long time needed for one scan to obtain the full high resolution. In this study, small size samples are scanned to get the detailed damage information inside the tested CFRP specimens.

2.9 Summary

Fabric-reinforced polymer composites have found a wide variety of applications in different areas. Their mechanical behaviour and the ensued damage mechanisms under static and dynamic in-service loading scenarios were presented and discussed in detail. The microstructural damage characterisation techniques for damage assessment in the composites were also reviewed. However, majority of the studies are related to damage caused by tensile loading in quasi-static conditions, whereas in dynamic conditions, the studies are mostly for drop-weight impact tests. Thus, the experimental studies of the behaviour of woven laminates under large-deflection bending in static and dynamic conditions are still very limited. Therefore, this work is focused on understanding of various damage mechanisms of woven composites under large-deflection bending. In order to comprehend the complex process of deformation and damage of woven composites linked to their architecture, it is important to characterise the material behaviour through experimental tests and microstructural damage evaluation techniques. The next chapter highlights the numerical methods used to model woven composite laminates and various types of damage modes under different loading scenarios.
Chapter 3: Finite-element Modelling of Composites and their Damage

3.1 Introduction

Numerical methods have been employed increasingly thanks to a growing computational power for the analysis of composite laminates and evaluation of various damage modes in them. Most of the approaches have tried to analyse composites within the framework of a 2D boundary value problem assuming plane-stress or generalized plane-strain conditions or full 3D stress analysis using the finite-element method (FEM), finite-difference method, boundary-element method, and finite-strip method. The benefit of these approaches is that they provide more accurate solutions for fields of stresses and displacements in damaged laminates at macro and micro scale levels. These computational tools have also been used to verify, compare or validate existing or proposed analytical and experimental methods for damage initiation and progression. Since damage in composites is a complex phenomenon, in recent years, the finite-element method has attracted much attention to model this damage behaviour at various scale levels.

An accurate prediction of composite fracture often depends on modelling a progressive development of all modes of damage, such as matrix cracking and delamination, as well as their interactions. The combined effect of various damage modes acting concurrently on complex shaped composite structures subjected to arbitrary loading conditions cannot be handled properly with analytical techniques. The problem becomes even more complicated if one has to account for stresses induced due to processing, thermal loading and moisture absorption. Therefore, one has to resort to numerical techniques such as the finite-element method to predict the damage behaviour of composite structures accurately. With advances in computing resources, damage modelling at various length scales and its effect on the macroscopic failure of laminated structures has become the focus of research. In this chapter, methodologies for modelling composite laminates and investigation of damage in these materials based on the finite-element method are presented.
3.2 Finite-element method

The finite-element method (FEM) is an approximate numerical technique for solution of continuum-mechanics problems. The continuum is discretized into a finite number of parts called elements connected at their common points called nodes. The finite-element discretization transforms a continuous boundary-value problem into an algebraic system of equations for discrete nodal variables of a given finite-element mesh. Variational principles are used to transform the governing differential equations of the problem under study to a weak form (integral form). In numerical methods, there are generally two approaches of formulating kinematics of continuum mechanics - Lagrangian or the material description of behaviour, and Eulerian or the spatial behaviour. In the Lagrangian approach, the material is associated with an element throughout the entire analysis, and the material cannot flow across element boundaries. In the Eulerian approach, elements are fixed in space and the material flows through them [48]. Thus, in the Lagrangian mesh, the mass in each element is constant while its volume varies; in the Eulerian, the mass can vary in each element, but not its volume. The Lagrangian formulation is preferable in simulations, where boundary conditions have to be fixed, contacts defined between the solids, or if an analysis is to be made of solids formed of several layers of material as in composite laminates. The Lagrangian mesh allows the history of the material to be followed which is particularly useful for the study of damage behaviour of composite materials [12]. Although the Lagrangian approach suffers from large-deflection of structures, when the mesh is highly distorted, still, the mesh control techniques available in the FE codes, e.g. Abaqus, resolve this issue. Eulerian methods are used commonly in fluid mechanics simulations. In this study, continuum elements based on the Lagrangian formulation are used for the analysis of damage in woven composites.

The finite-element method has found an increasing application in the solution of complicated engineering problems. The most attractive feature of the FEM is its versatility and ability to handle complicated geometries under various loading scenarios with relative ease [49]. It is powerful tool for solving problems having no exact analytical solutions. Although it is an approximate solution method, the approximation can easily be improved by refining the mesh at regions where field
gradients are high or if edge effects are to be included in the analysis. This local mesh refinement is known as \textit{h-refinement}. The solution accuracy can also be increased by using elements with higher-order shape functions known as \textit{p-refinement}. Combination of both these refinements is most efficient for simulating complex structures with irregular shapes and boundary conditions. It should be noted that the choice of element and mesh is problem dependent. When solving a specific problem using FEM, the accuracy of mesh refinement can be checked by carrying out a series of runs with different mesh densities or different element types and checking the results for convergence. On the other hand, modelling, discretisation and numerical errors affect greatly the solution of the problem.

Most real world composite structures do not admit exact solutions, requiring one to find approximate but representative solutions [50]. FEM is an effective approximate method for predicting the response of composite laminates. Application of FEM to composites requires specific element formulations that adequately represent their orthotropic behaviour, stiffness and strength, as well as the lamination of plies often used. Unlike isotropic materials, composites exhibit complicated mechanical behaviour requiring knowledge of anisotropic elasticity, lamination theories, and failure and damage criteria. This complex behaviour of composites results in need of in-depth research for their design and application. The next section highlights various approaches for the analysis of composite laminates using FEM.

3.3 FE modelling of composite laminates

Fibre-reinforced composites are manufactured in the form of thin layers bonded together to form a laminate with desired geometry and material properties. The properties of a laminate are very much dependent on properties of individual plies and their corresponding angles. Thickness of these laminates is usually small compared with their other dimensions so that it forms a plate type structure. Therefore, two-dimensional theories are used to analyse composite laminates for stresses and subsequent failure. Zhang and Yang [51] classified the laminated plate theories into the two categories: equivalent single layer shell theories and a continuum-based 3D elasticity theory. Computational modelling of composites in the commercial FE softwares is usually based on elements formulated according
to these theories. A short description of these modelling approaches is given below.

3.3.1 Equivalent single layer shell theories

Formulation of shell finite-elements is usually based on a classical lamination theory (CLT), a first-order shear deformation theory (FSDT) (also called Mindlin plate theory) and higher-order shear deformation theory (HSDT). The two-dimensional classical lamination theory (CLT) derived from the three-dimensional elasticity theory is found to be adequate for most applications where the thickness of a laminate is small and shear deformation effects are negligible [50]. This theory is based on assumptions that the in-plane displacements vary linearly through the thickness and the transverse displacement is assumed to be constant through the thickness. Stiffness coefficients of the laminate are derived from stiffnesses of individual plies using CLT. Usually, stress-strain relations for a thin lamina of orthotropic material are written down, and subsequent transformation and integration procedures yield stiffness moduli for the whole laminate. Determination of laminate’s stiffness moduli has been made easy by the application of FEM. Since, the material properties vary from ply to ply, the stress variation through thickness of the composite laminate will be discontinuous. That is why a laminate theory is used instead of simple material stiffness for a laminated material [52]. Further, the theory ignores edges whereas real-life plates have edges subjected to stresses. Therefore, CLT should not be used for composites that are likely to fail in transverse shear and delamination. In a commercial FE package Abaqus, formulation of thin conventional shell elements is based on CLT (Kirchhoff) theory [48]. The conventional shell is a planar 2D representation of a solid element, even if deformable in a 3D space. Thickness is attributed to a planar element by assigning a section or a composite layup. Since, the geometry is defined in the two-dimensional space, thus an element cannot be assigned to each ply of the composite. A continuum shell element is also available for modelling thick solid parts in Abaqus; however, its kinematic and constitutive behaviour is similar to conventional shell. Continuum shell also has a single element through its thickness, which contains multiple plies defined in the layup [48]. The first-order shear deformation theory (FSDT) or Mindlin shell theory is used for thin and moderately thick laminated plates, providing a balance between
computational efficiency and accuracy for the structural response of the laminated composites. But this theory does not predict local effects such as interlaminar stress distribution between layers and delamination. To overcome these deficiencies, higher-order shear deformation (HSDT) theories have been developed, that satisfy free boundary conditions of the transverse shear stresses on the upper and lower ply surfaces. Interlaminar stresses can be predicted accurately but at the expense of computational cost as the number of unknowns depends on the number of plies of the laminate. Thus, 3D representation of each ply in the FE model is necessary.

3.3.2 Continuum-based 3D elasticity theory

Composite laminates are typically modelled using elements based on shell theories, which are limited in modelling their through-thickness behaviour. Through-thickness bonding between the plies is provided by a weak matrix, which is susceptible to delamination caused by high transverse shear stresses. The solid elements formulated on 3D continuum-based elasticity theory have a capability to predict delamination and interlaminar shear as well as normal stresses in a composite laminate. In such layer-wise modelling, each separate layer of the laminate is explicitly represented by at least one continuum element with its own degrees of freedom (DOFs). The 3D continuum-based modelling approach relies classically on the use of solid brick elements that are stacked on each other so as to form the whole laminate. In Abaqus, if the FE model contains multiple continuum solid elements through the thickness of a region, correct results can be obtained by sectioning the solid region of the composite for each ply and defining a separate composite layup for each layer of elements [48]. Since, this description introduces a number of DOF that depends on the number of layers constituting the laminate; the computational cost for 3D models becomes higher.

The fibre direction in a ply and the stacking sequence of plies forming a composite laminate have a significant effect on its response [52]. While defining a composite laminate in a FE package, the fibre direction, number of plies and their stacking sequence are specified in the pre-processor. Similarly, the orthotropic material properties of each lamina are also defined in the material model. Most composite laminates are manufactured by stacking sheets of material with the fibres directions changing from ply to ply. In this type of structure, all the fibres are at
known fixed angles to each other, and the direction of the reference ply is defined to specify the alignment in the lay-up. The fibre direction is controlled by the angle with respect to the reference ply and the global coordinate system. By incorporating appropriate boundary and loading conditions, the structural problem is solved as in the case of isotropic material model. In post processing, the obtained results are viewed and manipulated at ply level, and each ply failure is thus investigated using an appropriate failure criterion.

Composite structures may be subjected to nonlinear behaviours in various ways. Structures made of thin laminates are prone to a large-deflection nonlinear response in static as well as dynamic loading scenarios. This geometric nonlinear behaviour is modelled employing large-deflection options in FE packages. Buckling instability of thin structures is also one of geometric nonlinear problems. Similarly, material nonlinearity has a significant effect on stability and failure of composite structures. Many composite structures are made of brittle materials and thus can exhibit a nonlinear behaviour involving brittle fracture. Delamination modelling based on material’s softening behaviour also poses a nonlinear phenomenon in FE models.

Finite-element formulations based on CLT, FSDT, HSDT and layer-wise theories are capable to predict in-plane failure of composite plates. They are generally used to investigate a mechanical behaviour and failure of laminated composite based on first-ply failure criterion. However, in our research problem, a structure is subjected to transverse static and dynamic loading, and the major failure mode is that of delamination because of interlaminar stresses. Therefore, finite-element models based on the 3D elasticity theory are used to investigate the behaviour of a 3D stress field at the edges of laminated plate and to predict through-thickness failure of composite laminates.

A macroscopic behaviour of composite materials is determined by the properties of their constituents, i.e. fibres and matrix, at microscopic level. Fibrous composites show a high degree of spatial variation in their microstructure, resulting in their non-uniform and anisotropic properties. In contrast to more traditional homogeneous materials like metals and ceramics, composites demonstrate multiple modes of fracture and damage due to their heterogeneity and microstructure. Damage evolution affects both their in-service properties and
performance that can deteriorate with time. Therefore, to design and engineer composite materials for specific end uses, an in-depth knowledge of material properties at various length scales is required. Thanks to availability of high-performance computing resources, FEM is largely used to model composite materials at various scales. The next section highlights modelling strategies for fibrous composites using FE techniques at various length scales.

3.4 Modelling of composite materials at various length scales

The advantage of fibre-reinforced composites as structural components lies in the fact that the material’s behaviour can be tailored. Desired properties of composites at macroscopic level can be obtained by manipulating the constituents at microscopic level. Although composite laminates appear to be homogeneous macroscopically, they show various heterogeneities at microscopic level. Since the mechanical behaviour of composite laminates varies at different length scales, therefore, it is required to study their behaviour at various scales. Micromechanical analysis facilitates understanding of the effect of local properties of constituents and their arrangement on macroscopic material and structural behaviour, thereby accelerating the development cycle of a material system for a specific application. The task of micromechanics is to link mechanical relations at various scales. The aim of computational micromechanics is to determine relationships between the microstructure and the macroscopic response of a material using models on the micro scale that are as simple as possible [53]. Microstructural mechanics combines approaches of computational mechanics and materials sciences to estimate local deformation mechanisms in heterogeneous materials at the level of heterogeneity, to predict overall properties of heterogenous materials based on homogenisation techniques and to simulate local damage in heterogeneous materials [54]. Multi-scale material models are employed to predict the properties at various scale levels and then correlate them using approaches of continuum mechanics. With a rapid growth in computational power of computers, multi-scale modelling of fibre-reinforced composites has become an important means of understanding the behaviour of such materials. A multi-scale approach considers three scale levels for the analysis of heterogeneous composite materials [34] as described below:
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1. Micro-scale: The micro-scale is the lowest observation scale taking into account the behaviour of constituents (fibres and matrix) of the material. Here, fibre and matrix phases are modelled separately, and the average properties of a single reinforced layer are determined based on properties of individual constituent using a homogenisation technique. Interaction between constituents and the resulting behaviour of the composite (fields of micro-strain and -stress) is the main concern at this scale. In woven composites, damage mechanisms such as matrix micro-cracking, fibre/matrix debonding and fibre failure within tows can be modelled at micro-scale.

2. Meso-scale: The meso-scale considers the ply as a basic homogeneous continuum entity for mechanical analysis of, and failure prediction in laminated composites. Each ply is modelled separately as a homogeneous material and the fibre direction is taken into account in terms of orthotropy of the homogenous material. The ply’s mechanical and elastic properties can be determined through experimentation, but modelling at this scale does not provide any information about a character of interaction between the constituents. However, this scale can be much more easily implemented in analysis of large structures than the micro-scale due to lower computational effort. In woven composites, damage such as intra-yarn cracking and inter-ply delamination is normally predicted using this scale. In this approach, a virtual laminate is built by stacking plies with different fabric orientations, and the FE model explicitly includes each ply as well as interfaces between them. Meshing of the laminate is carried out with solid elements for the plies, while cohesive interface elements can be used to account for ply interfaces in the model. This modelling strategy presents two main advantages. Firstly, full 3D stress states can be considered contrary to simulations based on the use of shell elements for composite plies, which are limited to 2D stress states. Secondly, intralaminar and interlaminar damage can be introduced separately together with a complex interaction between them. However, the main limitation of this approach lies in computational power required to carry out such simulations for large structures [55]. Ladeveze was among the pioneers to propose meso-level modelling approach for damage in composites [56].
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3. Macro-scale: The macro-scale is defined at the level of components at which the structure is a completely homogeneous continuum and its material behaviour is described by an anisotropic constitutive law. The overall structural response to external loading makes a continuum mechanics problem and can be investigated by using FE models with effective (average) material properties. At this scale, composite's failure is characterised by a strong interaction of intra-ply cracking and inter-ply delamination and ultimate fabric rupture [39]. Main advantages of macro-scale models are their simplicity as well as capability to be adapted to different geometries and types of fabric reinforcements, provided the respective mechanical tests that define them are performed. The main disadvantage is that since the reinforcement is not modelled explicitly, the actual damage mechanisms are not captured, leading to an arguable lack of physical representativeness [57].

Multi-scale modelling is based on two different analysis procedures of homogenisation and localisation. The homogenisation technique allows the behaviour of heterogeneous material to be regularised (homogenised) as a continuum. This technique is designed to estimate the effective properties of composite materials at macro-scale based on the known properties of several constituents and microstructural morphology. The approach is to compute a constitutive relation between volume-averaged field variables. Then, the homogenised properties can be used in a macroscopic analysis of the material. The volume averaging takes place over a statistically representative sample of the studied material referred to as a representative volume element (RVE). This statistical representative volume has to be large enough to reproduce information concerning the material's global behaviour. The choice of RVE depends upon the detail required to characterise the phenomena occurring inside the material [24]. However, the homogenisation process based on average values does not account for damage at micro level triggered by localised stress filed. Therefore to deal with physical events in the microstructure such as fibre damage etc., a localisation technique or a periodic microfield approach can be used to evaluate local stresses and strains at micro-scale. Analysis of periodic materials is based on the repeating unit cell concept (RUC) and the associated periodic boundary conditions. Here,
spatial placements of the reinforcements are assumed at regular locations in space. Numerical simulations of the mechanical response of a unit cell with prescribed loading and boundary conditions are carried out to determine the material’s macroscopic properties [58, 59]. Apart from the material characterisation and constitutive modelling, the micromechanical approach is also used to study local phenomena in heterogeneous materials such as initiation and evolution of microscopic damage, nucleation and growth of cracks, effects of local instabilities, stresses at intersections between macroscopic interfaces and free surfaces, and the interactions between phase transformations and microstresses [60]. Ernst et al. [61] carried out multi-scale progressive failure analysis of woven laminates. They first determined the material constants by using micro-scale RVEs and then implemented the material data in a macro-scale model of three-point bending test. Finite-element models incorporating various damage modes at the constituent scale of yarns and matrix in woven composites were developed in [17, 62-64] among others, yet the problem domain was limited to a representative volume or unit cell and the full laminate was not modelled. The meso-level modelling approach was employed [65-69] to characterise the damage behaviour of fabric-reinforced composite structures.

However, this hierarchical multi-scale approach faces major difficulties in selecting length scales for damage initiation and progression, and modelling multiple damage modes occurring simultaneously and interacting with each other. Talreja [16] proposed a synergistic damage mechanics approach, an alternative to hierarchical approach, combining continuum damage mechanics for macro- and meso-scale modelling and micro damage mechanics for microstructure modelling. Silberschmidt [70] showed that the effect of random distribution of fibres in composite laminates resulted in non-uniformity of damage processes and cracking evolution and their effect on the composite’s response to external loading using multi-scale damage models.

As mentioned earlier, computational modelling is used to investigate damage at various scale levels. Subsequently, multi-scale modelling transfers the damage information from lower to higher scale to predict the catastrophic collapse of laminates. However, multi-scale failure analysis using the finite-element models combining coarse meshes and finer meshes at the macro and micro levels
respectively, is still computationally costly because of a large volume of calculations. Further, a complex weaving architecture as well as multiple modes of damage at various length scales in textile laminates makes micro-mechanics based constituent-level modelling more computationally expensive for problems of real life. Therefore, meso-level models coupled with continuum damage mechanics still works well in design and failure prediction of composite laminates: it is employed also in this work. In the following section, the analysis procedures for FE modelling of composites and there damage are described.

3.5 Implicit vs. Explicit Schemes

Implicit and explicit techniques are widely used in finite-element methods to solve linear and nonlinear systems of equations [71]. Implicit methods are based on static equilibrium. The basic statement of static equilibrium is that the internal forces exerted on the nodes \( I \) (resulting from the element stresses) and external forces \( P \) acting at every node must balance [48]:

\[ P - I = 0 \]  

(3.1)

The solution procedure iterates successively until convergence is achieved. This incremental-iterative solution technique is usually based on the Newton-Raphson method as in Abaqus/standard. The method is unconditionally stable i.e. any size of increments can be used. This method requires a solution of a banded set of simultaneous equations at a series of load increments, and needs constant updating of a global stiffness matrix and a series of iterations in order to achieve convergence. However, every iteration requires a tangent stiffness matrix that needs careful handling in numerical aspects for its solution. One of the most challenging issues in the cohesive-zone method using the implicit solver is convergence of the FE model during its softening behaviour [72]. Also, implicit methods require much more computational resources and a higher computation time per cycle than explicit methods. Large matrices need to be stored and a large system of algebraic equations should to be solved in each cycle. Consequently, implicit methods are mostly used for static and low-rate dynamic analyses.

An explicit method uses a central difference rule to integrate the equations of motion explicitly with regard to time, using kinematic conditions at one increment
to calculate those at the next increment [48]. The explicit methods are based on
dynamic equilibrium which includes the inertial forces:

$$ P - I = m\ddot{u} $$

(3.2)

where $m$ is the mass and $\ddot{u}$ is the acceleration of the structure. The explicit
method does not deal with the tangent stiffness matrix, and, therefore,
convergence is not an issue. Each time increment is relatively inexpensive as
there is no need to solve a set of simultaneous equations. The explicit method is
conditionally stable i.e. for the solution to be stable; the time step has to be small
enough, such that information does not propagate across more than one element
per time step. Thus, this method needs very small time increments in order to
have stable solutions in time, leading to higher CPU times for time-dependent
loading [71]. This restriction makes the explicit method inadequate for long
duration dynamic problems. The explicit formulation is often appropriate in cases
with severe changes in stiffness matrix, such as analysis with failure or
degradation of the material. An explicit dynamic analysis approach is typically
adapted to model large deflections, material nonlinearities and contact behaviours
in high-velocity transients but it can be also employed effectively in modelling
dynamic phenomena with severe discontinuities in the structural response, as is
the case in unstable crack propagation. Since time integration is easy to
implement, the material nonlinearity can be cheaply and accurately treated, and
the computational resources required are small even for large problems. Menna et
al. [67] performed numerical simulations of low-velocity impact tests on glass
fabric/epoxy laminates through the explicit FE code LS-DYNA. Iannucci and
Willows [65] modelled impact induced damage such as delamination in woven
composite using the explicit code LS-DYNA 3D. Similarly, Johnson et al. [66]
studied impact damage in woven GFRP composites panels for marine applications
using Abaqus/Explicit solver. The method can also be used for quasi-static
analyses by artificially increasing the load rate or material's mass to increase the
stable time increment for fast solutions. Pinho et al. [73] employed the explicit
method to model mixed-mode delamination in composites using cohesive-zone
elements under quasi-static loading conditions Gözlüklü et al. [74] modelled
delamination propagation in L-shaped laminated CFRP composites under quasi-
static bending using the explicit solver of Abaqus. Based on the advantages, the
explicit method is ideal for nonlinear dynamic problems such as impact and penetration and large-deformation quasi-static simulations and it is employed in this study. The following section highlights various FE modelling and analysis techniques for damage behaviour of composite laminates.

3.6 Finite-element modelling of damage mechanisms

Failure prediction of fibre-reinforced composite laminates is complicated by their heterogeneous nature, which gives rise to various types of multiple cracks, interacting strongly as failure progresses. These damage mechanisms may cause significant redistribution of stresses and thus affect the load level, at which final structural failure occurs. Design and certification of most composite structures are based on empirical approaches because of the difficulty of complete damage-process prediction, with relatively little use of simulations. Therefore, there is a need for models, capable to simulate the entire damage process from its initiation through evolution to complete failure of the composite structure. Analytical models are impractical and, probably, unable to model this complex damage process, initiating from matrix cracking, evolving in delamination and fibre breakage to composite structural ultimate failure. The most promising and suitable tool is a computational approach based on the finite-element method (FEM). This approach unlocks a full potential of composites resulting in more rational and optimised designs of composite laminated structures. However, the development of proper numerical model representing the physics of damage mechanisms is a challenging task [75].

Reliable and accurate simulations of discrete damage behaviour of composite laminates require guidance from experimental and theoretical studies of damage mechanisms. Understanding a sequence of different damage modes in ply-scale damage and defining physical parameters in material’s constitutive laws that determine which mode will dominate is a challenge of respective simulations. According to Cox and Yang [76], the difficulty in composite damage modelling is linked not only to insufficient computational power. A more serious challenge is to categorise and characterise many possible mechanisms of damage and represent them in a model in a realistic and physical way. Similarly, understanding the origins of numerical instabilities that often occur in simulations of heterogeneous materials poses another challenge. It is critical to know whether these instabilities
are due to numerical approximations or rather they reflect physically unstable damage propagation, such as the dynamic crack propagation that is often observed in experiments. Modelling of cracking sequences and potential instabilities successfully in a computationally cost-effective way is of key interest to developing tools for use as virtual tests [77]. Various approaches are implemented in finite-element models to characterise the onset and progression of damage for analysis of composite structures. The studies involve monitoring of a particular type of parameter such as stiffness degradation for prediction and monitoring of damage growth. In the next section, various damage characterisation and analysis approaches based on numerical techniques are presented.

The damage in composite laminates is a complex phenomenon and results in various failure modes that interact in a unique pattern. Usually, failure of the first ply represents the damage initiation and evolution of the composite laminate, but does not lead to the ultimate structure failure. According to Puck and Schürmann [78] analysis of damage initiation and evolution in composite laminates requires (a) analysis of strains and stresses ply by ply; (b) failure criteria for single lamina; (c) degradation models to include the effects of damage, which often does not lead to ultimate failure of the laminate; and (d) a computer program, which simulates the gradual failure process by applying the above sequences. The approaches to model crack initiation and damage evolution are discussed below.

3.6.1 Strength-based approach

Strength of a material can be used to characterise the onset of damage in composite laminates. Strength-based failure criteria predict the onset of different damage mechanisms in composites and, depending on the material, its geometry and the loading conditions, can also predict final structural collapse. According to this approach, microcracks form when the stress reaches the transverse strength of the ply material or some multi-axial stress criterion is satisfied [19, 79]. Application of the strength approach is based on defining one or more strength criteria, and the structure is deemed to have been irreversibly damaged once these criteria are met. These models fail to account for difference in crack initiation and progression for specimens of different ply thicknesses since the stress state at the onset of transverse cracking is not constant for different laminates. Moreover, the main drawback of strength-based criteria is a lack of agreement with the
experimental predictions as shown by the failure exercise of Hinton and Soden [80]. It is important to note that strength-based characterisation of damage is most commonly applied to define the damage initiation, and not the progression of an existing damage region such as delaminations between plies [81]. A large number of strength-based criteria have been derived to relate stresses and experimental measures of material strength to the onset of failure. One of the first models to predict microcracking was based on the first ply failure theory, where it is assumed that the first crack develops when the applied strain in the plies reaches the ply’s failure strain. Another simplest approach is the ply discount method, which completely neglects transverse stiffness of the cracked plies; it underestimates stiffness of the cracked laminate. Predictions based on both models were not in agreement with experimental observations [34]. Highsmith and Reifsnider [82] developed shear lag models, where a load transfer between plies was assumed to take place in shear layers between neighbouring plies. The shear lag theory neglects variations in the stresses and strains through ply thickness. However, this theory is based on approximations that render predictions subject to error. To overcome these limitations, Hashin [28] and Nairn [29] developed models of stress transfer in cracked cross-ply laminates based on variational method. This approach attempts to solve a two-dimensional boundary value problem, and thus yields much better results than shear lag models. Prediction of theromechanical parameters such as stiffness, local ply stresses and coefficients of thermal expansion of cross-ply laminates with regularly spaced ply cracks based on these models were in good agreement with experimental data. Failure criteria such as the maximum stress, Hashin, Hoffman, Yamada-Sun, Puck, Tsai-Hill and Tsai-Wu [83] were developed over the past five decades for strength and failure analysis of laminates, but no universally accepted failure criterion exists. Among these, interactive failure criteria such as Hashin [84], Puck [78] and Tsai-Wu [85] are widely used to characterise separate damage types in composites. Chang and Chang [86] developed a progressive damage model for notched laminated composites subjected to tensile loading, capable of assessing damage in laminates with account for material’s nonlinearity. Recently, Pinho et al. [87] developed failure criteria denoted as LaRC04 that was based on 3D stress state and included nonlinear matrix shear behaviour, providing good correlation with experimental data.
Most of these failure criteria are implemented in commercial FE codes such as Abaqus, Ansys etc. to determine whether a composite structure will fail. Numerous researchers have applied strength-based criteria to predict damage in woven composites. Daggumati et al. [17] detected damage initiation at yarns with Hoffmann criteria in a meso-level FE model of 5-harness satin woven composite subjected to tension. Menna et al. [67] employed Tsai-Wu criteria to predict intra-ply damage in simulations of woven GFRP laminates under impact loading. Santiuste et al. [88] developed a progressive failure model based on Hou and Hashin criteria in Abaqus/Explicit to predict failure modes of composite laminated beams subjected to low-velocity impacts in a three-point bending configuration. However, the use of strength-based failure criteria for composite materials has the major disadvantage that the scale effect relating to the length of cracks subject to the same stress field cannot be modelled correctly. Hence, the use of fracture mechanics- or energy-based approaches is attractive for modelling of matrix cracks or delaminations [89].

### 3.6.2 Fracture mechanics-based approach

Strength and failure analysis of laminated composites is rather different from the analysis of strength of a single ply. Failure of laminates usually involves matrix cracking and delamination between the plies. It has become a common practice to investigate progression of these damage modes using fracture mechanics-based approaches. Fracture mechanics deals with the influence of defects and cracks on the strength of a material or structure. The main objective of fracture mechanics analysis is to predict the onset of crack growth for a structure containing a flaw of a given size. It has generally been assumed that the size of a plastic zone at the crack tip is small compared to the crack length while calculating the critical load for a cracked composite. Linear elastic fracture mechanics has been found useful for certain types of cracks in composites, i.e., interlaminar cracks such as delamination, or matrix cracks in a unidirectional composite [90]. In the fracture mechanics theory, the growth of a macroscopic defect is controlled by the rate of strain energy released in propagation, as compared to a threshold maximum strain energy release rate for that material, also known as *material toughness* [81]. The strain energy released in crack propagation is typically split into three components linked to separate mechanisms of crack growth. Mode I refers to opening or
peeling of crack surfaces, Mode II refers to sliding, and Mode III refers to tearing as shown in Fig. 3.1. This approach has proved to be highly successful, when complemented with an accurate stress analysis approach for damage initiation. However, the fracture mechanics approach cannot be easily incorporated into a progressive failure methodology [89].

![Modes of crack growth: (a) Mode I (opening); (b) Mode II (sliding); (c) Mode III (tearing) [90]](image)

Computational modelling of delamination requires the account for progression of damage area in the analysis. Fracture mechanics analysis has been limited in this respect due to complexities involved in monitoring crack growth and a typical requirement for a fine mesh around the crack front, which usually means either a highly dense mesh or computationally expensive re-meshing [81]. Further, prediction of fracture properties requires special techniques such as J-integral proposed by Rice [91]. But this approach cannot be employed to calculate the energy release rate for complex 3D laminated structure since it is limited only to the case for plane structure and also requires a high quality dense mesh at the crack tip when using finite-element method. These factors make the application of fracture mechanics techniques limited to predicting the initiation of crack growth and not its progression [81]. However, recent approaches based on fracture mechanics such as virtual crack closures technique (VCCT) were developed and implemented successfully in commercial FE codes such as Abaqus, for crack propagation analysis. This technique is applied during every increment of a nonlinear analysis, and uses single-mode and mixed-mode fracture criteria to determine when attached nodes should be released to represent crack growth.
The VCCT was initially proposed by Rybicki and Kanninen [92] based on Irwin’s crack tip energy analysis for linear elastic materials [93]. The method employs an assumption that the strain energy released in the process of crack growth by a certain amount is the same as that required to close the crack by the same amount. The VCCT is widely used to compute the energy release rates at the crack tip based on results from continuum (2D) and solid (3D) finite-element analyses to provide the mode separation energy required when using the mixed-mode fracture criterion [94]. Further, this technique does not require a high mesh density like other fracture mechanics based approaches such as J-integral. Pereira et al. [95] implemented the VCCT in Abaqus using 3D 8-node brick C3D8R elements to determine the energy released in Mode I double cantilever beam (DCB) tests performed on woven glass/epoxy multidirectional laminates. Shindo et al. [96] employed the VCCT in a FE model to calculate the energy release rate in Mode II fatigue delamination growth in woven GFRP laminates at cryogenic temperatures. Hallet et al. [97] used VCCT to determine the applied load that would cause free edge delamination in modelling interaction between matrix cracks and delamination in scaled quasi-isotropic specimens under tensile loading. Marsavina and Sadowski [98, 99] determined stress intensity factors at the tip of a crack in bi-material ceramic interfaces under bi-axial state of stress using finite-element method.

The VCCT is one of the most commonly applied methods for determining components of the strain energy release rate along a crack front in delamination propagation in composite laminates. The strain energy release rates can be calculated from the results of a single analysis performed on the FE mesh illustrating the similarity between crack extension from \( i \) to \( j \) and crack closure in Fig. 3.2. The Mode I, and Mode II components of the strain energy release rate, \( G_I \) and \( G_{II} \) are determined for four noded elements of unit thickness as shown in Fig 3.2 in the following way:

\[
G_I = F_{yj}(v_i - v_i)/(2\Delta a),
\]

\[
G_{II} = F_{xj}(u_i - u_i)/(2\Delta a),
\]

where \( \Delta a \) is the length of the elements at the crack front and \( F_{xj} \) and \( F_{yj} \) are the forces at the crack tip to hold the nodes together at point \( j \). The relative
displacements behind the crack tip are calculated from the nodal displacements at the upper crack face $u$ and $v$ at node $j$ and the nodal displacements $u'$ and $v'$ at the lower crack face node $j'$, respectively. However, the VCCT is only meaningful when a predefined crack exists in the laminate as it is based on linear elastic fracture mechanics (LEFM) approach; a separate strength-based failure criterion must be defined to predict the onset of crack growth. Further, it fails to take into account the effect of a zone of nonlinear fracture process at the crack tip, which is important when this zone is of comparable size to that of a specimen, or when multiple cracks interact [77]. An alternative approach to treat such a nonlinear behaviour is to model crack and delamination propagation in FE models by employing cohesive-zone elements that will be discussed in the later sections. The next section elaborates the continuum damage mechanics approach to model the damage and failure of composite laminates.

3.6.3 Damage mechanics approach

Strength-based failure criteria are commonly used to predict initiation of different failure mechanisms in composite laminates and can predict the ultimate structural failure depending on the given geometry, material and loading conditions. Since, composite structures accumulate damage before their ultimate collapse, the application of failure criteria is insufficient to predict this behaviour. The simplest
way to model this damage behaviour is based on continuum damage mechanics (CDM) first proposed by Kachanov [100].

Damage mechanics approach has been widely used to predict the stiffness degradation and damage evolution in composite laminates. Damage can be interpreted as the creation of microcracks and microvoids in a loaded material. Such microscopic damage behaviour is characterized by reduction of material stiffness at macroscopic level. According to Talreja and Singh [1], damage mechanics is “a subject dealing with mechanics-based analysis of microstructural events in solids responsible for changes in their response to external loading”. Hence, it is an approach for modelling a material response that attempts to quantify the physical events contributing to the evolving damage state. The application of damage mechanics involves introducing a phenomenological damage variable $D$ in the material’s constitutive model to represent initiation and progression of damage, which is monitored throughout the analysis. Multiple damage variables can be implemented to represent separate damage mechanisms, or a single damage variable can be used to capture the effects of all types of damage. A constitutive model relating the stress tensor $\sigma$ to strain $\varepsilon$ for a damaged composite laminate is given as

$$\sigma = (1 - D) E \varepsilon,$$  \hspace{1cm} (3.5)

where $0 \leq D \leq 1$ and $E$ is the material stiffness matrix. When the damage variable $D = 0$, it corresponds to a perfect undamaged material, and $D = 1$ to a completely damaged material. After initiation of laminate damage, the applied loads are resisted only by the undamaged ligaments such as fibres in the laminate. Stresses in fibres continue to increase until all fibres are severed and the laminate fails. In progressive failure analysis based on CDM analysis, mechanical properties of damaged material are replaced with those of a homogeneous material by associating the damage mechanisms with their effects on the mechanical behaviour of laminated materials. The failure of the first layer indicates initiation of failure and damage evolution in the composite, but it does not represent the ultimate damage since the composite has still residual load-bearing ability [83].

Various continuum damage models were developed by many researchers, e.g., Talreja [101], Matzenmiller et al. [102] and Miami et al. [103] among others. In these models, various damage modes in composites were assumed and
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constitutive relationships between the damage tensor, conjugate forces and internal stress/strains were formulated to describe a progressive failure process and to interpret stiffness degradation of composite laminates. However, such thermomechanics-based models were usually developed for elastic-brittle behaviour, neglecting the effect of plasticity in some matrix-dominated off-axis laminates such as ±45 symmetric ones, and were usually limited to plane structures. Barbero [104] presented a model by coupling CDM with a classical thermodynamic theory to predict inelastic effects such as reduction of stiffness and increments of damage and unrecoverable deformation; unrecoverable deformations and damage were coupled by the concept of effective stress. Talreja [105] used an alternative approach based on CDM to describe the mechanical behaviour such as stiffness degradation of cracked laminates. The reduction of laminate stiffness was modelled in terms of internal damage state parameters. It was necessary to fit certain parameters to experimental or numerical data for application of this model.

CDM-based modelling has found a wide application in analysis of damage and fracture of woven composites subjected to static and dynamic loads. Johnson et al. [106] developed a CDM model for fabric-reinforced composites as a framework within which both in-ply and delamination failures in impact loading were modelled using FE code Pam-Crash. The fabric ply was modelled as a homogeneous orthotropic elastic or elastic-plastic (in-plane shear) material with damage, whose properties were degraded on loading by microcracking prior to ultimate failure, giving a good agreement with experiments. Iannucci and Willows [65] presented an energy-based damage mechanics model for woven carbon composites under impact loading by introducing various damage variables for in-plane damage along the warp and weft directions as well as shear nonlinearity into the FE code LS-Dyna. The evolution of each mode of damage was controlled by a series of damage-strain equations, thus allowing the total energy dissipated for each damage mode to be set as a material parameter. Hochard et al. [107] developed a CDM-based model and a non-local ply scale criterion for failure prediction for woven CFRP. The model was implemented in Abaqus using a user-defined material routine (UMAT) and validated with experiments for a notched composite plate. Johnson et al. [66] implemented a CDM-based model in Abaqus/explicit
employing a user-defined material subroutine VUMAT to study the in-plane damage whereas inter-ply delamination was modelled with cohesive-zone elements in woven GFRP composites under impact loading. A recent approach of modelling damage in composites is that of combining the CDM approach for the intralaminar (bulk) damage and cohesive-zone models (CZMs) for interlaminar damage in the commercial FE codes.

However, several recent studies, such as [75, 108, 109] showed that computational models based on CDM are inadequate to capture the interaction of discrete cracks in a composite laminate. These models are incapable to represent complex matrix failure behaviour because of homogenization that is inherent in continuum models, irrespective of the applied failure criteria and material degradation laws [75]. Such process leads to the loss of key information on multiple-mode damage coupling at the macroscopic scale and, thus, may result in inaccurate prediction of a crack path [75, 108]. It is essential to have an explicit kinematic representation of all damage mechanisms in global models of composite structures to accurately predict interaction between these discrete failure modes, [109]. This inadequacy of CDM can be overcome with a recent trend of representing all major damage modes in FE models by cohesive-zone elements. Advanced numerical damage schemes based on cohesive-zone models (CZMs) also have the capability to couple directly the bulk (intra-ply) and interface (inter-ply) cracks to achieve unification of crack initiation and propagation [109]. Still, it had to be recognised that CDM performs well in damage analysis of composites since the multi-scale failure analysis using FEM costs more than the CDM in terms of computational time and efficiency. The next section describes the damage modelling technique based on cohesive-zone elements.

3.6.4  Cohesive-zone models

Computational simulation of delamination requires a capability to model initiation and progression of damage during analysis. Delamination initiation in composite laminates is usually assessed by strength-based criteria. Several techniques based on fracture-mechanics are employed in the finite-element method to simulate a delamination growth such as the J-integral, the virtual crack extension technique and the virtual crack closure technique (VCCT) [110]. Fracture-mechanics analysis is limited in this respect since it neglects material's
nonlinearity in most cases and requires a position of delamination crack to be known in advance [111]. Further, typically, a fine mesh around the crack front is required, which makes the analysis of three-dimensional composite structures rather computationally expensive. Therefore, numerical prediction of the effects of interlaminar damage on the behaviour of composite laminate requires a finite-element scheme that is capable to model strength as well as toughness of the inter-ply layers.

A reliable and promising approach to overcome the above issues and model the material as well as geometric nonlinearities is to employ cohesive elements at the interface between the composite laminae. Cohesive-zone elements are based on the model proposed by Dugdale [112], who introduced the concept that stresses in the material are limited by the yield stress and that a thin plastic zone is generated in front of the crack. Barenblatt [113] introduced an idea of cohesive forces on a molecular scale in order to solve the problem of equilibrium in elastic bodies consisting of cracks.

**Traction-separation constitutive laws**

Cohesive-zone damage models define relationships between tractions and displacement at an interface, where a crack may occur. Various constitutive models defining the traction-displacement behaviour of cohesive elements have been developed. Needleman [114] was one of the first to use polynomial and exponential types of cohesive-zone models to describe the process of void nucleation from initial debonding to complete decohesion in metal matrices. Tvergaard and Hutchinson [115] proposed a trapezoidal law to calculate the crack growth resistance in elasto-plastic materials, whereas Cui and Wisnom [116] defined a perfectly plastic rule. The irreversible, bi-linear, softening constitutive behaviour used in this study, was developed in previous works by Mi et al. [117], Alfano and Crisfield [111], and Camanho and Davila [118]. Other formulations have been proposed such as Yang and Cox [119] developed cohesive models for damage evolution in laminated composites.

Damage initiation in the bi-linear cohesive law is related to interfacial strength, i.e., the maximum traction on the traction-displacement jump relation, at which reduction of material’s stiffness starts as shown in Fig 3.3. Stiffness degradation continues until the interface elements attain zero stiffness, corresponding to
complete separation of adjacent layers. After this the interface elements act only as a contact region without transferring loads from them. The work required to reduce the material’s stiffness to zero is equal to fracture toughness, i.e. the area under the traction–displacement curve [120]. The bi-linear relationship, which is preferred due to its simplicity and used for quasi-brittle polymer based composites, can be defined as

\[
\sigma = \begin{cases} 
K\delta & \delta \leq \delta_0 \\
(1 - d)K\delta & \delta_0 < \delta < \delta_f \\
0 & \delta \geq \delta_f 
\end{cases} \quad (3.6)
\]

\[
d = \frac{\delta_f(\delta - \delta_0)}{\delta(\delta_f - \delta_0)}, \quad d \in [0,1] \quad (3.7)
\]

The traction-relative displacement curve shown in Fig. 3.3 can be divided into three main parts that are described below [121]:

(a) The portion for \( \delta \leq \delta_0 \) presents the elastic part where the traction across the interface increases linearly until it reaches the maximum. The stress in this portion of the law is linked to the relative displacement via the interface penalty stiffness \( K \) described by the first part of Eq. (3.6).

(b) The part of the curve for displacements \( \delta_0 < \delta < \delta_f \) is known as the softening part. Here, the traction across the interface decreases until it vanishes,
and the two layers begin to separate. The damage accumulated at the interface is characterized by a variable $d$ (see Eq. 3.7), which has a zero value in the undamaged state and reaches a value of one when the material is fully damaged.

(c) The portion of the graph for $\delta \geq \delta_f$ is called the *decohesion part*. Separation (or decohesion) of the two layers is complete, and there is no more bond between them. Interpenetration is prevented by reapplying only normal stiffness, in the case when a crack closure is detected.

The interfacial maximum strength in either Mode I or II for the bi-linear softening law can be determined as

$$\sigma_0 = K \delta_0$$  \hspace{1cm} (3.8)

Similarly, the critical energy dissipated per unit area of the softening curve which is equal to the fracture toughness $G_c$ for either Mode I or II can be determined as:

$$G_c = \frac{\sigma_{\text{max}} \delta_f}{2}$$  \hspace{1cm} (3.9)

Thus, at least five properties are required to define the interfacial behaviour of composites under quasi-static conditions: penalty stiffness $K$, corresponding fracture toughnesses in Modes I and II, $G_{IC}$ and $G_{II IC}$, the corresponding interlaminar normal tensile and shear strengths $\sigma_{I0}$ and $\sigma_{II0}$, respectively. In dynamic simulations of damage, the interface material density will also be required.

**Damage initiation and propagation**

The fracture mechanisms that the cohesive law represents are often mode-dependent, and, in general, a cohesive law should be defined for both normal and shear tractions as a function of both normal and shear openings. A delamination growth is likely to occur under mixed-mode loading in structural applications of composites, in which crack faces simultaneously open and slide relative to each other. Therefore, a general formulation for cohesive elements dealing with an onset and propagation of mixed-mode delamination is also required [118]. A mixed-mode traction separation law shown in Fig. 3.4 is used when a structure fails under combined fracture modes such as in bending. Various strength-based criteria are developed for damage initiation such as a maximum nominal stress
criterion and quadratic stress criterion. Cui et al. [122] highlighted the importance of interactions between interlaminar stress components when predicting delamination. It was shown that poor results were obtained using the maximum stress criterion. Therefore, the mixed-mode quadratic stress criterion accounting for the effect of interaction of traction components in the onset of damage proposed by Cui et al. [122] was used in this study. The nominal quadratic stress criterion for damage initiation used in this study is given as

\[
\left( \frac{\sigma_I}{\sigma_{I0}} \right)^2 + \left( \frac{\sigma_{II}}{\sigma_{II0}} \right)^2 + \left( \frac{\sigma_{III}}{\sigma_{III0}} \right)^2 = 1
\]

(3.10)

where \(\sigma_{I0}, \sigma_{II0}\) and \(\sigma_{III0}\) are cohesive interface strengths in opening Mode I and shear Modes II and III, respectively. The Macaulay bracket \(<\rangle\) shows that compressive normal tractions do not affect the damage initiation.

Delamination propagation is usually predicted by criteria established in terms of the energy release rates and fracture toughness under mixed-mode loading. Traditionally, two types of fracture energy based criteria are employed; one is the power law criterion proposed by Reeder [123] and the second developed by Benzeggagh and Kenane (B-K) [124]. Camanho and Davila [118] recommended the use of the B-K criterion for epoxy and thermoplastic based polymer composites. The B-K criterion used in this study for damage propagation is given as:

\[
G_C = G_{IC} + (G_{IIIC} - G_{IC}) \left[ \frac{G_S}{G_T} \right]^{\eta}
\]

(3.11)

where \(G_T\) is the work done by interface tractions; \(G_S/G_T\) is the fraction of cohesive energy dissipated by shear tractions; \(G_S\) is the work done by shear components of interface tractions; \(G_{IC}\) and \(G_{IIIC}\) are critical energy release rates in Modes I and II, respectively, and \(\eta\) is the material’s mode-mixity parameter. The parameter \(\eta\) is found by a least-square fit of experimental data points of fracture toughnesses under different mixed-mode ratios. There are established test methods to obtain interlaminar fracture toughness for Modes I and II. The double cantilever beam (DCB) test is used for Mode I fracture toughness \(G_{IC}\), whereas the end notched flexure (ENF) is employed to determine the Mode II fracture toughness \(G_{IIIC}\). For a
mixed-mode case of Modes I and II, the mixed-mode bending (MMB) test is normally used.

**Stiffness of cohesive-zone elements**

Interface elements act as load-transfer connections between continuum elements before the onset of delamination. Since cohesive elements represent interfaces with zero or infinitesimally small thickness, high stiffness is required to model the connections. Therefore, the interface stiffness should be large enough to avoid relative displacements between the connected ply elements but also not too large to cause numerical problems such as spurious oscillations in interfacial traction of the cohesive element [72, 126, 127]. This stiffness is usually calibrated by numerical simulations as it cannot be measured directly through the experiments. Several authors have proposed different methods to calibrate the cohesive element stiffness. Camanho et al. [128] obtained accurate results by using a value of $10^6$ N/mm$^3$ for graphite-epoxy specimens. Zou et al. [129] proposed a value for the interface stiffness between $10^4$ and $10^7$ times the value of the interfacial strength per unit length. Turon et al. [72] demonstrated that elastic properties of the composite would not be affected if the interface stiffness is defined as

$$K = \frac{\alpha E_{33}}{h} \quad (3.12)$$
where $E_{33}$ is the material's through-thickness stiffness, $h$ is the thickness of ply connected by the cohesive element, and $\alpha$ is a non-dimensional parameter, which should be greater than 50 for accurate simulation of various problems. Daudeville et al. [130] defined the cohesive-element stiffness as

$$K = \frac{E_{33}}{t_i}$$

(3.13)

where $E_{33}$ is the material's through-thickness stiffness and $t_i$ is the thickness of resin-rich interface between plies. The interface stiffness obtained from the Daudeville et al. equation is high. In this study, cohesive element stiffness is based on Eq. 3.13.

**Length and number of cohesive-zone elements**

In order to conduct an accurate delamination analysis using CZM, a mesh must be sufficiently fine to ensure that enough interface elements exist within the cohesive-zone length at the point of crack propagation. In case of a coarse mesh used for a cohesive-zone, the distributions of tractions ahead of the crack tip are not represented accurately. Different models were proposed in the literature to estimate the length of the cohesive-zone, $l_{cz}$ [72, 112, 113]. According to Yang and Cox [119], for delamination crack in slender bodies, the cohesive-zone length becomes a material and structural property [33]. For constitutive models that prescribe non-zero tractions when the displacement jump is zero, the length of the cohesive-zone was estimated by Yang and Cox [119] as:

$$l_{cz} = \left(\frac{E G_c}{\sigma_0^2}\right)^{1/4} h^{3/4}$$

(3.14)

where $E$ is the Young’s modulus of the material, $G_c$ is the critical energy release rate, $\sigma_0$ is the maximum interfacial strength and $h$ is the half thickness of the laminate (or thickness of the ply in case of inter-ply delamination). For the case of orthotropic materials with assuming a plane stress state, the value of the Young’s modulus in Eq. (3.7) is the transverse modulus of the material $E_{33}$. This length $l_{cz}$ is defined as the distance from the crack tip to the point where the final failure point is reached. The number of elements $N_e$ in a cohesive-zone according to Turon et al. [72], is given by

$$N_e = \frac{l_{cz}}{l_e}$$

(3.15)
where $l_e$ is the mesh size in the direction of crack propagation.

With respect to the minimum number of interface elements required within the cohesive-zone for an accurate damage analysis, recommendations in the literature show a wide variation, ranging from two to more than ten elements [72]. Recent analyses of composite delamination using cohesive elements have suggested that 2–3 elements are sufficient for an accurate load displacement analysis to be performed for cases of Mode I, Mode II and mixed-mode loading [72, 131]. To obtain suitable results by using cohesive-zone models, the tractions in the cohesive-zone have to be represented correctly by a proper number of elements. However, for large structural models, it may be computationally expensive to use excessively fine mesh. For this reason, it was suggested in [72, 131] to use the value of maximum interfacial stress significantly lower than the true physical value because it allows a relatively coarse mesh to be used, whilst still ensuring a sufficient number of interface elements within the cohesive-zone for an accurate delamination analysis.

The application of cohesive-zone models is becoming more and more popular for simulation of damage and fracture processes in composite materials. Various studies implemented the interface elements successfully for delamination analysis in composites under quasi-static loads such as [73, 74, 116, 128, 132-137] among others. The interface damage behaviour of fabric-reinforced composite structures subjected to dynamic loads was studied in [65-69]. In case of modelling composite delamination, cohesive-zone schemes offer a number of advantages over other modelling approaches, as they have the capacity to model both initiation and growth of damage in the same analysis, incorporating concepts of both the damage-mechanics and fracture-mechanics theories. However, application of cohesive-zone elements to model progressive delamination in composite structures poses numerical difficulties related to the proper definition of stiffness of the interface layer, the requirement of highly refined finite-element meshes, and convergence difficulties associated with a softening behaviour of the interface material [72]. Moreover, an ideal cohesive model should be able to model stable as well as unstable crack propagations and the transitions between the propagation regimes [127]. Still, these physical and numerical issues can be properly handled by defining various parameters for accurate simulations of
damage in composites, which makes the application of interface elements ever increasing. In this study, both the interlaminar and intralaminar damage modes are analysed by means of cohesive elements.

It is obvious that many research works using CZMs are available in the area of damage analysis of laminated composite. Here, some of the important contributions along with techniques for defining parameters of CZMs were summarized. Finally, an overview of some new techniques for damage analysis of composites is presented in the following section.

**3.6.5 Further techniques**

Apart from the methods mentioned above, there are some other new techniques that can be applied to simulate damage using different approaches, such as extended finite-element method (XFEM), augmented finite-element method (AFEM) and phantom node techniques. XFEM makes modelling of cracks easier and accurate and does not require a mesh to match the geometry of discontinuities (cracks) and partitioning of geometry at the crack location as in the case of CZMs. This can be used to simulate initiation and propagation of a discrete crack along an arbitrary, solution-dependent path without the requirement of remeshing and can also be employed in conjunction with the CZMs or VCCT. Predefined element boundaries are not needed for crack propagation as it fractures the element’s interior. XFEM extends the piecewise polynomial function space of conventional FEM with extra functions called “enrichment functions”. The numerical technique was introduced by Belytschko and Black [138] based on the partition of unity method of Melenk and Babuska [139]. Applications of this technique include modelling of bulk fracture and failure in composites. Moës and Belytschko [140] employed XFEM to simulate growth of arbitrary cohesive cracks, governed by a requirement for the stress intensity factors at the tip of the cohesive-zone to be vanished. Motamedi and Mohammadi [141] used XFEM for dynamic propagation analysis of moving cracks in composites by introducing time-independent orthotropic enrichment functions. Kästner et al. [142] developed a multi-scale XFEM-based model to simulate the inelastic material behaviour of textile GFRP composites. Abdel-Wahab et al. [143] analysed the fracture of cortical bone at various scales using XFEM in the FE code Abaqus. However, the method is still
limited in modelling branching and interaction of cracks, fatigue and dynamic crack growth as well as a single crack in an element.

The phantom node method for handling arbitrary cracking problems was developed by Hansbo and Hansbo [144]. In this method, a crack is represented as a mesh-independent discontinuity in the displacement field. This discontinuity is constructed with two overlapping elements, which have independent displacement fields, because additional ‘phantom nodes’ are placed on top of the existing nodes [145]. The two discontinuous domains are connected either by linear or nonlinear springs [144] or cohesive failure tractions [146]. Van de Meer and Sluys [145, 147] developed FE models based on the phantom-node method, implementing a mixed-mode cohesive law in an effort to simulate and analyse progressive matrix cracking, splitting, delamination and their interaction in notched cross-ply laminates under tension. The proposed method was capable of introducing a discontinuity in the displacement field at arbitrary locations, analysing matrix cracking with the phantom node method and delamination with interface elements. One major advantage of this method is that it uses only standard finite-element (FE) shape functions making it completely compatible with standard FE programs. However, adding elements as a crack evolves changes the dimension of numerical model dynamically, thus requiring full access to the source code, which is not always possible with commercial codes [109].

Another latest numerical method for simulation of cracking problems is augmented finite-element method A-FEM developed by Ling et al. [148], who proposed an element with double nodes that can treat an arbitrary intra-element discontinuity. They mathematically proved that a physical element, containing a discontinuity such as a cohesive crack or a bi-material interface, can be augmented internally as two combined mathematical elements, each using standard FE shape functions only. Further, they demonstrated that material’s heterogeneity in composite materials can be conveniently considered with this element. Fang et al. [109] exploited the capability of A-FEM in the analysis of an orthogonal double-notched tension composite specimen, in which delaminations interacted with transverse ply cracks, intra-ply splitting cracks, non-localized fine-scale matrix shear deformation and fibre breaks. The nonlinear processes of each fracture mode were represented by a cohesive model, which provided a unified description of crack
initiation and propagation and their interaction. The calibrated simulations were mesh independent and correctly reproduced all qualitative aspects of the coupled damage evolution processes. They also correctly predicted delamination sizes and shapes, the density of transverse ply cracks, the growth rate of splitting cracks, softening of the global stress–strain curve, and the ultimate strength. Fang et al. [149] also developed an augmented cohesive-zone element for arbitrary crack coalescence and bifurcation in heterogeneous materials, which was capable to accurately maintain a nonlinear coupling between various cracking modes.

Although these new numerical methods were developed for arbitrary cracking problems in composites, still they are not standardised based on their benefits and shortcomings. Therefore, this work incorporates the damage modelling capability of the advanced CZMs, which is well established for simulations of various damage modes in composites.

3.7 Summary

The background and the basic theory for FE modelling of composite laminates in 2D based on shell theories and 3D elasticity were described. Modelling of the damage mechanisms of these materials at various length scales under static and dynamic conditions were also described. Various damage modes were observed at each length scale in composites due to their heterogeneity and anisotropy. From a computational point of view, the question arises about the ways to realise this in a FE model, considering efficiency on the one hand, and regarding the physical structure properly on the other hand. Thus, a compromise must be found between computational efficiency and preciseness in modelling, which was a reason for the meso-scale modelling approach to be adopted in this study.

The basic principles of available methods and their implementation in the analysis of damage in laminated composites were discussed. Various damage modelling techniques such as strength-based criteria, VCCT, CZMs and some latest numerical approaches were reviewed and some advantages and disadvantages were mentioned. However, it is evident that there is no definitive failure strength-based criterion for composites. Still, damage mechanics-based models have shown significant success in modelling structural degradation in laminated composites.
Based on the review presented, it can be seen that extensive work has been done on the modelling and analysis of damage in laminated composites with their failures being modelled with approaches based on CZM, CDM and combination of both. Among these techniques, CZM seems to be more promising as it combines both strength- and toughness-based fracture criteria in a single element to simulate various damage and fracture modes. However, various damage modes exhibited by woven laminates under large-deflection loads and their effects on a structural global response is an area requiring further attention. Further, these damage modes are strongly interacting due to heterogeneity of composites, which are analysed in the context of CDM-based models. Due to the inherent homogenisation in the CDM models, they are incapable of modelling coupling of various damage mechanisms. Advanced CZM-based models have a potential to realise the individual damage modes at the ply and interface explicitly as well as their coupling. Therefore, developing progressive models based on CZM for interlaminar and interlaminar damage in woven composite materials having the capability to account for coupling of these damage modes is the main focus of this research. This goal will be achieved by a progressive damage modelling of woven CFRP laminates under large-deflection static and dynamic loading scenarios using the FE package Abaqus. To the author’s knowledge such descriptions of woven laminates under large-deflection bending are very limited. Further, three-dimensional multi-body dynamics models of such composites are rarely developed to investigate their behaviour under impact loading, which will be studied in this work.
Chapter 4: Mechanical Behaviour of Fabric-reinforced Composites under Quasi-static Loading

4.1 Introduction

Unlike metals, composites exhibit different mechanical properties in various directions due to their inherent anisotropy and heterogeneity. Characterisation of the mechanical behaviour of woven composites is necessary to evaluate their structural performance, provide data for analysis and design of composite laminates and understand their response to different loading conditions. In addition, these properties are necessary to develop numerical models to analyse deformation and fracture of composites at different structural levels. The experimental characterisation is based on determination of material’s behaviour and properties through tests on suitable material specimens. Elastic constants and strengths are the mechanical properties of composites to be known before evaluating their structural performance for specific applications. A significant amount of information has been acquired in the literature on mechanical properties of CFRP woven composites concerning their stiffness, strength and toughness. However, most experimental work is concerned with their tensile behaviour while the large-deflection flexure is rarely studied. Further, composites demonstrate variation in their mechanical properties with their size and volume, which makes it necessary to test specimens of specific size to get the accurate material’s data for the specific end use.

The objective of this chapter is to provide details of the experimental methods used in this study for characterisation of the stiffness and strength properties of CFRP laminates in various directions under quasi-static tensile and large-deflection flexural loading. Full field strain analysis for determination of in-plane shear behaviour of these laminates will be discussed. Reliable and accurate numerical modelling of discrete damage behaviour of composite laminates should be based on experimental characterisation of damage mechanisms corresponding to real-world loading scenarios. This objective was achieved by investigating the damage behaviour of the CFRP laminates tested under flexural loading using optical microscopy and X-ray Micro-CT techniques.
4.2 Materials and specimen preparation

The materials studied were laminates of woven fabric made of carbon fibres reinforcing thermoplastic polyurethane (TPU) polymer matrix. Due to their high strength and stiffness, carbon fibre composites were selected as compared to those with glass and Kevlar fibres. The TPU resin is tough providing an excellent impact and fatigue resistance behaviour and is applicable from −30°C up to 90°C [150], making it favourable for application in sports products. In addition, the TPU-Carbon laminate is easy to mould, obtaining an excellent surface quality. The laminates were produced from 0°/90° prepregs in the form of six and four plies designated as [0°,90°]_3s and [0°,90°]_2s, respectively. Here [0°,90°] holds for a single ply in which 0° and 90° represent yarns in the warp and weft directions, respectively. In this study, the stacking order of the form [0°,90°]_3s and [0°,90°]_2s will be designated as warp laminates (specimens), whereas [90°,0°]_3s and [90°,0°]_2s will be represented as weft laminates (specimens). All the woven laminates had a 2/2 twill 2D balanced and orthogonal weaving pattern; the fabric had the same number of yarns in the warp and weft directions as shown in Fig. 4.1. The parameters of the 2/2 twill weave fabric composite are listed in Table 4.1. As damage is sensitive to geometrical parameters of woven laminates such as the size and location of tows and resin within the laminate, thus, cross-sections of the laminate at various locations showing dimensions of tows and inter-ply resin-rich regions are illustrated in Fig. 4.2.

In this study, three types of fabric-reinforced CFRP laminates were used. The two main on-axis laminates, used in our tensile and flexural tests and numerical simulations, were 1.5 mm- and 1 mm-thick laminates designated as warp and weft specimens, respectively. The third type of CFRP laminates was the off-axis [+45/-45]_2s 2D woven laminate of 2 mm thickness consisting of 8 plies (that are the same as those in the main material); employed for measuring in-plane shear properties of the material.

The most commonly used specimen geometries for tensile testing are dog-bone specimen and the straight-sided rectangular specimen with end tabs. The dog-bone specimen may fail at the neck radius due to stress concentration and poor axial shear properties of the specimen. The flat specimens, used generally for
Chapter 4

**Fig. 4.1. Architecture of 2/2 twill weave**

**Fig. 4.2.** Through-thickness micrograph showing dimensions of weft tows and resin thickness measured at tows in resin-rich region (a) and surface (b) of on-axis 4-ply CFRP laminate
tensile tests of composites relieve these problems when used with end tabs held properly in serrated jaws. As compared to the dog-bone specimens, flat specimens with reduced gauge length are usually used as there is no stress concentration effect. Here, flat specimens were delivered in batches with nominal dimensions of 250 mm length and 25 mm width. Flat rectangular specimens of 80 mm length and 25 mm width were cut and ground to remove notches, undercuts and rough edges. The Labtom cut off wheel was used, consisting of a water cooled diamond cutting wheel that prevented heating associated deformation of

<table>
<thead>
<tr>
<th>Manufacturer’s Fabric ID</th>
<th>TEPEX® dynalite 208-C200(x)/45%</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Fibres</strong></td>
<td></td>
</tr>
<tr>
<td>Fibre type</td>
<td>Carbon T200</td>
</tr>
<tr>
<td>Filament diameter (µm)</td>
<td>7.0</td>
</tr>
<tr>
<td>Carbon fibre density (g/cm³)</td>
<td>1.75</td>
</tr>
<tr>
<td><strong>Yarns</strong></td>
<td></td>
</tr>
<tr>
<td>Yarn filament count</td>
<td>3000</td>
</tr>
<tr>
<td>Width, mm, warp/weft</td>
<td>1.58±0.06/1.78±0.08</td>
</tr>
<tr>
<td>Thickness, mm, warp/weft</td>
<td>0.17/0.13</td>
</tr>
<tr>
<td><strong>Fabric</strong></td>
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</tr>
<tr>
<td>Weave type</td>
<td>Twill 2/2</td>
</tr>
<tr>
<td>End/pick count, yarns/cm</td>
<td>5.2/5.14</td>
</tr>
<tr>
<td>Fabric areal density (g/m²)</td>
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<tr>
<td>Weight rate (%), warp/weft</td>
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</tr>
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<td>Thermoplastic polyurethane (TPU)</td>
</tr>
<tr>
<td><strong>Laminate</strong></td>
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</tr>
<tr>
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<tr>
<td>Thickness per ply (mm)</td>
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</tr>
<tr>
<td>Number of fabric layers</td>
<td>4, 6, 8</td>
</tr>
</tbody>
</table>
the material. This cutting technique resulted in high surface finish of the samples without causing any damage in the fibres and matrix. High strength materials such as CFRP need to be gripped firmly in the machine jaws to prevent slipping. However, stress concentrations are induced in the unprotected specimen ends, promoting premature failure of the specimen in the grips [90]. This problem can be alleviated by using end tabs in an attempt to reduce these stress concentrations and to protect the specimen ends from grip damage. Thus, end tabs prepared of the same material and orientation as that of the tensile specimen were adhesively bonded to the ends of the specimen. These tabs allowed a smooth load transfer from the grip/jaws of the testing machine to the specimen and also reduced the stress concentration. In all the tensile and flexural tests except in-plane shear tests, the specimens for both the warp and weft orientations had the same length and width as advised by their respective standards [151, 152].

4.3 Tensile tests

The uniaxial tensile test is the most widespread and most studied test for characterising the mechanical behaviour of composites. The popularity of this test method is explained mainly by its ease of processing, and analysis of the test results. The characteristics obtained from uniaxial tensile tests are used both for material specifications and for estimation of its load-carrying capacity. Practically all strength criteria include tensile strength. The test methods are used to investigate the tensile behaviour of composites and for determining the tensile strength, tensile strain, tensile modulus and other aspects of the tensile stress/strain relationship under the conditions defined. Although the focus of the current study is to investigate the mechanical behaviour of woven composites under large-deflection bending, the tensile tests are carried out to compare their behaviour to flexural. Still, in most studies material properties obtained through tensile tests are used for the flexure behaviour. The validity and comparison of this approach will be discussed in later sections.

In this study, all the tensile tests were performed on Instron 5569 machine in accordance with the ISO 527-4/5 standard [151]. The tests were conducted on 1.5 mm thick warp and weft specimens in displacement control at jaw speed of 100 mm/min, which is equivalent to strain rate 0.0417s⁻¹. This strain rate was used so that the tensile test results are compared with the flexural tests carried out at
similar strain rates as described in Section 4.5.1. The nominal axial strain rate can be calculated by dividing of the speed of the jaw of the machine by the gauge length of the specimen. As highlighted in the previous section, slippage of the CFRP specimens may occur in the grips, causing an error in the load cell reading of displacements and strains. Thus, for accurate strain measurement, an extensometer capable of 25 mm extension was used as shown in Fig. 4.3. To protect the extensometer against any possible damage, it was removed from the specimen before its ultimate failure. Five samples having a gauge length of 40 mm for both warp and weft orientations were tested until fracture. The load was applied to the specimens and the resulting extension was recorded over the test duration. From the measured data, the tensile strain was calculated as:

$$\varepsilon = \frac{\Delta L}{L_0} \times 100\%$$  \hspace{1cm} (4.1)

where $\Delta L$ is the increase in length and $L_0$ is the original (gauge) length of the test specimen.

Fig. 4.3. CFRP specimen in Instron 5569 with extensometer

The ultimate tensile strength $\sigma_u$ of the specimen was calculated by dividing the maximum applied load $F$ by its original cross sectional area as:
\[ \sigma_u = \frac{F}{w t} \]  

(4.2)

where \( w \) is the specimen width and \( t \) is its thickness.

The modulus of elasticity (Young’s modulus) \( E \) that provides a measure of stiffness of the materials, was defined based on the two specified strain values in the elastic region of the stress-strain curve according to ISO 527-1 standard [153]:

\[ E = \frac{\sigma_2 - \sigma_1}{\varepsilon_2 - \varepsilon_1} \]  

(4.3)

which is also the slope of the linear region of the stress – strain curve. Here, \( \sigma_1 \) is the stress, in megapascals, measured at the strain value \( \varepsilon_1 = 5000 \mu \varepsilon \) (0.5%); \( \sigma_2 \) is the stress, in megapascals, measured at the strain value \( \varepsilon_2 = 1000 \mu \varepsilon \) (1.0%).

Fig. 4.4. Stress-strain diagram from tensile tests of CFRP at 100 mm/min

Results of the tensile tests for both types of CFRP specimens are shown in Fig. 4.4 and presented in Table 4.2. The low strain regions of the curves in Fig. 4.4 are initially nonlinear. These portions of the curves represent the straightening of initially slack fibres within the tows. After the initial straightening of slack fibres, the stiffness of the fibre bundle is constant and the curves are linear until applied
stress reaches the failure stress and then a sudden load drop is observed. Both warp $[0^\circ,90^\circ]_{3s}$ and weft $[90^\circ,0^\circ]_{3s}$ samples exhibited a brittle failure response. Usually the on-axis laminates such as $[0^\circ,90^\circ]_{3s}$ and $[90^\circ,0^\circ]_{3s}$ undergo quasi-brittle fibre dominated damage due to the brittle fibres being aligned to the loading direction. The tensile strengths determined from the tests are 762 MPa and 764 MPa for the warp and weft samples (Table 4.2), respectively, which correlates with the manufacturer's quoted tensile strengths of 710 MPa and 705 MPa [150] for the same orientations with 1.5 mm specimen thickness, respectively. Similarly, the elongation at specimen fracture and elastic moduli also match well with the manufacturer's data. As shown in Table 4.2, tensile tests resulted in approximately the same elastic moduli and ultimate tensile strengths for warp and weft samples.

<table>
<thead>
<tr>
<th>Orientation</th>
<th>Ultimate tensile strength (MPa)</th>
<th>Elongation (%)</th>
<th>Elastic modulus (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Warp, $[0^\circ,90^\circ]_{3s}$</td>
<td>762.0 ± 12.0</td>
<td>1.55 ± 0.05</td>
<td>48.6 ± 1.3</td>
</tr>
<tr>
<td>Weft, $[90^\circ,0^\circ]_{3s}$</td>
<td>764.0 ± 10.0</td>
<td>1.56 ± 0.06</td>
<td>49.4 ± 1.2</td>
</tr>
</tbody>
</table>

This similarity is due to the symmetry of fibres in both warp and weft directions, i.e. the fabric is a balanced one in both types of samples. In fibre-reinforced polymer composites, the axial tensile strength is very sensitive to the fibre alignment with respect to the external load direction. This anisotropic mechanical behaviour of the material is explained by the fibre orientations affecting the load sharing mechanism. In on-axis laminates where the fibres are oriented at $0^\circ/90^\circ$, the maximum load is taken by the reinforcing fibres, and the material shows high failure strength and low ductility. Usually, the polymer matrix is much weaker in strength than the reinforcing fibres. When the reinforcing fibres fail at the ultimate strength, the polymer matrix alone is unable to sustain the applied load and thus fails abruptly showing no plastic flow. However, in off-axis laminates where the fibres are oriented at various angles, e.g. $\pm45^\circ$ with respect to the tensile loading axis, the applied load is resolved into axial and transverse components and induces shear flow in the matrix even at low tensile loads. When the local shear stress exceeds shear strength of the polymer matrix, it starts flowing plastically.
and dominates the material behaviour [154], which can be observed in the next section.

4.4 Tensile shear tests using digital image correlation method

A shear test of a composite material is performed to determine the in-plane shear modulus, the ultimate shear stress and strain as well as the in-plane Poisson’s ratio of the material. The shear response of a composite material is commonly nonlinear, and its full characterization thus requires generating the entire shear stress–strain curve to failure. Although many shear test methods are described in the literature, only a relatively few are standardised and in common use such as Iosipescu shear test, the two- and three rail shear tests, the \([\pm45]_\text{ns}\) tension shear test, and the short beam shear test [90]. Among these test procedures, the \([\pm45]_\text{ns}\) tensile shear test which is employed in this study, requires no special test fixture other than standard tensile grips and preparation of test specimen is relatively simple. Tensile tests were conducted according to ASTM standard D3518 [155] to investigate the in-plane shear response of off-axis \([\pm45/-45]_\text{2s}\) CFRP laminates. Although the standard outlines the procedure for using strain gauges in longitudinal and transverse directions, they only provide the localised strain data. Therefore, an ARAMIS digital image correlation (DIC) system was used to obtain the full-field in-plane displacement and strain data.

Digital image correlation is a non-contacting full-field deformation and strain measurement technique that uses no mechanical gauges and requires no mechanical interaction with the specimen; it was developed by Sutton et al. [156, 157]. The measurement technique digitally records a random speckle pattern on the specimen surface. The speckle pattern is applied by staining the surface of the specimen. DIC works by comparing images of a component obtained at various stages of loading and tracking blocks of pixels to measure surface displacement and build up full-field 2D and 3D deformation vector fields and hence in-plane strain maps. The position of the centre of pixel blocks is determined to sub-pixel accuracy over the whole image using correlation functions, from which the vector and strain components can be calculated. The advantages of this method are the availability of the full strain/displacement field, as well as the ability to experimentally determine all components of strain (normal and shear) at any point in the region of interest. This is particularly useful in characterisation of material
behaviour, crack tip and crack propagation studies, and mapping of strain fields around features and defects. If mechanical gauges are used, these data would require several tests with different gauges to collect. Other advantages are reduced vibration interference, fast data collection and a higher dynamic range of deformation measurement.

Fig. 4.5. Specimen and shear fracture of twill 2/2 [+45/-45]_{2s} woven CFRP laminates in tensile tests

4.4.1 In-plane tensile shear tests of CFRP

Tensile tests were carried out to determine the in-plane shear behaviour of CFRP laminates using the DIC technique. This technique has already been used to characterise the material behaviour of woven laminates, e.g. in [158]. Before performing tests, the specimens were sprayed with white paint, and then black speckles were marked on them to provide contrast necessary for image correlation as shown in Fig. 4.5. The aim of this patterning process is to provide enough variation in the recorded images that each area can be uniquely identified and its relative displacement observed. The woven off-axis [+45/-45]_{2s} CFRP specimens were loaded in the Instron 5569 machine; while two 1 megapixel digital cameras took photographs of the specimens at frequency of 50 frames per second during loading. Specimens were tested at a crosshead speed of 10 mm/min equivalent to a strain rate of 0.0014 s\(^{-1}\). Rectangular specimens of 200 mm length, 25 mm width and 2 mm thickness were prepared, each symmetric laminate having eight plies of 0.25 mm thickness each. The gauge length of samples was kept to 120 mm. In these tests, end tabs were not used because the tensile strength of a [±45]_{ns} laminates is low relative to that of an axially loaded on-axis composite laminates. The specimens were tested in tension to ultimate failure following the procedures outlined in [155]. The GOM DIC system consisted of two high speed PHOTRON SA 1.1 monochrome cameras, which can take photographs of the specimen at
frequencies ranging from 10 to 2000 Hz. The cameras were positioned at a distance of 1 m away from the specimen surface. Two standard halogen lamps on either sides of the camera guaranteed an even illumination of the specimen surface. The ARAMIS system with two-dimensional registration and the image size of 1024 x 1024 pixels was used to obtain an in-plane full field strain response of the laminates. The Instron machine had an integrated load cell, and load measurements were synchronised to each photograph at each time interval. Figure 4.6 shows the specimen inserted in the Instron machine and the DIC recording setup.

To obtain the in-plane strain field from the deformation of the speckle-patterned specimen’s front surface, digital images of the specimens were taken at 50 frames per second with an image resolution of 300 x 62 pixels during loading. Nonlinear shear stress-strain behaviour of off-axis CFRP specimen up to load level of 6.5 kN is presented in Fig. 4.7. The stress-strain curve becomes nonlinear at shear strain of approximately 0.01, which is due to matrix cracking and fibre trellising. Hence, the off-axis laminates such as [+45°/-45°]_3 exhibited matrix dominated damage resulting in nonlinear material behaviour similar to an elasto-plastic one. This
nonlinearity is attributed to the plasticity because of the matrix cracking and fibre trellising after crack saturation occurs in a laminate as explained in [159, 160]. In trellising, the fibres re-orient towards the direction of the loading vector and the angle between reinforcement directions changes from 90°, i.e. a square pattern becomes like rhombus. Contour plots of major (axial), minor (transverse) and shear strain from digitally imaged tension tests of CFRP samples in the linear elastic range are shown in Fig. 4.8. The variation in the strain contour plots might be due to more compliant regions such as resin pockets between the tows, or tows aligned along the loading directions (fibre trellising). These contour plots from the same test of CFRP samples at higher loads are shown in Fig. 4.9. The corresponding load level is slightly less than the laminate ultimate fracture load. It is noted that strain contours for the major, minor and shear plots are varying due to the material heterogeneity. The missing parts of the images represent the areas of paint flaking, which was speckled, and thus the data at these regions was not collected. Since the paint was too fragile, the speckles flaked off from the laminate at the time of specimen fracture. The localised plastic flow behaviour of the matrix is reflected as heterogeneous strain contours in the form of high strain regions in the strain maps of the tested CFRP material.

Fig. 4.7. Shear stress-strain diagram from digitally imaged tensile test of [+45/-45]$_{2s}$ twill 2/2 CFRP laminate
Fig. 4.8. Digital images of major (a), minor (b), and shear (c) strains from [+45/-45]_{2s} CFRP tensile test in linear range (50 MPa tensile stress)

Fig. 4.9. Digital images of major (a), minor (b) and shear (c) strain for CFRP sample at 240 MPa tensile stress
Figure 4.10 presents the applied tensile stress vs. major and minor strains recorded at points 1, 2 and 3 on the specimen surface. The tensile stress versus transverse strain is given on the left side of the figure and the tensile stress versus longitudinal strain is on the right side. The Poisson’s ratio of CFRP laminates is determined from the longitudinal and transverse strains in the linear elastic range of stress-strain curves in Fig. 4.10. At about 1% strain, the specimen starts to yield due to matrix cracking, and the density of matrix cracks keeps increasing with increasing load resulting in a nonlinear response. This crack density reaches a saturation level at about 7% strain, where the nonlinearity from matrix cracking is no longer prevalent. Now the material behaviour is controlled by fibre trellising, and the nonlinearity is due to fibre reorientation towards the loading direction. Fibre trellising continues up to strain of 28% for CFRP specimens, where eventually necking and the fibre failure begins. The final nonlinearity is a likely result of a random fibre failure process over a range of axial strain. Experimental tests show brittle matrix failure after considerable hardening. In a sudden event, the specimen breaks, with the crack running in fibre direction, as shown in Fig. 4.5.

Fig. 4.10. Applied tensile stress versus minor (transverse) and major (longitudinal) strain in [+45/-45]_2s twill 2/2 CFRP laminate tensile test.
The ultimate rupture of the specimens is approximately at 45°, as usually observed in [+45/-45]_{2s} laminates (Fig. 4.5). As the specimens were loaded axially and stretched upwards, the maximum strains occurred at the top of the specimens as can be observed at strain levels at points 1 in Fig 4.10. However, the strains in central parts such as point 2 were also high, which might be due to more compliant regions such as matrix, or yarns aligned across the loading directions.

### 4.4.2 Determination of in-plane shear properties

The tension tests of [+45/-45]_{2s} laminates provide an indirect measure of the in-plane shear stress-strain response in the material coordinate system. The in-plane shear response was obtained from tests with off-axis specimens by transformation of the applied stress and strain from the global coordinate system into the material coordinate system as shown in Fig. 4.11. Here xx coincides with the loading direction of the specimen, whereas axes 1 and 2 are along the fibre direction and perpendicular to it, respectively, in the plane of the ply. The stress and strain transformation relations in the elastic region, similar to those presented in [161] are

\[
\sigma_{11} = \sigma_{xx} \cos^2 \theta \tag{4.4}
\]

\[
\sigma_{22} = \sigma_{xx} \sin^2 \theta \tag{4.5}
\]

\[
\tau_{12} = -\sigma_{xx} \sin \theta \cos \theta \tag{4.6}
\]

\[
\varepsilon_{11} = \varepsilon_{xx} \cos^2 \theta + \varepsilon_{yy} \sin^2 \theta + 0.5 \gamma_{xy} \sin 2\theta \tag{4.7}
\]

\[
\varepsilon_{22} = \varepsilon_{xx} \sin^2 \theta + \varepsilon_{yy} \cos^2 \theta - 0.5 \gamma_{xy} \sin 2\theta \tag{4.8}
\]

\[
\gamma_{12} = (\varepsilon_{yy} - \varepsilon_{xx}) \sin 2\theta + \gamma_{xy} \cos 2\theta \tag{4.9}
\]

Here, \(\sigma_{xx}\) and \(\varepsilon_{xx}\) are the axial components of the stress and strain tensors, respectively, \(\sigma_{yy}\) and \(\varepsilon_{yy}\) are the transverse components of the stress and strain tensors, respectively, while \(\tau_{12}\) and \(\gamma_{12}\) represent the shear components of the stress and strain tensors in the material coordinate system.

For the 45° off-axis specimens, the elastic part of shear strain can be determined using Eq. (4.9), which is given as:

\[
\gamma_{12} = \varepsilon_{yy} - \varepsilon_{xx} \tag{4.10}
\]
Similarly, the maximum in-plane shear stress from Eq. (4.6) for 45° off-axis laminate is given as:

$$\tau_{12} = -\frac{\sigma_{xx}}{2} \quad (4.11)$$

The in-plane shear modulus $G_{12}$ can be calculated by dividing the shear stress Eq. (4.11) by the engineering shear strain Eq. (4.10) in the elastic region of the shear stress-strain curve:

$$G_{12} = \frac{\tau_{12}}{\gamma_{12}} \quad (4.12)$$

According to the definition of Poisson's ratio, it is calculated using the following formula

$$\nu = \frac{-\epsilon_{yy}}{\epsilon_{xx}} \quad \text{(Loading in the xx direction)} \quad (4.13)$$

Fig. 4.11. Specimen configuration of the off-axis tensile test

The CFRP in-plane shear modulus $G_{12}$ was determined from the linear shear stress–strain part of Fig 4.7 at strains in the range of 5000 – 10,000 µε (i.e.0.5 - 1%); which is the slope of the curve represented by Eq. 4.12. The Poisson’s ratio $\nu_{12}$, defined by Eq. 4.13 as the ratio of the minor to major strain components, was determined from the strain plots of Fig 4.10 in the linear elastic range. The in-plane shear strength was determined using Eq. 4.11 based on the maximum value at 45°. All the properties are listed in Table 4.3. The Poisson’s ratio is almost negligible because of the nature of fabric materials where the tows are identical in both warp and weft directions and the tow counts are also identical. The average values of in-plane shear properties obtained from the tensile tests are close to those reported in [39, 162] for a similar woven CFRP material.
Magnitudes of major (axial) and minor (transverse) strains for specific regions of CFRP specimens were almost the same indicating that the specimen failure was due to large shear deformations (Fig. 4.10). Full-field strain measurements on the surfaces of CFRP woven composite samples under tension proved to be a useful tool in studying their mechanical behaviour at the macro-scale, providing in-plane shear data of the composite. The next section describes the behaviour of CFRP laminates under quasi-static bending loads.

<table>
<thead>
<tr>
<th>Specimen type</th>
<th>In-plane shear modulus $G_{12}$ (GPa)</th>
<th>Poisson’s ratio $\nu_{12}$</th>
<th>In-plane shear strength $\tau_{12}$ (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CFRP</td>
<td>3.8±0.2</td>
<td>0.05±0.014</td>
<td>120±8</td>
</tr>
</tbody>
</table>

4.5 Bending tests

Bending tests are extensively used to determine mechanical properties of composite materials. The objective of testing is to work out the ultimate flexural stress, flexural strain and flexural modulus of woven CFRP composites under large-deflection bending and investigate the effect of varying deformation rates on these values. As compared to tensile tests, a complex stress state is experienced by composites in flexural tests, so that tensile, compressive and shear stresses may all be present simultaneously in a test specimen. However, because of the complex stress state in the specimen, it is typically not possible to relate directly the flexural properties obtained to the fundamental tensile, compressive, and shear properties of the material [90]. The main difficulty in flexure tests is connected with processing of the test results. Analysis of flexure test results for anisotropic materials is not as simple and obvious as in the uniaxial tensile test. Still, this loading type is particularly relevant as sports products are usually subjected to flexural loading in service. Due to the important influence of shear effects on displacements, large span-to-thickness ratios are used in to eliminate these effects.

In composite laminates, the stiffness properties of each ply of a laminate can be different because of different fibre orientations. Therefore, laminated beam theory
yields in a discontinuous distribution of normal stresses across the beam thickness, even though the strain variation is linear. The stress variation across a single ply thickness is linear, but discontinuous at boundaries of plies in a laminate [161]. As in bending tests of composites, the maximum flexural stresses that occur at the top and bottom plies of the specimen, are determined, for which the homogeneous beam theory is a reasonable choice, as employed in flexure test standards for composites [152, 163]. Also the longitudinal moduli are almost similar in the CFRP woven laminates, which makes its behaviour like UD composites, which can be studied by the simple beam theory. The most popular three-point bending test configuration for a homogeneous material is shown schematically in Fig. 4.12a. In these tests, a flat rectangular specimen is simply supported at the two ends and is loaded by a central load. Variation of flexural and shear stresses across the beam thickness induced in the specimen of a homogeneous isotropic material as a result of the applied bending force is shown in Fig. 4.12b. The simple beam theory suggests that the applied bending moment is balanced by a linear distribution of normal stress, $\sigma_x$, in Fig. 4.12b. Here, the top layer of the beam is in axial compression while the bottom surface is in tension. The neutral axis is located at the mid-plane where zero flexural stress is experienced. Conversely, the interlaminar shear stress $\tau_{xz}$ is maximum at the mid-plane and demonstrates a parabolic distribution vanishing at the free surfaces (Fig. 4.12b). In 3-point flexure,

![Figure 4.12](image.png)

**Fig. 4.12.** (a) Three-point flexure loading configuration; (b) Distribution of bending and shear stresses across thickness of a homogeneous rectangular cross-section beam subjected to three-point flexure

the shear stress is constant along the length of the beam, and directly proportional to the applied force $F$. However, in addition to being directly proportional to $F$, the
flexural stresses vary linearly with position along the length of the beam, and vanish at each end support and reach a maximum at the centre. Thus, the stress state is highly dependent on the support span length-to-specimen thickness ratio (S/h). Beams with small S/h ratios are dominated by shear whereas beams with long spans usually fail in tension or compression. Also, the concentrated load is applied at the point of maximum compressive stress in the beam, often inducing a local stress concentration. Thus, the composite beam may fail in compression at the mid-span loading point. Such stress concentrations are also observed at the support locations studied in detail in Chapter 6.

During testing, load data was recorded, from which several variables were determined to characterise the test materials. From the simple beam theory, the tensile and compressive flexural stresses at the surfaces of specimens, where the bending moment was maximum, was found using:

\[ \sigma_x = \frac{M_m y}{I} \]  

(4.14)

where \( M_m = \frac{F_m L}{4} \) is the maximum bending moment at the beam mid-span, \( y = \frac{h}{2} \) is the distance from the neutral axis to beam surface, and, \( I = \frac{bh^3}{12} \) is the moment of inertia of a beam of rectangular cross section, with \( b \) being the beam width. Substituting these values into the flexural stress Eq. (4.14) gives:

\[ \sigma_x = \frac{3F_m S}{2bh^2} \]  

(4.15)

where \( F_m \) is the maximum flexural load, \( S \) is the span between the supports and \( h \) is the specimen thickness. This formula is for the simple beam theory, where deflection is assumed to be small compared with the span. When deflections are large, as is frequently the case for long beams, Eq. (4.14) for stress is corrected, as discussed in ASTM D 790 [163]. The standard recommends that if support span-to-thickness ratios greater than 16 are used such that deflections in excess of 10% of the support span occur, the stress in the outer layer of the specimen for a simple beam can be reasonably approximated with the following equation:

\[ \sigma_{corr} = \sigma_x \left[ 1 + 6 \left( \frac{\delta}{S} \right)^2 - 4 \left( \frac{h}{S} \right) \left( \frac{\delta}{S} \right) \right] \]  

(4.16)
where $\delta$ is the maximum deflection of the specimen. This equation also takes into account the additional bending moment due to the horizontal components of the support reactions in case of large-deflection bending.

Similarly, the flexural strain at the point of maximum flexural load is given by

$$
\varepsilon_f = \frac{6\delta h}{S^2}
$$

(4.17)

Finally, the flexural modulus for the specimen can be calculated using the following equation:

$$
E_f = \frac{\sigma_2 - \sigma_1}{\varepsilon_{f2} - \varepsilon_{f1}}
$$

(4.18)

where $\sigma_1$ is the flexural stress, in MPa, measured at deflection $\delta_1$ and $\sigma_2$ is the flexural stress, in MPa, measured at deflection $\delta_2$.

Apart from the three-point flexural test, four-point test configurations are also used in order to obtain bending properties of composite laminates. However, there is no clear advantage of one test over the other, although the two test configurations differ from each other with respect to the state of stress and may give slightly different results. In a 3-point bending test, the bending moment on the beam varies linearly from zero at the supports to a maximum value at the specimen’s centre (along the span). The shear force and the resulting interlaminar shear stress are uniform all along its span between the supports. This may develop interlaminar shear failure [161]. To avoid these shear failures at the mid-plane rather than tensile or compressive failures at the beam surfaces, higher span length-to-thickness ratios of specimens are selected. In four-point bending tests, the bending moment increases linearly from zero at the supports to a maximum value under the load, this remains constant between the loads. To attain the same maximum bending moment as that of three-point bending, this configuration requires twice the testing machine force. In four-point bending, the shear force and the resulting interlaminar shear stress between the loads vanish, and, hence, this portion of the beam is in a pure bending state. Thus, from the stress state point of view, the four-point bending test is more desirable whereas the three-point bending test is easier to conduct. Further, three-point flexure is the most common, perhaps only because it requires the simplest test setup. Since the focus of this study is to characterise the material behaviour under the complex state of flexural
as well as shear stresses as experienced in sports products, three-point bending tests are a reasonable choice.

Fig. 4.13. CFRP specimen in three-point bending test fixture of Instron 5569

4.5.1 Testing procedure

Quasi-static flexural tests were used to determine the ultimate flexural strength, flexural strain and flexural modulus of the CFRP materials under three-point bending conditions. Two types of specimens of 80 mm length and 25 mm width were prepared with thicknesses of 1.5 mm and 1mm to observe the effect of thickness on flexural properties of CFRP in both warp and weft directions. The loading span between the supports was kept to 40 mm in all the tests. The flexural tests were carried out at indenter speed of 100 mm/min equivalent to strain rate 0.0417/s, using the Instron 5569 machine in accordance with the ISO 178 standard [152]. The three-point bending test set-up is shown in Fig. 4.13. Samples for each orientation were tested under large-deflection bending until their ultimate fracture; the tests were performed in the displacement-control regime. Five specimens were tested for each material type to ensure that results were statistically significant. The data was recorded by the machine load cell in the form of applied
load and deflection of the specimen. Since in our study, the support span-to-thickness ratios were 26.7 and 40 for specimens with thickness of 1.5 mm and 1 mm, respectively, and also the maximum deflection $\delta$ was greater than 10% of span $S$, so Eq. (4.16) was used to determine flexural stress. The flexural strain and modulus were determined from Eq. 4.17 and 4.18, respectively.

4.5.2 Results and discussion

The flexural stress-strain diagrams for each orientation of CFRP specimens of 1.5 mm and 1.0 mm thickness are shown in Figs. 4.14 and 4.15, respectively. In both types of specimens, flexural tests of warp $[0^\circ,90^\circ]_3s$ and weft $[90^\circ,0^\circ]_3s$ on-axis samples exhibited a quasi-brittle response. The reason for the material behaviour is the same as explained in Section 4.3, in the on-axis laminates, the applied load is carried by the fibres, which are stronger but brittle. Hence, the stress-strain curves are almost linear till ultimate failure which is represented by a sudden drop of stress level in Figs. 4.14 and 4.15. For each type of specimens, the flexure tests resulted in approximately the same elastic modulus and ultimate tensile strengths in both warp and weft directions as shown in Table 4.4. This similarity was due to the symmetry of fibres in both warp and weft directions in both types of samples. Stiffness degradation due to internal cracks and delamination occurred in the warp samples at about 90% of the ultimate load, whereas the weft samples showed no stiffness reduction before ultimate fracture (Figs. 4.14 and 4.15). Although the samples underwent matrix cracking and interlaminar damage before the structure lost its load-carrying capacity, the development of such inter-ply and intra-ply damage mechanisms was not reflected in the stress-strain diagram, i.e. the effect of these usually hidden and barely visible damage mechanisms was small. The elastic flexural moduli such as $E_{11}$ and $E_{22}$ were calculated from the mechanical tests of both types of specimens in the warp and weft directions, respectively, at strains in the range of 5000 $\mu$ε to 10000 $\mu$ε (i.e. 0.5% to 1%). The elastic moduli as well as material strengths obtained from flexural tests are listed in Table 4.4. In the flexural tests of both 1.5 mm and 1.0 mm thick specimens, the weft specimens showed earlier fracture due to a lesser stiffness in this direction (Figs. 4.14 and 4.15). This may also be due to the fact that the number of yarns in the warp direction is slightly more than the weft, since they are aligned to the longitudinal plane resisting the bending more than the weft. Thus, the flexural strength as well
Fig. 4.14. Stress-strain diagram from flexural tests of 1.5 mm thick twill 2/2 CFRP laminates

Fig. 4.15. Stress-strain diagram from flexural tests of 1.0 mm thick twill 2/2 CFRP laminates
as flexural modulus of warp specimens is also more than that of the weft specimens (Table 4.4). Here, the deflection is also somewhat higher for the warp specimens. The tests of 1.5 mm thick specimens resulted in flexural moduli of 44.1 GPa and 43.6 GPa, and ultimate flexural strengths of 826 MPa and 816 MPa, for warp and weft direction, respectively. The respective flexural moduli of 42.5 GPa and 41.5 GPa, and strengths of 745 MPa and 735 MPa in the warp and weft directions, respectively, reported by the manufacturer [150] were lower than those obtained from the tests.

Table 4.4. Flexural test data for CFRP samples

<table>
<thead>
<tr>
<th>Thickness (mm)</th>
<th>Orientation</th>
<th>Flexural modulus (GPa)</th>
<th>Ultimate flexural strength (MPa)</th>
<th>Maximum deflection (mm)</th>
<th>Elongation (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.5</td>
<td>Warp, [0°,90°]_3s</td>
<td>44.1 ± 1.3</td>
<td>826.0 ± 22</td>
<td>3.64 ± 0.22</td>
<td>1.96 ± 0.08</td>
</tr>
<tr>
<td></td>
<td>Weft, [90°,0°]_3s</td>
<td>43.6 ± 1.2</td>
<td>816.0 ± 20</td>
<td>3.6 ± 0.2</td>
<td>1.90 ± 0.1</td>
</tr>
<tr>
<td>1.0</td>
<td>Warp, [0°,90°]_2s</td>
<td>55.0 ± 1.5</td>
<td>993 ± 25</td>
<td>5.1 ± 0.3</td>
<td>1.85 ± 0.07</td>
</tr>
<tr>
<td></td>
<td>Weft, [90°,0°]_2s</td>
<td>52.0 ± 1.4</td>
<td>885 ± 23</td>
<td>4.75 ± 0.26</td>
<td>1.76 ± 0.06</td>
</tr>
</tbody>
</table>

Although the tests were conducted under large-deflection conditions, however, this nonlinearity was not reflected in the stress-strain curves. In thin laminates, large-deflection bending causes stress stiffening, which is reflected by the stress-strain curve. According to Timoshenko and Woinowsky-Krieger [164], if the ratio of maximum deflection to plate thickness becomes greater than 0.4 for clamped or 1.0 for simply supported isotropic plates, the effect of membrane stretching (stress stiffening) should be considered. These stress stiffening effects are also responsible for enhancement in ultimate flexural strength of the thinner 1 mm thick laminate (Table 4.4). A typical failure pattern of the 0° CFRP specimen in Fig. 4.16 shows that damage is distributed through the width at the centre of laminate. The character of ultimate fracture demonstrates that fibres of the specimen’s top
surface, which are in compression, remain intact (Fig. 4.16a), whereas fibres of the bottom surface, experiencing tension, are fractured (Fig. 4.16b).

Fig. 4.16. Fracture of 1.5 mm thick warp CFRP specimen in three-point bending test: (a) top surface; (b) bottom surface

The average flexural strengths of 826 MPa and 816 MPa for 1.5 mm thick specimens in the warp and weft directions, respectively, were higher than their average tensile strengths of 762 MPa and 764 MPa as listed in Table 4.2. Several investigators have shown that, in general, the flexural strength is greater than the tensile strength of the material from which it is made [88]. This behaviour is explained as a consequence of the nonuniform stress distribution through the thickness of the bending specimen. In the flexural specimen, the maximum stress occurred only in the outermost fibres and, thus, the probability of a flaw to be located within this region of higher stress and thus initiating failure, was lower than it would be for the entire specimen’s cross section in a tension specimen. This can also be explained in terms of the statistical variation in fibre strength which dominates failure initiation. In the tension tests, all the fibres aligned parallel to the loading direction experience a uniform stress, and so the weakest fails first. In the bending tests, only the back face experiences the maximum tensile stress and the weakest fibres are not necessarily at the back face [88]. Similarly, the elongation
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of 1.96\% and 1.9\% of the warp and weft specimens, respectively, in bending is more their respective values of 1.55 \% and 1.56 \%, listed in Table 4.2, in tension. This may be due to the large-deflection caused by flexural loads. This behaviour results in higher values of Young’s moduli obtained from tensile than bending tests (Table 4.2). Further, usually in textile laminates, relative sliding of tows and inter-tow friction results in a higher strain to fracture than UD & MD composites.

The effect of thickness on the flexural behaviour of CFRP laminates can be observed from the comparison of data for both 1.5 mm and 1.0 mm thick specimens in Table 4.4. When the strength of a composite material is reduced with the increasing volume of the material under the same testing conditions, it is said to exhibit a size effect. Size effects in composites are usually explained on the basis of Weibull statistical theory in terms of the probability of finding larger defects when the stressed volume is greater [165]. Hence, larger specimen volumes have a higher probability of having larger defects, than smaller specimen volumes. Although the variation in the thickness of CFRP specimens was small, still such effect was manifested by them as presented in Table 4.4. Here, by reducing the thickness from 1.5 mm to 1.0 mm, the flexural strength of the material is increased. Wisnom and Atkinson [166] carried out tensile tests with different gauge lengths and flexural tests with fully scaled specimens of unidirectional glass fibre/epoxy and found similar size effects in each case.

The mechanical response of composite materials is sensitive to the rate at which they are loaded, so for the effective use of composites, their response under different strain rates should be clearly understood. Generally, material characterization under high strain rate loading is carried out with the Split Hopkinson Pressure Bar or Kolsky system. However, the sports products experience loadings at low strain rates; thus, the behaviour of CFRP on-axis laminates is investigated at various speeds of load application to determine their rate dependency in the quasi-static range. This objective was fulfilled by testing 1.0 mm thick specimens of CFRP under large-deflection bending in the warp and weft directions. The specimen’s dimensions were the same as in the previous tests. Both types of specimens were tested at the indenter speeds of 100, 200, 300 and 400 mm/min equivalent to strain rates of 0.0417s\(^{-1}\), 0.0834 s\(^{-1}\), 0.125 s\(^{-1}\) and 0.167 s\(^{-1}\), respectively. Results for the warp and weft specimens are shown
Fig. 4.17. Flexural behaviour of 1.0 mm thick warp specimens at various test speeds

Fig. 4.18. Flexural behaviour of 1.0 mm thick weft specimens at various test speeds
in Figs. 4.17 and 4.18, respectively. Flexural stress-strain plots for both type of specimens showed no rate dependent behaviour. Usually, the strain-rate sensitivity of woven composites is due to inertia and strain rate dependency of the epoxy matrix [62]. Carbon fabric-reinforced composites are essentially strain rate-independent in the fibre-dominated modes such as on-axis laminates $[0^\circ,90^\circ]_{2s}$ and $[90^\circ,0^\circ]_{2s}$ subjected to in-plane tension, compression and bending in the low-velocity (low-strain-rate) regimes [89]. The tests were conducted on the on-axis laminates, which showed strain rate insensitive behaviour. However, woven CFRP exhibits strain-rate dependency in the matrix-dominated modes such as the off-axis laminates $[+45/-45]_{2s}$ subjected to in-plane shear [62, 89].

4.6 Analysis of damage in tested composites

Damage mechanisms such as matrix cracking, delamination and fibre fracture in usually occur inside the composite laminates and are barely visible. These damage mechanisms need examination for reliable and accurate modelling and analysis of composite laminates. In this study, the damage and fracture of composites was first investigated by using destructive evaluation technique such as optical microscopy of cross sections. Since, optical microscopy is limited to provide information about the internal damage of the laminates, a recent technique such as X-ray micro computed tomography (Micro-CT) was used to visualise 3D internal deformation and damage behaviours of tested composite laminates. Since the main focus of the study is to investigate the CFRP behavior under large-deflection bending, damage analysis of some of the specimens tested under these conditions is carried out and presented in this section. Specimens of 1.0 mm thickness in the warp direction tested under bending conditions were used to characterise the type and location of damage in the specimen by employing both types of the techniques.

4.6.1 Optical microscopic analysis of damage

Microscopic analysis of composite specimens is a suitable technique for evaluation of damage in woven-fabric laminates. In order to visualize the type and location of damage, optical microscopic analysis allows for the study of the comprehensive damage behaviour of the composite specimen. In this work, an OLYMPUS BX-60M microscope was used to capture the images of the through-thickness polished edge of the specimen fractured in bending to observe various
damage mechanisms and ultimate fracture. Image-Pro Plus software was used for the analysis of the captured microscopic images.

Figure 4.19 presents the damage modes observed at various locations (Fig.4.19a) in CFRP specimen fractured in bending. Major damage modes observed at the beam’s fractured location A were inter-ply delamination between the warp tows, weft cracking and fracture of fibers in the warp tow that was responsible for the final catastrophic failure, as shown in Fig. 4.19b. The warp tows in three bottom plies subjected to tensile load were completely fractured. At low stressed regions away from the beam centre (indenter), the major damage observed in the twill weave laminate was longitudinal weft tow cracking and matrix cracks. At the beam location B, Fig. 4.19c shows weft tow longitudinal cracking (intra-ply delamination).

Fig. 4.19. Damage at various locations of CFRP laminate in bending: (a) schematic of fractured specimen; (b) fracture mechanisms at location A; (c) cracking in weft tows (intra-ply delamination) connected by a transverse matrix crack at location B; (d) inter-ply delamination between tows near location B; (e) weft tow cracking (intra-ply delamination) at location C
that traversed transversely through the matrix to propagate within another weft tow; thus, delamination cracks interacted through this mechanism. Here, it can be observed that damage initiation occurred at the weft tow edge (crimp location) and then propagated into the weft tow accompanied by inter-ply transverse matrix cracking. Interlaminar delamination along the specimen’s longitudinal axis due to high in-plane shear stress can also be observed in Fig. 4.19d near location B. Intra-ply delamination in the weft tow can also be clearly seen in the microscopic image of Fig. 4.19e at location C of the beam. Unlike the transverse cracking observed in weft tows of woven laminates under tensile loading studied in [35, 39], longitudinal cracking of weft tows along the specimen axis was observed under flexural load in this study. From the microscopic analysis, it can be asserted that weft tow longitudinal cracking is the predominant damage mechanism at low stressed regions before the final failure of the laminate. Inter-ply delamination and ply fracture were the prominent damage modes at the maximum loaded regions in the beam centre.

Fig. 4.20. Damage on compression side of CFRP laminate in bending at location A in Fig. 4.19(a): (a) fibre buckling in warp tow of top ply of specimen; (b) fibre fracture due to buckling (magnified view)

Interlacing of tows in the woven laminates reduce the critical length of tows subjected to axial compression, reducing thus the probability of fibre buckling as compared to UD laminates. However, such a failure mode still occurs under high compressive load in the woven laminates. In our large-deflection bending tests, the top ply below the indenter was subjected to compression, which resulted in fibre buckling in the warp tow aligned to the beam axis as shown in Fig. 4.20a.
This failure mechanism can be seen in a form of fracture of fibres in short pieces in the compressed side of the specimen. Apparently, the fibres near to the specimen neutral axis with less compressive forces were less fractured as can be seen in Fig. 4.20b. The damage mechanisms observed in the microscopic analysis were: (i) matrix fracture and delamination at the interface between orthogonal tows, (ii) inter-ply delamination, (iii) longitudinal cracking within the weft tow (intra-ply delamination), (iv) tensile fibre fracture in warp tows and (v) fibre buckling in warp tow under compression.

4.6.2 Micro computed tomography (Micro-CT) analysis of damage

Micro computed tomography (Micro-CT) - a non-destructive damage evaluation technique - was used to study a 3D deformation and damage behaviour in the tested CFRP laminates at microstructural level. Micro-CT is usually used for imaging of material’s internal structure based on X-ray absorption, which is related to the material density [167].

In this study, X-ray Micro CT measurements were performed using an XT H 225 X-ray scanner machine. The system consists of an X-ray detector and an electronic X-ray source, creating two-dimensional images of cross-sections of the object. The source is a sealed X-ray tube operating at 25–225 kV with a 3 μm spot size. The specimen was positioned by an object manipulator with two translations and one rotation, rotating and raising/lowering the sample to a specific region of interest for adjustment of the sample magnification and acquisition of tomographic data. As denser materials absorb more X-rays than voids and air, this attenuation contrast allows detection and characterization of cracks and flaws in the tomographic images. High scan resolution is required to obtain maximum internal details of damage in the composite laminate. As the resolution is increased, the field of view of the sample is reduced. However, samples must remain within the field of view to obtain the radiographs of the region of interest; thus, there is a trade-off. A small sample with dimensions of 30 mm x 7 mm was prepared from the damaged region of the tested CFRP laminate to meet these requirements. The data for the sample (tomogram) was collected at 58 kV and 80 μA. For each tomogram, 3016 radiographs were acquired with an exposure time of 1000 ms. Transmission X-ray images were acquired from 3600 rotation views over 360° of rotation (0.1° rotation step) for 3D reconstruction. The data was reconstructed
using CT-Pro software, and the resulting 3D volumes were analysed using the commercial software package VG Studio Max 2.2 to segment and highlight regions of interest. These settings resulted in tomographs with resolution of 11 µm. Visualization of the data was presented in two different ways to provide a better understanding of the damage modes in the material. They include 3D volumes of the damaged specimen and cross-sectional views showing the damage in one plane as shown in Figs. 4.21 and 4.22, respectively.

![Fracture](image)

**Fig. 4.21.** Ultimate fracture of 1 mm thick 2 x 2 twill CFRP laminate (resolution 11 µm): (a) damage in three-dimensional volumetric reconstruction; (b) bottom surface

The reconstructed 3D and 2D images of the ultimate transverse fracture in the 1 mm thick CFRP \([0^\circ,90^\circ]_2s\) specimen are shown in Fig. 4.21. Figure 4.22 demonstrates damage in the resin-rich regions between the plies of the sample at various locations across its width. Dark regions in images represent cracks and
Chapter 4

voids whereas brighter regions - the virgin material. Cracks and damage were resolved based on the difference in the X-ray absorption between the carbon/epoxy material and air. Realization of damage mechanisms at the outer edge, 50% and 75% of the sample width is shown in Figs. 4.22 a, b, and c, respectively. Apparently before ultimate fracture, the laminate exhibited matrix cracking, and then delamination in the form of tow debondings. Matrix cracks developed in the weak resin-rich pockets around the tows. Inter-ply delamination and intra-ply delamination (tow debonding) can also be observed. Such delaminations normally appeared near the matrix cracks area, which suggested that formation of the cracks initiated delamination. In the fibre-rich regions, the damage was apparently associated with debonding of the fibre/matrix interface. The examination of internal structure showed that almost every ply was delaminated at the time of fabric fracture. All the tomographs showed that delamination and, subsequently, transverse ply fracture are the prominent failure modes.

Fig. 4.22. Damage mechanisms at various locations across width of sample (resolution 11 µm): (a) edge; (b) 50% of width; (c) 75% of width

Both types of microstructural analysis techniques demonstrated almost similar types of damage mechanisms in the woven CFRP laminates such as matrix
cracking, delamination, tow debonding and ply fracture. Due to low resolution, the Micro-CT scans did not provide any information about the fibre buckling mode. However, both of the methods were limited in their analysis of the full-scale tested specimens due to their fields of views restricting the size of the samples. To obtain high resolution images to observe the microstructural damage modes, a small sample is needed, while for a large size sample, the information of damage at the microstructural level is lost. Therefore, the damage analyses were used to identify the location and type of various damage modes to model accurately their behavior in FE models. Comparison of the experimentally investigated damage area with that obtained from the numerical simulations was limited by the scanning of small size sample.

4.7 Conclusions

The mechanical behaviour of woven CFRP laminates was characterised by performing tensile, in-plane shear and large-deflection flexural tests under quasi-static conditions. On-axis specimens of [0°,90°]2s and [90°,0°]2s in the warp and weft directions, respectively, were tested in tension and flexure whereas shear tensile test were carried out on off-axis laminates. The on-axis laminates exhibited quasi-brittle linear behaviour before their ultimate fracture in both tensile and flexural tests. Further, the on-axis specimens demonstrated a strain-rate insensitive behaviour while tested at various indenter speeds under large-deflection bending. The off-axis [+45/-45]2s laminates showed a nonlinear behaviour due to matrix cracking and fibre trellising in the in-plane tensile shear tests. The in-plane shear properties were determined in the elastic region of the obtained data. In these tests, DIC proved to be a very useful non-contact technique to acquire full-filed strain data. Tensile tests on 1.5 mm thick specimens were conducted to compare the obtained material data with those from the bending tests. A substantial difference was found between the material tensile and flexural properties due to the size effects and stress stiffening of thin laminates. Therefore, application of tensile material properties to study the bending behaviour of composites is not recommended. The obtained material data can be used for numerical simulations of the CFRP laminates under various types of loading.

Damage in woven CFRP composites under large-deflection bending was studied using optical microscopy and Micro-CT techniques. Microscopic analysis of
damage at various locations in the fractured specimen asserted that the low stressed regions demonstrated weft yarn longitudinal cracking, whereas the high stressed regions underwent inter-ply delamination and ply fracture. The internal damage mechanisms in the tested specimens were also investigated by means of X-ray Micro-CT imaging. Considerable matrix cracking leading to tow debonding and delamination were observed along with tow fibre fracture. Among all the observed inter-ply and intra-ply damage modes, delamination and fabric fracture are critical to the load-bearing capability of the CFRP laminates. Therefore, these two types of damage modes are incorporated in our FE models of damaged CFRP specimens under large-deflection bending. In the damage analysis, the identification of type and location of damage mode served as an input to the FE models.
5 Chapter 5: Mechanical Behaviour of Fabric-reinforced Composites under Large-deflection Dynamic Bending

5.1 Introduction

Research of low-velocity impact loading of composites is aimed at reducing the degree of damage in order to improve their damage tolerance in various applications. A number of approaches have been used to improve the impact damage resistance and tolerance of composite materials. These include control of fibre/matrix interfacial adhesion, laminate design and through the thickness reinforcement. Regarding the laminate design option, woven-fabric composites have been widely used in impact-prone structures as they offer better resistance to impact damage than unidirectional (UD) tape laminates due to their interlacing (weaving) tow architecture limiting the damage growth, particularly delamination between layers [168]. Due to these properties, woven laminates have been incorporated in the design of sports products that could be subjected to bending caused by impact loading. For example, sports products such as athletic footwear insoles [150, 169] and a custom-built Flex-Foot Cheetah blade for elite amputee athletes [170] made of CFRP are subjected to large-deflection bending caused by low-velocity impacts during running. In sports products, kinematics, velocity and energy of dynamic events are significantly different to traditional impact events experienced by aerospace structures; hence, it should be properly studied. However, these events are also characterised by realisation of multiple damage modes in composite laminates due to their microstructural heterogeneity and anisotropy. The damage mechanisms typically caused by out-of-plane impact loads are matrix cracking, fibre breakage, delamination between adjacent plies and, ultimately, fabric fracture in the composite structures [67]. Impact damage and, in particular, delamination cause a significant decrease in the material’s in-plane compressive strength and stiffness. Such internal damage mechanisms that often cannot be detected by visual inspection degrade the load-bearing capacity of the structures. Therefore, it is important to study the damage suffered by the composites under impact bending conditions. The low-velocity impact response of woven-fabric composite laminates has been extensively studied in the literature. Iannucci and Willows [171] investigated the dynamic behaviour and the
subsequent damage in woven carbon composites under impact loading at various
energy levels using experimental and numerical techniques. Atas and Sayman [2]
studied experimentally the impact response of woven fabric composite plates
under various impact energies. Evci and Gülgeç [172] investigated the impact
response of UD and woven glass fabric composites and found that woven
laminates were superior than UD ones in resisting damage propagation. Reyes
and Sharma [8] studied experimentally and numerically a low-velocity impact
damage behaviour of woven GFRP laminates under various levels of impact
energies. However, majority of these studies are dedicated to the impact
behaviour of composites tested with instrumented drop weight impact towers,
which usually cause localised damage such as penetration and perforation in
impacted laminates.

In service, the sports products are subjected to non-penetration dynamic loads
causing large-deformation bending. Therefore, the drop-weight impact tests do not
replicate the real loading scenarios of these products. Such dynamic bending
behaviour can be accurately characterised by impact tests on un-notched samples
using pendulum-type hammers. A large-deflection dynamic bending behaviour of
composite laminates caused by a pendulum-type impactor is rarely investigated.
A few studies can be found where an instrumented impactor was used by
Silberschmidt et al. [173] and Casas-Rodriguez et al. [174, 175] to study the
behaviour of adhesively bonded CFRP joints under repeated impacts, however,
the loading mode there was tensile. In the previous chapter, the composites’
mechanical properties were quantified and analysed under large-deflection quasi-
static bending and tensile loads. In this chapter, the dynamic behaviour of the
woven laminates is characterised in impact tests employing pendulum-type
hammer. The tests are performed on un-notched specimens at various energy
levels up to the fracture of the specimens. Damage induced by dynamic bending in
CFRP laminates is investigated using the Micro-CT technique.

5.2 Dynamic bending tests

The energy absorbing capability and dynamic strength of sports products exposed
to suddenly applied forces can be quantified with impact tests of the materials; the
required data about deformation and damage of the materials during dynamic
loading cannot be determined in quasi-static regimes. Dynamic bending tests are
carried out to characterise the behaviour of composite laminates under low-velocity impact regimes as experienced by the sports products in their service.

5.2.1 Test principle and experimental equipment

Dynamic testing is usually performed with drop-weight, Charpy, Izod or tensile-impact test machines. An Izod test system incorporates a swinging pendulum (hammer) that impacts a notched specimen fixed in a cantilever-beam position with the notch facing the hammer. In this study, the test principle is similar to that of Izod test except an un-notched specimen is used. In Izod tests the specimen is fractured by a single oscillation of the hammer whereas in impact tests the specimen may or may not fracture depending on the impact energy level, material’s flexural rigidity and support conditions. Here, a standard pendulum-type hammer mounted in a standardised machine CEAST Resil Impactor, shown in Fig. 5.1, was used to conduct dynamic tests in order to determine impact characteristics of the woven laminates under large-deflection bending conditions. The specimen was fixed firmly in the fixture as shown in Fig. 5.2, and impact test was performed by a single swing of the pendulum hammer at a fixed distance between the specimen’s clamp and the centre line of the impact. The maximum
potential energy due to the hammer mass and drop height is converted to kinetic energy. Part of the impact energy is absorbed by the specimen and part is released due to the elastic behaviour of the materials.

![Image of specimen holding fixture for impact bending test]

**Fig. 5.2.** Specimen holding fixture for impact bending test

The absorbed energy by the specimen in impact bending causes deformation and fracture of the specimen and consists of the following components:

1. Energy required to bend the specimen;
2. Energy required to indent or deform plastically the specimen at the line of impact;
3. Energy required to initiate fracture;
4. Energy required to propagate fracture;
5. Energy consumed in friction losses;
6. Energy consumed to separate the pieces of the specimen.

The instrumented test procedure adopted in this study involves direct measurement of the impact force by using a piezoelectric load cell attached to the hammer’s striker. Even though the entire impact event takes place in milliseconds, the instrumented tests enable capturing a complete force-time curve and provide enough information to calculate deformation and, in turn, velocity and absorbed energy [176]. The Resil pendulum impactor is an example of the instrumented-test
system. The used system is equipped with a pneumatic brake, and the range of impact energy is from 1 to 25 J. It is provided with two devices: pneumatic actuator applied to the hammer brake that permits automatic braking of the hammer after impact and another pneumatic device to release the hammer (Fig. 5.1). The instrumented hammer is connected to a data acquisition system DAS 8000 and a computer. DAS 8000 is a high-speed system designed to record data during fast events, such as impact test. DAS 8000 consists of a modular structure housing the CPU board (8 bit microprocessor), CPU power supply board, display power supply board and LCD graphic display. Eight independent acquisition boards can be equipped in the main CPU board. Each one can be attributed to one signal channel with separate system of acquisition (2 MHz A/D converters) and a memory able to store up to 8000 point acquired [176]. The signal of variable intensity coming from the piezoelectric sensor is sampled by DAS 8000, then it is sent through a USB port to the DAS8WIN extended version CEAST program on the PC [176]. The sampled signal is processed through the program resulting in the final graphical data.

5.2.2 Materials and specimen preparation

The same fabric-reinforced CFRP materials having a balanced 2/2 twill weaving pattern studied under the quasi-static bending in Chapter 4 were tested under the dynamic conditions. Two types of laminate orientations such as \([0^\circ, 90^\circ]_2s\) and \([90^\circ, 0^\circ]_2s\) were selected. The laminates with \([0^\circ, 90^\circ]_2s\) and \([90^\circ, 0^\circ]_2s\) stacking orientations are usually known as the warp and weft samples because 0° and 90° represent orientations of yarns in the warp and weft directions, respectively. Un-notched rectangular specimens of 40 mm length, 25 mm width and 1.0 mm thickness were prepared from each orientation; each laminate had four layers of 0.25 mm thickness each. Two specimens were used for each orientation to ensure reproducibility of the experimental results. A detailed description of the material is given in Section 4.2.

5.2.3 Experimental procedure

Dynamic bending tests in a low-velocity range from 0.7 m/s to 1.9 m/s as experienced by the sports products were carried out according to the ASTM D4812 standard [177] using an instrumented pendulum type CEAST Resil
impactor as shown in Fig. 5.1. In the impact tests, the bottom of the specimen was fixed firmly in a machine vice as a cantilever beam (Fig. 5.2). The upper 30 mm of the specimen was hit by the striking nose of the pendulum hammer with a controlled level of initial energy, resulting in dynamic large-deflection out-of-plane bending. The distance between the fixed support and the line of contact of the hammer’s striking nose was kept at 22 mm according to the standard. In this work, a calibrated impact hammer with a mass of 0.6746 kg and a length of 0.3268 m was used. The hammer can generate an impact of maximum energy of 2 J at impact velocity of 3.46 m/s corresponding to the pendulum’s initial angle of 150° to the striking position. The magnitudes of initial impact energy and velocity can be varied by changing the initial angle of the hammer resulting in variation of the level of impact energy and, thus, peak loads. A piezoelectric force transducer was fixed rigidly to the hammer’s striking nose to capture the impact-force signal. When the pendulum hammer was released from the pre-defined initial angle, the impact with the specimen generated a change in electrical resistance of the piezoelectric sensor that was captured by the data acquisition system (DAS 8000) connected to the Resil impactor. The signal was registered with a pre-defined sampling frequency of 227 kHz, with up to 5000 data points recorded per impact test. In order to decrease the data noise, a 1 kHz filter was used. CFRP specimens were tested at various impact energy levels as shown in Table 5.1 to determine their transient response and energy absorbing capability. These energy levels were chosen to investigate the general trends in the response of thin woven laminates subjected to dynamic bending. At least, two specimens were tested for each energy level. During the impact, the impactor and the specimen react to the impact by oscillating at their natural frequency, recorded by the instrumentation. These oscillations, often called ringing effect in the force-time response, were reduced by sticking a thin layer of elastomer at the hammer’s striking nose as advised in [178]. The force-time history of the dynamic event was recorded. The rest of the dynamic test parameters can be calculated using the equations given below.

The hammer initial velocity at the moment just before the contact takes place, can be determined from the potential energy using equation:

$$v_i = \sqrt{2gh_d} \quad (5.1)$$
where \( v_i \) is the velocity at impact, \( h_d \) is the drop height, and \( g \) is the acceleration of gravity. By integrating the force history, the hammer velocity \( v(t) \), deflection, and absorbed energy \( E_a \) can be calculated, respectively, using the following equations:

\[
v(t) = v_i - \frac{1}{m} \int_0^t F(t) \, dt \tag{5.2}
\]

\[
\delta(t) = \int_0^t v(t) \, dt \tag{5.3}
\]

\[
E_a = \frac{m}{2} (v_i^2 - v^2) \tag{5.4}
\]

As mentioned earlier, variation of contact force versus time was obtained by a data acquisition system. Therefore, other impact parameters such as velocity and deflection were calculated by using real contact force–time history, based on the above equations with the assumption that the impactor is perfectly rigid.

Table 5.1. Experimental parameters of dynamic bending tests of CFRP specimens

<table>
<thead>
<tr>
<th>Sample</th>
<th>Hammer angle</th>
<th>Impact velocity (m/s)</th>
<th>Impact energy (J)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Warp/Weft</td>
<td>25°</td>
<td>0.77</td>
<td>0.1</td>
</tr>
<tr>
<td>Warp/Weft</td>
<td>36°</td>
<td>1.11</td>
<td>0.2</td>
</tr>
<tr>
<td>Warp/Weft</td>
<td>44°</td>
<td>1.34</td>
<td>0.3</td>
</tr>
<tr>
<td>Warp/Weft</td>
<td>51°</td>
<td>1.54</td>
<td>0.4</td>
</tr>
<tr>
<td>Warp/Weft</td>
<td>58°</td>
<td>1.74</td>
<td>0.5</td>
</tr>
<tr>
<td>Warp/Weft</td>
<td>64°</td>
<td>1.9</td>
<td>0.6</td>
</tr>
</tbody>
</table>

5.3 Results and discussion

The low-velocity dynamic bending response of twill woven CFRP laminates in the warp and weft directions was evaluated using an instrumented impact testing machine at energy levels ranging from 0.1–0.6 J. The transient response of the sample includes the velocity, deflection, load and energy as functions of time. Diagrams of evolution of load with time and load–deflection graphs were plotted for representative samples at each energy level. Key impact parameters such as peak load, time to peak load, deflection at peak load were evaluated by the data.
acquisition system. Results of experimental tests obtained for the dynamic bending behaviour of woven CFRP laminates are presented in this section.

5.3.1 Force-time response

The impact force-time responses of CFRP warp and weft specimens tested at six different energy levels are presented in Figs. 5.3 and 5.4, respectively. From these plots, it is clear that for two types of laminate configurations, the slope of the force-time plot increases with increase in impact energy. The peak load increased as the energy level increased. The initial loading parts of the curves were smooth as the oscillations due to elastic vibration induced by initial contact between the impactor and the specimen were damped by the thin elastomer layer attached to the hammer’s striking nose. There is a linear increase of force with the time at the start of loading, indicating a purely elastic response of the specimen; no damage is expected to occur if the impact loading is terminated at this stage. After reaching the load peaks, the impactor bounced back and the load reduced to zero. The loading and unloading portions of the curves for warp specimens up to 0.4 J and weft specimens up to 0.3 J, had a nearly symmetric parabolic shape. This indicates that the respective stages during the contact duration were almost the same and very little damage occurred. At most, there might be a low-level dent or matrix cracking at the location of impact due to contact forces.

As the impact energy was increased to 0.5 J and 0.4 J for warp and weft specimens, respectively, fluctuations in the force-time history could be observed before the peak load was reached. These fluctuations associated with load drops are designated as $F_{di}$ in Figs. 5.3 and 5.4, which represent the internal damage initiation such as delamination at low incident energy with an associated loss in the laminate stiffness [8, 172, 179]. As the load continued increasing, the damage grew and multiple delaminations at different ply interfaces could occur. The curves became unsymmetric at 0.5 J with regard to their peaks as can be seen in Figs. 5.3 and 5.4. In these tests, fabric fracture occurred in both types of specimens as the impact energy was increased to 0.6 J. The force-time curves showed more oscillations due to significant damage after the damage threshold $F_{di}$ at higher impact energy. This was followed by the ultimate ply fracture. The final fabric failure was represented by a sudden drop in contact force implying a momentary loss of contact between the impactor and specimen due to tensile fracture of fibres.
Fig. 5.3. Force-time response of CFRP warp laminates at various impact energies

Fig. 5.4. Force-time response of CFRP weft laminates at various impact energies
at the front face (impacted tension side) of the specimen. After the ply fracture the contact was lost between the hammer and the specimen. The load did not drop to zero in warp specimens, as the distal plies in compression were not fractured causing a residual load of about 22 N. However, in the weft specimen, the load dropped to zero representing the complete fracture of the laminate. A typical failure of warp specimen at 0.6 J impact energy is shown in Fig. 5.5. The specimen showed no penetration damage on the front surface at the impact location (Fig. 5.5a). The character of ultimate fracture at the bending location demonstrated that tows of the specimen’s front surface (impact side), which were in tension, were fractured (Fig. 5.5a, c), whereas some of the tows of the back surface, experiencing compression, remained intact and some were fractured (Fig. 5.5b, c). The fabric fracture interacted with delamination at the bending location is also evident in Fig. 5.5c.

Fig. 5.5. Fracture of 1.0 mm thick CFRP warp specimen in dynamic bending test at 0.6 J: (a) front surface (impacted surface); (b) back surface; (c) side

In both types of specimens, the impact force increased with the impact energy but had a little effect on the contact durations. The reason is that higher impact velocities induced larger deformations and therefore, larger impact forces. This
also indicated that the impact velocity dominated the impact energy as the striker mass was the same in all the tests. Further, the contact durations remained almost the same with increase in the impact velocity for each type of specimens. Such a response indicates that the material’s behaviour is strain rate-independent under dynamic loading in fibre-dominated modes as was also observed under quasi-static loading in Section 4.5. As identified in the quasi-static flexural tests (Section 4.5), stiffness as well as strength of the warp specimens was slightly greater than those of weft specimens as can be seen in Table 4.4. Thus, for the corresponding energy levels, the warp specimens demonstrated a higher damage initiation $F_{di}$ as well as peak loads than the weft specimens. Damage initiated at low impact energy of 0.4 J in weft specimens due to their lower stiffness than that for warp specimens at 0.5 J. However, the maximum contact duration of the warp specimens was 13 ms, which was less than 14 ms observed for the weft specimens. This corroborated that stiffness (Young’s modulus) of the impacted specimen had a significant effect on the peak load as well as contact duration. Similar observations were made in [180] by studying the effect of contact stiffness between a striker and a composite shell on their dynamic behaviour. As the weft specimen’s stiffness was lower, the contact duration became longer. These results suggest that in dynamic tests, a given peak impact force can be obtained by varying impact velocity for a constant hammer mass, while a required contact duration can be obtained by choosing the appropriate material’s stiffness.

5.3.2 Force-deflection response

The force-deflection (F-d) response for the warp and weft configurations of CFRP samples at different impact energy levels is given in Fig. 5.6 and 5.7, respectively. If impact causes no damage, peak load relates to the maximum resistance provided by the specimen as an indication of specimen’s flexural stiffness [8]. The slope of F-d curves represents the contact stiffness, while the enclosed area under the curves gives the absorbed energy. Clearly, the area under the curves increased with impact energy indicating an increase in the energy absorbed by CFRP laminates. Energy absorption in composites is mainly through elastic strain energy, plastic deformation and formation of various damage modes [181]. In the material under study, the TPU matrix material is highly compliant and usually absorbs more energy without causing any appreciable damage. The descending
Fig. 5.6. Force-deflection response of CFRP warp laminates at various impact energies

Fig. 5.7. Force-deflection response of CFRP weft laminates at various impact energies
sections of the curves, where both the load and deflection decrease, represent rebounding of the hammer. Here, again there is a notable difference in the behaviour of warp laminates as compared to weft laminates. For warp specimens, stiffness (slope of load–deflection curve) showed an increase with the increase in the impact energy, whereas, for weft specimens, only a small increase in stiffness was observed. The increase in stiffness can be associated with laminate thickness effects (stress stiffening) on the impact event. It was also shown in [8, 180] that the contact stiffness for relatively thin laminates did not remain constant but increased with the impact force (impact energy), and that the magnitude of the change was dominated by the deflection difference between the contact centre and the support boundary. However, stiffness of warp specimens was more than that of weft specimens. Due to lower stiffness, the maximum deflection of weft specimens was slightly more than that of the warp at the respective impact energy levels. The initiation and progress of damage were evident by oscillations in the load at higher energy levels. The specimen’s fracture was represented by sudden load drop at 0.6 J impact energy. As load–time and load–deflection plots essentially gave similar inferences, thus only load–time plots are used for comparison with numerical analysis of the dynamic bending of CFRP composites.

In the above impact events, the laminate suffered bending as a result of the interface pressure in the contact area. The bending mode was significant even at points far from the contact area (e.g. tip of the cantilevered specimen), especially when the specimen’s thickness was small in comparison with other dimensions of the specimen resulting in lower flexural rigidity. Obviously, flexural rigidity of the composite specimens depends on material’s stiffness, their geometry (size) and boundary conditions. Thus, a larger plate of the same material with only one end fixed (cantilever) has a considerable decrease in flexural rigidity and membrane stiffness resulting in large-deflection than a small plate with its edges fully clamped as in the case of specimens for drop-weight tests. However, still, the membrane component was significant as the specimens were very thin. This was also confirmed by examination of the peak deflections, which were several times greater than the thickness of the specimens. Hence, in the Izod-type bending tests, where only one side was clamped and the rest were free resulting in low
membrane stresses, thin laminates failed at low energy levels. Experimental results highlighted the energy absorbing capabilities of CFRP composites.

5.4 Micro-CT analysis of damage

Micro-CT analysis was used to obtain detailed information about the actual damage mechanisms and their location through the thickness in the impacted specimens. In this analysis, the same set-up as described in Section 4.6.2, was employed to scan and develop tomographs of CFRP warp specimens.

Fig. 5.8. Reconstructed 3D images of CFRP warp specimen at impact location across height of sample (resolution 14.7 µm): (a) edge; (b) 50% of height; and (c) 75% of height

5.4.1 Analysis of damaged specimen

A sample from the impact region of the damaged CFRP warp specimen tested at 0.5 J was prepared for 3D scanning. To obtain a reasonable resolution and remain within the system field of view, the size of the sample was 20.6 mm length x 10.2 mm width x 1 mm thickness. The data for the sample was collected at 75 kV and 80 µA. Transmission X-ray images were acquired from 3600 rotation views over 360° of rotation (0.1° rotation step) for 3D reconstruction. These settings resulted
in tomographs with a resolution of 14.7 µm for the impact specimen. The reconstructed 3D images of the transversely impacted CFRP specimen are shown in Fig. 5.8. The tomographs showed matrix cracking and delamination at the hammer impact location along the specimen’s height. Dark grey regions in the images represent cracks and damage whereas light grey regions represent carbon-fibre yarns. Matrix cracks developed in weak resin-rich pockets around the tows. Inter-ply delamination can also be observed. Micro-CT analysis showed that matrix cracking and inter-ply delamination were the prominent damage modes at the specimen’s impact location.

5.4.2 Analysis of fractured specimen

The CFRP warp specimen fractured at 0.6 J impact energy was also scanned employing Micro-CT at the impact location as well as fracture (bending). Here, to obtain high resolution images, two small samples, one with dimensions of 10.14 mm x 6.0 mm x 1.0 mm (W x H x T) from the impact region, and second with dimensions of 8.35 mm x 6.10 mm x 1.0 mm from the fractured region were prepared. The data for the samples was collected at 80 kV and 85 µA. Here, too, for 3D reconstruction, transmission X-ray images were acquired from 3600 rotation views over 360° of rotation (rotation step of 0.1°). These settings resulted in tomographs with resolution of 6.7 µm and 6.1 µm for the impact and fractured specimens, respectively.

The reconstructed tomographs of the fractured specimen at the impact location are presented in Fig. 5.9. The images show matrix cracking, delamination and tow debonding at the hammer’s impact location along the specimen height. At the impact location (Fig.5.9c), the interface on the impact side (specimen’s front side) was more damaged than the back. Dents in the tows below the striker are also visible. All the resin-rich interfaces underwent inter-ply delamination below the impactor. The reconstructed 3D images of the CFRP specimen at the bending (fractured) location are shown in Fig. 5.10. Realisations of inter-ply and intra-ply damage mechanisms at outer edge, 25%, 50%, and 75% of the sample’s width are shown in Figs. 5.10a, b, c, and d, respectively. It is evident that the laminate exhibited matrix cracking and then delaminations and tow debondings before ultimate fracture. Matrix cracking was dominant in the weak resin-rich pockets around the tows. Inter-ply delamination and intra-ply delamination such as tow
debonding can also be observed. These delaminations were initiated by the matrix cracks. In the fibre-rich regions, damage was apparently linked to debonding at the tow/matrix interface. The analysis of internal structure showed that at the time of fabric fracture which was triggered by tensile fibre failure, almost every ply was delaminated. This delamination is more pronounced in Figs. 5.10a and c than in 5.10b and d. The reason is that warp tows, which are aligned along the specimen’s axis, bear a higher load than transverse weft tows in bending of the specimen under impact. All the tomographs showed that matrix cracking, delamination and tow debonding were the prominent damage modes at the specimen’s impact location, whereas at the bending location, these modes were coupled with fabric fracture.

Fig. 5.9. Reconstructed 3D tomographs of twill 2/2 CFRP specimen at impact location across height of sample (resolution 6.7 µm): (a) edge; (b) 25% (1 mm above impact centre line) (c) 50% (at impact centre); (d) 75% (below impact centre line) of height

In this damage analysis, both large and small size samples from the impact region were scanned in an attempt to investigate various damage modes at different resolutions. At lower resolution of 14.7 µm, only some major damage modes such as matrix cracking and delamination were observed (Fig. 5.8). Reducing the sample size by almost half resulted in images of 6.7 µm resolution that showed
tow debonding apart from the matrix cracking and delamination (Fig 5.9). Hence, it is evident that increasing the sample size results in losing the realistic damage information, while reducing the sample size limits the analysis to a localized region. This prevents the Micro-CT technique to allow the study of the complete damage picture in a full-scale specimen. Therefore, as mentioned in Section 4.6.2, the Micro-CT analysis is used to identify the location and type of various damage modes to model accurately their behavior in FE models. Still, the size of delamination cracks at the impact locations is compared to those obtained from the FE analysis in Chapter 7. Further, the damage mechanisms observed in the impact tested specimens were almost similar to those observed in the CFRP specimens tested under quasi-static bending conditions.

![Fig. 5.10. Reconstructed 3D tomographs of twill 2/2 CFRP specimen at bending (fracture) location across width of sample (resolution 6.1 µm): (a) edge; (b) 25%; (c) 50%; (d) 75% of width](image)

5.5 Comparison of large-deflection dynamic bending with drop-weight impact tests

In the dynamic bending tests, thin laminates failed at lower energy levels as compared to their failure in drop-weight impact tests. This is due to the fact that all the edges of the specimens are usually fully clamped in drop-weight tests and, thus, in-plane membrane stresses are developed in composite plates. In contrast,
in the Izod-type bending, only one side is clamped and the rest are free, resulting in low membrane stresses. Thus, a stress-stiffening effect in fully clamped laminates augments their damage resistance and, thus, more energy is required to initiate damage.

Kinematics of the large-deflection bending is different due to low level of constraints in cantilever type specimens, defining significant differences in realisation of damage processes in it as compared to drop-weight tests. In the latter, for comparable materials, damage formation is usually conical diminishing from bottom to top of the specimen as observed in [6, 182]. In this study, at the impact location (Fig. 5.9c), the interface on the impact side (specimen’s front side) was more damaged than the back side. The drop-weight impact tests results in penetration damage at the impact location causing fibre tensile and shear fracture at the back of the laminate at higher impact energies [171, 181]. On the other hand, no such failure modes occurred at the impact location in the large-deflection dynamic bending tests as shown in Fig. 5.5. Minor dents below the striker were observed on the impacted face (Fig. 5.9a). The impact location experienced matrix cracking and delamination as shown in Figs. 5.9c and d. Here, the fabric fracture was observed at the bending location. It should also be noted that the force-time curves of dynamic bending tests were similar to those in drop-weight impact testing but the implications are fundamentally different. In the case of drop-weight impact events, load drops are associated with major damage processes such as fracture of fibre bundles or complete plies. As far as dynamic bending events are concerned, the load drops were associated with both elastic deformation and formation of matrix cracking and delamination below the critical impact energy of 0.6 J.

5.6 Conclusions

Experimental tests were performed using the Izod type impactor to study the dynamic behaviour of CFRP laminates under large-deflection bending. Two different configurations of the laminates - in warp and weft directions - were subjected to low levels of impact energies up to their ultimate fracture. Load versus time and load–deflection diagrams were used to evaluate the dynamic behaviour of the materials at various energy levels. The load histories obtained from the impact tests also yielded important information concerning damage
initiation and growth. At sub-critical energy levels such as 0.5 J, the load drops before the peaks exhibited damage initiation. The ultimate laminate fracture was characterised by a sudden load drop in the force history plots at the critical energy level of 0.6 J. The warp specimens absorbed more energy than the weft ones. Due to their lower stiffness and strengths, the weft specimens showed earlier damage initiation and lower peak loads than the warp specimens. The contact duration, a direct measure of the effective stiffness of the impactor and the specimens, was more for the weft specimens than that of the warp specimens. The load–deflection responses provided the energy absorbing capability of each specimen type. The transient tests gave a better understanding of the load-carrying and energy absorbing capabilities of CFRP composites under large-deflection bending conditions. Since, at low impact energy levels, the specimens showed almost undamaged behaviour, therefore, the dynamic behaviour of warp specimens at 0.4 J is studied as an undamaged response of the materials in FE simulations. The subcritical and critical loading conditions are also be modelled numerically to investigate the various damage modes exhibited by the laminates.

Damage in the tested woven CFRP composites was studied using the Micro-CT technique. Damage analysis at low resolution of the impact region at sub-critical impact energy of 0.5 J revealed matrix cracking as well as delamination. The high resolution images of the impacted region at critical energy level of 0.6 J revealed matrix cracking, tow debonding and delamination at the impact lactation. Minor dents were also observed below the striker. Inter-ply and intra-ply delamination along with fabric fracture were observed at the specimen bending location. The low and high resolution scanning confirmed the limitations of Micro-CT analysis for obtaining the detailed damage information for a full-scale specimen. For the first time, a substantial difference was found in damage mechanisms at the impact location between the non-penetration dynamic bending and drop-weight tests. In drop-weight tests, damage formation was in conical form, whereas in this study, it was distributed almost uniformly at every ply interface. Further, in drop-weight tests, the back face experiences fibre fracture and splitting due to tensile loading; whereas, in dynamic bending tests, there were minor dents on the impacted face. Micro-CT analysis provided a deeper understanding of various damage modes developed during testing on the specimen’s surface and inside it. This study
formed a basis for realistic simulations of the damage behaviour of CFRP laminates under dynamic bending conditions.
6 Chapter 6: Finite-element Modelling of Damage and Fracture in Fabric-reinforced Composites under Quasi-static Bending

6.1 Introduction

As observed in the experimental damage analysis in Chapter 4, quasi-static large-deflection bending of composite laminates resulted in complex damage mechanisms such as matrix cracking, inter-ply delamination, tow debonding and intra-ply fabric fracture. These damage modes usually cause significant stiffness reduction and thus affect the ultimate load level, at which the structural failure occurs. Therefore, it is necessary to study the effect of these damage modes and their interaction on the global load-carrying capability of fabric-reinforced composites for their reliable performance under the given loading conditions. The finite-element method (FEM) has proved to be the most efficient and widely used technique to model these damage mechanisms from initiation through progression to ultimate composite failure. Damage initiation and growth in composite structures subjected to bending loads have been studied by numerous researchers, using various FE models [88, 183, 184]. However, these models were usually developed in the context of a static or steady-state crack propagation using implicit FE tools. Further, the effect of multiple delaminations on the global structural load-carrying capability under large-deflection bending is rarely studied.

Traditionally, the inter-ply and intra-ply damage mechanisms have been treated separately in damage analysis of composite components and structures. Delamination is usually analysed with interface modelling techniques such as cohesive-zone models (CZMs) [72, 119, 185]. The intra-ply damage mechanisms are normally predicted using ply-based strength failure criteria coupled with continuum damage mechanics (CDM) [99, 186-188]. Since, the interlaminar and intralaminar damage modes are strongly interacting energy-dissipation mechanisms in composite structures as observed in Section 4.6, continuum damage modelling is incapable to capture such interaction between these discrete damage mechanisms. A homogenisation process inherent in CDM models leads to the loss of key information on coupling of multiple damage modes at the macroscopic scale and, thus, may result in inaccurate prediction of a crack path [75, 189]. To accurately predict interaction between these discrete failure modes, it
is essential to have an explicit kinematic representation of all damage mechanisms in global models of composite structures [190]. Improved CZMs have the capability to directly couple the bulk (intra-ply) and interface (inter-ply) cracks to achieve unification of crack initiation and propagation. Recently, this interactive damage modelling approach was studied in [97, 190-193]. Hallet et al. [97, 191] simulated fracture of quasi-isotropic laminate specimens in tension tests using CZMs accounting for both inter-ply delamination and intra-ply splitting damage mechanisms and their interaction. Khokhar et al. [192] developed a method of stochastic cohesive-zone elements (CZE) to study the interaction of matrix cracking and delamination in CFRP cross-ply laminates. Okabe et al. [193] applied CZEs to model transverse cracking and delamination, and truss elements to simulate fibre breakage in cross-ply glass-epoxy laminates loaded in tension. The numerical models reproduced complex progressive damage mechanisms in the laminates very well. However, majority of the mentioned work is related to simulations of composite specimens' fracture in tension; research on damage interaction in woven fabric laminates subjected to bending deformation is limited. Therefore, a further research work is needed to develop reliable FE models capable of simulating progression of various damage modes and their interaction behaviour in laminated composites under large-deflection bending resulting in more rational and optimised designs. However, such high-fidelity simulations of discrete damage behaviour of composite laminates should be based on experimental studies of damage mechanisms in order to represent adequately their in-service performance. To address these problems, this chapter first presents numerical simulations of interlaminar damage propagation in woven CFRP laminates under transverse loading, using a cohesive-zone element method. For this purpose, two-dimensional (2D) FE models were developed in the commercial code Abaqus/Explicit to analyse the onset and growth of single and multiple delaminations in CFRP specimens under large-deflection bending. The second part of the chapter addresses modelling of initiation, progression and interaction of two dominant damage modes such as delamination and fabric fracture as observed in the CFRP specimens tested under bending conditions. In these interactive damage simulations, 2D FE models were implemented in Abaqus to study the behaviour of multiple delaminations, fabric fracture and their interaction using CZEs at the identified crack locations. In all these models,
locations of inter-ply and intra-ply damage were identified by micro-structural examination of tested CFRP specimens using Micro-CT and optical microscopy analyses, as elaborated in Section 4.6. Based on the experimental results that showed a nearly similar behaviour for both the warp and weft specimens (Section 4.5), it was decided to develop numerical models of warp CFRP specimens in this chapter. Here, modelling of interaction between inter-ply delamination and intra-ply fabric damage, based on CZEs rather than a CDM scheme, is a novel approach for damage assessment under large-deflection bending. Thus, the aim of this part of the study is to develop and validate numerical models for analysis of the damage and fracture behaviour of CFRP laminates subjected to large-deflection bending. Most parts of the research work presented in this chapter have been published in refereed journal [194, 195].

6.2 FE modelling of single and multiple delaminations

Finite-element models were developed in the commercial FE package Abaqus/Explicit to investigate large-deflection bending of tested composite laminates and the resulting interlaminar damage. An explicit dynamic analysis approach is typically adapted to model large deflection, material nonlinearity and contact in high-velocity transients but it can be also employed effectively in modelling dynamic phenomena with severe discontinuities in the structural response, as occurs in unstable crack propagation. However, an explicit time integration scheme requires a small time increment $\Delta t$ that depends on the highest natural frequency $\omega_{max}$ of the structure and is given as [196]

$$\Delta t \leq \frac{2}{\omega_{max}} , \quad (6.1)$$

The size of time step can be chosen automatically by Abaqus or set manually by the user. Therefore, first, an eigenvalue analysis of the undamaged laminate was carried out to determine the structure’s natural frequency in order to define the first estimate of time increment for stable solution. However, in nonlinear problems, the highest frequency of the model will continually change, which consequently changes the stability limit. Thus, the time increment value is reduced to obtain a converged solution with activating automatic time incrementation. Since the
analysis is quasi-static, the computational time of the simulation is directly proportional to the number of time increments required and size of the FE mesh. The number of increments required is \( n = \frac{T}{\Delta t} \), and depends upon \( \Delta t \) for a constant time period \( T \) of the simulated event. Hence, in a two-dimensional analysis, refining the mesh by a factor of two in each direction will increase the run time in the explicit procedure by a factor of eight (four times as many elements and half the original magnitude of time increment). However, this magnitude should not be too large to lose the accuracy and convergence of nonlinear large-deformation simulation involving damage. Thus, for most simulations conducted in this study, the automatically calculated time step was found to be adequate, since a higher value led to problems with convergence whereas a shorter time step increased the calculation time without improving the accuracy significantly.

6.2.1 Modelling strategy

In this part of simulations, three FE models - A, B and C as shown in Fig. 6.1- were developed representing the bending tests on 80 mm long, 25 mm wide and 1.5 mm thick CFRP warp specimens. Model A contained a single cohesive layer above the beam’s neutral axis (NA), Model B had two interface layers - one above NA and second coinciding with it, whereas Model C had three cohesive layers - above, on and below the NA to simulate multiple delamination scenarios. The cohesive layer above the NA is referred to as top cohesive layer (TCL), the cohesive layer on the NA is referred to as mid cohesive layer (MCL) and the one below the NA is referred to as bottom cohesive layer (BCL). Theoretically, many interlaminar layers may be included in the model at each resin-rich interface between two plies as the location of damage initiation is a priori known. However, in such a case, the modelling effort, the complications that are relevant to the calibration of the penalty stiffness, and the computational times may increase, and the solution convergence becomes rather complicated. Further, the number of cohesive layers in FE models should be such that the structure may be able to carry the applied load without losing its global stiffness before the damage starts in the actual laminate. That is why numerical predictions of composite damage should be based on experimental evidence. To avoid the effect on the actual global stiffness of the structure, the maximum of three cohesive layers were introduced in Model C. Further, in three-point bending, the layers above and below
the NA experience axial compression and tension, respectively, due to bending load, whereas the mid layers are subjected to shear. Therefore, the cohesive layers at these locations were defined to simulate both the single and mixed-mode damage mechanisms.

Fig. 6.1. FE models for single and multiple delaminations simulation in three-point bending

The laminate had a considerable length in z-direction, thus, the generalised plane-strain conditions were assumed. Two-dimensional FE models based on plane-strain elements with linear shape functions were developed to represent the out-of-plane bending behaviour in a computationally cost-effective manner. In all finite-element models, homogeneity at the meso-scale (ply level) was considered for the general description of the composite laminas. The constituent plies were considered to be linear elastic and generally orthotropic; therefore, the concept of engineering constants was used to describe the laminas elastically. The composite laminas were meshed with plane-strain reduced integration and hourglass control CPE4R elements capable of eliminating the shear locking in bending problems,
using the structured meshing technique. Interlaminar cohesive layers were meshed with two-dimensional COH2D4 elements using sweep mesh control. The interface region was discretized with a single layer of cohesive elements with infinitesimally small thickness having shared nodes with ply elements. The beam’s meshes included two elements per lamina along the thickness to reproduce the stacking sequence of the laminate, and capture the normal and shear stress distributions through thickness and control the hourglassing. Application of the indenter load was represented by a circular arc at the centre of the beam that was also laterally supported by two other circular arcs, which were set at a distance of 40 mm along the beam’s axis to replicate experimental tests. All three arcs were considered to be rigid with a diameter of 5 mm. Surface-to-surface explicit contact was defined between the rigid arcs and the laminate top and bottom surfaces. The overhanging length of the beam L (edge) in Fig. 6.1 is 20 mm where the distance between the supports and indenter L (mid) is also kept at 20 mm.

6.2.2 Material properties

All composite laminas were assigned elastic properties shown in Table 6.1. The elastic flexural moduli such as \( E_{11} \) and \( E_{22} \) calculated from the mechanical tests of 1.5 mm thick warp and weft specimens, respectively, (Table 4.4), are listed in Table 6.1. Similarly, the CFRP in-plane shear modulus \( G_{12} \) and its Poisson’s ratio \( \nu_{12} \) determined from the tests in Section 4.3 (Table 4.3) are reproduced here. Rest of the elastic properties were taken from [162], where a similar CFRP woven laminate was studied. Since the matrix makes the weakest link between the plies in a composite laminate, the initiation and progression of interlaminar delamination is controlled by its properties. Thus, in this study, the inter-ply normal \( \sigma_{10} \) and shear \( \sigma_{110} \) strengths used were close to typical values of the TPU matrix, presented in Table 6.1. These properties are usually difficult to determine experimentally with good accuracy; therefore, they can also be assessed by means of numerical calibration. The strength values used here were also validated through numerical analyses, the detail of which can be found in our published work [194]. Further, it was studied in [69] that the variation of the damage initiation parameters within 15% from the typical values did not influence the analysis results. Hallet et al. [191] argued that since it is the propagation of the matrix cracks and delaminations that are the significant events rather than their initiation, the exact value used for the
interfacial strengths is not critical. The values of critical energy release rates of interface matrix are either determined through experimental fracture tests such as DCB, ENF and MMB (see Section 3.6.4), or taken from the published work, or calibrated numerically. In this study, the latter two approaches were employed. The fracture toughness values in Mode I \( (G_{IC}) \) and Mode II \( (G_{IIc}) \) were taken from [127] as first estimates, where numerical damage analysis of a similar CFRP material was carried out. There is a significant variation in values of critical energy release rates for interface delamination in composite systems published in the literature. Therefore, the fracture toughness values were also calibrated based on numerical optimisation so that the corresponding FE model was capable to represent damage in the bending test of the laminate.

Table 6.1. Material properties of CFRP considered in the FE model

<table>
<thead>
<tr>
<th>Elastic property</th>
<th>Interlaminar Strength and Toughness</th>
</tr>
</thead>
<tbody>
<tr>
<td>( E_{11} ) (GPa)</td>
<td>( \sigma_{10} ) (MPa)</td>
</tr>
<tr>
<td>( E_{22} ) (GPa)</td>
<td>( \sigma_{110} ) (MPa)</td>
</tr>
<tr>
<td>( G_{12} ) (GPa)</td>
<td>( G_{IC} ) (J/m(^2))</td>
</tr>
<tr>
<td>( \nu_{12} )</td>
<td>( G_{IIc} ) (J/m(^2))</td>
</tr>
<tr>
<td>( E_{33} ) (GPa)</td>
<td>8.0</td>
</tr>
</tbody>
</table>

6.2.3 Interlaminar damage modelling using cohesive zone elements

Interlaminar damage in the resin-rich region between plies of a laminate is usually an invisible threat to structural integrity of composites. Computational models with the capability to predict the initiation and progression of delaminations can reduce the number of costly experimental tests and can lead to improved designs. Cohesive-zone elements (CZEs) have the ability to capture the onset and propagation of delamination [120, 128]. Cohesive elements can be defined at various locations in FE models, and the analysis determines which one, or what combination of potential delaminations develops. The elements can also be placed between every ply of a laminate, although it is not necessary to place them at interfaces between plies of the same orientations where delaminations occur rarely [108]. Delamination failures in composite laminates initiate and propagate under the combined influence of normal and shear stresses. To account for mode interactions under mixed-mode loading, the quadratic failure criterion given by Eq.
3.10 was used for damage initiation in this study. Delamination propagation was based on the B-K criterion (Eq. 3.11). The value of empirical mode-mixity parameter $n$ characterizing coupling between the modes was taken as 2. The interface element stiffness was calibrated using Eq. 3.13, where the material’s through-thickness stiffness of 8.0 GPa (Table 6.1) and average thickness of the resin interface from Fig. 4.2 were used. The interface stiffness determined from Eq. 3.13 was used as a first estimate, which was increased gradually until convergence was achieved at $4 \times 10^6$ N/mm$^3$.

6.2.4 Discretization and mesh convergence

Presence of CZEs in the FE model defines the crack propagation path. The extent of crack growth along the prescribed path defined by CZEs depends on the size of these elements as the cohesive-zone model is a local approach. Thus, the application of CZE requires a fine spatial discretization at the cohesive-zone to capture the damage growth properly. However, such refinement may be prohibitive since it needs a significant increase in computational efforts [197]. An optimum number of elements are required in the cohesive-zone to obtain accurate numerical results. In case of a coarse mesh used for a cohesive-zone, the distributions of tractions ahead of the crack tip are not reproduced accurately. Various models were proposed in the literature to determine the cohesive-zone length [112, 113, 119, 120]. However, the minimum number of CZEs is not well established. Further, all these models are based on pre-existing and pre-defined cracks in the laminate such as in simulating DCB, MMB and other tests. No accurate model is available to determine the cohesive element size for the FE model of the undamaged state before load application.

In this work, the model proposed by Turon et al. [120] was used in the numerical analysis to obtain an initial estimate of the cohesive-zone length and the size of interface elements defined by Eq. 3.15. Before performing further simulations, a mesh convergence study was performed on Model A with a single cohesive layer. Four FE models were developed with different element lengths ranging from 0.05 mm to 0.4 mm. The thickness of all interface elements was 16 µm; the total thickness of the laminate remained unchanged. The laminate was meshed with fine structured elements of aspect ratio 1 to avoid premature solution termination.
because of elements distortion due to geometric as well as material nonlinearities. The performed mesh study with different element lengths is summarized in Fig. 6.2. The results indicated that by decreasing the element length, the damage zone along the laminate length increased and solution convergence was achieved. Mesh 3 of 0.1 mm element length was selected for computationally effective simulations of single and multiple delaminations in numerical Models A, B and C. A similar behaviour of CZEs was also shown in [126], indicating that as long as the interface element size was kept less than 1 mm, numerical results were in agreement with experiments and a better solution convergence was achieved.

![Fig. 6.2. Damage sensitivity to cohesive element size](image)

**6.2.5 Boundary conditions and solution**

Simply supported boundary conditions were applied at the reference points of the rigid supports below the laminate representing the test fixture. A displacement-controlled load was applied at the central rigid arc representing the indenter, which was in contact with the top ply of the laminate, for better convergence of the solution. Boundary conditions were applied at rigid surfaces instead of constraining the ply nodes as the local stresses due to the constraints-induced edge effects disperse over greater distances of the structure because of the
composite’s anisotropy. As shown by Horgan et al. [198] for anisotropic composite materials, the application of St. Venant’s principle for plane elasticity problems involving anisotropic materials is not justified in general. The displacement was applied gradually to obtain a stable and converged solution at each equilibrium iteration for a particular time step. The model was solved using the explicit solver capable of overcoming convergence difficulties due to the material softening behaviour and stiffness degradation after the onset of damage. Since Abaqus/Explicit is a dynamic solver, thus, for quasi-static analysis, the beam was loaded slowly enough to eliminate significant inertia effects. This was ensured by keeping the kinetic energy low throughout the response when compared with the internal energy of the system. Quasi-static analysis was carried out for 0.1 second with large-deflection effects by applying the load in small time increments of $10^{-7}$ seconds to capture the damage process in the CZEs. The final FE Model C contained a total of 11,715 elements which took 4 hours on a dual core machine with two 2.7 GHz processors each. To obtain uniform interlaminar shear stress distribution, Model C was finely meshed with 20 µm size on each side, resulting in 31,600 elements. Solution of this model took 11 hours and 20 minutes on quad core machine with two 3.47 GHz processors each. The computational cost is a direct consequence of a fine mesh coupled with the highly nonlinear behaviour of interface damage elements.

6.2.6 Results and discussion

Results of numerical simulations of the large-deflection bending behaviour of CFRP laminates and comparison with experimental tests are presented in this section. Damage initiation and progression along a single cohesive layer above the neutral plane of the beam in FE Model A is shown in Fig. 6.3. Damage is represented by normalised length ($L_d/L$) of the cohesive layer along the beam axis against the normalised displacement loading ($\delta/\delta_f$), where $\delta_f$ is the displacement at the ultimate failure of the test specimen. The damage in the overhang initiated at 22% of the total load, lower than that for the mid-region, which was at about 50%. The delamination process in this model was of mode-I type triggered by normal stresses above the neutral plane. Similarly, delamination progressed more in the overhang than the mid-region before the ultimate failure. The initiation and progression behaviours of multiple delaminations in Model B are demonstrated in
Fig. 6.4. Here too, delamination initiated faster in the overhang regions of the top and mid cohesive layers than in the mid-region. The overhang exhibited mode-I fracture whereas the mid-region was in mode-II state. However, the delamination grew more rapidly in the MCL than TCL in the beam’s mid-region until the MCL was completely damaged. This behaviour was more pronounced in the results of Fig. 6.3. Damaged zone at the edge and middle of the specimen - Model A

Fig. 6.4. Damaged zone at the edge and middle of the specimen - Model B
multiple delaminations in Model C, shown in Figs. 6.5 and 6.6. Figure 6.5 demonstrates that although delamination initiated earlier in the beam’s edges, it grew more in the mid-section. The MCL was more damaged due to mode-II shear fracture as shown in Fig. 6.6. The reason for this is that the maximum through-thickness shear stresses generally occur in the mid-section of the laminate and drive the mode-II delamination process. Further, the mid-region exhibited mode-II delamination as the shear stresses outside the beam’s supports diminished as shown in Fig. 6.8. Variation of flexural stress $\sigma_{xx}$ along the beam thickness in Model C is shown in Fig. 6.7. Figure 6.8 displays the contour of interlaminar shear stress $\tau_{xy}$ in Model C under three-point bending. Interlaminar shear stress existed between the supports as shown in Fig. 6.8, and its value diminished outside the beam supports, as the shear force vanished there. The letters L, M and S in Fig. 6.8 indicate the locations of loading, mid-span between loading and support, and support, respectively, for investigation of the interlaminar shear stress profile and distribution through the thickness. Contour plots of the shear stress at these locations are presented in Fig. 6.9. The shear stress distribution through the thickness computed at the nearest Gauss points to points L, M and S in Fig. 6.8
Fig. 6.6. Development of interlaminar damage in multiple cohesive layers (Model C) under bending load

Fig. 6.7. Contour of bending stress $\sigma_{xx}$ in Model C of three-point bending test

Fig. 6.8. Contour of interlaminar shear stress $\tau_{xy}$ in Model C of three-point bending test
are illustrated in Fig. 6.10. The shear stress concentrations in the vicinity of the loading point and the support rollers (points L and S) are very high but they rapidly decrease to values below the failure threshold at approximately 1/10 of the thickness for the CFRP laminate. Feraboli and Kedward [199] found this value as 1/12 of the thickness for carbon/epoxy composite laminates. It is evident from Figs. 6.9b and 6.10b, that at a distance away from the rollers, the shear stress increases toward the middle i.e. the neutral axis of the beam. The transverse distribution of shear stresses is parabolic with a maximum value of 25 MPa at the central cohesive layer, which experienced mode-II fracture. A sharp increase in stress at the locations of cohesive layers i.e. crack tips, is apparent in Fig. 6.10, when the interlaminar failure occurred, as expected.

![Shear stress contour plots](image)

Fig. 6.9. Contour of interlaminar shear stress $\tau_{xy}$ at load application point L (a), mid-span between loading and support, point M (b) and support point S (c) in Fig. 6.8

Similarly, contour plots and distribution of normal stress $\sigma_{yy}$ are presented in Figs. 6.11 and 6.12. Here, too, the loading and support points (L and S) experienced high stress concentration as shown in Figs. 6.11, 6.12a and 6.12c but it decreased...
rapidly to lower values away from these locations. At the mid-span between the loading point and support, normal stress is almost uniformly distributed but varies at the cohesive layers as depicted in Figs. 6.11 and 6.12. In the bending theory of beams, this normal stress is assumed to be negligible; however, at the loading and support locations, as was observed, this along with the interlaminar shear stress can cause localised damage in composite beams. In all the numerical models, delamination initiated at above 30% of the failure load and then propagated at a higher rate. Damage suddenly spread after attainment of 70% of the failure load, especially in the middle section of the beam.

![Graphs](image)

Fig. 6.10. Through-thickness distribution of Interlaminar shear stress at load application point L (a), mid-span between loading and support, point M (b) and support point S (c) in Fig. 6.8

The developed numerical models were validated through comparison with experimental test data. The load-deflection behaviour obtained numerically for three different models and experimentally from three-point bend tests of CFRP 1.5 mm thick warp specimens is presented in Fig. 6.13. A good agreement was achieved between experiments and numerical simulations indicating that the numerical models were capable to reproduce the damage of the composite.
Fig. 6.11. Contour of normal stress $\sigma_{yy}$ at load application point L (a), mid-span between loading and support, point M (b) and support point S (c) in Fig. 6.8

Fig. 6.12. Through-thickness distribution of normal stress at load application point L (a); mid-span between loading and support, point M (b); and support point S (c); in Fig. 6.8
specimens. It can be observed that the development of such interlaminar damage did not affect noticeably the force-deflection curves until the stiffness degradation occurred at points P, Q, R and T (Fig. 6.13). At these points the damage saturation occurs, followed by instantaneous loss of structure’s load-carrying capability. However, such internal barely visible delamination damages can reduce the compressive strength of the composite and can result in buckling of the plies. Figure 6.13 shows that by increasing the number of cohesive layers in FE models, the structure lost its load-carrying capability earlier and ultimate failure occurred at a lower load level. Thus, a reasonable number of cohesive layers should be defined in FE models to capture the real damage behaviour especially between the plies with different orientations, where the laminate is more susceptible to delamination initiation. Models A and B, with one and two cohesive layers, respectively, reproduced the specimen’s failure behaviour observed in tests. Further, all of the performed numerical simulations and the presented numerical–experimental comparison indicated that the response of composite laminates before the onset of delamination was adequately reproduced by the calibrated stiffness of the interface elements.

Fig. 6.13. Numerical and experimental load-deflection response of CFRP laminates under bending
6.3 FE modelling of damage interaction

Various damage modes grow in composite laminates with load application intercatively rather than independently. Such copuling of various damage mechanisms was observed in the damage analysis of tested CFRP specimens in Section 4.6. It was also found that inter-ply delamination and ply fracture were the domeinant damage modes. In the bending tests of 1.0 mm thick CFRP specimens presented in Section 4.5, it was observed that some specimens fractured completely, while in others, the top ply under compression remianed intact. In this section, the behaviour of both types of failures along with the coupling of damage modes is analysed numerically. To accurately predict interaction between delamination and ply fracture, the damage mechanisms were represented explicitly in global models of composite specimens by defining CZEs at the predefined damage locations.

![Fig. 6.14. FE models for simulation of damage interaction in three-point bending](image)

6.3.1 Modelling strategy

The finite-element models – Models D and E as shown in Fig. 6.14 - were developed in Abaqus/Explicit to investigate large-deflection bending of the tested
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CFRP composite laminates as well as resulting inter-ply and intra-ply damage and their coupling. These models represent the bending tests on 80 mm long, 25 mm wide and 1.0 mm thick CFRP warp specimens. The meso-level models consisted of homogeneous orthotropic plies and a cohesive interface connecting two adjacent plies. Models D and E contained three longitudinal cohesive layers - one above the beam’s neutral axis (NA), the second coinciding with it, and the third below the NA, to simulate multiple delamination scenarios. An additional transverse cohesive layer was inserted at the laminate centre to simulate coupling between delamination and ply fracture in both models. However, in Model D, the transverse interface layer crosses three plies to simulate the tested laminate with an intact top ply under compression, whereas in Model E the transverse cohesive layer crosses the whole thickness to simulate the complete laminate rupture as shown in Fig. 6.14. Cohesive layers designated in Fig. 6.14 were included in the model since the location of damage initiation was a priori known from our Micro CT and microscopic analyses (Section 4.6).

Here, too, 2D FE models based on plane-strain formulation were developed to represent out-of-plane bending. Composite plies were meshed with plane-strain CPE4R elements whereas cohesive layers with two-dimensional COH2D4 elements. As found in Section 4.5, the flexural moduli were increased for 1.0 mm thick specimens in both warp and weft laminates to 55.0 GPa and 52.0 GPa, respectively. These elastic flexural properties were assigned to composite plies. The rest of elastic constants were the same as listed in Table 6.1. The interlaminar normal $\sigma_{10}$ and shear $\sigma_{20}$ strengths as well as fracture toughness values in Mode I $G_{IC}$ and Mode II $G_{IIIC}$ listed in Table 6.1 were used for delamination modelling. The stiffness of the interface elements representing delamination was $4 \times 10^6$ N/mm$^3$, similar to that used in Section 6.2. The initiation of intralaminar ply fracture was based on the average ultimate flexural strength of the warp specimen taken as 993 MPa from the bending test. Intralaminar damage evolution was based on fabric fracture energy $G_{IC}$ of 40 kJ/m$^2$ in both warp and weft directions as reported in [65] for a similar carbon fabric. The stiffness of interface elements defined for intralaminar fracture was based on the longitudinal elastic stiffness $E_{11}$ of the ply instead of $E_{33}$. The value of $8 \times 10^6$ N/mm$^3$ was found to be a reasonable estimate for this simulation. A similar interface element length of 0.1 mm as defined in
Model C was selected for damage simulations in Models D and E. The boundary conditions as well as the analysis procedure were the same as in Section 6.2 for interlaminar damage simulations. The displacement-controlled load was applied gradually in small time increments of $10^{-6}$ s using automatic time stepping to capture the damage process, keeping the amount of kinetic energy small with respect to the total internal energy. The final FE Model E contained a total of 12,832 elements and it took 20 minutes on a six core machine with two 3.33 GHz processors each to run the simulations.

### 6.3.2 Results and discussion

In this section, the results of simulations of damage interaction in CFRP laminates under large-deflection bending and their comparison with experimental data are presented. The initiation and progression behaviours of multiple inter-ply delaminations and intra-ply fracture in Model E are demonstrated in Fig. 6.15. Various damage modes are presented by respective normalised lengths $L_d/L$ of the cohesive layer against the normalised deflection $\delta/L_f$, where $L_f$ is the deflection at the ultimate failure of the test specimen. Apparently, inter-ply delamination initiated earlier, at about 30% of the failure load in the mid-regions $L$ (mid) of the top, mid and bottom cohesive layers, than in overhang regions $L$ (edge). After initiation, the delamination propagated at a slower rate up to 91% of the failure load. After this, a faster stable delamination growth was predicted in the specimen's mid region until the ultimate failure. As studied in Section 6.2, in three-point bending, the maximum shear stresses occurred in the vicinity of loading and support rollers due to stress concentration, and the through-thickness high shear stress at the mid-plane of the laminate due to the shear force associated with bending as shown in Fig. 6.16. A sharp increase of shear stress $\tau_{xy}$ at the locations of crack tips is apparent in Figs. 6.16b and c. These interlaminar shear stresses were responsible for damage initiation in the mid-region. Therefore, as the applied load reached 30% of the failure load, mode-II type delamination initiated in the specimen's mid-region where the shear stress equaled to the interface shear strength $\sigma_{II0}$, and grew more in MCL than TCL and BCL. Delamination in the beam's overhang region started at about 70% of the failure load, and here too, MCL was more damaged than TCL and BCL. In this case, the
Fig. 6.15. Numerical evolution of damage in the edge, middle and ply-crack zones of the specimen (Model E)

Fig. 6.16. Contour plots of interlaminar shear stress $\tau_{xy}$ in Model E of three-point bending at various load levels: (a) 400 N; (b) 408 N; (c) 416 N
entire mid-interface layers were totally damaged before the ply fracture started as shown by damage in CCL in Fig. 6.15. As the specimen's span-to-thickness ratios were 26.7 and 40 for specimens with thickness of 1.5 mm and 1.0 mm in Models C and E, respectively, the Model C resulted in a higher interlaminar shear stress due to larger specimen thickness. This caused an earlier damage initiation and faster growth in Model C than in Model E. However, still, the MCL in both models was completely damaged before the ultimate failure load.

Fig. 6.17. Calculated evolution of inter-ply and intra-ply damage interaction in Model E of three-point bending at various load levels: (a) 400 N; (b) 408 N ;(c) 412N; (d) 416 N; and (e) after fracture (scaling factor 0.5)

The sequence of the inter-ply delamination and intra-ply fabric breakage in Model E is illustrated in the deformed plots of damage shown in Fig. 6.17. Here, the interface layers were fully delaminated before the fabric breakage initiated, as represented by the laminate mid-region in Fig. 6.17a. The first ply’s fracture occurred at 98% of the ultimate load as shown in Fig. 6.17b followed by fractures of the second, third and fourth ply in mode-I as illustrated in Figs. 6.17c, d and e, respectively, at higher loads. In damaged areas, the ply elements moved relative
to each other along the delaminated interface layers as in mode-II fracture. The mismatch of element boundaries showed delamination and ply fracture.

Fig. 6.18. Comparison of delamination crack lengths: (a) observed in microscopic analysis of tested CFRP specimen; (b) numerically predicted in Model D (scaling factor 0.5)

In Model D, the fourth ply was left intact, since in bending it was not fractured because of compressive stresses on top of the specimen. Although, fibre breakage due to buckling instability in the top ply under compression was observed (see Fig. 4.19), this behaviour was not studied in these meso-level models. The length of delamination cracks observed microscopically in the tested specimen is compared with that obtained numerically from Model D in Fig. 6.18. The lengths of the delamination cracks matched reasonably well; the delamination length of 2.53 mm in simulation was somewhat smaller than that of 2.73 mm in the specimen. Here, in the simulations, only those elements which were fully damaged and deleted from the model when damage variable D attained the value ‘1’, were considered for the crack length. However, if the red elements in the damage process zone with ‘D’ near to ‘1’ were also considered, then the crack length might have matched to the experimental one. In Model D, the damage sequence was the
same as in Model E except that the damage progressed on both sides of the cut plies along the delaminated TCL in the form of mode-II fracture as shown in Fig. 6.18. Variation of flexural stress $\sigma_{xx}$ along the specimen’s thickness in Model E is illustrated in Fig. 6.19. Although the composite was fully delaminated (Fig. 6.19a), the distribution of flexural stress was still uniform. This stress reduced due to reduction in flexural load as the first, second and third plies were fractured (Figs. 6.19b, c, and d, respectively). Stress concentration at the locations of inter-laminar crack tips is apparent in Figs. 6.19b, c and d. As expected, it could be noted that each delaminated ply behaved like a cantilever beam in bending constrained at the delamination tips. Bending stress decreased drastically as the laminate fractured (Fig. 6.19e).

Fig. 6.19. Contour plots of bending stress $\sigma_{xx}$ in Model E of three-point bending at various load levels: (a) 400 N; (b) 408 N; (c) 412 N; (d) 416 N; and (e) zero (scaling factor 0.5)
Figure 6.20 shows an energy balance composed of the kinetic energy, the strain energy, the damage dissipation energy (energy released during crack propagation) and the external work obtained with the explicit analysis of Model E. It can be observed that the kinetic energy quickly developed during the crack propagation at about 91% of the ultimate deflection; the kinetic energy was negligible indicating that the solution was quasi-static before the ply fracture started. At the same instant a quick increase in the damage dissipation energy and decrease in the strain energy could be identified showing the first ply fracture. Although delamination occurred before ply fracture, the damage dissipation energy was negligible during delamination propagation. The oscillatory behaviour of the kinetic and strain energy after the ply fracture may be due to elastodynamic effects of wave motion in the ply material from the sudden release of the interface in the transverse direction during rapid crack propagation as no contact elements were used between the plies. The significant kinetic energy term indicated the importance of dynamic effects during crack propagation. Hence, it showed that the unstable crack growth during the ply fracture was dynamic even though the loading was quasi-static. Similar behaviour of dynamic crack propagation was observed in the numerical analysis of L-shaped unidirectional composites in [200].

![Energy balance](image)

Fig. 6.20. Evolution of energy in Model E
Comparison of numerical and experimental load–deflection curves of CFRP woven laminates under flexure is presented in Fig. 6.21, demonstrating a good agreement. Here, too, the development of inter-ply damage did not affect noticeably the force–deflection curves till the stiffness degradation occurred at points X, Y and Z in the test, Models E and D, respectively (Fig. 6.21). Delamination damage occurred at those points characterised by small drops in load. The ultimate fabric fracture in both Models D and E is identified by a quick load drop representing instantaneous loss of structure’s load-carrying capability. As Model E represented full rupture of the specimens, the load dropped to zero. Both in the FE models and experiments, during delamination propagation the laminate carried the load, but with a lower slope in the load-deflection curves that indicated a loss of flexural stiffness due to delamination. Apparently, for on-axis laminates made only of 0° and 90° plies, the model including both geometric and material nonlinearities produced a nearly linear response but also properly predicted the onset of total failure observed in the experiments. A transition to another orientation such as in a case of 45° plies will contribute to plasticity.
resulting in a pronounced softening behaviour of the specimen. Although the numerical models showed earlier stiffness degradation due to delamination, still, the main features of behaviour were the same.

6.4 Conclusions

Damage in CFRP textile composites under large-deflection bending was studied using numerical simulations. Two-dimensional plane-strain finite-element models were implemented in the commercial code Abaqus using the explicit solver. A series of simulations was performed first to study the onset and progression of the inter-ply delamination process under mixed-mode large-deflection bending and then the interaction of inter-ply and intra-ply damage modes observed in the specimens. In the developed FE models, single and multiple layers of cohesive-zone elements were defined at the damage locations identified from the microstructural examination.

In the first section of interlaminar damage modelling, the numerical results were in agreement with the experimental ones, and the numerical models had the capability to reproduce the failure mechanisms in composite laminates. The FE models provided more information than the experimental tests and helped to gain a better understanding of the damage initiation and evolution processes in woven laminates. The numerical simulations showed that damage initiation and growth was sensitive to the mesh size of cohesive-zone elements. The top and bottom layers of the laminate specimen experienced mode-I failure whereas the central layers exhibited the mode-II failure behaviour. Application of the suggested numerical approach to the test cases proved the capability to model complex patterns of damage development in originally undamaged specimen. Damage that suddenly propagated and subsequently led to an immediate loss of load-carrying capability was captured with the explicit dynamic approach. Overall, it should be noted that all of the modelled interlaminar layers in the finite-element schemes represented potential zones of damage nucleation and propagation. The results indicated the suitability of the developed numerical approach to study the onset and propagation of interlaminar damage. However, the calibration of numerical models based on interface layers proved to be highly mesh and stiffness sensitive. The numerical results also revealed that in order to achieve a response closer to that observed in the experimental tests, there must be some limitations on the
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values of cohesive elements’ stiffness and size, which, in turn, influence the computational cost of simulations. It is also significant that all interface parameters must be calibrated and specified correctly in order to avoid long computational times, solution oscillations or even premature termination and obtain better convergence.

In the second part of modelling of damage interaction, the meso-scale 2D FE models were developed to simulate some modes of inter-ply and intra-ply damage and their subsequent interaction observed in woven composites. The obtained numerical results were close to the experimental ones, and the numerical models had the capability to reproduce the damage sequence and pattern observed experimentally in composite laminates. The length of delamination crack was predicted close to that observed in the tested specimen. The load level, at which the load drops occurred, caused by delamination damage, also showed a good correlation as did the predicted fabric failure load. Numerical analyses showed that dynamic effects were significant during ply crack propagation even under quasi-static loading. Based on the results of the simulations of CFRP specimens with longitudinal and transverse cohesive layers, it is concluded that the models accurately represented the onset and propagation of delamination, coupling between delamination and ply cracking and the final fracture of the laminate as a result of fabric fracture. The finite-element simulations provided important information such as stress and strain distributions, crack length variation with applied load and interaction of various damage modes in composite specimens, which is cumbersome to measure in quasi-static large-deflection bending tests.
Chapter 7: Finite-element Modelling of Damage and Fracture in Fabric-reinforced Composites under Large-deflection Dynamic Bending

7.1 Introduction

Composite structures, when exposed to dynamic loading conditions in service, suffer more damage than equivalent metallic structures, which can lead to loss of stiffness and load-carrying capability. In low-velocity out-of-plane impacts, the contact duration is relatively long and there is enough time for the structure to absorb energy elastically allowing evolution of various damage mechanisms. The microstructural study of woven laminates in Section 5.4 revealed that various damage modes such as matrix cracking, inter-ply delamination, tow debonding and intra-ply fabric fracture and their subsequent interaction in thin composite laminates were ensued by large-deflection dynamic bending. These damage modes, in particular, delamination cause a significant decrease in the material's in-plane compressive strength and stiffness. The framework of numerical models based on the finite-element method offers an opportunity to describe accurately and predict these complex damage mechanisms in a more cost- and time-effective manner than physical tests or microstructural studies. In such FE models of dynamic events, inter-ply delamination is usually analysed with interface modelling techniques such as cohesive-zone models (CZMs) [171, 183, 201], whereas the intra-ply damage mechanisms are mostly predicted using continuum damage mechanics (CDM) models [65-67, 99, 202]. As highlighted in Chapter 6, CDM, due to its inherent homogenisation process, is limited in modelling the coupling between various damage modes as observed experimentally. Advanced FE models based on CZMs have the potential to couple directly the inter-ply and intra-ply cracks resulting in a unified framework for analysis of crack initiation and propagation. A modelling-based study of interacting damage modes was implemented in [97, 109] by employing cohesive-zone elements (CZEs) in simulations of composite fracture in tension. However, application of the modelling approach to analysis of the behaviour of fabric-reinforced laminates with interacting damage modes under conditions of dynamic bending is limited. Most of
studies focus on the impact behaviour of fabric-reinforced composites for conditions of drop-weight impact loading. A large-deflection dynamic bending behaviour of composite laminates as experienced in sports products is rarely studied.

Failure and damage in fabric-reinforced composites is usually analysed by either using a micro-mechanics approach that considers failure and damage at the constituent (micro) level or a CDM-based meso-level approach, with composites presented as homogeneous orthotropic plies and cohesive interfaces connecting two adjacent plies. Hence, failure and damage is studied at the ply and interface level in the latter approach. Such a meso-level approach coupled with CDM was used in [65-69] to characterise the damage behaviour of woven composites under impact loading. The complex weaving architecture as well as multiple modes of damage at various length scales in woven laminates makes the micro-mechanics-based constituent-level modelling more computationally expensive for a real-size problem. Although FE models incorporating various damage modes at the constituent scale of tows and matrix in woven composites were developed in various studies, e.g. [17, 62-64], however, the problem domain there was limited to an RVE.

The focus of this study is to investigate the macro-level response of real-size specimens of woven laminates for sports products under dynamic bending. Here, a discrete modelling approach based on the CZM rather than CDM is employed to overcome the deficiency of coupling of various damage modes inherent to the CDM approach. The approach adopted here is different from CDM-based modelling: apart from the interface delamination, the out-of-plane fabric fracture is also idealised by inserting CZEs at the fracture location. Therefore, some of the major damage modes such as delamination and ply fracture and their interaction, as observed in Micro CT analysis of the tested CFRP specimens, are studied in a computationally cost-effective manner, without any compromise on the global behaviour of the structure. The location and type of inter-ply and intra-ply damage was identified by micro-structural examination of tested CFRP laminates with Micro-CT technique as described in Section 5.4. Three-dimensional (3D) multi-body-dynamics FE models were developed in Abaqus/Explicit for undamaged, damaged and fractured specimens for various impact-energy levels. The FE
models were developed for warp CFRP specimens only as both the warp and weft specimens behaved almost similarly under dynamic loading (Section 5.3). Cohesive-zone elements were employed at the identified crack locations between the lamina interfaces to model delamination, particularly at the hammer’s impact location, and ply fracture near the specimen’s fixture as well as their subsequent interaction. In this chapter, modelling of coupling between delamination and fabric fracture, based on the CZM rather than a CDM scheme, is used; it is the latest approach for damage simulation under impact bending. The damage modelling approach employed in this work is an extension to three-dimensional impact bending scenarios of the plane strain formulations used to predict the progressive damage interaction under quasi-static bending in Chapter 6. Another element of novelty is the study of behaviour of woven laminates subjected to impacts with a pendulum.

7.2 FE modelling strategy

For this study, three-dimensional multi-body-dynamics FE models consisting of a specimen and hammer assembly were developed in the commercial FE software Abaqus/Explicit to investigate impact bending of tested composite laminates and the resulting inter-ply and intra-ply damage as studied in Chapter 5. Three FE models – Models A, B and C as shown in Fig. 7.1 – were developed to simulate the dynamic behaviour of undamaged, damaged and fractured CFRP specimens at impacts with energy levels of 0.4 J, 0.5 J and 0.6 J, respectively. Model A was developed first to validate a dynamic contact behaviour of the hammer and specimen without incorporating any damage. In the damaged specimen (Model B), representing the impact test at 0.5 J impact energy, CZEs were introduced at the resin-rich interfaces between the laminate’s plies. Hence, Model B contained three longitudinal cohesive layers - one in front of the beam’s neutral plane (NP), the second coinciding with it, and the third on the back of the NP to simulate multiple delamination scenarios. Model C of the fractured specimen at 0.6 J included the same three interface layers with one additional through-thickness transverse cohesive layer at the specimen’s fracture location to model both the inter-ply and intra-ply damage and their subsequent interaction. The cohesive layer in Models B and C in front of the NP (towards impact face) is referred to as front cohesive layer (FCL), the cohesive layer coinciding with the NP is referred to as middle cohesive layer (MCL).
layer (MCL), the one to the back of the NP is referred to as *back cohesive layer* (BCL) and the through-thickness transverse layer is referred to as *crack cohesive layer* (CCL). These cohesive layers were included in the models since the location of damage initiation was *a priori* known from Micro CT analysis in Section 5.4.

**Fig. 7.1.** Schematics of FE models for undamaged (Model A) (a), damaged (Model B) (b) and fractured (Model C) (c) behaviours of CFRP specimens tested at impact energy of 0.4 J, 0.5 J and 0.6 J, respectively.

The material behaviour of the on-axis $[0^\circ, 90^\circ]_{2s}$ warp CFRP specimens was strain rate-independent as observed in the quasi-static bending tests (Section 4.5) as well as dynamic tests (Section 5.3). Such behaviour of woven CFRP composites was also observed in [62, 89] in fibre-dominated modes. Since low-velocity impact bending caused mostly in-plane tension and compression (fibre-dominated modes) in warp specimens, the material constants obtained from the quasi-static tests were used in the FE models. All composite plies were assigned elastic flexural properties of 55.0 GPa and 52.0 GPa in warp and weft directions, respectively.
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The rest of elastic constants were the same as listed in Table 6.1. The hammer was assigned properties of steel with the Young’s modulus of 200 GPa, Poisson’s ratio 0.3 and density 8100 kg/m$^3$. Similarly, the cohesive law was assumed to be strain-rate insensitive, and, thus, static fracture-toughness values were assumed for a dynamic crack growth; this was proven to be sufficient in previous dynamic fracture studies [74, 203]. Thus, as in the previous works, all the rate-dependent effects were due to inertia and to the intrinsic time scale typical of cohesive models in the explicit analysis. Hence, the initiation and progression properties for delamination modelling were the same as listed in Table 6.1. The intra-ply fracture properties were also similar to those described in Section 6.3.1. The stiffness values for the interface elements were the same as employed in the interactive damage models developed in Section 6.3.

7.3 Geometry and boundary conditions

Numerical simulations are based on multi-body dynamics approach, which is concerned with the modelling and analysis of constrained deformable bodies that undergo large deflections during their interaction (contact). In the experimental tests, the hammer and specimen interacted with each other inducing large deformation in the specimen under dynamic bending. To represent the real world physics of the problem, both the hammer and specimen were modelled as deformable bodies in the frame work of multi-body dynamics FE models. Appropriate geometrical models were built, and kinematic and loading boundary conditions were defined to represent the experimental impact bending set up. The used geometry, mesh and boundary conditions of FE models are shown in Fig. 7.2. The specimen and the hammer were modelled as deformable bodies in the multi-body-dynamics FE model. A reference point at the pivot of the hammer along the axis of rotation was created and then tied with the hammer’s cylindrical surface through kinematic constraints. All translations and rotations of the pivot node were constrained except rotation about z-axis to simulate the hammer’s centre of rotation as shown in Fig. 7.2a. Global and local coordinates were defined to account for ply orientations and describe correctly the laminate’s and material’s behaviours. In the FE simulations, the initial position of the hammer’s striking nose was just in contact with the specimen to avoid the computational cost of bringing the hammer from the initially inclined position. The initial angular velocities of 4.72
rad/s, 5.33 rad/s and 5.81 rad/s were applied to the whole hammer, corresponding to the impact energies of 0.4 J, 0.5 J and 0.6 J in Models A, B and C, respectively. The degrees of freedom of the specimen’s bottom part were constrained in such a way that the supports allow shrinkage due to the Poisson’s effect and replicate its boundary conditions in the vice (Fig. 7.2a).

Fig. 7.2. Multi-body impact-test FE model: (a) 3D geometry with boundary conditions; (b) hammer-specimen contact interaction; (c) mesh of hammer; (d) mesh of specimen in Models B and C (specimen is shown larger)

7.4 Type of elements and mesh density

The hammer was discretised with 4-noded linear tetrahedron C3D4 elements, whereas the specimen was meshed with 8-noded linear brick C3D8R elements using a structured meshing technique. Such reduced-integration elements are capable to control hourglass and eliminate shear locking in bending-dominated problems. Each ply of the laminate was modelled with a single element in Models A and B, and two elements in Model C, through its thickness. In Model A of the undamaged response, a mesh size of 0.4 mm x 0.4 mm was defined in the plane
of the specimen. Inter-ply and intra-ply cohesive layers were meshed with 8-noded bilinear COH3D8 elements to model damage initiation and growth. A variable mesh density was introduced in different regions of the specimen in FE Models B and C to reduce the computational time. A structured converged mesh of 0.2 mm x 0.2 mm was defined in the plane of the specimen at the impact and fracture locations. Other regions of the laminate were modelled with a biased mesh. Similarly, a refined mesh was introduced at the contact surface of the hammer’s striker. The striker’s surface mesh density was changed to see its effect on the damage progression in the specimen. Thus, two types of meshes referred to as coarse and refined meshes of size 1.5 mm and 0.2 mm, respectively, were employed in simulations of the damaged specimen (Model B) to assess the effect of a master-surface mesh on the damage propagation (Fig. 7.2c). Excessive deformation and localised stiffness reduction due to damage were controlled by activating the hourglass and distortion controls in Abaqus/Explicit. Also, the damage threshold value was limited to 0.99 thus helping to maintain some residual stiffness as mentioned in [204, 205]. The stable time increment of the whole model is defined by its smallest element, which is linked to the thickness of cohesive elements. Usually, the time step in a dynamic analysis is very short compared to a static one, since it is necessary to capture stress waves moving at high speed in the FE model. The calculated time step for the dynamic analysis is therefore similar to the crossing time of a stress wave over the smallest element in the model. The stable time increment $\Delta T$ can be redefined using the wave speed of the material, $C_d$, and element length, $L_e$, as $\Delta T = \frac{L_e}{C_d}$. The wave speed is given as $C_d = \sqrt{\frac{E}{\rho}}$, where $E$ is the Young’s modulus and $\rho$ is the density of the material.

The stable time increment calculated automatically by Abaqus was found to be adequate for most simulations conducted in this study. The FE meshes of Models B and C contained a total of 187,634 and 268,365 elements, which took 22 and 30 hours, respectively, to run simulations on the HPC cluster of 36 CPUs. The angular velocity was applied gradually in small stable time increments of $2.6 \times 10^{-9}$ s using automatic time stepping to capture the damage process. Such a time increment, required by the explicit solver, resulted in the long computation times.
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7.5 Contact interactions

The contact between the steel hammer’s striking nose and the laminate’s surface was defined by a surface-to-surface kinematic contact algorithm with finite sliding, available in Abaqus/Explicit [48]. In this algorithm, the tangential behaviour was controlled by the Coulomb friction-based penalty contact enforcement method, whereas the ‘Hard’ contact method with allowed separation option was used for normal behaviour of interactions. The friction coefficient of 0.3 was taken at contact between the steel and composites laminate as used in [204-206]. The impactor’s striking surface and the specimen’s surface referred as S1 and S2 in Fig. 7.2c and 7.2d, respectively, were defined as master and slave surfaces, respectively, in all FE models.

7.6 Energy balance

The evaluation of energies associated with impact is required to elucidate the development of various events in numerical simulations. The onset and progression of each damage mechanism of the composite laminate can also be assessed with energies associated with damage mechanisms. The energy balance is given as

\[ E_I + E_K + E_V + E_F - E_W - E_{CW} = E_T \]  

(7.1)

where \( E_I \) is the internal energy, \( E_K \) is the kinetic energy, \( E_V \) is the energy dissipated by bulk viscosity damping, \( E_F \) is the frictional dissipation energy, \( E_W \) is the work of the external forces, \( E_{WC} \) is the work done by contact and constraint penalties and \( E_T \) is the total energy of the system that should be constant. The internal energy can be split in the following terms:

\[ E_I = E_E + E_A + E_{CD} + E_{Dd} + E_{Dr} \]  

(7.2)

where \( E_E \) is the recoverable strain energy (elastic energy), \( E_A \) is the artificial strain energy, and \( E_{DC} \) is energy dissipated by distortion control, \( E_{Dd} \) and \( E_{Dr} \) are the energies dissipated by delamination and fibre damage, respectively. The energy terms can be computed for the entire model or for its specific element sets. The artificial strain energy that includes energy stored in resisting hourglass and transverse shear locking in bending dominated problems, should be negligible compared to the internal energy of the system, i.e. less than 2%. Large values of
artificial strain energy indicate that mesh refinement or other changes to the mesh are necessary [48].

7.7 Inter-ply and intra-ply damage modelling

Apart from inter-ply damage modelling, CZEs have also been used to model intra-ply damage mechanisms such as splitting and ply fracture in composite laminates. Hallet et al. [97, 191] simulated fracture of a quasi-isotropic laminate in tension tests using CZMs to represent both inter-ply delamination and intra-ply splitting damage mechanisms and their interaction. Li et al. [207] employed CZEs to explicitly model delamination, intra-ply splitting and fibre failure in progressive damage analysis of overheight compact tension test notched specimens. These studies formed a basis for modelling intra-ply fracture based on CZEs. Here, too, the damage initiation and propagation were based on quadratic stress criterion (Eq. 3.10) and the B-K criterion (Eq. 3.11), respectively. To capture the damage growth accurately, a reasonable mesh density is needed. For this purpose, the mesh convergence study was performed by developing four FE models with different mesh sizes in the plane of the specimen ranging from 0.15 mm x 0.15 mm to 0.3 mm x 0.3 mm for damaged response. In all these models, two elements per ply were used in through-thickness direction of the laminate. The damaged area at the impact location of the mid layer (MCL) is plotted against the size of CZEs in Fig. 7.3. Here, mesh convergence was achieved with interface elements of size 0.2 mm x 0.2 mm and 16 µm thickness, which was selected for damage simulations in Models B and C. Again, this figure shows that the damage evolution is sensitive to the size of CZEs in a FE simulation. Similarly, the effect of interlaminar shear strength (ILSS) $\sigma_{i10}$ on delamination propagation was studied in Model B by varying its magnitude. In this study, the effect of cohesive elements density on the dynamic response of the specimens was also investigated. To the author’s knowledge, many numerical simulations of impact-induced damage in composites have been performed, but, the effect of density of cohesive elements on the dynamic response of the composites was not studied. Although, the cohesive elements in this study are of very small thickness, still, they are associated with a mass per length that is prescribed by the user through the density of the cohesive material multiplied by the nominal thickness. If the density is very small, it will affect the automatically calculated time step and the convergence of the model.
(see the stable time step equation), whereas if the value is large, it will add an additional mass to the system, affecting the global dynamic response of the laminate. In this study, first, the density was assumed to be equal to that of the bulk material, which resulted in rather high peaks of the impact load of the order of 3-4 kN. When the mass of the model was checked in the status file, it was greater than the actual mass of the system. This density was reduced gradually to obtain the actual mass as well as the level of load peaks observed in the experiments. For the present FE damage models, the density of cohesive elements was chosen so that the mass of the FE models matched the actual mass of the hammer and specimen. Thus, the used cohesive density was found to be a reasonable compromise which did not affect the time step, nor the dynamic response of the specimen.

Fig. 7.3. Mesh convergence study for cohesive elements in Model B

### 7.8 Results and discussion

Results of numerical simulations for the impact bending behaviour of woven CFRP laminates are presented in this section. The evolution of experimental and calculated impact forces with time was compared to assess the accuracy and validity of the proposed modelling approach that attempted to estimate the impact resistance of the woven CFRP composites. It should be noted that the simulation results were filtered using the Butterworth filter with same cut-off frequency of 1
kHz that was used for experimental results. Initiation and propagation of inter-ply delamination and the extent of damage around the impact site and the bending locations as predicted by the interface elements in the FE models are also examined. The predicted delamination crack lengths are also compared with those observed experimentally in the test specimens. Finally, the ability of the modelling approach to capture the interaction between delamination and ply fracture is also discussed.

7.8.1 Response of undamaged specimen

The force-time history of a CFRP specimen tested and simulated in Model A at low impact energy of 0.4 J is shown in Fig. 7.4. The experimental response is compared with the numerical results of Model A, demonstrating a good agreement. The load peak of 70 N in the FE model is slightly higher than the experimentally measured one - 66 N. This discrepancy may be due to the fact that the FE model did not take into account the real physical energy losses due to material viscous damping and friction between the interacting parts of the experimental setup. However, the contact duration is almost the same implying that stiffness of the impacted specimen was accurately modelled. Figure 7.5 shows a time history of

![Fig. 7.4. Experimental and numerical (Model A) force-time diagram for CFRP laminates at impact energy of 0.4 J showing response without visible effects of damage](image)
the energies for Model A. These include the kinetic energy, the strain energy, the viscous dissipation energy, the artificial strain energy, the energy dissipated by friction and the total energy balance for the model. The total energy is seen to remain almost constant during the analysis, as it should. As it is evident, at the start of impact event, kinetic energy is maximum and the strain energy is minimum. Gradually, the kinetic energy is imparted to the specimen and its strain energy increases. This strain energy is then converted entirely back to kinetic energy as the hammer rebounds. As no damage is incorporated in Model A, so no energy loss can be observed in Fig. 7.5. Viscous energy, caused by the small amounts of bulk viscosity included in the Abaqus element formulation and viscous contact damping, is negligible. Artificial strain energy is also very small indicating that the mesh is adequate and not suffered from hourglassing and shear locking. Artificial energies in addition to viscous and frictional effects are less than 2%, compared to the strain energy of the system.

![Energy Distribution](image.png)

Fig. 7.5. Evolution of energies in Model A at impact energy of 0.4 J

### 7.8.2 Response of damaged specimen

The force-time history of a CFRP specimen tested and analysed at impact energy of 0.5 J is shown in Fig. 7.6. Fluctuations in the experimental force-time history (load drops in Fig. 7.6) can be observed before the peak load is reached. These
load drops designated as $F_{di}$-Exp represent the internal damage initiation such as delamination at low incident energy with an associated loss in the laminate stiffness. A comparison between the experimentally measured and calculated force response of the damaged specimen in FE Model B is presented in Fig. 7.6. It can be observed that the global response and the contact times are reasonably well predicted for the damaged specimen. However, the numerical damage thresholds $F_{di}$-FE, at which a significant change in the laminate stiffness is detected, are clearly under-predicted in the finite-element models, and from these points, the force histories separate and follow different paths. This discrepancy was studied by increasing the interface elements stiffness in the FE models; however, no improvement was obtained. It can be argued that the underestimation of damage thresholds and peak forces may be due to the meso-level consideration of the plies in FE formulations. Apparently, in experiments, the woven laminates absorbed more energy due to the resin-rich pockets and the interlacing of fibres in two mutually perpendicular directions, and, thus, offered more resistance to damage initiation, which are not taken into account in the FE models.

Fig. 7.6. Experimental and numerical (Model B) force-time diagrams for CFRP laminates at impact energy of 0.5 J showing response of damaged specimen
Effect of Interlaminar shear strength on initiation and propagation of damage

Composite laminates subjected to low-velocity impacts exhibit delamination usually due to a low level of transverse interlaminar shear strength (ILSS) or fracture toughness of the matrix between the plies. This behaviour of composite materials was studied by Davies and Zhang [208], who proposed a model based on the Mode-II interlaminar fracture toughness of the composite. A similar observation was made by Sutherland and Guedes Soares [209], who developed a simplified shear delamination model to predict the onset of interlaminar damage in composite materials under impact loading. They showed that the critical load for delamination initiation is related to ILSS of the composite given as

\[ P_{\text{crit}} = \frac{6\, \text{ILSS}^3 \pi^3 \, t^3}{E} R \]  

where \( E \) is the equivalent in-plane modulus, \( t \) is the laminate thickness and \( R \) is the radius of the indenter. They predicted that the delamination occurred as the interlaminar shear strength was exceeded.

In this work, the effect of ILSS on the initiation and propagation of delamination was studied numerically instead of using the above analytical approaches. Here, two simulations were performed using different levels of interlaminar shear strength - 12 MPa and 26 MPa - denoted as ILSS–I and ILSS–II, respectively. At low ILSS, delamination initiated earlier at 29 N compared to 35 N at its higher value. Similarly, the load peak was also lower at low ILSS (Fig. 7.6). This may be due to the fact that the increased damage resulted in lower effective stiffness of the laminate. Figure 7.7 shows the predicted delamination area at three interfaces – FCL, MCL and BCL – in simulations with Model B for ILSS–I presented at three different time intervals: at 1 ms (just at \( F_{\text{di}} \)-FE; Fig. 7.7a); 3 ms (Fig. 7.7b) and at 6 ms (near the peak load; Fig. 7.7c). The images shown illustrate the extent of damage by different colours: red corresponds to a completely delaminated region, blue to the undamaged region; while colours between these two indicate the damage process zone with the interface not yet delaminated but locally weakened thus contributing to energy dissipation. In this simulation, delamination initiated first at the point of hammer impact and then at the bending location of the cantilevered CFRP specimen, i.e. the end of the fixture. Due to sharp corners of the hammer’s striker, the ends of the trace were more damaged. A large and quick
Fig. 7.7. Delamination evolution in Model B with ILSS–I (12 MPa) at impact energy of 0.5 J at 1 ms (a), 3 ms (b) and 6 ms (c)
increase in the delamination area was observed simultaneously in all the layers beyond the time corresponding to $F_{di}$-FE. It can be seen that the largest area of delamination occurred at the mid-plane interface layer MCL. In bending of the laminate, the high level of through-thickness shear stress at the mid-plane of the laminate is responsible for damage initiation in the mid-region. Therefore, a Mode-II type delamination initiated in the specimen’s mid-region where the shear stress reached the interface shear strength $\sigma_{II0}$, and grew more in MCL than FCL and BCL at the impact location. Similarly, the upper (front) and lower (back) plies experienced tension and compression, respectively, in cantilever bending. Thus, besides the extensive through-thickness delamination seen in mid layer MCL, the maximum bending stresses were foremost responsible for damage formation at FCL and BCL of the specimen, especially at the bending location. In BCL, subjected to compression, failure of cohesive elements occurred due to shear stresses only because the contribution of normal transverse stress during the compression phase was excluded from the quadratic failure criterion. Damage evolution due to ILSS–II at 6 ms is shown in Fig. 7.8. Here, due to the higher level of ILSS, the damage area is less as compared to that for ILSS–I presented in Fig. 7.7c.

Fig. 7.8. Delamination in Model B with ILSS–II (26 MPa) at impact energy of 0.5 J at 6 ms

The evolution of damage area calculated from Model B with ILSS–I is plotted in Fig. 7.9. As the impactor came in contact with laminate, the sharp corners of the
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A striker caused stress concentration in the laminate, which induced localised damage in the FCL at the beginning. Delamination in the other layers at the impact and bending locations started at approximately 1 ms. After initiation, delamination propagated at a faster rate in the bending region than in the impact region up to the maximum load at 6.2 ms. The maximum shear stress at the neutral plane of the laminate induced more damage to MCL at the impact location, whereas the maximum bending stress at the top and bottom plies caused more damage in FCL and BCL at the bending regions. After the load peak was reached at 6.2 ms, the damage growth stopped in the unloading region of the load-time response.

![Fig. 7.9. Evolution of inter-ply delamination area at impact and bending locations of laminate in FE Model B with ILSS–I (12 MPa) at impact energy of 0.5 J](image)

Figure 7.10 demonstrates the energy time history of the whole FE Model B: the total energy, the kinetic energy, the recoverable strain energy, the artificial energy and the energy due to viscous dissipation and delamination damage. The total energy remained almost constant during the analysis. The kinetic energy of the hammer was completely transferred to the specimen when its velocity reached zero. After this point, the elastic energy stored by the specimen was transferred back into the hammer's kinetic energy, which caused its rebound. A small amount...
of this transferred energy was dissipated by the specimen through elements’ viscosity, delamination damage and friction. Although, the artificial energy was negligible, the viscous energy increased with the damage growth especially after the rebound. As the damage initiated, the default viscosity of elements introduced localized damping to overcome the convergence difficulties and, thus, the viscous energy increased. However, it should be ensured that the viscous damping energy is reasonably low compared to the total strain energy of the model. The viscous energy continued to grow after the rebound as a result of damping of stress waves that usually persist in the specimen after the hammer’s rebound starts.

Fig. 7.10. Evolution of energies in Model B with ILSS–I (12 MPa) at impact energy of 0.5 J

Effect of impactor mesh on initiation and propagation of damage

As mentioned earlier, two types of mesh – coarse and fine – were generated at the hammer’s striker to analyse its effect on the damage evolution. The results obtained for the coarse mesh (with 10 elements along the striker length) with ILSS-I are presented in Fig. 7.11 at 6 ms. Here, the damage contours are only shown at the Gauss points of the elements, whereas the damage in the case of
the fine mesh shows a smooth contour in Fig. 7.7c. Therefore, the mesh of both the master and slave surfaces should be carefully designed in the multi-body deformable contact simulations to predict properly the damage initiation and progression.

Fig. 7.11. Effect of impactor mesh on interface damage in Model B with ILSS–I (12 MPa) at impact energy of 0.5 J at 6 ms

**Comparison of experimental and numerically predicted damage**

Different impact parameters, such as impactor shape and size, laminate thickness, and the type of support and loading conditions of the specimen usually result in different responses and damage mechanisms (both in terms of their size and evolution) [209]. In drop-weight impact tests and simulations, where the fully clamped specimen is hit with a hemispherical impactor, the damage formation is from the back to the front, in a shape of a cone, and the back face of the laminate is usually more damaged due to tensile fracture as elaborated in Section 5.5. The shape of damage in the plane of the woven-fabric laminates varies such as the damage of star shape with the point lying in the warp and weft directions was observed in [6], whereas a diamond shape was detected in [67]. In UD nonwoven composites, delamination in peanut shapes usually result from low-velocity drop-weight impacts [204]. However, in this study of impact bending of the cantilever specimen, the damage formation is front-to-back, and the front face of the
lamine is more damaged than the back at the impact location as can be seen in Figs. 7.7, 7.8 and 7.14. Also the damaged area was of the rectangular shape, where the regions below the corners of the striker were more damaged than those below the striker (Figs. 7.7 and 7.14).

Fig. 7.12. Comparison of experimental (a) and numerically predicted (b) delamination lengths in Model B at impact energy of 0.5 J

Similar front-to-back damage formation was also observed in the Micro-CT analysis of the tested specimens (Section 5.4). The size and shape of the individual delaminations could not be compared to their experimental counterparts as such experimental results were not provided by the Micro-CT for full-scale specimen. Thus, the measurements of delamination crack lengths at the impact location of the specimen tested at impact energy of 0.5 J are compared with the numerical prediction in Fig. 7.12. The arrows in both images show the position of the impactor. This figure also confirms the front-to-back formation of delamination

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cracks. As the width of Micro-CT sample was 6 mm, thus, a comparable region of the impact location from Model B is presented here. Although there is a variation in the lengths of the delamination cracks, still, the mid interface layers are more damaged in both the scan and simulations. This also corroborates that transverse shear stresses, reaching the maximum at the mid-plane, were probably largely responsible for the greater amount of delamination seen inside the specimen. The delaminated length in the Micro-CT image shown in Fig. 7.12 is larger than that predicted, which may be due to damage properties or the mesh size of the CZEs used in simulations.

![Diagram showing impact force-time relationship](image)

**Fig. 7.13.** Experimental and numerical (Model C) force-time diagrams for CFRP laminates at impact energy of 0.6 J showing response of fractured specimen

### 7.8.3 Response of fractured specimen

Fabric fracture occurred at 78 N as the impact energy was increased to 0.6 J in experimental tests. The predicted evolution of impact force in Model C is compared with experimental results at this level of energy in Fig. 7.13. Here, too, delamination initiation is represented by the load drops designated as $F_{di}$-Exp and $F_{di}$-FE for the experiments and numerical results, respectively. The final fabric failure is presented by a sudden drop in the contact force implying that the
Fig. 7.14. Delamination evolution in Model C with ILSS–I (12 MPa) at impact energy of 0.6 J at 1.5 ms (a), 3 ms (b) and 4.8 ms (c)
impactor has lost its contact with the specimen due to tensile fracture of fibres at the front face (impacted tension side) of the specimen. However, the load did not drop to zero, as the distal plies in compression were not fractured resulting in a residual load of about 22 N. The simulation showed the ply fracture at a higher load of 83 N and somewhat earlier - at 4.38 ms. Such an earlier fracture might be due to the low ply fracture initiation strength defined in the FE Model C. However, the prediction of residual load is almost the same as experimental. In general, the numerical analysis gave a good prediction of the important features of the failure process.

The extent of delamination damage at three interfaces - FCL, MCL and BCL - in simulation with Model C with ILSS–I (12 MPa) at impact energy of 0.6 J is presented in Fig. 7.14 for three different times: at 1.5 ms (just at F_{di}-FE; Fig. 7.14a), at 3 ms (Fig. 7.14b) and at 4.8 ms
(after ply fracture; Fig. 7.14c). The sequence of delamination initiation is the same as in Model B. Delamination progressed more after the first threshold load was reached at $F_{cr}$-FE. Again, the MCL was more damaged due to higher shear stress at the mid-plane of the beam. However, in this model, the edges of the MCL were fractured at 4.38 ms, the damage spread along the delaminated interfaces and most of the interface area was damaged as can be seen in Fig. 7.14c. After the instant of ply fracture, bending stresses caused more damage in the FCL and BCL. The damage area due to delamination as well as ply fracture calculated for Model C is plotted in Fig. 7.15 against the impact duration. Delamination started earlier in the interface planes at both the impact and bending locations. Once more, the mid layer (MCL) at the bending location was more damaged before the ply fracture started. The ply fracture represented by CCL started at 4.2 ms. At 4.38 ms an abrupt increase in the damage area at the impact and bending locations can be observed. Since the localised fracture initiated at the bending location, this region

Fig. 7.16. Comparison of experimental (a) and numerically predicted (b) delamination lengths in Model C at impact energy of 0.6 J
was more damaged than the impact one. After ultimate fracture at 4.8 ms, the
damage growth remained constant in the unloading region of the load-time
response. After the ply fracture, almost every interface was damaged. This
behaviour can also be observed in Fig. 7.16, where the predicted delamination
length is compared with that obtained from Micro-CT analysis. A region of 6 mm
length from the impact location in Model C was selected, the same as the width of
the Micro-CT sample. Both the experimental and numerical analyses gave the
same damage behaviours. The Micro-CT image also display that almost every
interface was damaged after the ply fracture which corroborates the result plotted
in Fig. 7.14c.

![Energy balance during delamination and ply fracture in Model C at
impact energy of 0.6 J showing various dissipative mechanisms](image)

Fig. 7.17. Energy balance during delamination and ply fracture in Model C at
impact energy of 0.6 J showing various dissipative mechanisms

The energy balance for Model C is shown in Fig. 7.17; it demonstrates evolution of
total energy, the kinetic energy, the strain energy, the viscous energy, the artificial
energy and the damage dissipation energy due to delamination and ply fracture
(energy released during crack propagation) as well as the external work done. The
initial kinetic energy of the hammer was converted into the specimen’s strain
energy as the impact progressed. Meanwhile, delamination started at about 1 ms,
dissipating some of the energy represented as DE - Delamination in Fig. 7.17.
Although the cohesive elements at the fracture location were in the damage
process zone at 2 ms, some energy was still absorbed by these elements. When
the plies were fractured at about 4.38 ms, a substantial drop in the laminate strain
energy took place, together with a sudden increase of the fracture dissipation energy (DE - Fracture) to 0.1 J. Since the ply elements moved along the delaminated interface after the fracture, the energy dissipation due to delamination further increased to 0.14 J. Thus, the main energy-dissipation mechanism was delamination after the ply fracture occurred; this is also corroborated by the damage progression in Fig. 7.14c. Similarly, the viscous energy continued to dissipate after the fracture as a result of damping of stress waves in the specimen after fracture. These energy increases resulted in an increase in the total energy, which did not remain constant after the laminate fracture. The strain energy dropped at the fracture point showing some residual energy due to the remaining intact plies of the laminate. However, the kinetic energy showed a slight increase at the fracture point. The increase in the kinetic energy term indicates the importance of transient effects during crack propagation, demonstrating that the unstable crack growth during the ply fracture was dynamic. A similar behaviour of dynamic crack propagation was observed in the numerical analysis of damage interaction in CFRP under quasi-static load in Section 6.3. A small increase in the artificial energy can also be observed in Fig. 7.17 after the ply fracture caused by element distortion.

Fig. 7.18. Evolution of inter-ply and intra-ply damage interaction in bending area in Model C at impact energy of 0.6 J at 4.2 ms (a), 4.38 ms (b), 4.62 ms (c) and 6.0 ms (d) (side view; scaling factor 0.5)

The sequence of inter-ply delamination, intra-ply fabric breakage and interaction of these two modes in Model C is illustrated in the deformed contour plots of damage
shown in Fig. 7.18. Apparently, the interface layers subjected to tensile stress in bending were delaminated before the fabric breakage initiated, as represented by the laminate mid-region in Fig. 7.18a. The first ply’s fracture occurred at 4.38 ms as shown in Fig. 7.18b followed by fracture of the second ply in Fig. 7.18c at 4.62 ms. The third and fourth plies remained intact because of compressive stresses on back side of the specimen in bending. As the load increased, the ply elements moved relative to each other as in Mode-II fracture along the delaminated interface layers as shown in Fig. 7.18d. There may be fibre kinking due to buckling instability in the back plies under compression; however, this behaviour is not studied in these meso-scale models.

Fig. 7.19. Contours of bending stress $\sigma_{11}$ in Model C showing interaction of inter-ply and intra-ply damage at impact energy of 0.6 J at 4.2 ms (a), 4.38 ms (b), 4.62 ms (c) and 6.0 ms (d) (side view; scaling factor 0.5; fractured interfaces are represented by white colour)

Variation of flexural stress $\sigma_{11}$ at various loading intervals along the beam axis in Model C is illustrated in the deformed contour plots shown in Fig. 7.19. Although the laminate was fully delaminated in Fig. 7.18a, the flexural stress contours were almost uniform. The stress is reduced as the first ply’s fracture occurs at 4.38 ms
as shown in Fig. 7.19b followed by fractures of the second ply in Figs. 7.19c at 4.62 ms. The remaining intact third and fourth plies carried the residual load and were in tension and compression states, respectively, as can be seen in Figs. 7.19c and d. Stress concentration at the locations of inter-ply and intra-ply crack tips is apparent in Figs. 7.19b-d. The distribution of flexural stress through the thickness of laminate along line P-P’ in Fig. 7.19 near the fabric fracture location is presented in Fig. 7.20. At 4.2 ms, when the interface is fully delaminated, the stress variation across a single ply thickness is linear and smooth, except at the interface regions, and with similar tension and compression portions. The value of bending stress reduced to zero as the second and third ply fractured (Figs. 7.20b and c). The neutral plane also shifted from its initial position at 0.5 mm thickness to the centre of the remaining intact plies.

Fig. 7.20. Through-thickness distribution of bending stress $\sigma_{11}$ in Model C along line P-P’ in Fig. 7.19 at impact energy of 0.6 J at 4.2 ms (a), 4.38 ms (b), 4.62 ms (c) and 6.0 ms (d)

Contour plots for interlaminar shear stress $\sigma_{13}$ in Model C at 0.6 J impact are shown in Fig. 7.21 at various time intervals. The inter-ply distributions of shear
stress through the specimen’s thickness are also presented (See Fig. 7.22). It is evident from Figs. 7.21a and 7.22a that away from the laminate’s fracture location, shear stress increases toward the middle (i.e. the neutral plane) of the specimen. The transverse distribution of shear stresses is almost parabolic with a maximum value of 12 MPa, equal to shear strength of the interface defined in the model. Sharp spikes of stress at the locations of cohesive layers, i.e. crack tips, are apparent in Fig. 7.21, when the interlaminar failure occurred, as expected. Shear stress changes its sign opposite to that in the plies after fracture of the first ply (Figs. 7.21b and 7.22b). Here, too, the location of maximum shear stress shifted from the initial neutral plane as the ply fracture progressed (Figs. 7.21c, d and 7.22c, d). It is evident that after the first ply fracture, the tensile side of the specimen under bending lost its capacity to carry the shear load near the crack tip.

Fig. 7.21. Contours of interlaminar shear stress $\sigma_{13}$ in Model C at impact energy of 0.6 J at 4.2 ms (a), 4.38 ms (b), 4.62 ms (c) and 6.0 ms (d)

7.9 Conclusions

The numerical models for analysis of the dynamic flexural behaviour of woven CFRP laminates were developed for undamaged, damaged and fracture behaviours observed in the tests at three levels of impact energy. The dominant
damage modes such as delamination and ply fracture and their interaction observed in the microstructural analysis of tested specimens were incorporated in the developed 3D FE models. The numerical models provided a deeper understanding of the transient behaviour of the material under large-deflection dynamic conditions. The pattern of damage formation observed from the Micro-CT analysis (Section 5.4) was also validated by the FE models. Here, too, the damage formation in the specimens at the impact location was from its front to the back in the large-deflection impact tests, unlike to that of back-to-front in drop-weight tests. Further, a good correlation was obtained between the experimental and numerically predicted delamination crack lengths.

The meso-scale 3D FE models of CFRP laminates were developed to simulate an undamaged behaviour as well as some modes of inter-ply and intra-ply damage observed in the woven composites. The observations of damage type and location
in Micro-CT examination served as an input to the FE models. A series of simulations were performed to study the onset, progression and interaction of inter-ply and intra-ply damage processes under mixed-mode impact bending by employing multiple layers of CZEs in the FE models of the damaged and fractured specimens. The transverse bending fracture of the laminate was also modelled with CZEs instead of the traditional CDM approach. To accurately predict the onset and growth of the damage mechanisms, refined meshes were introduced, especially in the identified damage regions, which required a high computing power to run the simulations. The accuracy of the FE models was determined by comparing experimental results for contact-force histories with respective numerical predictions for several impact energies. The obtained numerical results were close to the experimental ones, although micro-cracking (e.g. matrix cracks) and fibre buckling were not accounted for by the meso-scale modelling approach. The cohesive-based numerical models demonstrated their capability to reproduce the complex three-dimensional damage patterns observed experimentally in impacted composite laminates. The load level, at which delamination initiated, was under-predicted, whereas the ultimate load drops due to fabric rupture showed good correlation with the predicted failure load. The type, location and extent of damage with respect to impact time were also identified by the models. The numerical simulations helped to gain a better understanding of the complex damage phenomena that occurred at various stages of impact in multi-body dynamics. For instance, interaction of delamination and ply fracture, which cannot be assessed in real tests and by non-destructive inspection, was captured by the discrete modelling approach presented in this study. Similarly, splitting of the total energy of the model resulted in clear understanding of the role of each energy dissipative mechanism. It was also observed that a certain number of delamination areas propagated simultaneously at the threshold corresponding to the first load drop in the force-time response. Based on the results of simulations of CFRP specimens with longitudinal and transverse cohesive layers, it is concluded that the models represented reasonably well the onset and propagation of delamination, coupling between delamination and ply cracking and the final fracture of the laminate as a result of fabric fracture in impact bending. The proposed modelling approach for interacting damage modes can serve as a benchmark for modelling damage coupling in composite laminates efficiently in a
computationally cost-effective manner as compared to CDM. The simulations also confirm that FE models based on the appropriate constitutive material behaviour can serve as a powerful tool in design and validation of damage-tolerant composite structures subjected to dynamic loading.
Chapter 8: Conclusions and Future Work

8.1 Summary

The overall aim of this research was to study the deformation and fracture behaviours of carbon fabric-reinforced polymer (CFRP) composites under bending loads. The research was motivated by a need to develop further understanding of the mechanical behaviour of these materials, as a step in improving the predictability of their response under flexural loading scenarios. This is necessary to advance their implementation in a range of structural applications, where they are subjected to large bending deformations, especially in sports products. The aim was achieved by conducting a critical literature review followed by experimental and numerical studies. Several types of experiments were performed in order to quantify the mechanical properties of the material under quasi-static and dynamic loading regimes for different orientations. Those experiments were carried out for two purposes: (1) to characterise its mechanical properties in order to develop finite-element models to simulate the deformation and damage of CFRP under flexural loading scenarios; (2) to investigate the load-bearing and energy-absorbing capability of the material under the foreseeable in-service conditions. The ensued damage mechanisms in the composite laminates were investigated using optical microscopy and Micro-CT techniques. Based on the results of experiments and microstructural examination, a number of finite-element models were developed using Abaqus, in order to analyse large-deflection deformation and fracture of CFRP. Among those models are (i) 2D FE models used to simulate its multiple delamination scenario under quasi-static large-deflection bending, (ii) 2D FE models for analysis of experimentally observed inter-ply delamination, intra-ply fabric fracture and their subsequent interaction under quasi-static bending conditions, and (iii) 3D FE models based on multi-body dynamics used to simulate interacting damage mechanisms in CFRP under large-deflection dynamic bending conditions for the Izod impact test setup. The developed models provided a better understanding of the deformation and damage behaviours of the material, and, most importantly, they adequately replicated the experimental data. It is suggested that with careful consideration of material’s input parameters, the FE models could have a potential for use as a
design tool for applications that involve bending loads. Following the objectives of this study (see Chapter 1), the PhD research has brought the results described below.

8.1.1 Experimentation

Quasi-static mechanical behaviour

The material’s mechanical behaviour under quasi-static loading conditions was characterised using tensile, in-plane shear and bending tests. The tests were performed on specimens of different orientations to determine the material’s elastic constants as well as their strengths. The experiments resulted in the following conclusions:

- The on-axis specimens demonstrated almost similar mechanical properties in warp and weft directions due to the symmetry of yarns in both directions. The warp and weft specimens exhibited a quasi-brittle linear behaviour before their ultimate fracture in both the tensile and flexural tests.
- Based on comparison of tensile and flexural tests, it can be concluded that the flexural strength of the CFRP was greater than the tensile strength. Therefore, it is not recommended to use tensile material properties to study the bending behaviour of composites.
- The off-axis \([+45/-45]_{2s}\) laminates showed a significantly nonlinear behaviour due to matrix cracking and fibre trellising in the in-plane tensile shear tests. In these tests, DIC proved to be a very useful non-contact technique to acquire the full-field strain data for determination of the material’s in-plane shear properties.
- Both the warp and weft specimens demonstrated a strain-rate independent behaviour while tested at various indenter speeds under flexural loading. The material also exhibited a size effect when the flexural behaviour of specimens of various thicknesses was examined.
- The bending tests were conducted under large-deflection conditions; however, this nonlinearity was not reflected in the stress-strain curves due to the stress stiffening effects caused by high deflection in thin laminates.
- The microscopic analysis of the fractured specimens asserted that the low stressed regions showed weft yarn longitudinal cracking, whereas the high
stressed regions underwent inter-ply delamination and ply fracture. The internal damage mechanisms in the tested specimens observed by means of X-ray Micro-CT scanning were matrix cracking, tow debonding, delamination and ply fracture. However, the Micro-CT technique was limited to provide full-scale damage information of the tested specimens.

**Dynamic mechanical properties**

In order to investigate the dynamic behaviour of CFRP laminates under large-deflection bending, experimental tests were performed using the Izod impact test setup. Two types of laminates in warp and weft directions were subjected to low levels of impact energies up to their ultimate fracture. The dynamic behaviour of the material was evaluated by analysing the impact load–time as well as load–deflection responses. Also, Micro-CT analysis of CFRP specimens was carried out in order to gain adequate knowledge of their damage mechanisms. Based on the obtained results, the following conclusions can be formulated:

- In dynamic bending tests of both types of the specimens, the impact force increased with the impact energy but had a little effect on the contact duration. Hence, higher impact velocities induced larger deformations and therefore, larger impact forces. This also implied that the impact velocity dominated the impact energy as the impactor mass was the same in all the tests. Further, the impact duration remained almost the same with increase in the impact velocity for each type of specimens, which indicated the strain rate-independent behaviour of the material in fibre-dominated modes under dynamic loading, as was also observed under quasi-static loading.
- The load-time and load-deflection responses obtained from impact testing also provided important information concerning the damage initiation and growth. At sub-critical energy levels such as 0.5 J and 0.4 J, the load drops before the peaks corresponded to damage initiation in warp and weft specimens, respectively. The ultimate fracture of both types of specimens occurred at the critical energy level of 0.6 J, that, was characterised by a sudden load drop in the force and energy history plots.
- Due to their lower stiffness and strengths, weft specimens showed earlier damage initiation and lower peak loads than the warp specimens. Also, the
contact duration was somewhat longer for the weft specimens than that for the warp specimens. In the cantilever-type dynamic bending tests, the warp and weft specimens experienced large bending deformations such as 9.5 mm and 10.5 mm, respectively, at energy level of 0.5 J.

- Thin laminates failed at low energy levels in dynamic bending tests as compared to their failure in drop-weight impact tests, which may be due to less stress-stiffening effects because of less constrained edges of the specimen in cantilever bending.
- Micro-CT analysis of the tested specimens showed that matrix cracking, delamination and tow debonding were the dominant damage modes at the specimen’s impact location, whereas at the bending location, these modes were coupled with fabric fracture.
- A substantial difference in damage patterns was found at the impact location between the non-penetration dynamic bending and penetration type drop-weight tests for the first time. In specimens subjected to drop-weight tests, the damage formation is usually from the bottom to the top in a conical form with the back face experiencing fibre fracture and splitting due to tensile loading. However, in this study, it was distributed almost uniformly at every ply interface, and only minor dents were observed on the impacted face of the specimens.

8.1.2 Simulations

Quasi-static numerical models

Following the experimental testing and microstructural examination of CFRP composites, the observed damage behaviour under quasi-static bending was studied by developing 2D plane-strain FE models in Abaqus. A series of simulations was performed, first to study the initiation and evolution of inter-ply delamination process under mixed-mode bending and then the interaction of inter-ply and intra-ply damage modes observed in the specimens under similar loading conditions. Based on the results obtained from the numerical simulations, the following conclusions can be drawn:
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- The simulation results exhibited a good agreement with the experimental ones, and the numerical models A, B and C had the capability to reproduce the failure mechanisms in composite laminates. The interlaminar layers at the top and bottom of the laminate specimens experienced Mode I failure whereas the central layers exhibited the Mode II failure behaviour.

- The FE models helped to gain a better understanding of the damage initiation and evolution processes in woven laminates. Numerical simulations demonstrated that damage initiation and growth was sensitive to the size of cohesive-zone elements.

- It was observed that all of the modelled interlaminar layers in the finite-element schemes represented zones of damage nucleation and propagation identified in the microstructural damage analysis. The results indicated the suitability of the developed numerical approach to study the onset and propagation of interlaminar damage.

- In Models D and E of damage interaction, the obtained numerical results were close to the experimental ones, and the numerical models had the capability to reproduce the damage sequence and pattern observed experimentally in composite laminates. The novel CZM-based models accurately represented the onset and propagation of delamination, coupling between delamination and ply cracking and the final fracture of the laminate as a result of fabric fracture.

- A good agreement was obtained between the length of delamination cracks predicted numerically and that determined from the Micro-CT analysis in the tested specimen. The predicted and experimental load level, at which the fabric failure occurred, also showed a good agreement.

- The numerical analyses showed that dynamic effects were significant during ply crack propagation even under quasi-static loading.

- The numerical simulations provided important information such as stress and strain distributions, variation of crack lengths with applied load and interaction of various damage modes in composite specimens, which is hard to obtain with quasi-static large-deflection bending tests.
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Dynamic numerical models

In this part of the study, various finite-element models were formulated for the Izod test setup to investigate the dynamic behaviour of CFRP material under large-deflection bending. Three different models - A, B and C - were developed for three levels of impact energy representing undamaged, damaged and fracture behaviours observed in the tests. The dominant damage modes observed in the Micro-CT analysis of tested specimens such as delamination and ply fracture and their interaction were incorporated in the developed 3D FE models. The damage interaction was modelled by employing the CZM-based approach instead of the traditional CDM-based one. The numerical models provided a deeper understanding of the transient behaviour of the material under large-deflection dynamic conditions. Based on the obtained results from the numerical simulations, the following conclusions can be formulated:

- The obtained simulation results for CFRP composites under dynamic bending were close to the experimental ones, validating the meso-scale modelling approach. The accuracy of the FE models was determined by comparing the experimental results for contact-force histories with respective numerical predictions for several impact energies.

- The numerical models demonstrated their capability to reproduce the complex three-dimensional damage patterns and their interaction observed experimentally in impacted CFRP specimens. The load level at which delamination initiated was under-predicted, whereas the ultimate load drops due to fabric rupture showed a good agreement with the predicted failure load.

- Detailed information about the damage and fracture processes was obtained using the FE models that enhanced understanding of evolution of these processes. For instance, certain behaviours such as interaction of delamination and ply fracture, and splitting of the total energy of the model for identification of each damage mechanism, which were not possible experimentally, were captured by the FE models. The type, location and extent of damage with respect to impact time were also identified by the models.
The FE models also validated the pattern of damage formation observed with the Micro-CT analysis. Here, too, the damage formation in the specimens at the impact location was from the front to the back similar to that observed with Micro-CT scanning. Further, a good correlation was obtained between the numerically predicted and experimentally delaminated crack lengths.

The simulations also demonstrated that a certain number of delamination areas propagated simultaneously at the threshold corresponding to the first load drop in the force-time response. The growth of delamination area at the impact location was strongly dependent on the interlaminar shear strength (apart from the mesh size). The multiple delamination scenario was similar to that observed in the Micro-CT analysis.

Based on the results of CZM-based FE models, it was found that the models represented reasonably well the onset and evolution of delamination, coupling between delamination and ply cracking and the final fracture of the laminate as a result of fabric fracture in impact bending.

8.2 Conclusions

The overall major conclusions can be summarised as follows:

- The material characterisation revealed that for applications demanding strength and stiffness, on-axis CFRP composites is a better choice. Where energy absorbing capability is required, the off-axis CFRP laminates are favourable candidates. However, the applications demanding strength, stiffness and energy absorbing as in sports products, a combination of both types of plies as in a quasi-isotropic lay-up would be an optimum choice.

- A better understanding of mechanical behaviour, deformation and damage modes of CFRP composites under quasi-static and dynamic loading conditions is gained in this research which is required for damage-tolerant design of CFRP laminates for applications in foreseeable loading scenarios. The knowledge of novel type of damage formation pattern under dynamic bending can also be used in tailoring the composite design under such conditions.
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- The developed meso-level numerical models for simulation of deformation and damage of composites would provide insights and quantitative exploration of design space and would substantially reduce the costly experimental testing thus reducing the product design time.

- The simulations also confirm that FE models based on the appropriate constitutive material behaviour could serve as a powerful tool in design and validation of damage-tolerant composite laminates subjected to various loading conditions in sports products.

- The proposed CZM-based modelling scheme for interacting damage modes can serve as a benchmark for modelling the coupling of various damage modes in composite laminaes explicitly and accurately as compared to the CDM approach. The approach presented in this thesis has the flexibility and potential to be applied to the analysis of complex interaction of damage mechanisms in composite materials under any type of loading conditions.

8.3 Future work

The need for further work follows from the results and outcomes of this thesis. In the next sections, some future investigations are suggested for experimentation and simulations.

8.3.1 Experimentation

- The effect of varying thickness of on-axis laminates should be examined in more detail under quasi-static and dynamic flexural loads using a greater range of thickness and dimensions. This would be useful in further understanding the size effect on the behaviour of CFRP laminates.

- The in-plane material properties were obtained in this study. An experimental scheme should be developed to determine the material’s out-of-plane properties.

- Although it is established that the behaviour of CFRP is strain-rate independent in the fibre dominated mode, it should still be investigated at higher strain rates under large-deflection bending as well.

- This thesis focused on the study of material’s behaviour under single impact. The behaviour of CFRP under multiple impacts as experienced by sports products in service conditions should be investigated. The material’s
behaviour should be investigated by developing diagrams of fatigue strength versus life and damage versus number of impacts.

- Experimental work should be conducted for characterisation of the fracture toughness properties of the composite in the failure Modes I, II and a mixed-mode regime.
- The material’s behaviour of off-axis laminates should also be further studied under quasi-static and dynamic large-deflection bending.
- Apart from optical microscopy and Micro-CT, ultrasonic C-scan technique could be used to obtain the damage picture of the full-scale specimens for comparison with simulations.

8.3.2 Simulations

- Meso-level 3D FE models should be developed for CFRP specimens under quasi-static large-deflection bending, capable of simulating various damage modes and their interaction.
- A multi-scale modelling approach should be adopted for simulating damage mechanisms in CFRP laminates under quasi-static and dynamic bending loads. Although, development of full-scale models may be prohibited by currently available level of computational power, localised modelling at the scale of yarns and matrix will be possible at the indenter position in quasi-static bending, and impact and fracture locations in dynamic bending simulations.
- The CZM-based modelling approach could be extended to handle multiple damage modes and their interactions in the multi-scale FE models such as matrix cracking apart from delamination and ply fracture.
- For multi-scale modelling, the architecture and geometry of woven laminate can be developed e.g. with an open source code TexGen. The geometric details of yarns can be obtained from the Micro-CT analysis. The geometric model can be imported in Abaqus to develop microstructural FE model of the specimen. The improvements of the multi-scale models should go side by side with improvements in computing power.
- Developing a modelling strategy for impact fatigue and the subsequent damage in the material using FE simulations.
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- The reasons for underestimation of delamination thresholds in the FE models of dynamic bending should be further investigated.
- Models for the nonlinear behaviour of off-axis laminates capable of coupling the observed plasticity and damage could be developed.
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