

**THE PERFORMANCE OF THERMOPLASTIC
AND THERMOSET COMPOSITES SUBJECTED
TO IMPACT LOADING**

by

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Abstract

Carbon Fibre Reinforced Polymer (CFRP) composites play an increasingly important role in the aerospace industry where CFRPs are extensively employed in the primary structures of aircrafts. These composite materials have demonstrated relatively high stiffness-to-weight and strength-to-weight ratios, as well as good fatigue and corrosion resistance, compared with the conventional metallic materials. However, these composite primary structures of aircrafts face great challenges when encounter with an impact with foreign bodies. Due to the anisotropic property and interaction between fibres and matrix, the failure of composite structure is generally a complex process, which requires a further investigation and better understanding.

Bearing the above in mind, the aim of this project is to investigate the modes and extent of damage inflicted on the unidirectional carbon fibre and woven carbon fibre reinforced thermoplastic and thermoset composites, under various impact loadings. A series of low-velocity (drop-weight) and high-velocity (gas-gun) experiments were firstly performed, using rigid impactors at different energy levels, on the unidirectional Carbon Fibre Reinforced Poly (ether-ether-ketone) (CF/PEEK) and toughened Epoxy (CF/Epoxy) cross-ply composites. Particularly, for the high-velocity experiments, the three-dimensional (3D) Digital Image Correlation (DIC), coupling with high-speed cameras, is employed to measure the out-of-plane displacement from the rear face of the composite specimens during the impact events. The ultrasonic C-scan results are utilised for assessing damage in the composites and reflecting the strain rate effects on the impact response of the composites. A finite element model, based on the continuum shell elements, is developed to simulate both the low- and high-velocity impact test results. The modelling results are quantitatively validated against the experimentally obtained impact behaviour of the composites, such as the loading response, interlaminar and intralaminar damage etc. The experimental validation enables the model to be used with confidence in the future industrial applications.

To better understand the failure mechanisms of woven-fabric composites, woven-fabric CF/PEEK and CF/Epoxy specimens are employed to study the behaviour of woven composite laminates subjected to impact loading by soft gelatine and hard metallic projectiles to represent the impact of foreign object, such as a small bird, hail-stone or runway debris. For this investigation, both the 3D DIC measurements and C-scan assessments are employed in the experiments performed on the CF/PEEK and CF/Epoxy specimens. A FE model is developed, based on Smoothed Particle Hydrodynamics (SPH) method, Abaqus built-in Hashin's criteria and bilinear cohesive law, to predict the viscoelastic-plastic fluid behaviour of the gelatine project and the behaviour of composite specimens during the impact events. Good agreement is shown between the predictions from using the FE model and the experimental results. The FE model has produced accurate predictions of the deformation of the gelatine projectiles, the major strain and out-of-plane displacement of the composites, contact pressure and the damage degrees resulting in the rear-face of the composite specimens. The dynamic characteristics of the thermoplastic and thermoset composite laminates, investigated in the project through a series of experimental tests and numerical predictions, can assist the design of lightweight composite structures for energy-absorbing applications.

Declaration of Originality

I hereby affirm that the work presented in this thesis was produced by me during the course of my Ph.D. studies at Imperial College London. Where contributions by others are used, they have been appropriately referenced in the text and in the bibliography. No part of the work presented in this thesis has been submitted elsewhere for another degree or qualification.

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Nomenclature

d	General form of the damage variable
d_f	Damage variable in fibre damage
d_f^t	Damage parameter in tensile fibre failure
d_f^c	Damage parameter in compressive fibre failure
d_m	Damage variable in matrix damage
d_m^t	Damage parameter in tensile matrix failure
d_m^c	Damage parameter in compressive matrix failure
d_s	Damage variable in shear damage
k	Stiffness of cohesive law
m	Mass of projectile
p	Positive pressure in compression
p_H	The Hugoniot pressure
s	Slope of the linear relationship between U_s and U_p
t	Time
t_e	Thickness of an adjacent composite ply
t_i^0	Initial cohesive law strengths ($i = 33,31,32$)
v_i	Initial impact velocity
v_r	Residual velocity
C_d	Damaged elasticity matrix
E_{ii}	Elastic modulus in the fibre or transverse directions ($i = 1,2$)
E_a	Kinetic energy absorption
E_m	Internal energy per unit mass
E_H	Specific energy per unit mass
F_f^t	Tensile damage criterion in fibre failure

F_f^c	Compressive damage criterion in fibre failure
F_m^t	Tensile damage criterion in matrix failure
F_m^c	Compressive damage criterion in matrix failure
G_{12}	Shear modulus in the longitudinal-transverse direction
G_c	Interlaminar fracture energy
G_I	Mode I energy-release rate
G_{Ic}	Interlaminar Mode I fracture energy
$G_{Ic} _{ft}$	Tensile intralaminar fracture energy in the longitudinal direction
$G_{Ic} _{fc}$	Compressive intralaminar fracture energy in the longitudinal direction
$G_{Ic} _{mt}$	Tensile intralaminar fracture energy in the transverse direction
$G_{Ic} _{mc}$	Compressive intralaminar fracture energy in the transverse direction
G_{II}	Mode II energy-release rate
G_{IIc}	Interlaminar Mode II fracture energy
S_d	Damage area
S^T	Shear strength in the transverse direction
S^L	Shear strength in the longitudinal direction
U_p	Particle velocity of the projectile
U_s	Shock-wave velocity
X^T	Tensile strength in the longitudinal direction
X^C	Compressive strength in the longitudinal direction
Y^T	Tensile strength in the transverse direction
Y^C	Compressive strength in the transverse direction
α	A constant in the cohesive law
σ	Stress
σ_n	Stress in the normal direction of the composite ply
σ_n^0	Strength in the normal direction of the composite ply
$\hat{\sigma}$	Effective stress in the composite ply
$\hat{\sigma}_{ii}$	Effective stress in the longitudinal or transverse directions ($i = 1,2$)
τ_s	Shear strength of the composite ply
τ_s^0	Shear strength of the cohesive surface
τ_t	Tensile strength of the composite ply
τ_t^0	Tensile strength of the cohesive surface

NOMENCLATURE

$\hat{\tau}_{12}$	Effective stress tensor in the longitudinal-transverse shear direction
ν_{ij}	The Poisson's ratio ($i, j = 1, 2, i \neq j$)
ε	Strain
ε^0	Initial failure strain
ε^f	Final failure strain
ε_{ii}	Strains in the longitudinal or transverse directions ($i = 1, 2$)
ε_{ii}^0	Initial failure strains in the longitudinal or transverse directions ($i = 1, 2$)
ε_{ii}^f	Final failure strains in the longitudinal or transverse directions ($i = 1, 2$)
ε_{ii}^{ft}	Tensile final failure strain in the longitudinal or transverse directions ($i = 1, 2$)
ε_{ii}^{fc}	Compressive final failure strain in the longitudinal or transverse directions ($i = 1, 2$)
l_c	Characteristic length of the element
δ	Displacement
δ^o	Displacement of damage initiation
δ^f	Displacement of element at failure
η	Nominal volumetric compressive strain
η_{BK}	Benzeggagh-Kenane coefficient mode-mix exponent
Γ	The Grüneisen ratio
Γ_0	A material constant to define the Grüneisen ratio
ρ	Density of the gelatine projectile
ρ_0	Reference density of the gelatine projectile

Abbreviation

3-D	Three-dimensional
APC	Aromatic Polymer Composite
ASTM	American Society for Testing and Materials
B-K	Benzeggagh-Kenane
CAI	Compression-After-Impact
CDM	Computational Damage Mechanics
CF/Epoxy	Carbon Fibre Reinforced Epoxy
CF/PEEK	Carbon Fibre Reinforced Poly (ether-ether-ketone)
CFRP	Carbon Fibre Reinforced Polymer
CT	Computed Tomography
DA	Damage Area
DIC	Digital Image Correlation
EPFM	Elastic-Plastic Fracture Mechanics
FAA	Federal Aviation Administration
FE	Finite Element
FOD	Foreign-object Damage
FRP	Fibre Reinforced Polymer
FVF	Fibre Volume Fraction
HDPE	High-Density Poly-ethylene
IE	Impact Energy
ISO	International Organization for Standardization
KEA	Kinetic Energy Absorption
NDT	Non-destructive Testing
OOP	Out-of-plane
PEEK	Poly-ether-ether-ketone

PPS	Poly-phenylene-sulphide
PTFE	Poly-tetra-fluoro-ethylene
SPH	Smoothed Particle Hydrodynamics
UD	Unidirectional
VCC	Virtual Crack Closure

Chapter 1. Introduction

1.1 Background

Since carbon fibre and Kevlar fibre were developed in the last few decades, composite materials have played an increasingly important role throughout industrial development. The usage of composite materials has permeated into daily lives and been expected to contribute to a tremendous global cross-sector market of \$105.8 billion in 2020 [1]. The outstanding performances, such as high strength-to-weight ratio, no corrosion, excellent fatigue resistance, temperature tolerant to excessive heat, low thermal expansion [2–7], of composite materials prompted such materials as competitive alternatives to the conventional materials, i.e. metallic materials. Nowadays, composite materials are critical to many industries and have been extensively used in numerous scientific and commercial applications, especially in transportation industry, where the weight of the structure has direct influence on operating cost and performance. Example applications in railway, automotive and aerospace industries are shown in Figure 1-1 [8–10].

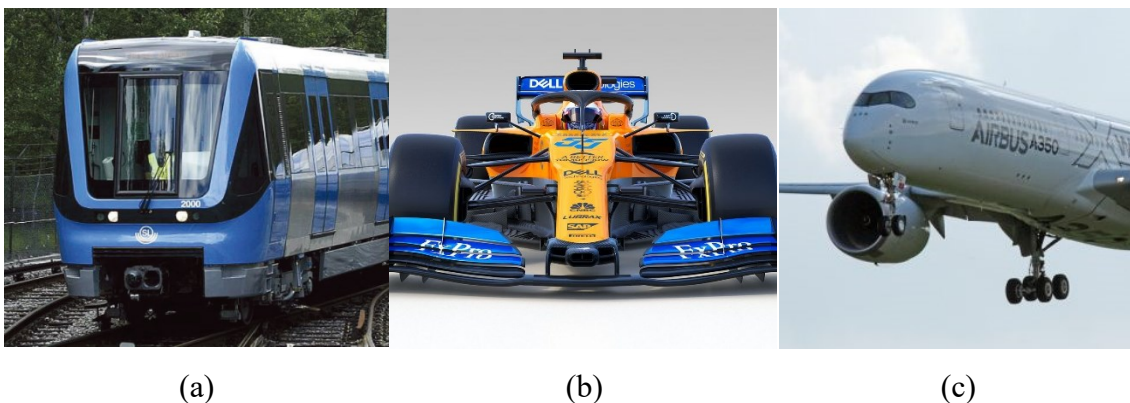


Figure 1-1. Applications of composite materials: (a) Bombardier Transportation C20 metro train with composite car body; (b) a Formula One car with carbon fibre chassis; (c) Airbus A350-XWB aircraft employed 53 wt% of composite materials [8–10].

In aerospace industry, as shown in Figure 1-2 [10,11], the use of composite materials on aircrafts has steadily grown since Airbus applied such materials on the secondary structures of the world's first twin-engine widebody aircraft A300. Until now composite materials have been utilised extensively within primary structures of aircraft such as outer wing and fuselage. After nearly a century of dominance as the material of choice for aircraft, metals are seeing increased competition from composite materials in use on such aircraft as Airbus A350 XWB and Boeing Dreamliner 787 jetliner, of whose composite materials account for 53 wt.% [10] and 50 wt.% [11], respectively. Thanks to the use of composite materials on such aircrafts, it is not only improving flight performances but also reducing global environment impact.

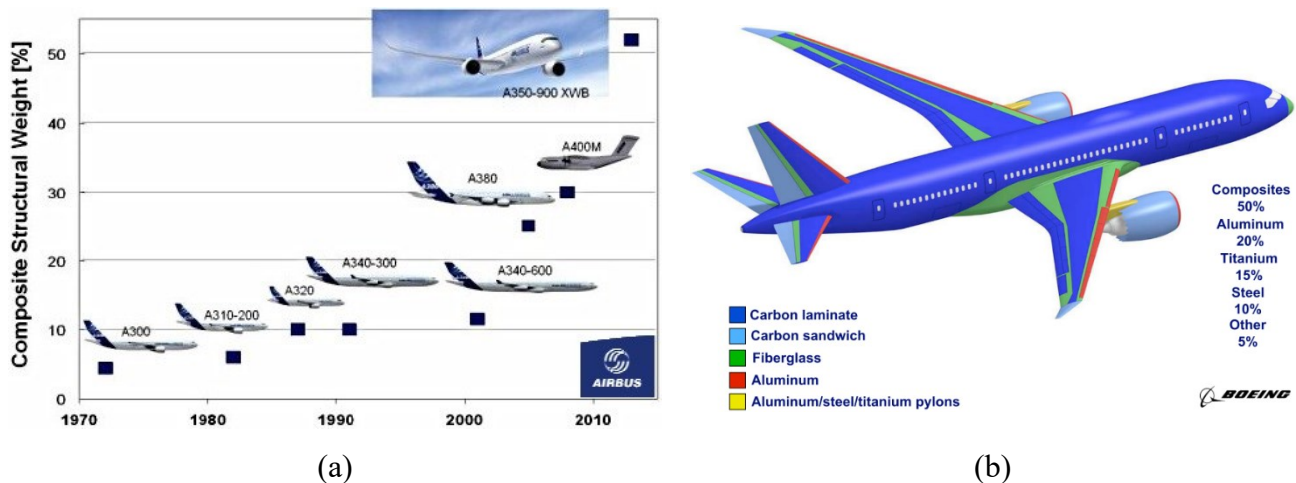


Figure 1-2. Composites in aerospace industry: (a) composite structural weight development on Airbus aircrafts; (b) composite solutions applied throughout the Boeing Dreamliner 787 [10,11].

The great performance of composite materials supports the revolution of commercial aircraft. Correspondingly, the need from aerospace industry has significantly drove various types of polymer matrix to be invented and developed. In broad generalities, thermoset matrix composites are more predominant than thermoplastic composites for a long history, because they frequently have lower raw material costs and are easy to be processed and formed into complex geometries. However, in recent years, thermoplastic composites have an increasing market share in the aerospace industry, due to their excellent properties of higher strength and impact resistance, better recyclability and eco-friendly manufacturing [5–7]. In other word, they offer aircraft designers an opportunity to utilise less material to satisfy the structural strength requirement so that the overall weight of an aircraft can be further reduced.

However, the damage mechanisms of these thermoplastic composite, subjected to various damage loading, i.e. impact loading, are not fully understood yet [12–16]. Foreign-object debris (FOD) damage, one of the most critical damage mechanisms to thermoplastic composites, poses serious threats to aircraft, as such damage can, lead to damage which may not be necessarily visible but which may, nonetheless, lead to a significant structural strength reduction. Additionally, in the worst case, lead to major structural damage in critical aircraft components, as in Figure 1-3 [17,18]. The US Federal Aviation Administration (FAA) stated in July 2019, that there were 214,048 bird-strikes reported to the FAA from 1990 to 2018 [15]. More than two-hundred and sixty three aircraft were seriously damaged, and over two-hundred and eighty two people were killed due to wildlife strikes from 1988 to 2018 [15]. In the case of runway debris hazards, the Boeing reported that the resulting damage to airplanes, equipment and personnel is estimated to cost approximately four billion dollars annually [19].



Figure 1-3. Aircraft structures collapsed after (a) hail impact, or (b) bird-strike [17,18].

As a result of the increasing number of these incidents, more attentions have been paid to the impact damage resistance of these structures on aircrafts which are constantly threatened by FOD under not only low and high velocity impact but also the soft and hard impact. Both the damage mechanisms and structural integrity need to be studied experimentally and numerically to obtain a better understanding of the impact behaviour of composite materials, which will then contribute to the wider application of composite materials in the latest generation of aircraft.

1.2 Aims and objectives

In this Ph.D. project, the main aim is to seek experimental and numerical methods to evaluate the performances and investigate the damage mechanisms of thermoplastic and thermoset composites under impact loading, which can be finally employed in the future industrial applications, to assist the design and maintenance of composite structures. To achieve the objectives, the research programme has been divided into the distinct tasks, which are detailed as below.

- i. To conduct a comprehensive literature review and learn about the state-of-the-art in impact testing and modelling on polymer composites.
- ii. To design and manufacture thermoplastic poly(ether-ether ketone) and thermoset epoxy composite specimens to study mechanical behaviours under impact.
- iii. To investigate the impact performance of unidirectional and woven-fabric composite specimens under low velocity (i.e. $<10 \text{ m.s}^{-1}$) and high velocity (i.e. up to 100 m.s^{-1}) impact.
- iv. To study the perforation behaviour of such composites impacted at high-velocity using soft and hard impactors.
- v. To develop a high-fidelity computational model to predict the impact behaviour of composite structures under such impact.
- vi. To quantitatively analyse the loading responses and failure mechanisms of the investigated thermoplastic and thermoset composites based on the results obtained from the impact experiments and simulations.

1.3 Structure of the thesis

The chapters in this thesis are outlined as follows.

Chapter 1 introduces the background, the objectives and the structure of this Ph.D. project.

Chapter 2 addresses the fundamentals of fibre-reinforced composite materials, including classification, fibre reinforcements, polymer matrices, manufacturing methods and typical failure modes, as well as the advanced impact performance assessment methods,

including characterisation testing methods, non-destructive testing (NDT) methods and numerical modelling of fibre composites.

Chapter 3 introduces the unidirectional and woven-fabric carbon fibre composites employed that are based upon a thermoplastic poly(ether-ether ketone) matrix (termed CF/PEEK) or a thermoset epoxy matrix (termed CF/Epoxy), detailing their mechanical properties and manufacturing processes, and the experimental equipments and technologies, including drop-weight tower, gas-gun, three-dimensional (3D) Digital Image Correlation (DIC) and ultrasonic C-scanning.

Chapter 4 details the low velocity and high velocity impact experiments that are performed using a rigid impactor, at different energy levels, using unidirectional thermoplastic and toughened-thermoset composites which possess the same cross-ply lay-up. All impacted specimens are inspected using ultrasonic scan device to evaluate the impact damage. The measured loading response, energy dissipation and damage area are employed to compare the impact performance of these tested thermoplastic and thermoset composite specimens.

Chapter 5 presents an elastic, two-dimensional finite-element (FE) model, which is computationally efficient, for simulating both the low velocity and high velocity impact tests on the thermoplastic and thermoset composites in Chapter 4. The modelling results are quantitatively validated against the experimentally-measured impact responses of the tested composites.

Chapter 6 investigates the performance of woven-fabric carbon fibre composites under high-velocity impact loading at impact velocities of up to 100 m.s^{-1} . The composites are impacted using soft and hard projectiles. Both visual inspection and C-scanning examination are conducted to assess the extent of impact damage in the tested specimens. The perforation resistance and energy absorbing capability of the woven thermoplastic and thermoset composites are studied by performing high-velocity impact experiments using the hard-projectile and the resulting modes and extents of damage are identified.

Chapter 7 presents experimental and numerical studies on the behaviour of woven-fabric composite laminates subject to impact loading by soft gelatine projectiles. High velocity impact experiments are performed to investigate the effect of the matrix system on the impact response of the composites. A FE model is developed to simulate the impact by a soft projectile on the composite specimens. Good agreement is shown between the predictions from using the FE model and the experimental results.

Chapter 8 summarises the main outcomes of this Ph.D project.

Chapter 9 suggests the future work to continue this study.

Chapter 2. Literature review

2.1 Introduction

In this chapter, the fundamentals of composite materials, including classification, fibre reinforcement, polymer matrix, fabrication methods, as well as an overview of the past and current research work on the composite materials. This overview addresses the impact performance of the composite materials under low-velocity and high-velocity impact loading and introduces the typical failure modes shown by composite materials. This is followed by presenting the Non-Destructive Testing (NDT) methods for assessing the structural damage occurred in composites. The numerical approaches for high-fidelity modelling of damage in composites, for enabling the predicting of impact behaviour, are discussed at the last part.

2.2 Composite materials

2.2.1 Classification of composite materials

Composite materials are composed of two or more constituent materials with significantly different advantageous properties that, when combined, produce a material with characteristics different from the individual components. Among the constituent materials, one is called reinforcement which provides strength and stiffness of composite, and the one in which it is embedded is called matrix which gives overall shape and contribute to the toughness of composite [4,7]. Such constituents remain their identities and possess an physically identical interface between each other in the composite. A typical structure of composite is illustrated in Figure 2-1.

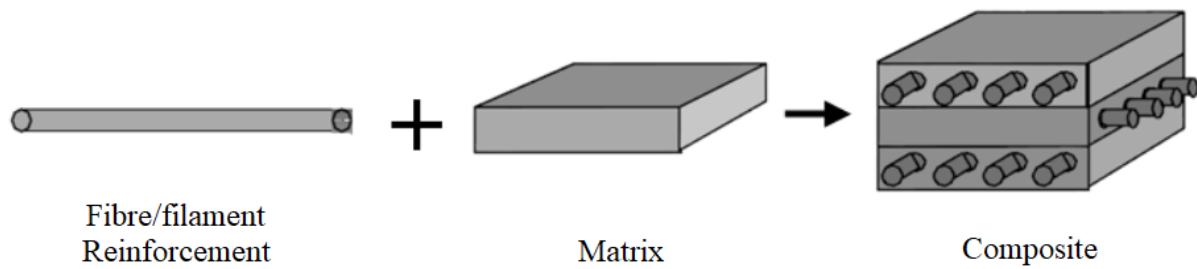


Figure 2-1. Typical structure of composite [4].

Composites are generally classified by either the geometry of the reinforcement or the type of matrix. Particulate composites, flake composites and fibre composites are commonly classified by the geometry of the reinforcement and designed for various application purposes [20]. Among these composites, due to the high aspect ratio between length and diameter of fibres, which can provide effective shear stress transfer between the matrix and the fibres, fibre reinforced composites are most widely used in the industrial applications and are emphasised in this thesis. Such composites can be further classified by the type of matrix, e.g. polymer, metal and ceramic [21]. Considering the excellent properties of high strength and friendly manufacturing principles, polymer matrix composites are the most common advanced composites in industries [21].

2.2.2 Fibre reinforcement

2.2.2.1 Types of fibres

In a composite structure, fibre reinforcement gives strength, stiffness and other mechanical properties to the composite, and dominates other properties, e.g. the coefficient of thermal expansion, conductivity, and thermal transport. In the history of fibres, natural fibres were the first uses of fibre reinforcements when Egyptians and Mesopotamians combined straw and mud for building construction. Along with the arrival of modern era, plastics were developed and further improved by Slayter Games to enhance the strength and rigidity of composite materials with glass fibres [22]. Since carbon fibre (also known as graphite fibre) and aramid fibre (known as Kevlar) were developed in 1970s [23,24], Fibre Reinforced Polymer (FRP) became the indispensable part of composite material system. In present time, various types of carbon fibre, glass fibre and aramid fibre are most commonly used fibres which have different properties, for example, as shown in Figure 2-2.

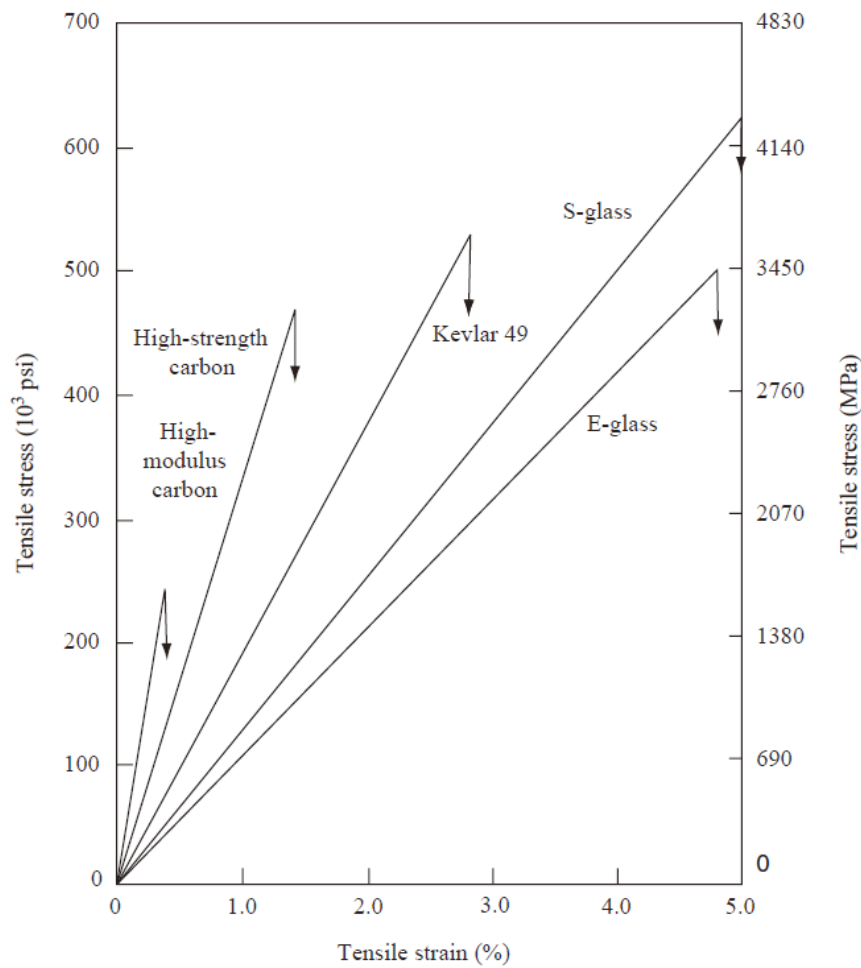


Figure 2-2. The diagram of tensile stress-strain for various reinforcing fibres [3].

Among these fibres, glass fibre is the most common fibre applied polymer matrix composites. The advantages of glass fibre include low raw material cost, relatively high strength, high toughness, excellent heat and corrosion resistance [25]. In contrast to glass fibre, carbon fibre exhibits great improved strength and stiffness at a lower weight, but with a higher material cost. Such fibre contain at least 90 wt.% of carbon and is produced from polymeric precursor materials. From a macro perspective, carbon fibre is very thin filament with a diameter of approximately 5-10 μm , which can transfer shear stress effectively between the matrix and fibres. Furthermore, carbon fibre provides excellent resistance to corrosion, chemical and fatigue. Such properties make it an outstanding choice as fibre reinforcement for the advanced load-bearing applications [26]. Similarly, Kevlar offers high specific tensile strength, excellent abrasion and fatigue resistance. However, it has poor compressive strength and degraded performance against ultraviolet radiation or acid corrosion.

2.2.2.2 Fabric Architecture

Fabric architecture is aimed as the structural arrangement of fibres in a composite, which not only influences the properties of the composite, but also its processing. In general, the fibre layers in various forms can be embedded into a matrix either in continuous lengths or discontinuous lengths. Continuous FRPs normally have a preferred orientation and high aspect ratios, while discontinuous fibres have a random orientation and low aspect ratios, which dramatically reduces their strength and modulus [27]. Continuous FRP composite is often made of multi layers of the fibre-reinforced plastics by stacking the unidirectional (UD) fibre or woven fibre layers and curing them into the desired thickness at certain temperature and pressure. Because of this, continuous FRP composite provides good strength in axial loading but less in transverse loading [25].

UD fibre composites are the most commonly used continuous FRPs in commercial aerospace due to their exceptional specific properties. The in-plane performance of UD composites is generally superior to any other class of FRP, e.g. woven-fabric. The reason for this is largely attributable to the way fibres are arranged within these materials. A lamina is the building block of FRP structures, which may be oriented in different directions that will enhance the strength in the primary load direction. Figure 2-3 illustrates how fibres in UD laminates might be arranged. UD prepregs have highly aligned fibres with a regular packing arrangement that allows for a high Fibre Volume Fraction (FVF). These properties give UD laminates their exceptional in-plane properties. In contrast, a woven-fabric lamina is usually offered as a cross-ply construction. Weaves are made on a loom by interlacing two orthogonal sets of yarns. This specific waviness of woven-fabric composites attributes to the most resistant to in-plane shear movement, but reduces the strength and stiffness of the composite [27]. Note that one major problem with multilayered laminates is that their interlaminar properties can be low and they can be prone to early failure by delamination, in which cracks originated at the interface between the layers due to high interlaminar tensile and shear stresses cause separation of layers [26]. To overcome this problem, 3D weaving structures have been developed to provide out-of-plane reinforcements and produce composites with enhanced through thickness properties. However, this structure may cause negative influence on the in-plane fibres and thus the degraded primary properties of composite.

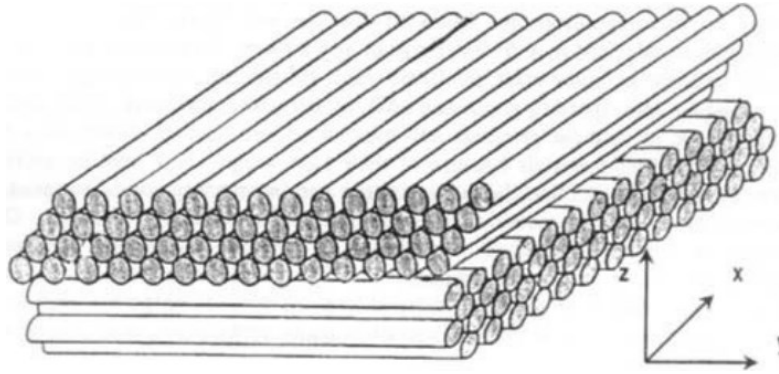


Figure 2-3. Illustration of fibre alignment in two layers of a cross-ply UD prepreg laminate [25].

2.2.3 Polymer matrix

Matrix, in general, gives shape to the composite part, protects the reinforcements from the environment, transfers loads to the reinforcements, and contributes to mechanical properties, e.g. toughness, that depend upon both the matrix and the reinforcement [3]. Polymer matrix, as the most popular matrix employed in the advanced high performance FRP composites, can be classified into two categories: thermoplastics and thermosets, according to their molecular structures which are illustrated in Figure 2-4.

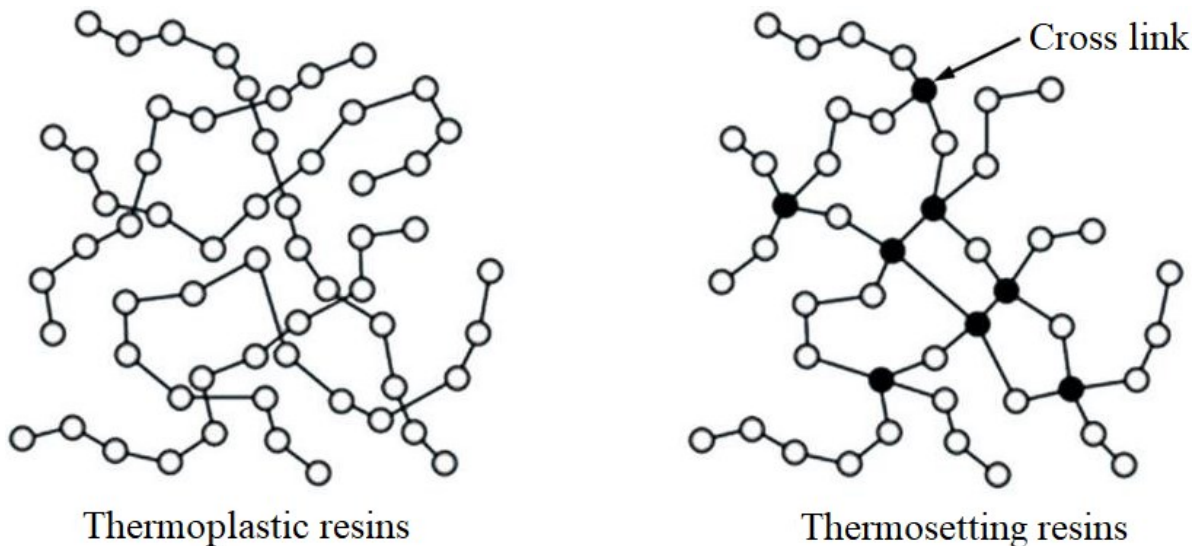


Figure 2-4. Molecular structure of thermoplastic and thermoset polymers [28].

Polymers are composed of long-chain molecules consisting of many repeating units. In thermoplastics, linear polymer arrays with amorphous, semicrystalline or mixed morphology exhibit weak intermolecular forces between polymer chains. No chemical bonding or cross links usually take place during the fabrication process. Therefore, thermoplastic resins can be repeatedly moulded when exposed to heating [2,5]. In thermosetting resins, strong covalent bonds from cross links between polymer chains are formed from an irreversible chemical reaction during the fabrication process. The cured or crosslinked material cannot be reprocessed even heated [3].

Due to the crosslinked polymer networks, thermosets exhibit excellent heat and chemical resistance, but with a resultant compromise of characteristics including formability and toughness. They are relatively brittle, resulting in low fracture toughness, reduced damage tolerance and low transverse tensile, interlaminar tensile and in-plane shear failure strains. Thermosets are also prone to microcracking during processing and when subjected to mechanical loading[26]. On the other hand, thermosets offer advantages over thermoplastics in ease of processing, addition polymerisation with no evolution of volatiles, high creep resistance, good fluids resistance and low resin costs.

In contrast, thermoplastics typically offer excellent toughness, high damage tolerance and excellent shelf life. However, their resistance to chemical, heat and creep are lower than thermosets[27]. In addition to thermoplastics, high molecular weights common in high-temperature thermoplastics result in high melt viscosity and low flow in the matrix that can make processing difficult. Poor management of volatiles generated from solvent or polymer condensation reactions can also produce voids in the composite, degrading the mechanical properties of the material. The comparison of properties between thermoplastics and thermoset are summarised in Table 2-1.

Table 2-1. The comparison of properties between thermoplastics and thermosets [20,21,29].

Thermoplastics	Thermosets
Soften on heating and pressure	Decompose on heating
Higher fabrication temperature and viscosities	Lower fabrication temperature
Can be remoulded and recycled	Cannot be remoulded or recycled
High strains to failure	Low strains to failure
Short cure cycles	Long cure cycles
Fair heat resistance	Excellent heat resistance
High resin cost	Low resin cost
Indefinite shelf life at ambient temperature	Definite shelf life at low temperature

2.3 Fabrication processes and their effects

The fabrication stages for FRPs, including lay-up, moulding and curing, are covered in this section. In lay-ups, wet lay-up, prepreg lay-up and spray lay-up are the major processes. In the stages of moulding and curing, a number of fabrication processes [6,27], such as vacuum bag forming, injection moulding, compression moulding and autoclave curing, have been established for both thermoplastic and thermoset composites. Over the past few decades, the manufacturing processes have been developed which make use of the economic processing characteristics. In particular, the effect of fabrication process of thermoplastic and thermoset composites on mechanical behaviour of the composites have been well studied and will be discussed in this section.

2.3.1 Fabrication processes

For thermosets, prepreg lay-up, vacuum bag forming and autoclave curing are the predominant fabrication method used in the aerospace industry. Typically, a prepreg tape, fibre reinforcement impregnated with a partially-cured thermosetting resin, is stored at low temperature around -18 °C as the viscosity and tack of the thermosetting resin is quite sensitive to temperature and humidity. After prepreg lay-up, the laminate is sealed in a vacuum bag for curing, as shown in Figure 2-5. Vacuum bag forming and then cured at an autoclave. The aerospace industry mostly uses carbon-epoxy prepreg which is designed to be cured at around 175 °C to 180 °C.

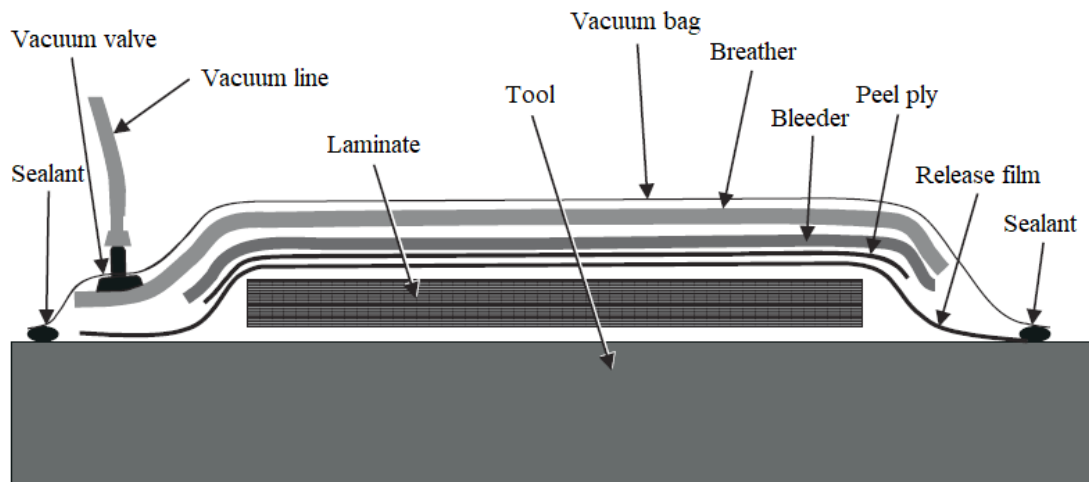


Figure 2-5. Schematic of a vacuum bag system [4].

In overall, the fabrication of thermoplastic composites is more friendly to be handled. Unlike thermosets, which undergo a chemical curing reaction for geometry stabilisation during forming, the forming process for thermoplastic composites is relatively simple and involves purely physical changes to the material. Fully consolidated prepreg tapes are mostly used and processed via compression moulding, as shown in Figure 2-6. The compression moulding is used here to denote processes where the forces required to form a part are provided by a high-pressure press, containing matched metal plates. The compression moulding process is non-isothermal and capable of high production rates with cycle times typically under ten minutes but at elevated temperature between 280 °C to 400 °C.

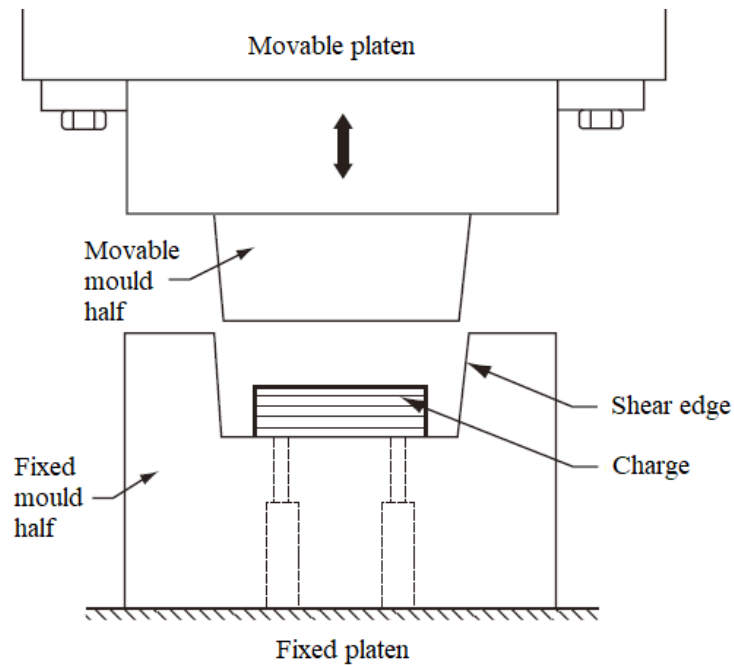


Figure 2-6. Schematic of a compression moulding system [3].

2.3.2 Effects of manufacturing process

During processing, matrix flow through the fiber architecture determines the void content, fibre wetting, fibre distribution, dry area and others in the final composite, which in turn, also affect its properties and performance. The industries therefore have increasingly focusing on optimising the composites properties and improving the economical efficiency.

Thermoplastic composites due to their inherent properties, a number of fabrication processes have provided a cost-effective way to enable a significant reduction in manufacturing cycle times by eliminating the time-consuming curing step which is necessary for thermosets. The influence of the consolidation temperature, holding time, contact pressure, cooling rate on the mechanical properties of the thermoplastic composites have been investigated and the optimum mechanical properties of the composites were studied [30–39]. For example, crystallinity, as a function of cooling rate, can lead to enhanced stiffness and promoted good mechanical properties. The crystallisation of PEEK depends on first on the state of the melt from which it is crystallised. It is preferred that this melt should be at a temperature at least 360 °C [31]. The crystallisation of the matrix slowed down after time in the melt state but the crystallisation behaviour cannot be affected by longer period heating, which concluded by Saliba et al. [32] who held times of 2 hours at 400 °C for Aromatic Polymer Composite (APC)-2. Leicy et al. [33] have studied the impact properties of standard processed and quenched

mouldings of carbon fibre/PEEK. They observe that the incompletely crystallised samples absorb 10% less energy at punch through, but that they are significantly less prone to delamination on impact. Fujihara et al. [34] have shown that a relative lower consolidation temperature (380-410 °C) with a short holding time (20 min) will achieve better mechanical performance for APC-2 prepregs. Moreover, Jar et al. [35] have revealed that the interfacial strength between matrix and fibre will be improved while the consolidation temperature increased from 380 °C to 400 °C, but this property will decrease dramatically if the temperature increased over 400 °C.

For thermoset composites, it is well known that the fabrication of thermoset composites is mainly depend on the autoclave process, which requires long cycle time and high energy consumption. Therefore, many investigators [36–39] have been focused on the effect of cure cycle time and temperature in autoclave processing. Kim and Daniel [37] studied the effect of curing cycle on the development of material properties and claimed that the residual stress developed at the excessive curing temperature due to the constraint-induced strain. Coenen et al. [38] developed a novel out-of-autoclave process for CF/epoxy composites, and Davies et al. [39] tested this process and claimed that the flexural strength of the novel processed specimen was found to be 10 % lower than the specimen manufactured by conventional autoclave process.

2.4 Impact performance

Impact performance of composites materials is deemed as a critical and fundamental field in the aerospace industry. The impact strength of a composite material is the ability of the material, under impact loading, to absorb mechanical energy without considerable deformation occurring [40]. Before considering the effects of impact loading on composite materials, it is wise to consider the different forms of impact loading and typical failure modes that can be achieved. Impact loading can be classified into low and high velocity impacts, whereby researchers [41–44] have collectively agreed that the threshold velocity between the two forms of impact loading, which also defines the threshold of low velocity impact, is 10 m.s⁻¹. Davies and Robinson [45] provide a more detailed definition of low and high velocity impacts, whereby they define a low velocity impact as one in which through-thickness stress waves in the specimen play no significant part in the stress distribution at any time during the impact event, whereas a high velocity impact is stress wave dominated. Another classification was based on the structural response of the impacted plate. Thereby, low velocity impact speeds

were impact speeds at which the contact duration between impactor and target was long enough for the structure to deform in phase with the impactor [46], as illustrated in Figure 2-7. In contrast to that were high velocity or ballistic impacts where material damage was very localised and mainly caused by through-thickness stress waves.

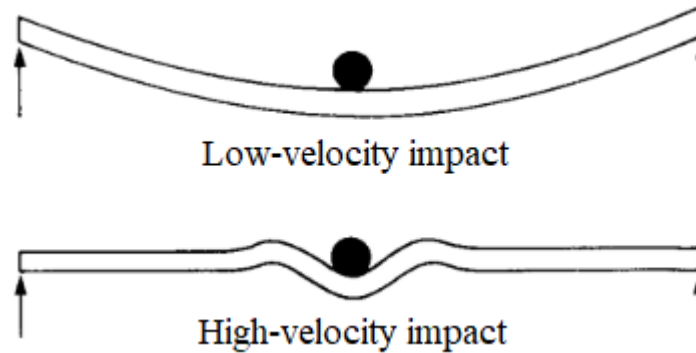


Figure 2-7. Schematic representation of impact response under low and high-velocity impact [46].

Low velocity impact is usually classified as being due to dropped tools used during maintenance services, whereas hailstorms, runway debris impacting the fuselage during take-off and landing, bird strikes and ice from propellers striking the fuselage are sources of high velocity impact [44]. Various modes of damage can occur due to the sources of impact mentioned above.

2.4.1 Under low-velocity impact loading

The low-velocity tests are generally performed using a drop weight or a swinging pendulum. A number of standard testing procedures, e.g. International Organization for Standardization (ISO), American Society for Testing and Materials (ASTM) standards, are available to assess the impact performance of the composite materials by applying the standard on specimens of fixed size and geometry, and testing conditions employed. There are many data, such as load-time curve, load-displacement curve, can be obtained from a low-velocity impact test via the data acquisition system equipped.

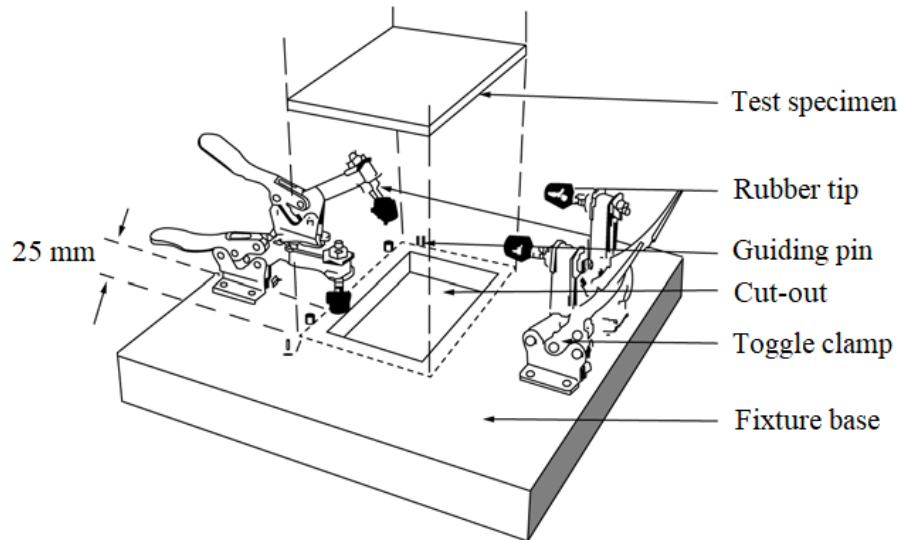


Figure 2-8. Schematic of drop-weight impact fixture in the ASTM 7136 standard [47].

2.4.1.1 Effect of matrix resin on impact performance

Considering previous research on the impact resistance of thermoplastic and thermoset composites, as early as 1985, researchers [48–50] first reported that thermoplastic composites possessed a superior resistance to impact damage, when subjected to relatively low-velocity drop-weight tests, compared with the then currently available CFRPs which employed a matrix of a relatively very brittle, thermosetting epoxy polymer. Vieille et al. [51] have undertaken an experimental study to compare the performances of a CF/Epoxy composite and carbon-fibre reinforced poly(phenylene sulphide) (PPS) and CF/PEEK composites, all with a woven-fibre architecture, when subjected to a relatively low-velocity impact using a drop-weight test. They found that the CF/Epoxy composite experienced a larger area of delamination damage compared with the thermoplastic-matrix CF/PPS and CF/PEEK composites. They suggested that this reflected the tougher nature of the thermoplastic PPS and PEEK matrices compared with the thermosetting epoxy matrix. This also further confirmed that a tougher matrix system can facilitate composite laminates to present a better impact performance.

Moreover, Chamis et al. [52] executed a range of Izod tests on composite structures. They concluded that ductile thermoplastic composites are considered tougher than thermoset composites, that are considered brittle due to their brittle thermoset matrix such as epoxy resin, due to thermoplastic composites exhibiting a greater area underneath their stress-strain curve when compared to thermoset composites. This agrees with recent observations made by Sundaram et al. [53], which states that thermoplastic composites, such as CF/PEEK, who fail

in a ductile manner and hence possess a large area under their stress-strain curve, possess greater mechanical stability during impact and greater fracture toughness than thermoset composites. This is due to the plastic nature of the thermoplastic composite's resin which causes the impact damage to propagate in a ductile manner through the composite. This statement is further supported by Jogur et al. [44] and Faur-Csukat [54], whereby they both managed to conclude that thermoplastic composites possess a greater impact resistance than brittle thermoset composites, due to their higher toughness characteristics.

2.4.1.2 Effect of laminate lay-up configuration and projectile geometry on impact performance

Caminero et al. (52), who investigated the impact performance of unidirectional and multi-directional laminate stacking sequences, found out that a sequence of laminate plies oriented at $\pm 45^\circ$ exhibited the best impact performance out of all the orientations tested, due to the fact that the impact load is shared between the matrix and fibre reinforcement. This finding was reinforced by Hu et al. (58), who stated that as the proportion of $\pm 45^\circ$ layers of fibres increase, the energy absorption of the laminate increases, while the maximum deformation exhibited by the composite structure decreases.

Mitrevski et al. [17] have reported the effect of impactor shape on the impact behaviour of woven CF/Epoxy composites. Their studies involved the low-velocity impact of the composites using 12 mm diameter hemispherical, ogival and conical head-shaped impactors at impact energies of 4 J and 6 J. They found that the energy absorbed by the specimen was the highest for the conical impactor which also produced the largest penetration depth. However, the peak load was greatest for the hemispherical impactor which also produced the shortest contact duration. In a subsequent paper [18] it was reported that different impactor head-shapes produced different levels of the various types of damage that were observed, such as fibre breakage, matrix cracking and delamination, which affected the residual properties of the composite. It was found that the relatively blunt hemispherical impactor produced the largest damage area and the type of damage was dominated by delamination. Robinson and Davies [19] have reported results from drop-weight tests using a hemispherically-shaped impactor at impact velocities of up to 6.0 m.s^{-1} striking woven glass- and carbon-fibre composites based on a matrix of an acrylic polymer. They found that the impact damage caused was mainly a function of the impact energy, and not the velocity or mass, separately, of the impactor. Further, the

measured peak load was also shown to correlate with the impact energy independent of the impactor mass used, indicating that the low-velocity impact of these specimens was a quasi-static process. Their test results also indicated that the peak load versus damage size relationship was independent of the diameter of the impactor. More recently, Sevkatt et al. [20] have undertaken drop-weight tests at an impact velocity of 4.1 m.s^{-1} on woven glass-fibre/carbon-fibre hybrid epoxy composites. They reported that the contact area between the impactor and the composite test specimen was a critical factor in the response of the composites. Impactors with relatively large contact surfaces produced a higher initial load peak, a higher maximum load, a shorter contact duration for the time-scale of the impact event and more extensive delaminations between the hybrid layers.

2.4.2 Under high-velocity impact loading

In contrast to the low-velocity test, at present, no acceptable standard testing procedures are available for high-velocity impact testing of composite materials. However, over the past few decades, numerous research has been undertaken in an attempt to better understand the impact response and failure mechanisms of CFRPs under high-velocity impact. As Cantwell and Morton [41] claimed that it is difficult to separate the effects of mechanical properties, such as strength and stiffness of the composite matrix and reinforcement respectively, and geometrical factors, such as projectile shape and laminate lay-up configuration, on the overall impact performance of the composite structure. However, it is very beneficial to present various observations made by numerous researchers on the sole effect that mechanical properties of the composite, the projectile geometry and the laminate lay-up configuration have on the impact performance of the composite material.

2.4.2.1 High-velocity impact responses

The physical phenomena occurring in laminates under high velocity impact is deemed complex. As the effect of velocity or intensity of dynamic loading is increased, the potential for damage and subsequent reduction in mechanical performance is increased. Numerous observations were made by various researchers that conducted high velocity impact simulations and experiments on composites. In terms of high strain rate effects, a number of researchers observed that the tensile and compression modulus and strength increased with the increase in

the strain rates. However, the shear modulus and strength were discovered to decrease with the increasing strain rates, and the strain rate effects in the composite appeared to be influenced primarily by the matrix viscoelasticity, fiber-matrix interfacial properties, the composite woven reinforcement architecture, and the time-dependent nature of damage accumulation. Jogur et al. [44] suggested that the dominating factor in high velocity impact is the propagation of stress waves throughout the target during the impact event. The duration of impact during a high velocity impact event between the impactor and target is very low, and hence does not allow stress waves to propagate through the thickness of the laminate and reach the rear surface. This idea of the stress waves not reaching the rear laminate surface was challenged by Will et al. [55], as they revealed that the stress waves reach the rear laminate surface and get reflected forming tensile waves, which act on neighbouring ply interfaces initiating delamination. Additionally, they concluded that a larger damage area is found on the rear surface, due to delaminated layers sliding over each other as the impact projectile reaches them, thus causing the layers of laminate to fail completely due to tensile failure. This agrees with Abrate [56] damage area depiction in Figure 2-9 which is further supported by Raimondo et al. [57]. This idea was supported by Cantwell and Morton [46] and Higuchi et al. [58], who both discovered that stress wave reflections from the rear laminate surface caused a phenomenon called spalling, which involves the removal of a large number of the outermost fibres.

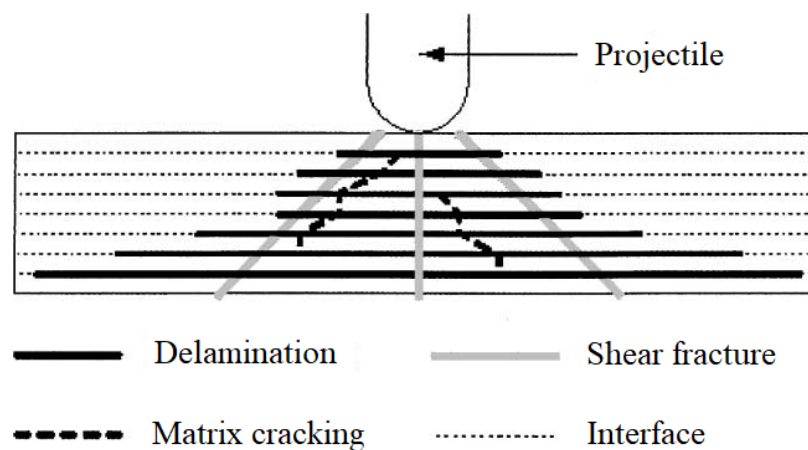


Figure 2-9. Illustration of impact damage in a laminate [55].

2.4.2.2 Effect of matrix resin on impact performance

Many reasearchers have investigated the effect of matrix system on impact performance of composites. Among the composites, thermoplastics are emerging in the aerospace industry

as promising competitors to typical thermosets thanks to their higher impact resistance and toughness. Morita [59] investigated the impact resistance of CF/PEEK and CF/epoxy composites when subjected to a high-velocity impact using a hard-projectile with a hemispherical end, and claimed that the CF/PEEK exhibited superior impact resistance than the CF/Epoxy. Wagner et al. [60] performed similar high-velocity impact tests and concluded that, at high impact energy level, CF/Epoxy laminates exhibited similar size of damage area to the CF/PEEK specimens but with severe delamination behaviour. This may lead to a reduction of the structural strength of the specimen. However, several researchers challenged the fact that the impact resistance of composites is reliant on the mechanical properties of the matrix. Husman et al. [61] discovered, after executing high velocity impact tests on carbon fibre composites, that the impact performance of the composite material does not depend on the toughness of the polymeric matrix. Moreover, Cantwell [41] investigated the independence of the impact performance on the toughness of the polymeric matrix. He suggested that, in brittle composites, the resin toughness is fully transferred to the composite, however, in ductile composites, the resin toughness is not fully transferred to the composite and therefore the positive effect of having a tough matrix is reduced. This idea has been challenged by various recent studies that suggest methods by which the mechanical properties of composite structures can be enhanced [62–65]. Bull et al. [65] suggested that the toughness of composites with thermosetting resins can be improved by adding a toughener, causing the composite to have three phases, consisting of the thermosetting matrix, the fibre reinforcement and the toughener. This was supported by Nash et al. [62], who confirmed that by modifying a thermosetting composite through the inclusion of a thermoplastic phase, the impact performance of the composite can be improved and enhanced.

2.4.2.3 Effect of laminate lay-up configuration on impact performance

A number of studies have investigated the effect of the laminate lay-up configuration on the impact performance of composite structures [55,60,66,67]. For example, Cantwell and Morton [66] studied the effect of laminate stacking sequence on the impact performance of CFRP composites. They determined that the extent of delamination observed in the composite material was related to the energy absorbed during impact and was affected by the orientation angles of laminates stacked on top of each other. This was explained and supported by Will et al. [55], who suggested that delamination is more likely to occur at interfaces where there is a

mismatch in stiffness. Since a mismatch in stiffness occurs at the interface of fibres oriented at different angles, delamination is deemed to be highly likely to occur at these interfaces. Wang et al. [67] determined that as the fibre orientation angle increases, the debris from the high velocity impact decreases in size. It was suggested that this is due to the fact that the circumferential stiffness of the composite increases with increasing fibre orientation angle, and hence adds resistance to interlaminar crack growth. Additionally, it was observed that the laminates with low fibre angles were less prone to fracture than the laminates with greater values of fibre angles.

2.4.2.4 Effect of projectile geometry on impact performance

Many researchers have investigated the effect of projectile geometry on the impact performance of composite structures [68–70]. Appleby-Thomas et al. [68] observed, by conducting a series of high velocity impact tests on composites structures, that blunt nosed projectiles caused extensive deformation and fragmentation of the target composite, due to plugging and shearing of the laminate. On the other hand, for projectiles of sharper nose geometries, they observed that the projectile had penetrated the composite to a greater degree in comparison with the blunter nosed projectiles. Therefore, it was concluded that as the sharpness of the projectile nose increases, less kinetic energy is lost by the projectile, hence allowing it to penetrate the laminate more efficiently, causing a smaller damage area in comparison with blunter nosed projectiles. Conversely, Babu et al. [69] conducted a series of high velocity impact tests using hemispherical, conical, and conical with round nose and truncated tip projectiles. It was observed that the projectile nose geometry had insignificant influence on the damaged area on the composite target, which agrees with Seifoori's et al. [70] observation. Furthermore, they were able to establish from their experimental results that the maximum amount of energy absorbed by the composite was exhibited in the target plate struck by the truncated tip projectile. This is due to the sharper conical projectiles having a point contact between them and the target, whereas for the truncated nose projectile, the contact area is greater, and hence the load will be distributed across the flat portion of the tip. Consequently, this causes higher impact energies to be needed in order to observe laminate penetration for truncated tip projectiles in comparison with normal conical projectiles.

2.4.3 Typical failure modes

The failure modes, presented by the fibre reinforced composite laminates, are generally fibre fracture, matrix cracking and delamination [43,71], Figure 2-10. In these three types of failure, the fibre fracture and matrix cracking can be summarised as intralaminar damage, and the delamination can be defined as interlaminar damage [72,73]. Furthermore, all failure modes contain two processes: initiation and propagation. Laffan et al. [74] pointed out that the critical strain energy release rates associated with these failure modes are properties that are intrinsic to the material system, which need to be measured for completely characterising the damage tolerance of the material system [75–77]. Due to the properties of matrix material, composite laminates tend to present nonlinear behaviour under shear loading, which has significant effects on the performance of composite structures, and should also be accurately characterised [78–80].

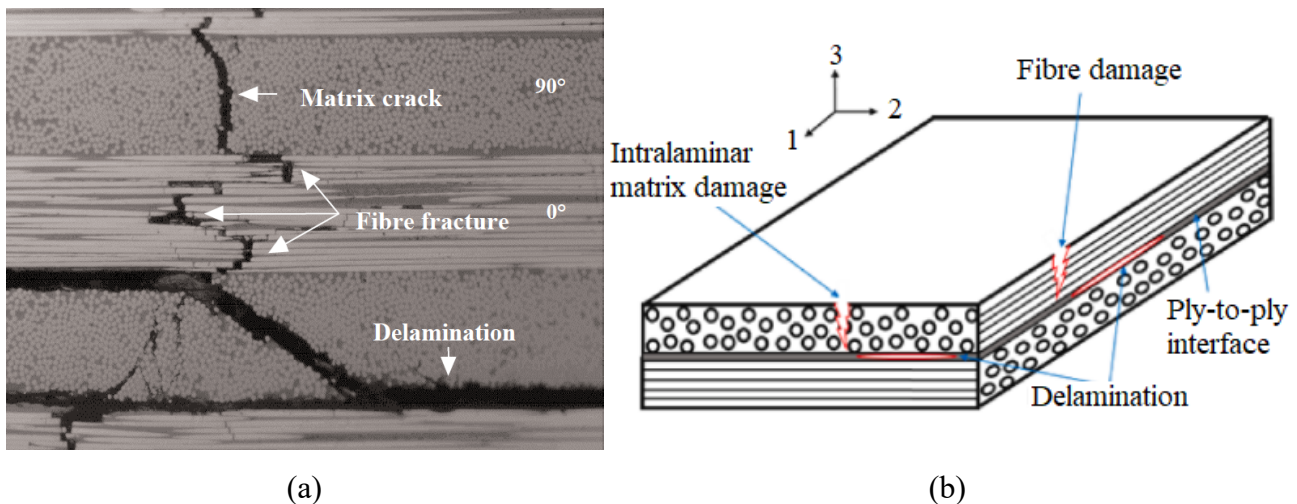


Figure 2-10. The main failure types shown in the unidirectional FRP laminates: (a) experimental observed failure modes [43] and (b) schematically categorised failure modes.

2.4.3.1 Interlaminar failure

Delamination is a type of failure between plies and may result in significant loss of compressive strength and stiffness of the laminate [81,82]. This is due to the fact that delamination has the effect of dividing the laminate into sub-laminates that are more susceptible to buckling when loaded in compression. The total strain energy release rate associated with the mode of failure, can be decomposed into G_I , G_{II} and G_{III} based on the different failure modes: Opening mode (Mode I), Shearing mode (Mode II) and Tearing mode

(Mode III), as shown in Figure 2-11, relating to interlaminar tension, sliding shear and tearing shear, respectively [81,83,84].

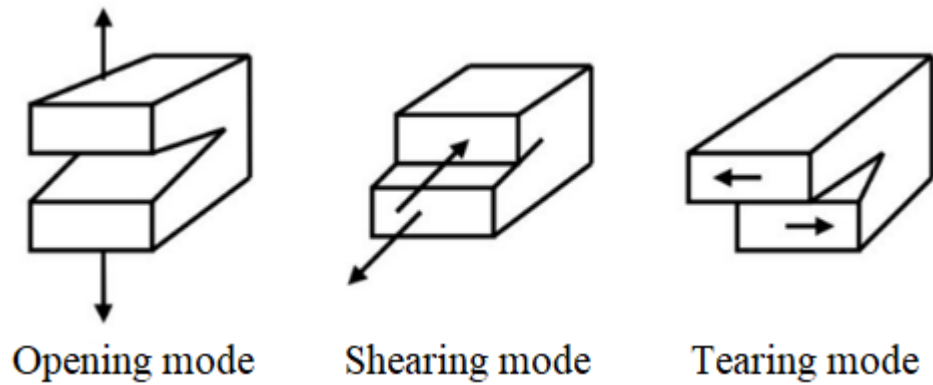


Figure 2-11. Modes of interlaminar failure [81].

2.4.3.2 Intralaminar failure

The failure modes of intralaminar failure consist of intralaminar fibre tensile/compressive failure and matrix tensile/compressive failure [85–87]. In a unidirectional composite ply, the material response in the longitudinal (0°) direction is dominated by the fibres. In woven plies, the fibres dominate the material response along warp (0°) and weft (90°) directions. For fibre-dominated failure, the large amounts of fibre-matrix de-bonding and subsequent fibre pull-out contribute to the homogenised ply-level fracture toughness [88,89]. For fibres in compression, failure may initiate from small misalignments to the loading axis which propagate as shear driven fibre failure and kinking.

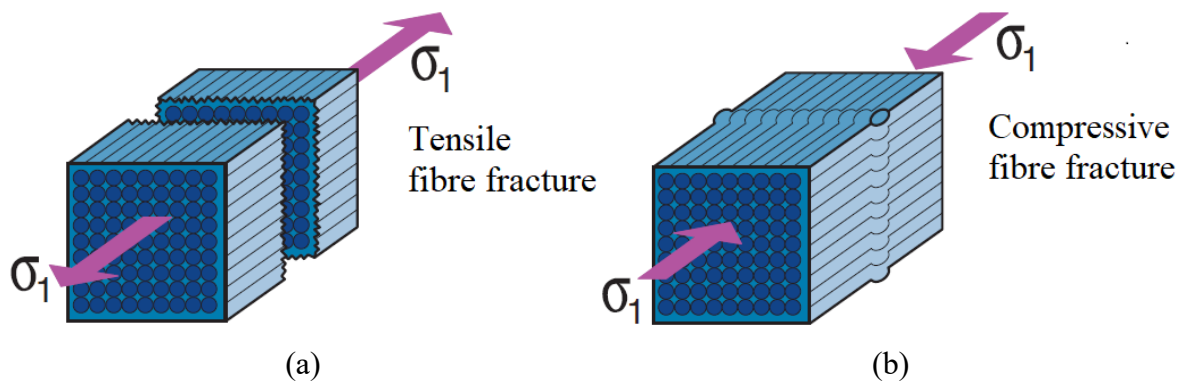


Figure 2-12. Fibre-dominated (a) tensile and (b) compressive failure modes [90].

For matrix-dominated failure, fracture plane will be formed on the specimen under tensile or compressive loading perpendicular to the fibre direction. As exemplified in Figure 2-13, the failure patterns of the woven-carbon/epoxy specimens tested under tensile and compressive loading in the through-thickness direction [91,92]. Both tensile and compressive failure showed fracture planes perpendicular to the loading direction.

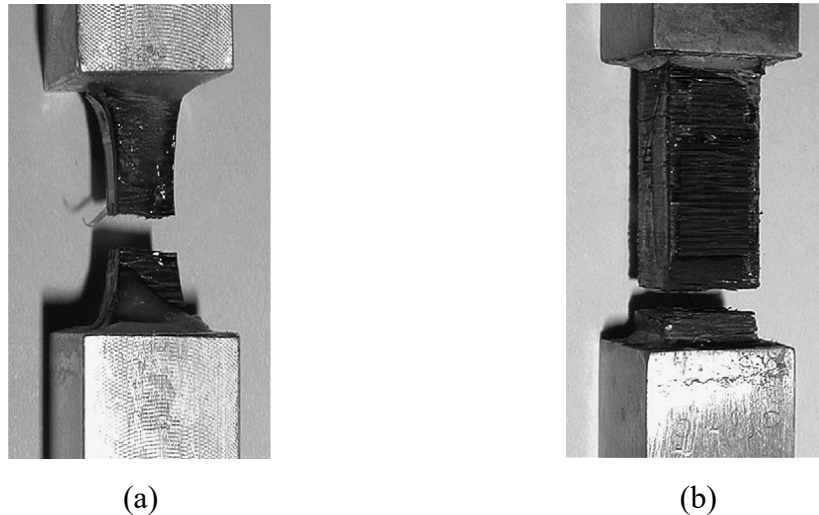


Figure 2-13. Fracture planes obtained from woven CFRPs under (a) tensile and (b) compressive loading [91].

2.4.3.3 Nonlinear shear behaviour

Under shear loading, both unidirectional and woven carbon-fibre reinforced composites behave non-linearly [78,93,94], which can be demonstrated by the stress-strain curves, for example, in Figure 2-14. This nonlinear response, which is mainly caused by the yielding of the resin, can affect the distribution of stress and the onset of failure [95–97]. The corresponding strain localisation can be readily observed along the ligament region, where failure modes including matrix cracking, fibre/matrix debonding, fibre pull-out and fibre breakage can be found [94,98,99].

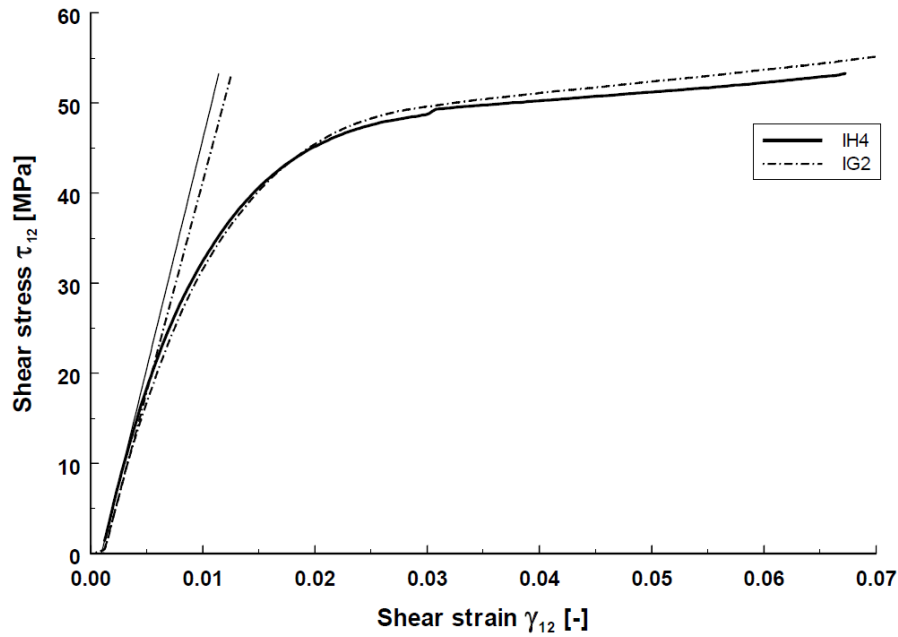


Figure 2-14. Shear stress-strain curve for the static $[+45^\circ, -45^\circ]_{2s}$ composite specimens under tensile tests [97].

2.5 Techniques for detecting impact damage in fibre reinforced polymer composites

There are various defects and damage modes that occur in composite structures which may have been introduced from the raw materials, manufacturing, assembly or during service due to impact loading. Damage to composite structures is not always visible under visual inspection. Therefore, it is critical to be able to effectively detect these flaws and modes of damage without structurally impairing the composite structures. This can be achieved using non-destructive testing (NDT) techniques, which are an essential part of the quality assurance philosophy of large commercial aerospace companies [100]. A number of NDT techniques, such as ultrasonic C-scanning and X-ray Computed Tomography (CT) scanning, whereby each possesses a set of advantages and disadvantages when compared with each other.

2.5.1 Ultrasonic C-scanning

Ultrasonic C-scanning technique has been used as one of the most widespread NDT techniques for composite damage evaluation, and as such is often used to benchmark other techniques [100,101]. Ultrasonic C-scanning involves transmitting sound waves through a

medium which is to be tested, and then analysing the reflected sound waves. Ultrasonic waves with frequencies ranging between 0.5 to 50 MHz are transmitted through the composite that is to be tested, and upon encountering a discontinuity or delamination in the composite material, the waves are reflected back and received by a transducer. Ultrasonic C-scanning results are usually viewed in the form of either an A-, B-, or C-scan, whereby the C-scan, which is most frequently used, is a planar view of the A scan's displacement (of the signal through the thickness of the material) versus amplitude information, but represented in the form of a colour map, whereby variations in the colour map indicate the presence of flaws or defects. The A-scan and C-scan representations of ultrasonic data are demonstrated in Figure 2-15.

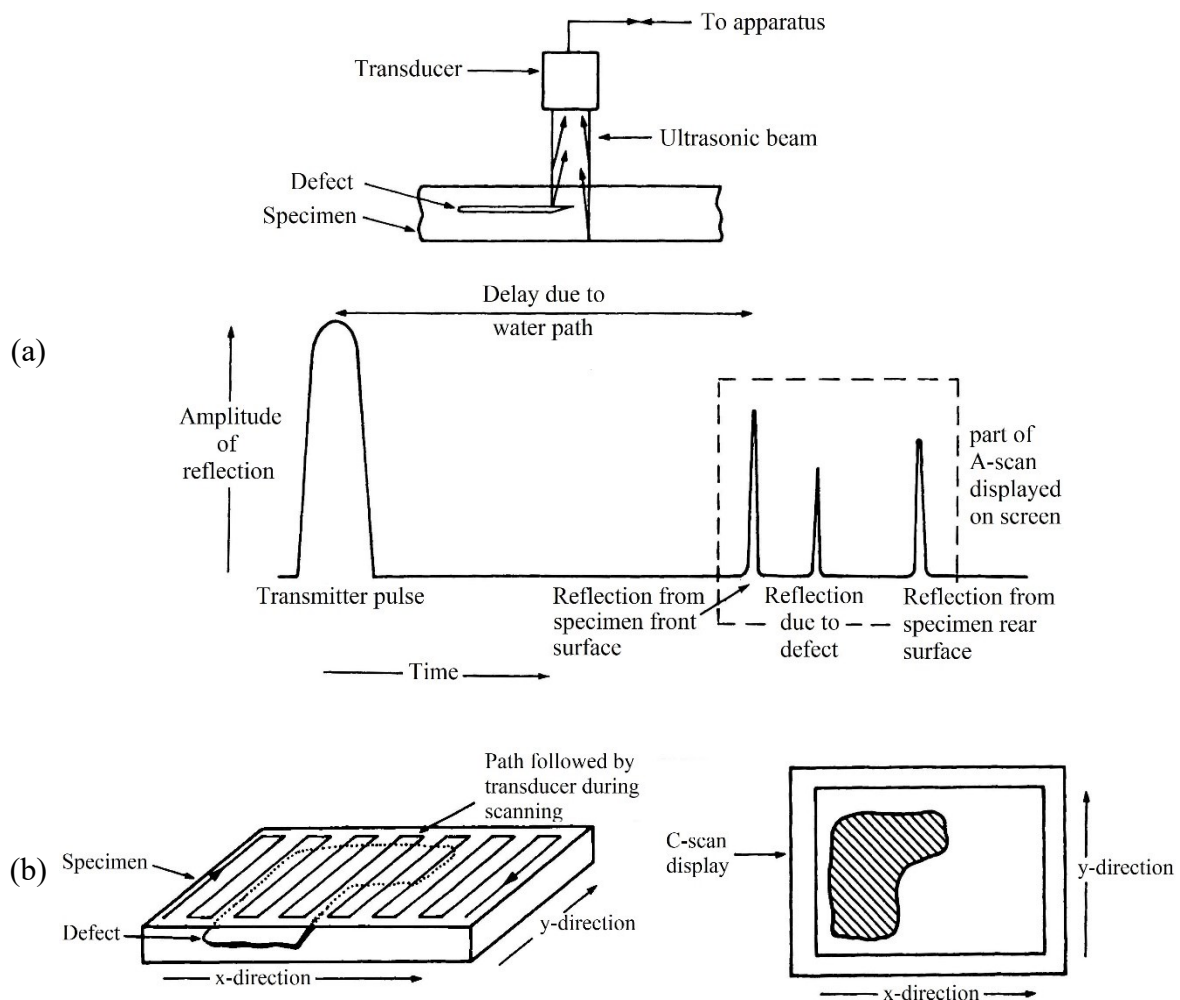


Figure 2-15. Representations of (a) A-scan data and (b) C-scan data [21].

The C-scan maps, presented in the form of a colour map, provide an integrated view and detailed depth distribution of the damage area. The position and size of a delamination can

be determined by comparing the total travel time (delamination position) and amplitude (delamination size) of the wave reflected from both the delamination and the back wall of the material [100]. Because of this principle, the C-scan maps can only present the superposed delamination areas in various plies. In addition to this, due to the propagating mechanisms of impact stresses, the interlaminar fracture extends considerably from the point of impact until the bottom ply, which causes a conical fashion from the impact point, i.e. the smaller delamination area is in the upper (near-impact-surface) ply while the largest delamination area is in the lowest ply [3,102]. Therefore, it is considerably important to scan from the impact surface which would allow to determine the shape and depth of the delamination in various plies. Moreover, the presence of delamination can be detected by observing a sudden reduction in amplitude of the back-wall echo, while for smaller voids and inclusions, reflected signals which are weaker than that of a crack or delamination are detected. In addition to this, ultrasonic C-scanning has the disadvantages of relatively slow inspection speed and the requirement of a coupling liquid.

2.5.2 X-ray Computed Tomography (CT) scanning

X-ray CT scanning relies on the computational reconstruction of radiographs acquired using different angles of illumination on the test object, while the image contrast depends on the differences in the attenuation of X-ray radiation passing through the investigated specimen [103]. This popular method allows effective assessment and visualisation, in 3-D, of failure modes such as fibre breakage, delamination and matrix cracks, thus preventing various misinterpretations caused by previously generated 2-D images. In addition, X-ray tomography allows the user to export composite laminate structures, in order to generate image-based meshes, which can then be used to produce finite element models of the composite structure [103]. However, X-ray tomography possesses numerous disadvantages, such as the fact that it is highly expensive and time consuming [104]. In addition, the size of the test chamber greatly limits the size of the sample to a few centimetres, causing the technique to be unsuitable for aerospace applications and components. Also, X-ray tomography relies on producing a test sample from the manufactured component and cannot be performed in situ [104], and therefore is impractical for aerospace components, as specimens cannot be obtained from an in-service aircraft.

2.5.3 Other techniques

Generally, in all NDT methods, the composite is evaluated by the detection of a test signal, which is used to assess the structural health of the composite structure. However, none of which are universally applicable. The choice depends on cost, ease of testing, types and size of defect. Therefore different techniques are chosen to detect different types of flaws. A number of other NDT techniques, such as eddy current, acoustic emission and shearography, have different advantages on various types of damage detection.

The eddy current technique, which is classed as a radio frequency NDT technique, induces current in a conductive material, according to Faraday's law [105]. For the case of composite materials such as CFRP, the induced currents flow, in the composite structure, along fibres and from one fibre to another through the fibre contact points. The impedance of the current flow through the fibres will change at a discontinuity or flaw in the composite structure, and hence flaws in the composite structure can be detected using an impedance analyser [105,106]. This technique is designed to detect delamination, fibre breakage, resin rich zones and even observe the fibre orientations within the composite structure, but not to detect matrix cracking [106].

The acoustic emission technique, whereby it relies on detecting, using an array of piezoelectric transducers, the mechanical vibrations that stem from flaw and defects in the composite material, such as matrix cracking, fibre-matrix debonding, delamination and fibre pull-out. This technique relies on detecting the energy released from the medium rather than supplying it with energy, which is the case with most other NDT techniques [107]. It is not only able to detect the various modes impact damage but also fatigue damage. However, it requires highly trained and skilled individuals to be able to use the information of mechanical vibrations released from flaws, and correctly associate it with a specific defect or damage mode in the composite material [107].

Shearography, is considered as one of the most advanced NDT techniques, relies on the fact that light waves, of different path lengths, produce a fringe pattern when interference occurs [101]. A shearing interferometer is used to laterally shear the reflected beam, causing interference between the sheared and direct images [100]. If a delamination is present in the composite material, under some form of loading, the location where the delamination is present will deform differently than other locations in the material. Therefore, when compared with the unloaded specimen reference image obtained, allowing detection of the presence of a

delamination in the composite structure. However, it possesses the disadvantage of not being able to detect and characterise other forms of impact damage other than delamination [107].

2.6 Composite damage modelling

Over the past few decades, a number of techniques have been proposed to develop physically-based composite material damage models predicting the ultimate strength of composite structures. Most are broadly based on distinguishing between interlaminar and intralaminar damage, with the latter further accounting for fibre-dominated and matrix-dominated failure modes.

2.6.1 Interlaminar modelling

Delamination, which may result from impact damage or manufacturing defects, can lead to a significant reduction in the load-carrying capacity of composite structure [108–110]. An understanding of the mechanism of delamination and the capability to build an effective failure model for this kind of failure deserve particular attention. Most evaluations of delamination growth apply a fracture mechanics approach and evaluate energy release rates for self-similar delamination growth [111–114]. At present, there are primarily two methods, which are widely used, to predict interlaminar damage—the Virtual Crack Closure (VCC) technique, or the use of cohesive zone model.

2.6.1.1 Virtual crack closure technique

According to the formulations and assumptions of Elastic-Plastic Fracture Mechanics (EPFM), it is possible to predict crack growth with the VCC technique, proposed by Rybicki and Kanninen [115]. Despite the valuable information, regarding the onset and the stability of delamination, which can be acquired using the VCC technique, there are limitations. The VCC technique may require complex moving mesh techniques to advance the crack front when the local energy release rates reach a critical value [116]. VCC technique analysis also requires that an initial delamination be defined, while for particular geometries and load situations, it is not easy to confirm the locations of the delamination front [117].

2.6.1.2 Cohesive elements model

The application of cohesive elements placed at the interfaces between plies can overcome some of the shortcomings mentioned above. Cohesive elements are based on the Dugdale-Barenblatt cohesive zone approach [118], which can be related to Griffith's theory of fracture when the size of the cohesive zone is negligible compared with the characteristic dimensions of the structure being analysed [119]. A main advantage of using cohesive elements is the capability to predict both initiation and propagation of delamination without previous knowledge of the crack location and propagation direction, provided that cohesive elements are defined at all ply interfaces.

As for the mixed mode cases, the quadratic delamination criterion is used to predict the initiation of the softening process. The most extensively used criterion to predict delamination propagation under mixed-mode loading is the power law criterion. Similar expression can be derived from a more accurate criterion presented by Benzeggagh and Kenane (B-K Law) [120]. The development of a cohesive element allows a faithful description of delamination under mixed mode and the variation of mode participation during the delamination process [121].

2.6.2 Intralaminar modelling

A number of researchers have proposed numerical models for capturing intralaminar failure. The majority of current intralaminar damage models are based on the more empirical approaches. For different types of damage in composite structures, intralaminar fibre failure and matrix failure, these intralaminar models can predict the response reasonably well. More importantly, these models provided a basis for developing more complicated cases including crush and impact damage [122–124].

2.6.2.1 Modelling fibre-dominated damage

On the basis of Mohr's and Hashin's considerations on brittle fracture, Puck and Shurmann [125] developed a new inter-fibre fracture criterion for different fracture modes in fibre-reinforced plastics. Results relating to crack initiation and fracture as well as an indication of the direction of cracks and the fracture angle can be gained from this phenomenological model. By drawing on Dvorak's [126] fracture mechanics analyses of cracked plies and Puck's action plane concept, Davila and Camanho [86] proposed a new physically-based criteria which

can predict fibre failure without requiring parameters of curve-fitting. Specifically, an in-situ effect was considered in their model. They made an assumption that only longitudinal shear strength is subjected to in-situ effect. The result shows a significant improvement over the Hashin criteria.

2.6.2.2 Modelling matrix-dominated damage

Criteria for determining a matrix fracture plane, in a unidirectional ply under various damaging loads, were proposed by Puck and Schurmann [125]. The criteria suggested that for transverse tension, this fracture plane will be normal to the load, while for transverse compression or shear, the fracture plane would be determined by the ability of the matrix to resist the shear loading. The stress state, at a fracture plane, is characterised by a normal stress, σ_n and shear stresses σ_{nt} and σ_{n1} , as shown in Figure 2-16. A stress-based damage criterion can be employed to capture the matrix damage initiation in composite laminates. It is generally assumed that the fracture toughness associated with intralaminar matrix propagation is similar to interlaminar Mode II fracture toughness [127].

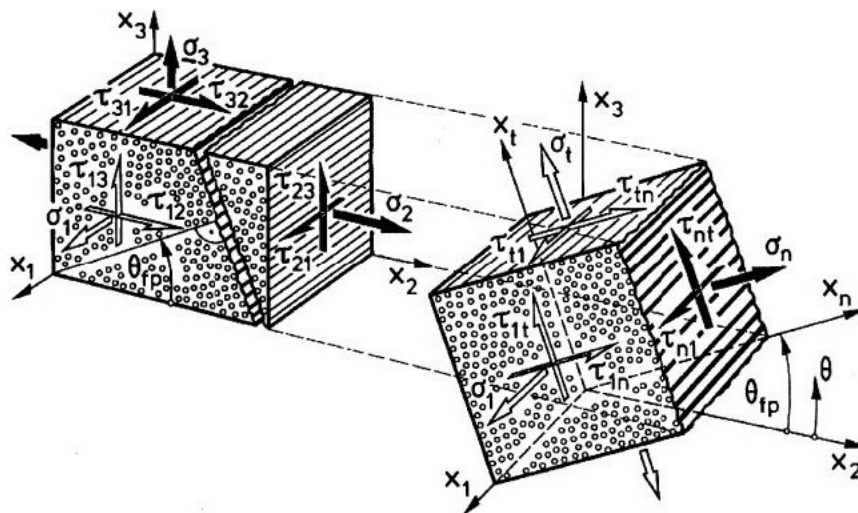


Figure 2-16. Stresses and fracture plane on a composite material [125].

2.6.3 Modelling impact behaviour

A selection of FE based composite damage models is available in commercial packages. Examples include the Abaqus built-in progressive composite damage model based on the work

by Matzenmiller et al. [128]; and the LS-DYNA [129] material model type 262 which uses an approach based on the failure criteria presented by Chang and Chang [130]. Despite the widespread application of these commercial packages, calibration of non-physical parameters to control the damage propagation, is generally required.

Y. Shi et al. [109,131] used stress-based criteria and fracture mechanics techniques to capture composite laminate damage initiation and evolution of damage during an impact event. The nonlinear shear properties of composites were defined by a shear stress–strain semi-empirical formula. X-ray radiography was used to validate the proposed numerical model. Ansari and Chakrabarti [132] conducted a numerical investigation on the penetration and perforation behaviour of FRP composite laminates under impact loading. The effects of boundary conditions and thickness-to-span ratio were discussed. Donadon et al.[133,134], Faggiani et al. [135] and Falzon et al. [136,137] proposed a 3D Computational Damage Mechanics (CDM) based material damage model to capture the intralaminar degradation of composite laminates with nonlinear shear behaviour. This model was combined with cohesive elements to investigate impact damage and captured the permanent indentation caused by low velocity impact.

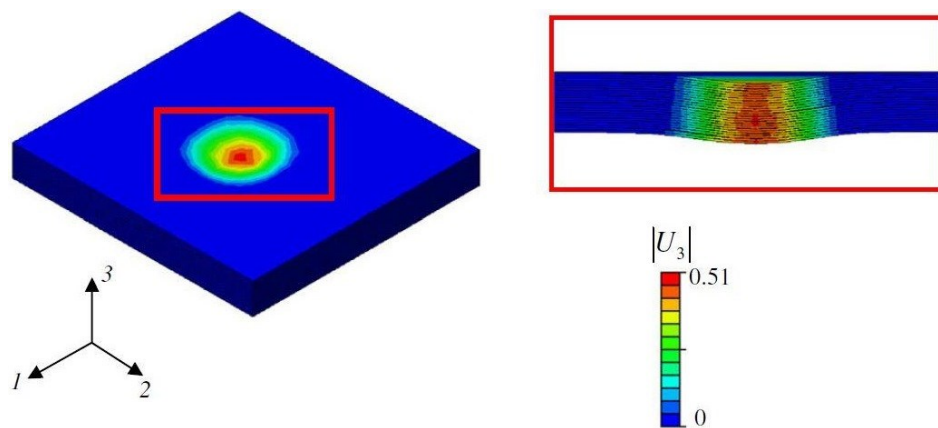


Figure 2-17. Out-of-plane displacement magnitude for stiffened composite panel FE model after impact showing permanent indentation as a result of the damage process (in mm) [135].

Along with the development of the experimental methodology, various numerical approaches have been employed to model the composites under high-velocity impact events [138,139]. For example, Nishikawa et al. [138] presented a numerical simulation based on the Lagrange multiplier method to predict the impact-induced deformation and damage of

composites subjected to high-velocity soft-body impacts. In this method, severe contact-induced deformation caused by a relatively large contact force can be effectively determined by adjusting the time-scale by using a small time step. Liu et al. [139] used an explicit FE code to predict damage in a composite leading edge. The Smoothed Particle Hydrodynamic (SPH) method was used to model the deformation and flow of the soft-body gelatine projectile, which was employed to simulate a bird-strike. In the SPH method, the FE mesh for the impacting body is replaced by interacting particles, that are able to better represent the deformation and flow of the projectile over the target. Cantwell et al. [46,48,66,140] have performed a number of studies on the high-velocity impact responses of CF/Epoxy composites, in which specimens with various lay-up configurations and thicknesses were evaluated. The energy-absorbing capability and failure mechanisms of these composites were determined and a model was developed to predict the degree of perforation of the impacted target specimen. This model enabled the evaluation of the effects of target thickness, and other factors, on the threshold impact velocity, and the associated impact energy, for damage initiation and for perforation of the composite target.

2.7 Summary

This chapter introduced the fundamentals of the composite materials, and gave a review of relevant past and current research work on composite materials subjected to impact loading. A brief overview of historical background and classification of composite materials, including fibre reinforcement, polymer matrix, were given. The fabrication processes for thermoplastic and thermoset composites were introduced, and the effect of fabrication parameters on mechanical properties of composites were elaborated.

This is followed by a review of the impact performance of composite materials under low-velocity and high-velocity impact. The impact response and failure mechanism, as well as the effect of mechanical properties, e.g. strength and stiffness of composite matrix and reinforcement, and geometrical factors, e.g. projectile shape and laminate lay-up configuration, on the overall impact performance of the composite structures, were discussed. Subsequently, a number of Non-Destructive Testing (NDT) techniques for damage assessment of composites were introduced and compared their applications and limits. There has been relatively little research on comparing the effects of changing the impact velocity and energy levels over a relatively wider range when using a rigid or soft impactor, such as might arise from debris,

birds or tool impacts, to strike a thermoplastic composite and a thermoset composite with the same fibre architecture.

Finally, this chapter also introduced the fundamental Finite Element (FE) modelling techniques, including the interlaminar modelling, intralaminar modelling, non-linear shear modelling and impact behaviour modelling. A number of numerical research work based on various modelling methods were reviewed, which characterised and predicted the response of impact and energy-absorption of the composite materials. Furthermore, a number of factors that affect the impact response of composite structures, including the loading rate, projectile geometry and material constituents of the composite structures were reviewed, which contributed to increase the design efficiency and reduce the maintenance costs of composite structures.

Chapter 3. Materials and Experimental procedures

3.1 Introduction

The carbon fibre reinforced thermoplastic and thermoset composite specimens, used in this study, were fabricated by employing either a unidirectional-fabric prepreg or a woven-fabric prepreg. Various properties of these specimens, including fibre type, fibre volume fraction, thickness, toughening agent, dimension, moulding method etc., were introduced in this chapter. Various types of experiments and technologies, e.g. low-velocity (drop weight) testing and high-velocity (gas gun) testing were performed to understand the failure mechanisms and determine the performance of these specimens under impact loading.

3.2 Materials

3.2.1 Unidirectional fibre reinforced composites

The unidirectional AS4 carbon fibre reinforced poly (ether-ether-ketone) (CF/PEEK) prepregs (APC-2) and the T700 carbon fibre reinforced epoxy (CF/epoxy) prepregs (BA9913) were employed to fabricate the thermoplastic and thermoset specimens, respectively. The photographs of these unidirectional CF/PEEK and CF/epoxy specimens are shown in Figure 3-1. Note that specimen categories can be determined visually from external features, as the pale crystalline polymer by precipitation can be seen on the surface of the CF/PEEK specimens while the CF/epoxy specimens present good gloss of carbon fibres.

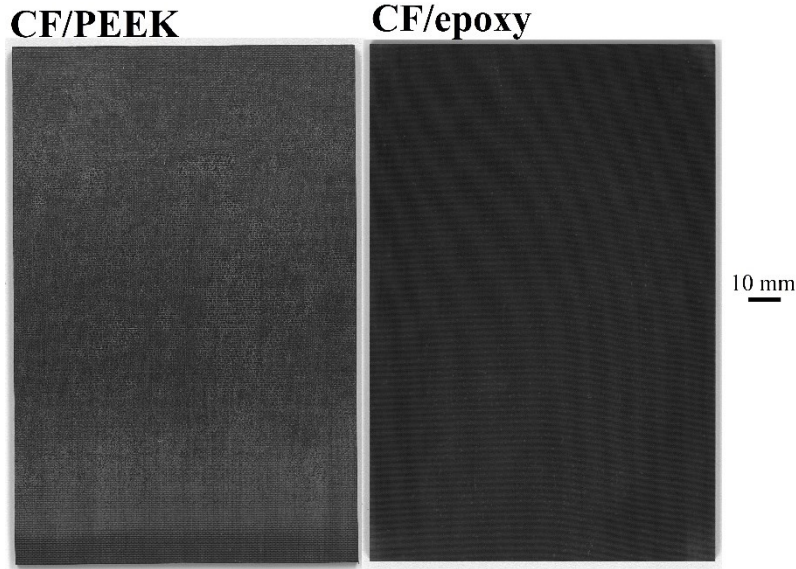


Figure 3-1. Photographs of the unidirectional CF/PEEK and CF/epoxy composite specimens.

In the CF/Epoxy specimens, 20-30 wt% thermoplastic polyether-sulfone toughening agent was used to toughen the matrix system. Note that CF/PEEK employed AS4 carbon fibres and the T700 carbon fibres, employed in the thermoplastic and thermoset laminates respectively, have very similar mechanical properties [141,142], which confirms the facility of the comparative study. Both the CF/PEEK and CF/epoxy specimens have the same layup of $[90_3/0_3]_{2s}$ and the same overall thickness of 3 mm. The overall density of the CF/PEEK and CF/Epoxy composite specimens was 1.60 g.cm^{-3} and 1.62 g.cm^{-3} , respectively. It should be noted that the objectives here were (a) to achieve virtually the same density for the two types of composites (i.e. the CF/PEEK and the CF/Epoxy composites) and (b) for the specimens made from the two types of composites to have very similar values of thickness. The primary properties of both specimens are summarised in Table 3-1.

Table 3-1. Primary properties of the unidirectional CF/PEEK and CF/Epoxy laminates.

	CF/PEEK	CF/Epoxy
Prepreg	APC-2	BA9913
Matrix system	Poly(ether-ether ketone)	Epoxy (toughened)
Fibre type	Unidirectional AS4	Unidirectional T700
Fibre volume fraction	60 %	68 %
Overall density	1.60 g cm ⁻³	1.62 g cm ⁻³
Moulding	Press moulding	Autoclave consolidation
Lay up	[0 ₃ /90 ₃] _{2s}	[0 ₃ /90 ₃] _{2s}
Thickness	3 mm	3 mm
Dimensions	150 × 100 mm	150 × 100 mm

3.2.2 Woven-fabric fibre reinforced composites

The woven-fabric CF/PEEK and CF/Epoxy composite specimens were made from the same T300 carbon fibres but two different matrix systems (Vitrex PEEK 150PF and non-toughened C-M-P epoxy CP009-3, respectively). The photographs of these woven-fabric CF/PEEK and CF/Epoxy specimens are shown in Figure 3-2. The specimens were machined, using a waterjet cutter, from large plaques into 140 x 140 mm test specimens.

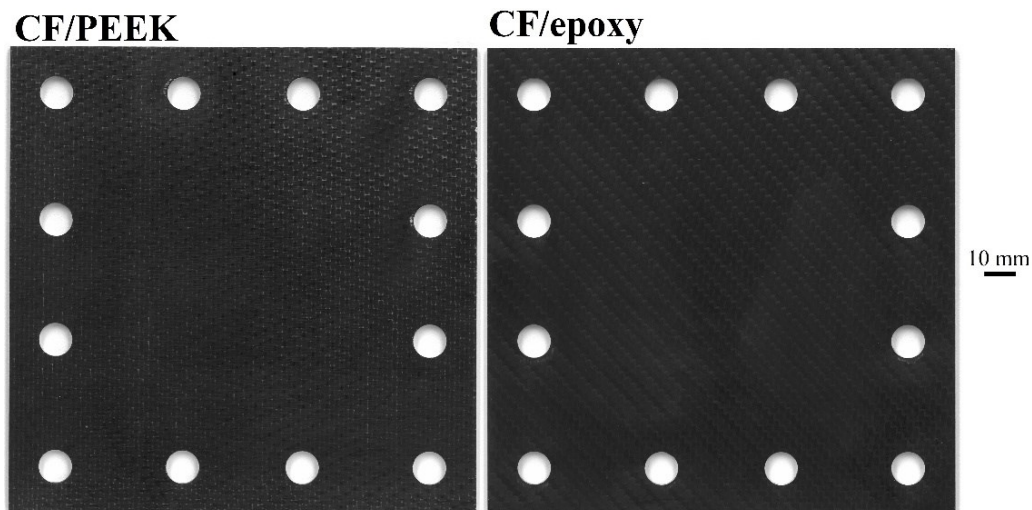


Figure 3-2. Photographs of the woven-fabric CF/PEEK and CF/Epoxy composite specimens.

Both types of specimen had a nominal thickness of 2 mm. To achieve this thickness, seven plies of the thermoplastic prepreg, with a nominal 50% volume fraction of fibres, were used for the CF/PEEK specimens and eight plies of the thermoset prepreg, with a nominal 55% fibre volume fraction, were used for the CF/Epoxy specimens. The overall density of the CF/PEEK and CF/Epoxy composite specimens was 1.54 g.cm^{-3} and 1.57 g.cm^{-3} , respectively. The primary properties of both specimens are summarised in Table 3-2.

Table 3-2. Primary properties of the woven-fabric CF/PEEK and CF/Epoxy laminates.

	CF/PEEK	CF/Epoxy
Matrix system	Poly(ether-ether ketone)	Epoxy (non-toughened)
Fibre type	Woven-fabric T300	Woven-fabric T300
Fibre volume fraction	50 %	55 %
Overall density	1.54 g cm^{-3}	1.57 g cm^{-3}
Moulding	Press moulding	Autoclave consolidation
Thickness	2 mm	2 mm
Dimensions	$140 \times 140 \text{ mm}$	$140 \times 140 \text{ mm}$

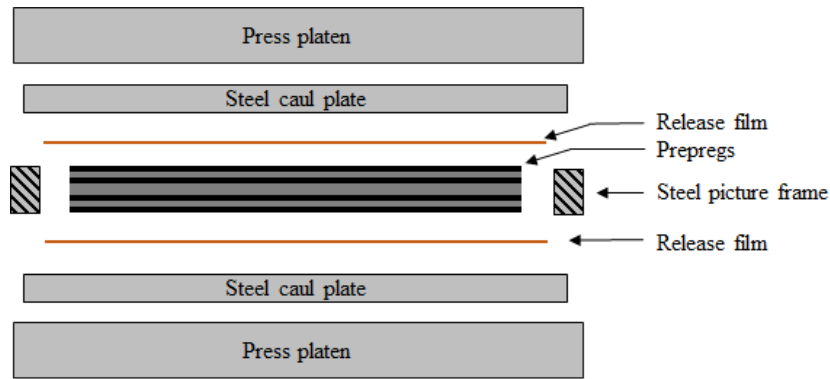
3.3 Manufacturing processes

3.3.1 Thermoplastic polymer composites

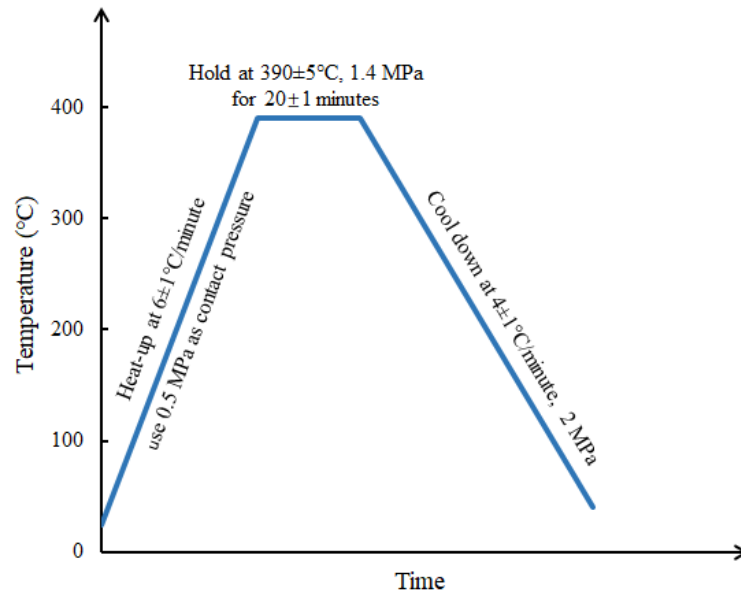
The unidirectional CF/PEEK laminates were fabricated at the University of Nottingham using a hot press device provided by Mackey Bowley, United Kingdom. Whilst the woven-fabric CF/PEEK laminates were manufactured by Haufler Ltd., Germany and used same press moulding process. The hot press device at the University of Nottingham, shown in Figure 3-3 (a), produces a consolidation temperature of up to $410 \text{ }^{\circ}\text{C}$ and a contact load of up to 50 tonnes. This device also employs a water-cooling system which aids cooling the press platens at a maximum rate of $4\pm 1 \text{ }^{\circ}\text{C/minute}$.



(a)



(b)



(c)

Figure 3-3. Photograph of (a) the thermo compressor and the diagram of (b) the consolidation jig and (c) the cure cycle for the CF/PEEK laminates consolidation.

Before the hot press moulding, unidirectional prepregs were cut into square shape by utility knife and stacked as desired configuration. The individual plies were pre-tacked by using soldering iron to weld around the periphery of the stack. The tacked prepregs were placed into the consolidation jig as presented in Figure 3-3 (b). A steel ‘picture frame’ was used to control the thickness of the consolidated specimens and stop the flow of material. The release agent and release films were used for easy demoulding. The cure cycle of the CF/PEEK laminates consolidation is plotted in Figure 3-3 (c). Note that there is no restrictions on the rate of heat up and cool down.

3.3.2 Thermoset polymer composites

The unidirectional CF/Epoxy laminates were manufactured at the AVIC Composite Corporation using an autoclave, shown in Figure 3-4(a), provided by Beian, China. Whilst the woven-fabric CF/Epoxy laminates were manufactured by Haufler Ltd., Germany and followed the similar autoclave consolidation process. Typical consolidation jig used for this process which illustrated in Figure 3-4(b). Note that every three layers of prepreg were vacuumed in separate bag for 4 hours and overlaid in desired lay-up and kept for over 12 hours before the consolidation process. The cure cycle of the CF/Epoxy laminates consolidation is plotted in Figure 3-4 (c).

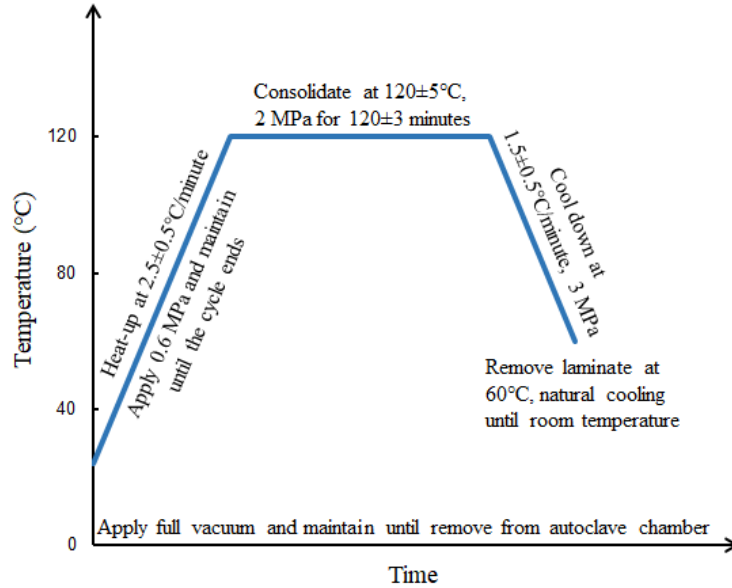
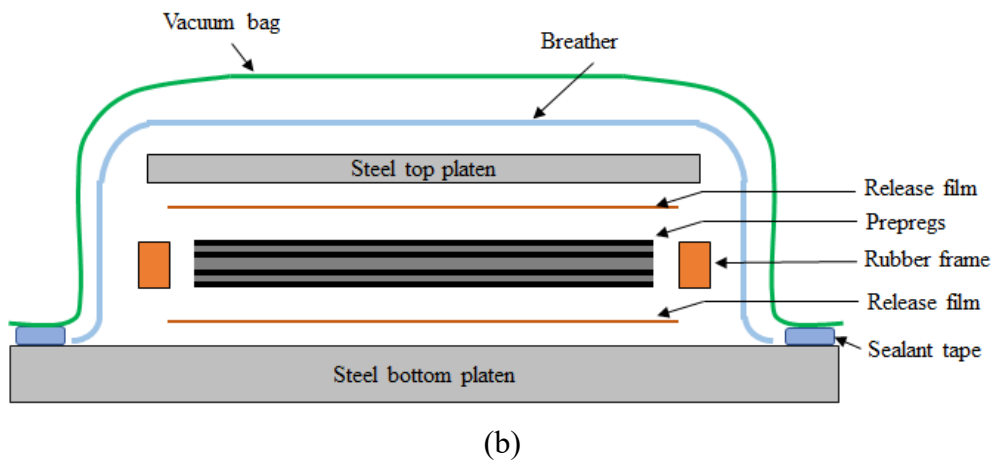
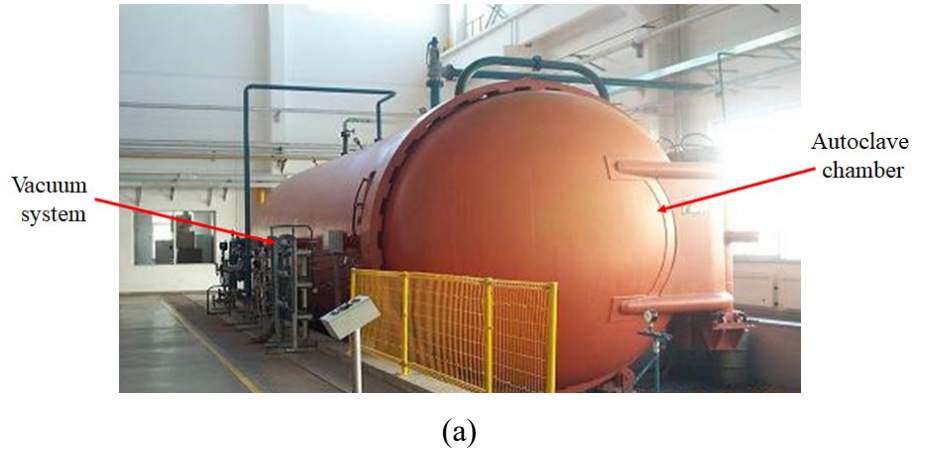


Figure 3-4. Photograph of (a) the autoclave and the diagram of (b) the consolidation jig and (c) the cure cycle for the CF/Epoxy laminates consolidation.

3.4 Overview of experiments

To determine and compare the performance of the thermoplastic and thermoset laminates under different rates of impact loading, low- and high-velocity impact testing were performed. 3D Digital Image Correlation (DIC) measurement was employed with gas gun testing to monitor the out-of-plane displacement of the specimens during impact. Note that all experiments were performed in a laboratory equipped with a constant environmental system holding a temperature of 22 ± 0.5 °C and a humidity of 50 ± 2 %.

3.4.1 Low-velocity (drop weight) testing

The low-velocity impact experiments were performed by using Instron CEAST 9340 drop weight tower, as shown in Figure 3-5, which is designed to deliver up to 405 J of impact energy. Each composite specimen was placed on a steel picture-frame which had outer dimensions that matched those of the 150 mm × 100 mm composite specimens and with a 125 mm × 75 mm window cut-out. This assembly was clamped to the base of the drop-weight tower using four toggle clamps with rubber tips, which prevented slippage of the composite specimen during the impact test. The load cell, with a data sampling rate of 500 kHz, was located in the forward section of the impactor. An anti-rebound system as required in the ASTM standard was employed to catch the impactor in order to prevent it from impacting the specimen a second time.

This floor standing drop weight tower also equipped with the CEAST DAS 64K data acquisition system which designed to run the tower and collect experimental data. This data acquisition system enables the data collection from an instrumented tup insert with built-in high-frequency piezoelectric sensor. This sensor is particularly ideal to test fairly brittle materials such as composite materials.

Before testing, a configuration including the overall mass of drop weight, desired impact energy, testing mode, specimen information and data acquisition should be established in the CeastVIEW software. Based on the known information, this software calculates the desired velocity upon impact and then moves the impactor to certain height off the clamped specimen in return for the gravity potential energy required. For testing mode, all drop weight experiments in this study were carried out in Compression After Impact (CAI) testing mode which is common for composite materials and a majority of standards. After

impact testing, the load/time traces for each test were recorded while the load/displacement and energy/time traces could be calculated from the data recorded.

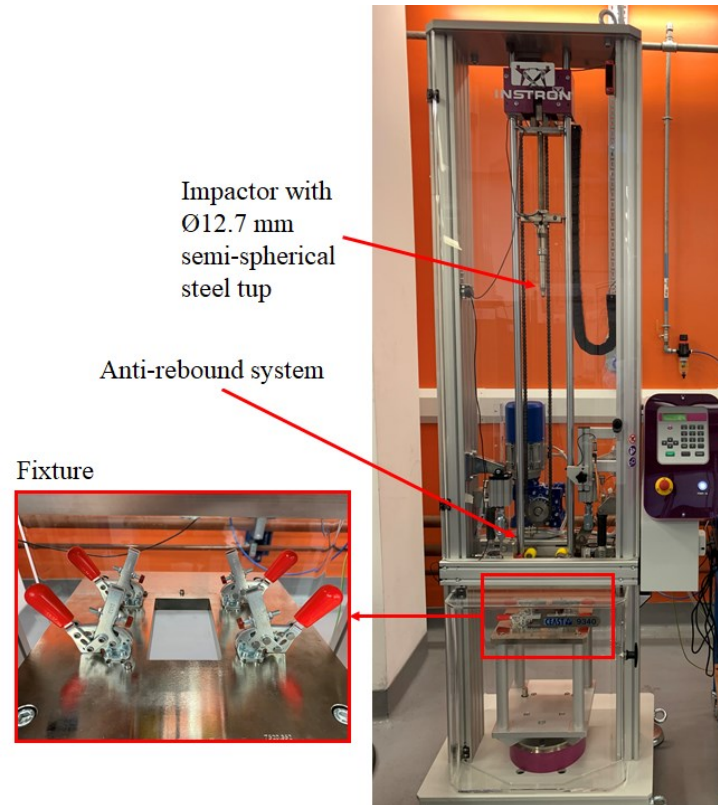


Figure 3-5. Set-up for the drop weight (low-velocity) experiments.

3.4.2 High-velocity (gas gun) testing

The air-feed gas gun was employed to preform high-velocity impact experiments. The facilities for gas gun experiments are shown in Figure 3-6. Helium gas was used to feed a four-litre pressure vessel which provides air pressure up to 10 bar and was released into to a three-metre long barrel by a fast-acting pneumatic valve. In order to fire the projectiles or sabots in different diametres, this barrel can change its inner diameter to 10 mm, 25 mm or 40 mm by switching the specific size of the inner barrels.

The velocity of the projectile was controlled by changing the pressure of the vessel and showed high consistency at the same specified pressure. The oscilloscope recorded the time difference of the projectile passing two pairs of infrared radiation sensors located at the end of the barrel. Therefore, the velocity of the projectile can be determined from the time difference

and the known distance between two pairs of the sensors. Note that the distance from the end of the barrel to the specimen was 10 cm and any difference in the velocity of the projectile measured at the exit point from the barrel compared with that upon it impacting the composite specimen was less than $\pm 1\%$. A target chamber consists of steel plates and perspex panels to provide safety guarantee and the clear view between the specimen fixture and the camera stand for 3D DIC measurements.

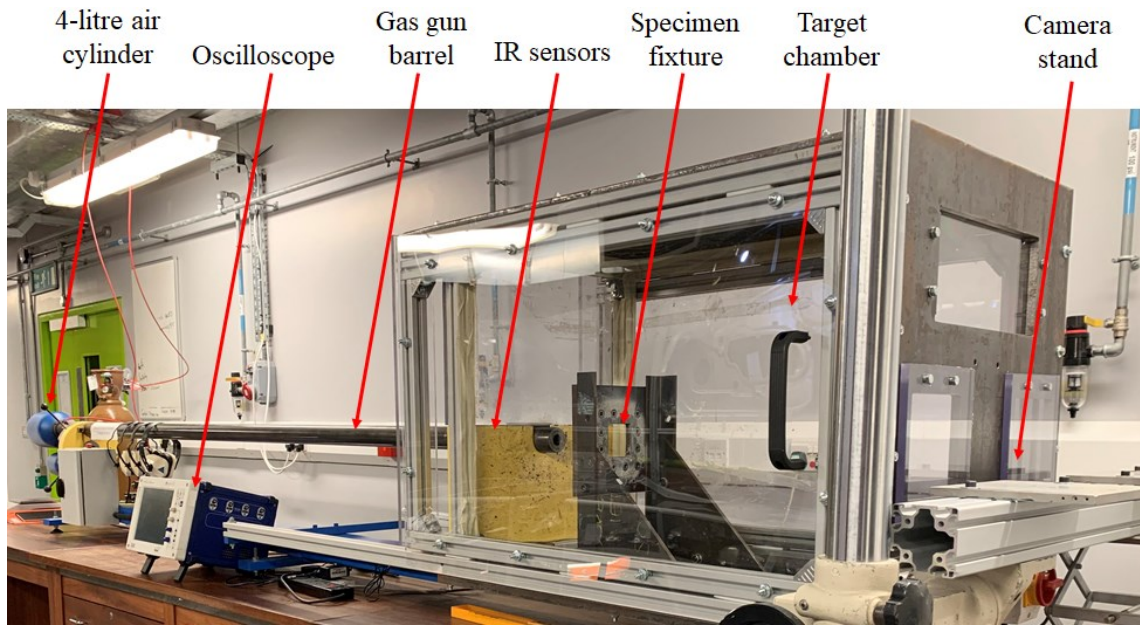


Figure 3-6. Set-up for the gas gun (high-velocity) experiments.

3.4.3 3D Digital Image Correlation (DIC)

A 3D DIC system was employed to monitor the out-of-plane displacements of the specimens during gas gun impact, which was measured from the deformations of the non-impacted (i.e. rear) surface of the composite specimen via the random black speckle-pattern. As the schematic shown in Figure 3-7, two synchronized high-speed cameras were placed behind the target chamber at a distance of 925 mm from the centre point of the target. The distance and angle between two cameras were 410 mm and 25° , respectively, which is as recommended by GOM Ltd., Germany [143]. A pair of 50-mm fixed focal length lenses, supplied by Nikon, UK, was employed for both cameras. These two cameras were triggered simultaneously by the signal generated from the infrared sensors and recorded the deformation of the rear of the composite specimen. Two light-emitting diode lamps were employed to

illuminate the rear surface of the CFRP specimens, which were only turned on a few seconds before each experiment was initiated.

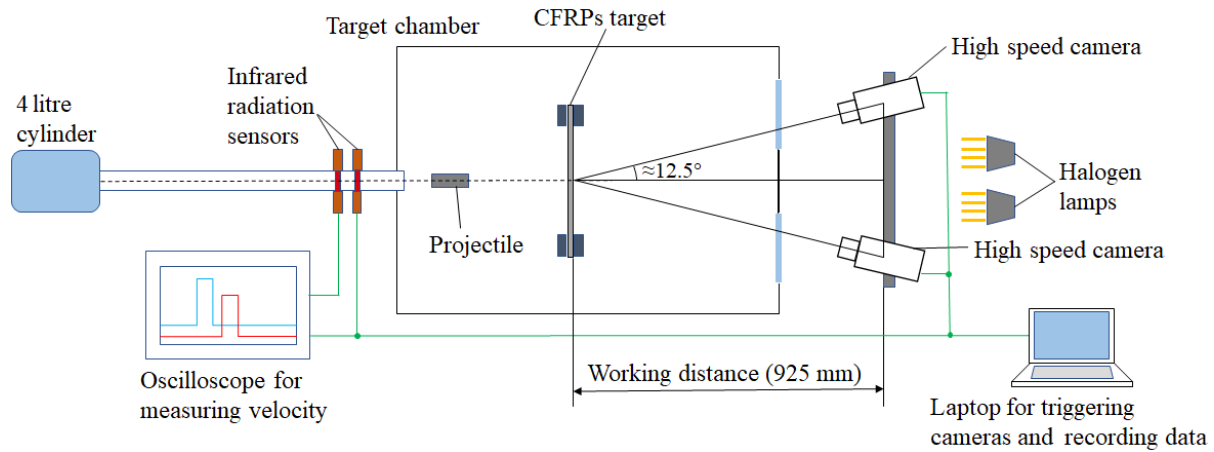


Figure 3-7. Schematic for the high-velocity impact test with 3D DIC system.

Before DIC measuring process, a calibration object should be measured so that the DIC measuring system is adjusted and its dimensional consistency is ensured. During this process, the system determines geometrical parameters of each camera, such as position and orientation, and calculates the reference point deviation, i.e. intersection error. According to the measuring volume and distance in the existing set-up, a 120×96 mm calibration panel, with 4 mm diameter reference points, was used as calibration object to be recorded in 13 defined positions, in order to determine the geometrical parameters of the cameras and the reference point deviation. A qualified calibration file would be created via the 'GOM ARAMIS Professional' software if the orientation between two cameras is within $25 \pm 0.3^\circ$ and the reference point deviation is under 0.05. This software was employed also for the data processing of the image series recorded by the high-speed cameras. Note that a new qualified calibration file generated for each testing day and only be used with the DIC data recorded on same day. For processing the measured DIC data, the surface components employed a facet size of 19 pixels, with 15 pixels between each displacement point, following the recommendations for the optimum speckle size [143].

It should be noted that the 3D DIC technique only records surface displacements and strains but this is useful in recording the overall displacement response of the panel before damage and can detect surface damage when it occurs.

3.4.4 Ultrasonic C-scan

Ultrasonic C-scan technique was employed to examine all CFRP specimens before experiments, in order to avoid that any potential damage occurred during the process of manufacturing, transportation and storage. Further damage assessments, for the specimens after impact, were conducted. A portable 'Prisma 16:64 TOFD' ultrasonic C-scan device with a 5 MHz probe, supplied by Sonatest Ltd., UK, as shown in Figure 3-8, was employed to assess potential damaged area of the specimens. This phased array probe integrated with a 16 /mm encoder which is used to provide encoded position of the probe along the scan axis.

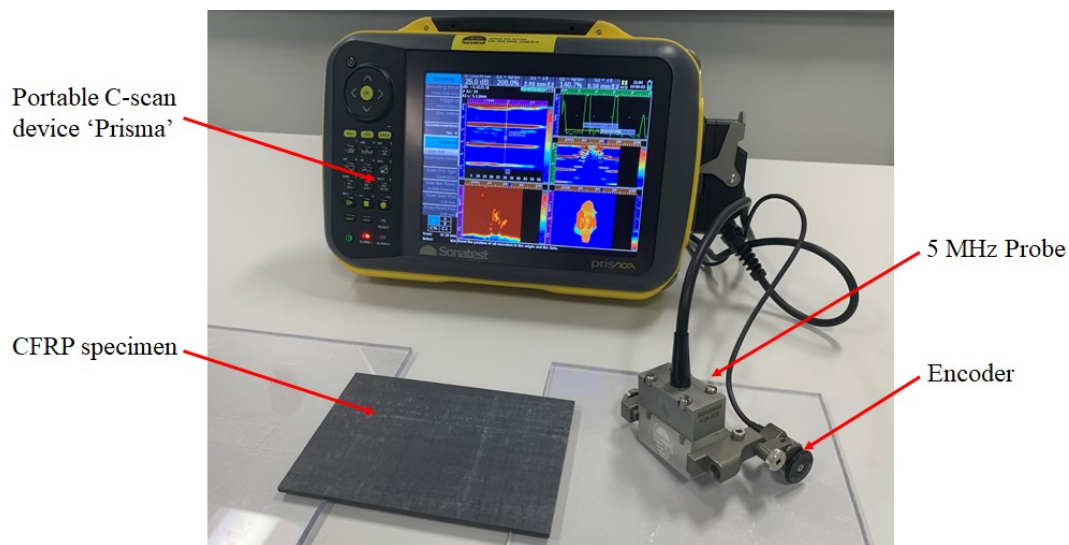


Figure 3-8. Portable C-scan system.

To precisely assess potential damage, a configuration is set for each type of specimen with different matrix systems and thicknesses. In the configuration, A-scan, refers to a one dimensional amplitude modulation scan, used to determine the output volume of ultrasonic signal and the gate measurements for pulses. A gate measurement with adjustable length is set to rectify the extent of signal extraction covering the positive waves reflected from the bottom surface of the specimen or the potential damage. Therefore, a pre-damaged specimen was used to understand the A-scan profile of a damaged area. During scanning, C-scan maps, presenting the damage area as a function of the depth through the thickness of the specimen, show the damage outline supporting the resolution of up to 0.1 mm². Last but not least, water spray used to apply a layer of water, as contact agent, on the top surface of the specimen to eliminate the effect of air and ensure the stability of ultrasonic transmission.

Chapter 4. Experimental study on unidirectional thermoplastic and thermoset laminates under low-velocity and high-velocity impact loading

4.1 Introduction

The present chapter describes the results from experimental studies on the behaviour of continuous carbon fibre polymer matrix composites subjected to a relatively low-velocity or high-velocity impact, using a rigid, metallic impactor. Drop-weight and gas-gun tests are employed to undertake the low-velocity and high-velocity impact experiments, respectively. The carbon-fibre composites are based upon a thermoplastic poly(ether-ether ketone) matrix (termed CF/PEEK) or a thermoset toughened-epoxy matrix (termed CF/Epoxy), which have the same fibre architecture of a cross-ply $[0_3/90_3]_{2s}$ lay-up (detailed in Chapter 3.2.1). The studies clearly reveal that the CF/PEEK composites exhibit the better impact performance and that, at the same impact energy of 10.5 ± 0.3 J, the relatively high velocity test of 54.4 ± 1.0 m.s⁻¹ leads to more damage in the composites than observed from the low-velocity test where the impactor struck the composite at 2.56 m.s⁻¹.

Note: The work presented in this chapter was previously published in the Journal of Materials Science as “The behaviour of thermoplastic and thermoset carbon-fibre composites subjected to low-velocity and high-velocity impacts”.

4.2 Test specification and methods

4.2.1 Drop weight testing

The drop-weight impact experiments were performed using an Instron CEAST 9340 drop tower, as detailed in Chapter 3.4.1, with an instrumented stainless-steel impactor having a hemispherical head with a diameter of 12.7 mm. The drop-weight impact experiments were conducted at three energy levels, i.e. 4.5 J, 7.5 J and 10.5 J, and three replicate CF/PEEK and CF/Epoxy specimens were tested at each energy level. The impact energy was varied by adjusting the height of the impactor. For the impact energies of 4.5, 7.5 and 10.5 J, with the 3.2 kg impactor, then the corresponding impact velocities were 1.68, 2.16 and 2.56 m.s⁻¹, respectively. No software filtering of the output force-signal was employed and the software for the instrumented drop-weight test provided the impact load and displacement as a function of the time-scale of the impact event.

4.2.2 Gas gun testing

The schematic of the set-up for gas-gun experiments is given in Figure 3-7. In the gas-gun experiments, the specimens were fixed by four toggle clamps on a steel supporting plate with a 125 × 75 mm opening. This fixture provides the same boundary condition as used in drop weight testing. Two high-speed cameras ('Photron AX200' cameras supplied by Photron, Japan) recorded the testing process at their maximum rate of 50,000 frames per second at a resolution of 384 x 256 pixels. For the 3D DIC measurements, the target composite specimens were painted with a matt-white acrylic paint, which was used to avoid excessive light reflection from the specimens. A random speckle pattern was hand-painted on the white surface using a matt-black marker to achieve the maximum contrast. The size of the black speckles was approximately 1.5 mm.

The Ø12.7 mm projectiles, used in the gas-gun experiments, were machined as cylinder objects with hemispherical noses, which have the same geometry as the ones used in the drop-weight impact experiments. These projectiles were manufactured using the 7075-T6 aluminium alloy and had a nominal mass of 7.1 ± 0.1 g. The high-density polyethylene (HDPE) sabot, with an inner diameter of 12.7 mm and an outer diameter of 24.9 mm, were used for accelerating the projectiles to a desirable velocity. An aluminium alloy stopper, located at the

end of the barrel, was employed to separate the projectile from the sabot and secure the sabot during the gas-gun experiments.

4.2.3 Post-test inspection

In order to assess the damage suffered by the composite specimens, a visual and an ultrasonic C-scan inspection were undertaken. The latter used the ‘Prisma’ portable C-scan equipment (Sonatest Ltd, UK) and the instrumented probe had a scanning frequency of 5 MHz.

4.3 Experimental results

4.3.1 Low-velocity drop weight tests

4.3.1.1 Load versus time and load versus displacement results

Figure 4-1 shows the results for the measured load as a function of time for the CF/PEEK and CF/Epoxy composite specimens impacted at energy levels of 4.5, 7.5 and 10.5 J using the low-velocity drop-weight test. The corresponding impact velocities were 1.68, 2.16 and 2.56 m.s⁻¹, respectively. Clearly very good reproducibility is recorded for the three replicate composite specimens. Now, several interesting observations may be made from the results shown in Figure 4-1.

Firstly, considering the CF/PEEK composites, Figure 4-1(a) at an impact energy (IE)=4.5 J shows the load versus time curves for the three replicate tests and all these curves reveal that no detectable damage occurred in the CF/PEEK composites, which is confirmed by the C-scan results as discussed below. The relatively small amplitude, sinusoidal oscillations on the rising part of these force versus time curves at an IE=4.5 J are indicative of mass-spring oscillations, as first analysed in detail in [144–146]. However, the load versus time curves shown in Figure 4-1(a) for a higher impact energy of IE=7.5 J clearly reveal that the three replicate curves now all exhibit a very major drop in load at $3.4 \pm 4\%$ kN, which is indicative of the initiation of damage in the composite specimens as is confirmed by the C-scan results below. Likewise, for Fig. 4(a) at an IE=10.5 J, all three replicate curves show a major decrease in load at $3.1 \pm 10\%$ kN, which again is indicative of damage initiation, again as is confirmed by the C-scan results shown below.

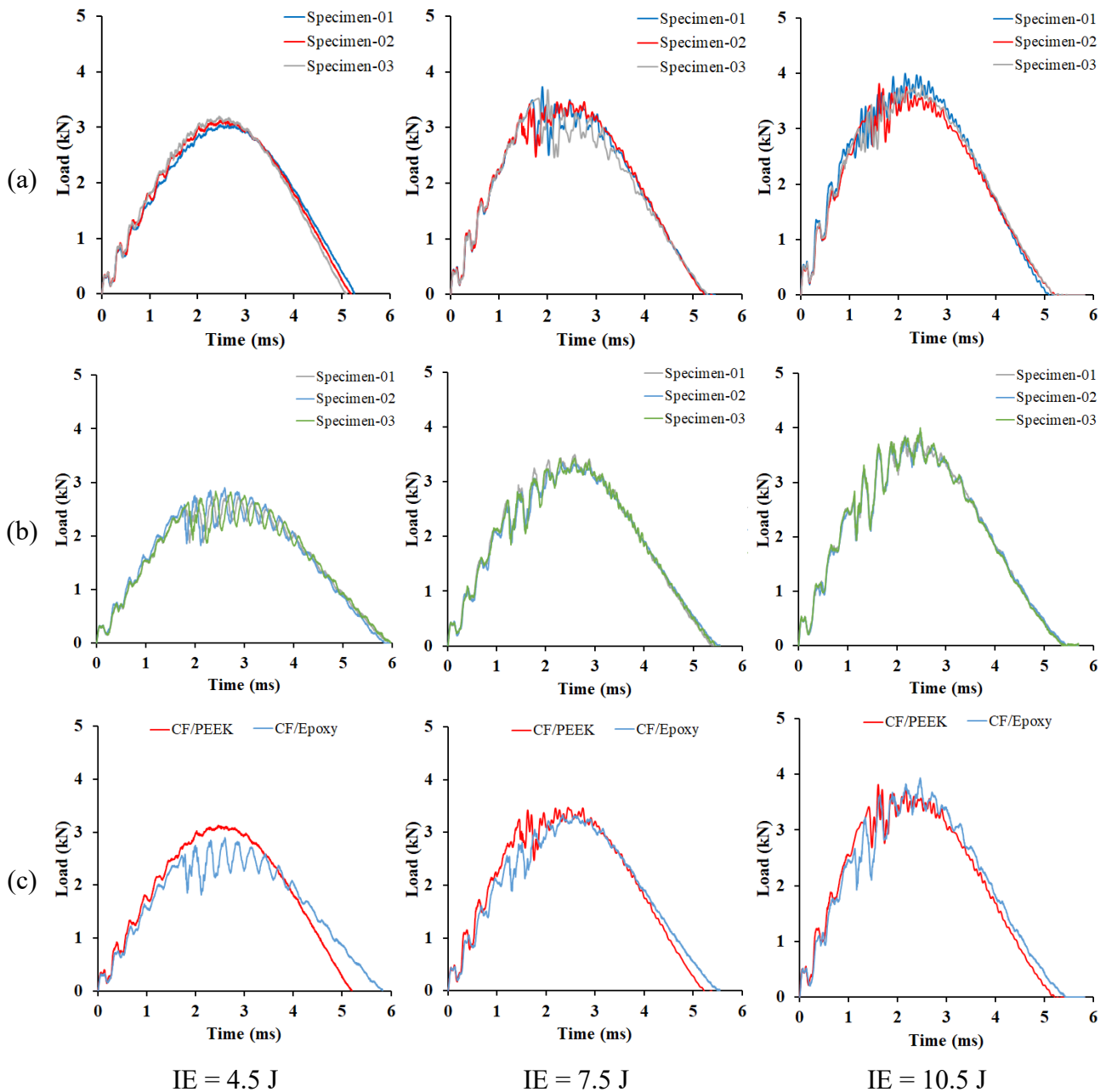


Figure 4-1. Measured load as a function of time curves for the low-velocity drop-weight tests at impact energies (IE) of 4.5, 7.5 and 10.5 J: (a) CF/PEEK, (b) CF/Epoxy and (c) a comparison of typical load versus time curves for the CF/PEEK and CF/Epoxy composites.

Secondly, Figure 4-1(b) shows that a similar behaviour exists for the CF/Epoxy specimens but for this composite damage occurs even at the lowest impact energy of 4.5 J. Indeed, the loads at which a major decrease occurs are $2.6 \pm 3\%$, $2.6 \pm 2\%$ and $2.8 \pm 3\%$ kN for the values of impact energy of 4.5, 7.5 and 10.5 J, respectively. These interpretations of the load versus time curves are again confirmed by the C-scan results, as discussed below. Also, as for the CF/PEEK results, at all three impact energy levels there are relatively minor sinusoidal oscillations associated with mass-spring oscillations prior to a major decrease in the

load associated with damage initiation. Thirdly, Figure 4-1(c) compares a typical load versus time curve for the CF/PEEK composite with one for the CF/Epoxy composite for each impact energy level and in Table 4-1 values are given for the average loads for damage initiation and the maximum recorded loads. It is apparent from these results that at the lowest energy level, i.e. IE=4.5 J, the CF/PEEK composite exhibited a significantly higher maximum load than recorded for the CF/Epoxy composite. Further, at this impact energy, whilst damage occurred in the CF/Epoxy composite, no damage was initiated in the CF/PEEK composite specimens and their stiffness at relatively high loads was therefore not compromised. In contrast, at the highest impact energy of 10.5 J, the results for the CF/PEEK and CF/Epoxy composites reveal very similar values of the maximum load for both types of composite, although the load for damage initiation is somewhat lower for the CF/Epoxy composite. Indeed, the results show that for the impact energies of both 7.5 J and 10.5 J the load for damage initiation is consistently somewhat higher for the CF/PEEK than that for the CF/Epoxy composite specimens, but once damage has been initiated and has propagated the general shape of the load versus time curves shown in Figure 4-1(c) for the two types of composite are similar.

Table 4-1. Comparison of the measured damage initiation load (i.e. the first major drop in the measured load versus time curve) and the maximum load obtained from the low-velocity drop-weight tests. (Note: there was no evidence of a damage initiation load for the CF/PEEK composite for an impact energy of 4.5 J. The error given is the coefficient of variation from the replicate experiments).

Type of composite	Damage initiation load (N)			Maximum load (N)		
	4.5 J	7.5 J	10.5 J	4.5 J	7.5 J	10.5 J
CF/PEEK	-	3.4±4%	3.1±10%	3.1±7%	3.6±4%	3.9± %
CF/Epoxy	2.6±3%	2.6±2%	2.8±3%	2.8±6%	3.4±2%	3.9 ±2%

Turning to the load as a function of displacement curves shown in Figure 4-2 for the composite specimens impacted at energy levels of 4.5 J, 7.5 J and 10.5 J, then as may be seen the maximum out-of-plane displacement increases with increasing impact energy, as would be expected. No significant difference was found between the values of the maximum displacements undergone by the CF/PEEK and the CF/Epoxy composite specimens, apart from possibly at the lowest impact energy of 4.5 J where the CF/Epoxy test specimen underwent a somewhat higher displacement. This is due to the CF/PEEK composite specimen suffering no damage at all at this lowest impact energy level, unlike the CF/Epoxy composite. Finally, the

intersection point between the load versus displacement curve and the x-axis is the out-of-plane displacement corresponding to separation between the impactor and the composite specimen, since for both types of composite the test specimens were never penetrated by the impactor and a significant rebound of the drop-weight impactor occurred at the end of the impact event.

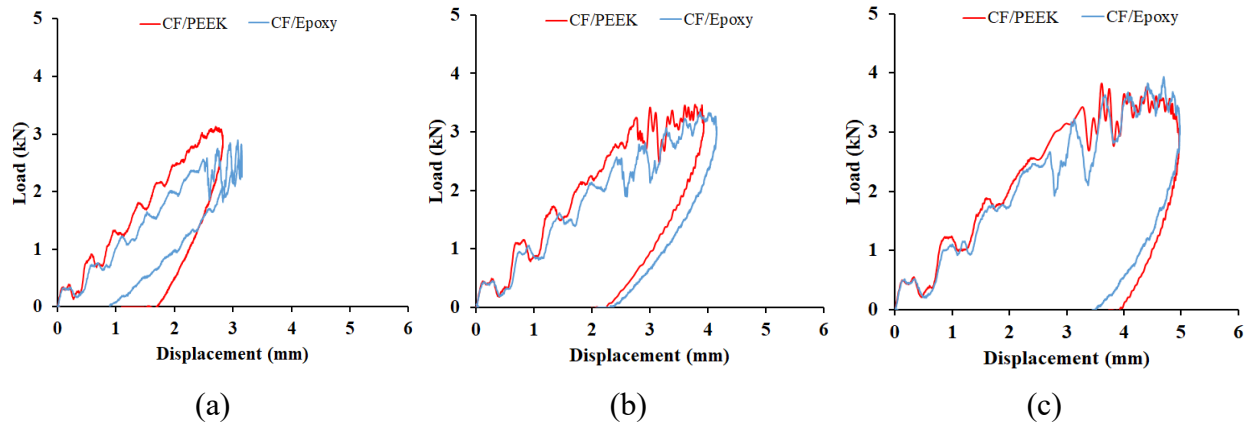


Figure 4-2. Measured load as a function of displacement curves for the low-velocity drop-weight tests for the CF/PEEK and CF/Epoxy specimens at impact energies of (a) 4.5 J, (b) 7.5 J and (c) 10.5 J.

4.3.1.2 Visual and C-scan inspection

For all the composite test specimens, no damage could be detected from a simple visual inspection of the specimens after the impact test. However, a slight indentation mark was apparent on the front face of all the test specimens at the impact site.

Figure 4-3 shows the C-scan damage maps obtained from the composite specimens, impacted at energy levels of 4.5, 7.5 and 10.5 J. At each impact energy, the three replicate CF/PEEK and three replicate CF/Epoxy composite test specimens were subjected to ultrasonic C-scanning to identify any interlaminar delamination damage that might have resulted. The right-hand side scale in Figure 4-3 indicates the location of the delamination damage, as a function of the depth through the thickness of the specimen, where the dark red colour represents the front (impacted) face and the dark blue colour represents the rear (non-impacted) face of the composite specimen. The areal footprint of the damage detected is given in the top right-hand corner for each specimen and was determined by counting the number of pixels which had a colour which was not dark blue, as the rear surface simply reflects the ultrasound and appears as being dark blue in colour.

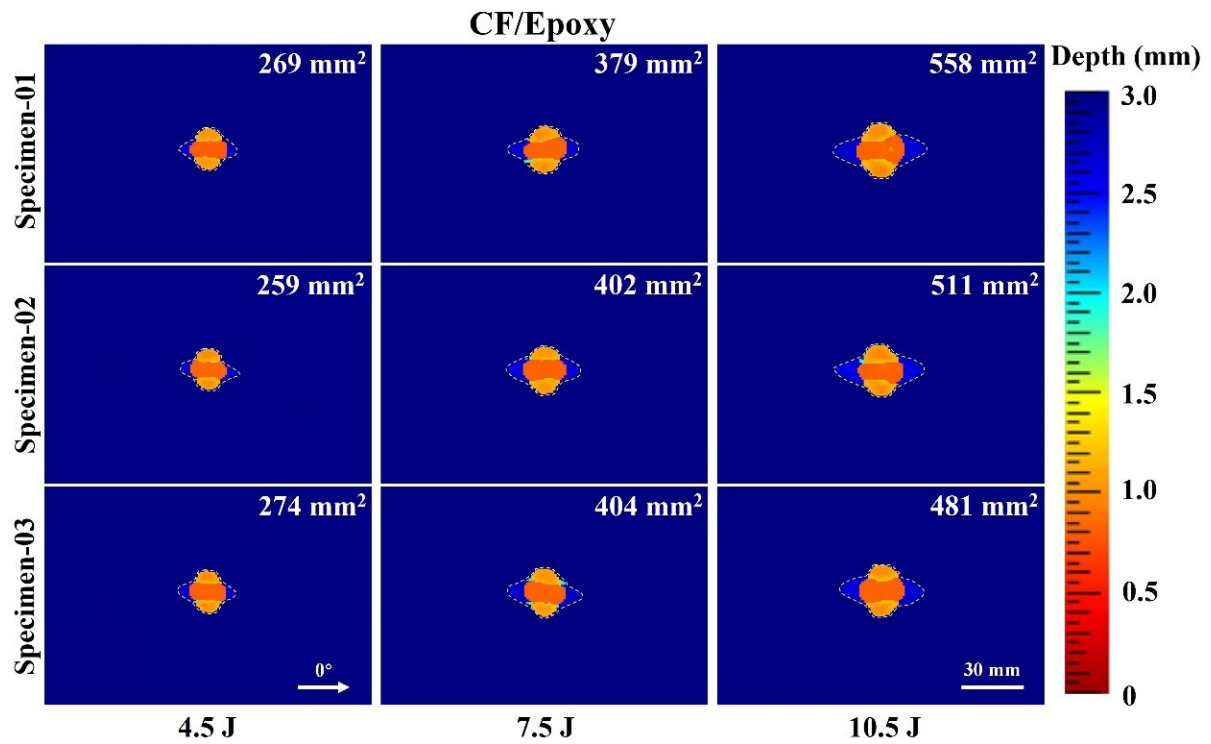
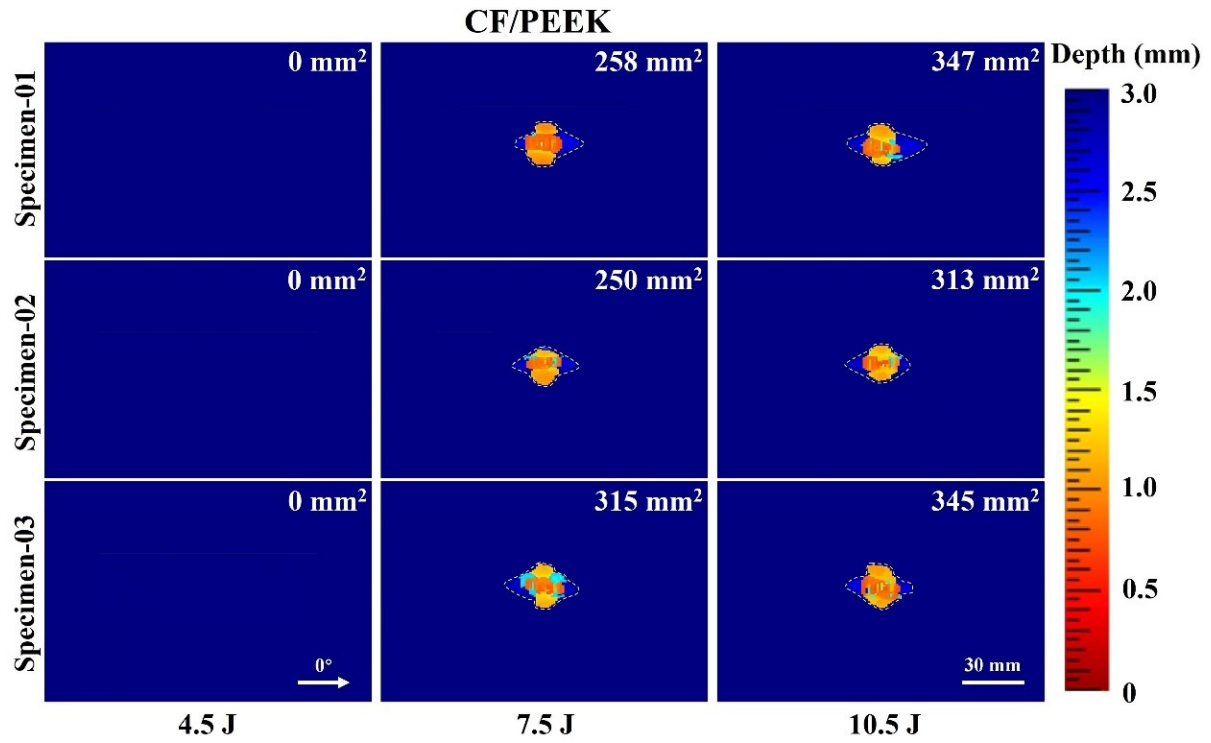


Figure 4-3. C-scan damage maps obtained from the (a) CF/PEEK and (b) CF/Epoxy replicate composite test specimens impacted at a low-velocity (i.e. the drop-weight test) at impact energies of 4.5 J, 7.5 J and 10.5 J. The corresponding impact velocities were 1.68, 2.16 and 2.56 m.s⁻¹, respectively.

In Figure 4-3(a) for the CF/PEEK composites impacted at 4.5 J it is apparent that there is no delamination damage, as expected from the force versus time curves shown in Figure 4-1(a). For the CF/PEEK specimens impacted with energies of 7.5 and 10.5 J delaminations are detected. Such delamination damage would be expected mainly to occur preferentially between the adjacent 0° and 90° ply directions in the $[0_3/90_3]_{2s}$ cross-ply lay-up, since at these interfaces the values of stiffness, in any one direction, of the adjacent ply layers change dramatically from one ply to the next. Thus, the interfacial stresses thereby generated tend to initiate delaminations [147]. Further, the area of the delaminations increases in extent as the impact energy is increased from 7.5 J to 10.5 J. In contrast, delaminations are observed to occur in the CF/Epoxy composites at all the impact energies employed and again the damage area steadily increases in size as the impact energy is increased, see Figure 4-3(b). For both types of composite, which have the same fibre lay-up of $[0_3/90_3]_{2s}$, the delamination damage that is furthest from the impact face is a slightly lighter blue in colour than the background and occurs at a depth of almost 3 mm. This damage is aligned along the longer dimension of the rectangular specimen, consistent with the direction of the 0° outer-ply layer since, in general, delaminations propagate in a direction determined by the orientation of the ply beneath the delamination [147]. The values of the average damage areas, i.e. the area of the delamination footprint, measured from the C-scan maps are summarised in Table 4-2. This comparison clearly shows that the CF/PEEK composite, when impacted at these relatively low velocities, suffers significantly less damage than the CF/Epoxy composite when the results are compared at the same energy level.

Table 4-2. Comparison of the footprint of the damage areas obtained from the low-velocity drop-weight tests as measured from the C-scans. (The error given is the coefficient of variation from the replicate experiments.)

Type of composite	Average damage area (mm ²)		
	4.5 J	7.5 J	10.5 J
CF/PEEK	0±0%	274±11%	335±5%
CF/Epoxy	267±6%	395±3%	517±6%

4.3.2 High-velocity gas gun impact tests

4.3.2.1 Visual and C-scan inspections

As in the case of the low-velocity drop weight tests discussed above, for all the composite test specimens no damage could be readily detected from a simple visual inspection of the specimens after being impacted using the gas gun test. However, a slight indentation mark was apparent on the front face of all the test specimens at the impact site.

Also, as for the low-velocity drop-weight tests, C-scan damage maps were determined for all the composite specimens that had been impacted using the high-velocity gas-gun and it is convenient to first discuss such results. Figure 4-4 shows C-scan damage maps obtained from the CF/PEEK and CF/Epoxy composite specimens using the gas-gun and impacted at an average high-velocity of $54.4 \pm 1.0 \text{ m.s}^{-1}$, which gives a corresponding average impact energy of $10.5 \pm 0.3 \text{ J}$. Duplicate impact tests were undertaken for each type of composite, as shown. The areal footprints of the damage for each test specimen are given in the top-right corner of the C-scan maps with the respective impact energy given in the top left-hand corner. These results demonstrate that the CF/PEEK and CF/Epoxy composites exhibit a similar pattern with values of the average damage area of $1462 \pm 37 \text{ mm}^2$ and $1898 \pm 395 \text{ mm}^2$, respectively. Thus, the CF/Epoxy composite clearly has the higher average damage area but with a significantly larger degree of scatter.

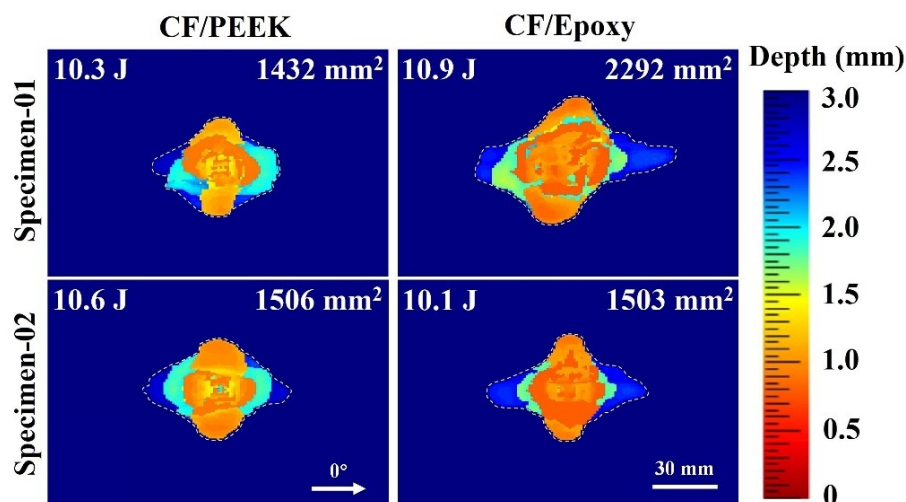


Figure 4-4. C-scan damage maps obtained from the CF/PEEK and CF/Epoxy duplicate composite test specimens impacted at a high-velocity (i.e. the gas-gun test). The average impact velocity and energy were $54.4 \pm 1.0 \text{ m.s}^{-1}$ and $10.5 \pm 0.3 \text{ J}$, respectively. (The actual impact energy for each test is shown in the top left-hand corner for each specimen.)

4.3.2.2 Out-of-plane displacement results

It is difficult to measure the load versus time relationship in a high-velocity gas-gun test and such curves can show considerable dynamic effects, as discussed below when the modelling of such tests is considered. For these reasons, measuring the displacement of the impacted specimen is typically undertaken. Figures 4-5 and 4-6 show the typical out-of-plane displacement contours obtained from the DIC measurements with respect to the time-scale of the impact event for the CF/PEEK and CF/Epoxy test specimens impacted at a velocity of $54.4 \pm 1.0 \text{ m.s}^{-1}$ and an impact energy of $10.5 \pm 0.3 \text{ J}$. For both composites, it is clear that the deformation is initially localised around the area of the composite specimen that was struck by the impactor. However, as the maximum out-of-plane displacement in the central region of the specimen increases, the deformation then extends elliptically, with the major axis of the ellipse aligned with the lengthwise dimension of the specimen, which is the direction of the 0° plies. For an impact energy of $10.5 \pm 0.3 \text{ J}$, the maximum out-of-plane displacements of the specimens are lower in the case of the high-velocity gas-gun test compared with the low-velocity drop-weight test, as may be seen from comparing Figures 4-2(c), 4-5 and 4-6. Indeed, these results reveal that the values from the high-velocity test are approximately 3 mm for the CF/PEEK composite and 3.4 mm for the CF/Epoxy composite, which may be compared with the values of approximately 5 mm for both types of composite from the low-velocity tests at an impact energy of 10.5 J. These observations will be discussed later in the context of the modelling studies in Chapter 5.

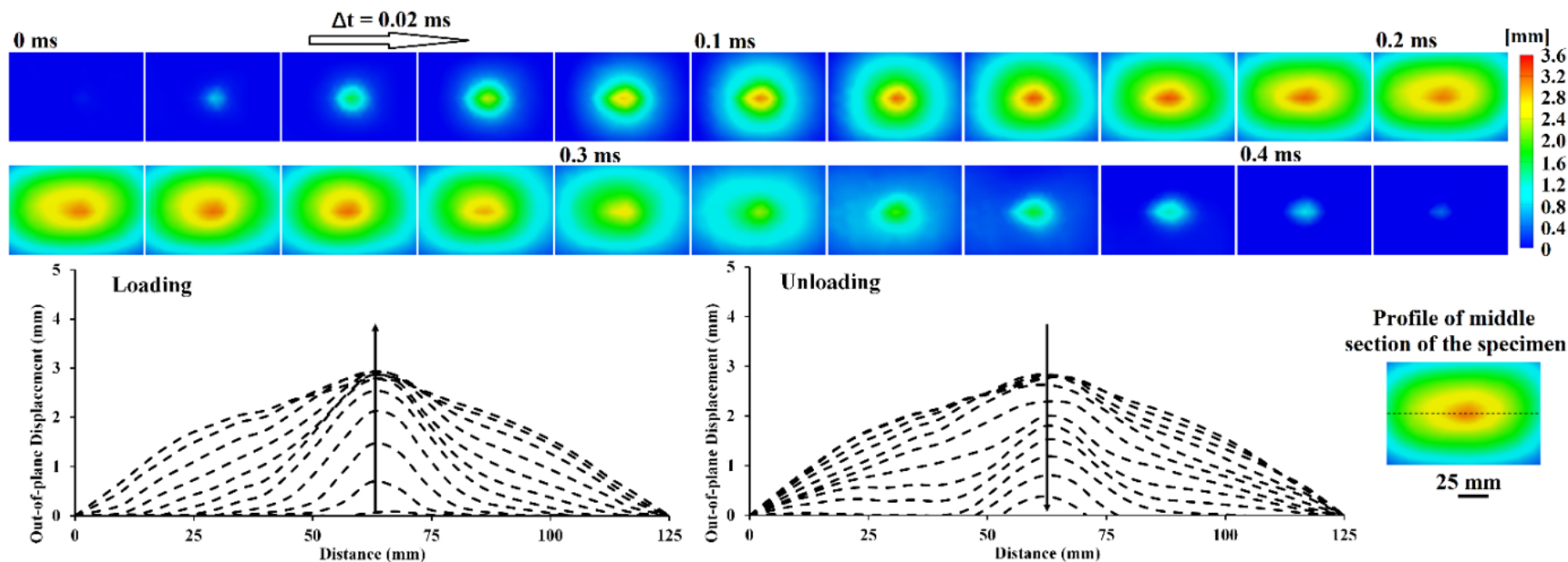


Figure 4-5. Maps and diagrams of the typical full-field out-of-plane displacements of the rear surface of the composite specimens measured for both the loading (upper map strips) and unloading (lower map strips) phases for the high-velocity gas-gun experiments at $54.4 \pm 1.0 \text{ m.s}^{-1}$ and an impact energy of $10.5 \pm 0.3 \text{ J}$ for the CF/PEEK composite test specimens. (For the displacement maps the value of the displacement is given by the scale bar, with the brightest-red colour representing in the range of about 3.2 to 3.6 mm. For the out-of-plane displacement diagrams the time interval between the dashed lines is 0.02 ms. The diagrams give the displacement values taken across the middle section of the specimen, as illustrated schematically in the scaled map to their right.)

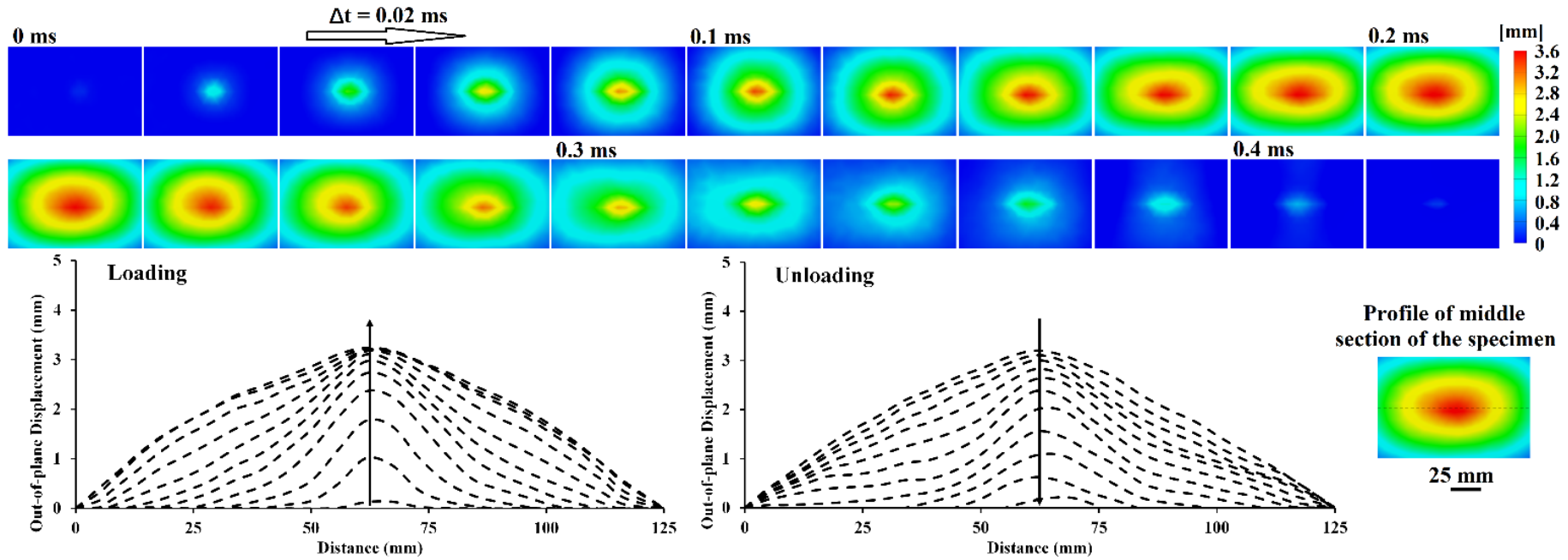


Figure 4-6. Maps and diagrams of the typical full-field out-of-plane displacements of the rear surface of the composite specimens measured for both the loading (upper map strips) and unloading (lower map strips) phases for the high-velocity gas-gun experiments at $54.4 \pm 1.0 \text{ m.s}^{-1}$ and an impact energy of $10.5 \pm 0.3 \text{ J}$ for the CF/Epoxy composite test specimens. (For the displacement maps the value of the displacement is given by the scale bar, with the brightest-red colour representing in the range of about 3.2 to 3.6 mm. For the out-of-plane displacement diagrams the time interval between the dashed lines is 0.02 ms. The diagrams give the displacement values taken across the middle section of the specimen, as illustrated schematically in the scaled map to their right.)

4.3.3 Comparison of damage inflicted by the low-velocity and high-velocity tests

In Table 4-3 results are shown comparing the damage areas, as measured from the C-scans, for the CF/PEEK and CF/Epoxy composite specimens from both the low-velocity (drop-weight) and the high-velocity (gas-gun) tests but with a similar impact energy of approximately 10.5 J.

Table 4-3. Comparison of the footprint of the damage areas as measured from the C-scans obtained from the low-velocity drop-weight and high-velocity gas-gun tests at a similar impact energy of 10.5 ± 0.3 J. (The error given is the coefficient of variation from the replicate experiments.)

Type of composite	Damage initiation load (N)	
	at 2.56 m.s^{-1}	at $54.4 \pm 1.0 \text{ m.s}^{-1}$
CF/PEEK	$335 \pm 5\%$	$1462 \pm 3\%$
CF/Epoxy	$517 \pm 6\%$	$1898 \pm 21\%$

Firstly, as noted previously, for the low-velocity tests at 2.56 m.s^{-1} then the CF/Epoxy composites suffer a significantly greater extent of damage than the CF/PEEK composites. Secondly, at the higher test velocity of $54.4 \pm 1.0 \text{ m.s}^{-1}$ the average damage area for the CF/Epoxy composites is again greater than for the CF/PEEK composites. Notwithstanding, as was noted above, the CF/Epoxy composite specimens show a relatively large degree of large scatter for the values recorded for the damage areas from the C-scan tests, compared with the far more consistent values for the CF/PEEK composite. Thus, it cannot be definitely stated that the CF/Epoxy composite exhibited statistically significantly more damage than the CF/PEEK composite when they were both tested using an impact energy of 10.5 ± 0.3 J, i.e. at the higher test velocity of $54.4 \pm 1.0 \text{ m.s}^{-1}$. Thirdly, both the CF/PEEK and CF/Epoxy composite specimens clearly exhibit very significantly larger damage areas when subjected to the high-velocity tests of approximately $54.4 \pm 1.0 \text{ m.s}^{-1}$, when compared with the damage areas associated with the low-velocity tests at 2.56 m.s^{-1} . This observation undoubtedly arises from several causes, since in the high-velocity tests (a) the CF/PEEK and CF/Epoxy composite specimens have excessive contact pressure between the projectile, i.e. less time to absorb the impact energy, and therefore the deformation in the specimen is initially more localised around the impact site with an

increased curvature of the specimen local to the impact site, which will induce more extensive localised delaminations throughout the composite plies, (b) these extensive localised delaminations will then readily propagate along the various $0^\circ/90^\circ$ ply interfaces and (c) the values of the interlaminar and matrix fracture energies will be somewhat lower for both the CF/PEEK and CF/Epoxy composites due to the relatively high strain-rate effects associated with the high-velocity test and therefore delaminations and matrix damage will initiate and evolve more readily in such tests. The overall consequence is that the damage inflicted is more extensive in the high-velocity tests compared with the low-velocity tests, as indeed is observed from the greater area of delaminations in the C-scans of the high-velocity tests, see Figures 4-3 and 4-4, and Table 4-3.

4.4 Summary

The impact behaviour of a thermoplastic CF/PEEK composite and a toughened-thermoset CF/Epoxy composite, which possessed the same cross-ply lay-up of $[0_3/90_3]_{2s}$, has been investigated. The relatively low-velocity impact experiments, at impact velocities of 1.68, 2.16 and 2.56 m.s^{-1} giving impact energies of 4.5, 7.5 and 10.5 J, were undertaken using a drop-weight test. For the high-velocity experiments, a gas-gun was employed using an impact velocity of $54.4 \pm 1.0 \text{ m.s}^{-1}$ which resulted in an impact energy of $10.5 \pm 0.3 \text{ J}$. In all the tests a rigid, metallic impactor was used with a hemispherical-shaped head. The damage inflicted in the composites was assessed by both a visual inspection and by ultrasonic C-scanning.

For the low-velocity drop-weight tests, the experimental results demonstrated that, at the same impact energy, the CF/PEEK composite always possessed a superior impact performance compared with the CF/Epoxy. This was reflected by the CF/PEEK composite requiring a higher load to initiate damage and a smaller area of delamination damage being present in the CF/PEEK composite post-impact.

In the high-velocity tests using the gas-gun, the results revealed that the CF/PEEK and CF/Epoxy composites exhibited a similar damage pattern with values of the average damage area of $1462 \pm 37 \text{ mm}^2$ and $1898 \pm 395 \text{ mm}^2$, respectively. Thus, the CF/Epoxy composite clearly exhibits the higher average damage area but with a significantly larger degree of scatter. Furthermore, these damage areas are significantly greater than those recorded from the low-velocity drop-weight test, at the same impact energy of $10.5 \pm 0.3 \text{ J}$, by a factor of about four for both types of composite. The out-of-plane displacements of the composite specimens were

measured as a function of time during the impact event with the maximum values being approximately 3 mm for the CF/PEEK composite and 3.4 mm for the CF/Epoxy composite.

Therefore, the present detailed study has revealed the effects of the impact energy and velocity on the behaviour of a thermoplastic- and a toughened thermoset-matrix CFRP composite, where both have the same cross-ply fibre-architecture.

Chapter 5. Numerical modelling on unidirectional thermoplastic and thermoset laminates under low-velocity and high-velocity impact loading

5.1 Introduction

The present chapter develops a numerical model on the behaviour of continuous carbon-fibre/polymer matrix composites subjected to a relatively low-velocity or high-velocity impact, using a rigid, metallic impactor (experiments detailed in Chapter 4). This computationally-efficient, finite-element model that has been developed is generally successful in capturing the essential details of the impact test and the impact damage in the composites, and has been used to predict the loading response of the composites under a high-velocity impact.

5.2 Composite damage model

5.2.1 Brief overview of the damage model

It is obviously of great interest to attempt to model the experimental result from the impact tests described in Chapter 4 and, once validated, to use the model to undertake predictive studies. The impact event for the relatively hard impactor striking the CFRP

Note: The work presented in this chapter was previously published in the Journal of Materials Science as “The behaviour of thermoplastic and thermoset carbon-fibre composites subjected to low-velocity and high-velocity impacts”.

composite specimen was therefore modelled using the ‘Abaqus/Explicit 2018’ finite-element (FE) code [148]. An overall flow chart of the basic two-dimensional FE methodology is shown in Figure 5-1 and the model is shown schematically in Figure 5-2.

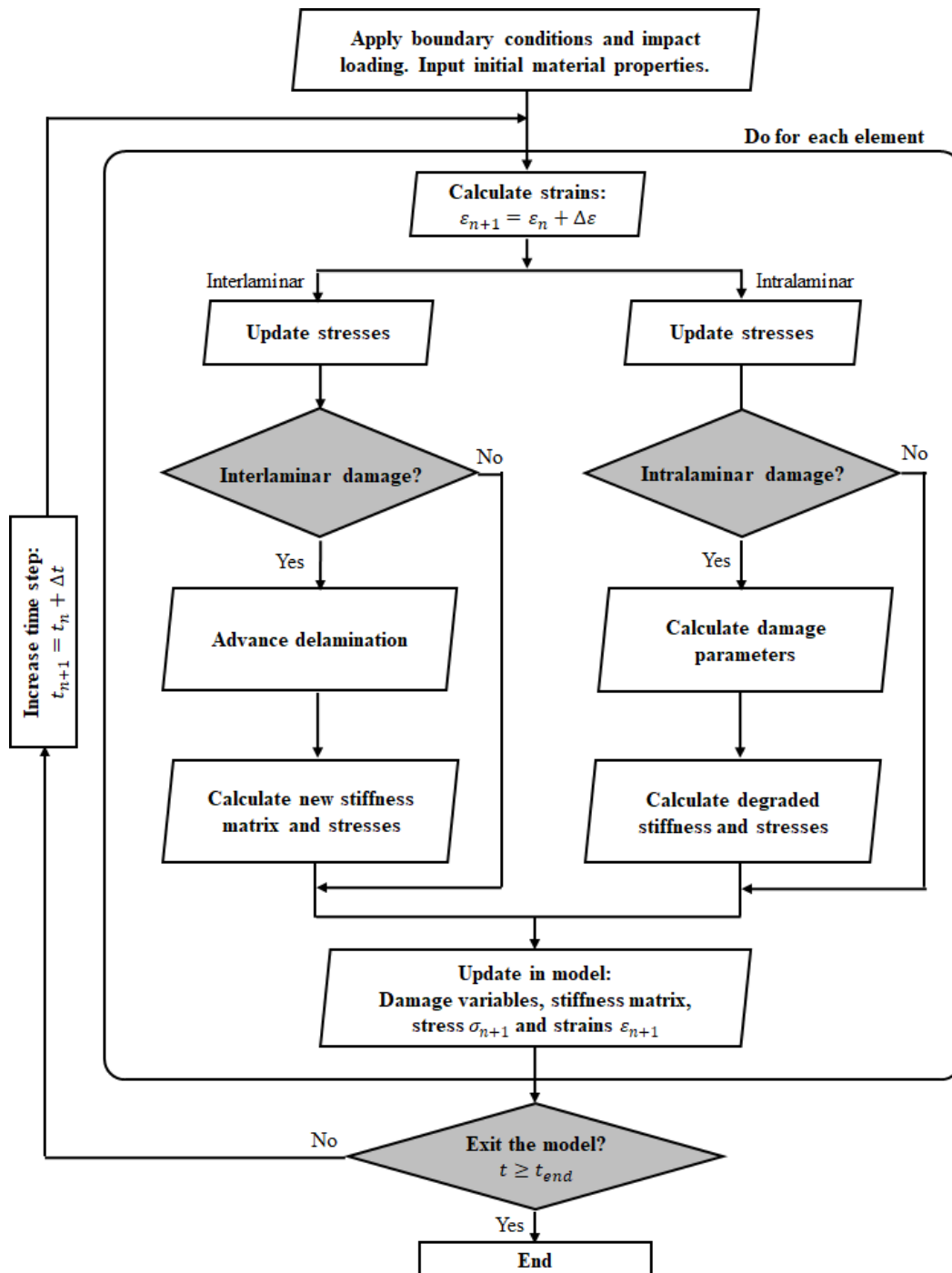


Figure 5-1. The implementation of the FE model showing schematically the flow chart, for one computation time-step, for a single element for modelling the interlaminar and intralaminar damage

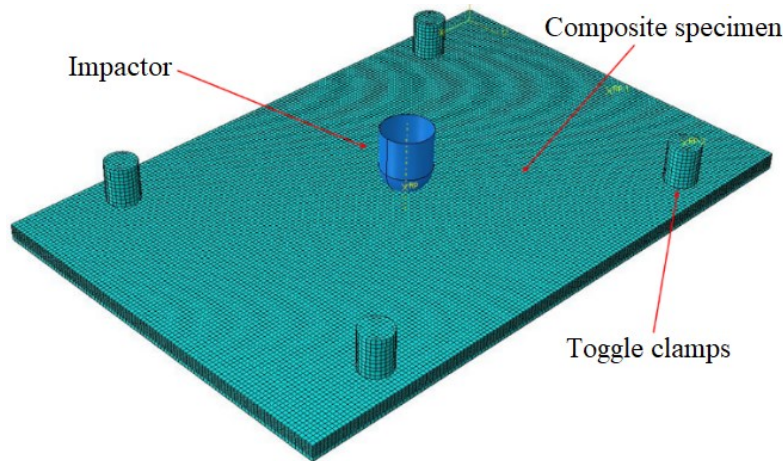


Figure 5-2. Schematic of the FE model.

The impactor was modelled as a spherically-shaped, rigid surface, with a reference lumped mass of 3.2 kg or 7.1 g for the low- and high-velocity tests, respectively. The composite specimen was defined using continuum shell elements and thus only the in-plane material properties are required for the numerical modelling. The elements had a size of $1 \text{ mm} \times 1 \text{ mm}$ and this mesh size was found to give mesh-independent results. Cohesive surfaces were defined at the $0^\circ/90^\circ$ interfaces to capture the interlaminar damage. A general contact algorithm was employed to govern the global contact and the friction coefficients for the metal/composite-ply and composite-ply/composite-ply interfaces were set as 0.2 and 0.25, respectively [149,150]. The computational accuracy was set as ‘double procession’ to reduce the accumulation of error during the simulation. The FE model for the low-velocity drop-weight test was typically run as individual time steps, Δt , of 0.06ms for about 100 steps, which represents the order of the experimentally measured time of about 6 ms for the complete impact event to take place. Whilst, for the high-velocity gas-gun test, the model was typically run as individual time steps, Δt , of 0.005ms for about 100 steps, which represents the order of the experimentally measured time of about 0.5 ms for the complete impact event to take place. A computational time-step was performed for every appropriate single element in the FE model. The modelling runs were stopped when the defined total time for the impact event had expired. Computations were undertaken using 32 CPUs on a ‘Linux Cluster’ with a run time of 12 to 15 hours. The material coordinate system for the CFRP laminate was defined with the lengthwise direction of the specimen being the 11-direction and the widthwise direction being the 22-direction. For the 0° plies, the fibre direction was aligned with the 11-direction, and for the 90° plies, the fibres were aligned in the 22-direction.

5.2.2 The intralaminar damage model

5.2.2.1 Initiation of intralaminar damage

The model for predicting the initiation of any intralaminar damage was based upon Hashin's 2-D theory [151,152] which assumes that the undamaged composite exhibits linear-elastic behaviour. In Hashin's damage model, four different types of damage mechanisms, which arise from tensile fibre failure, compressive fibre failure, tensile matrix failure and compressive matrix failure, are employed to capture the initiation of intralaminar damage in the unidirectional-fibre sub-ply. The general forms of the damage criteria in Hashin's approach to model the initiation of these four different types of damage are given by:

for tensile fibre failure ($\hat{\sigma}_{11} \geq 0$):

$$F_f^t = \left(\frac{\hat{\sigma}_{11}}{X^T} \right)^2 \quad (5-1)$$

for compressive fibre failure ($\hat{\sigma}_{11} < 0$):

$$F_f^c = \left(\frac{\hat{\sigma}_{11}}{X^C} \right)^2 \quad (5-2)$$

for tensile matrix failure ($\hat{\sigma}_{22} \geq 0$):

$$F_m^t = \left(\frac{\hat{\sigma}_{22}}{Y^T} \right)^2 \quad (5-3)$$

for compressive matrix failure ($\hat{\sigma}_{22} < 0$):

$$F_m^c = \left(\frac{\hat{\sigma}_{22}}{2S^T} \right)^2 + \left[\left(\frac{Y^C}{2S^T} \right)^2 - 1 \right] \frac{\hat{\sigma}_{22}}{Y^C} + \left(\frac{\hat{\tau}_{12}}{S^L} \right)^2 \quad (5-4)$$

In the above equations, the indices on the terms F_f^t , F_f^c , F_m^t and F_m^c represent the four types of damage of tensile fibre failure, compressive fibre failure, tensile matrix failure and compressive matrix failure, respectively, and failure is predicted to occur when $F \geq 1$. The parameters, X^T and X^C denote the tensile and compressive strengths in the longitudinal fibre-direction, respectively. The terms Y^T and Y^C are the tensile and compressive strengths in the transverse direction, respectively; S^L and $S^T = Y^C/2$ denote the shear strengths in the longitudinal and transverse directions to the fibres, respectively; and the term $\hat{\sigma}_{11}$, $\hat{\sigma}_{22}$ and $\hat{\tau}_{12}$ are components of the effective stress tensor, $\hat{\sigma}$, that are used to evaluate the above criteria.

5.2.2.2 The evolution of intralaminar damage

Corresponding to the damage initiation criteria defined in Hashin's theory, Equations (5-1) - (5-4), four damage parameters, d_f^t , d_f^c , d_m^t and d_m^c , were implemented in the damage evolution model, which arise from tensile fibre failure, compressive fibre failure, tensile matrix failure and compressive matrix failure, respectively. These parameters have the value of 0 when the element is undamaged and 1 when fully damaged. A general form of the damage variable, d , once a particular damage mechanism initiates, is given by [148]:

$$d = \frac{\varepsilon^f (\varepsilon - \varepsilon^0)}{\varepsilon (\varepsilon^f - \varepsilon^0)} \quad (5-5)$$

where the strain, ε , is the applied strain and the strain values ε^0 and ε^f are those corresponding to the initiation of damage and final failure, respectively. For tensile fibre or compressive fibre failure, the terms ε , ε^0 and ε^f are assigned to be $\varepsilon = \varepsilon_{11}$, $\varepsilon^0 = \varepsilon_{11}^0$ and $\varepsilon^f = \varepsilon_{11}^f$, respectively; with the second superscripts of t or c being used to indicate tensile or compressive stresses, respectively. Similarly, for tensile or compressive matrix failure, the terms ε , ε^0 and ε^f are assigned to be $\varepsilon = \varepsilon_{22}$, $\varepsilon^0 = \varepsilon_{22}^0$ and $\varepsilon^f = \varepsilon_{22}^f$, respectively; with the second superscripts of t or c being used to indicate tensile or compressive stresses, respectively.

Now, the applied strain, ε , may be deduced from interrogating the FE output for any element for a given time step. In the damage evolution model, the values of the initial failure strains, ε^0 , are equal to the strain values corresponding to damage initiation, which may be

directly obtained from the computation via implementing Equations. (5-1) - (5-4). The final failure strains, ε^f , are given by [153]:

for tensile fibre failure:

$$\varepsilon_{11}^{ft} = 2G_{Ic}|_{ft}/(X^T l_c) \quad (5-6)$$

for compressive fibre failure:

$$\varepsilon_{11}^{fc} = 2G_{Ic}|_{fc}/(X^C l_c) \quad (5-7)$$

for tensile matrix failure:

$$\varepsilon_{22}^{ft} = 2G_{Ic}|_{mt}/(Y^T l_c) \quad (5-8)$$

for compressive matrix failure:

$$\varepsilon_{22}^{fc} = 2G_{Ic}|_{mc}/(Y^C l_c) \quad (5-9)$$

where the terms $G_{Ic}|_{ft}$ and $G_{Ic}|_{fc}$ are the tensile and compressive intralaminar ply fracture energies in the longitudinal fibre-direction, and $G_{Ic}|_{mt}$ and $G_{Ic}|_{mc}$ are the tensile and compressive interlaminar ply fracture energies in the transverse to the fibre-direction. The characteristic length, l_c is equal to the edge length of the element.

Three damage variables, d_f , d_m and d_s , which reflect fibre damage, matrix damage and shear damage, respectively, may then be derived from these damage parameters, d_f^t , d_f^c , d_m^t and d_m^c , as follows [148]:

for fibre damage:

$$d_f = \begin{cases} d_f^t, & \hat{\sigma}_{11} \geq 0 \\ d_f^c, & \hat{\sigma}_{11} < 0 \end{cases} \quad (5-10)$$

for matrix damage:

$$d_m = \begin{cases} d_m^t, & \hat{\sigma}_{22} \geq 0 \\ d_m^c, & \hat{\sigma}_{22} < 0 \end{cases} \quad (5-11)$$

for shear damage:

$$d_s = 1 - (1 - d_f^t)(1 - d_f^c)(1 - d_m^t)(1 - d_m^c) \quad (5-12)$$

Now, prior to damage initiation the composite laminate is taken to be linear-elastic, with the stiffness matrix of a plane-stress orthotropic material. Once the damage initiates and starts to evolve the response of the material is computed from:

$$\sigma = C_d \varepsilon \quad (5-13)$$

where C_d is the damaged elasticity matrix, which has the form [148]:

$$C_d = \frac{1}{D} \begin{bmatrix} (1 - d_f)E_{11} & (1 - d_f)(1 - d_m)v_{21}E_{11} & 0 \\ (1 - d_f)(1 - d_m)v_{12}E_{22} & (1 - d_m)E_{22} & 0 \\ 0 & 0 & (1 - d_s)G_{12}D \end{bmatrix} \quad (5-14)$$

where $D = 1 - (1 - d_f)(1 - d_m)v_{12}v_{21}$, and E_{ii} ($i = 1, 2$) is the elastic modulus in the longitudinal or transverse directions, G_{12} is the shear modulus and v_{ij} ($i, j = 1, 2$ $i \neq j$) are the Poisson's ratios. Now, the values of the damage variables, d_f , d_m and d_s , reflect the current state of fibre damage, matrix damage and shear damage, respectively, and may be calculated

from Equations (5-10) to (5-12). Thus, the degraded stresses acting in any element for any time step for an applied strain, ε , can now be computed from Equations (5-13) and (5-14). These degraded stresses and strains may then be updated, as being the ‘new model state’, for a given element in the next time step of the run of the FE model, see Figure 5-1. For the simulations of the extent of intralaminar damage as a function of the time-scale of the impact event that are deduced from the model then, following earlier work [135], a value for the damage parameter of equal to, or greater, than 0.9 is used to define the relatively intense intralaminar damage, i.e. to calculate the areas indicated by a red colour in the figure shown later. This value is determined based on a number of attempts from previous work to capture sufficient intralaminar damage without interfusing excessive element distortion.

From the above discussions, it is evident that if any of the four intralaminar damage mechanisms have been activated as determined from the Hashin criteria, see Equations (5-1) - (5-4), then the elastic properties in those elements in the FE model will start to degrade, as defined above in Equation (5-14), which is based upon the damage parameters calculated in Equations (5-6) to (5-12). Finally, when any of these three damage parameters meet the condition, as stated in Equation (15) below, then these damaged elements are deleted from the model [148]:

$$\text{Delete element if: } \begin{cases} d_f > 0.99 \\ d_m > 0.99 \\ d_s > 0.99 \end{cases} \quad (5-15)$$

5.2.3 The interlaminar damage model

Interlaminar damage typically involves the initiation and growth of delamination between the plies that make up the composite laminate and this was captured using the Abaqus built-in surface-based cohesive (i.e. interface) element using an energy-release approach [148]. The interface element was described via a cohesive (i.e. damage) surface law [118,131,154–156] where the traction, σ , is a function of the displacement, δ , as shown schematically in Figure 5-3, and is in the form of a bilinear cohesive law for a linear-softening material model. This damage law is divided into two steps. Before the initiation of any delamination, the relationship possesses linear-elastic behaviour. Once the damage criterion is satisfied, at a value of the displacement of δ^0 , the cohesive stiffness degrades linearly until separation of the

interface, i.e. the delamination now propagates when the maximum failure displacement of δ^f is attained. The energy under the bilinear cohesive law is equivalent to the interlaminar fracture energy, G_c , and the evaluation of this term, and of the initial stiffness, k , of the interface element, are discussed below. Finally, the initiation and growth of any intralaminar damage, as discussed above, significantly influences the extent of interlaminar damage and hence these two damage modes were modelled to be interactive in the Abaqus simulation, see Figure 5-1.

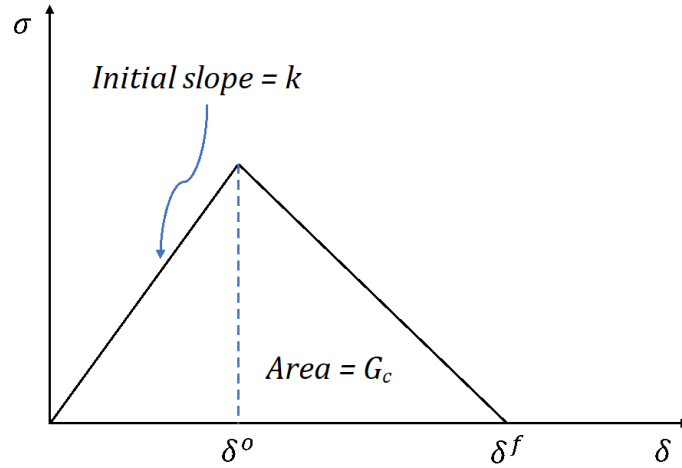


Figure 5-3. Schematic of the bilinear cohesive surface law.

5.2.3.1 The initiation of interlaminar damage

To determine when interlaminar damage initiates, a quadratic-stress criterion was again implemented within the Abaqus FE code, and is given by [148,157]:

$$\left(\frac{\langle\sigma_n\rangle}{\sigma_n^0}\right)^2 + \left(\frac{\tau_s}{\tau_s^0}\right)^2 + \left(\frac{\tau_t}{\tau_t^0}\right)^2 \geq 1 \quad (5-16)$$

where σ_n represents the current stress that is acting normal to the ply, and τ_s and τ_t represent the current shear stresses that are acting on the ply. The values of the cohesive strengths, σ_n^0 , τ_s^0 and τ_t^0 denote the interface shear and tensile strengths, respectively. Employing the quadratic-stress criterion, the value of the damage initiation displacement, δ^o , see Figure 5-3, may be calculated.

5.2.3.2 The evolution of interlaminar damage

The embedded cohesive surface law, see Figure 5-3, requires a value of the interlaminar fracture energy, G_c , and this represents the area under the bilinear law. The energy-based Benzeggagh-Kenane (B-K) [84,148,158] mixed-mode propagation criterion was used to derive a value G_c for the growth of the delamination between the composite plies, as given by:

$$G_c = G_{Ic} + (G_{IIc} - G_{Ic}) \left[\frac{G_{II}}{G_I + G_{II}} \right]^{\eta_{BK}} \quad (5-17)$$

where G_{Ic} is the Mode I (opening tensile) interlaminar fracture energy, G_{IIc} is the Mode II (in-plane shear) interlaminar fracture energy and η_{BK} is the B-K Mixed-mode interaction exponent. The values of all these terms may be experimentally measured [83,84,158,159]. The parameters G_I and G_{II} are the current Mode I and Mode II energy-release rates, respectively, as calculated from the FE code by multiplying the relevant local stress by its conjugate displacement. The stiffness, k , of the cohesive law is given by [155]:

$$k = \frac{\alpha E_{22}}{t_e} \quad (5-18)$$

where α is a constant much larger than unity, i.e. $\alpha \gg 1$, and t_e is the thickness of an adjacent ply. From the known values of k , δ^0 and G_c the cohesive surface law, as shown in Figure 5-3, may now be completely and quantitatively described. Finally, complete fracture of the interface element is assumed to occur, and delamination results, when the cohesive traction vanishes at the end of the degradation step. That is when the displacement, δ , of the interface element, as determined in the FE code, attains the criterion:

$$\delta \geq \delta^f \quad (5-19)$$

where δ^f is the displacement of the element at failure, see Figure 5-3. For the simulations of the location and extent of interlaminar damage as a function of the impact velocity and energy,

and the time-scale of the impact event, that are deduced from the model then, following earlier work [135], a value corresponding to the displacement ratio of δ/δ^f of equal to, or greater, than 0.9 is used, i.e. to calculate the areas shown by a red colour in the figures shown later.

5.3 Modelling results

5.3.1 Material properties

The material properties of the CF/PEEK and CF/epoxy composites required for the FE modelling studies are given in Table 5-1. These properties were typically measured at strain-rates in the range of 10^{-3} s^{-1} to 10^0 s^{-1} .

5.3.2 Simulations of the low-velocity drop weight impact test results

As for the experimental tests, simulations were undertaken at three different energy levels of 4.5, 7.5 and 10.5 J, which correspond to the impact velocities of 1.68, 2.16 and 2.56 m.s^{-1} , respectively. The strain-rate for the composite test specimens for the low-velocity drop-weight tests is about 3 s^{-1} for an impact velocity of 2.56 m.s^{-1} . It was considered that this still relatively low value of strain-rate would not lead to significantly different values of the material properties to those shown in Table 5-1 and these properties were therefore used in the FE modelling studies for the low-velocity impact tests.

Table 5-1. Input properties for the FE modelling studies of the CF/PEEK and CF/Epoxy composites [133,141,147,153,160–164].

Property	Unidirectional CF/PEEK sub-ply	Unidirectional CF/epoxy sub-ply
Moduli (GPa)	$E_{11} = 127; E_{22} = 10.3; G_{12} = 5.7$	$E_{11} = 130; E_{22} = 7.7; G_{12} = 4.68$
Poisson's ratio	$\nu_{12} = 0.3$	$\nu_{12} = 0.3$
	$X^T = 2070; Y^T = 85$	$X^T = 1950; Y^T = 75$
Strength values (MPa)	$X^C = 1360; Y^C = 276$	$X^C = 1015; Y^C = 220$
	$S^L = 186$	$S_{12} = 150$
Ply fracture energies (kJ.m⁻²)	$G_{Ic} _{ft} = 218; G_{Ic} _{fc} = 104$	$G_{Ic} _{ft} = 133; G_{Ic} _{fc} = 40.0$
	$G_{Ic} _{mt} = 1.7; G_{Ic} _{mc} = 2.0$	$G_{Ic} _{mt} = 0.5; G_{Ic} _{mc} = 1.6$
Interlaminar fracture energies (kJ.m⁻²)	$G_{Ic} = 1.7; G_{IIc} = 2.0$	$G_{Ic} = 0.5; G_{IIc} = 1.6$
Benzeggagh-Kenane coefficient mode-mix exponent	$\eta_{BK} = 1.09$	$\eta_{BK} = 1.45$
Cohesive strength (MPa)	$\sigma_n^0 = 43.0; \tau_s^0 = \tau_t^0 = 50.0$	$\sigma_n^0 = 43.0; \tau_s^0 = \tau_t^0 = 50.0$
Cohesive law stiffness (MPa.mm⁻¹)	$k = 6.4 \times 10^5$	$k = 6.4 \times 10^5$

The predicted damage maps from the modelling studies for the three different impact energies are compared with the corresponding experimental results in Figures 5-4 and 5-5 for the CF/PEEK and CF/Epoxy composites, respectively. (The experimentally measured maps which are shown represent the mean value of the damage area that was measured from the replicate tests, see Figure 4-3 in Chapter 4.)

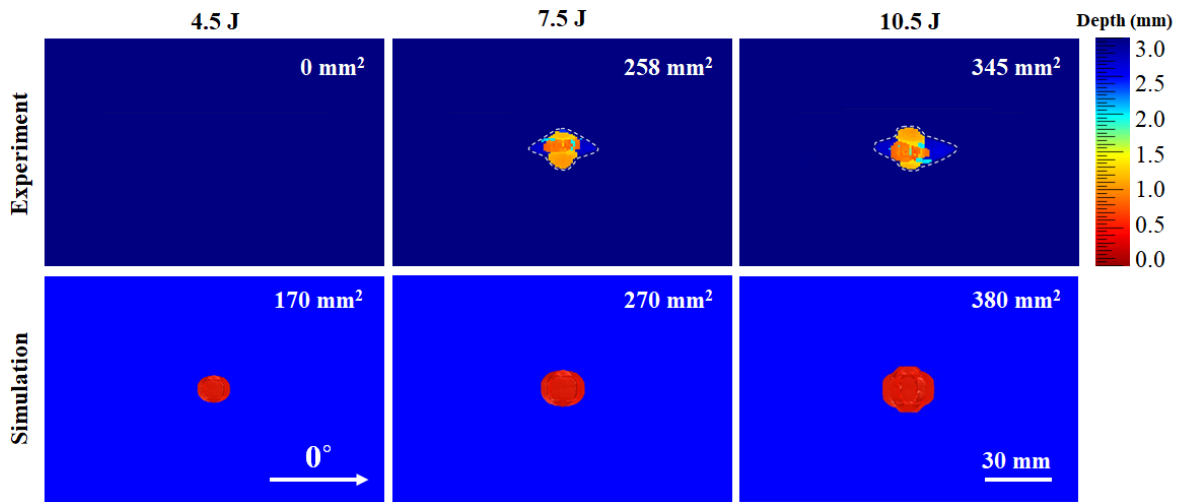


Figure 5-4. Experimental and simulated (red-coloured) footprints of the damage area for the CF/PEEK composite specimens for impact energies of 4.5, 7.5 and 10.5 J for the low-velocity drop-weight impact test. (See Figure 4-3 in Chapter 4 for the explanation of the experimental C-scan depth scale, etc.)

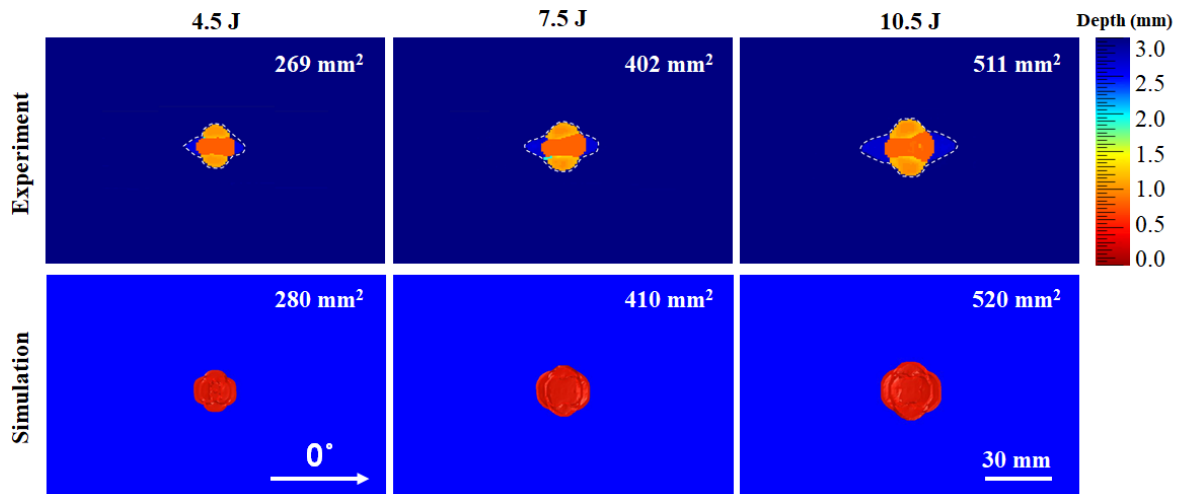


Figure 5-5. Experimental and simulated (red-coloured) footprints of the damage area for the CF/Epoxy composite specimens for impact energies of 4.5, 7.5 and 10.5 J for the low-velocity drop-weight impact test. (See Figure 4-3 in Chapter 4 for the explanation of the experimental C-scan depth scale, etc.)

As may be seen, the increase in the extent of damage as the impact energy is increased for a given composite is in good agreement with the damage maps that were experimentally measured using the C-scan test. The general shapes and the locations of the damage areas from the modelling studies are also in good agreement with the experimental results. The one exception to the good agreement between the modelling and experimental results is the prediction of damage occurring in the CF/PEEK composite specimen at an impact energy of 4.5 J, when none was experimentally observed, see Figure 5-4. This was most likely due to the predicted stress required to initiate such damage, as deduced from Equation (5-16), at the lowest impact energy used for the CF/PEEK composite being somewhat lower than that actually required. Hence, interlaminar damage is predicted when none is actually observed experimentally at the lowest values of impact velocity and energy used in the present study. Further, as may be seen, the FE model lacks the fidelity to reproduce the exact shape of the experimentally measured delaminations. Nevertheless, the quantitative predictions of the damage areas obtained from the FE simulation, demonstrate that the proposed model has the capability to predict reasonably well the overall area of damage due to a low-velocity impact.

The modelling results for the typical evolution of interlaminar (i.e. delamination) damage in the CF/PEEK composite specimens subjected to the drop-weight impact test is presented in the time-scale series of images in Figure 5-6(a) as a series of cross-section images obtained from the simulations at an impact velocity and energy of 2.56 m.s⁻¹ and 10.5 J, respectively. The damage evolution of intralaminar (i.e. matrix) damage in the CF/PEEK composite specimen for the same test conditions is shown in Figure 5-6(b). The intralaminar matrix damage, such as matrix cracking, arises from transverse stresses acting perpendicular to the fibres in the plies and there were no predictions of fibre-failure from the modelling studies during either the low-velocity, or the high-velocity, impact tests. Therefore, it appears that the stresses induced in the plies are insufficiently high to cause fibre-failure, which is in agreement with the experimental observations of the tested composite specimens. Finally, an interesting point to note is that the intralaminar matrix damage under the impactor is much more localised than the corresponding delamination damage.

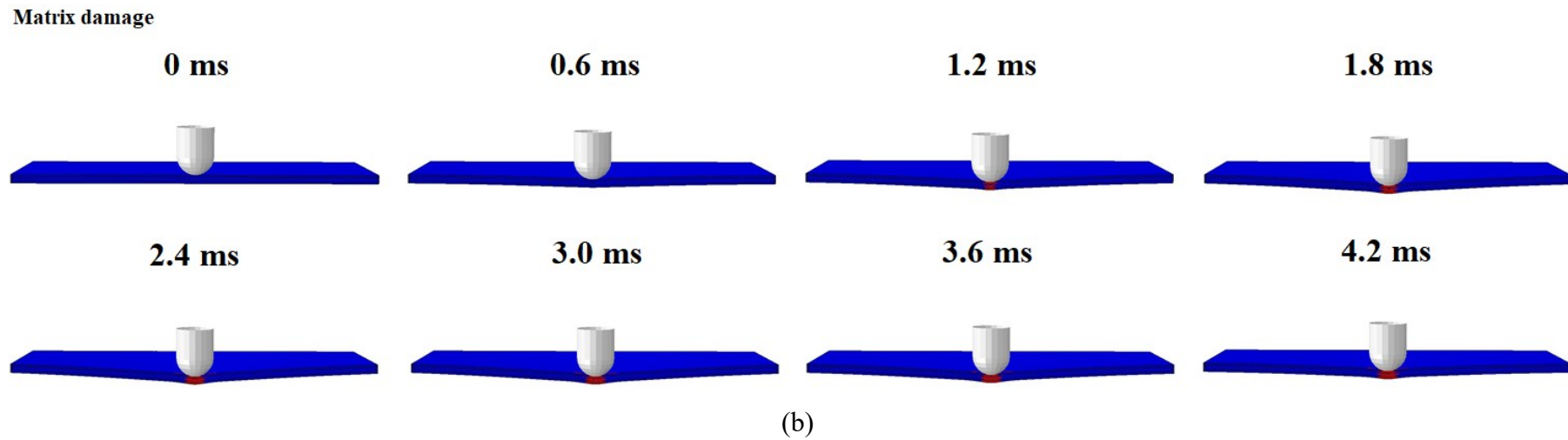
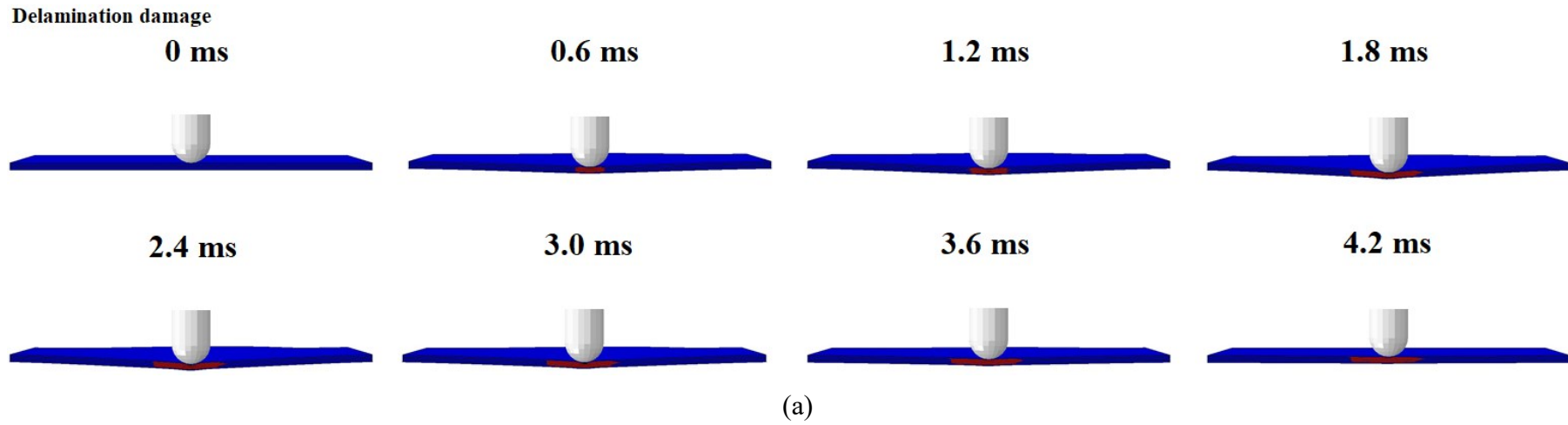


Figure 5-6. Simulations of the (a) interlaminar delamination (red-coloured) damage and (b) intralaminar matrix (red-coloured) damage as a function of time-scale for the CF/PEEK composites for low-velocity drop-weight tests at an impact energy and velocity of 10.5 J and 2.56 m.s⁻¹, respectively.

The load versus time curves obtained from the modelling studies and the experimental tests for the CF/PEEK and CF/Epoxy composites are compared in Figures 5-7 and 5-8, respectively. A very good agreement is obtained between the experimental and modelling results for all three energy levels for both types of composite. Indeed, the modelling results even capture the small amplitude, sinusoidal oscillations on the rising part of load versus time curves which are indicative of mass-spring oscillations [144–146], as discussed earlier.

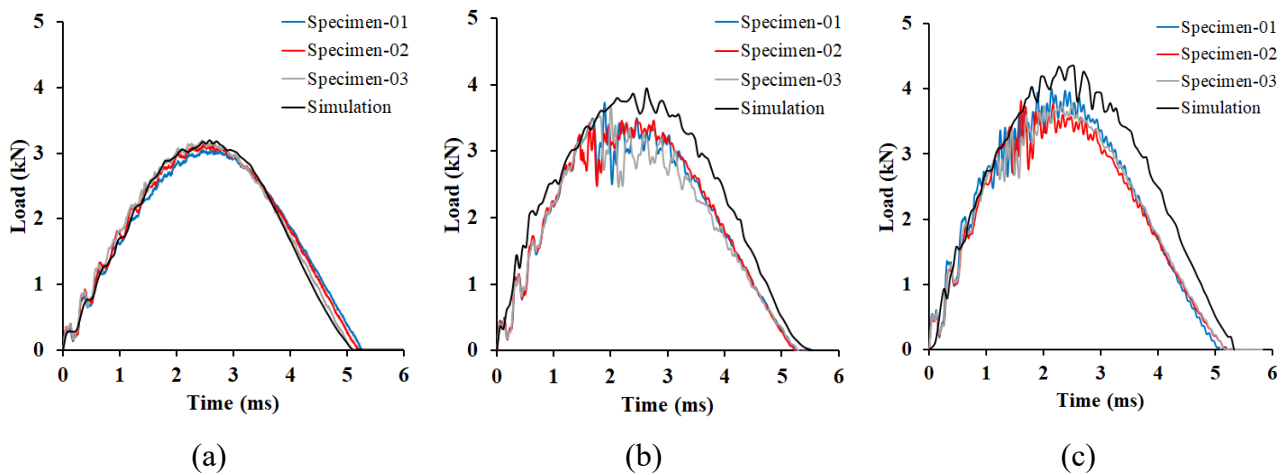


Figure 5-7. Experimental and simulated load versus time curves for the low-velocity drop-weight tests for the CF/PEEK composite specimens for: (a) 4.5, (b) 7.5 and (c) 10.5 J impact energy levels.

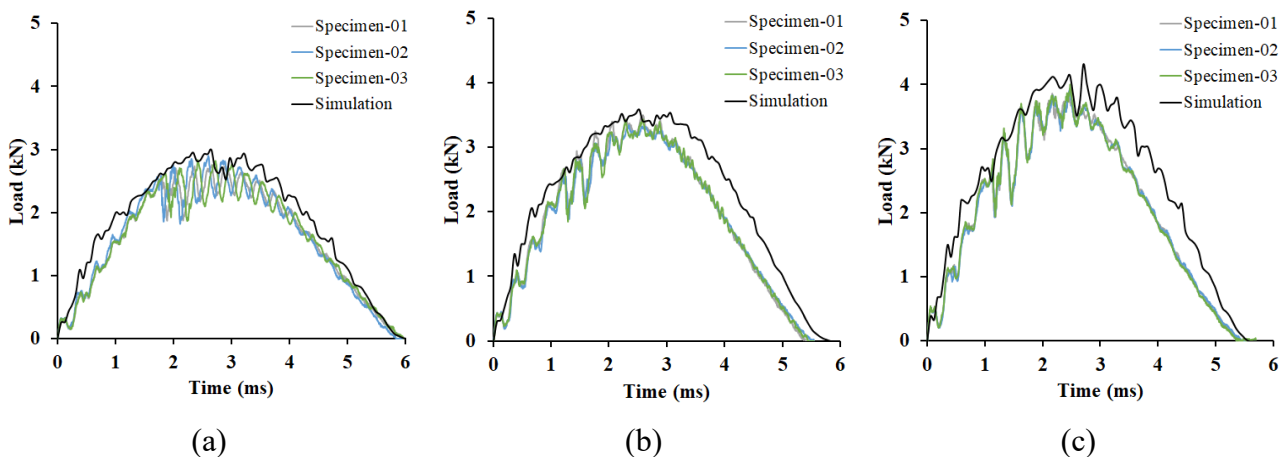


Figure 5-8. Experimental and simulated load versus time curves for the low-velocity drop-weight tests for the CF/Epoxy composite specimens for: (a) 4.5, (b) 7.5 and (c) 10.5 J impact energy levels.

The key experimental and numerical modelling results are compared in Table 5-2 and these results highlight the good agreement for both types of composite. Indeed, the modelling

studies clearly agree with the experimental observations that the CF/PEEK composite, when impacted at these relatively low velocities, suffers significantly less damage than the CF/Epoxy composite when the results are compared at the same energy level.

Table 5-2. Comparison of the experimental and numerically-simulated results obtained from the low-velocity drop-weight tests at an impact energy and velocity of 10.5 J and 2.56 m.s⁻¹, respectively. (The error given is the coefficient of variation from the replicate experiments.)

Items	CF/PEEK		CF/Epoxy	
	Experiment	Simulation	Experiment	Simulation
Maximum load (kN)	3.9±3%	4.3	3.9±2%	4.2
Average damage area (mm ²)	335±5%	380	517±6%	580

5.3.3 Simulations of the high-velocity gas gun impact test results

Many researchers, for example [165–167], have shown that there can be a reduction in the values of the interlaminar fracture energies for composite materials of up to about 20% when the strain rate is increased from quasi-static test conditions (i.e. about 10⁻³ to 10⁰ s⁻¹) to high-velocity impact velocities of about 55 m.s⁻¹, which in the present tests correspond to a strain-rate of about 300 s⁻¹. (It was possible to extract from the DIC experimental results the maximum strain experienced by the composite specimens and, from knowing the time-scale of the impact event, this information gives a strain rate of about 250 s⁻¹ for the CF/PEEK and about 350 s⁻¹ for the CF/Epoxy composite test specimens when struck by a rigid impactor at a velocity of about 55 m.s⁻¹.) As no fibre breakage was observed, or predicted, in the impacted composite specimens, only the interlaminar and matrix-ply fracture energies, for both Mode I and Mode II failure, were decreased in the present models of the high-velocity impact tests. Now, from the literature [165–167] on the effects of strain-rate on the properties of composite laminates, it is considered that for these high-velocity tests a reduction in the values of the interlaminar and matrix fracture energies of about 20% should be made to account for this strain-rate effect. Thus, two different sets of values for the values of the interlaminar and matrix-ply fracture energies were considered for use in the simulations, i.e. 100% and 80% of the quasi-static values of $G_{Ic|mt}$, $G_{Ic|mc}$, G_{Ic} and G_{IIc} given in Table 5-1.

In the simulations of the gas-gun impact tests an impact velocity of 54.4 m.s^{-1} was employed which gives an impact energy of 10.5 J. The simulations of the gas-gun tests were performed for both the CF/PEEK and the CF/Epoxy composite specimens. The simulation results were obtained from the numerical FE model with the two different levels of fracture energy values, as noted above. The predicted damage footprints are shown in Figure 5-9, together with the experimentally measured C-scan damage maps. The modelling results accurately predict that (a) the damage areas in the composites are always significantly higher from the high-velocity gas-gun tests than from the low-velocity drop-weight tests, at the same impact energy of 10.5 J, see Figures 5-5 and 5-9, and (b) the CF/Epoxy composite undergoes somewhat greater damage compared to the CF/PEEK composite. However, the results shown in Figure 5-9 reveal that the areas of the damage footprints predicted by the FE model using the quasi-static fracture energies are much smaller than those obtained from the experiments. Obviously, as expected, the models using the reduced values of 80% of the fracture energies give larger footprints of the damage area and the simulation results are now in somewhat better agreement with the experimental results. However, the simulations shown in Figure 5-9 reveal that, even when using 80% of the quasi-static fracture energies, the footprints of the predicted damage areas are always lower than the values measured experimentally from the C-scans tests. This is likely to arise from several factors. Firstly, in the high-velocity model simulations, only the interlaminar and matrix-ply fracture energies were reduced to 80% of their quasi-static values and the other material properties required for the model were kept at their quasi-static values. This was done since the fracture energies have been shown to decrease at relatively high strain-rates [165–167] and the fibre-dominated properties would certainly be expected to show relatively little change as the strain rate is increased. Secondly, without a detailed knowledge of the strain-rate effects for the other material properties, it was considered best to keep all the other material properties at their quasi-static values for consistency. Thirdly, in the model, the presence of intralaminar-matrix fracture damage is considered to lead to a reduction in stiffness of the elasticity matrix, i.e. the term C_d as defined by implementing Equations (5-10) to (5-14), which represents damage of the ply elements. Thus, intralaminar matrix cracks, as such, are not physically present in this ‘smeared-crack’ model but only their effects on the ply element stiffnesses. However, in the experiments, intralaminar matrix cracks could be diverted at the ply interfaces to then generate more interlaminar delaminations and this may aspect also account for the extra extent of delamination that was experimentally observed.

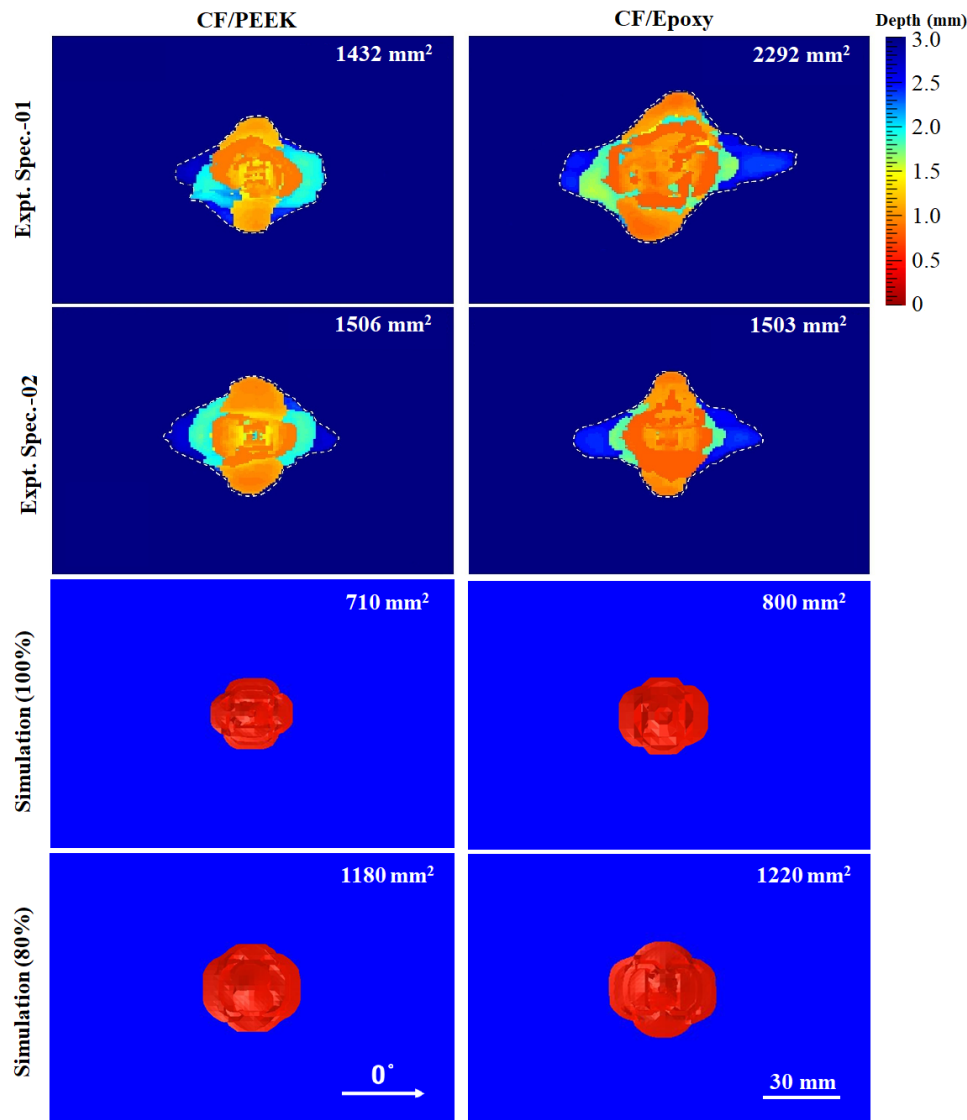
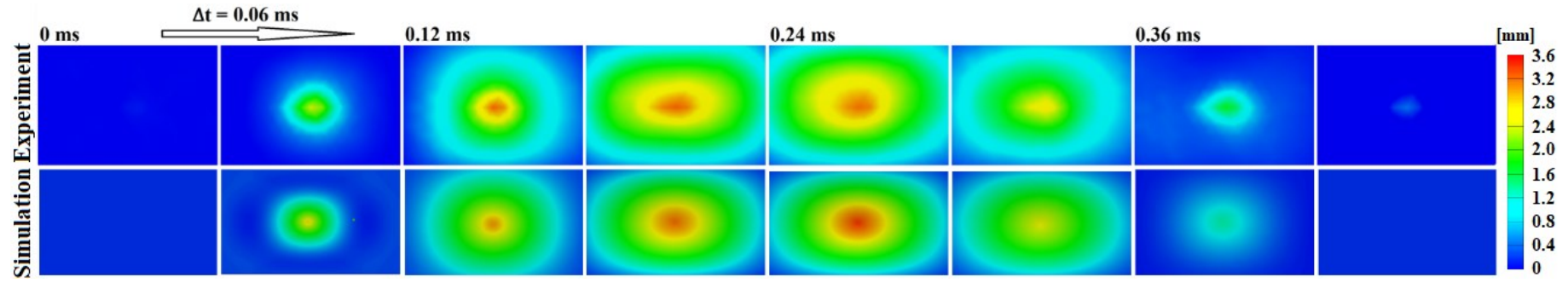


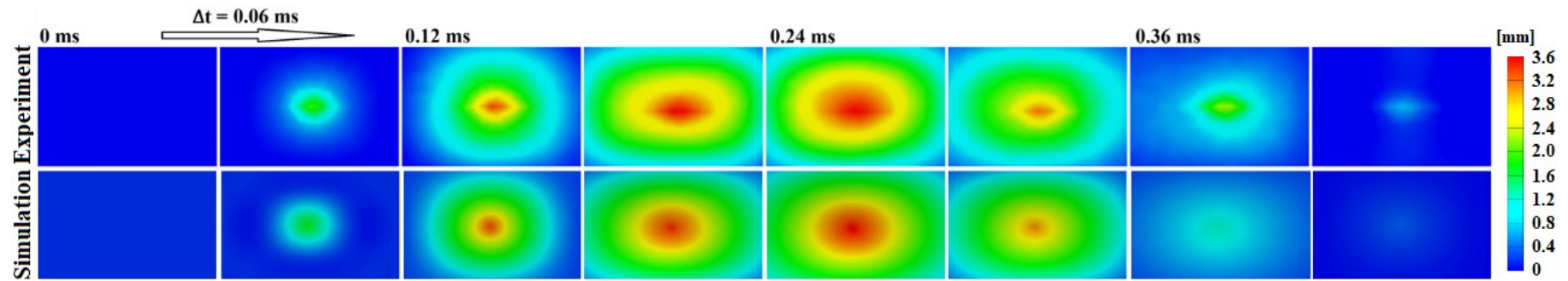
Figure 5-9. Experimental and simulated (red-coloured) footprints of the damage areas for the CF/PEEK and CF/Epoxy composite specimens from the high-velocity gas-gun test using an impact energy and velocity of 10.5 ± 0.3 J and 54.4 ± 1.0 m.s⁻¹, respectively. (See Figure 4-3 in Chapter 4 for the explanation of the experimental C-scan depth scale, etc. Results are shown for both 100% and 80% of the quasi-static interlaminar and matrix fracture energies.)

The FE model, using the 80% of the values of the quasi-static fracture energies, was also employed to predict the values of the out-of-plane displacement and the results are compared to the experimental values from the DIC tests, as shown in Figure 5-10. As may be seen, good agreement is obtained between the experimental and simulation results. Furthermore, the predicted maximum out-of-plane displacements of 3 mm for the CF/PEEK and 3.5 mm for the CF/Epoxy composite are in very good agreement with the maximum values

of 3 mm and 3.4 mm, respectively, that were experimentally measured, see Figures 4-5 and 4-6 in Chapter 4. It should be noted that the deflection of the specimen by the impactor is a structural response of the specimen to the loading by the impactor and, as such, it would not be expected to be strongly affected by the reduction of the quasi-static fracture energies. To confirm this, the out-of-plane displacements were also determined at 100% of the quasi-static fracture energies and it was indeed found that the displacements of the composite specimens are virtually identical to those predicted when using only 80% of the quasi-static fracture energies in the model. Thus, although the composite specimens are damaged by the delaminations that occurred, their overall structural responses are not dramatically modified, as is indeed confirmed by these modelling results.



(a)



(b)

Figure 5-10. The experimental and simulated out-of-plane displacement maps obtained from (a) the CF/PEEK and (b) the CF/Epoxy composite specimens when subjected to the high-velocity gas-gun test at an impact energy and velocity of 10.5 ± 0.3 J and 54.4 ± 1.0 m.s⁻¹, respectively.

For these high-velocity impact tests, the load versus time curves of the CF/PEEK and CF/Epoxy composites could not be experimentally measured, as discussed earlier. However, they may now be predicted using the FE model, as shown in Figure 5-11. Interestingly, the load versus time curves are not very sensitive to the effect of the percentage values of the fracture energies that are used, since the load versus time response is mostly a dynamic effect. The presence of the two major peaks in Figures 5-11(a) and (b) is due to the initial rapid acceleration of the specimen upon being struck by the impactor [168,169]. This rapid loading leads to a temporary loss of contact between the impactor and the composite specimen which is then followed by contact being re-established when there is a further phase of loading. The load versus time impulse (i.e. the area under load versus time curve) is about 0.6 N.s which will be responsible for the total change in momentum of the incoming impactor. The total change in momentum would therefore be expected to be 0.6 kg.m.s⁻¹ during the impact event. The incoming momentum of the impactor is 0.38 kg.m.s⁻¹ and so it would be expected that the impactor would rebound with momentum of 0.22 kg.m.s⁻¹. This implies that the rebound velocity of the impactor at the end of the impact event would be about 60% of the impact velocity, which was not measured but this is consistent with the significant rebound that was observed. Thus, the gas-gun, and also the drop-weight, impact events were reasonably elastic in nature with a high proportion of kinetic energy being returned to the impactor.

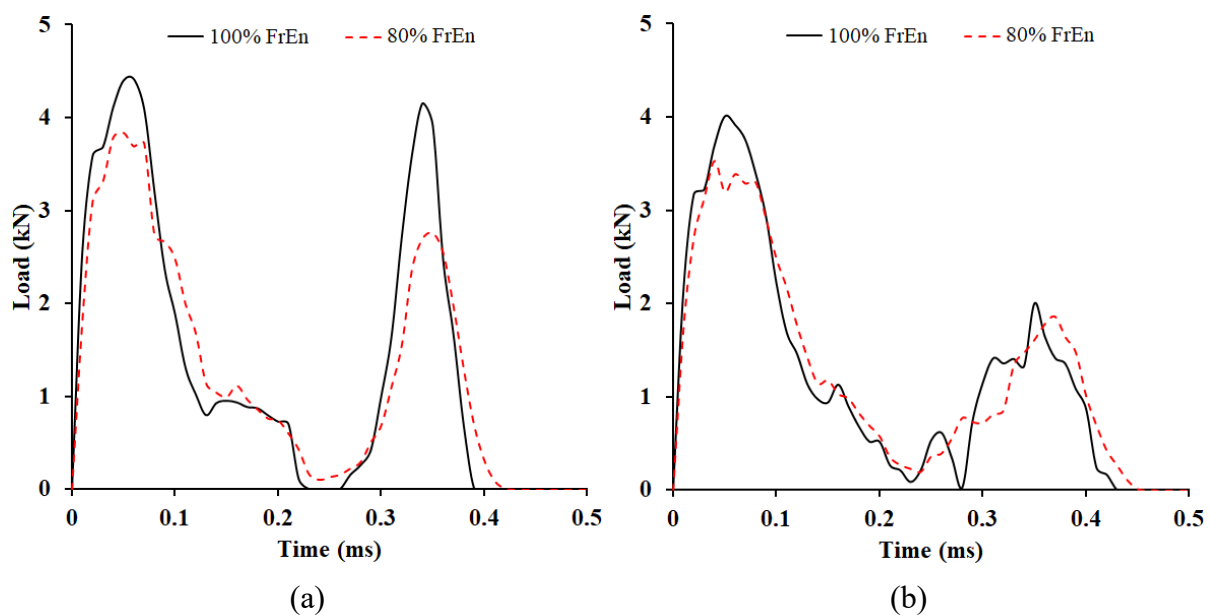


Figure 5-11. Load versus time curves as predicted from the FE model for the CF/PEEK and CF/Epoxy composite specimens subjected to a high-velocity gas-gun test at an impact energy and velocity 10.5 J and 54.4 m.s⁻¹, respectively. (Results are shown for using 100% and 80% of the quasi-static interlaminar and matrix fracture energies (FrEn)).

5.4 Summary

In the present study, an elastic, two-dimensional finite-element (FE) model has been developed to simulate the experimental results obtained from both the low-velocity and high-velocity impact tests in Chapter 4. The model, which is relatively computationally efficient, has been shown to simulate (a) the loading responses of the composites by the impact event and (b) the interlaminar and intralaminar damage induced.

For the low-velocity drop-weight tests, the experimental results demonstrated that, at the same impact energy, the CF/PEEK composite always possessed a superior impact performance compared with the CF/Epoxy. The results from the FE model for the load versus time curves from all the tests were in very good agreement with the experimental measurements and even captured the small amplitude, sinusoidal oscillations on the rising part of the load versus time curves which are indicative of mass-spring oscillations. Considering the FE modelling of the high-velocity tests, the simulations have accurately predicted (a) the shape and values of the out-of-plane displacements as a function of the time-scale of the impact event, (b) that the damage areas in the composites were always significantly higher from the high-velocity gas-gun tests than from the low-velocity drop-weight tests, at a comparable impact energy of 10.5 J, and (c) that the CF/Epoxy composite suffered somewhat greater damage compared to the CF/PEEK composite. However, the actual footprints of the damage areas predicted from the modelling studies for the high-velocity tests were always somewhat lower than the values measured experimentally from the C-scans tests, and reasons for this have been proposed. For these high-velocity impact tests, the load versus time curves of the CF/PEEK and CF/Epoxy composites could not be experimentally measured, nevertheless they were predicted using the FE model. The presence of two major peaks in the load versus time curves has been identified and ascribed to the initial rapid acceleration of the specimen upon being struck by the impactor. This rapid loading leads to a temporary loss of contact between the impactor and the composite specimen which is then followed by contact being re-established when there is a further phase of loading, before the impactor finally rebounds at the end of the impact event at about 60% of its impact velocity. Therefore, the numerical FE model has been developed which is relatively computationally efficient. The results from this model has been quantitatively validated against the impact response of the composites and captures the essential aspects of their impact behaviour and can be used (a) to simulate aspects of the high-velocity tests that cannot be directly measured and (b) to optimise the impact performance of such materials in industrial applications.

Chapter 6. Experimental study on woven-fabric thermoplastic and thermoset laminates under high-velocity soft and hard impact loading

6.1 Introduction

The present chapter investigates the impact performance of woven-fabric carbon fibre composites based upon both thermoplastic- and thermoset-matrix polymers under high-velocity impact loading by conducting gas gun experiments at impact velocities of up to 100 m.s⁻¹. The carbon fibre reinforced polymers (CFRPs) are impacted using soft (i.e. gelatine) and hard (i.e. aluminium-alloy) projectiles to simulate either a soft bird-strike or a hard foreign-body impact (e.g. runway debris), respectively, on composites employed in civil aircraft. The out-of-plane displacement of the impacted composite specimen is obtained by means of a three-dimensional Digital Image Correlation (DIC) system for the soft projectile impact on the composites and the extent of damage is assessed both visually and by using portable C-scan equipment. The perforation resistance and energy absorbing capability of the composites are also studied by performing high-velocity impact experiments using the hard-projectile and the resulting extent and type of damage are identified.

Note: The work presented in this chapter was previously published in the Applied Composite Materials as “The impact performance of woven-fabric thermoplastic and thermoset composites subjected to high-velocity soft- and hard-impact loading”.

6.2 Test specification and methods

6.2.1 Test specimens and projectiles

The woven-fabric CF/PEEK and CF/Epoxy specimens employed in this study were detailed in Chapter 3.2.2. The 3D DIC system was employed in the soft-projectile impact experiments to record the deformation of the CF/PEEK and CF/Epoxy specimens. Therefore, the non-impacted surface (i.e. the rear surface of the sample) of the CFRP specimens were prepared according to the instructions of the manufacturer of the DIC system [143] which is detailed in Chapter 3.4.3. Prior to undertaking the experiments, all the specimens were kept under a temperature of 22 ± 0.5 °C and a humidity of 50 ± 2 % for 24 hours.

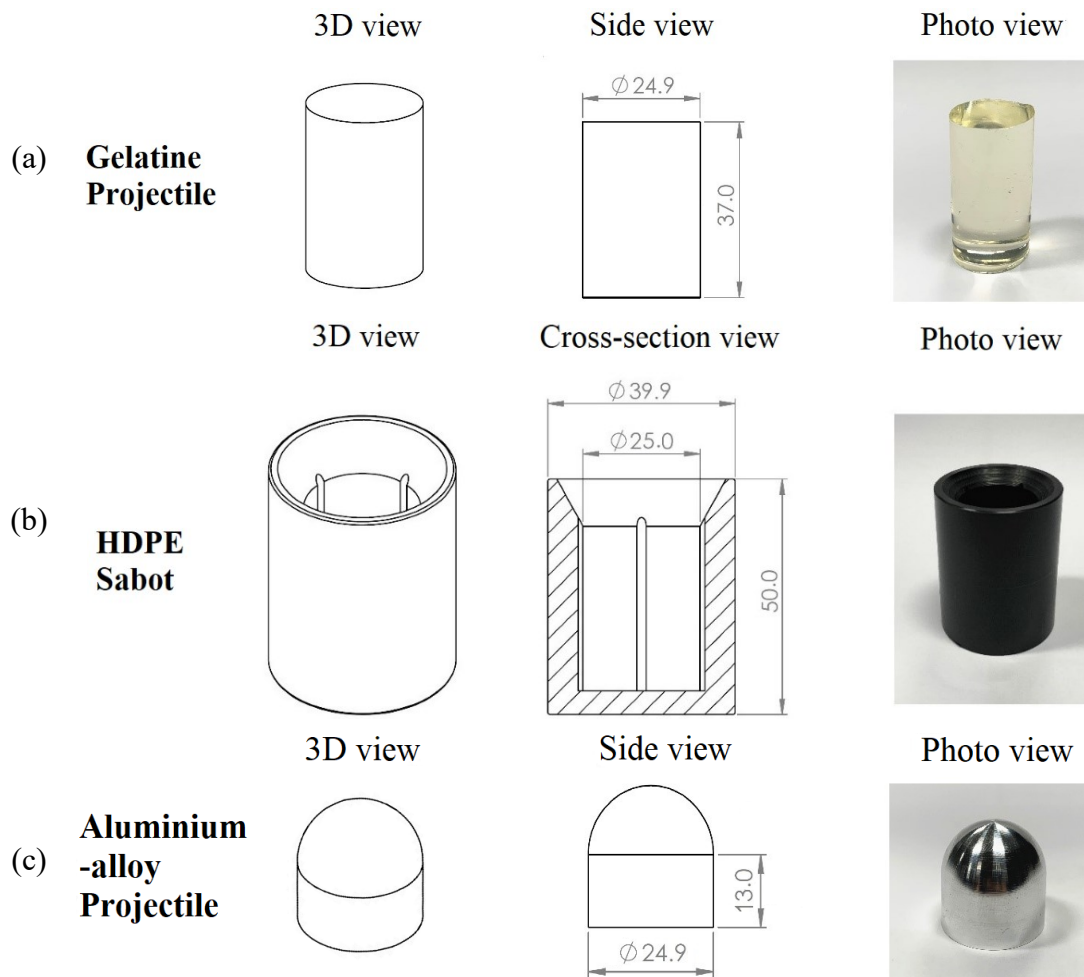


Figure 6-1. Projectiles and sabot for soft- and hard-impact tests: (a) flat-fronted soft gelatine projectile and (b) HDPE sabot and (c) hemi-spherical head of the hard aluminium-alloy projectile. (Dimensions are in mm).

In order to simulate the impact of a bird-strike and runway debris on the composites, gelatine projectiles (to represent a relatively soft projectile, e.g. a bird-strike) and aluminium-alloy projectiles (to represent a hard projectile, e.g. runway debris) were used in the gas-gun experiments, as shown in Figure 6-1. New projectiles were employed for each impact test.

Now gelatine has been identified as a good substitute material to simulate birds to overcome the drawbacks of using real birds for the impact tests, as discussed above. The gelatine projectile generates a similar level of initial shock pressure as that experienced in a typical bird-strike, which is important in determining the extent and type of local damage that results in the impacted composite [170]. The gelatine projectiles employed were based on the bird model developed by Wilbeck and Rand [170].

A well-defined process, which is relatively simple and controllable, has been developed for preparing the soft-gelatine projectiles to a uniform standard. The ingredients used to prepare these projectiles were gelatine powder and distilled water. The gelatine powder was supplied by Honeywell Specialty, Germany. The detailed preparation procedure of the gelatine projectiles is introduced in Table 6-1. These 3 wt.% gelatine projectiles had a nominal mass of 20 g and possessed a Shore hardness A of 0 to 5 and this enabled them to withstand the inertial loads during being launched from the gas gun. It is important to note that the shape of the gelatine projectile from being fired to impacting the target remained essentially unchanged, i.e. no stretching or compressing of the projectile occurred. To achieve this a sabot, 40 mm in diameter, was used to provide the projectile acceleration that was required and to reduce the friction between the soft-projectile and the barrel of the gas-gun.

Table 6-1. Detailed preparation procedure of the gelatine projectiles.

Steps	Operations
I	Raise the temperature of the distilled water to 80°C and maintain this temperature using a water bath with a thermocouple to monitor the temperature of the distilled water.
II	Mix the gelatine powder with distilled water at a specific mass ratio of 3:100, and then stir the mixture, using a magnetic stirrer at a stirring rate of 5 revolutions per second, until the gelatine powder is completely dissolved.
III	Transfer the solution to a beaker and let the solution cool down to room temperature.
IV	Transfer the solution at room temperature to a poly(tetrafluoroethylene) (PTFE) cylindrical mould, which has a paraffin-oil coated onto its surface to prevent leakage and facilitate the subsequent removal of the solid gelatine projectile.
V	Seal the mould with cling film placed over the top of the mould to prevent dehydration.
VI	Place the sealed mould into an environmental chamber where the temperature is kept between 4 to 6°C for at least 12 hours.
VII	Carefully push the solid gelatine cylinder out from the mould and use a digitally-controlled disc saw to cut the gelatine cylinders to the required length for the gelatine projectiles.

The hard aluminium-alloy projectiles were machined from 7075-T6 aluminium-alloy. These gave a hard-impact event, for example as might be experienced by impact from runway debris. The aluminium-alloy projectiles had a nominal mass of 28 g and a nominal diameter of 24.9 mm, which is slightly smaller than the diameter of 25 mm of the barrel of the gas gun. This was necessary in order for the aluminium-alloy projectiles to achieve a reproducible maximum velocity.

6.2.2 Gas gun testing

Two different set-ups of the gas gun were used to assess the dynamic deformation and ballistic response of the composites dependent upon whether the high-velocity impact was via the relatively soft- or the hard-projectile. Both set-ups fixed the specimens between two steel

supporting plates, both with a 70×70 mm opening, and twelve M8 bolts were used to fasten the steel plates together. A schematic drawing of the steel plates and the specimen is shown in Figure 6-2.

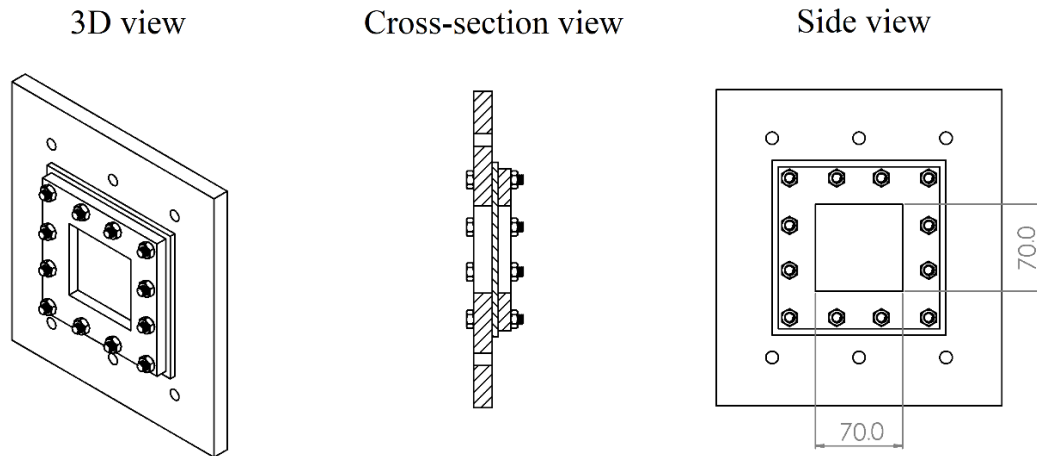


Figure 6-2. Schematic of the support for the composite test specimen.
(Dimensions are in mm.)

For ‘Set-up I’ as mentioned in the section 3.4.3, for the impact by the soft-gelatine projectile, it employed a pair of high-speed cameras, i.e. ‘Phantom Miro M/R/LC310’ cameras supplied by Vision Research Phantom, USA, to track the deformation history of the rear surface of the target specimen and record at their maximum rate of 39,000 frames per second. In order to monitor the full-field deformation of the specimens, the image size was selected as 256 x 256 pixels.

For ‘Set-up II’, the aim was to study the ballistic response of the CF/PEEK and CF/Epoxy composites subjected to a high-velocity impact by the relatively hard aluminium-alloy projectile. As shown in Figure 6-3, two high-speed cameras were located at the side of the target chamber and recorded at their maximum rate of 39,000 frames per second. The kinetic energy absorbed during the impact could be determined via measuring the initial and residual velocities of the projectile. Two pairs of infrared sensors were also used to validate the initial velocity of the projectile as measured using the high-speed cameras.

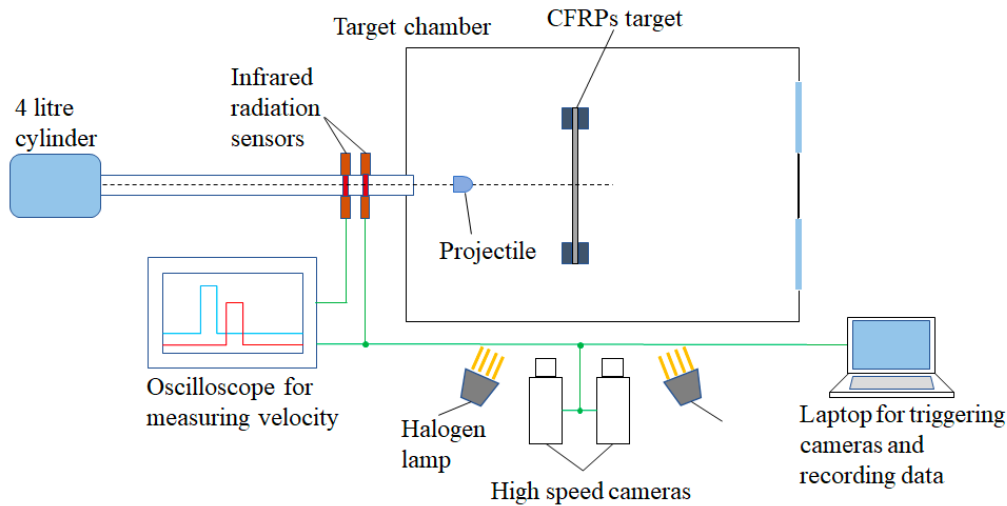


Figure 6-3. Set-up II for the high-velocity hard-projectile impact: (a) photograph and (b) schematic.

6.2.3 Post-test inspection

The portable C-scan system were used to examine all the impacted specimens after the soft-projectile impact experiments. A configuration designed for the 2-mm woven specimens was operated in the C-scan system. Visual inspections, for both soft and hard impact experiments, were performed on the rear surfaces of the specimens where the impact damage is often found. To investigate the failure mechanisms of the specimens, the cross-section of the mid-plane of the composites was created via a ‘Brillant 220’ cut-off machine provided by QATM, Germany. This laser alignment aided machine, as shown in Figure 6-4, had a precision of 1 μm and equipped with a 0.5 mm thick high concentration diamond cut-off wheel. The wheel forwarded 0.2 mm per second and set at 2000 rpm to minimum additional damage to the specimens.

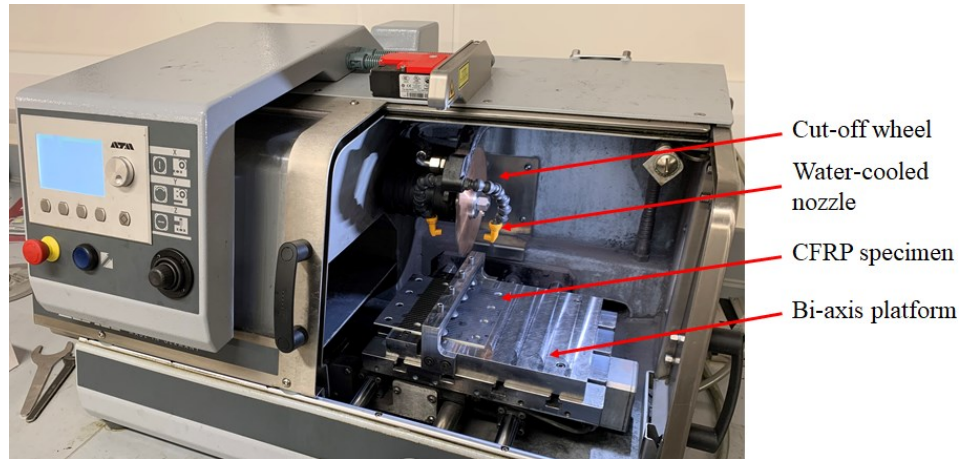


Figure 6-4. The cut-off machine equipped with high concentration diamond saw.

6.3 Experimental results

6.3.1 The soft projectile impact testing

6.3.1.1 Impact response

The impact responses of the CFRP specimens were investigated over a range of impact velocities from 42 to 100 m.s⁻¹ using the soft-gelatine projectiles, and were therefore studied at different impact energy levels. Figures 6-5 and 6-6 show results for a soft-gelatine impact at 100 m.s⁻¹ on the CF/PEEK and the CF/Epoxy specimens, respectively. Figure 6-5(a) shows the out-of-plane displacement maps obtained from the CF/PEEK specimen under soft impact loading, where both the loading and the unloading phases are included. The deformation profile across the middle section of the specimen is shown in Figure 6-5(b). This CF/PEEK specimen, which was subjected to an impact velocity of 100 m.s⁻¹, exhibited a maximum out-of-plane displacement of approximately 3.7 mm.

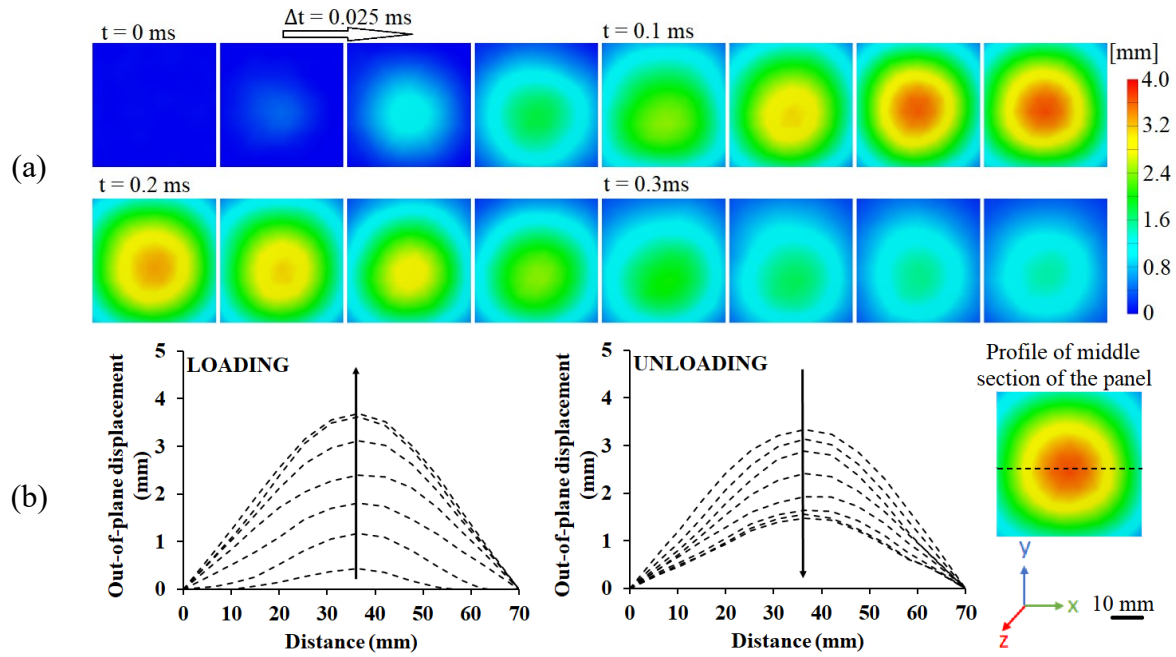


Figure 6-5. Experimental results for the CF/PEEK panel impacted using a soft-gelatine projectile with a velocity of 100 m.s^{-1} showing the (a) out-of-plane displacement maps and (b) out-of-plane displacement profile during the loading and unloading phases. (Inset picture, on right, shows a horizontal dash line where the profile section is taken.)

The out-of-plane displacement maps and deformation profiles of the CF/Epoxy specimen, under a soft impact loading at an impact velocity of 100 m.s^{-1} , are shown in Figures 6-6(a) and 6-6(b), respectively. In comparison to the DIC results of the CF/PEEK specimen, a 10% higher maximum out-of-plane displacement was determined for the CF/Epoxy specimen. In addition, a slower rebound sequence for the CF/Epoxy composite was found from the deformation profiles for the unloading phase. These differences can be explained from the photographs of the rear surface and cross-section of the CF/Epoxy specimen shown below, where significant impact damage was found in the central area of the specimen. The presence of damage, which was observed to arise from matrix cracking, delamination and fibre breakage, as discussed below, leads to a significant loss of the stiffness of the specimen. This, in turn, leads a higher maximum out-of-plane displacement and results in a slower rebound of the specimen in the unloading phase for the CF/Epoxy composite, compared to the CF/PEEK composite.

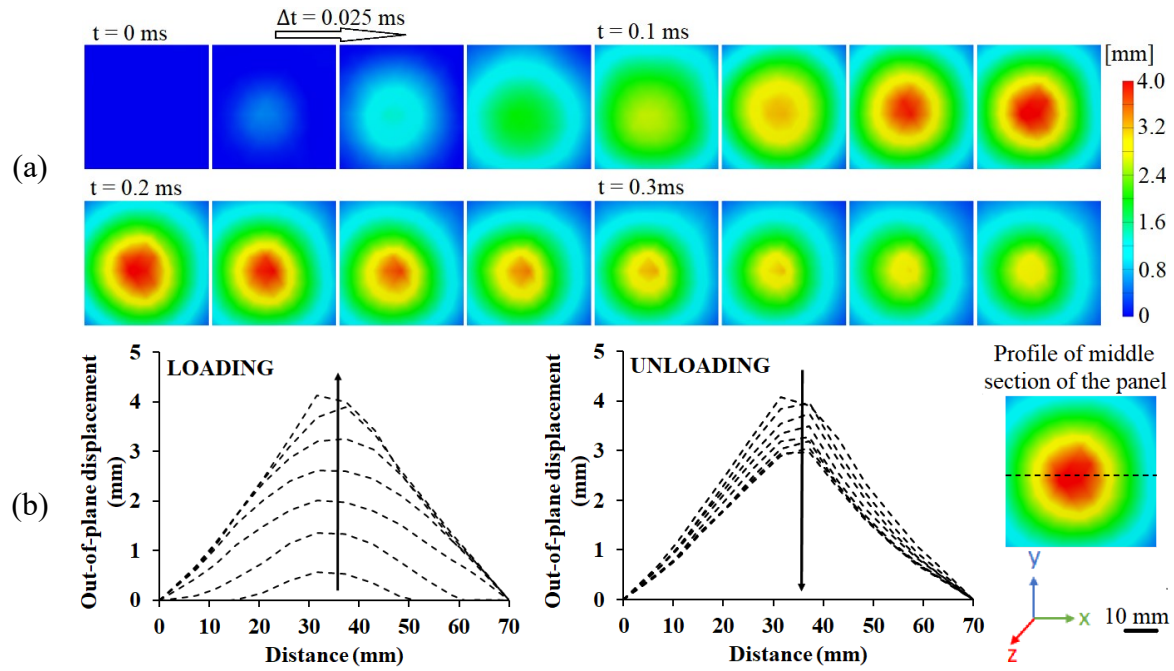


Figure 6-6. Experimental results for the CF/Epoxy panel impacted using a soft-gelatine projectile with a velocity of 100 m.s^{-1} showing the (a) out-of-plane displacement maps and (b) out-of-plane displacement profile during the loading and unloading phases. (Inset picture, on right, shows a horizontal dash line where the profile section is taken.)

The 3D DIC results of the maximum out-of-plane displacement of the specimens under four different impact velocities for both types of composites are summarised in Table 6-2. In addition, the maximum out-of-plane displacement is plotted against the impact energy in Figure 6-7. As may be seen, the values of the maximum out-of-plane displacements increase with increasing impact velocity. Also, at a similar impact velocity (i.e. at a similar impact energy level) the CF/Epoxy composite undergoes a greater maximum out-of-plane displacement, compared to the CF/PEEK composite, and this difference is most striking at the highest impact velocity that was employed of 100 m.s^{-1} .

Table 6-2. Results of the high-velocity soft-gelatine projectile impact tests.

Specimen Code	Impact Velocity* (m.s ⁻¹)	Projectile Mass (g)	Impact Energy* (J)	Maximum Out-of-plane Displacement (mm)
CF/PEEK_1	42.4	19.9	17.9	1.6
CF/PEEK_2	58.1	19.9	33.6	2.2
CF/PEEK_3	78.1	20.0	61.0	2.9
CF/PEEK_4	100.0	19.8	99.0	3.7
CF/Epoxy_1	45.9	19.8	20.9	1.8
CF/Epoxy_2	57.5	20.0	33.1	2.3
CF/Epoxy_3	74.6	19.7	54.8	3.1
CF/Epoxy_4	100.0	19.9	99.5	4.2

*Note: Experimental error: $\pm 2.5\%$ in impact velocity and $\pm 5\%$ in impact energy.

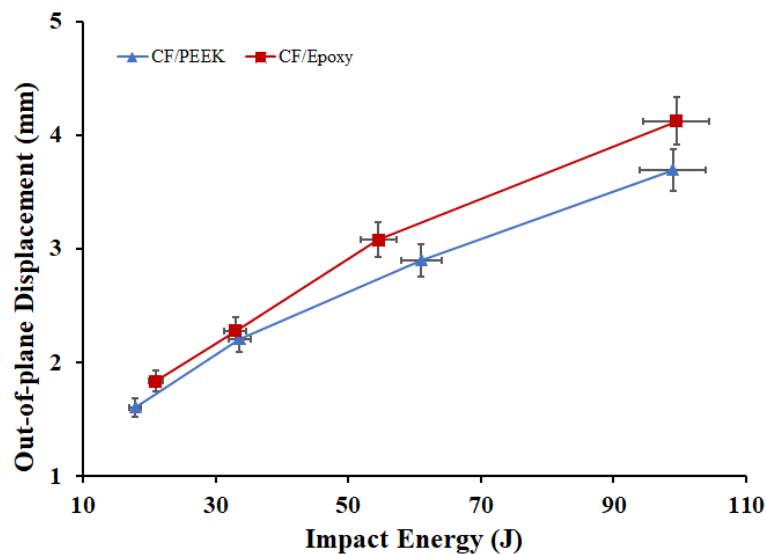


Figure 6-7. Maximum out-of-plane displacement versus the impact energy for the CF/PEEK and CF/Epoxy composites upon being impacted using soft-gelatine projectiles.

6.3.1.2 Visual inspections

Visual inspections were performed on the rear surfaces of the CF/PEEK and the CF/Epoxy specimens, as shown in Figure 6-8(a) and 6-8(b), respectively, for the different impact velocities used for the soft-gelatine projectile. Firstly, no visible damage was observed in any of the CF/PEEK specimens. Secondly, however, matrix cracking and fibre breakage

were observed, at the highest impact velocity of 100 m.s^{-1} on the rear surface of the impacted CF/Epoxy composite. The cross-sections of the CF/PEEK and CF/Epoxy specimens impacted at 100 m.s^{-1} are shown in Figures 6-9(a) and 6-9(b), respectively. As may be seen, again no visible damage was observed in the CF/PEEK composites, whereas different types of damage including matrix crack, delamination and fibre breakage were particularly observable in the CF/Epoxy composite impacted at the highest test velocity of 100 m.s^{-1} .

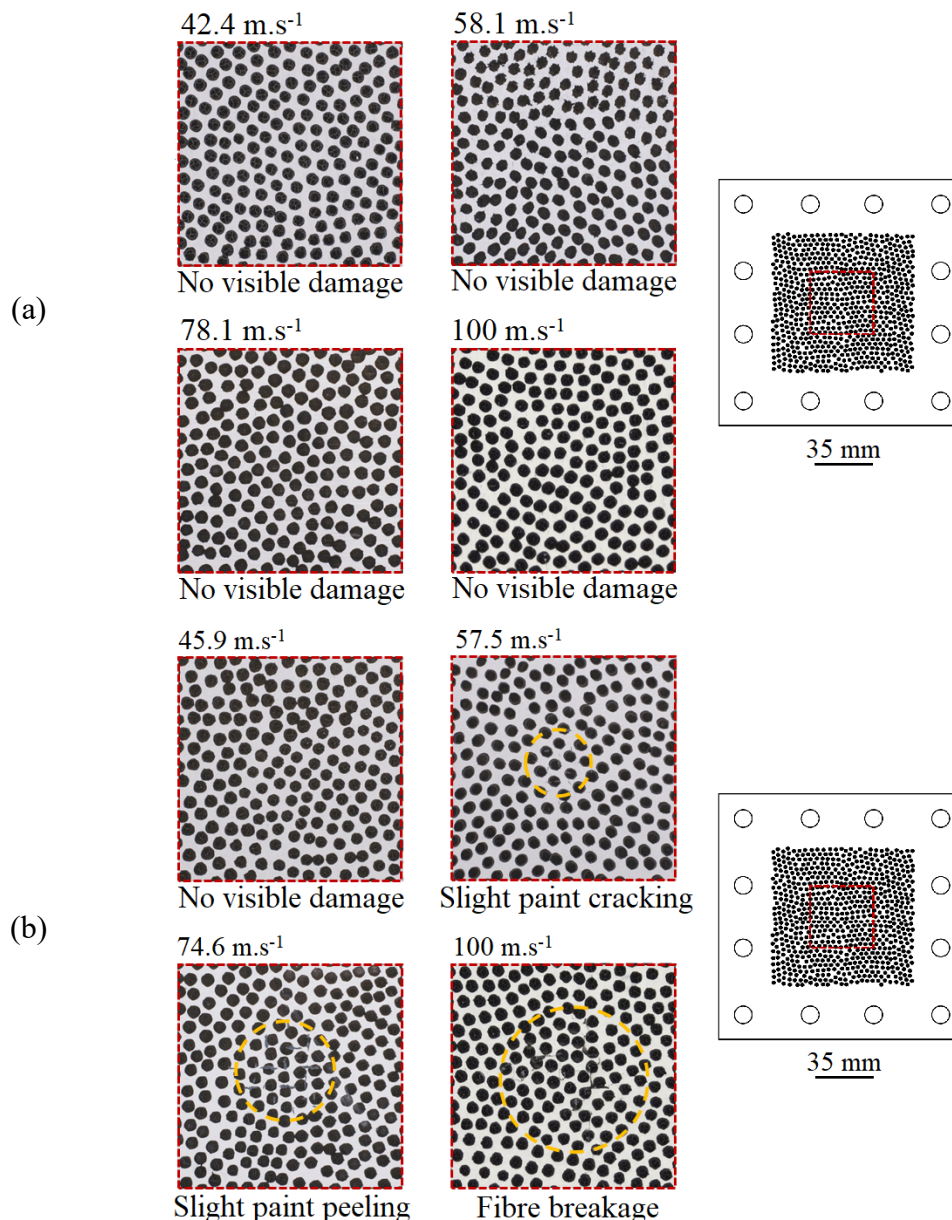


Figure 6-8. Visual inspection of the rear surface of the (a) CF/PEEK specimens and the (b) CF/Epoxy specimens subjected to different impact velocities by the soft-gelatine projectile. (The area represented by the red square corresponds to the area shown on the schematic of the full specimen at the far right-hand side. The yellow circles indicate the extent of visible damage.)

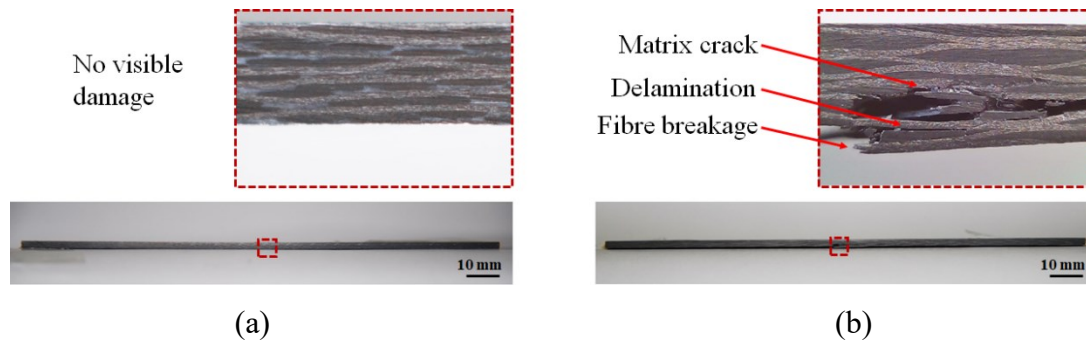


Figure 6-9. Photographs of the cross-section of the mid-plane of the composite: (a) CF/PEEK and (b) CF/Epoxy specimens impacted using a soft-gelatine projectile with an impact velocity of 100 m.s^{-1} .

6.3.1.2 Ultrasonic C-scan maps

As noted above from Figures 6-8 and 6-9, a visual inspection of the impacted specimens showed that significant damage was only observed in the CF/Epoxy composite tested at the highest impact velocity, whilst no damage was ever observed in the CF/PEEK composite at any of the impact velocities employed. The relevant C-scan images in Figure 6-10 show very clearly an area of delamination, in both the relatively low magnification and the higher magnification C-scan maps, for the CF/Epoxy composite impacted at the highest velocity of 100 m.s^{-1} . (These maps show delamination as a function of the depth through the thickness of the specimen, where the red colour is representative of the front (impacted) surface and blue colour is representative of the rear (non-impacted) surface of the specimen.) Thus, the ultrasonic C-scan maps reveal sub-surface damage only for the CF/Epoxy composite, and then only significant damage when impacted at the highest velocity of 100 m.s^{-1} . These findings are clearly in excellent agreement with the visual inspections discussed above. Therefore, it may again be concluded that the CF/PEEK laminates have a superior resistance to delamination when subjected to a soft impact at 100 m.s^{-1} .

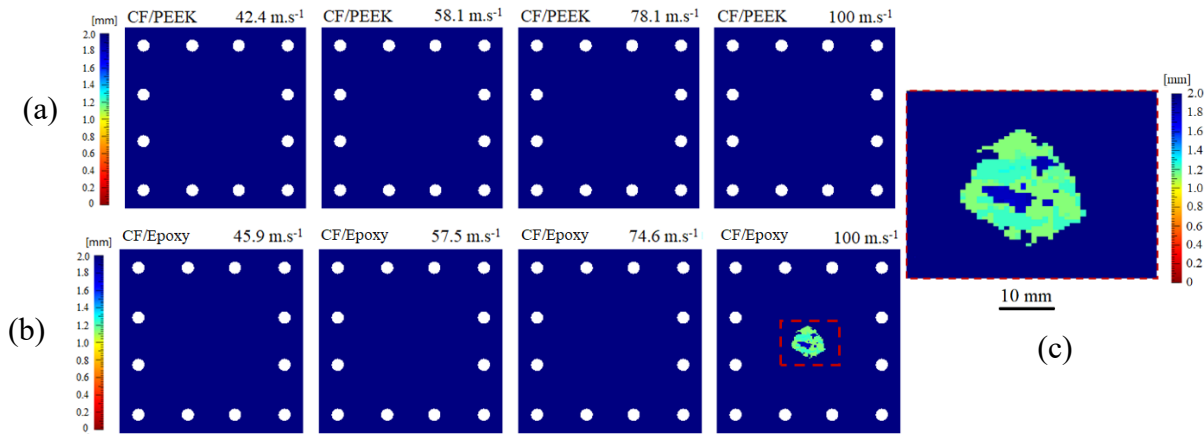


Figure 6-10. Ultrasonic C-scan maps of the (a) CF/PEEK specimens and (b) CF/Epoxy specimens subjected to impact velocities using the soft-gelatine projectile and (c) magnified C-scan image of the CF/Epoxy specimen impacted at the highest velocity of 100 m.s⁻¹.

6.3.2 The hard projectile impact testing

The perforation resistance and energy absorption of both the CF/PEEK and CF/Epoxy specimens were studied when these composites were subjected to an impact by the relatively hard aluminium-alloy projectile over a range of impact velocities from 30 to 77 m.s⁻¹. The kinetic energy absorption, E_a , of the test specimen can be determined from:

$$E_a = \frac{1}{2}mv_i^2 - \frac{1}{2}mv_r^2 \quad (6-1)$$

where m is the mass of projectile, and v_i and v_r are the initial impact velocity and residual velocity (i.e. the exit or rebound velocity of the projectile), respectively. Three different impact velocities below and above the ballistic limit, which is the lowest velocity required for the projectile to penetrate consistently through the test specimen, were selected to investigate the relationship between the energy absorption and the damage suffered by the two types of composites.

The photographs obtained from the rear surface of the tested specimens are shown in Figure 6-11. As may be seen, minor surface damage, with similar areal extents of the damage, can be observed from both types of specimens impacted at about 30 m.s⁻¹ (i.e. with an associated impact energy of 12-13 J). For an impact velocity above about 50 m.s⁻¹ (i.e. with an

associated impact energy above about 40 J) the CF/PEEK specimens perforated and exhibited damage areas which were slightly larger in size than the cross-sectional area of the projectile. In contrast, for impact energies above about 40 J, the damage area was about three times larger in the perforated CF/Epoxy specimens, compared to the CF/PEEK specimens. For both types of composite there was no significant further increase in the extent of the damage area when the impact velocity was raised to about 70 m.s⁻¹.

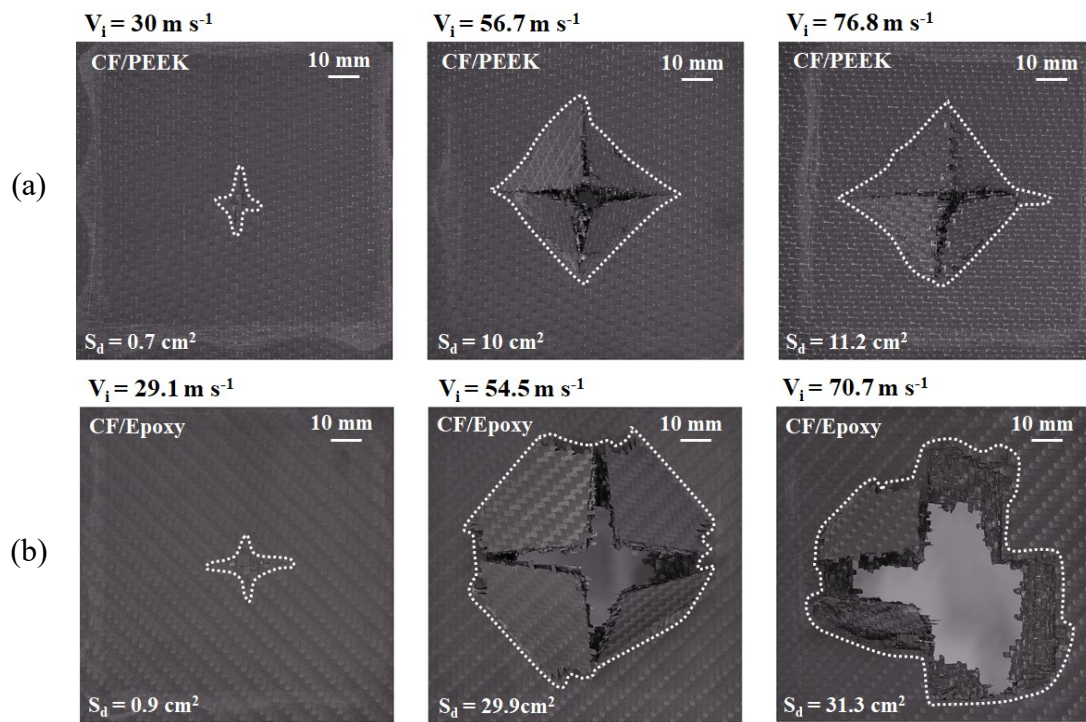


Figure 6-11. Photographs of the rear surface of the (a) CF/PEEK and (b) CF/Epoxy composites impacted using the hard aluminium-alloy projectile at various initial impact velocities. (The visible damage area, S_d , outlined in by the broken white line, was determined by measuring the number of pixels in the damage area compared to the total number in the specimen of known area.)

The experimental results from using the hard aluminium-alloy projectile are summarised in Table 6-3. (It should be noted that in Table 6-3 a positive value in the residual velocity indicates that full perforation has occurred, whereas a negative value means that the projectile rebounded from the front surface of the specimen. Also, the damage area, S_d , of each specimen was determined, via a data processing programme, by measuring the number of pixels in the damage area compared to the total number in the specimen of known area.) The kinetic energy absorption and the damage area suffered by the composites are also plotted

against the impact energy in Figure 6-12. In addition, it should be noted that no visible deformation was observed in the aluminium-alloy projectiles post-impact.

Table 6-3. Results of the high-velocity hard aluminium-alloy projectile impact tests.

Sample Code	Initial Velocity* (v_i) (m.s^{-1})	Projectile Mass (g)	Impact Energy* (J)	Residual Velocity* (v_r) (m.s^{-1})	Energy Absorption* (E_a) (J)	Damage Area* (S_d) (cm^2)
CF/PEEK_5	30.0	28.4	12.8	-14.8	9.7	0.7
CF/PEEK_6	56.7	28.2	45.3	23.2	37.7	10.0
CF/PEEK_7	76.8	28.1	82.9	50.0	47.8	11.2
CF/Epoxy_5	29.1	28.1	11.9	-15.5	8.6	0.9
CF/Epoxy_6	54.5	28.2	41.9	24.8	33.2	29.9
CF/Epoxy_7	70.7	28.3	70.7	47.8	38.4	31.3

*Note: Experimental error: $\pm 2\%$ in velocities, $\pm 4\%$ in impact energy and $\pm 5\%$ in damage area.

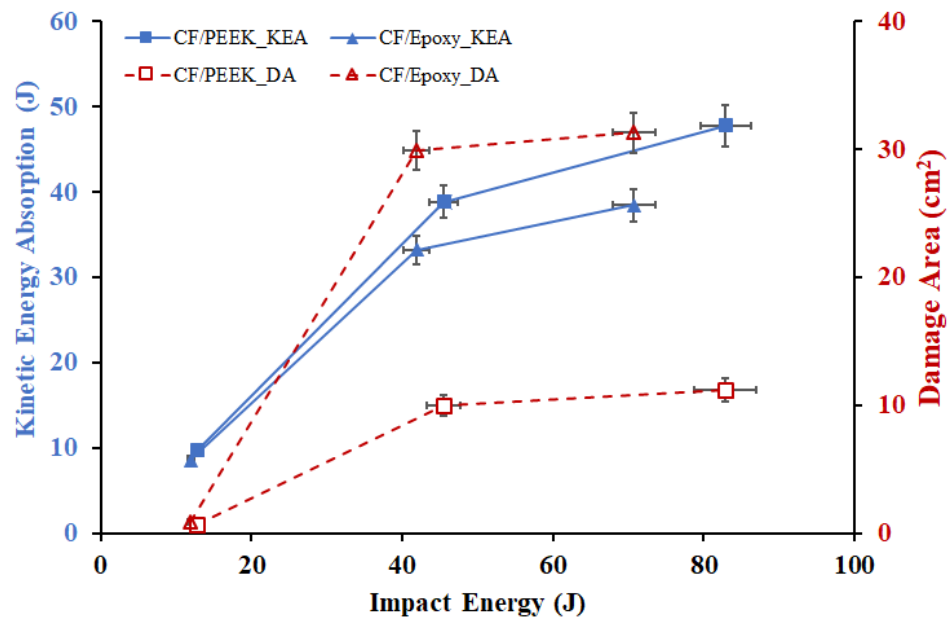


Figure 6-12. Comparison of the kinetic energy absorption (KEA) and damage area (DA) versus the impact energy for the CF/PEEK and CF/Epoxy composites impacted using the hard aluminium-alloy projectile at different velocities. (Note: perforation of the composites occurred at, and beyond, an energy level of about 40 J.)

In Figure 6-12 it can be seen that, as the impact energy is increased, then both the kinetic energy absorption and the damage area increase. Once perforation occurs at an impact energy of about 40 J, and beyond, there is, however, a far smaller increase on the energy absorption and damage area as the impact velocity is further increased. This phenomenon has previously been observed in the impact behaviour and failure mechanisms of composites when they are perforated by an impacting projectile [66]. Now, the damage is considered to arise from compressive stress waves which are generated and transmitted through the volume of the composite material after impact. These waves are reflected when reaching the rear layer of the composite panel to form tensile waves which now act upon the neighbouring ply interfaces and can initiate delamination, and so lead to extensive interlaminar failure. Also, shear failures occur when the projectile starts to penetrate the uppermost layers of the composites. Subsequently, when the projectile reaches the delaminated layers, these layers can slide over each other and ultimately fail completely due to tensile failure. It is for these reasons that a larger damage area is found on the rear surface. Now, many researchers [66,67,171] have indeed reported the phenomenon that the damage area in the composite increases relatively rapidly with an initial increase in the impact velocity up to the ballistic limit velocity, and then the damage area remains relatively constant in extent beyond this velocity. As was indeed observed in the present high-velocity tests using the hard aluminium-alloy projectile.

6.4 Summary

In the present chapter, the impact performance of woven-fabric carbon-fibre composites based upon both thermoplastic- and thermoset-matrix polymers under high-velocity impact loading have been studied by conducting gas-gun experiments at impact velocities of up to 100 m.s^{-1} . The thermoplastic matrix was poly(ether-ether ketone) (PEEK) and the thermoset matrix was a non-toughened Epoxy. The carbon-fibre reinforced-polymers (CFRPs) were impacted using a soft (i.e. gelatine) or a hard (i.e. aluminium-alloy) projectile to simulate either a soft bird-strike or a hard foreign-body impact (e.g. runway debris), respectively, on composites employed in civil aircraft.

For the high-velocity soft gelatine impact tests, the 3D Digital Image Correlation (DIC) technique was successfully employed to determine the out-of-plane displacement of the woven-fabric composites specimens at different impact velocities, and hence different impact energy levels. The woven-fabric CF/PEEK composites, with the same lay-up and thickness and

approximately the same density as the CF/Epoxy composites, exhibited a reduced maximum out-of-plane deformation compared to the CF/Epoxy composites. Indeed, fibre and matrix damage on the rear of the specimen, and sub-surface damage, indicated that significant damage was only observed in the CF/Epoxy composite, and only at the highest velocity of 100 m.s^{-1} .

Delamination damage was indeed observed in the C-scan map of the CF/Epoxy composite impacted at the highest velocity of 100 m.s^{-1} . By contrast, no delamination area was found in any C-scan maps of the CF/PEEK composite specimens. It was therefore concluded that the CF/PEEK composites have a superior damage resistance compared to the equivalent CF/Epoxy composites when subjected to a high-velocity impact by the soft gelatine projectile, which simulates a bird-strike.

For the high-velocity impact tests using a hard aluminium-alloy projectile, with a hemispherical head, to simulate runway debris, superior impact resistance was observed in the CF/PEEK composites compared to the CF/Epoxy composites. The efficiency of kinetic energy absorption decreases with the increasing impact velocity from about 30 to 75 m.s^{-1} . The CF/PEEK composite was more efficient in absorbing kinetic energy for these velocities and the damage area was less extensive than for the CF/Epoxy composite. As to be expected, at similar impact energy levels, composite specimens are more vulnerable to the harder aluminium-alloy projectile than the soft-gelatine projectile. This is attributed to the higher contact pressures, and their localisation, associated with the aluminium-alloy projectile.

It is important to note that the CF/PEEK composite showed improved impact performance both when struck by hard aluminium-alloy projectiles (to simulate runway debris) and soft gelatine projectiles (to simulate bird-strike) compared to the CF/Epoxy composite.

Chapter 7. Experimental study and numerical modelling on woven-fabric thermoplastic and thermoset laminates under high-velocity impact loading

7.1 Introduction

This chapter presents experimental and numerical studies on the behaviour of woven-fabric composite laminates subject to impact loading by soft projectiles to represent the impact of a small bird or hail-stone. In this research, gas-gun experiments are performed to study woven carbon-fibre reinforced poly (ether-ether ketone) (CF/PEEK) composites subjected to an impact by soft-gelatine projectiles. In addition, woven carbon-fibre reinforced epoxy (CF/Epoxy) composite specimens are also evaluated using gelatine projectiles to investigate the effect of the matrix system on the impact response of the composites. A high-speed camera is employed to capture the deformation of the projectiles and a three-dimensional (3D) Digital Image Correlation (DIC) system is used to record the deformation of the impacted composite specimens. A Finite Element (FE) model is developed to simulate the impact by a soft projectile on the composite specimens. Good agreement is shown between the predictions from using the FE model and the experimental results.

Note: The work presented in this chapter was previously published in the Engineering Failure Analysis as “The behaviour of fibre-reinforced composites subjected to a soft impact-loading: An experimental and numerical study”.

7.2 Numerical modelling methods

7.2.1 Modelling the response of the projectile

The SPH approach [110] was employed to model the behaviour of the gelatine projectile within the ‘Abaqus/Explicit 2018’ code, as discussed later. For the SPH method to capture the response of the soft-gelatine projectile upon impact of the composite, a constitutive law is required with suitable material properties employed for the gelatine projectile. The model used was originally developed for ballistic impact in metals and describes an isotropic elastic-plastic material subjected to relatively low pressures with an equation of state (EOS) describing the hydrodynamic pressure versus volume behaviour at high pressures. The linear Mie-Grüneisen EOS was employed to define the coupled equations for pressure and internal energy [148]. The most common form for this EOS is given by:

$$p - p_H = \Gamma \rho (E_m - E_H) \quad (7-1)$$

where p is the pressure which is defined as positive in compression. The Hugoniot pressure, p_H , is a function only of the density and can be ascertained from fitting to the experimental data. The parameters E_m and E_H are the internal energy per unit mass and the specific energy per unit mass (i.e. the Hugoniot energy), respectively. The parameter, ρ , is the current density of the gelatine projectile. The parameter, Γ , is the Grüneisen ratio and is defined by:

$$\Gamma = \Gamma_0 \frac{\rho_0}{\rho} \quad (7-2)$$

where Γ_0 is a material constant and ρ_0 is the reference density of the gelatine projectile. The specific energy per unit mass, E_H , is related to the Hugoniot pressure by:

$$E_H = \frac{p_H \eta}{2 \rho_0} \quad (7-3)$$

where $\eta = 1 - \rho_0/\rho$ and η is the nominal volumetric compressive strain. The elimination of Γ and E_H from the above equations yields:

$$p = p_H \left(1 - \frac{\Gamma_0 \eta}{2} \right) + \Gamma_0 \rho_0 E_m \quad (7-4)$$

In the above equation, the pressure, p , is a function of the Hugoniot pressure, p_H , and the nominal volumetric compressive strain, η . Once the relationship between p_H and η is defined, the pressure, p , can be expressed as a single-variable function. To achieve this, the linear U_s versus U_p relationship was employed to fit the curve of the Hugoniot pressure versus the nominal volumetric compressive strain. The term U_s is the shock-wave velocity. The term U_p is the particle velocity of the projectile and the measured value of U_p was assigned to all the 8-node linear-brick (C3D8R) elements when the FE model was started, as discussed in detail later. However, immediately after initial contact of the projectile with the composite, these elements for the projectile were converted to continuum particle (PC3D) elements and the value of U_p assigned to the particle elements was then continually updated based upon the loading conditions on the particles after the initial contact. Assuming the usual linear U_s versus U_p relationship, then the Hugoniot pressure versus the nominal volumetric compressive strain equation is given by:

$$p_H = \frac{\rho_0 c_0^2 \eta}{(1 - s\eta)^2} \quad (7-5)$$

where the fitting coefficient, s , is the slope of the linear relationship between U_s and U_p :

$$U_s = c_0 + sU_p \quad (7-6)$$

where c_0 is the reference speed of sound in the gelatine projectile. With the above assumptions, the relationship between the pressure, p , and the nominal volumetric compressive strain, η may now be written as:

$$p = \frac{\rho_0 c_0^2 \eta}{(1 - s\eta)^2} \left(1 - \frac{\Gamma_0 \eta}{2}\right) + \Gamma_0 \rho_0 E_m \quad (7-7)$$

Thus, in the FE model, see below, Equations (7-5) and (7-6) were employed to define the parameters in the EOS for modelling the gelatine projectile and Equation (7-7) was used to predict the contact pressure between the gelatine projectile and the composite.

7.2.2 Modelling the response of the composites

7.2.2.1 The intralaminar damage model

The model for predicting the initiation of any intralaminar damage was implemented within the ‘Abaqus/Explicit 2018’ FE code, and was based upon Hashin’s theory [148,151,172], which assumes that the undamaged composite exhibits linear-elastic behaviour. In Hashin’s damage model, four different types of damage mechanisms, which arise from tensile fibre failure, compressive fibre failure, tensile matrix failure and compressive matrix failure, are employed to capture the initiation of intralaminar damage in the unidirectional-fibre sub-ply. Corresponding to the damage initiation mechanisms defined in Hashin’s criteria, four damage parameters were implemented in the damage evolution model, which arise from four failure modes mentioned above. Three damage variables which reflect fibre damage, matrix damage and shear damage can then be derived from these damage parameters. When any of these three damage variables meet the stated condition, then the these damaged elements are deleted from the model [148]. The related equations and derivation are detailed in Chapter 5.2.2.

7.2.2.2 The interlaminar damage model

Interlaminar damage typically involves the initiation and growth of delamination between the plies that make up the composite laminate and this was captured using the Abaqus built-in surface-based cohesive (i.e. interface) element using an energy-release approach [148]. The interface element was described via a cohesive (i.e. damage) surface law [118,131,154–156] and is in the form of a bilinear cohesive law for a linear-softening material model. This damage law is divided into two steps. Before the initiation of any delamination, the relationship possesses linear-elastic behaviour. Once the damage criterion is satisfied, the cohesive stiffness degrades linearly until separation of the interface, i.e. the delamination now propagates when

the maximum failure displacement is attained. Finally, the initiation and growth of any intralaminar damage, as discussed above, significantly influences the extent of interlaminar damage and hence these two damage modes were modelled to be interactive in the Abaqus simulation, see Figure 5-1 in Chapter 5. The related equations and derivation are detailed in Chapter 5.2.3.

7.3 Experimental test specification and methods

7.3.1 Projectiles and composites

A 6 wt.% gelatine projectile, as the soft impactor to simulate birdstrike, was employed in this high-velocity soft impact tests. The detailed procedure, to manufacture the gelatine projectiles, is presented in Table 6-1 in Chapter 6. The gelatine projectiles had a nominal weight of 20 g, a nominal diameter of 24 mm and a nominal length of 40 mm. Due to their relatively low hardness, the gelatine projectiles initially tended to deform during the launching event from the gas-gun. To eliminate this problem, a plastic sabot was developed to maintain the shape of the gelatine projectile during the acceleration phase of the impact tests.

A woven T300 carbon-fibre reinforced PEEK composite and a woven T300 carbon-fibre reinforced ‘Toray 3631’ epoxy composite (as detailed in Chapter 3.2.2) were studied. The woven carbon-fibre ply possessed a $[0^\circ-90^\circ]$ architecture and had a nominal thickness of 2 mm. For the DIC measurement, the specimens were first painted on the rear-face using a white matt paint and then ‘speckled’ using a paintbrush to form the matt-black pattern.

7.3.2 Experimental investigations

The gas-gun (as introduced in Chapter 3.4.2) was employed to accelerate the projectiles in the impact tests. The same steel fixture with an opening of 70×70 mm, used in the study illustrated in Chapter 6, was employed to support the specimens. A new projectile and a new sabot were employed for each impact test. A 3D DIC system was used to measure the deformation of the rear-face of the specimens during impact loading. Two ‘Phantom Miro M/R/LC310’ high-speed cameras, supplied by Vision Research Phantom, USA, were employed. A pair of identical ‘Nikon’ lenses, with a fixed focal length of 50 mm, supplied by Nikon, UK, were used with these two cameras. During the tests, the recording rate of these two

cameras was set at 39,000 frames per second and they were triggered simultaneously by the signal generated from the infrared sensors.

After the impact experiments, visual inspections were undertaken on the composite specimens and photographs were taken from the rear-faces of the post-impacted specimens. In general, the type of damage suffered by the composites on the rear-face could be categorised as: (a) no visible damage present, (b) cracking observed, (c) fracture having occurred, and (d) perforation (i.e. penetration of the projectile through the specimen) having occurred. The main difference between ‘type (b) cracking’ and ‘type (c) fracture’ is whether there was fibre breakage observed. For ‘type (b) cracking’ this was defined as when cracks were only observed in the matrix. However, for ‘type (c) fracture’, fibre failure was also observed. Schematics of these descriptions for status of the post-impacted composites are shown in Figure 7-1.

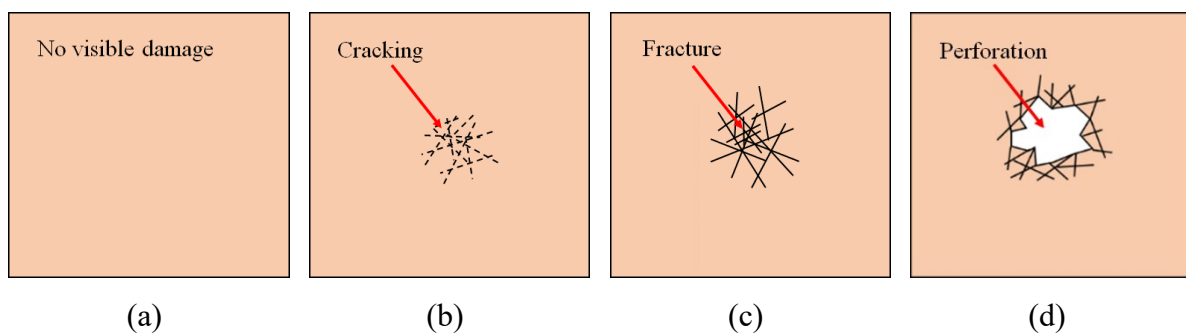


Figure 7-1. Schematics of the types of post-impact damage on the rear-face of the composites: (a) no visible damage, (b) cracking, (c) fracture and (d) perforation.

7.4 Experimental results

7.4.1 Deformation of the gelatine projectile

Figure 7-2 shows the deformation of the gelatine projectile recorded by a high-speed camera during an impact with the CF/PEEK composite specimen for an impact energy of 37 J. Within the resolvable time intervals, the time, t , corresponding to the initial contact was defined as 0.0 ms, as shown in Figure 7-2(b). It was found that at the beginning of the impact event (i.e. $t = 0.0$ ms) the shape of the gelatine projectile was well preserved, which ensured that the gelatine projectile impacted the centre of the specimen and then deformed symmetrically. However, in Figure 7-2(f) for $t = 0.4$ ms, the gelatine projectile can clearly be seen to be flowing freely after impact.

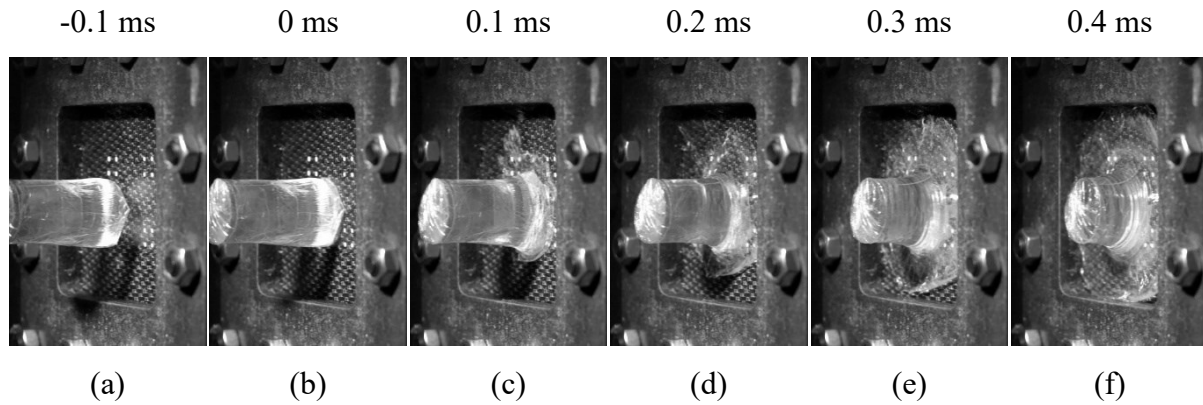


Figure 7-2. Deformation history of the gelatine projectile for a 37 J impact energy impacting the CF/PEEK composite.

7.4.2 Effects of the impact energy of the gelatine projectile

To study the effects of the impact energy on the response of the CF/PEEK composites subjected to soft impact-loading, these composites were impacted using gelatine projectiles fired at four different impact velocities, and hence with different impact energies. The testing configurations for investigating the effects of the impact energy on the impact response of the CF/PEEK specimens are given in Table 7-1.

Table 7-1. Test configurations for investing the effects of the impact velocity and energy on the CF/PEEK composites.

Test	Projectile	Projectile mass (g)	Impact velocity (m.s ⁻¹)	Impact energy (J)
GCP-I	Gelatine	20 ± 0.5	61 ± 2.5%	37 ± 5%
GCP-II	Gelatine	19 ± 0.5	75 ± 2.5%	53 ± 5%
GCP-III	Gelatine	20 ± 0.5	80 ± 2.5%	64 ± 5%
GCP-IV	Gelatine	20 ± 0.5	85 ± 2.5%	72 ± 5%

7.4.2.1 Comparison of the Digital Image Correlation (DIC) results

The 3D DIC system was employed to measure the major strain and out-of-plane displacement on the rear-faces of the composites. The main DIC results obtained from the CF/PEEK composites, impacted by gelatine projectiles at different energy levels, are summarised in Table 5. (It should be noted that due to fracture of the rear-face during ‘Test

GCP-IV' when an energy level of 72 J was used, no accurate value for the maximum major strain could be obtained from the DIC results for this test.) As the impact energy is steadily increased, the maximum major strain and maximum OOP displacement both increased in value.

Table 7-2. Main DIC results for the CF/PEEK composites impacted by the gelatine projectiles.

Test	Impact velocity (m.s⁻¹)	Impact energy (J)	Maximum OOP displacement (mm)
GCP-I	61 ± 2.5%	37 ± 5%	3.9 ± 3%
GCP-II	75 ± 2.5%	53 ± 5%	4.2 ± 3%
GCP-III	80 ± 2.5%	64 ± 5%	4.6 ± 3%
GCP-IV	85 ± 2.5%	72 ± 5%	4.8 ± 3%

Figures 7-3 present the typical DIC results obtained from the CF/PEEK composite impacted by a gelatine projectile with impact energy of 37 J. The OOP displacement contours, corresponding to different times during the impact tests, were also obtained from the DIC results and are shown in Figure 7-3(a) for an impact energy of 37 J. Similarly, the OOP displacements along the horizontal mid-section, during the loading and unloading process of the specimen, were also obtained and are shown in Figure 7-3(b).

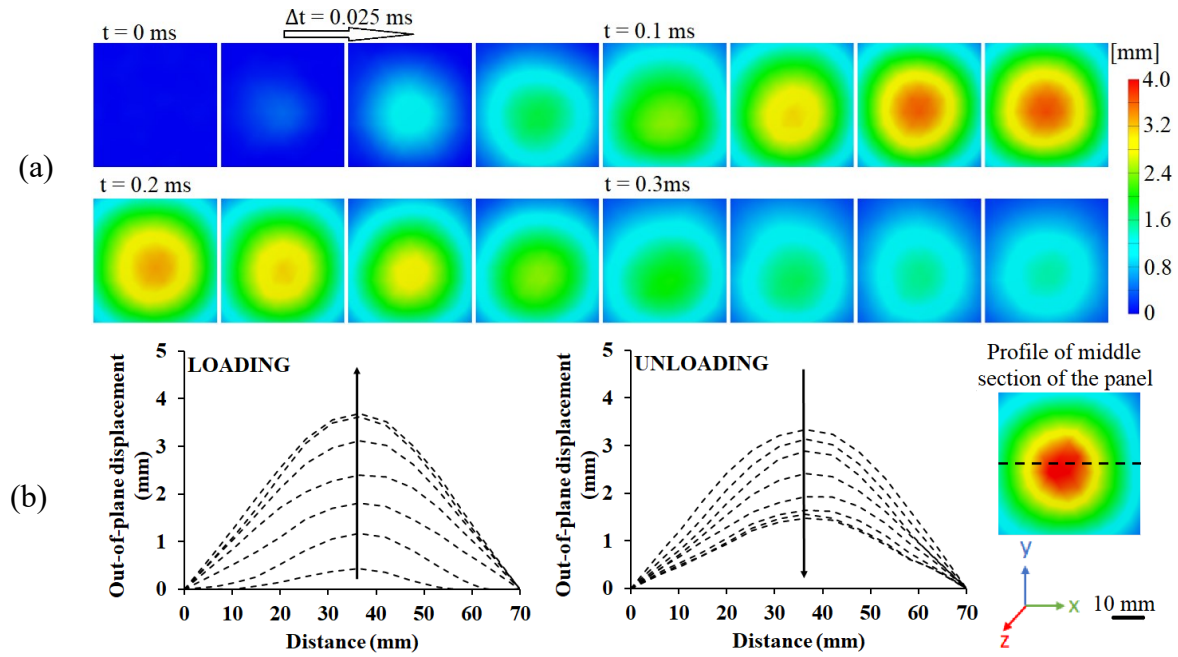


Figure 7-3. CF/PEEK composites impacted at a 37 J energy level: (a) the OOP displacement contours and (b) the evolution of the OOP displacement profiles (in intervals of 0.025 ms) during loading and unloading. (Inset picture, on right, shows a horizontal solid line where the profile section is taken.)

7.4.2.2 Comparison of the Digital Image Correlation (DIC) results

Representative photographs taken of the rear-faces of two of the gelatine-impacted CF/PEEK specimens are shown in Figure 7-4, along with corresponding magnified images of the central area. In Figure 7-4(a), where the CF/PEEK composite was impacted by a gelatine projectile with energy of 37 J, no visible damage was observed. The same observation, of no visible damage, was recorded for the CF/PEEK tests conducted at impact energy levels of 53 J and 64 J. In contrast, the CF/PEEK composite impacted using a gelatine projectile with an impact energy of 72 J has suffered ‘type (c)’ fracture damage, with cracking in the matrix mainly being confined to the central area of the specimen, as shown in Figure 7-4(b). Further, obvious fibre breakage was observed in this CF/PEEK composite specimen. Thus, it can be concluded that there is a critical impact energy between about 64 J and 72 J at which visible damage in the CF/PEEK composite is initiated.

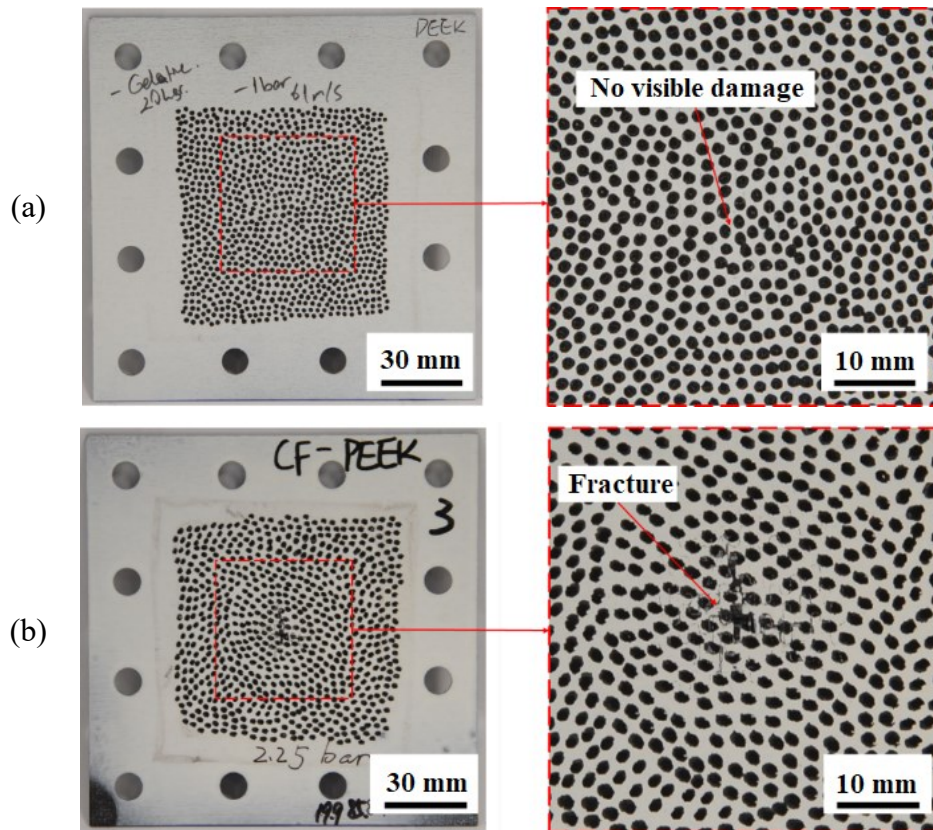


Figure 7-4. Photographs of the rear-faces of the CF/PEEK composites after impact: (a) for an energy of 37 J ('Test GCP-I') and (b) for an energy of 72 J ('Test GCP-IV').

7.4.3 Effects of the matrix system

To study the effects of the employed matrix system on the impact response of the composite laminates, CF/epoxy composite specimens were also impacted, using a soft-gelatine projectile, at an energy level of 38 J. The details of the testing conditions are summarised in Table 7-3 and the results are shown in Table 7-4 and Figure 7-5. As may be seen, the main effect of the matrix selected for the carbon-fibre composite is that the CF/PEEK composite ('Test GCP-I') impacted at an energy level of 37 J did not show any visible damage, whilst the CF/epoxy composite ('Test GCE-I') showed significant damage with 'type (b) cracking' being recorded.

Table 7-3. Gas-gun test conditions to study the effect of the matrix system.

Test	Projectile	Matrix system	Projectile mass (g)	Impact velocity (m.s ⁻¹)	Impact energy (J)
GCP-I	Gelatine	PEEK	20 ± 0.5	61 ± 2.5%	37 ± 5%
GCE-I	Gelatine	Epoxy	20 ± 0.5	62 ± 2.5%	38 ± 5%

Table 7-4. Results from the CF/PEEK and CF/Epoxy composites impacted by the gelatine projectiles.

Test	Matrix system	Impact velocity (m.s ⁻¹)	Energy (J)	Maximum major strain	Maximum OOP displacement (mm)
GCP-I	PEEK	61 ± 2.5%	37 ± 5%	0.013 ± 3%	3.9 ± 3%
GCE-I	Epoxy	62 ± 2.5%	38 ± 5%	N/A	4.0 ± 3%

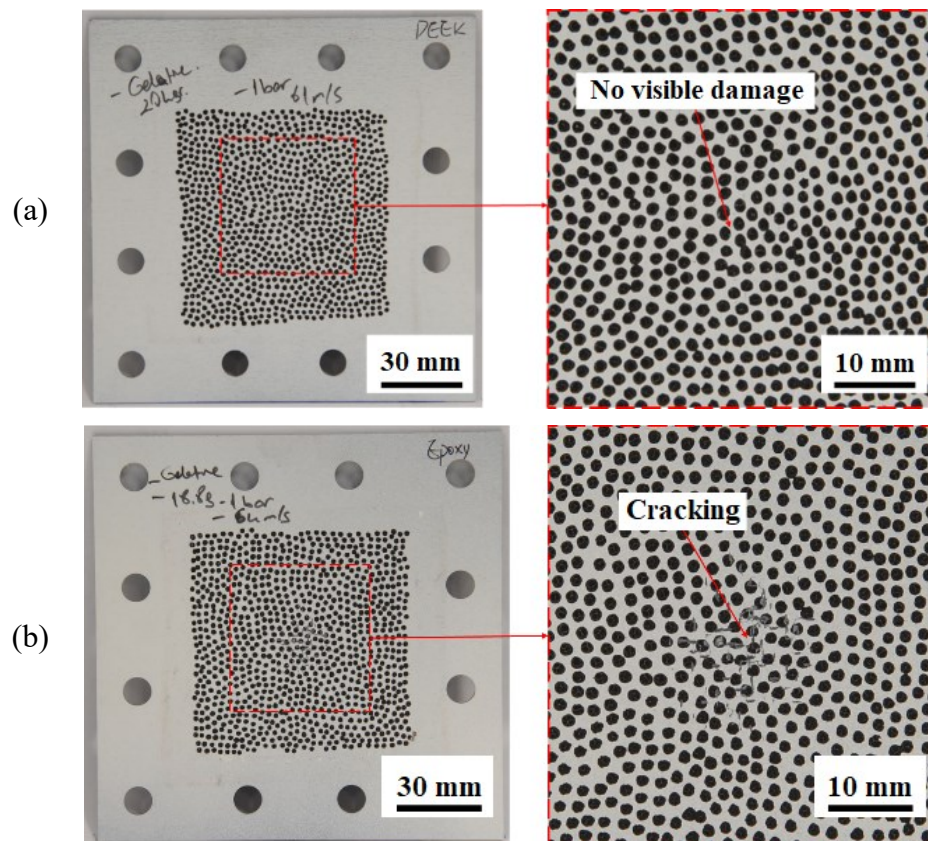


Figure 7-5. The rear-faces of the specimens after impact: (a) the CF/PEEK composite impacted at a 37 J energy level and (b) the CF/epoxy composite impacted at a 38 J energy level.

7.5 The Finite Element (FE) model

7.5.1 Model definition

As a discussed earlier, in order to model the soft-body impact on the composite test specimens a Finite-Element (FE) model was developed based upon a commercial software code, ‘Abaqus/Explicit 2018’. Within the FE model, the gelatine projectile was modelled using the Smoothed Particle Hydrodynamics (SPH) modelling technique [110]. The SPH method is a meshless Lagrangian technique where the solid FE mesh for the gelatine impactor is replaced by a set of discrete interacting particles. The gelatine projectile was first modelled using 8-node linear-brick (C3D8R) elements. However, upon initial contact of the projectile with the composite target specimen, these elements were converted to continuum particle (PC3D) elements, see Figure 7-6. The characteristic length for the PC3D elements was 0.5 mm, which was equivalent to half of the element size that was used for modelling the gelatine projectile with the C3D8R elements. The total mass of the projectile was equally distributed between all the 8-node linear-brick (C3D8R) elements or all the continuum particle (PC3D) elements.

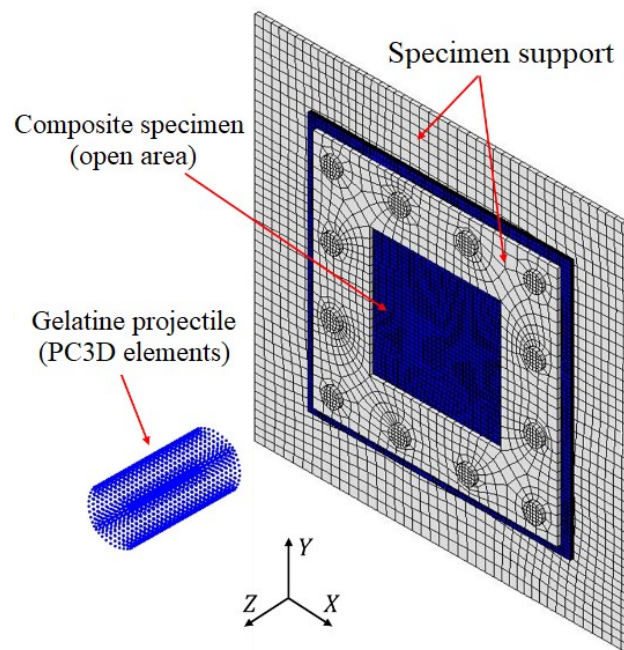


Figure 7-6. The FE model with PC3D elements.

Turning to the modelling of the composite specimen, the damage theories discussed earlier [148,151,172] were originally developed for unidirectional fibre-reinforced composite

plies. Hence, the $[0^\circ\text{-}90^\circ]$ woven carbon-fibre ply used for the CF/PEEK and CF/epoxy composites was represented as two unidirectional-fibre sub-plies, joined at right angles to the fibre direction. Thus, in the FE modelling, two unidirectional-fibre sub-plies were first created, with the thickness of each of the unidirectional-fibre sub-plies (i.e. 0.125 mm) being equal to half that of the thickness of the equivalent $[0^\circ\text{-}90^\circ]$ woven-fibre composite ply (i.e. 0.25 mm). These two unidirectional-fibre sub-plies were placed at right angles and then joined using ‘tie constraints’, to form a single equivalent $[0^\circ\text{-}90^\circ]$ woven-fibre composite ply, which has the same in-plane properties as the actual woven-fibre composite ply that was used in the composite specimens, see Figure 7-7. The elements employed in the FE model for the composite target test specimens were 8-node quadrilateral in-plane general-purpose continuum shell (SC8R) elements, with an element size of $1\text{ mm} \times 1\text{ mm}$. The interfaces between the composite plies were modelled using the cohesive surface law, which is again a built-in subroutine within the ‘Abaqus/Explicit 2018’ code [135,149,173]. The boundary conditions employed in the model were the same as those used in the gas-gun experiments. A general contact algorithm was used to govern the global contact in the numerical modelling and a friction coefficient of 0.2 was adopted for the global contact [174–176].

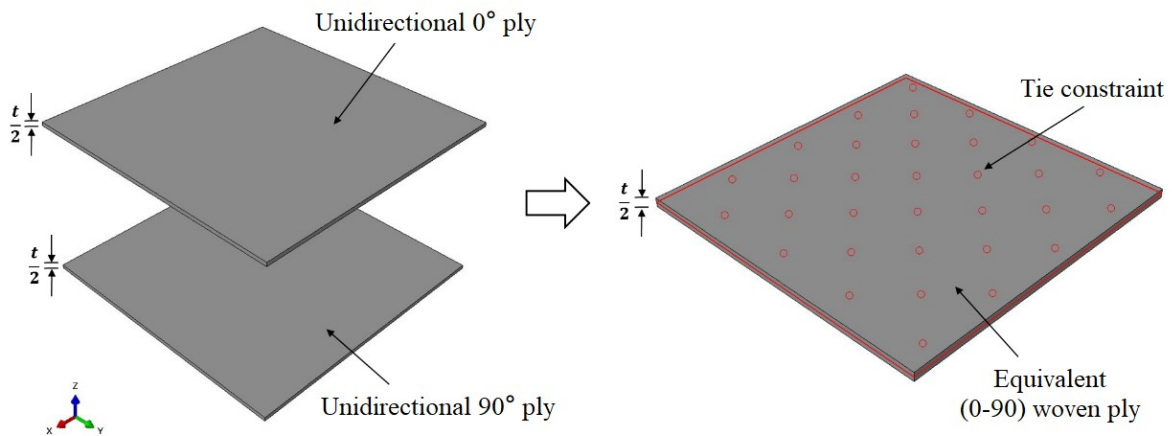


Figure 7-7. The creation of a single equivalent $[0^\circ\text{-}90^\circ]$ woven-fibre reinforced composite ply.

7.5.2 Input parameters

In order to use the SPH method for capturing the response of the soft-gelatine projectile, an equation of state (EOS) with suitable input parameters, as shown in Equations (7-5) and (7-6), is required for the modelling of the gelatine projectiles, see Chapter 7.2.1. The input

parameters required for the numerical modelling of the gelatine projectiles are shown in Table 7-5. For the composite specimen, it was defined using continuum shell elements and only the in-plane material properties are then required for the numerical modelling. However, the values of the cohesive stiffness, maximum cohesive strength and the various fracture energies do need to be inputted into the sub-routine which simulates the damage evolution in the composite via a linear-softening material model embedded in a bilinear cohesive law. The relevant material properties of the CF/PEEK and CF/Epoxy composites required for the FE modelling studies may be found from the literature [133,147,153,160–162,177,178] and are given in Table 7-6.

Table 7-5. Input properties for the FE modelling of the soft-gelatine projectile [37-39].

Properties	Reference density (g.cm ⁻³)	Dynamic viscosity (MPa.s)	Reference speed of sound (mm.s ⁻¹)	Slope of the U_s versus U_p curve	Grüneisen ratio
Values	1.06	1×10^{-6}	$c_0 = 1.45 \times 10^6$	$s = 1.87$	$\Gamma = 1.09$

Table 7-6. Input properties for the FE modelling studies of the composite [133,141,147,153,160–163].

Property	Unidirectional CF/PEEK sub-ply	Unidirectional CF/epoxy sub-ply
Moduli (GPa)	$E_{11} = 127; E_{22} = 10.3; G_{12} = 5.7$	$E_{11} = 125; E_{22} = 8.7; G_{12} = 4.3$
Poisson's ratio	$\nu_{12} = 0.3$	$\nu_{12} = 0.3$
Strength values (MPa)	$X^T = 2070; Y^T = 85$ $X^C = 1360; Y^C = 276$ $S^L = S^T = 186;$	$X^T = 1930; Y^T = 41$ $X^C = 1250; Y^C = 254$ $S^L = S^T = 110$
Ply fracture energies (kJ/m²)	$G_{Ic} _{ft} = 218; G_{Ic} _{fc} = 104$ $G_{Ic} _{mt} = 1.7; G_{Ic} _{mc} = 2.0$	$G_{Ic} _{ft} = 201; G_{Ic} _{fc} = 92;$ $G_{Ic} _{mt} = 0.6; G_{Ic} _{mc} = 1.5$
Interlaminar fracture energies⁺ (kJ/m²)	$G_{Ic} = 1.7; G_{IIc} = 2.0$	$G_{Ic} = 0.6; G_{IIc} = 1.5$
Benzeggagh–Kenane mode-mix exponent⁺	$\eta_{BK} = 1.09$	$\eta_{BK} = 2.09$
Initial cohesive law strength (MPa)	$t_{33}^0 = 43; t_{31}^0 = t_{32}^0 = 50$	$t_{33}^0 = 20; t_{31}^0 = t_{32}^0 = 34$
Initial cohesive law stiffness (MPa/mm)	$k = 6.4 \times 10^5$	$k = 6.2 \times 10^5$

Note: ⁺These properties are for interlaminar failure between two of the [0°-90°] woven-fibre plies.

G_{Ic} and G_{IIc} are the Mode I and Mode II interlaminar fracture energies between two [0°-90°] woven-fibre composite plies.

$G_{Ic}|_{ft}$ and $G_{Ic}|_{fc}$ are the tensile and compressive ply fracture energies of the unidirectional-fibre sub-ply in the longitudinal fibre-direction.

$G_{Ic}|_{mt}$ and $G_{Ic}|_{mc}$ are the tensile and compressive ply fracture energies of the unidirectional-fibre sub-ply in the transverse to fibre direction.

7.5.3 Implementation of the model

In the computation process a computation time-step was performed for every appropriate single element in the FE model. The numerical model is stopped when the defined total time for the impact event has expired. The flow chart of the main FE model is as shown in Figure 5-1 in Chapter 5.

7.6 Model validation and application

7.6.1 Validation of the model

7.6.1.1 The deformation of the gelatine projectile

The deformation histories of the gelatine projectile obtained from the experimental studies and predicted using the FE model for an impact test conducted at an energy level of 37 J on the CF/PEEK composite (i.e. ‘Test GCP-I’) are compared in Figure 7-8. The experimental results show that, after the initial contact with the composite specimen, the front of the gelatine projectile started to deform and flow to the periphery of the composite specimen. Correspondingly, the modelling results show a similar phenomenon, as shown in Figure 7-8(b). At a later stage of the impact event, see Figure 7-8(e), most of the gelatine projectile has deformed, flowed and spread over the surface of the composite specimen, and again the modelling studies accurately capture this behaviour of the gelatine projectile. Thus, the comparison between the experimental and numerical modelling results reveal that the SPH model for the relatively soft-gelatine projectile can indeed reproduce the experimental behaviour of the soft-gelatine projectile used in the gas-gun impact experiments.

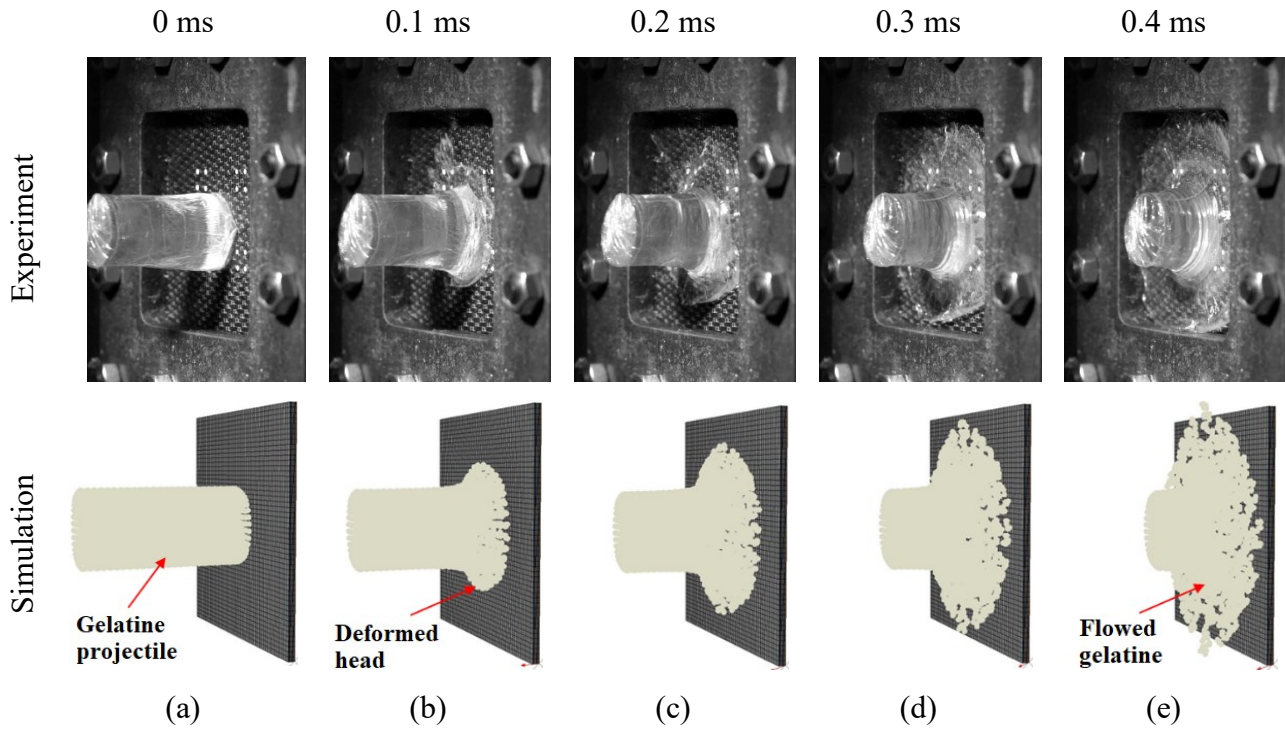


Figure 7-8. Deformation of the gelatine projectile obtained from the experimental studies and as predicted from the numerical FE model for the CF/PEEK composites at an impact energy of 37 J.

7.6.1.2 The CF/PEEK composites

Based on the DIC results obtained from the experiment conducted at an impact energy of 37 J using the gelatine projectile (i.e. ‘Test GCP-I’), the major strain and out-of-plane (OOP) displacement histories of the centre point for the rear-face of the composite test specimen can be extracted. The values of the maximum major strain and central OOP displacement predicted from the FE modelling studies are compared with the corresponding experimental results in Figure 7-9(a), and good agreement may be seen. To further confirm the accuracy of the numerical FE model, the predicted central OOP displacement versus time trace was also compared with the corresponding experimental results, see Figure 7-9(b). It can be seen from these results that, although the modelling studies gave somewhat lower maximum values than the experimentally measured values, the general trend and overall response of the composites were predicted extremely well using the numerical FE model. The slightly lower prediction values may be due to curvature effects in the woven material which the model could not fully capture.

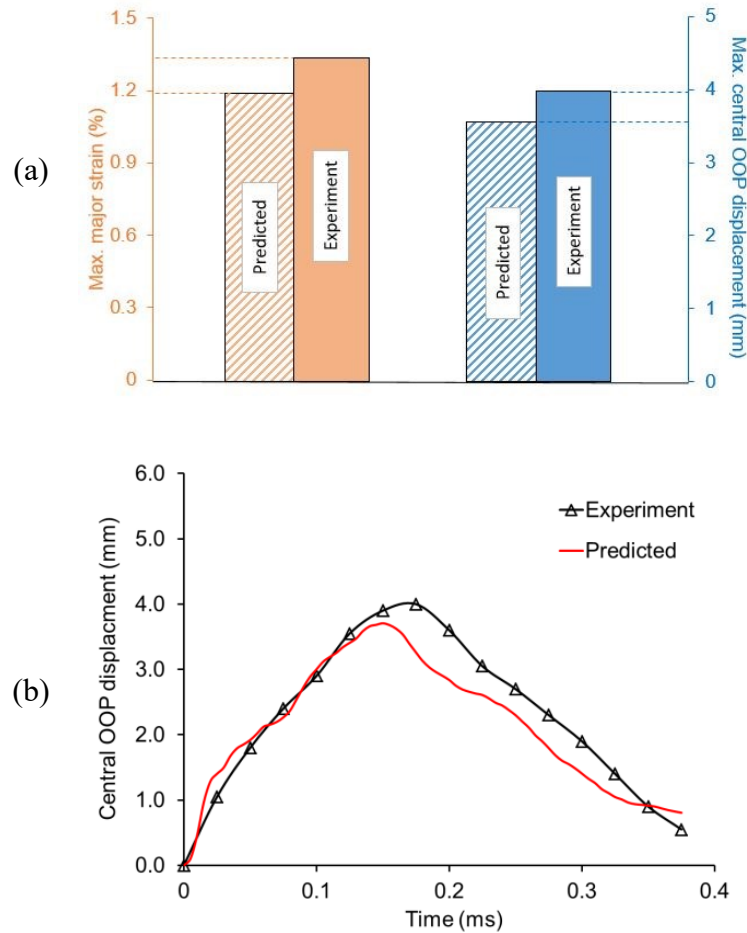


Figure 7-9. Comparison between the predicted and experimental results for the CF/PEEK composite at an impact energy of 37 J: (a) the maximum major strain and the out-of-plane (OOP) displacement and (b) the central OOP displacement versus time trace.

The next step is to assess the capability of the numerical FE model that has been developed to predict the impact damage created in the composite by the impact event, and two impact energies levels of 37 J and 72 J were so modelled. The experimental and predicted extents of damage at these two energy levels, which resulted in the CF/PEEK composites, are shown in Figures 7-10(a) and 7-10(b), respectively. (The ‘DAMAGESHR’ shown in the legend corresponds to the shear damage.) It was found that, at an energy level of 37 J, the prediction from the FE numerical modelling studies was that no visible impact damage would have been suffered by the composite specimen. This finding is in excellent agreement with the experimental results. When an impact energy level of 72 J was modelled, failure was predicted to be present only in the central area of the CF/PEEK composite, as shown in Figure 7-10(b). The experimental results revealed that some damage had indeed occurred in this region of the composite. In addition, the extent of the damage, as determined from the post-impact

experimental observations on this composite specimen, is accurately predicted by the numerical studies.

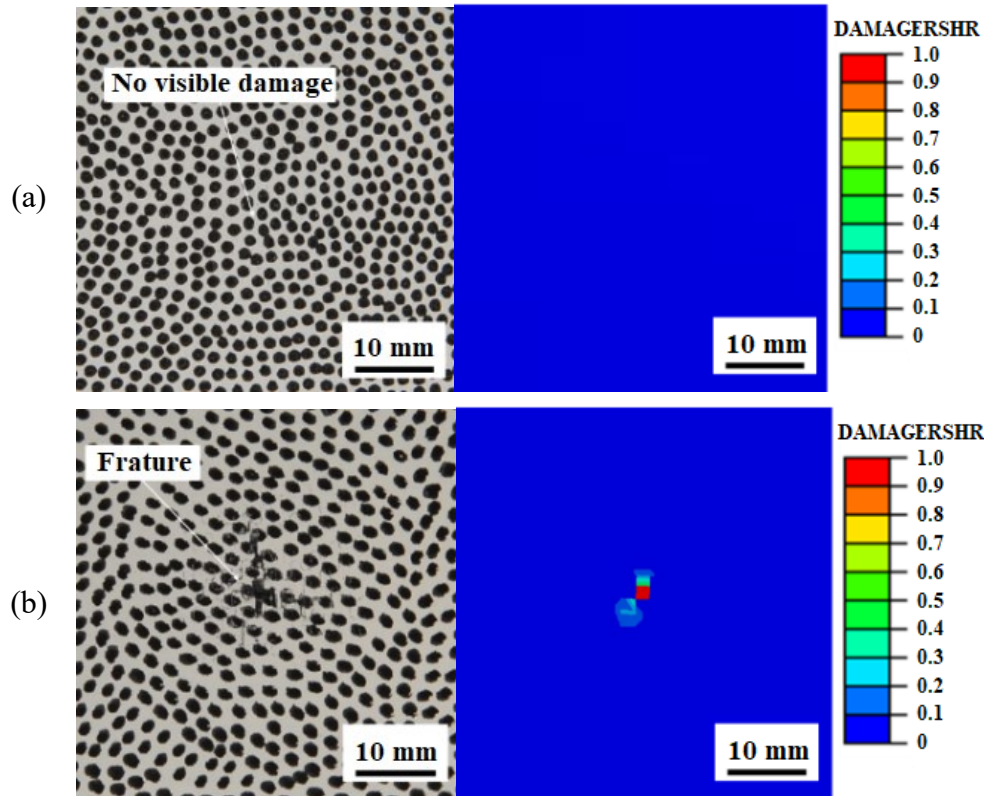


Figure 7-10. The experimentally measured and predicted degrees of damage resulting in the rear-face of the CF/PEEK composites at impact energies of (a) 37 J and (b) 72 J.

7.6.2 Application of the model

7.6.2.1 Predicting the deformation of the CF/PEEK and CF/Epoxy composites

To model the effects of the matrix system on the impact response of the composites, the central OOP displacement was predicted from the FE model for the CF/PEEK specimens impacted at a 37 J energy level and the CF/Epoxy specimens impacted at a 38 J energy level, as shown in Figure 7-11(a). The central OOP displacement versus time traces predicted for the CF/PEEK composite and for the CF/epoxy composite exhibited a very similar behaviour up to a peak value of the displacement followed by a gradual decrease. Figure 7-11(b) shows a comparison of the maximum central OOP displacements predicted in the FE model for the CF/PEEK ('Test GCP-I' at 37 J) and the CF/epoxy ('Test GCE-I' at 38 J) composites. When

impacted, the CF/epoxy composite ('Test GCE-I') is predicted from the FE modelling to undergo a maximum central OOP displacement of 3.9 mm, which is marginally higher than that of 3.7 mm for the CF/PEEK composite ('Test GCP-I'). These predicted values of the central OOP displacement for the two types of composite are also compared with the experimental results in Table 7-7, where very good agreement may be seen between the experimental measurements and the FE modelling simulations. The OOP displacement response for CF/PEEK and CF/Epoxy are very similar as both composites have the same carbon fibres with similar volume fraction.

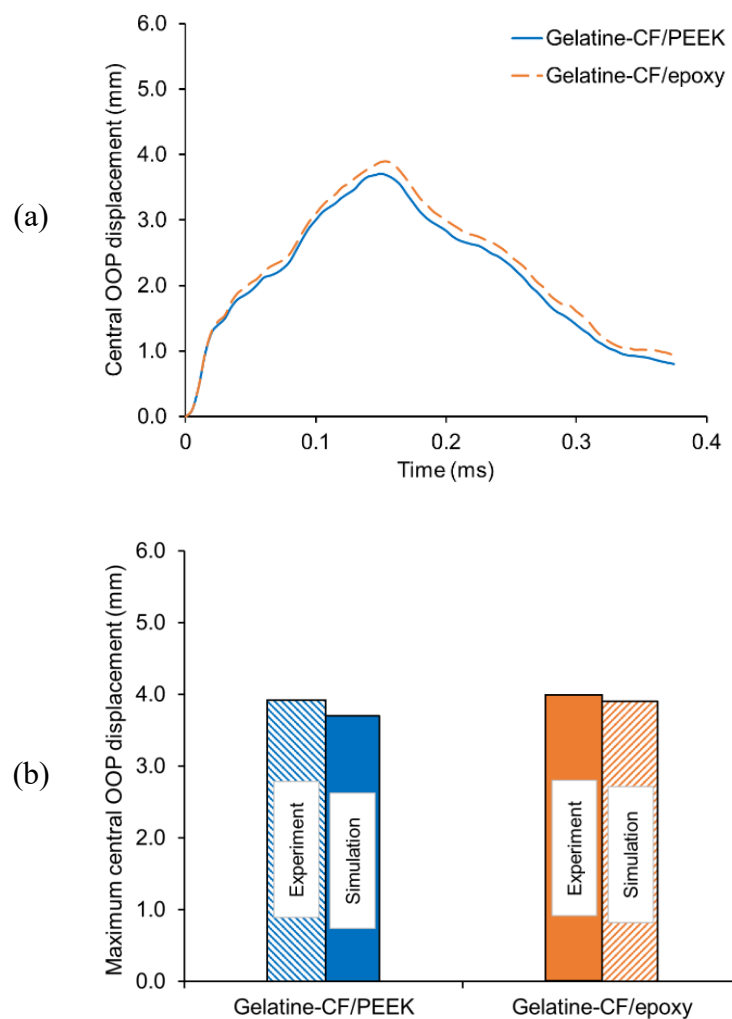


Figure 7-11. Predicted (a) central out-of-plane (OOP) displacement versus time trace and (b) the experimentally measured and predicted maximum OOP displacements for the CF/PEEK impacted at 37 J and the CF/epoxy impacted at 38 J.

Table 7-7. Comparison of the experimentally measured and the numerically predicted maximum central out-of-plane (OOP) displacement.

Composite	Energy level	Experiment (d_e)	Simulation (d_m)	Deviation ($ \frac{d_m-d_e}{d_e} \times 100\%$)
CF/PEEK	37 J	3.9 mm	3.7 mm	5.1%
CF/Epoxy	38 J	4.0 mm	3.9 mm	2.5%

7.6.2.2 Predicting the post-impact damage of the composites

A comparison of the post-impact damage in the composites obtained from the experiments and the FE numerical modelling results for the CF/PEEK and the CF/epoxy composites is shown in Figures 7-12(a) and 7-12(b), respectively. It can be seen that the predicted results for the CF/PEEK did not show any damage, which agrees fully with the experimental observations. On the other hand, the modelling results for the CF/epoxy predicted that some centrally-located damage would occur, which was indeed observed in the experimental studies. The evolution of damage was observed by plotting the derived damage variable, d_s , which is dependent on fibre and matrix failure, d_f and d_m respectively, as defined by Equation (5-12) in Chapter 5. The damage observed in CF/epoxy on the rear face was mostly localised matrix and fibre failure at the centre of the panel. The amount of energy expended in damage of the composite sample for CF/PEEK and CF/epoxy was very small relative to the incident impact energy. Most of the incident impact energy is transformed into elastic energy in the specimen which is then dissipated in friction at the support fixtures and in intrinsic damping, as the specimen vibrates after impact. Certainly, some of the incident impact energy is dissipated in plastic flow of the projectile.

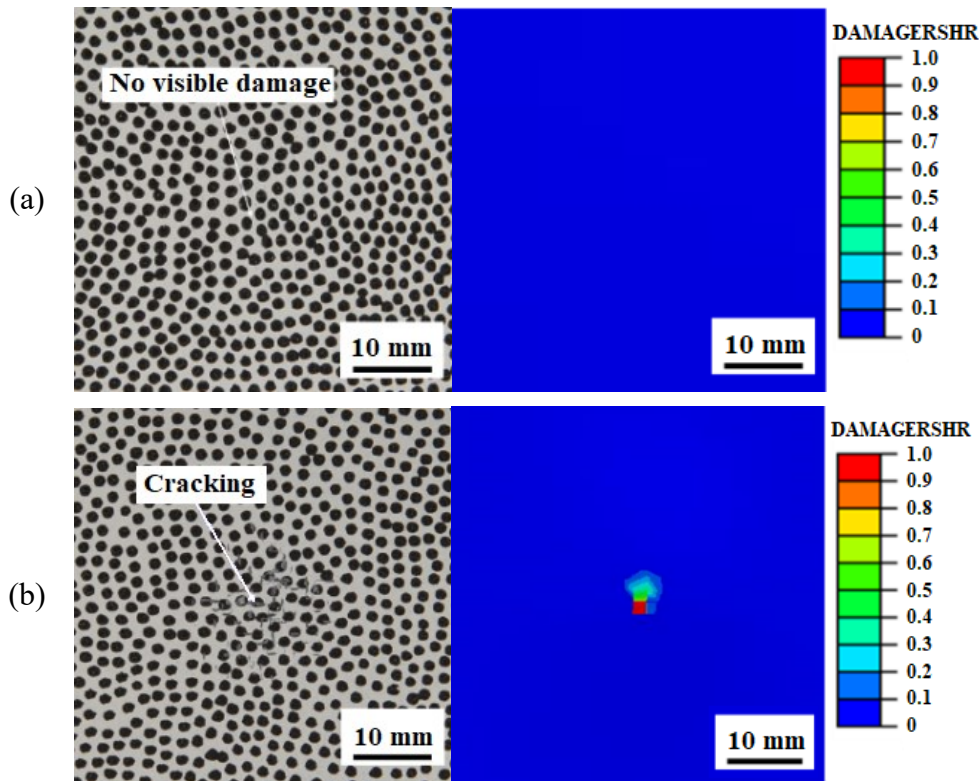


Figure 7-12. Comparison of the damage obtained from the experiments and the FE modelling: (a) the CF/PEEK composite impacted at 37 J and (b) CF/epoxy composite impacted at 38 J.

7.6.2.3 The deformation of the gelatine projectile

The numerical FE model was also employed to predict the average contact pressure between the soft-gelatine projectile and the composite specimen by using Equation (7-7), see Chapter 7.2.1. This parameter could not be readily experimentally measured in the gas-gun experiments. The contact pressure versus time histories were obtained from the FE models for (a) the CF/PEEK composite impacted at an impact energy of 37 J and (b) the CF/epoxy impacted an impact energy of 38 J, and the results are shown in Figure 7-13(a). It can be seen that the CF/PEEK and the CF/epoxy composites suffered a very similar average contact pressure history, with an initial short duration compressive phase giving rise to a relatively high initial contact pressure. The predicted maximum average contact pressures for the CF/PEEK and the CF/epoxy impact tests, when the relatively soft-gelatine projectile was used, are 10.7 MPa and 9.8 MPa, respectively, as shown in Figure 7-13(b).

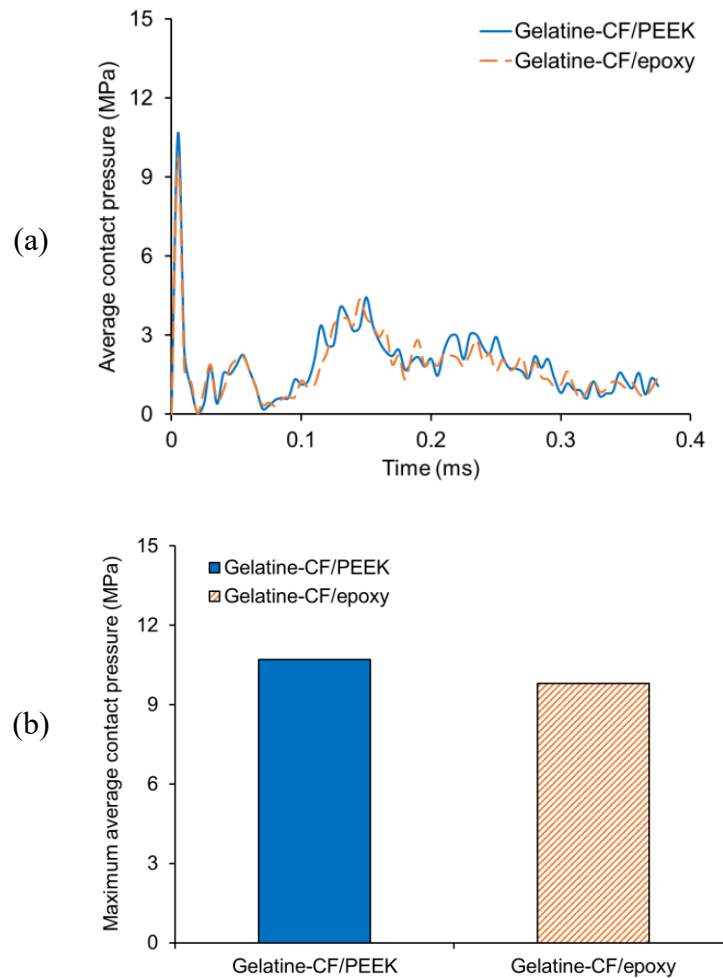


Figure 7-13. Numerical predictions from the FE model for: (a) the average contact pressure versus time history and (b) the maximum average contact pressure. For the CF/PEEK composite impacted at an energy of 37 J and (b) the CF/epoxy impacted an energy of 38 J.

7.7 Summary

This chapter has focussed on experimental and numerical studies of the response of woven-fabric CFRPs under impact loading by a soft projectile. A simple but reliable technique was proposed for the preparation of the relatively soft-gelatine projectiles and a plastic sabot was employed to maintain the shape of the gelatine projectile upon being launched from the gas-gun. A high-speed camera was used to record the deformation of the projectile during the impact event. The recorded frames showed that the gelatine projectile behaved as a viscoelastic-plastic fluid. The gas-gun tests were firstly performed using woven carbon-fibre reinforced poly(ether-ether ketone) (CF/PEEK) composite specimens, using the gelatine projectiles, at four different impact energy levels. Secondly, to investigate the effects of the matrix system on the impact response of the composites, woven carbon-fibre reinforced epoxy

(CF/Epoxy) were impacted using the gelatine projectiles. The experimental results demonstrated that the CF/epoxy composite exhibited a lower impact resistance and suffered more impact damage, compared with the CF/PEEK composite, when struck by the gelatine projectiles using a similar impact energy.

A finite-element (FE) numerical model was developed, which was based on the ‘Abaqus/Explicit 2018’ commercially-available software code, for predicting the behaviour of the projectile and the composite test specimen during the impact event. The FE numerical model has enabled (a) the deformation, (b) the initiation of damage, and (c) the evolution of such damage in the composite target specimens to be predicted, as well as the deformation and flow behaviour (and the contact pressure) of the projectile. The results from the numerical studies have been found to be in very good agreement with the experimental results.

In terms of design, woven architectures are often employed on the outside of composite laminates to generate a hybrid architecture. Woven composites do not have the stiffness of an equivalent laminate carbon-fibre material but they have the advantage that delamination and interfacial cracking does not occur so readily as has been shown by the above modelling and experiments. CF/PEEK is a very attractive material as the threshold for damage is higher than an equivalent CF/epoxy and this has also been confirmed by the modelling and experiments.

Chapter 8. Conclusions

8.1 Introduction

In this Ph.D. project, the performance of unidirectional cross-ply and woven-fabric carbon fibre composites, based upon thermoplastic-poly (ether-ether-ketone) (PEEK) and thermoset-epoxy polymers, under low-velocity and high-velocity impact tests using a rigid (i.e. steel or aluminium-alloy) impactor or a soft (i.e. gelatine) impactor, were studied. The three-dimensional (3D) Digital Image Correlation (DIC) technique was successfully employed to determine the out-of-plane (OOP) displacement of the composite specimens at different impact velocities, and hence different impact energy levels. The damage inflicted in the composite specimens was assessed by both a visual inspection and by ultrasonic C-scanning. An elastic finite-element (FE) model was developed to simulate the experimental results obtained from both the low-velocity and high-velocity impact tests. The results from this model were quantitatively validated against the impact response of the composites and captures the essential aspects of their impact behaviour and can be used to optimise the impact performance of such materials in industrial applications. The outcomes of the studies are summarised as follows.

8.2 Effects of the matrix system employed on the performance of CFRPs

The impact behaviour of a thermoplastic CF/PEEK composite and a toughened-thermoset CF/Epoxy composite, which possessed the same cross-ply lay-up of $[0_3/90_3]_{2s}$, under low-velocity and high-velocity impact using a rigid hemispherical-shaped impactor has been investigated. For the low-velocity drop-weight tests at the impact energies of 4.5, 7.5 and 10.5 J, the experimental results demonstrated that, at the same impact energy, the CF/PEEK

composite always possessed a superior impact performance compared with the CF/Epoxy. This was reflected by the CF/PEEK composite requiring a higher load to initiate damage and a smaller area of delamination damage being present in the CF/PEEK composite post-impact. In the high-velocity tests at an impact energy of 10.5 ± 0.3 J, the results revealed that the CF/PEEK and CF/Epoxy composites exhibited a similar damage pattern. However, the CF/Epoxy composite clearly exhibits the higher average damage area but with a significantly larger degree of scatter. The same conclusion can be found in the high-velocity impact tests on the woven-fabric composites using a soft gelatine projectile or a hard aluminium-alloy projectile with a hemi-spherical head. In the soft gelatine impact tests, the CF/PEEK composites exhibited a reduced maximum OOP deformation and less delamination damage compared to the CF/Epoxy composites. It was therefore concluded that the CF/PEEK composites have a superior damage resistance compared to the equivalent CF/Epoxy composites. In the hard impact tests, the CF/PEEK composite was more efficient in absorbing kinetic energy for these velocities and the damage area was less extensive than for the CF/Epoxy composite. In a word, CF/PEEK is a very attractive material as the threshold for damage is higher than an equivalent CF/epoxy and this has also been confirmed by the modelling and experiments.

8.3 Effects of the loading rate on the impact performance of CFRPs

As measured from the C-scans, the damage areas for the unidirectional cross-ply CF/PEEK and CF/Epoxy composite specimens from both the low-velocity and the high-velocity tests but with a similar impact energy of approximately 10.5 J, has been compared. Both the CF/PEEK and CF/Epoxy composite specimens clearly exhibit very significantly larger damage areas when subjected to the high-velocity tests of approximately 54.4 ± 1.0 m.s⁻¹, when compared with the damage areas associated with the low-velocity tests at 2.56 m.s⁻¹, by a factor of about four for both types of composite. This observation undoubtedly arises from several causes, since in the high-velocity tests (a) the CF/PEEK and CF/Epoxy composite specimens have less time to absorb the impact energy and therefore the deformation in the specimen is initially more localised around the impact site with an increased curvature of the specimen local to the impact site, which will induce more extensive localised delaminations throughout the composite plies, (b) these extensive localised delaminations will then readily propagate along the various 0°/90° ply interfaces and (c) the values of the interlaminar and

matrix fracture energies will be somewhat lower for both the CF/PEEK and CF/Epoxy composites due to the relatively high strain-rate effects associated with the high-velocity test and therefore delaminations and matrix damage will initiate and evolve more readily in such tests. The overall consequence is that the damage inflicted is more extensive in the high-velocity tests compared with the low-velocity tests, as indeed is observed from the greater area of delaminations in the C-scans of the high-velocity tests.

8.4 Effects of the hardness of the projectile on the impact response of CFRPs

The impact performance of woven-fabric CF/PEEK and CF/Epoxy under high-velocity impact using a soft gelatine or a hard aluminium-alloy projectile to simulate either a soft bird-strike or a hard foreign-body impact (e.g. runway debris), respectively. A high-speed camera was used to record the deformation of the projectile during the impact event. The recorded frames showed that the gelatine projectile behaved as a viscoelastic-plastic fluid. As to be expected, at similar impact energy levels, composite specimens are more vulnerable to the harder aluminium-alloy projectile than the soft-gelatine projectile. This is attributed to the higher contact pressures, and their localisation, associated with the aluminium-alloy projectile. In the soft impact case, due to the characteristics of the hydrodynamic loading, the impact energy is mainly absorbed by global bending of the laminates which also gives a longer contacting time allowing energy spread extensively from the impact zone. Conversely, the due to locally high stresses and indentation effects, the hard impact case, in which the target response is highly localised, presents delamination and matrix cracking at initial stage, followed by the fibre failure, which subsequently led to the catastrophic failure in the composites.

8.5 Modelling unidirectional CFRPs under low-velocity and high-velocity impact

An elastic, two-dimensional FE model has been developed to simulate the experimental results obtained from both the low-velocity and high-velocity impact tests. The model, which

is relatively computationally efficient, has been shown to simulate the loading responses of the composites by the impact event and the interlaminar and intralaminar damage induced.

For the low-velocity drop-weight tests, the results from the FE model for the load versus time curves from all the tests were in very good agreement with the experimental measurements. Considering the FE modelling of the high-velocity tests, the simulations have accurately predicted the shape and values of the OOP displacements as a function of the time-scale of the impact event, that the damage areas in the composites were always significantly higher from the high-velocity gas-gun tests than from the low-velocity drop-weight tests, at a comparable impact energy of 10.5 J, and that the CF/Epoxy composite suffered somewhat greater damage compared to the CF/PEEK composite.

The results from this model has been quantitatively validated against the impact response of the composites and captures the essential aspects of their impact behaviour and can be used (a) to simulate aspects of the high-velocity tests that cannot be directly measured and (b) to optimise the impact performance of such materials in industrial applications.

8.6 Modelling woven-fabric CFRPs under high-velocity impact

A FE numerical model was developed for predicting the behaviour of the soft gelatine projectile and the woven-fabric composite test specimen during the impact event. This model has enabled (a) the deformation, (b) the initiation of damage, and (c) the evolution of such damage in the composite target specimens to be predicted, as well as the deformation and flow behaviour (and the contact pressure) of the soft gelatine projectile. The results from the numerical studies have been found to be in very good agreement with the experimental results.

In terms of design, woven architectures are often employed on the outside of composite laminates to generate a hybrid architecture. Woven composites do not have the stiffness of an equivalent laminate carbon-fibre material but they have the advantage that delamination and interfacial cracking does not occur so readily as has been shown by the modelling and experiments.

Chapter 9. Future work

9.1 Introduction

Carbon fibre reinforced polymers (CFRPs) are, as the most promising materials for aerospace applications, facing great challenges when encounter with an impact with foreign bodies. Failure of CFRP structure is often a complex process and require a thorough understanding. Based on the understanding of the current studies, some further developments in terms of material characterisation methods, repaired CFRPs and composite damage model are suggested for further improving the understanding of such materials, and optimise the reliability and capability of the developed composite damage model.

9.2 Effects of impact loading on compression-after-impact (CAI) strength of CFRPs

An impact on a composite structure may result in invisible external damage but a dramatic degradation of compressive strength. Hence, a compression-after-impact (CAI) test is a crucial method to evaluate the residual strength of composite materials after impact. Its values are determined by the resistance to buckling of the sub-laminates, mainly affected by delamination, and rated as one of the major screening parameters for material election [64,179,180].

To extend the existing studies, the thermoplastic and thermoset composites under high-velocity impact and CAI loading conditions should be investigated. Such residual strength is an essential factor to be evaluated for critical composite aerostructures. In addition to that, the thermoplastic and thermoset composite specimens were impacted at a relatively low-velocity and a high-velocity by using a rigid metallic impactor or a soft gelatine impactor. The effects

of such impact, low-velocity impact versus high-velocity impact and soft impact versus hard impact, on CAI strength are worthy of further investigation.

9.3 Effects of environmental factors on impact performance of CFRPs

CFRPs are widely implementing on primary structural applications of the latest commercial aircrafts which operate over a broad range of temperatures from -56.4°C to 55°C at ground level and as low as -80°C at an altitude of 12,000 m [181–183]. Under such extremes of the ambient air temperature and thermo-mechanical conditions, the performance of aircrafts should be established and satisfy the airworthiness certification requirements.

With this in mind, many researchers have studied the impact performance of CFRPs under impact loading at low and high temperatures [171,184–187]. In a certain extent, low temperature has led to greater damage extension due to worse dissipation of energy, whereas the ductile and viscoelastic behaviours of CFRPs, being enhanced at elevated temperature, weaken their stiffness and residual strength and further influence failure response. In this Ph.D. project, all tests were performed in a lab environment at constant room temperature and humidity. Therefore, a better understanding of the impact behaviour of such composite materials under extreme environmental conditions is needed.

9.4 Performance of the repaired CFRPs under impact loadings

An area of interest when considering the use of composite materials for aircraft is the impact performance of repaired composites. The mechanical properties of these materials after repair are of great interest because the life of an aircraft is significantly longer than some of its components, meaning repair is critical. It is especially important for composite materials because the residual strength is significantly reduced after damage, but this drawback can be mitigated through the use of repair techniques. Currently, there are two main techniques, i.e. patch repair and scarf joint [188–190], that are well developed. Hence, the impact performance of the composites repaired via such techniques is of particular interest.

As there are a number of types of impact that an aircraft will be subjected to, the performance of repaired composites under drop-weight impact and gas gun impact are of great worth to be investigated. In addition, once a damaged sample has been repaired and the component replaced on the aircraft, it is not often hit in the same place again and so consideration has to be taken for how the repair performs when the impact is at a distance from the original impact. Therefore, the impact behaviour of joint area where has been approved as the weakest part of the repair should be well studied [191,192].

List of publication

Journal papers

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Appendices

Appendix A: Effects of impactor geometry on the low-velocity impact behaviour of fibre-reinforced composites: an experimental investigation

A.1 Introduction

The present appendix describes a detailed experimental investigation on the relatively low-velocity (i.e. $<10 \text{ m.s}^{-1}$) impact behaviour of such composite laminates. In particular, the effects of the geometry of the impactor have been studied and two types of impactor were investigated: (a) a steel impactor with a hemispherical head and (b) a flat-ended steel impactor. They were employed to strike the composite specimens with an impact energy level of 15 J. After the impact experiments, all the composite laminates were inspected using ultrasonic C-scan tests to assess the damage that was induced by the two different types of impactor.

A.2 Experimental programme

A.2.1 Impactors

Two types of impactors, a hemispherical steel (i.e. a round-nosed steel (RNS)) impactor and a flat-ended (i.e. a flat-faced steel (FFS)) steel impactor, were employed in the experiments. The hemispherically-headed steel impactor and the flat-ended steel impactor were manufactured from stainless steel and are illustrated in Figure A-1. They had masses of 5.20 and 5.25 kg, respectively, and the impact velocities used were 2.40 m.s^{-1} and 2.39 m.s^{-1} respectively. The impact energy was 15 J in both cases. The impactor heads had a diameter of

16 mm, with the head of the flat-ended impactor having a chamfer of about 45° around its periphery, as shown in Figure A-1. The diameter of the main body of the impactor was 20 mm and the load cell, with a data sampling rate of 500 kHz, was located in the forward section of the impactor.



Figure A-1. Illustrations of: (a) the hemispherically-headed (RNS) steel and (b) the flat-ended steel (FFS) impactor with a 45° chamfer on the edge to the flat-ended cylindrical section.

A.2.2 Composite specimens

Composite laminates were manufactured from unidirectional (T700) carbon-fibre reinforced epoxy-matrix pre-pregs. The stacked prepregs, provided by Beian Ltd, China, were cured in an autoclave. Composite specimens for the drop-weight impact experiments were produced from the manufactured composite panels and the cured composite laminates had a thickness of 3 mm. All the composite specimens possessed the same lay-up of $[+45_3/0_3/-45_3/0_3]_s$. The dimensions of the composite specimens were 150 mm $\times$ 100 mm, as defined in the ASTM D7136 standard [47].

A.2.3 Experimental procedures

The drop-weight impact experiments were performed using the CEAST drop-tower system, as shown in Figure 3-5. In the experiments the hemispherically-headed impactor and the flat-ended steel impactor were employed to strike the composite specimens with 15 J of energy. Each target composite specimen was placed on a steel picture-frame which had outer

dimensions that matched those of the composite specimens and with a 125 mm × 75 mm cut-out. This assembly was clamped to the base of the drop-weight tower using four toggle clamps with rubber tips, which prevented slippage of the composite specimen during the impact test [47]. For each test, three replicate composite specimens were tested and the signal from the load cell was not subjected to any software filtering. All the tested composite specimens were inspected using Sonatest C-scan device with 5 MHz probe, as shown in Figure 3-8, to assess the extent of any impact-induced damage.

A.3 Experimental results

A.3.1 Loading response

The loading responses of the impacted composite specimens when struck by the hemispherical steel and the flat-ended steel are shown in Figure A-2. At the same impact energy level of 15 J, the average contact time of the flat-ended steel impacted specimens was about 1 ms (i.e. about 15%) shorter than that when using the hemispherical steel impactor. An average maximum impact load of 7.1 kN was measured on the flat-ended steel impactor upon striking the composite specimens. This is about 50% higher than that for the hemispherical steel impactor, where the average maximum impact load was 4.7 kN. Now, the hemispherical steel impactor provides a more localised loading of the composite specimen, relative to the flat-ended impactor, and hence the load rises more slowly to the maximum load (i.e. 4.7 kN). The maximum displacement measured for the composite specimen struck with the hemispherical impactor was 6 mm. For the flat-ended impactor, the maximum load (i.e. 7.1 kN) is notably higher and the maximum displacement in this case is reduced to 4.7 mm. The reasons for these observations are that, as the flat-ended impactor has a larger area upon which to act on the specimen, this impactor is arrested more quickly and the initial effective stiffness of the specimen is higher.

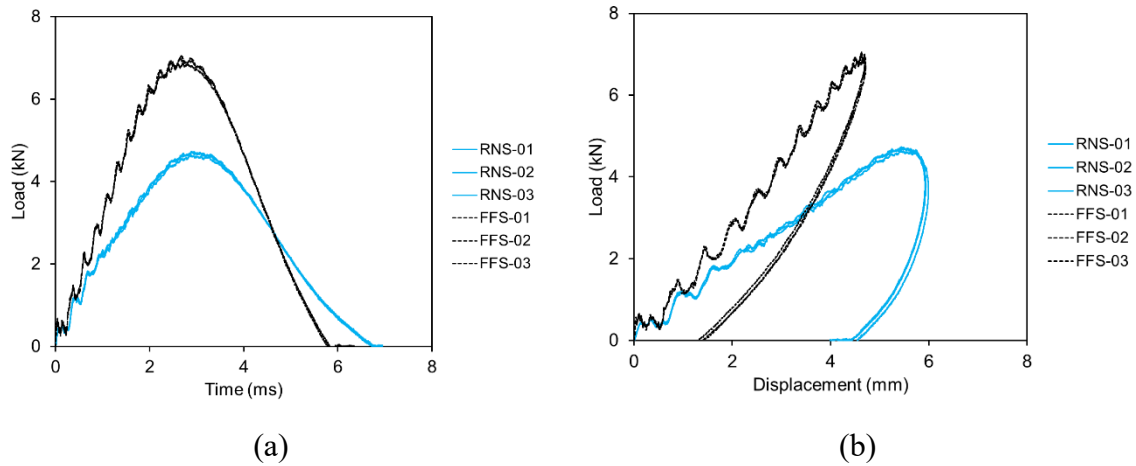


Figure A-2. Experimental results for (a) the load versus time and (b) the load versus displacement curves obtained from the impact experiments performed using the hemispherical steel (RNS) and the flat-ended steel (FFS) impactors. (Results from the three replicate tests are shown.)

A.3.1 Impact damage

For both types of impactor it was found that, since the composite has a lay-up of $[+45_3/0_3/-45_3/0_3]_s$, there was a propensity for the delamination to grow at the interfaces between the 45° and 0° plies, with the orientation of the lower ply determining the direction of delamination propagation. Figure A-3 shows the delamination maps, measured via the C-scan tests, obtained from the composite specimens impacted by the hemispherical steel and flat-ended steel impactors. The right-hand side scale indicates the location of the delamination as a function of the depth through the thickness of the specimen, where the red colour is representative of the front (impacted) surface and the blue colour is representative of the rear (non-impacted) surface of the composite specimen.)

The average areas of the damage maps produced by the hemispherical steel impactor and flat-ended steel impactors are similar. However, the hemispherical steel impacted specimens show a continuous damage area, which is centred around the point of impact, whilst the specimens impacted with the flat-ended steel impactor exhibit two separate damage areas, with a central zone exhibiting no damage. These differences in the extent and shape of the damage are explained below from the results of the numerical modelling studies.

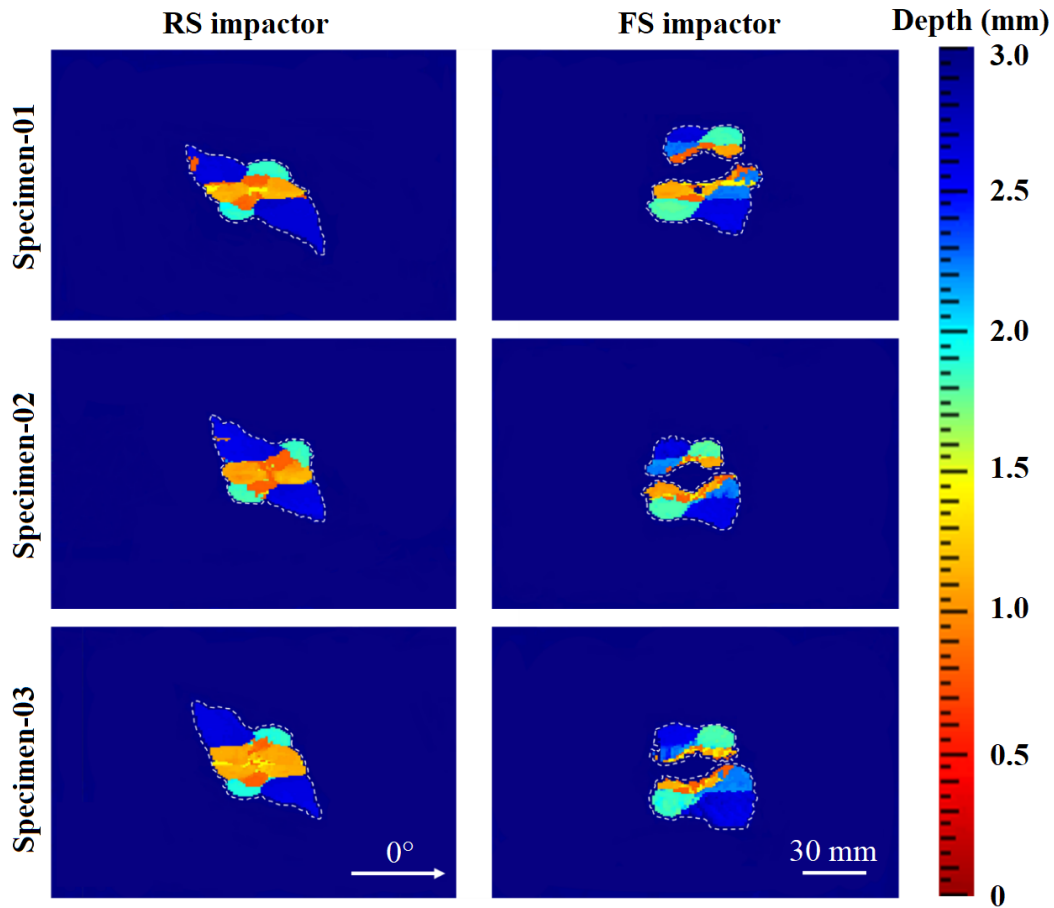


Figure A-3. Interlaminar C-scan damage maps obtained from the specimens impacted with the hemispherical steel (RNS) and the flat-ended steel (FFS) impactors. (Results from the three replicate tests are shown. The 0° fibre direction is indicated.)

A.4 Summary

A hemispherical steel impactor and a flat-ended steel impactor have been used to impact carbon fibre-reinforced/epoxy-matrix composite laminates with an energy level of 15 J. All the composite specimens were then inspected using ultrasonic C-scan testing to assess the impact damage. It was found that the composite specimens struck with a flat-ended steel impactor suffered a higher maximum load upon impact, but experienced a smaller out-of-plane displacement, than those struck with the hemispherical steel impactor. The comparison of the experimental C-scan images showed that the damage maps of the composite specimens impacted with the hemispherical steel and the flat-ended steel impactors were similar in area but with somewhat more areal damage occurring when the hemispherical steel impactor was used. However, the composite specimens impacted with the hemispherical steel impactor showed a continuous damage area, which was centred around the point of impact, whilst the flat-ended steel impacted specimens exhibited two separate damage areas, with a central zone

exhibiting no damage. These studies have increased our understanding of the behaviour of composite laminates when subjected to a relatively low-velocity impact and have proposed, and validated, an elastic-plastic three-dimensional numerical model for predicting the response of the laminate [193].

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