
Interleaved composites for controllable stiffness, shape memory and easy repair capabilities

Bohao Zhang

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Imperial College London

Department of Aeronautics

Imperial College London

South Kensington Campus,

London, SW7 2AZ

Declaration of Originality

This work is original research undertaken by the author in the Department of Aeronautics, Imperial College London under the supervision of Professor Paul Robinson and Professor Alexander Bismarck between May 2014 and May 2018. No part of this thesis has been submitted for consideration towards another degree at this or any other institution.

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April 2020

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Abstract

Thermoplastic interleaved carbon fibre/epoxy composites with controllable stiffness and shape memory capabilities have been developed in Imperial College London. Simple structures of this material have exhibited excellent re-shaping and shape recovery capabilities on heating. This thesis describes research exploring the properties and potential of these interleaved composites.

Firstly, an investigation was presented which examined the polystyrene (PS)-interleaved carbon fibre/epoxy composites in three deployable configurations. These were an integrated hinge, an interleaved linear actuator in a sine wavy profile and an expanded composite mesh. Experimental results showed that specimens of each configuration exhibited good re-shaping and shape recovery performance and confirmed the potential of interleaved composites for deployable structure applications.

The viscoelastic behaviour of the thermoplastic interleaf influences the stresses during re-shaping and the time required for shape recovery of these polystyrene interleaved composites. The viscoelastic behaviour of the interleaf material in shear and of interleaved composites in three-point flexure tests at elevated temperature was therefore investigated. Test results showed that the interleaved composites exhibited increasing stiffness when loaded at increasing crosshead speed. For specimens tested at higher crosshead speeds and then held at a fixed displacement, the load relaxed with time due to the stress relaxation of the thermoplastic interleaf. When unloaded, the specimens recovered towards the original shape. The shape recovery rate was highest at the beginning as the stored elastic stresses in the composites plies were released, but became slower as time progressed due to the viscoelastic behaviour of the thermoplastic interleaf.

A FE investigation was conducted into how finite element (FE) analysis can be used to model the shape memory capability of interleaved composites. A FE model was developed of a PS-interleaved specimen in the shape of a circular arc which was to be re-shaped to a flat configuration in its low

stiffness state at 120°C, cooled down to room temperature to return it to its high stiffness state and retain the flat shape and finally re-heated to 120°C to recover to the original shape. The 'Restart' facility in Abaqus was used to transfer data between the various stiffness states. Results showed that upon re-shaping, the bending stresses within composite plies were of right magnitude but were smaller than the predicted value from the simple beam theory. When cooled down, the FE predicted a 0.23 mm springback of the specimen upon removing the applied pressure. Finally, when re-heated, the specimen almost recovered to the original, circular arc configuration and the remaining bending stresses were small. Such a modelling approach could be applicable for the design of complex structures made of interleaved composites. The FE analysis was used to model the viscoelastic behaviour of polystyrene-interleaved composites subjected to three-point flexure tests at 120°C. Results predicted that the interleaved composites exhibited higher loads when loaded at higher crosshead speeds. When held at the 5.0 mm displacement, specimens of the higher crosshead loading speed cases relaxed in load with time towards the value in the low crosshead speed cases. Upon unloading, the specimens recovered rapidly at the beginning due to the release of stored elastic stresses in the composite plies, then gradually, more slowly towards the original shape, as the result of the viscoelastic behaviour of the PS interleaves. These predictions were consistent with the experiment results.

Laminated composites are susceptible to interlaminar damage due to impact. Conventional composite repair methods are often expensive, time consuming and require skilled operators. In the final part of this thesis, a concept is proposed for using a suitably interleaved carbon fibre/epoxy composite to enable easy repair of interlaminar damage. In this concept, the laminate is interleaved with a thermally bondable polymer with suitable mechanical properties so that damage due to low energy impacts will occur preferentially within the interleaves, i.e. the interleaf acts as a weak link. The damage could then be healed by applying heat and pressure to return the laminate to its pristine state. To demonstrate the concept, two interleaved carbon fibre/epoxy composites, one interleaved

with polylactide (PLA) and one with polycarbonate (PC), were subjected to three-point flexure tests and static indentation tests. It was showed that interlaminar damage occurred at the interleaf locations and this could be repaired by heating the laminate whilst applying external pressure. The shear strength, as measured in the flexure tests, showed a 66% overall recovery in the PLA-interleaved specimens after two damage-repair cycles and a 68% overall recovery in the PC-interleaved specimens after three cycles. In the static indentation tests, the critical load associated with the onset of interlaminar damage was measured and the recovery was 45% for the PLA-interleaved specimens after two damage-repair cycles and 30% for the PC-interleaved specimens after three cycles.

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List of publications and presentations

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Zhang B, Maples HA, Robinson P, Bismarck A, Li. W and Burgstaller C. Exploring deployable configurations using interleaved composites with shape memory capability. 21st International Conference on Composite Materials, 20th – 25th August 2017, Xi'an, China.

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List of nomenclatures

Amplitude	A	Melting temperature	T_m
Applied force	P	Overall strain energy	ALLSE
Austenite finish temperature	$T_{austenite, finish}$	Overall kinetic energy	ALLKE
Carbon fibre reinforced polymer	CFRP	Poisson's ratio	ν
Elastic modulus	E	Quasi-isotropic	QI
Flexural modulus	E_f	Recovery angle	$\theta_{recovery}$
Flexural stiffness	EI	Shear strength	τ^u
Flexural stiffness at high temperature	$(EI)_{HT}$	Springback angle	$\theta_{springback}$
Glass transition temperature	T_g	Storage elastic modulus	E'
Interlaminar shear strength	ILSS	Storage shear modulus	G'
Length of interleaved region	L	Thermoplastic interleaf	TI
Load causing shear failure	P_c	Thickness (Chapter 6)	t
Loss elastic modulus	E''	Time (Chapter 4)	t
Loss shear modulus	G''	Time at the end of constant loading crosshead rate phase	t_{cle}
Martensite finish temperature	$T_{martensite, finish}$	Time at the end of fixed displacement phase	t_{jde}
Maximum change in curvature	$\frac{1}{R_{max}}$	Unidirectional	UD
Maximum possible displacement	δ_{max}	Wavelength	λ
Maximum strain	ϵ_{max}	Width	w

List of abbreviations

bis(tricyclohexylphosphine)		Macro fibre composite	MFC
benzylidene ruthenium (IV) dichloride catalyst	Grubbs catalyst	poly(ethylene-co-acrylic acid)	PEAA
Bisphenol E cyanate ester	BECy	poly(methyl methacrylate)	PMMA
carboxyl-terminated butadiene acrylonitrile	CTBN	poly(methyl methacrylate-co-acrylamide)	p(MMA-co-AAm)
Compression after impact	CAI	poly(ϵ -caprolactone)	PCL
Computerised Numerical Control	CNC	poly[ethylene-co-(methacrylic acid)]	EMAA
dicyclopentadiene	DCPD	polyacrylamide	PAAm
Diels-Alder	DA	Polycarbonate	PC
Differential scanning calorimetry	DSC	Polylactide	PLA
diglycidyl ether bisphenol A	DGEBA	Polystyrene	PS
		polyurethane	PU
		retro-Diels-Alder	rDA
Double cantilever beam	DCB	Shape memory alloy	SMA
Dynamic mechanical thermal analyser	DMTA	Shape memory composite	SMC
		Shape memory hybrids	SMH
Elastic memory composites	EMC	Shape memory polymer	SMP
Ethylcyanoacrylate	ECA	Silicon Carbide	SiC
Hollow glass fibres	HGF		

List of units

Celsius	°C	Weight percentage	wt%
Celsius per minute	°C/min		
Degree	°		
Fahrenheit	°F		
Gigapascal	GPa		
Hertz	Hz		
Joule	J		
Kilogram	kg		
Kilonewton	kN		
Megahertz	MHz		
Megapascal	MPa		
Meter per second	m/s		
Micrometre	µm		
Milligram	mg		
Millimetre	mm		
Millimetre per minute	mm/min		
Newton	N		
Newton meter	Nm		
Percentage	%		
Percentage	%		
Pound per square inch	psi		
Radian per second	rad/s		
Second	s		

Chapter 1 Introduction

This thesis describes research exploring the potential of thermoplastic interleaved laminates. These materials have been shown to have controllable flexural stiffness and to exhibit shape memory capabilities, and may have the potential for space applications. Due to the capacity of launch vehicles, some components are required to be lightweight and possible to stow in a small volume (1). One approach to achieve this is to stow separate parts for launch and then to assemble these in space by astronauts (2). (Structures manufactured using this approach are defined as erectable structures.) However, such an approach increases the risk to astronauts in space and extra devices and equipment to assemble these parts increase launching weight and volume (3). Therefore, deployable structures have attracted significant interest. Deployable structures can be transformed from a small, stowed shape to a large, in-service shape and therefore can significantly contribute to the weight and volume saving for launch (2). Deployable structures can also find applications in morphing wings (4). A key challenge is to develop suitable materials for deployable structures. Such a material should be capable of deploying from a highly compacted configuration while providing significant stiffness to withstand launch load and to maintain the space components in their deployed, functional state. Thermoplastic-interleaved carbon fibre/epoxy composites with controllable stiffness and shape memory capabilities could be a suitable material for deployable structures. This material possesses a good flexural stiffness at room temperature and when heated to elevated temperature, the flexural stiffness is reduced significantly (For a polystyrene (PS) interleaved carbon fibre/epoxy composite, room temperature elastic flexural stiffness of up to 107 GPa and a reduction in this of over 98% at 120°C have been reported (5)). Such stiffness reduction allows this material to undergo large, reversible re-shaping with low actuation forces. Deployable space structures could therefore be a potential application for thermoplastic interleaved composites. Other applications could include

morphing structures and, by exploiting the re-shaping capability, the manufacture of simple 2D forms from flat interleaved composite sheets.

Another aspect of thermoplastic interleaved composites considered in this thesis is their potential to address the problem of repair of composite laminates. Laminated, fibre-reinforced polymer composites are vulnerable to interlaminar damage, for example when impacted, due to the relatively low strength of the polymer matrix. Such damage can significantly reduce the structural integrity of composites, particularly in compression. A common method to restore structural integrity with a flush or near-flush surface appearance is to remove the damaged material by machining, usually at a shallow angle, and then to adhesively bond a repair patch in its place. Alternatively, the damaged area may be repaired by drilling a hole in the centre of the damaged section and then injecting liquid resin which is subsequently cured. Unfortunately, these repair techniques are time-consuming, expensive and require considerable skill and experience. There has therefore been considerable interest in self-healing composites (6-21), but these introduce extra manufacturing complexity and while the healing efficiency can be good this reduces with further damage-repair cycles. Interleaved composites offer the possibility to tailor the interleaf properties to confine the damage due to a low energy impact to the interleaf layers and if these layers can be readily repaired by the application of heat and pressure then it may be possible to fully restore the composite to its pristine state.

1.1 Aims and objectives

The overall aim of this PhD study was to investigate thermoplastic-interleaved carbon fibre/epoxy composites for controllable stiffness, shape memory and easy repair capabilities.

1) The first objective was to explore three deployable configurations using the PS-interleaved carbon fibre/epoxy composites. The following tasks were conducted to achieve this objective:

- An integrated PS-interleaved carbon fibre/epoxy composite hinge was developed to

examine the shape memory capabilities of this structure.

- A linear actuator made of PS-interleaved carbon fibre/epoxy composites was manufactured to explore the linear actuation capability.
- An expanded mesh structure comprised of PS-interleaved carbon fibre/epoxy composites to demonstrate how the re-shaping capability can be exploited.

2) The second objective was to investigate the viscoelastic behaviour of PS-interleaved carbon fibre/epoxy composites at elevated temperature and to understand the importance of such behaviour on re-shaping and shape recovery capabilities of this composite. The following tasks were performed to achieve this objective:

- DMTA tests were conducted to investigate the frequency-dependent shear modulus properties of the PS interleaf.
- Three-point flexure tests were performed at 120°C to examine the loading rate dependence of PS-interleaved composites, to evaluate the load relaxation at a fixed displacement and to measure the shape recovery rate upon unloading.

3) The third objective was to investigate the possibility of using finite element (FE) analysis to model the shape memory capability of PS-interleaved carbon fibre/epoxy composites. It was then used to investigate the viscoelastic behaviour of PS-interleaved composites subjected to three-point flexure tests to compare prediction of the loading rate dependence, the load relaxation and the shape recovery rate upon unloading to the experimentally observed behaviour. The following tasks were performed to achieve this objective:

- A finite element model was developed of a PS-interleaved circular arc specimen as it was re-shaped to a flat configuration and then returned to its original shape.
- Stresses developed at re-shaping, cool down and re-heating stages were analysed.

- A finite element model was developed of PS-interleaved composite specimens subjected to three-point flexure tests at 120°C.
 - The loading rate dependence, the load relaxation at a fixed displacement and the shape recovery rate upon unloading were investigated. These were compared with the experimental results.
- 4) The fourth objective was to investigate an easy repair strategy for interlaminar damage using interleaved carbon fibre/epoxy composites. The following tasks were performed to achieve this objective:
- PLA-interleaved specimens were manufactured and then subjected to several damage-repair cycles in three-point flexure and static indentation tests to demonstrate the success of the concept.
 - PC-interleaved specimens were manufactured and tested in the same damage-repair cycles to investigate whether the PC improves the mechanical properties of the composite for this concept.

1.2 The structure of the thesis

This PhD thesis consists of seven chapters. Chapter 1 introduces the incentive, aims and objectives of this PhD study and describes the investigations to achieve the study of the thermoplastic interleaved carbon fibre/epoxy composite for controllable stiffness, shape memory and easy repair capabilities. Chapter 2 reviews the development of interleaved composites and materials with shape memory capability. This chapter also reviews the conventional composite repair and self-healing composite strategies for interlaminar damage of fibre-reinforced composites. Chapter 3 describes a study into the use of PS-interleaved carbon fibre/epoxy composites for three deployable configurations and includes the design, manufacture and testing of these structures. Chapter 4

describes the investigation of the viscoelastic behaviour of PS-interleaved carbon fibre/epoxy composites in flexural tests. This includes the investigation of the viscoelastic properties of the PS interleaf at 120°C. Chapter 5 describes a FE investigation into how FE analysis can be used to predict the shape memory capability of a PS-interleaved composite specimen. It also reports the FE investigation of the viscoelastic behaviour of PS-interleaved composite specimens subjected to three-point flexure tests at 120°C to examine the loading rate dependence, the load relaxation at a fixed displacement and the shape recovery rate upon unloading of the material. Chapter 6 describes the investigation of the easy repair concept of thermoplastic interleaved carbon fibre/epoxy composites by performing several damage-repair cycles in three-point flexure and static indentation tests. Chapter 7 reports the conclusions of this PhD study and suggests areas for further study.

Chapter 2 Literature review

2.1 Introduction

As already described in Chapter 1, this thesis presents an investigation of interleaved composites specifically focused on their controllable stiffness and shape memory characteristics and their potential for easy repair of interlaminar damage. This literature survey therefore commences with a review of previous work on controllable stiffness materials and this is followed by a summary of research into shape memory composites. Finally repair methods for composite laminates are described and the research related to self-healing composites is reviewed.

2.2 Composites with controllable stiffness

Controllable stiffness composites enable the stiffness to be reduced on demand and so have the potential for use in morphing and deployable structures (4). A reduction in stiffness results in lower actuation force required to achieve the morphing within a deployment. (Morphing and deployable structures are capable of significant shape changes on demand by mechanical actuation and by external stimuli such as heat and light (4).)

2.2.1 Controllable stiffness composites using polymer interleaves

Polymer interleaves were initially explored to improve the toughness of fibre-reinforced polymer matrix composites and it has been reported that interlaminar fracture toughness can be significantly improved (22-31). For example, the impact response of carbon fibre/epoxy laminates with interleaves was investigated (27). Specimens of 30-ply unidirectional carbon/epoxy prepreg tapes with 3 layers of poly(ethylene-co-acrylic acid) (PEAA) film at the midplane undertook drop-weight impact tests at a low velocity of around 1.1 m/s. Test results showed that the interleaved specimen

was able to suppress the initiation of delamination. In another study, the damage resistance of a carbon fibre/epoxy laminate interleaved with an elastomer-modified epoxy resin was investigated (25). The modified epoxy resin was prepared by adding either carboxyl-terminated butadiene acrylonitrile (CTBN) or polyurethane (PU) elastomers to epoxy resins. Both Mode I and Mode II critical strain energy release rates were significantly increased (25-50% and 63-130% respectively higher than the values of non-interleaved, baseline specimens) by the interleaf layer.

More recently, interleaved composites have been investigated to achieve a stiffness control capability on heating. A hybrid laminate with stiffness control capability was proposed in 2004 (32). This laminate, as shown in Figure 1, consisted of high-yield spring steel to provide structural stiffness and layers of PU-based shape memory polymer (SMP) for stiffness variation. Results from three-point bend tests in a dynamic thermal mechanical analyser showed that the flexural stiffness of this laminate reduced by over 90% on four laminate configurations, when the temperature was above the glass transition temperature of the SMP matrix. (This interleaved laminate also exhibited a shape memory capability due to the SMP matrix.)

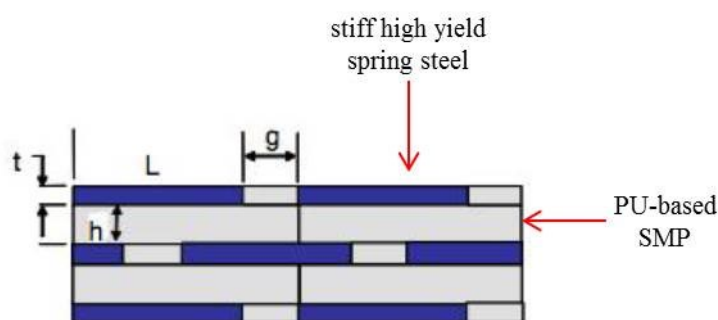


Figure 1. Sketch of a discontinuous platelet specimen. The blue and grey sections represent the stiff high yield spring steel and the polyurethane-based SMP respectively (32). (L and t are the length and thickness of spring steel elements, g is the gap of two steel elements and h is the thickness of SMP matrix.)

In another study, a multi-layered beam was developed to investigate the flexural stiffness variation capability upon heating (33). It consisted of an aluminium base beam adhesively bonded with polymer layers to upper and lower aluminium cover sheets (see Figure 2). Electrical heating blankets

were embedded within polymer layers to heat the beam. Specimens, with the base beam mounted as a cantilever, were subjected to tip deflection tests at room temperature and temperature above the glass transition temperature of the polymer. It was showed that the flexural stiffness of the multi-layered beam was reduced by 2 to 4 times, compared with the stiffness measured at room temperature. In addition, a further study investigated the actuation force required to deform the beam at elevated temperature, the energy needed to soften the polymer layer for the stiffness reduction and flexural strain developed during the deformation process (34).

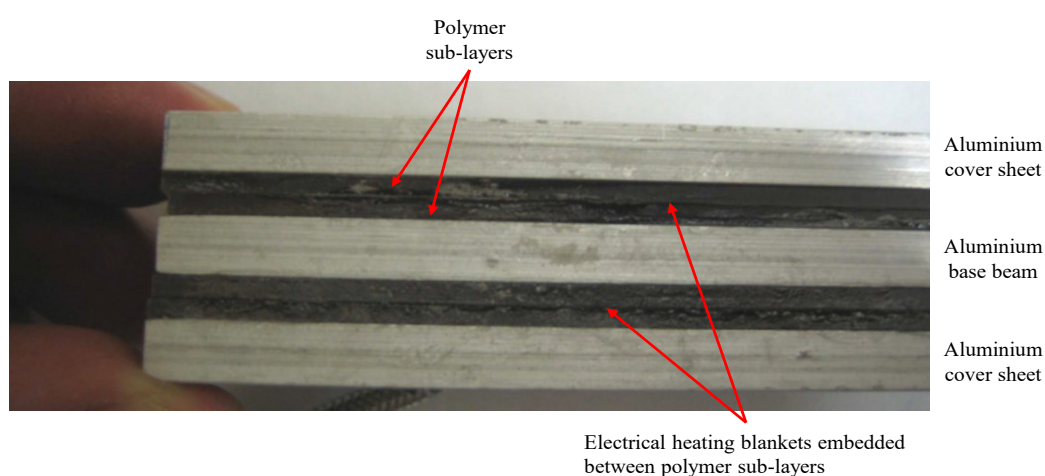


Figure 2. A photograph of a segment of a manufactured laminated beam demonstrating the aluminium sheets, the polymer layers and the embedded ultra-thin electric blanket heaters (33).

Finite element models were used to investigate the parameters affecting the flexural stiffness control of the beam (35). It was predicted that the flexural stiffness of the structure could be tuned at elevated temperature above the T_g of the polymer, by changing the thickness and modulus of the cover layer and polymer layer.

In addition to the flexural stiffness variation, a multi-layered composite structure was devised to demonstrate and characterise the adaptive bending-twist coupling stiffness on heating (36). It was comprised of a 0° unidirectional carbon fibre/epoxy base beam, two polymer interleaves and two unidirectional composite cover layers orientated at 23° (see Figure 3). Results from clamped

deflection tests conducted at three temperatures showed that the bending-twist coupling of this material was reduced by a factor of 10 when heated above the T_g of the polymer.



Figure 3. Schematic diagram of a polymer-interleaved composite beam specimen with different fibre orientations (36).
(Black layers are composite laminae and red layers are polymer interleaves.)

The stiffness control effect of a multi-layered carbon fibre/epoxy composite laminate with thermoplastic interleaves was also investigated at Imperial College London (5). This laminate was comprised of carbon fibre/epoxy laminate interleaved with polystyrene (PS) layers at various positions as shown in Figure 4a. Specimens of this material were subjected to three-point flexure tests at room temperature and at a temperature above the glass transition temperature (T_g) of the polystyrene and it was showed in Figure 4b that the flexural stiffness of the laminate was reduced to 2% of the value measured at room temperature. The loss of flexural stiffness was regained on cooling below the T_g of the polystyrene.

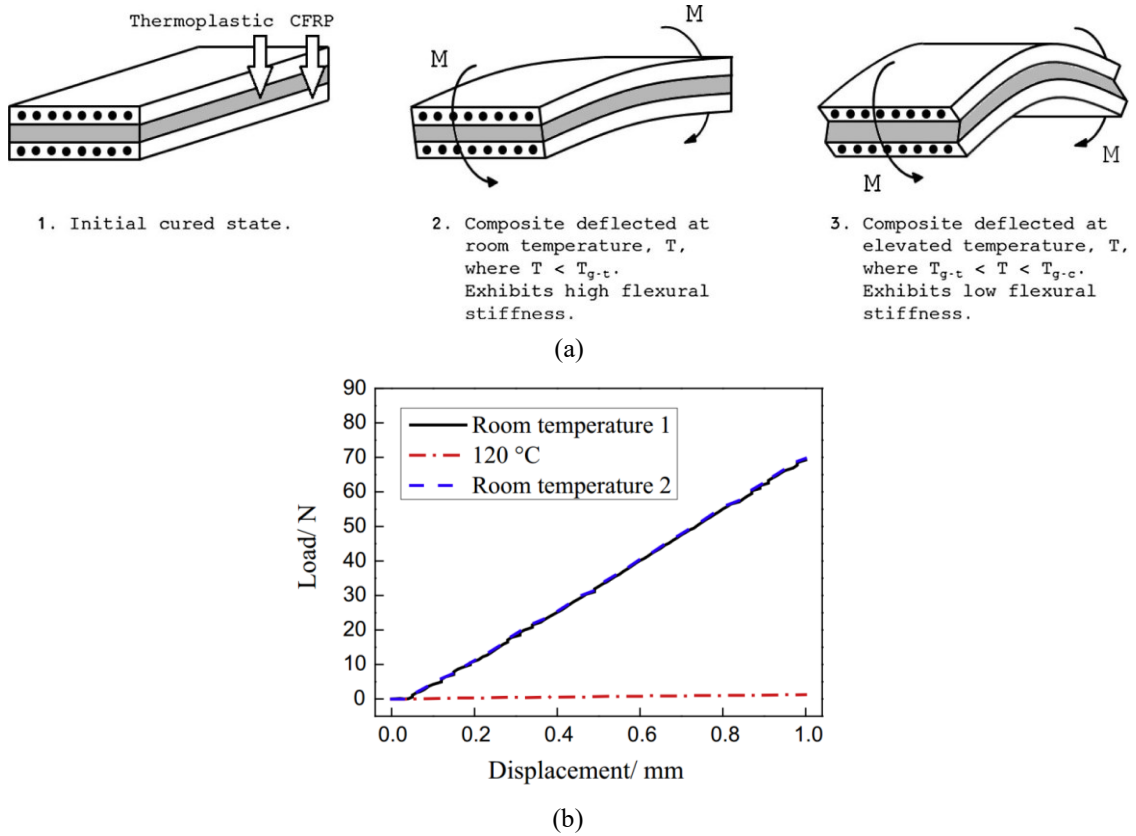


Figure 4. (a) A sketch of controllable stiffness mechanism of the polystyrene-interleaved composite and (b) three point flexure test results of a PS-interleaved specimen tested at room temperature and 120°C (5).

This PS-interleaved carbon fibre/epoxy composite laminate was then used to explore the benefit of the stiffness reduction in adaptive structures (37). A cantilever beam in which two 0° unidirectional carbon fibre/epoxy plies were separated by a single PS layer was manufactured (See Figure 5). A thin Kapton film encapsulated heater was bonded to the lower surface of the beam to raise the temperature to achieve the stiffness reduction and a Macro fibre composite (MFC) actuator (by Smart Material Corp under license from NASA) was attached to the upper surface to deflect the beam. For the same actuation voltage, the tip deflection of the beam was increased by 118% when the temperature of the cantilever beam was raised from room temperature to 120°C.

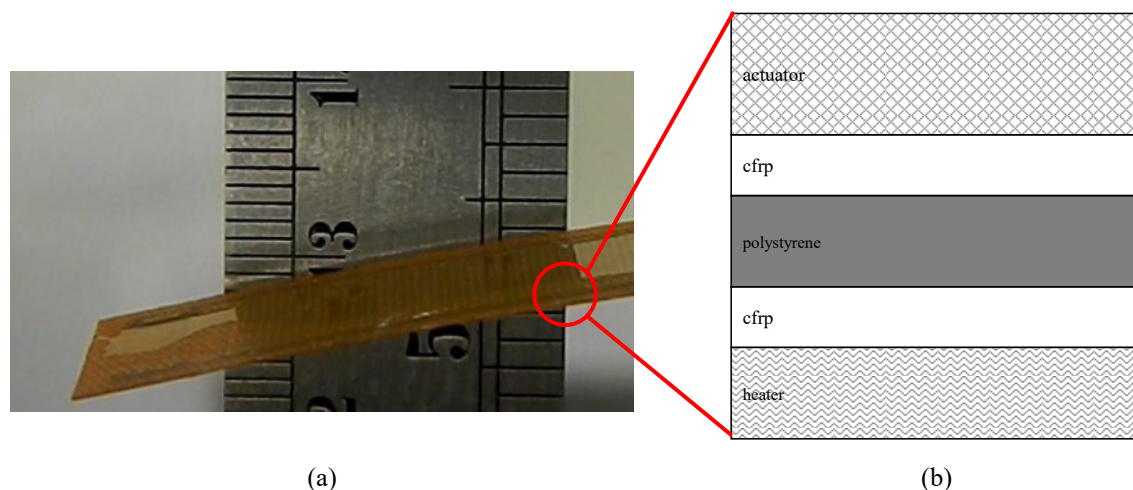


Figure 5. (a) The deflected shape of an interleaved composite cantilever beam with the embedded MFC actuator and (b) the cross section of the cantilever beam (37).

2.2.2 Controllable stiffness composites containing thermo-responsive fibre coatings

Another design of composites to exhibit controllable stiffness capability is to form a fibre/matrix interphase by coating a thermoplastic layer onto the fibres. Interphase modification has been used to toughen the bond of fibres to matrix and therefore to improve the interlaminar shear strength and fracture toughness of the composite (38, 39). For example, a CTBN layer was coated onto carbon fibres before resin infusion to investigate the effect on the impact performance (40). Impact test results showed that the composite failed at higher impact energy with the CTBN coating.

Interphase materials have also been used at Imperial College London to achieve controllable stiffness in fibre-reinforced composites (41). In this approach, a thin polyacrylamide (PAAm) layer was electrically coated onto non-sized carbon fibres which were then infused with epoxy matrix to manufacture the composite. Figure 6 illustrates how, when heated above the glass transition temperature (T_g) of the fibre coating, the fibres can slide relative to the matrix and so the stiffness is reduced. Three-point bending tests, conducted at the temperature above the T_g of PAAm, showed that the flexural stiffness of the composite was reduced by up to 88% compared with the value

measured at room temperature. It was also found that the composite recovered its full flexural stiffness when cooled down to room temperature.

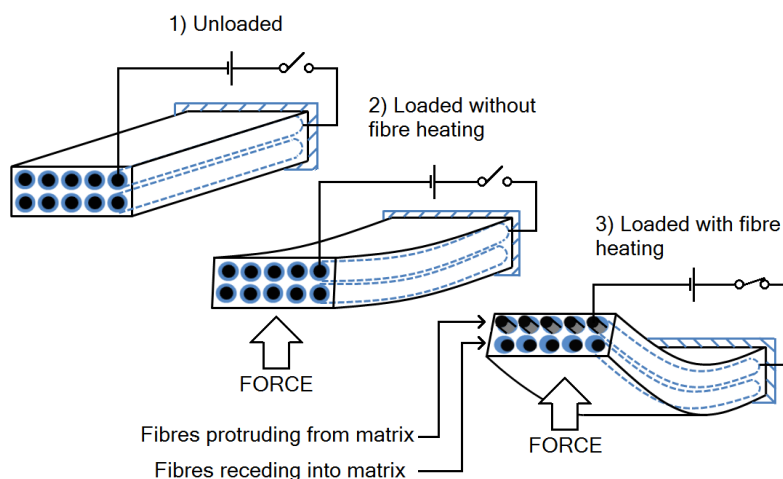


Figure 6. Schematic representation of the principle of the coated fibre/epoxy composite with controllable stiffness (41).

In another work conducted at Imperial College London (42), poly(methyl methacrylate) (PMMA) and poly(methyl methacrylate-co-acrylamide) [p(MMA-co-AAm)] layers were individually grafted onto carbon fibres and the grafted fibres were infused with an epoxy matrix to form composites. Specimens of the two composites were subjected to flexure test at 130°C and test results have shown that the flexural stiffness was reduced by 41% and 54% for PMMA- and p(MMA-co-AAM)-grafted carbon fibre/epoxy composites respectively. They also reported that the stiffness was fully recovered when cooled down to room temperature.

A carbon fibre/epoxy composite with a polystyrene layer electrically coated onto carbon fibres was also investigated (43). Flexure testing at elevated temperature (120°C) was performed on specimens of this material and a reduction of 36% in flexural stiffness was observed. Again, the flexural stiffness was fully restored when cooled to room temperature.

2.3 Materials with shape memory capability

Materials with shape memory capability are of interest for deployable structures. These materials can be manufactured in in-service configurations, then re-shaped into small, compacted configurations for launch from Earth and, finally, returned (or deployed) to the original, in-service configurations once in space. Materials which exhibit the shape memory capability include shape memory alloys (SMA), shape memory polymers (SMP) and shape memory composites (SMC). A brief summary of these materials is provided here.

2.3.1 Shape memory alloys

Shape memory alloys (SMAs) were among the earliest materials demonstrated to possess shape memory. These materials show re-shaping and shape recovery capability with the application of temperature and pressure (44). The shape memory effect of SMAs is due to the diffusionless solid-solid phase transformation between austenite phase and martensite phase. Figure 7 shows a typical shape memory process of a NiTi-based SMA material. On cooling to the martensite finish temperature ($T_{\text{martensite, finish}}$), the SMA changes from the austenite phase to the twinned martensite phase, but no macroscopic shape change can be observed. If external load is applied at this temperature, the martensite detwins and re-orient, and therefore the SMA is deformed. The ‘new’ shape is retained when the applied load is removed, as the detwinning process is irreversible. When the SMA is re-heated to austenite finish temperature ($T_{\text{austenite, finish}}$), the phase transforms from martensitic back to austenitic. The deformation induced in the detwinning process is recovered and thus the SMA recovers to its original shape. This process is known as one-way shape memory process because the material ‘memorises’ the shape at the austenite phase. A two-way shape memory process has also been developed. SMA with this capability can not only recover to the austenite shape, but also ‘memorises’ the configuration at the martensite phase when subsequently cooled down (44, 45). However, such two-way SMA is not commonly used due to the complexity of

training process, the reduction of the reversible strain and the degradation of the shape memory effect.

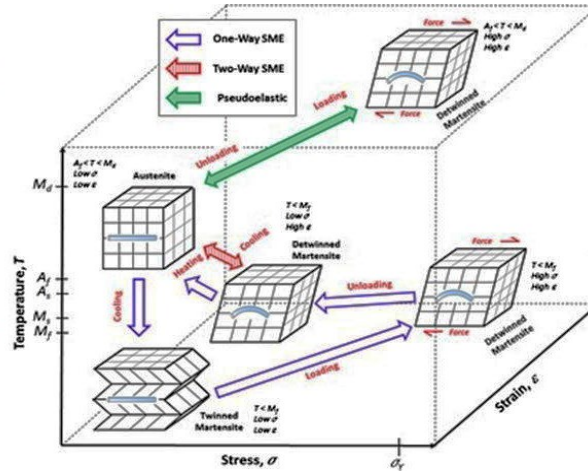


Figure 7. A sketch of shape memory mechanism of shape memory alloys (44).

A number of attempts have been successfully performed to use SMA as actuators for morphing wings (46, 47). For example, SMA wires were used to actuate aircraft engine inlet area change by heating the SMA wires to extend and contract (46). However, a significant disadvantage of the SMA is the operational recoverable strain (the reported highest recoverable strain is 8% (48)). This significantly limits the use of SMA for deploying large components such as telescope booms. Another drawback of the SMA is the low actuation efficiency (the work output is in the range of 10-15% of the total energy input from external sources such as light and heat to the SMA system in order to achieve shape change via the phase transformation) (49, 50). In addition, SMA shows a long response time and a slow cooling rate (51) which can substantially retard the deployment process.

2.3.2 Shape memory polymers

Another class of materials which exhibit shape memory capability are shape memory polymers (SMPs). The shape memory capability comes from two segments in the molecular structure of SMP; the elastic and transition segments. The elastic segment provides the high elasticity in the polymeric

structure, while the transition segment exhibits stiffness variation upon heating (52). The shape memory mechanism of SMP, as shown in Figure 8, is as follows. When heated above the glass transition temperature, T_g , the transition segment loses its stiffness and becomes soft. At this temperature, the SMP can be deformed with external load. If cooled down to room temperature under load, the SMP will retain the deformed shape. This is because the stiffness of the transition segment is fully regained. The cooling process is associated with large strain energy stored within the elastic segment. Finally, when re-heated while unconstrained to above the T_g , the transition segment is again in its low stiffness state, the stored strain energy is released and the SMP can return to its original shape. In addition to heat, the shape memory effect can be achieved by various stimuli including electricity (53), light (54-60), moisture (61-63), chemical reaction (64-66) and magnetism (67-74).

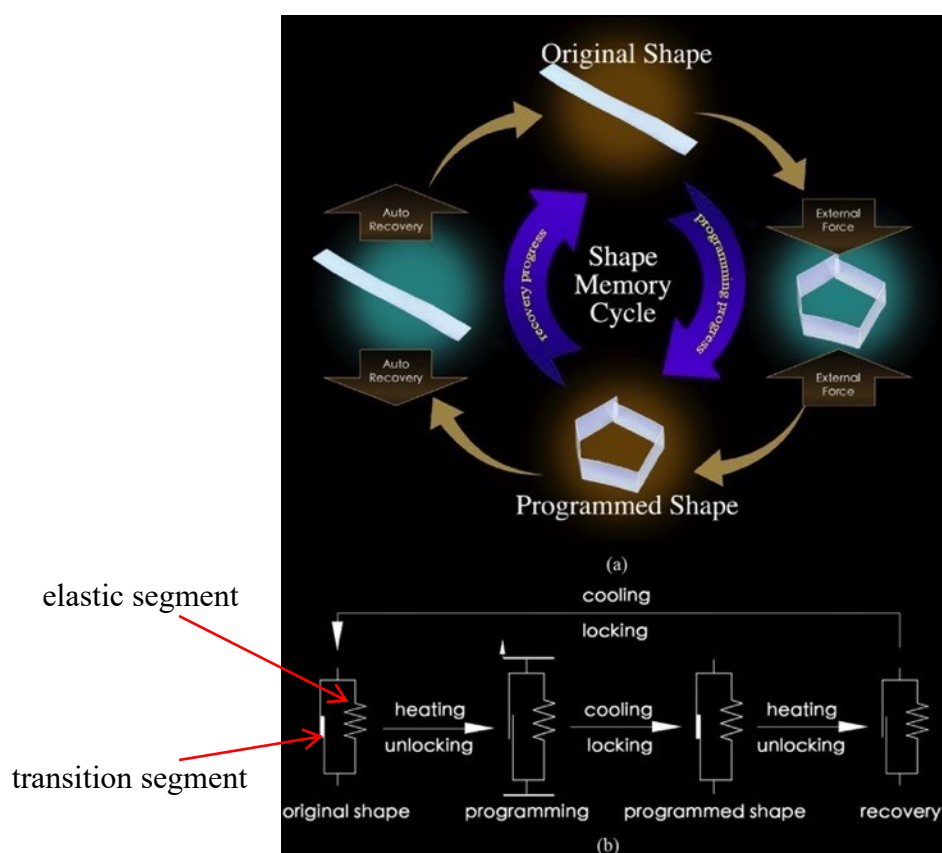


Figure 8. (a) A sketch of a thermomechanical re-shaping/shape recovery process of a SMP and (b) a diagram of spring-telescope rod system to demonstrate the mechanism of elastic segment (the spring) and transition segment (the rod) in SMPs (75).

SMP materials possess a substantial recoverable strain up to 1100% (76) which allows the material to be re-shaped into more complex configurations. Also, SMP materials are lightweight, have low material costs and also have low processing costs (52). However, a significant disadvantage of these materials is their poor mechanical properties below the glass transition temperature. For example, DiAPLEX polyurethane-based SMP and Veriflex polystyrene-based SMP (77) are commercially available SMP materials. The former possesses a tensile strength of around 30 MPa below its glass transition temperature (78), whereas the latter exhibits tensile strength of 23 MPa and tensile modulus of 1.2 GPa at room temperature (79). Due to these low mechanical properties, SMPs have limited potential for deployable structures.

2.3.3 Shape memory composites

A shape memory composite (SMC) is manufactured by hybridising a shape memory material with another shape memory material or with a material without shape memory capability. Two main types of shape memory composites have been developed; fibre-reinforced and nanoparticle-reinforced shape memory composites (80).

As mentioned before, SMP is not in general a suitable material for deployable structures due to its poor mechanical properties (77). To address this problem, carbon, glass and Kevlar fibres have been used as reinforcements to enhance the mechanical strength and stiffness of SMPs (81-83). Other reinforcements, including carbon nanotubes and carbon nanofibers (84-87), exfoliated nanoclay (88), silicon-based inorganic particles (89, 90) and cellulose nanowhiskers (91), have also been used to form nanoparticle-reinforced shape memory composites.

The shape memory effect of SMCs is identical to SMPs as mentioned earlier in section 2.3.2. Moreover, SMCs can activate the shape memory effect by using the inclusions or reinforcements as mediums for transferring heat. For example, continuous fibres can transfer heat to SMP matrix due to the electrical conductivity of the fibres (92, 93). In addition, carbon nanotubes (53, 94), carbon black

nanoparticles (95), optical fibres (59, 60) and magnetic nanoparticles (69-73) can implement the heat transfer process.

Studies on SMP-matrix composites reinforced with carbon powder (96), SiC particles (97) and carbon nanofibres (98) have been shown that the elastic modulus and strength can be significantly improved, compared with neat SMP matrix. However, such improvement is relatively weak when compared with what can be achieved with the use of continuous fibre reinforcements (81) and so continuous fibre-reinforced SMC materials are usually used as a structural material.

The next two sections will focus on the fibre-reinforced shape memory composites. The first section will describe fibre-reinforced shape memory composites which contain one or more constituents promising shape memory capability. The second section will introduce shape memory hybrid materials, which possess no shape memory constituents, and will describe research investigating the shape memory characteristics of an interleaved fibre-reinforced composite which is a type of shape memory hybrid materials.

2.3.3.1 Fibre-reinforced polymer matrix composites

A fibre-reinforced shape memory composite can be manufactured by embedding continuous or short fibres into a host matrix material (80) with either the fibres (such as SMA fibres), or the matrix (like SMP matrix), or both possessing the shape memory capability.

Both short SMA fibres (99, 100) and continuous SMA fibres (101-107) have been used for the improvement of the elevated-temperature mechanical performance of a host material. For example, an epoxy-matrix composite reinforced with SMA short fibres was developed (99). Dynamic flexure tests performed in dynamic mechanical thermal analyser at an elevated temperature above the austenite transformation temperature of the SMA fibres has shown that the storage modulus of the composite with 3.5 wt% of SMA fibres was up to 6 times higher than that of the pure epoxy matrix.

Despite the improved mechanical properties, the disadvantages of the SMA fibre-reinforced composite include interfacial failure which can occur between SMA fibres and matrix upon cooling down due to the different thermal expansion coefficients (108, 109), the complicated composite manufacture process where SMA fibres are pre-strained for the stress-induced austenite-martensite phase transformation (105) and the influence of surface treatment (110), matrix curing process (111) and pre-straining level of SMA fibres (112) on the properties of the composite. Relatively little published research literature has investigated the shape memory performance of the SMA fibre-reinforced composites. Instead studies have been conducted on SMA wire-embedded fibre-reinforced composites in which the SMA wires act as actuators (113-115). A problem with this material is the added weight to the composites because of the SMA wires. Other problems include the strength and durability of SMA wire-polymer matrix interface at different temperature conditions for load transfer and the SMA wire breakage when repeated shape memory is performed (115).

It has been mentioned previously that SMP shows poor stiffness and strength and so is not suitable for deployable structures. To address this problem, stiff fibres, including carbon fibres (116, 117), glass fibres and Kevlar fibres (81, 118, 119) have been used to reinforce the SMPs (i.e. to form fibre-reinforced SMP matrix composites). For example, a carbon fibre fabric reinforced styrene-based shape memory polymer composite was manufactured for the development of smart release devices (116). The tensile test results showed that the tensile stress at failure of the reinforced SMP matrix specimens was 4 times higher than that of the pure SMP matrix specimens. The smart release device prototype manufactured with this composite showed the re-shaping and shape recovery capability. A similar investigation was performed on a chopped strand glass fibre-reinforced shape memory polyurethane composites (118) and it was reported that the tensile strength of the SMP specimens with 10, 20 and 30 wt% of the chopped strand glass fibres content increased by 9, 23 and 44% respectively, compared with the pure SMP specimens.

Elastic memory composites (EMCs) are a form of fibre-reinforced SMP composites in which the thermoset SMP matrix possesses the shape memory capability (120). The shape memory mechanism of the material is illustrated in Figure 9 using an EMC tube upon heating. Studies have been performed on various shape memory thermoset resins to be used as matrix (4, 121). However, a significant disadvantage of this material is that the composite is susceptible to delamination at high temperature as the SMP matrix becomes soft. Another problem is that at elevated temperature fibre can freely move within the matrix, leading to the formation of fibre-rich areas which weaken the mechanical performance of the material.

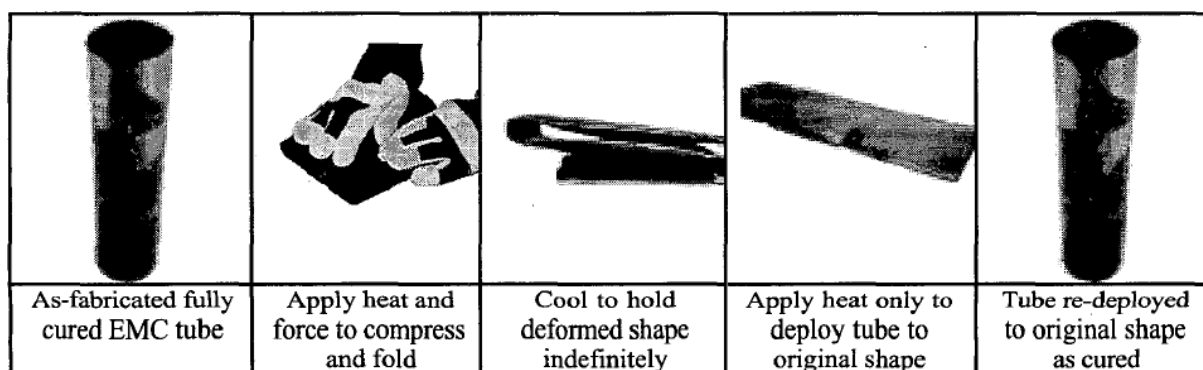


Figure 9. A series of images of the shape memory process of an EMC tube (120).

Composites containing layers of reinforced SMP laminates and SMP films have been developed to demonstrate the shape memory capability. An investigation of the shape memory capability of multi-layered shape memory composites subjected to bending at elevated temperature was performed (82). The composites consisted of carbon fibre-reinforced SMP prepreg and SMP films, arranged in several different configurations as shown in Figure 10a. The composites were bent at elevated temperature and subsequently cooled down to retain the bent shape. Test results from the bend recovery tests showed that all the configurations recovered from a bent shape when heated above the glass transition temperature of the SMP films (see Figure 10b).

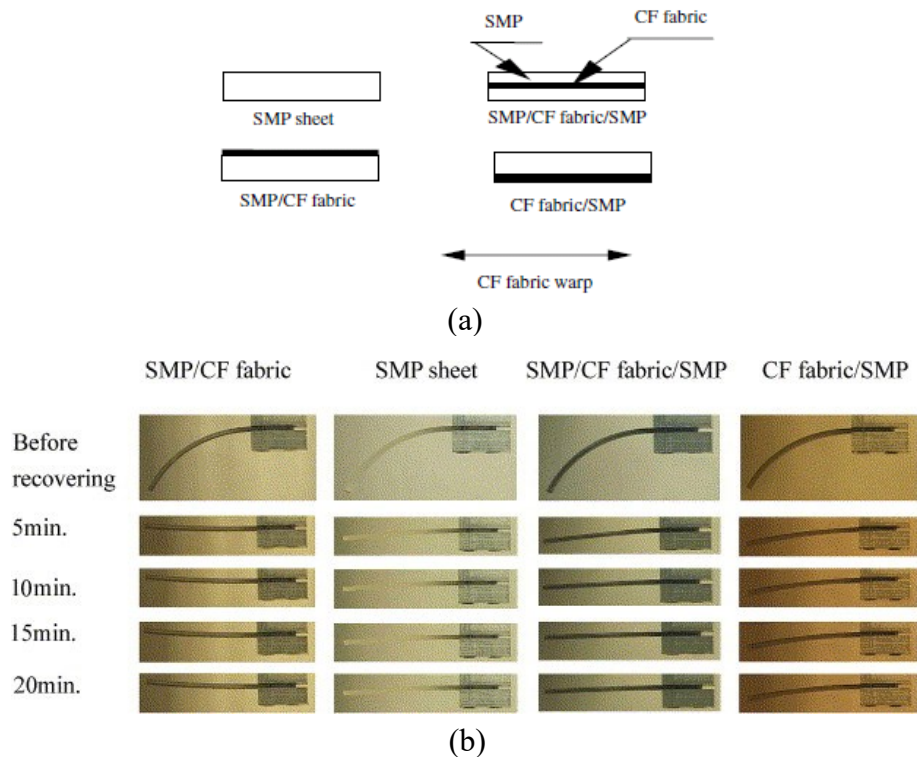


Figure 10. (a) The layup configurations of the multi-layered SMC and (b) the shape recovery tests for the specimens in these configurations (82).

Although reinforced SMP matrix composites exhibit better mechanical properties than neat SMPs, a significant disadvantage of the use of reinforcements in the SMP matrix is that these fillers substantially reduce the recoverable strain and therefore the shape memory performance (117, 122, 123). An investigation on chopped glass fibre-reinforced thermoplastic SMP composites has showed that the material exhibited the shape recovery reduced by 62% than the pure SMP (118). Another problem with such material is that the shape memory capability only takes place when the SMA fibres are heated or the SMP matrix is softened. Typically for fibre-reinforced SMP composites, once at elevated temperature, (i.e. the matrix is softened), the fibre reinforcements could be misaligned as they can freely slide. This can cause the change in local fibre volume fraction, resulting in premature damage. Therefore, SMCs could be a suitable material for deployable structures, provided that the possible fibre misalignment problem can be resolved.

2.3.3.2 Shape memory hybrids

Shape memory hybrids (SMH) are a class of shape memory materials which do not contain any shape memory constituents. SMH consists of a stiff material which provides high elasticity and a temperature-sensitive material which undergoes a stiffness reduction upon heating. Its shape memory effect is similar to that of SMPs where the stiff material represents the elastic segment and the temperature-sensitive material represents the transition segment. Figure 11 shows the shape memory mechanism of a silicon/wax SMH material (52). Here, the silicone matrix is the elastic segment, which retains its stiffness at elevated temperature and the wax inclusions form the temperature-sensitive component and when heated to its melting point will lose their stiffness. At this stage, the SMH can be readily deformed by compression as shown. It retains the compressed shape (with some spring-back) when cooled down to room temperature, as the wax solidifies and regains the stiffness. When re-heated, the SMH returns towards its original shape. A range of materials can be used for SMH materials (52). For example, a SMH material consisting of an elastic sponge with cupric sulfate pentahydrate inclusions was developed to demonstrate the water-activated hole closure process (124). A pre-bent specimen of a silicone/wax SMH material recovered to its flat shape in 15 seconds when placed in water at 45°C (52). The SMH examples mentioned so far are not suitable for structural applications due to their poor mechanical properties.

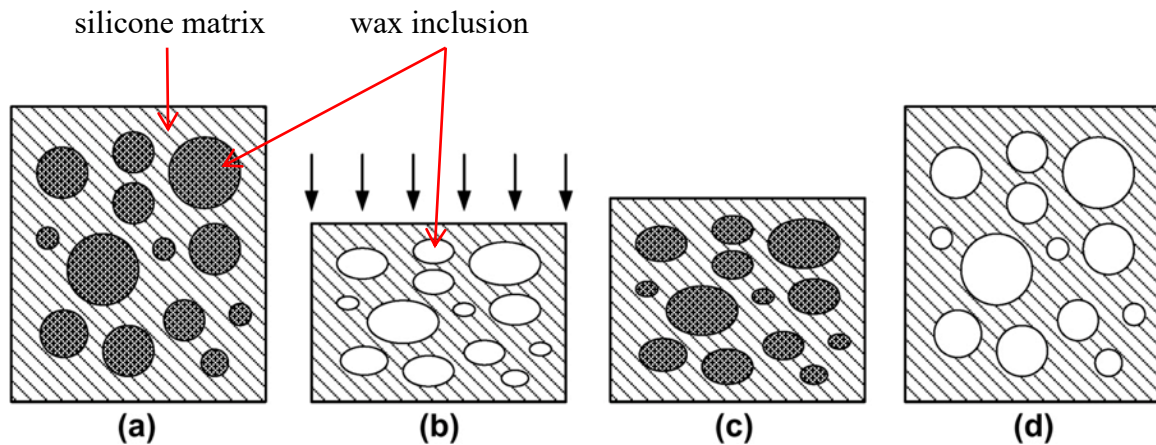


Figure 11. A sketch of the shape memory mechanism of a silicone/wax SMH system at (a) initial shape, (b) compression at elevated temperature, (c) cooling down and (d) shape recovery stages (52).

However, the PS-interleaved carbon fibre/epoxy composite laminate developed at Imperial College London which showed controllable stiffness capability (125) is also a type of SMH materials. In the same study, it has been shown that the displaced shape (see Figure 12) of a PS-interleaved composite laminate subjected to three-point flexure tests at high temperature could be ‘frozen in’ by cooling to room temperature while holding in this displaced shape and the unconstrained specimen would return to its original flat shape when reheated (125). In this case, the carbon fibre/epoxy composites are the stiff material and the PS interleaves are the temperature-sensitive material.



Figure 12. An image of the PS-interleaved flexure specimens in the initial and displaced shapes (125).

This shape-memory capability of PS-interleaved composite laminate was subsequently investigated by the author in his MSc project at Imperial College London (126). In that work, specimens of 90°-bend specimens made of PS-interleaved carbon fibre/epoxy composite laminate were re-shaped to 180°-bend and flat configurations (see Figure 13a). In subsequent shape recovery tests, the specimens recovered to the original, 90°-bend configuration quickly and with almost full recovery in 9 minutes (see Figure 13b).

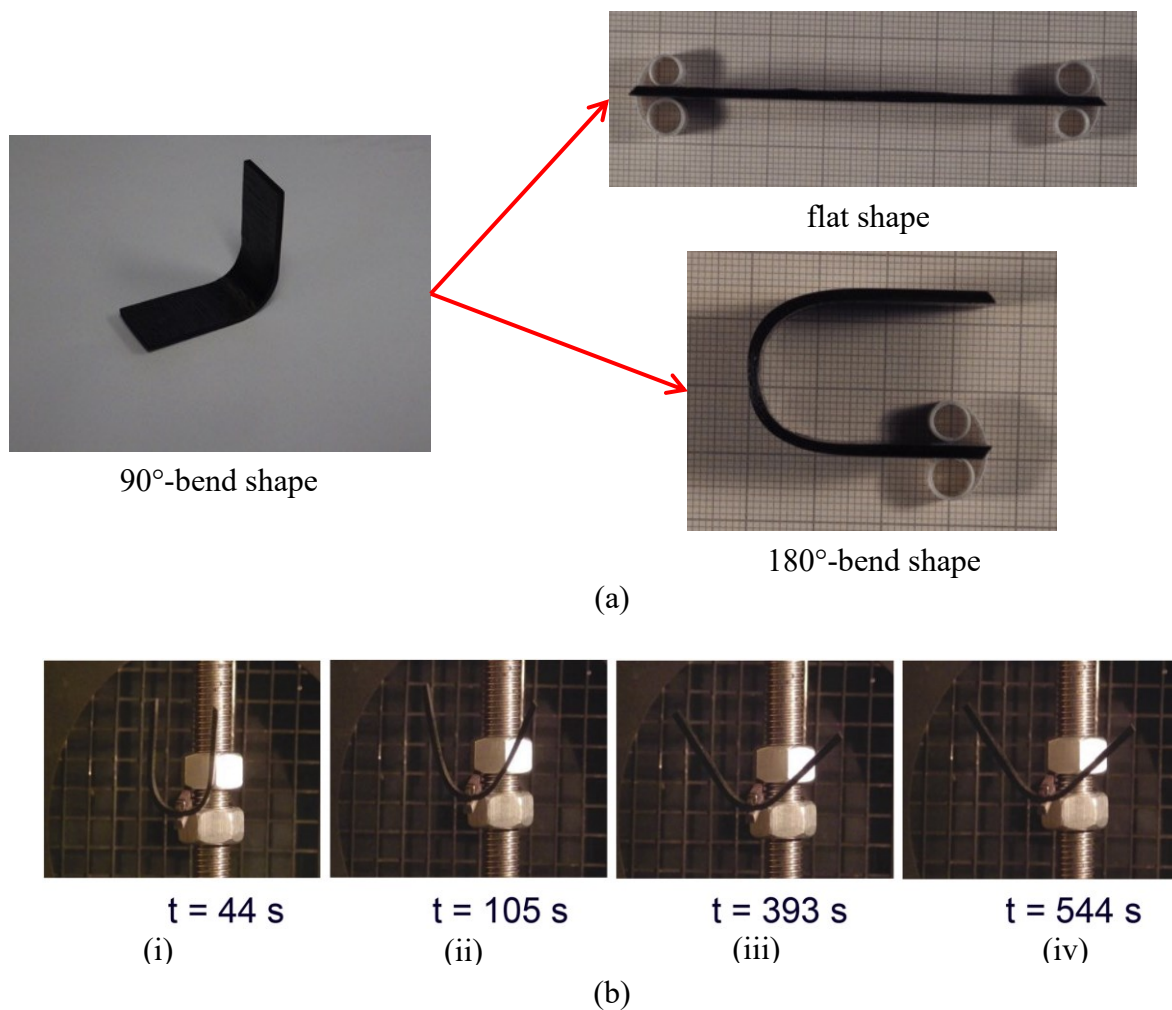


Figure 13. (a) Images of the initial 90°-bend specimen re-shaped to 180°-bend and flat configurations and (b) the shape recovery process of the 180°-bend, quasi-isotropic specimen to the initial 90°-bend configuration (126).

A simple structure consisting of this interleaved composite laminates as hinges and stiff, non-interleaved composites was manufactured (126, 127). The deployment test (Figure 14) showed that

this assembly can deploy in a quick and accurate way and therefore demonstrated the potential of this material for use in deployable structures.

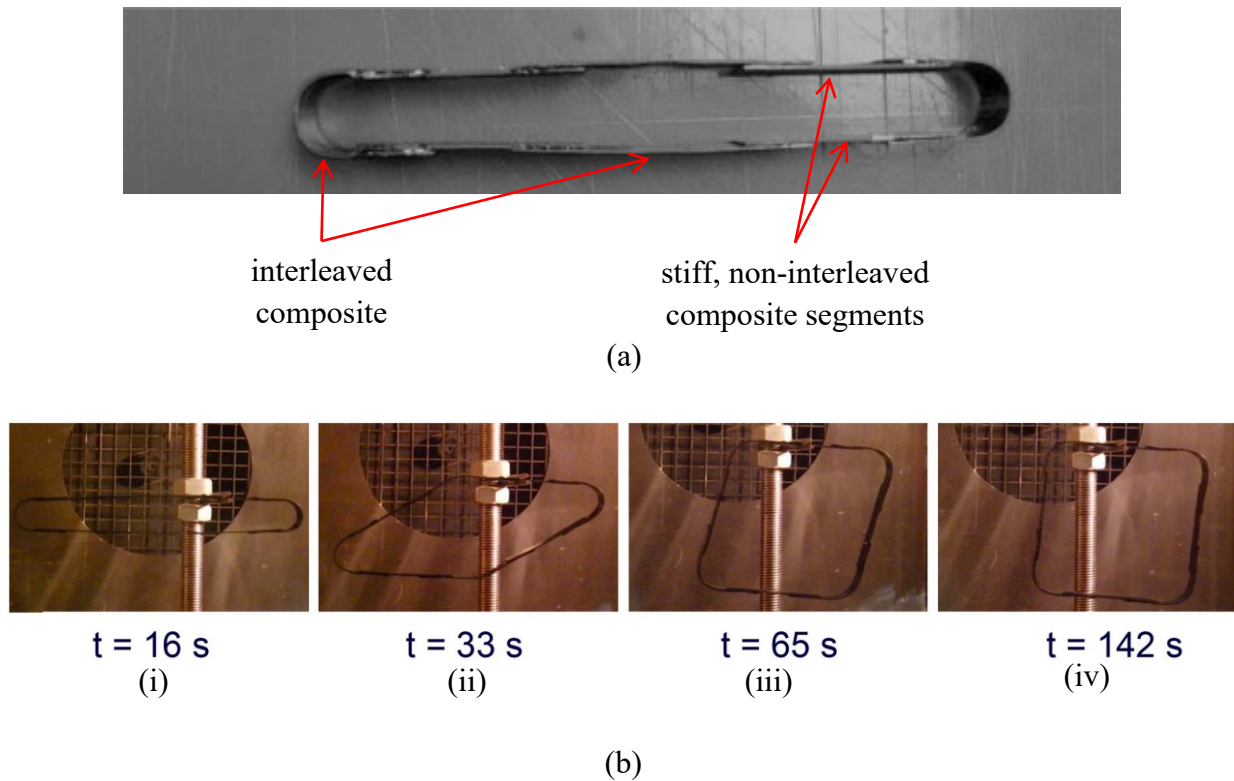


Figure 14. (a) A box section of an interleaved cross-ply laminates and (b) the shape recovery process of this box section (127).

However, one problem with the PS-interleaved composite laminate is that interior composite ply fracture (see Figure 15) was observed after shape recovery of the laminate from 180°-bend and flat configurations (127). This was attributed to the viscoelastic behaviour of the PS interleaves. (The PS-interleaved composite behaved as an integrated structure when re-shaped at a fast speed and so induced higher bending force in the composite plies. With slower re-shaping speed, such damage can be avoided.) Another issue for the PS-interleaved composites is the void formation in the PS layers that can be seen in Figure 15. This was due to the outgassing of the PS at high temperature.

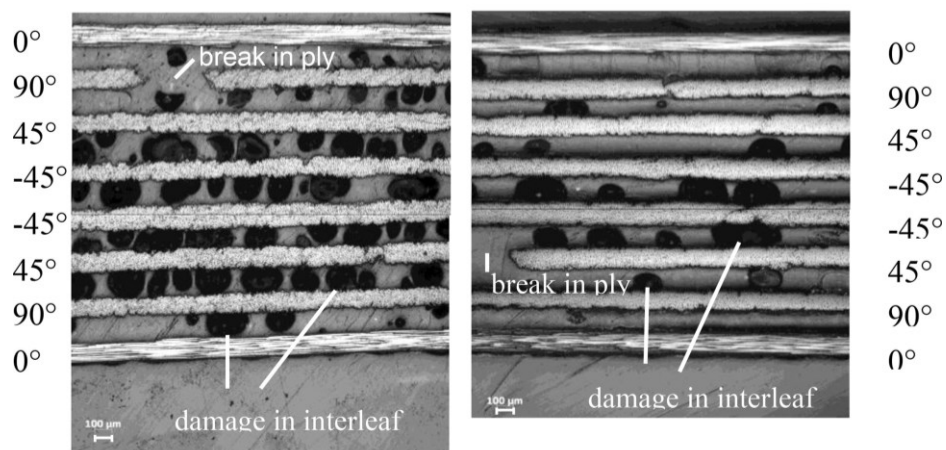


Figure 15. Microscope images of polished cross section of quasi-isotropic interleaved laminates after shape recovery showing interior composite ply damage (127).

2.4 Repair strategies for fibre-reinforced polymer matrix composites

Carbon fibre/epoxy composites have been widely used in high-performance applications as they possess excellent specific stiffness and strength, compared with metallic materials. However, various types of damage can occur in composite laminates, including defects which can arise during manufacture and delamination and fibre breakage due to mechanical overload or impact. Such damage can significantly reduce the structural integrity of composite structures (128). The aim of composite repair is to recover the structural integrity and to restore the strength and stiffness of the material (129). Conventional repair strategies, including patching and resin injection, have been used widely and have shown a good repair efficiency (130). More recently, self-healing composites have received considerable attentions. The following sections will review conventional composite repair techniques and self-healing composites.

2.4.1 Conventional repair approaches

Conventional composite repair methods are frequently based on adhesive bonding. There are four types of bonding-based composite repair, including external/overlap patching, scarf patching, injection repair and cosmetic repair (130). It is known that both external patching and scarf patching

are used to repair structural damage, while injection repair is used to repair disbond and delamination for non-critical structure applications (131). On the other hand, cosmetic repair removes damage such as scratch, dent or missing plies on the composite surface (i.e. to restore surface smoothness), but not to restore material strength (130). Therefore, this section will focus on external patching, scarf patching and injection repair.

2.4.1.1 External patching repair

In the external patching repair technique, damaged material is removed by drilling a hole and then an overlap patch is adhesively bonded over the damaged area on one or both sides, as shown in Figure 16 (132). The patch can be manufactured from uncured prepreg and subsequently co-cured with an adhesive layer in the damaged area, or it can be machined from pre-cured laminates and adhesively bonded into the damaged region (133).

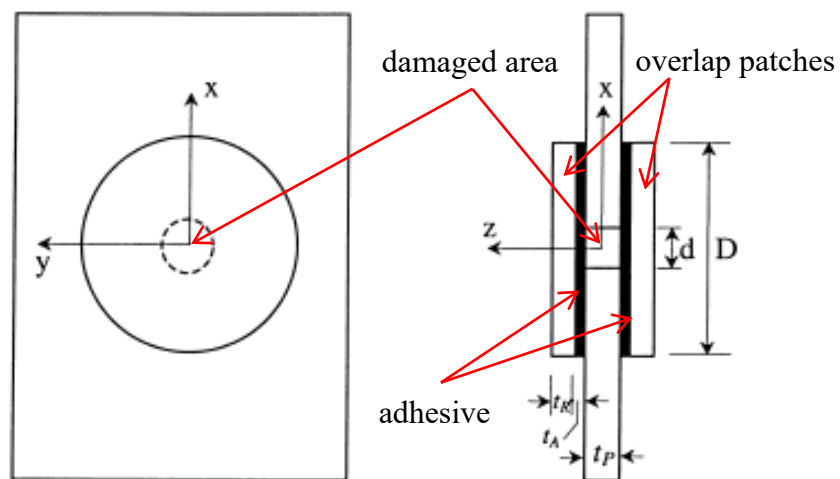


Figure 16. A sketch of external patching repair (132).

Extensive studies have been performed on the investigation of external patching repair for restoring the mechanical properties of laminated composites (131, 133-140). For example, the compressive behaviour of composite laminates repaired by external patching was investigated (131). Round and

square patches in two overlap sizes were considered. Results from double-lap joint tests showed that over 80% of the compressive strength of an unnotched specimen was restored for the larger overlap length. Patch parameters affecting the bonding strength, including overlap length, patch geometry and thickness and adhesive types, were studied in a double-lap joint model using finite element modelling and it was reported that over 90% strength recovery could be achieved for composites repaired with the optimum external patches.

The impact performance of woven glass fibre laminates repaired by external patching technique was investigated (134). Both single and double patching configurations were considered (see Figure 17). Low-velocity impact results showed that for a given 6 J impact energy (this energy was typically selected to induce full perforation of the specimens after multi-impacts), the double-patched specimens showed a 97.1% higher maximum impact load than the single-patched specimens, as the double-patched configuration possessed higher stiffness.

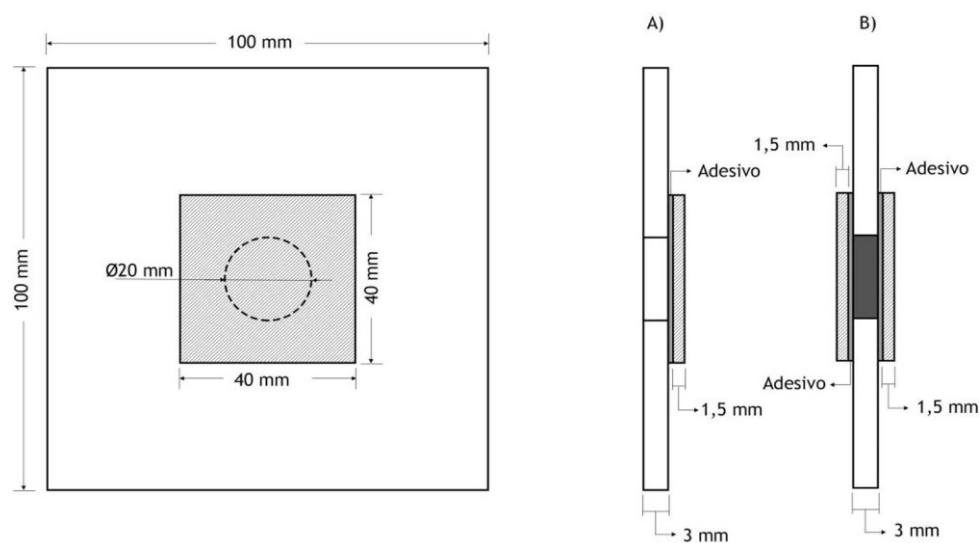


Figure 17. A sketch of the patch geometry and configurations (134).

The tensile behaviour of carbon fibre/epoxy laminates with an open hole repaired by adhesively bonding with composite patches of the same material was examined (133). Two layup configurations ($[+45^\circ/-45^\circ]_s$ and $[90^\circ/0^\circ/-45^\circ/+45^\circ]$) of the composite patches were examined to investigate the effect the layup configurations on the strength recovery. Test results showed that in both cases, the

repaired specimens exhibited a higher failure load than the notched, baseline specimens (80% higher for the $[+45^{\circ}/-45^{\circ}]_s$ patches and 50% higher for the $[90^{\circ}/0^{\circ}/-45^{\circ}/+45^{\circ}]$ patches). In the same study, FE modelling was performed to investigate the optimum external patches. It was reported that with the optimised patches, over 90% of the tensile strength of the parent material can be restored.

Moreover, the compressive strength of an external patch-repaired carbon fibre/epoxy laminate was investigated (135). Compression test results showed that the compressive strength was restored by almost 80% of the pristine value measured from the control specimens.

In addition to evaluating the repair efficiency, the hot/wet environmental effect on room-temperature static and fatigue tensile strength of repaired carbon fibre/epoxy composite laminates by immersing in distilled water at 50°C for up to 16 months was examined (136, 141). It was reported that the soaked repaired specimens recovered 84% of the static tensile strength of the baseline, parent specimens. However, the environmental effect on of fatigue tensile strength of the soaked repair specimens was significant with the loss of 50% of the fatigue strength of the parent specimens.

2.4.1.2 Scarf patch repair

In a scarf patch repair, material in the damaged region is removed in a stepped or tapered shape with a low scarf angle (see Figure 18). (Scarf angle is the angle at which damaged material is removed from the parent material (142).) Then a ‘hard’ patch which is pre-formed with moulds or by machining to match the configuration of the cavity, is adhesively bonded in place or the cavity is filled with an uncured ‘soft’ patch preform which is manufactured by hand layup from prepreg and subsequently co-cured with film adhesive by heating and vacuum pressure (143).

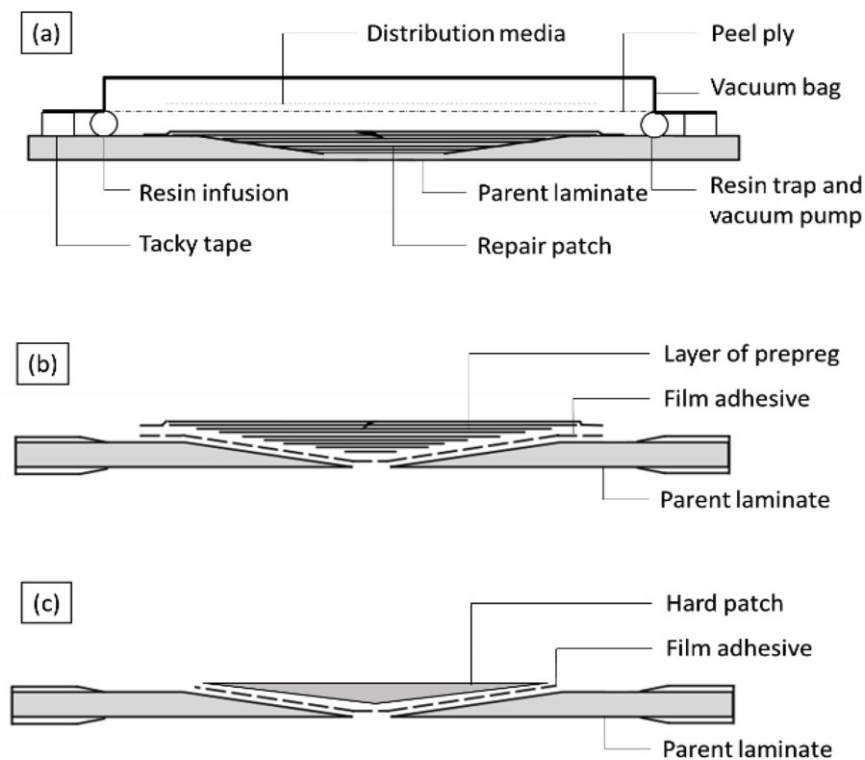


Figure 18. A sketch of scarf patching repair using (a) a stepped patch, (b) a tapered, soft patch and (c) a tapered, hard patch (142).

Studies have been conducted on the scarf patching repair for composites (143). For example, a scarf patching repair strategy was proposed for carbon fibre/epoxy composites with a pre-moulded hard patch (144). A scarf cavity was formed by CNC machining and a mould matching the cavity geometry was subsequently manufactured. A patch of carbon fibre/epoxy composite was then manufactured by manually laying up laminae, followed by consolidation in a vacuum bag. The cured hard patch was sand blasted on the bonding surface and placed into the cavity separated with an adhesive film. The assembly was finally heated under pressure to bond the patch. Unfortunately, no examination of the strength recovery was reported in this work.

Scarf composite patches were also used to repair graphite/epoxy composites (145). Patches of various layup configurations and damaged specimens were co-cured at 350°F (or 177°C) in vacuum for 2 hours, together with the FM 400 adhesive to investigate the hot/wet environmental effect. Results

showed that about 80% of the tensile and compressive strengths of the pristine composite were restored, even under environmental conditioning.

In order to optimise the joint strength of patches and bulk material, studies have also been conducted in both experiments and modelling on the parameters for a successful scarf patch repair. These include scarf angle (146, 147), patch geometry (147, 148), material machining technique and surface preparation (149) and patch stacking sequence (150). For instance, the geometry of stepped patch joints of a carbon fibre/epoxy composite was investigated (148). In this study, composite laminae were removed from specimens by machining in the step shape at an angle of 3° and the machined specimens were adhesively bonded with stepped patches of the same material with sharp (square) or blunt (round) corners. The patched specimens were subjected to anti-buckling compression tests and results showed that the specimens containing square-edge patches exhibited a higher failure load (16 kN) than those with round-edge patches (14 kN), as the sharp corners had a slightly higher bonding strength than the round corners.

2.4.1.3 Injection repair

In an injection repair (see Figure 19), a room temperature-curing resin of low viscosity is injected into the delaminated area of composites (149). An inlet hole is drilled at the centre of the delaminated area and some vent holes are drilled around the delaminated area, to inject and eject the resin. Air blasting and acetone rinsing are used to remove the scattered piece of dust from the delaminated area and holes. After this, resin is injected into the inlet hole in vacuum. (The vacuum assists the resin to fill the delaminated area.) The resin is subsequently consolidated to repair the delaminated area.

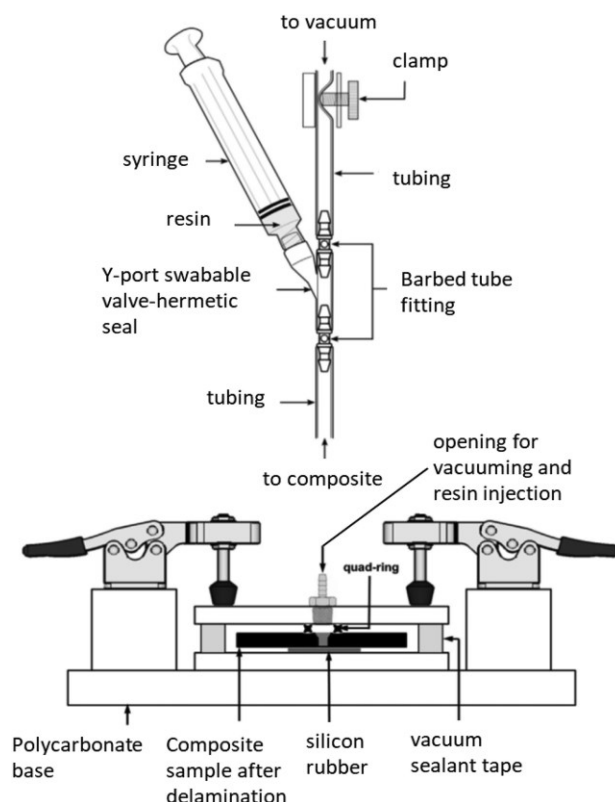


Figure 19. A sketch of the setup for injection repair (149).

Successful injection repair is very dependent on the selected resin. It should possess low viscosity and a high cured glass transition temperature (151-153). Resins used for injection repair include epoxy, cyanate ester and cyanoacrylate. Epoxy can provide good mechanical properties, as well as low cost. Cyanate ester, once cured, exhibits high cured glass transition temperature (220-270°C) and this allows the material to be used for high temperature applications (151). Bisphenol E cyanate ester (BECy) is a type of cyanate ester which possesses low viscosity at room temperature, high cured glass transition temperature, ease of processing and better mechanical properties than epoxy (152). Ethylcyanoacrylate (ECA), a type of cyanoacrylate, can also be used for injection repairs as this material can polymerise rapidly and readily in cured epoxy with the use of various mediums (154), but it shows thermal instability over 100°C and therefore cannot be used for high temperature applications (153).

Research has shown that injection repair can achieve good compressive strength recovery for fibre-reinforced composites (151-156). For example, the repair efficiency of carbon fibre/bismaleimide composite panels repaired with a BECy resin was investigated. Specimens were subjected to either static indentation or impact tests, then repaired by injecting the repair resin and finally re-tested in the static indentation test when the resin was cured. Results showed that the repaired specimens, which were initially subjected to static indentation tests, restored an average 91% of the overall stiffness of the pristine specimens, whereas the repaired specimens initially underwent impact tests restored the stiffness by 76% on average (155). The difference in repair efficiency was because discrete delamination occurred in the impacted specimens which the repair resin could not reach. Moreover, the flexural strength recovery of a carbon fibre/epoxy laminate by the injection repair with a low viscosity cyanoacrylate resin was investigated (156). In this study, specimens were initially impacted at an impact energy of 20 J, then repaired by infiltrating the repair resin and the repaired specimen were subjected to four-point flexure tests. Test results showed that repaired laminates almost fully restored the flexural strength.

2.4.1.4 Summary of conventional composite repair methods

Although conventional repair methods can achieve a good strength recovery efficiency, there are concerns with these methods. Despite a relatively easy, quick and low cost repair process (132), external patching repair requests well design of repair patches (e.g. repair patches with tapered edges to reduce peel stress (132)), adhesive types to achieve good patch bonding and the need for skilled operators (132, 137, 157).

Scarf patch repair achieves good surface smoothness and higher strength recovery (149) but requires extensive, careful surface preparation of the damaged area to achieve the low scarf angle for good strength recovery, correct fibre orientation of the patch to match that of the parent material (131) and the use of special facilities and equipment such as Computerised Numerical Control (CNC)

machines. Other issues of concern include wrinkling and ply distortion during co-cure of soft patches, the high cost of patch materials and the need for experienced operators.

In injection repair, the drilled holes can result in stress concentration (158, 159). A drilled hole should be well designed by considering cutting tool geometry and materials, feeding rate and cutting speed (160-164). Moreover, the injection repair technique cannot repair damage such as fibre breakage due to impact, which limits the use for critical structure applications (165). Other problems include the selected resin to have a reasonable curing time for a quick repair and the complexity of operation which involves the use of special facilities and equipment.

2.4.2 Self-healing composites

The concept of self-healing fibre-reinforced composites was motivated by the difficulties associated with the conventional repair techniques and was inspired by biological systems in which self-activated healing process occurs in response to damage whenever it takes place. To date, the most common healing mechanisms in fibre-reinforced composites are capsule-based, vascular (hollow tubes) and intrinsic healing methods (see Figure 20).

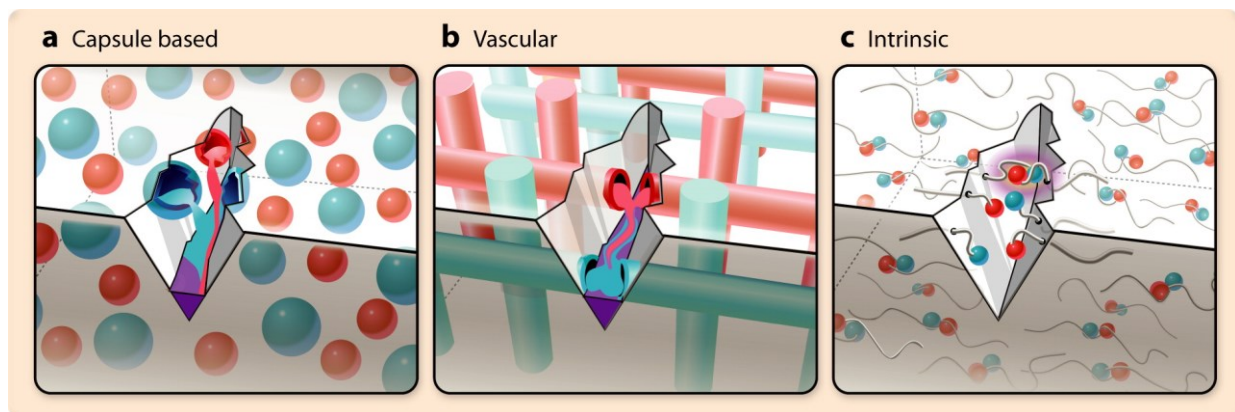


Figure 20. Images of (a) the capsule-based, (b) the vascular and (c) the intrinsic healing approaches, with the demonstration of the process of each method (6).

2.4.2.1 Capsule-based healing method

In the capsule-based approach (see Figure 20a), isolated capsules containing healing agent and hardener are embedded into a composite system. When these capsules are ruptured due to cracks, the healing agent and the hardener are released into the damaged region and react to trigger the local healing treatment (6). One drawback is that this cannot be repeated if the same area is subsequently damaged again since the healing agent and the hardener cannot be refilled (21, 166). Also, the use of microcapsules in the matrix may compromise the mechanical strength of composites.

A plain weave carbon fabric-reinforced epoxy composite containing microcapsules was proposed to exhibit the self-healing ability (8). The healing agent of dicyclopentadiene (DCPD) monomer was encapsulated in poly(urea-formaldehyde) microcapsules by in-situ polymerisation. Specimens of this material were subjected to mode I double cantilever beam test. As shown in Figure 21, during the test, the healing agent was released from the microcapsules that were ruptured by a propagating crack. The healing agent was then reacted with the bis(tricyclohexylphosphine) benzyldiene ruthenium (IV) dichloride catalyst (Grubbs catalyst), dispersed within the matrix (rather than contained in separate capsules), to generate a tough, highly cross-linked polymer to fill the cracks in the damage zone. Upon re-testing the healed specimens and measuring the interlaminar fracture toughness, an average 38% healing efficiency of specimens healed at room temperature was observed, whereas a higher value of 66% when the specimens were healed at elevated temperature of 80°C.

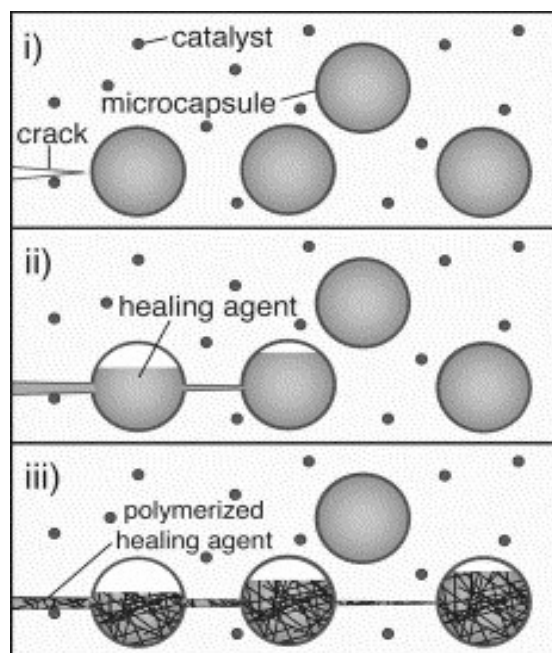


Figure 21. A diagram showing the mechanism of the capsule-based healing treatment to suppress a propagating crack (8).

A self-sealing E-glass fibre/epoxy woven composite was developed (9). Microcapsules containing the DCPD healing agent and wax-coated Grubb's catalyst microspheres were distributed inside the matrix. Specimens of this material were loaded in cyclic indentation tests to rupture the microcapsules. The healing mechanism was triggered by the reaction of the healing agent released from the ruptured microcapsules and the catalyst to fill the damage. A healing efficiency of 67% was recorded by quantifying the resulting Nitrogen flow through the cross section of the healed specimens clamped in a pressure cell machine. In another investigation of healing a plain weave glass fibre-reinforced epoxy composite using DCPD microcapsules and catalyst microspheres, specimens were subjected to low-velocity impact and compression-after-impact (CAI) tests (10). Impact results revealed that for a given impact energy (45.1 J), the total crack length per imaged edge of the self-healed specimens was 51% smaller than that of non-self-healing, control specimens containing the capsules and the catalyst. It was showed that the recovered compressive strength after impact of the specimens healed under no pressure and under lateral pressure was 68% and 78% respectively of the pristine compressive strength.

2.4.2.2 Vascular healing method

An alternative to the capsule-based concept is to deploy an interconnected network of vasculatures that sequester healing agent in one, two or three dimensions within the bulk matrix (6). The vasculatures can be either capillaries or hollow vascular sections. As shown in Figure 20b, when the vasculatures are ruptured, from which the healing agent and hardener are released, the healing mechanism is initiated. A significant feature of this approach is that multiple healing actions can be achieved as the vasculature can be refilled from remote sources or other regions inside the network (6).

An E-glass/epoxy composite containing an interconnected network of hollow glass fibres was proposed (167). This composite was stacked in the $([90/0]_{\text{solid}}, [90/0/90/0]_{\text{hollow}}, [90/0/90]_{\text{solid}})_S$ sequence. ('Solid' indicates that the fibres had a solid cross-section; 'hollow' indicates the fibres were hollow tubes.) Healing agent of an uncured modified epoxy resin was injected into the 0° hollow glass fibres, whereas the 90° hollow glass tubes were filled with the corresponding hardener. Figure 22a demonstrates the configuration of the material. Specimens were subjected to indentation tests to rupture the hollow glass fibres. Both the released healing agent and the hardener reacted in the damage region and healed cracks. The healed specimens were subsequently loaded in a four-point bending test to measure the flexural strength, in order to characterise the healing efficiency of this approach. Experimental results in Figure 22b show that the flexural strength of the healed specimens was significantly recovered to around 600 MPa, which was 97% of the pristine value of an undamaged specimen (167).

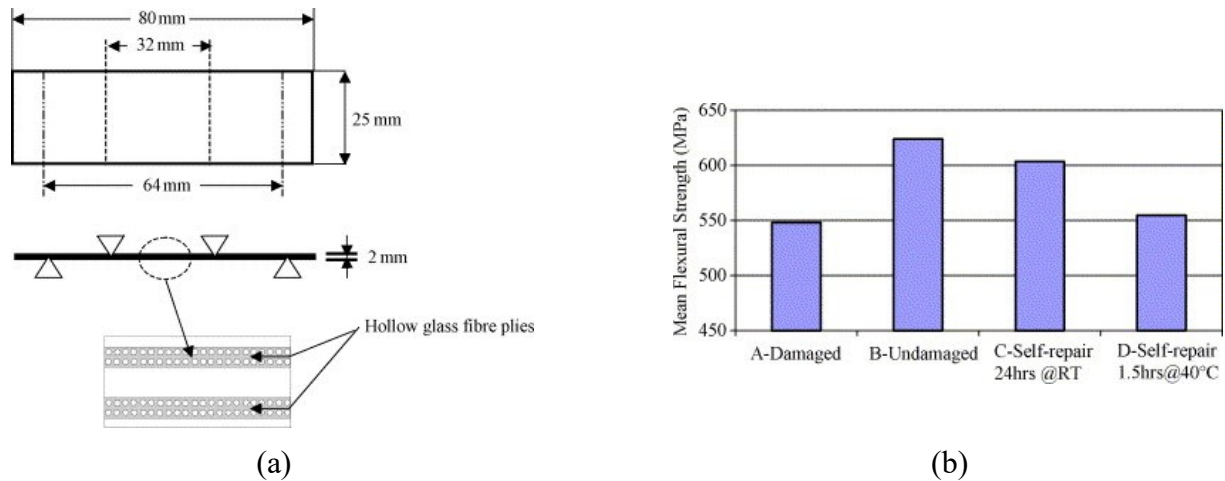


Figure 22. The diagrams of (a) the configuration of the material described here under a four-point flexure test (Only half of the layup configuration is showed.) and (b) the results of the average flexural strength of the composites under different conditions (167).

An examination of the healing efficiency of carbon fibre/epoxy laminates with resin-filled hollow glass fibres (HGF) was conducted (16). In their work, discrete hollow glass fibres were distributed in carbon fibre/epoxy prepreg at the interlaminar region of the 0° and the -45° plies, as shown in Figure 23. A healing medium of a two-part epoxy resin (Cytec Cycom 823) was injected into the HGF tubes, together with a UV fluorescent dye to highlight the healing process. Specimens initially were subjected to a quasi-static impact to trigger the healing process, followed by a four-point flexure test to measure the recovery of the flexural strength. Results from the flexure test showed a remarkable recovery of the flexural strength of the healed specimens, which was equivalent to 90% of the original flexural strength of an undamaged specimen.

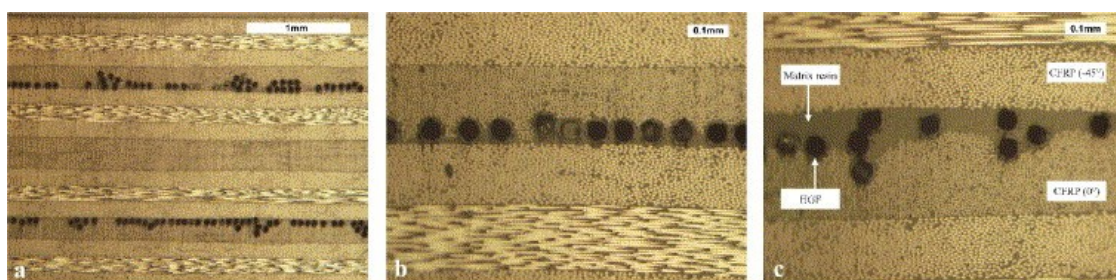


Figure 23. The cross section of a specimen with hollow glass fibres deployed at the interlaminar region between the 0° and -45° plies (16).

2.4.2.3 Intrinsic healing method

A third approach to self repair of damages in composites is to use a material which contains reactants capable of chemical bonding to heal the damage (6) (Figure 20c). These reactants are previously dispersed into the matrix when manufacturing the composite. This approach avoids the necessity of incorporating microcapsules or vascular networks.

A composite consisted of a thermally mendable polymer (Mendomer 401) reinforced by carbon fabrics was manufactured to examine the self-repair capability (168). Two chemical substances, Diels-Alder (DA) and retro-Diels-Alder (rDA), were sequestered within the mendable polymer matrix. Specimens of this material were subjected to a three-point bending test to induce microcracks in the matrix, followed by heating to 150°C for one minute using an electrical resistance heater attached on the upper surface of specimens. The healing mechanism was triggered by the reaction of Diels-Alder and retro-Diels-Alder. The efficiency of the healing approach was evaluated by measuring the strain energy at 14 mm mid-span displacement of a specimen in the three-point bending test for four times. They observed a generally good agreement among the load-displacement plots, with an overall 7% loss in the strain energy at 14 mm mid-span displacement.

A glass fibre-reinforced composite with the thermally-activated solid-state self-healing ability was developed (169). The thermosetting matrix of the composite was mixed with a thermoplastic polymer of polybisphenol-A-co-epichloro-hydrin at 10 wt%. This dissolvable thermoplastic material becomes soft and mobile at elevated temperature. Specimens of this composite were impacted in a drop-weight impact test and then heated to 130°C for one hour. The impact damage region was photographed before and after the heat treatment. Through the comparison of the images shown in Figure 24a, a significant reduction by 30% in the visible damage area was measured. In their second study (170), the multiple healing capability of this composite was examined by repeatedly loading a specimen in the same test protocol for three times. The experimental results revealed that the

multiple healing was feasible, with a healing effectiveness of approximately 30% after three impacts (Figure 24b).

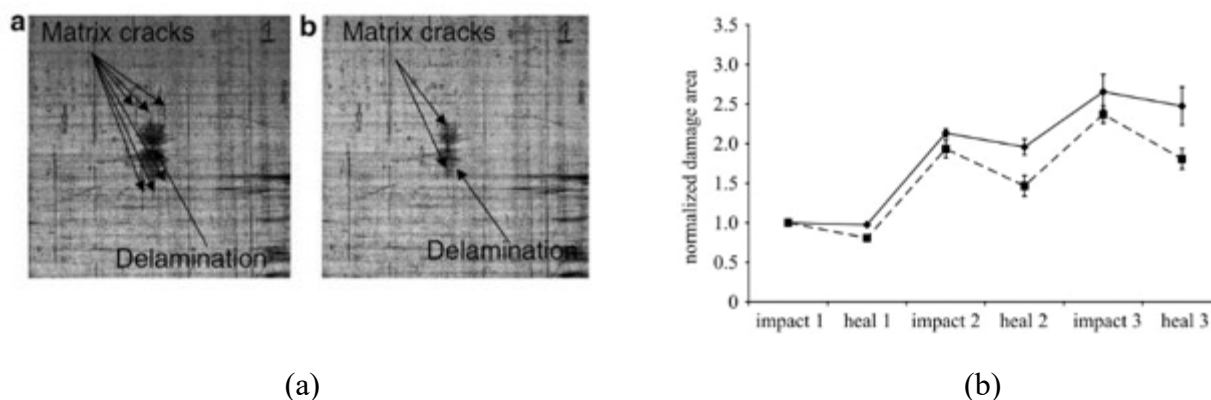


Figure 24. Figures of (a) a reduction in the damage area after healing (169) and (b) specimens with the healable matrix (dash line) and the normal matrix (solid line) at different times of impact (170).

A CFRP laminate containing mendable poly[ethylene-*co*-(methacrylic acid)] (EMAA) particles in the matrix was developed (171, 172). Double cantilever beam (DCB) specimens of this material were tested, then heated at 150°C for 30 minutes without the application of pressure, followed by further DCB testing to measure the healing effectiveness. The experimental results demonstrated that although the fracture toughness G_{Ic} after the first healing achieved in excess of 200% of that of the control specimen (not containing EMAA particles), subsequent healing showed that the repair efficiency reduced to around 120% due to the depletion of the function group of EMAA to re-bind the fracture surface.

A self-healing matrix consisting of a blend of diglycidyl ether bisphenol A (DGEBA) epoxy resin with 4,4'-diaminodiphenylsulfone hardener and poly(ϵ -caprolactone) (PCL) was developed (173, 174). This self-healing matrix was used to manufacture glass fibre reinforced composites (see Figure 25). Specimens of this composite were subjected to DCB tests, then heated at 150°C for 30 minutes with a compressive load of 0.39 kg perpendicular to the crack faces and re-tested in the DCB configuration. Three damage-repair cycles were performed to evaluate the multiple repair efficiency. DCB test results showed that the toughness recovery of the specimens after three cycles was between

40% to 60%. However, with the addition of PCL, the toughness was reduced by 58% of the pristine value of neat glass fibre/epoxy specimens.

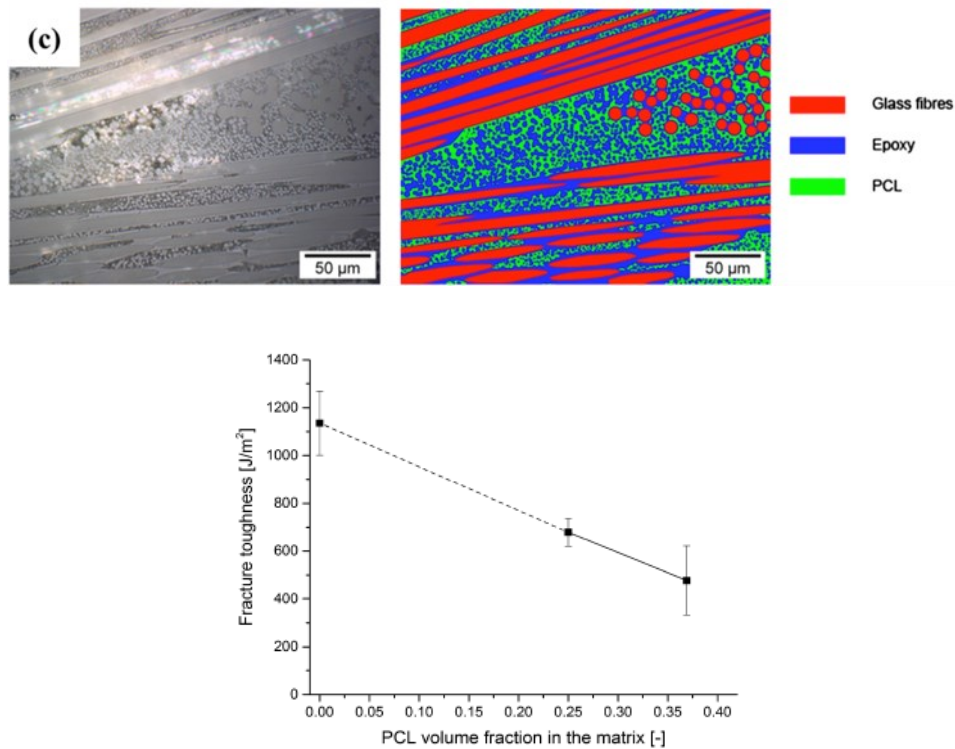


Figure 25. The glass fibre/self-healing matrix with PCL composite (top) and a plot of fracture toughness versus PCL volume content in matrix (bottom) (174).

Another type of intrinsic repair approach was heating elements embedded within a composite to join fracture surfaces by resistance welding (175). The heating elements, usually connected to external electrical circuits, generate heat to melt and then to bond the fracture surfaces of composites. For example, a glass fibre reinforced thermoplastic composite implanted with stainless steel mesh was developed to investigate the potential of resistance welding (176). A stainless steel mesh acting as a heating element was sandwiched in between two composite laminates. With the application of pressure and electricity, the thermoplastic matrix of the two laminates was softened and laminates bonded together. Results from lap shear tests on the bonded specimens showed that the shear strength was in the range of 27-30 MPa. However, this paper did not investigate the repair efficiency.

The resistance welding of carbon fibre/thermoplastic to carbon fibre/thermoset laminates using heating elements was also investigated (177). The bonded specimens were subjected to lap shear testing and it was reported that the shear strength was in the range of 10-25 MPa. Again, the repair efficiency was not considered. Resistance welding does have potential for the repair of interlaminar damage in thermoplastic matrix composites (and for bonding of thermoplastic to thermoset composites). However the inclusion of heating elements can reduce structural efficiency and integrity.

2.4.2.4 Summary of self-healing composites

Although the microcapsule and vascular self-healing approaches do avoid the complexity of the traditional repair strategies, the incorporation of such systems within the composite inevitably introduces additional complications (particularly the vascular systems) and the microcapsules and vascular network can reduce the stiffness and strength of the composite (10, 16). Also, for composite with embedded microcapsules, once self-healing takes place at one position, subsequent healing cannot occur at the same position (21). In contrast, although multiple healing can be performed on composite with a vascular network, the healing efficiency is reduced significantly after multiple cycles (20). Therefore, the self-healing composites may not be ideal for practical applications where both high mechanical properties and good repair efficiency are required. In this thesis, an interleaved carbon fibre/epoxy composite is proposed which has the potential for relatively easy repair.

Chapter 3 Investigation of interleaved carbon fibre/epoxy composites with shape memory capability for deployable structures

3.1 Introduction

In space engineering, some components are required to be stowed in a highly compact volume on Earth for launch vehicles and when delivered, they can deploy to their full, functional configurations (178). Materials for the use in this application should possess excellent flexibility and high shape formability (4). Shape memory materials, including SMAs, SMPs and SMCs, have been employed for deployable antenna (179), deployable booms and trusses (4, 180) and other deployable components (178, 181). SMAs possess good shape memory capability, but they also have several disadvantages including high material and processing costs, high density and limited recoverable strain (112). SMPs possess remarkable recoverable strain (82), but their mechanical strength and stiffness are insufficient for practical applications. To address this problem, fibre-reinforced shape memory polymer composites have been developed, but the fibre reinforcement, although it does enhance the mechanical performance of shape memory polymers, significantly reduces the shape memory effect (120, 182).

In comparison, the interleaved composites in the current study provide both the excellent shape memory capability and good mechanical strength and stiffness. The interleaf material is selected so that at a specific elevated temperature, it softens significantly. At this temperature, the composite laminae can slide easily relative to each other and consequently the interleaved composite exhibits a greatly reduced flexural stiffness. The shape memory capability of the interleaved composite does

not arise from shape memory properties in the constituent composite laminae or the interleaf material (it is a type of shape memory hybrids as described in Chapter 2).

To illustrate the source of the shape memory capability a simple interleaved laminate shown in Figure 26 will be used. This consists of two carbon fibre/epoxy laminae separated by a thermoplastic interleaf layer. When heated to an elevated temperature above the glass transition temperature (T_g) of the interleaf layer, the stiffness of the thermoplastic interleaf is significantly reduced. At this temperature, the laminate can be easily re-shaped by bending as shown. This bent shape is retained if the laminate is held at this shape while cooling down to below the T_g of the interleaf material, as the flexural stiffness of the laminate is now restored. In this new shape, elastic stresses are stored within the composite laminae. Upon re-heating, the laminate recovers towards its original shape, driven by the stored elastic stresses which is released, as the thermoplastic interleaf layer is again at its low stiffness stage.

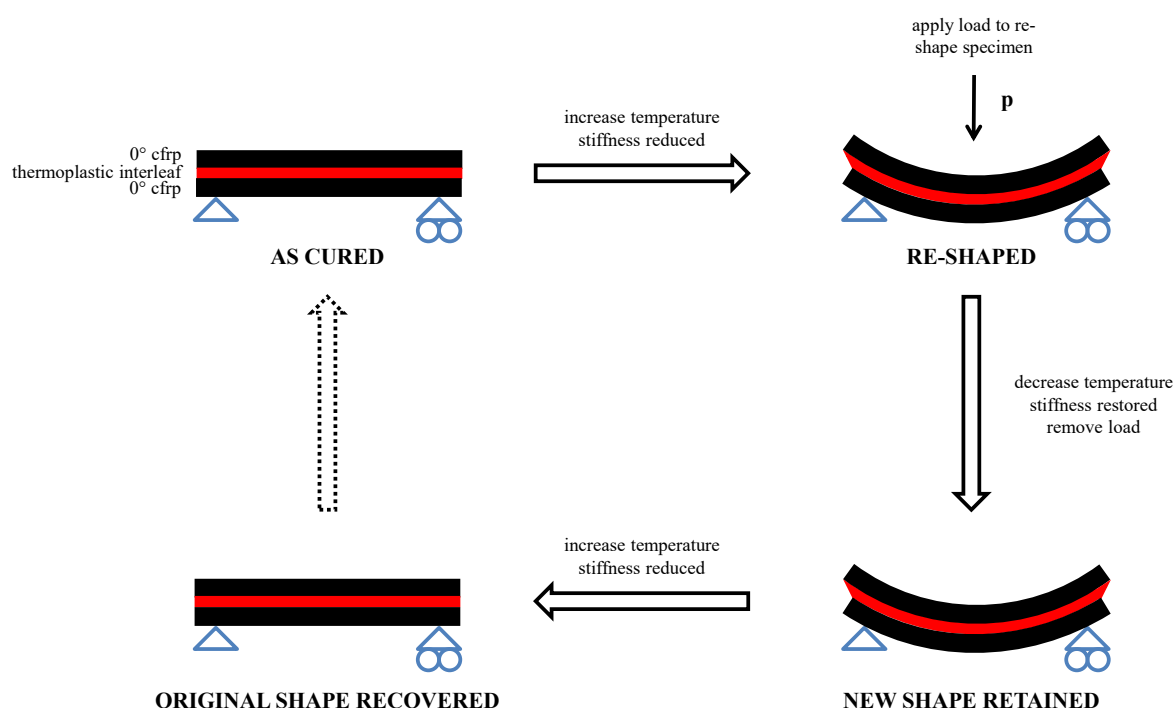


Figure 26. A sketch of a three-ply interleaved composite to demonstrate the source of the shape memory effect on heating.

In the previous work, 90°-bend specimens of interleaved composite were re-shaped to 180°-bend and to flat configurations and good recovery to the original, 90°-bend shape was observed on re-heating. Using the re-shaped 180°-bend and flat configurations a ‘collapsed’ box section was manufactured to explore the potential of this material for deployable structures (127). Figure 27 shows that the ‘collapsed’ box section deployed to the square shape in under 150 seconds.

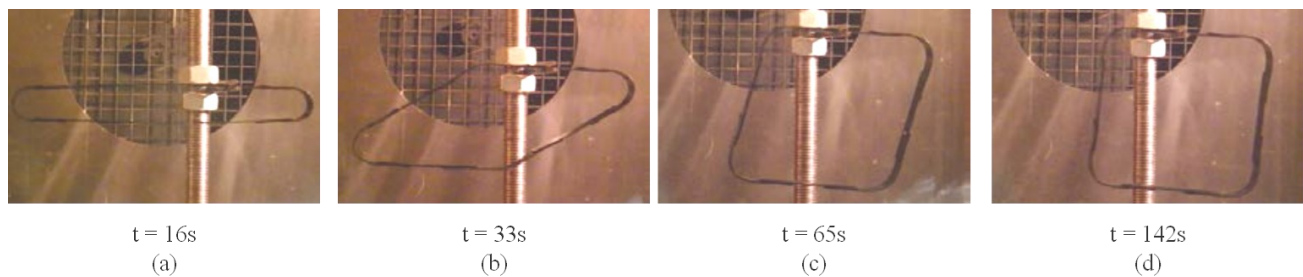


Figure 27. Still images of the deployment of the 'collapsed' box section to the square shape (127).

This chapter describes investigations into three deployable configurations using thermoplastic-interleaved carbon fibre/epoxy composites:

- a deployable composite structure with integrated interleaved composite hinges

This work was conducted in conjunction with D. Smaradhana, an MSc project student. The current author assisted in the supervision and helped in the manufacture and experiments. The work was reported in an MSc Project report (183).

- an interleaved composite configuration enabling linear actuation

This work was conducted in conjunction with W. Li, an MSc project student. The current author assisted in the supervision and helped in the manufacture and experiments. The work was reported in an MSc Project report (184).

- an expanded mesh made of interleaved composite

3.2 Design, manufacture and evaluation of deployable structures

3.2.1 Materials

A unidirectional T300 high-tensile, surface-treated carbon fibre/epoxy 914 composite with fibre volume fraction of 60% [HexPly, FIBREDUX 914C-TS-5-34% with a cured ply thickness of 0.125 mm and a curing temperature of 175°C, purchased in prepreg form from Hexcel, UK] was used in the current work. Polystyrene (PS) [Styrolution PS 124N/L, supplied as 0.1 mm thick films from TCKT, Austria] was selected as the interleaf material. (These materials had been successfully used in previous studies on the shape memory capability of interleaved composites (127).) The T_g of the PS is 85°C measured from DSC and that of the cured epoxy matrix is 180°C from Hexcel datasheet. The basic mechanical properties of these materials are shown in Table 1.

Table 1. The room-temperature properties of the materials used in this study.

Materials	Property	Value
HexPly carbon fibre/epoxy composite	E_{11} , GPa	135 ^a
	E_{22} , GPa	8.5 ^a
	ν_{12}	0.32 ^b
Styrolution 124N/L PS film (25°C)	E , GPa	3.2 ^c
	ν	0.35 ^d

^a both E_{11} and E_{22} of carbon fibre/epoxy composite were from Hexcel datasheet

^b ν_{12} of carbon fibre/epoxy composite were from (185)

^c E of PS film at room temperature was from Styrolution datasheet

^d ν of PS film was from the website (<http://polymerdatabase.com/polymer%20physics/Poisson%20Table.html>) [Accessed date: 12.03.2018]

3.2.2 Integral interleaved composite hinge investigation

3.2.2.1 Design concept and development trials

As mentioned earlier, a deployable box section had been previously manufactured, consisting of non-interleaved carbon/epoxy straight, side elements with adhesively bonded interleaved composite ‘hinges’ forming the corners. This showed the potential of the interleaved composites for deployable structures (127). However adhesively bonding the interleaved composite hinges is not ideal as it adds

an extra stage of complexity during the manufacturing process. Moreover, any excess adhesive which cures on the edge of an interleaved ‘hinge’ can inhibit the deployment process by preventing sliding of the carbon fibre/epoxy composite plies.

To address this problem, a composite structure with integrated interleaved composite hinges was investigated. Figure 28 shows an initial concept which consisted of two conventional composite laminate segments and an interleaved zone. The edge view shows there are seven composite laminae in the outer segments. Four of these laminae are continuous through the interleaved zone but the remaining three plies (between the four continuous plies) are stopped at the junction with the interleaved zone and replaced by the interleaf material.

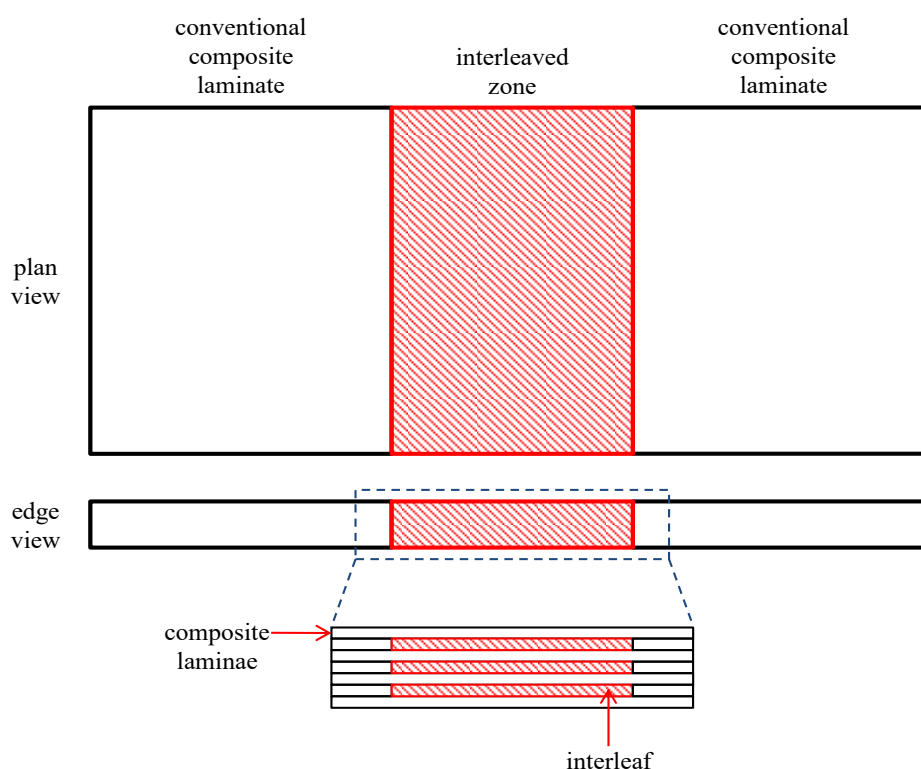


Figure 28. A sketch of a composite panel containing an interleaved zone. (White layers are composites and red layers represent interleaves.)

To reduce weight, it is envisaged that the interleaved zone could be machined to form three individual strips (see Figure 29). These strips would act as hinges to deploy the composite panel at elevated temperature.

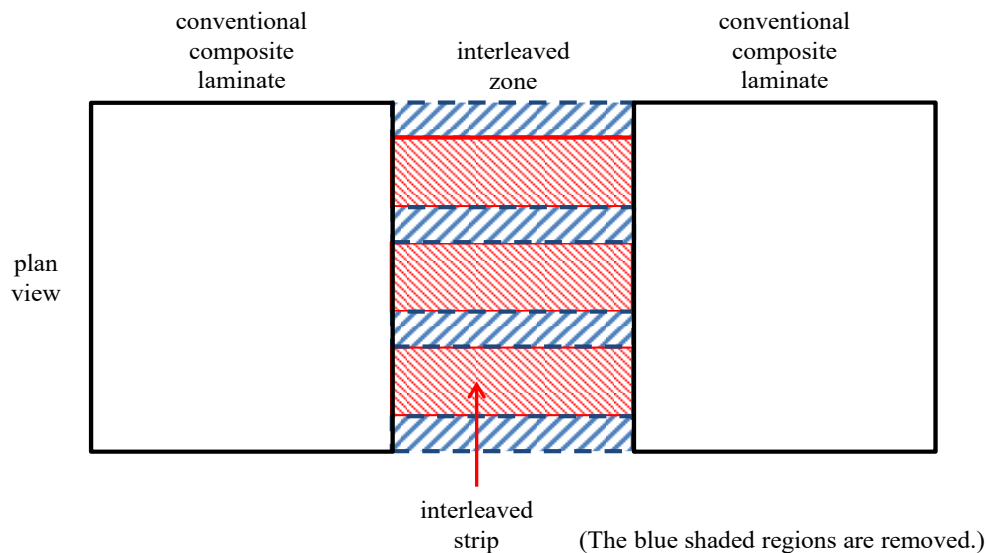


Figure 29. A sketch of the composite panel after machining at the interleaved zone.

However, there is a problem with this integrated hinge concept; it is not possible for the composite plies to slide relative to each other as the hinge is folded. This is illustrated in Figure 30 which shows that the conventional composite laminate segments prevent the composite laminae in the interleaved zone from sliding and so re-shaping of the hinge zone is not possible without damage to the composite laminae.

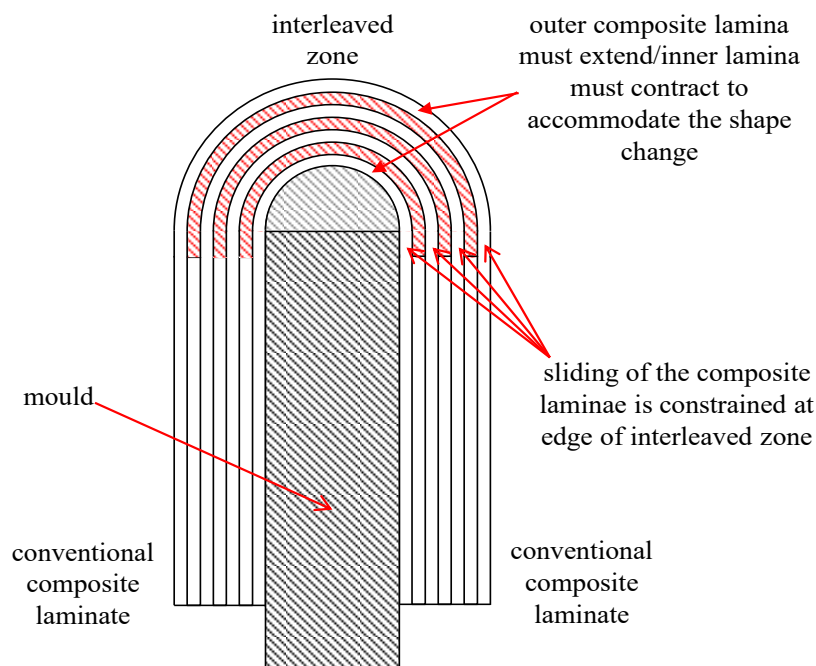


Figure 30. A sketch of the 180°-bend shape of a composite panel with integrated interleaved composite hinges. (White layers are composites and red layers represent interleaves.)

It is therefore important to incorporate some provision to ensure that the composite plies in the interleaved zone can slide during re-shaping at elevated temperature. Two trial configurations were investigated and both involved cutting three of the four continuous plies in the interleaved zone. Both trial configurations consisted of modifications to the base configuration shown in Figure 31. The base configuration had a central interleaved zone, the edges of which are constrained by non-interleaved unidirectional composite laminate zones. This base configuration is representative of the integral hinge concept shown earlier in Figure 28 and Figure 29.

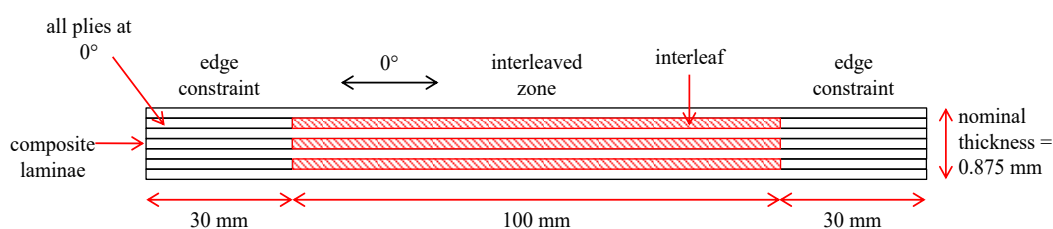
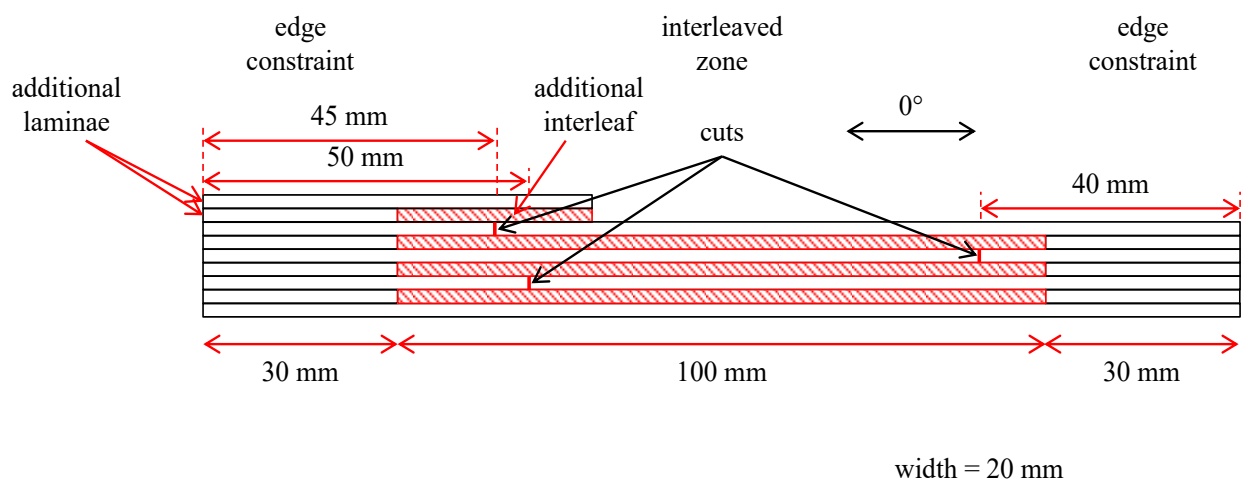


Figure 31. A sketch of the base configuration of an interleaved laminate hinge with edge constraints. (White layers are composites and red layers represent interleaves.)

In the first trial, cuts were introduced in the upper three composite plies (see Figure 32a). The bottom composite ply was not cut (i.e. this ply remained continuous). The lower surface was to be placed on the mould during re-shaping and so the cuts would permit the composite plies in the interleaved zone to slide relative to each other as shown in Figure 32b. The cuts are positioned so that they remain within the straight portion after re-shaping to reduce any tendency for the interleaved zone to delaminate. Additional laminae and an interleaf layer were provided at the left side to cover the uppermost cut and so reduce any tendency of the surface composite lamina to peel off during re-shaping.



(a)

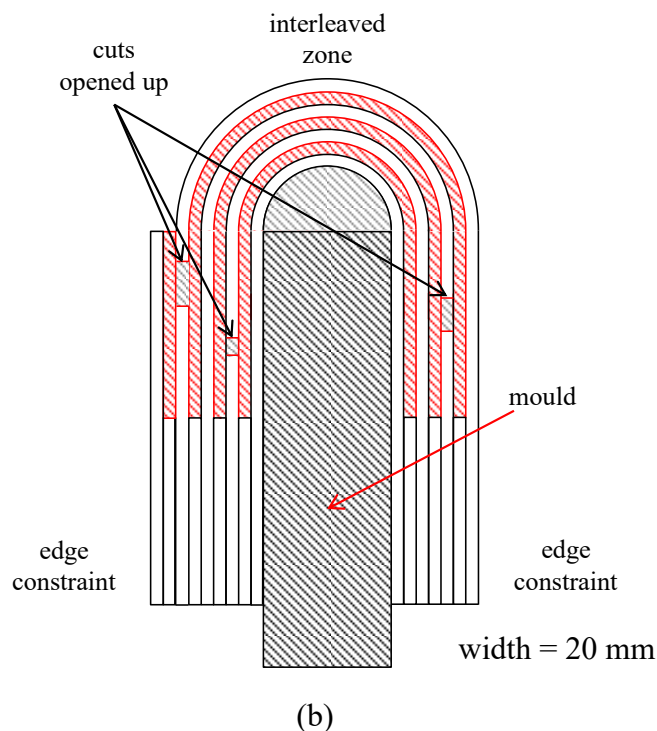


Figure 32. A sketch of (a) an interleaved hinge with cuts and (b) the re-shaped hinge. (White layers are composites and red layers represent interleaves.)

The trial configuration of Figure 32a was manufactured using the material specified in section 3.2.1 (see Figure 33a). After re-shaping to the 180°-bend configuration at elevated temperature and cooling, the specimen showed a small enclosed angle of around 2.5° (Figure 33b left). Optical microscope inspection showed that the cuts opened up as intended (Figure 33b right). Upon shape recovery, the hinge recovered towards the original, flat shape (Figure 33c). The enclosed angle after shape recovery was 176.7°. However, microscope examination showed that the cuts were not fully closed. It can be seen in Figure 33c that the PS had migrated into the cuts, perhaps due to the applied pressure during re-shaping process at 120°C which squeezed the softened PS into the opened cuts. This would have inhibited the complete shape recovery of this configuration.

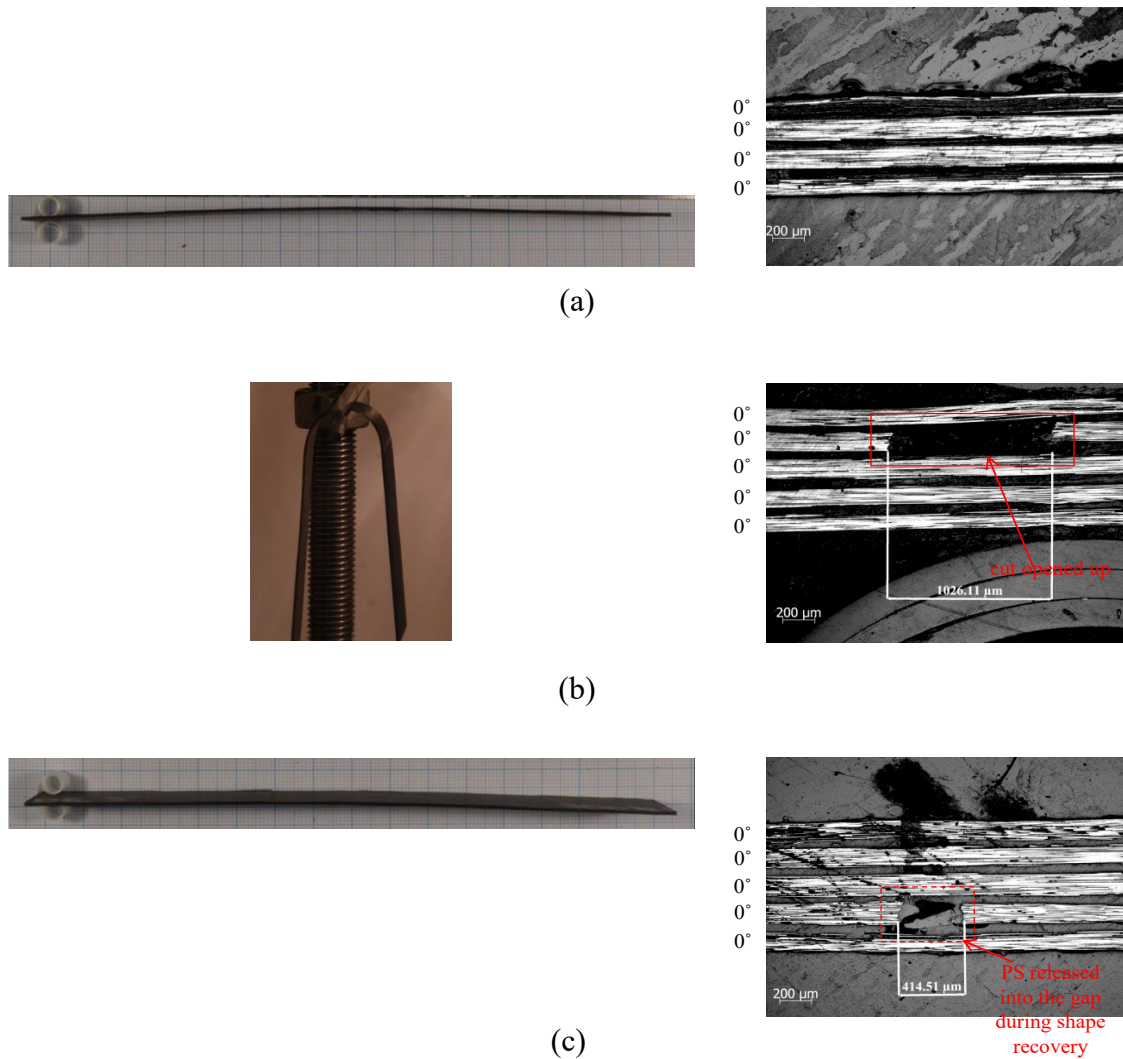
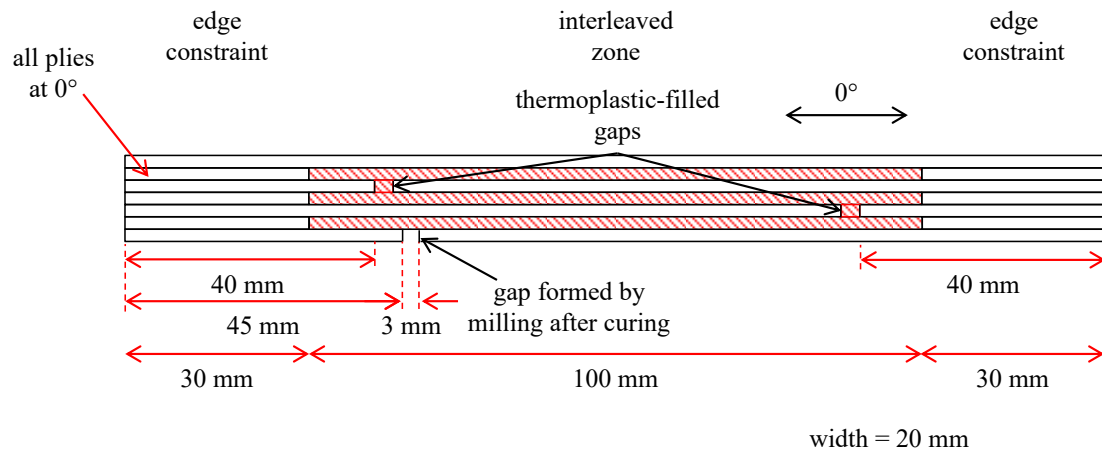
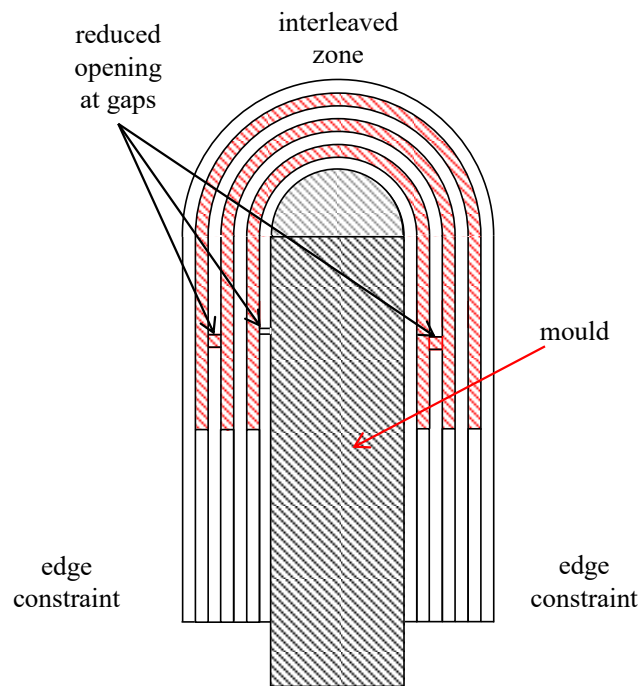


Figure 33. Images of an interleaved laminated hinge with cuts at (a) initial, (b) re-shaped and (c) recovery stages.

In the first trial, it was found that PS in the cuts prevented the shape memory effect. In the second trial, an alternative pattern of cuts was investigated. Gaps of 3 mm width were formed in the interior composite plies of the interleaved zone as shown in Figure 34a. These gaps were filled with PS films (the same material used as interleaves) to ensure that no epoxy resin would be released into these gaps during curing. An additional gap, as shown in Figure 34a, was to be introduced into the bottom composite ply by milling the specimen after curing. Only the outermost composite ply remained continuous in the interleaved zone.



(a)



(b)

Figure 34. A sketch of an interleaved hinge with thermoplastic-filled gaps at (a) as-cured and (b) re-shaped stages. (White layers are composites and red layers represent interleaves.)

In the second configuration the gaps would reduce in width on re-shaping to the 180°-bend configuration at elevated temperature (see Figure 34b). These gaps ensure that the composite plies can slide relative to each other at elevated temperature. Upon shape recovery, these gaps would open up and the hinge can recover to its initial, flat shape.

The second trial configuration was manufactured (see Figure 35a) and then re-shaped on the mould to the 180°-bend configuration. On cooling there was a small amount of springback giving an enclosed angle of around 1.7° (Figure 35b). The outermost, continuous composite ply, however, was fractured during the re-shaping process. Unfortunately, a microscope image of the cross section of the re-shaped hinge was not prepared. Upon shape recovery, the hinge almost fully recovered to the original, flat shape (Figure 35c). The enclosed angle after shape recovery was 178.3°. Microscopic investigation revealed that PS has been squashed out from the milled gap in the lower composite lamina during shape recovery and that a small amount of delamination had also occurred near the internal cuts (see Figure 35c).

Despite the upper ply fracture during re-shaping and the delamination, the second configuration recovered the original shape and it was decided to use this configuration for an integrated hinge demonstration specimen. The second configuration has the advantage that it does not require additional laminae and interleaf layer to be added to the base configuration.

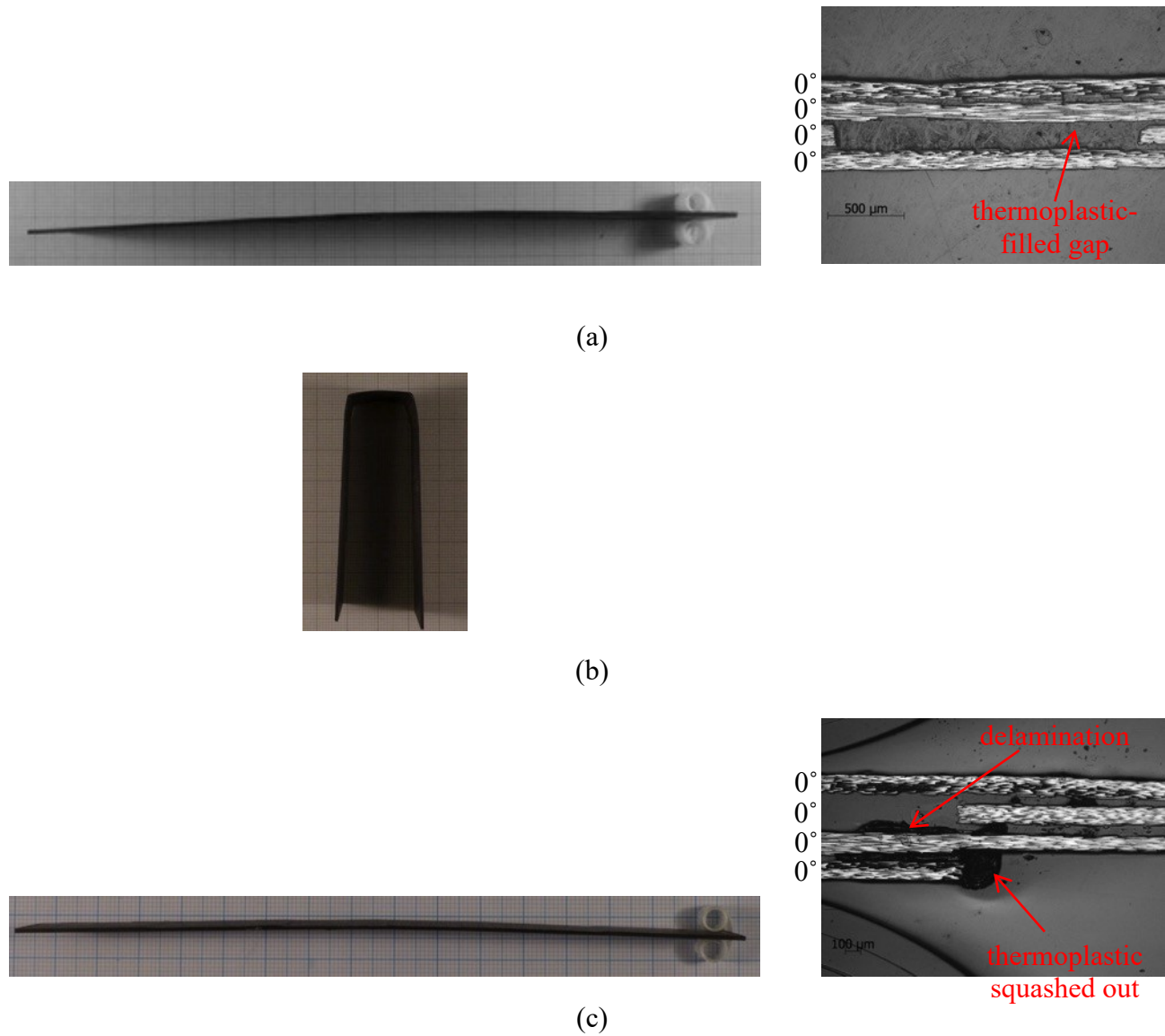


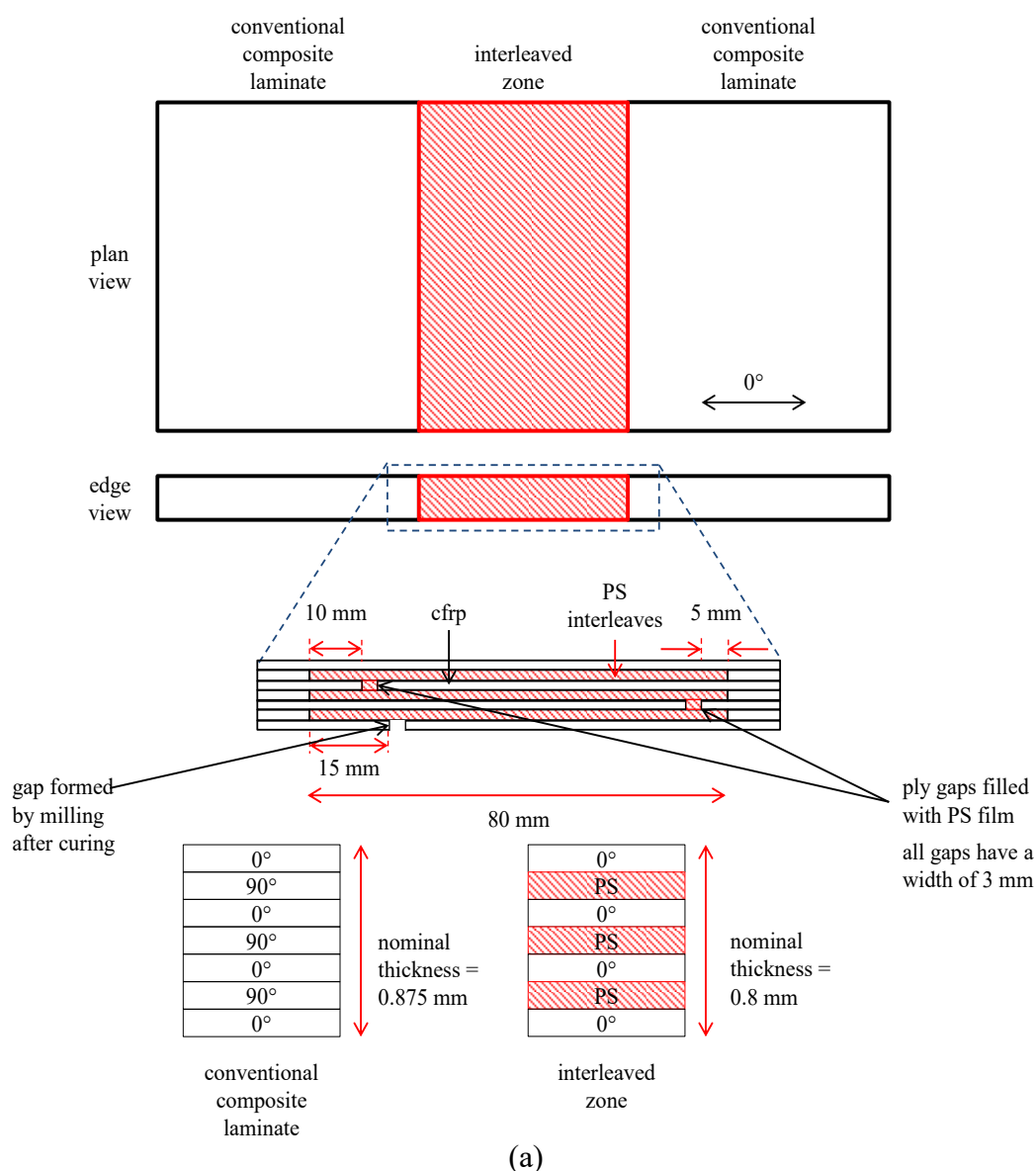
Figure 35. Images of an interleaved laminated hinge with thermoplastic-filled gaps at (a) initial, (b) re-shaped and (c) recovery stages.

3.2.2.2 Experimental details for manufacture and test of the integral hinge panels

A carbon fibre/epoxy panel (320 mm long and 120 mm wide) was manufactured by hand layup. The panel, as shown in Figure 36a, consisted of two non-interleaved composite segments with a cross-ply layup $[0^\circ/90^\circ/0^\circ/90^\circ/0^\circ/90^\circ/0^\circ]$ and a PS-interleaved segment with a 0° unidirectional layup $[0^\circ/\text{PS}/0^\circ/\text{PS}/0^\circ/\text{PS}/0^\circ]$. (The 0° fibre direction was along the longer side of the panel.) Two gaps of

3 mm width were introduced into the interior 0° composite plies. PS films (3 mm long and 0.1 mm thick) were used to fill these gaps to stop the epoxy resin seeping into these gaps during curing.

The panel was cured in an autoclave [Autoclave, LBBC, UK] at 175°C and a 100 psi pressure for 1 hour. After curing, the bottom ply of the panel was milled in a milling machine [Bridgeport, Adcock-Shipley, UK] to form a 3 mm wide ply gap at 15 mm away from the junction to the left conventional composite segment. After this, the interleaved segment of the cured panel was subsequently machined to form three hinge strips of 80 mm long and 20 mm wide. Figure 36b shows the panel prior testing. Two panels were manufactured for this investigation.



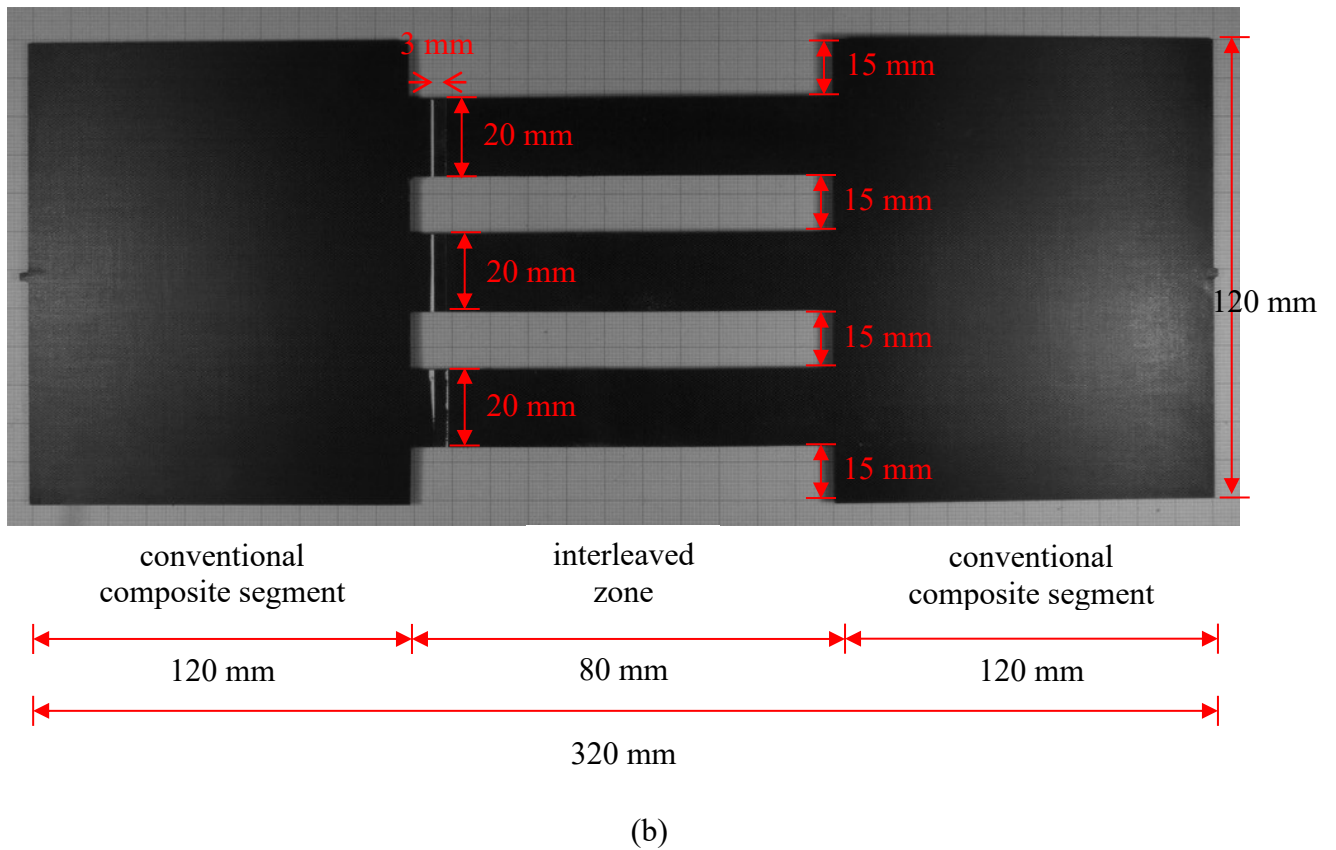


Figure 36. (a) A sketch of an interleaved composite panel with integrated interleaved composite hinges and (b) a photograph of a composite panel with integrated hinges after manufacture (plan-view).

As mentioned earlier, it was decided to re-shape the panel into a 180°-bend shape with an inner radius of 10 mm. (Simple beam theory had previously been used to show that the plies would not fail during re-shaping.) Figure 37 shows the procedure for re-shaping the panel. (In this figure, the full length of the composite segment is not shown to save space.)

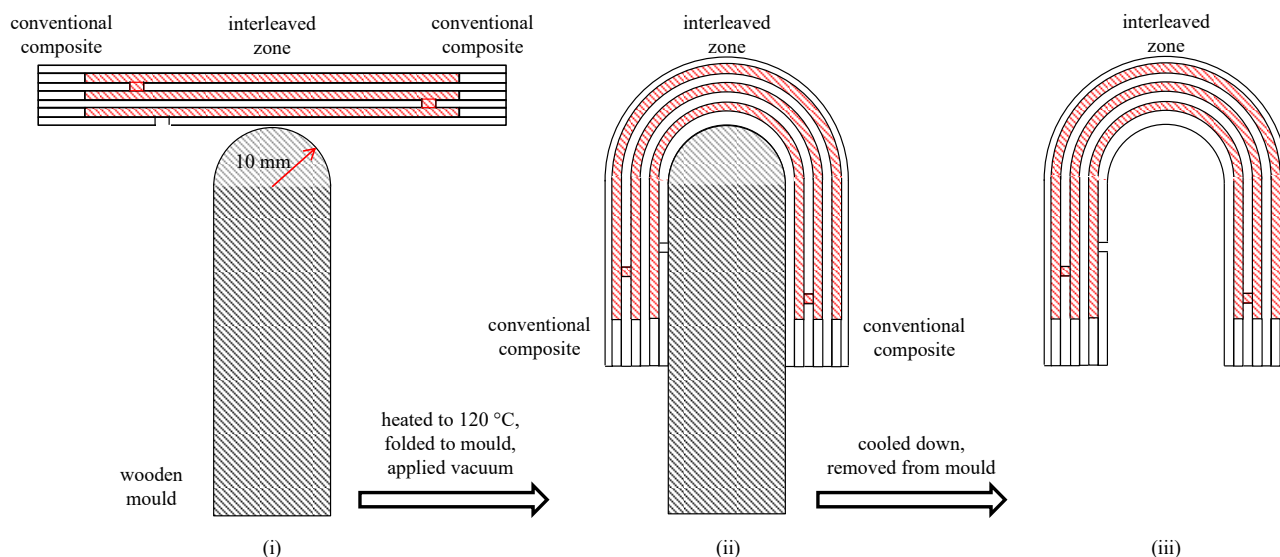


Figure 37. A sketch of the test procedure to re-shape an integrated composite panel to the 180°-bend configuration at 120°C.

To achieve this re-shaping, the panel was placed on a wooden mould with a radius of 10 mm. The assembly was placed on an aluminium plate, covered with PTFE release films, a breather cloth and a vacuum bag sealed to the edges of the plate. The assembly without vacuum applied was placed in an oven [Forced convection oven, OF-21E, Lab Companion, Jeio Tech, Korea], heated to 120°C and kept at this temperature for 15 minutes. It was then re-shaped manually towards the mould surface to form the 180°-bend configuration and vacuum pressure was applied to retain the panel in this configuration. The panel was cooled down to room temperature with the vacuum still applied. The re-shaped panel was then removed from the mould and its side profile was scanned in a flatbed scanner [HP Scanjet 4850, HP, UK] to measure the enclosed angle between the two composite segments.

To observe the shape recovery capability, the re-shaped panel was placed in an environmental chamber [SFL, Phoenix Calibration & Services Ltd, West Midland, UK], gripped between two steel bars (see Figure 38) and re-heated to 120°C. The shape recovery process was recorded by a Pentax camera [Pentax X-5 Digital Camera, Pentax Ricoh Imaging CO.,LTD, UK] until no further shape change of the panel was observed. After cooling, the side profile of the panel was again scanned

using the flatbed scanner to determine the enclosed angle between the two composite segments after shape recovery.

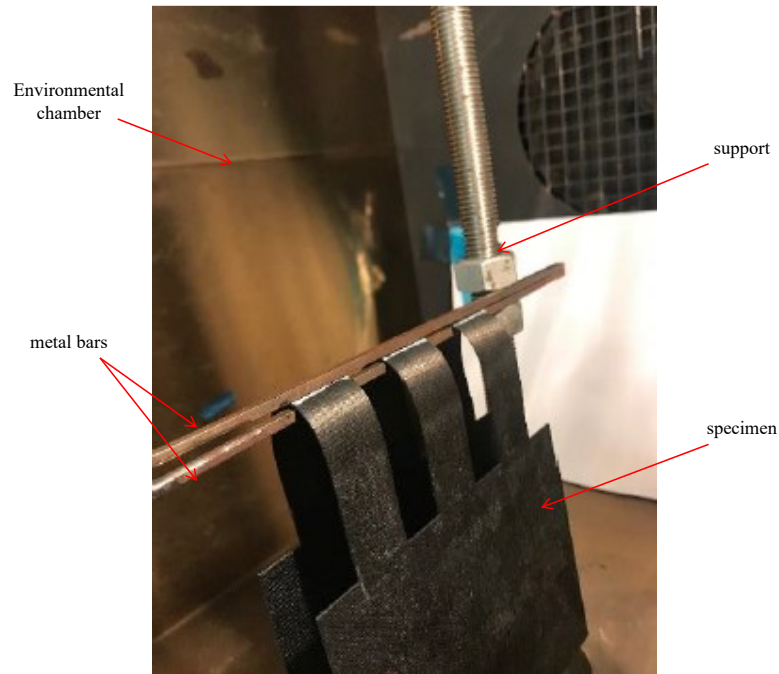


Figure 38. An image of the setup for the shape recovery test.

To examine the cross sections of these panels, small sections of the hinge elements of these panels were mounted in epoxy resin [Epoxicure, Buehler, UK] with a hardener [Kleer set hardener, 11 10 89, MetPrep, Buehler, UK] and cured for 8 hours at room temperature in air. The mounted sections were then polished in a grinder-polisher machine [MetaServ/MetaPol, Buehler, UK] with a SiC grinding paper in the grit level of P600 and 3 dispersions of diamond ($9\text{ }\mu\text{m}$), alumina ($0.6\text{ }\mu\text{m}$) and silco ($0.03\text{ }\mu\text{m}$) and then examined under an optical microscope [Axio Imager M2m, Zeiss, Germany].

To predict the springback angle of integrated hinges when released at room temperature after re-shaping and the recovery angle of the integrated hinges after shape recovery a beam analysis was performed. Consider an integrated hinge as shown in Figure 39b. This shows the integrated hinge in

the 180°-bend configuration. If the hinge is released now, there will be a small amount of deflection towards the initial flat shape, i.e. a small amount of springback can be observed.

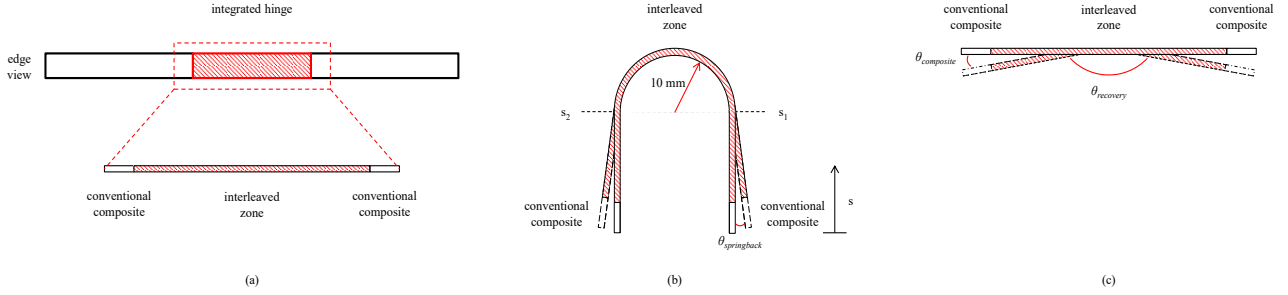


Figure 39. A sketch of the interleaved zone of the panel in (a) flat, (b) 180°-bend and (c) recovered shapes. The parameter s represents locations along the midline of the laminate.

The magnitude of the springback angle, $\theta_{springback}$, is related to the moment required to re-shape the panel. The moment, M_i , for a composite lamina i of thickness t and width b , to produce a change of curvature at the mid-plane of the lamina ($1/R_i$) is given by

$$M_i = \frac{E_i b t^3}{12 R_i} \quad (1)$$

where E_i is the longitudinal elastic modulus of the carbon fibre/epoxy lamina.

The total moment (M^T) for the re-shaping of the panel is therefore given by

$$M^T = \sum_{\text{for all composite lamina}} M_i \quad (2)$$

The maximum applied moment will occur in the part between S_1 and S_2 (see Figure 39b). However, a small reduction in the applied moment is expected when the panel is retained in the re-shaped form while cooling down, but this is neglected in the current calculation. The springback angle ($\theta_{springback}$) can be estimated by considering the total moment M^T applied in the opposite direction and is given by

$$\theta_{springback} = \frac{1}{(EI)_{RT}} \int_{for\ entire\ midline\ of\ panel} M^T ds \quad (3)$$

where s is the distance along the midline of the panel and $(EI)_{RT}$ is the flexural stiffness of the panel at room temperature. This can be calculated according to simple beam theory.

Ignoring self-weight, upon shape recovery, the panel, in the unconstrained state, would recover to the original flat shape (i.e. 180°), as the elastic stresses stored within the composite laminae are released. However, the recovery angle ($\theta_{recovery}$) is smaller than the intended value due to the weight of the composite segments (see Figure 39c).

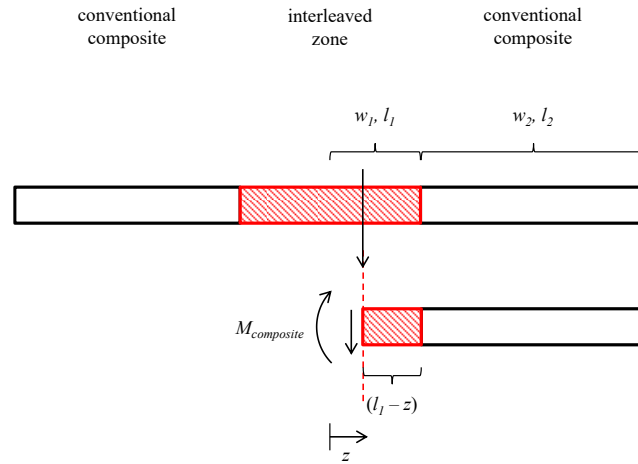


Figure 40. A sketch of the bending moment due to self weight of the composite.

The bending moment at position z (see Figure 40) due to the self-weight ($M_{composite}$) for z greater than 0 less than l_1 is given by

$$M_{composite} = w_1 \times \frac{(l_1 - z)^2}{2} + w_2 \times l_2 \times (l_1 - z + \frac{l_2}{2}) \quad (4)$$

where w_1 and w_2 are the weight of interleaved region and pure composite part per unit length, l_1 and l_2 are the length of interleaved region and pure composite part and z is the distance from the centre of the specimen.

The angle at the junction of the interleaved region to the composite segment due to the self-weight,

$\theta_{composite}$, in radian, is given by

$$\begin{aligned}\theta_{composite} &= \int_0^l \frac{M_{composite}}{(EI)_{HT}} dz \\ &= \frac{1}{(EI)_{HT}} \int_0^l [w_1 \times \frac{(l_1 - z)^2}{2} + w_2 \times l_2 \times (l_1 - z + \frac{l_2}{2})] dz\end{aligned}\tag{5}$$

where $(EI)_{HT}$ is the flexural stiffness of the interleaved region at 120°C.

Therefore the shape recovery angle ($\theta_{recovery}$) in degree is given by

$$\theta_{recovery} = 180^\circ - (2\theta_{composite} \times \frac{180^\circ}{\pi})\tag{6}$$

3.2.2.3 Results of the shape memory test for integral hinge panels

Table 2 shows the enclosed angle measured between the conventional composite segments of each panel after re-shaping and after shape recovery. On re-shaping the panels to the 180°-bend configuration, some springback was observed. The enclosed angles for the two panels were 3.5° and 2.8° and were within 2° higher than the predicted value of 1.6°. After shape recovery, both panels recovered towards the initial, as-cured shape, but did not fully recover. (Due to self weight of the panel, full recovery was not expected; the predicted enclosed angle was 168.7°.) The measured recovered enclosed angles of both panels were very close to the predicted value. However, panel 2 suffered a large amount of delamination during shape recovery (see Figure 41). It is suspected that the clamping mechanism was over-tightened and prevented the composite laminae from sliding

relative to each other. The resulting stress caused during shape recovery will have initiated the observed delamination failure.

Table 2. Enclosed angle measurement for re-shaping and shape memory of the integrated composite panels with in-built interleaved composite hinges.

Panels	State	Predicted enclosed angle* (°)	Measured enclosed angle (°)
1	Re-shaped to 180° bend shape	1.6	3.5
	Recovered to flat shape	168.7	169.3
2	Re-shaped to 180° bend shape	1.6	2.8
	Recovered to flat shape	168.7	167.5

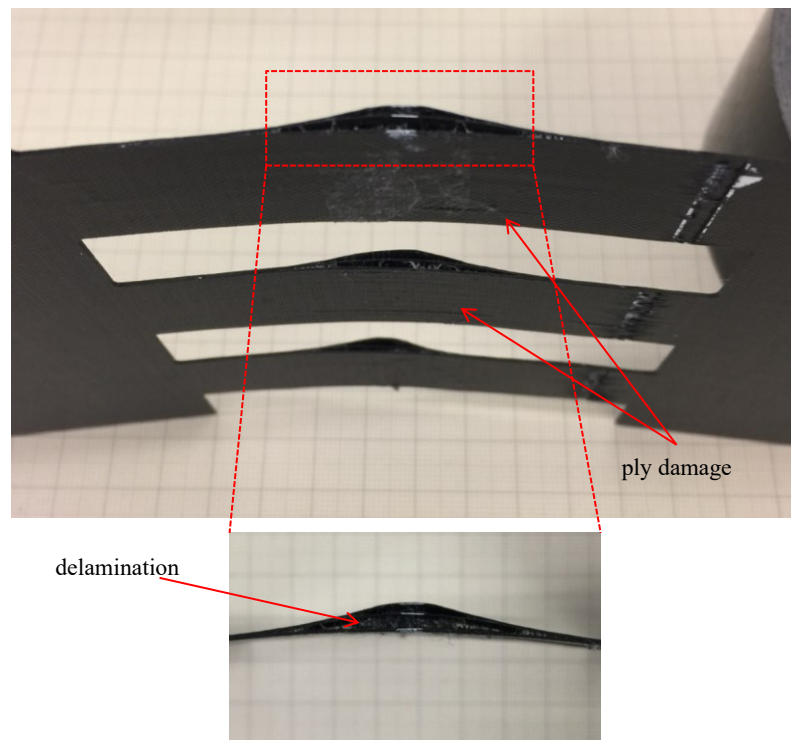


Figure 41. An image of the interleaved panel 2 after shape recovery to show the ply damage and delamination in the PS interleaves.

Figure 42 shows still images extracted from the video of panel 1 recovering from the 180°-bend configuration. The shape recovery duration of the 180°-bend panel was around 5 minutes after the panel was mounted on the supporting fixture in the chamber at 120°C. (The video was started after

closing the chamber door. Therefore, during the early stage of this video the chamber was recovering the temperature and so the precise shape recovery time for a constant temperature of 120°C could not be measured.)

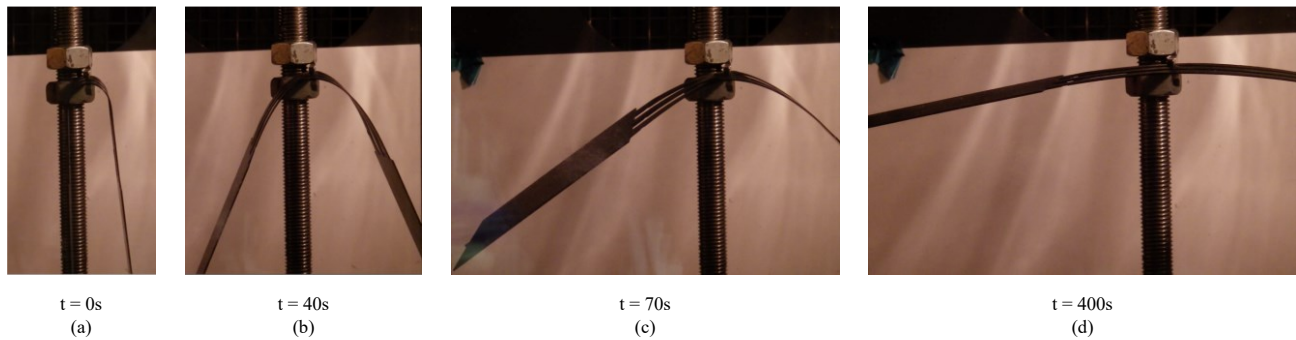


Figure 42. A sequence of still frames taken from the deployment video of an integrated composite structure from the 180°-bend shape.

Figure 43 plots the enclosed angle versus time of the panel during shape recovery. It can be seen that 80% of the shape recovery of the specimen was observed in less than 2 minutes. This is because the stored elastic stresses within the composite plies are released at 120°C. The recovery rate then became slower and this is due to the viscoelastic behaviour of the PS interleaves. (According to Ward and Sweeney (186), when a viscoelastic polymer is released from a fixed strain, it gradually recovers towards its initial configuration due to its viscous feature. In this case, after the release of the stored elastic stresses, the PS interleaves gradually recovered towards the initial shape as time proceeded.)

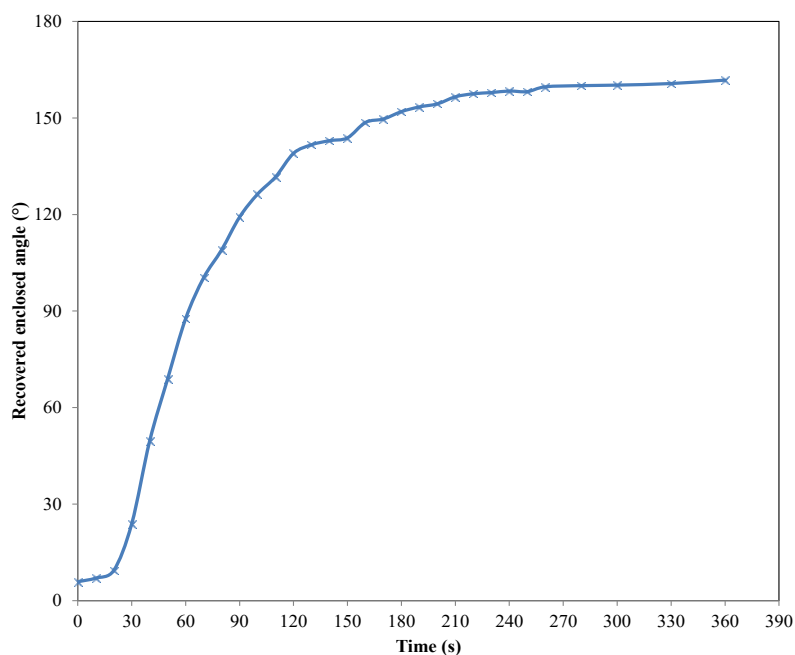


Figure 43. The recovered enclosed angle versus time plot of an integrated hinge specimen during shape recovery at 120°C.

Figure 44 shows the polished cross section taken from the optical microscope of the interleaved zone of panel 1 after curing and after shape recovery. It shows that after curing, PS interleaves were at every interface with the adjacent carbon fibre/epoxy laminae and at the gaps. After shape recovery, PS was squashed out from the gap on the bottom surface of the panel. (However, it is difficult to determine whether this phenomenon occurred after re-shaping or after shape recovery, due to the lack of microscope inspection of the panel after re-shaping.) Voids were also formed during shape recovery. (Voidage has been reported in earlier work and is due to the outgassing of the PS interleaves and the film shrinkage at the high temperatures during the re-shaping and shape recovery stage (127).)

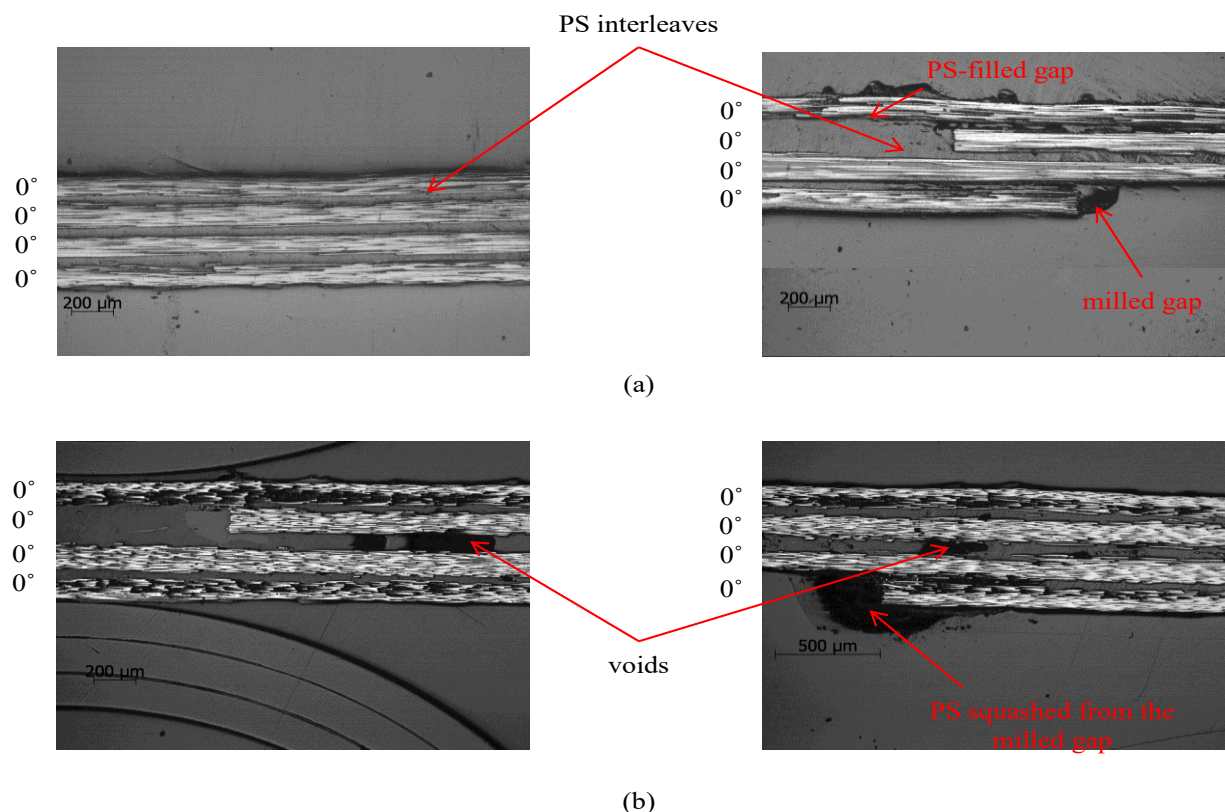


Figure 44. Microscope images of the cross section of the interleaved panel 1 (a) after curing and (b) after shape recovery.

3.2.2.4 Discussion and summary of integrated hinge investigation

It has been shown that an integrated deployable hinge composite panel was developed to exploit the shape memory capability of thermoplastic interleaved composites for deployable applications. However, it was observed that there was delamination in the panel 2. The delamination was probably due to the over-tightened clamping mechanism which inhibited internal ply sliding at elevated temperature. (Previous trials on specimens have shown that no delamination occurred during the shape recovery test, as they were not over-clamped and so inter-ply sliding was able to take place.) Voids were formed in the PS interleave of both specimens. The time for shape recovery is highly dependent on the viscoelastic behaviour of the PS interleave at 120°C. (It was observed that the recovery rate became slower as time proceeded.)

3.2.3 Linear actuator

3.2.3.1 Design concept

Previous works have shown that PS-interleaved composites exhibit a significant shape memory capability in flexure (127). This can be used to produce an angle change which can be employed in deployable hinges such as the integrated hinges explored in the previous section. In the study described in this section, the linear actuation capability of interleaved composites was investigated.

A laminated carbon fibre/epoxy composite actuator with thermoplastic interleaves in a sine wave profile is developed. As shown in Figure 45, this consists of unidirectional carbon fibre/epoxy laminae with thermoplastic interleaves at every interface in the gauge length region. (Gauge length is defined as the length of the wavy region between the straight ends.) The straight ends do not contain thermoplastic interleaves. This is to prevent sliding of the composite plies within the gripped ends, when being tested at elevated temperature. To ensure linear actuation and to avoid delamination, all the carbon fibre/epoxy laminae would need to possess the same sine wave geometry. (Delamination can occur if the laminae are not in the same length as the shorter laminae will be straightened before the longer laminae, resulting in a transverse tensile force acting across the actuator width and leads to delamination.) This implies that all the carbon fibre/epoxy laminae are of the same length and that the PS interleaves would vary in thickness along the wave; thickest at the crest and trough positions, and thinnest at the position midway between adjacent crest and trough. (The thickness of PS interleaves is the same when arrived, but varies after composite manufacturing as the PS is softened at elevated temperature and can be squeezed with the use of pressure.) To evaluate the linear actuation capability, the actuator would be pulled into a straight form by axial extension at elevated temperature and then cooled to retain this shape. On subsequent heating in the unconstrained state, the actuator should contract back to its original, sine wave geometry.

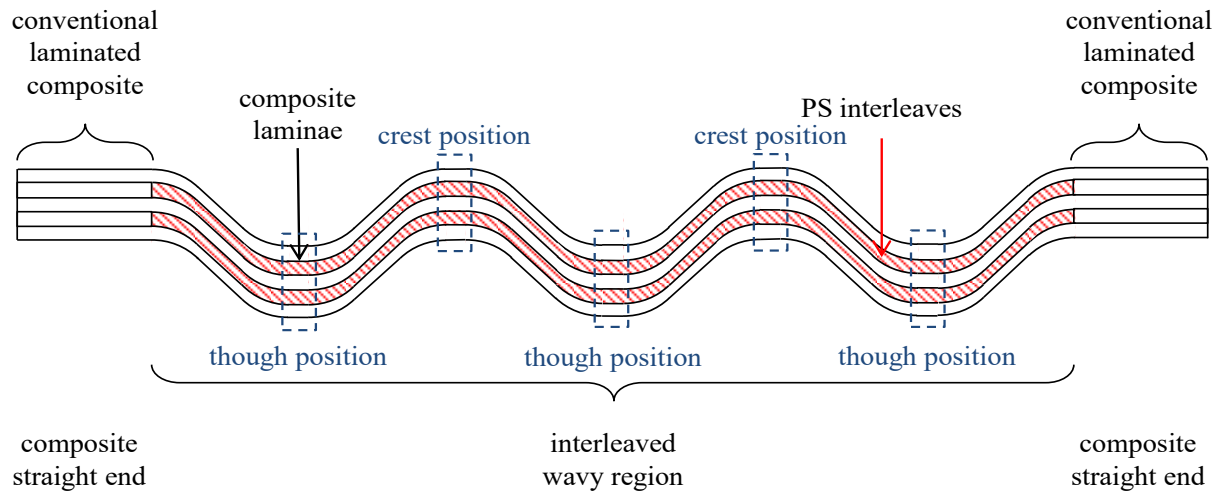


Figure 45. A sketch of the interleaved composite linear actuator concept.

It should be noted that even though sliding of the composite plies is prevented at the ends of the interleaved wavy zone, this sine wave concept does not require ply cuts or gaps to permit the shape memory capability as the relative ply sliding occurs between adjacent trough and crest positions of the wave with zero sliding at these positions and maximum sliding at the position of midway between crest and trough positions, as illustrated in Figure 46.

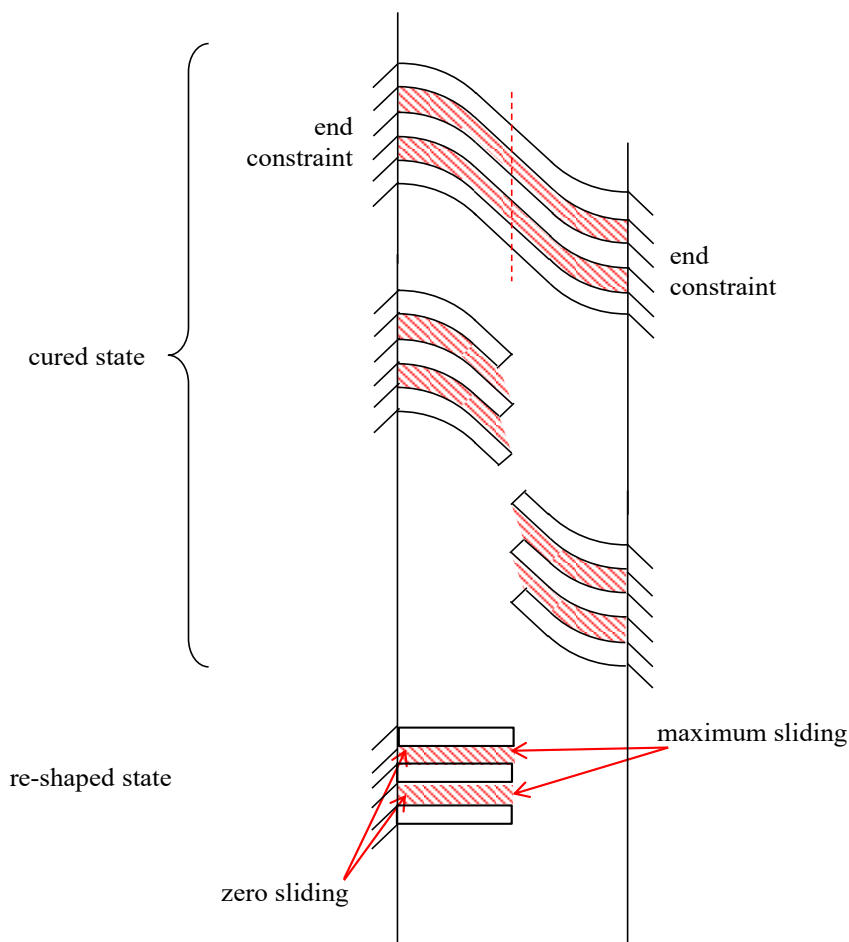


Figure 46. A schematic diagram to demonstrate the sliding due to shearing of the PS interleaves at 120°C.

3.2.3.2 Experimental investigation of linear actuator concept

Two sizes of unidirectional carbon fibre/epoxy lamina (designated as lamina A and B) and two sizes of PS interleaf films (designated as film C and D) were prepared. As shown in Figure 47, lamina A is 0° unidirectional carbon fibre/epoxy lamina of 250 mm length and 150 mm width and lamina B is 0° unidirectional carbon fibre/epoxy lamina of 250 mm length and 30 mm width. (Two laminae B were prepared.) Both composite laminae have a thickness of 0.125 mm. The 0° fibre direction was along the longer side of these laminae. PS film C is of 180 mm length and 150 mm width and PS film D is of 180 mm length and 30 mm width. Two of each film was prepared. Both PS films are of 0.1 mm thick.

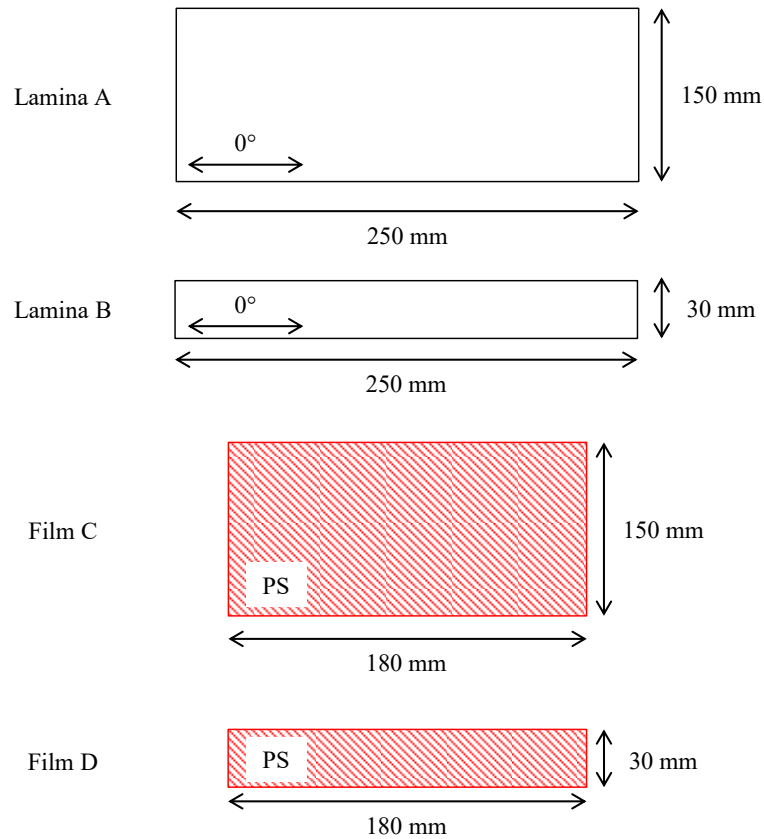


Figure 47. A sketch of the dimensions of composite laminae and PS films for the linear actuator.

The stacking configurations of these materials are showed in Figure 48. Lamina A was stacked with PS films C on its upper and lower surfaces in the gauge length region by hand layup (see the bottom sketch in Figure 48). For lamina B, one was laid up with PS film D at its upper surface in the gauge length region (the middle sketch), while the other was laid up with PS film D at its lower surface in the gauge length region (the upper sketch).

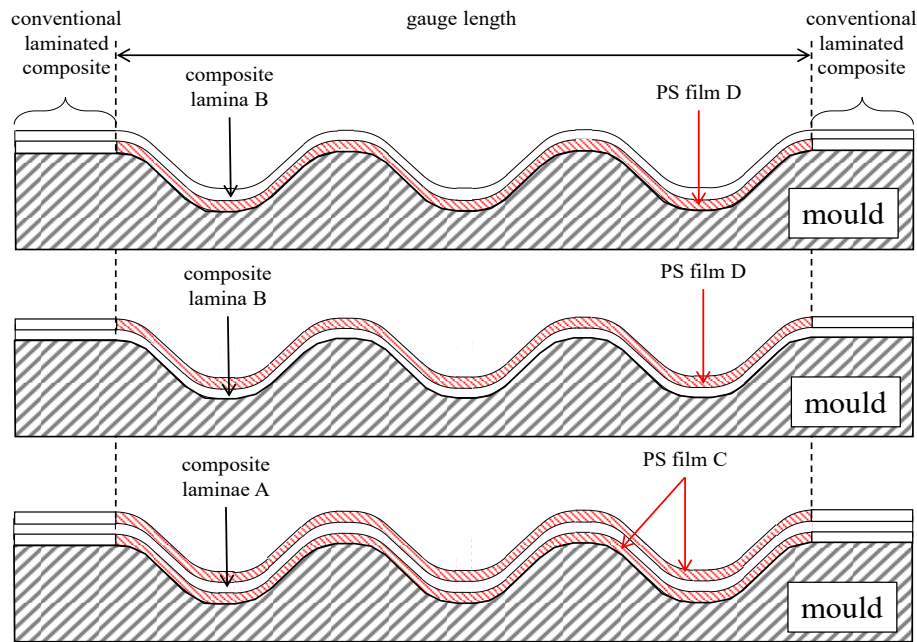


Figure 48. The stacking configurations for the manufacture of the linear actuator.

These stacked laminae were then placed separately on a corrugated aluminium mould in a sine wave geometry shown in Figure 49. The aluminium mould of 240 mm × 240 mm square and 20 mm thick possesses three sine waves at a wavelength (λ) of 60 mm and an amplitude (A) of 5 mm. They were manually driven down onto the surface of the aluminium mould to form a corrugated configuration. (The mould was coated with a thin layer of release agent [Chemlease[®] 41-90 EZ, Chem-Trend (Deutschland) GmbH, Germany] so that after curing these laminae can be easily removed from the mould.) Another aluminium mould of the same geometry was placed on the top of these laminae to retain them at this corrugated configuration.

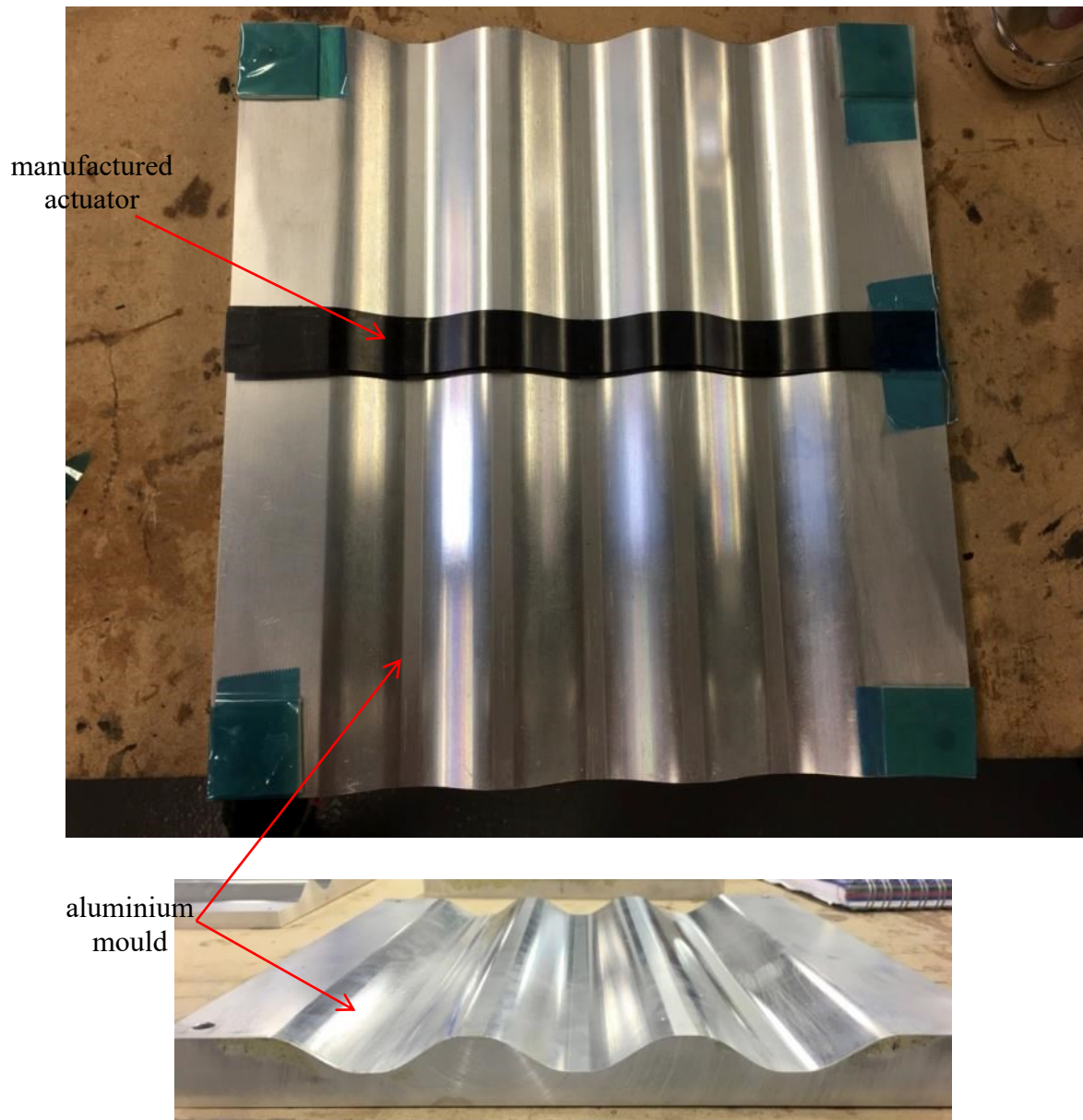


Figure 49. The corrugated aluminium mould.

The assembly was cured in an autoclave [Autoclave, LBBC, UK] at 175°C and a 100 psi pressure for 1 hour, following the manufacturer's instructions. After curing, the cured laminate A (with PS interleaf films C on both surfaces) was trimmed to four strips of 250 mm length and 20 mm width using a wire saw [MURG 24-BB, Well Diamond Wire Saw, Inc., USA] and the cured laminates B (with PS interleaf films D) were cut to the same dimension. These laminates were returned to the aluminium mould and stacked in the sequence of laminate B/(laminate A)₄/laminate B to form a new laminate in a unidirectional layup configuration ([0°/PS/PS/0°/PS/PS/0°/PS/PS/0°/PS/PS/0°/PS/PS

/0°]). (As shown in Figure 48, laminate A had PS layers on both surfaces and laminate B had one PS layer. Therefore, when laid up, two PS layers were stacked together between two adjacent composite plies.) FM300 epoxy film adhesive [FM300-2 film adhesive, Cytac, UK] was embedded at the interface between every two carbon/epoxy laminae at the conventional composite regions. The assembled laminate was then heated in a hot press [C969 Hydraulic press, Moore, UK] at 121°C for 90 minutes with a 40 psi pressure to cure the film adhesive. After heating, the specimen surface was lightly sanded at the straight ends using a sand paper with a grit level of P80 [241UZ, 3M, UK] and bonded to glass fibre end tabs using a two-part high temperature resistant adhesive cured at room temperature in vacuum pressure for at least 8 hours [Araldite 2011, Huntsman Advance Material (Europe) BVBA, Belgium]. After this, the specimen was polished along its side profile with sand papers in grit level of P320, P600, P1200, P2500 and P4000 [Metallographic Abrasive paper, Metprep, UK] for microscope examination. In this way, an interleaved carbon fibre/epoxy composite actuator in a sine-wave form was manufactured (see Figure 50).

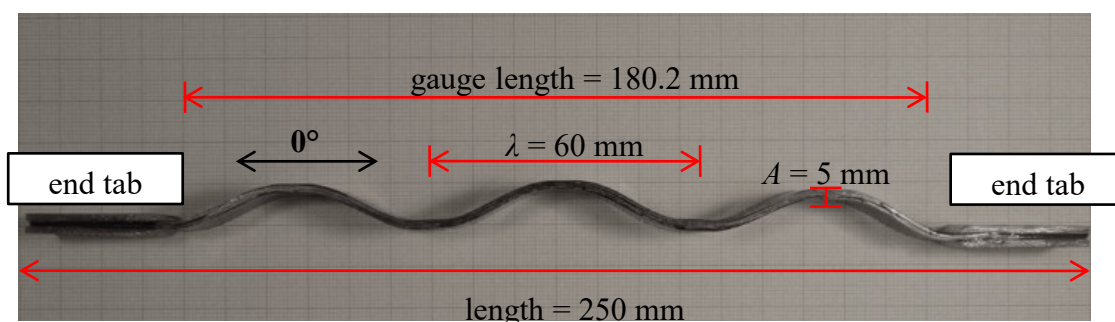


Figure 50. An as-manufactured interleaved composite actuator.

As mentioned earlier, the specimen was re-shaped into a straight form in a tensile test at elevated temperature. It was gripped in tension fixtures mounted on a universal mechanical testing machine [Instron 5985 machine, Buck, UK] fitted with a 1 kN load cell. The specimen and tension grips were embedded within an environmental chamber [SFL, Phoenix Calibration & Services Ltd, West Midland, UK] (see Figure 51). The specimen was heated to 120°C in the chamber and kept at this

temperature for 15 minutes. It was then subjected to a tensile test at a constant crosshead speed of 1 mm/min until a rapid increase of the load was observed to indicate that the actuator was almost straight. The re-shaped specimen was cooled down to room temperature under load to retain the straightened shape. The gauge length of the specimen was measured using a calliper before testing, after testing while mounted in the grips and after removing it from the grips to determine any springback.

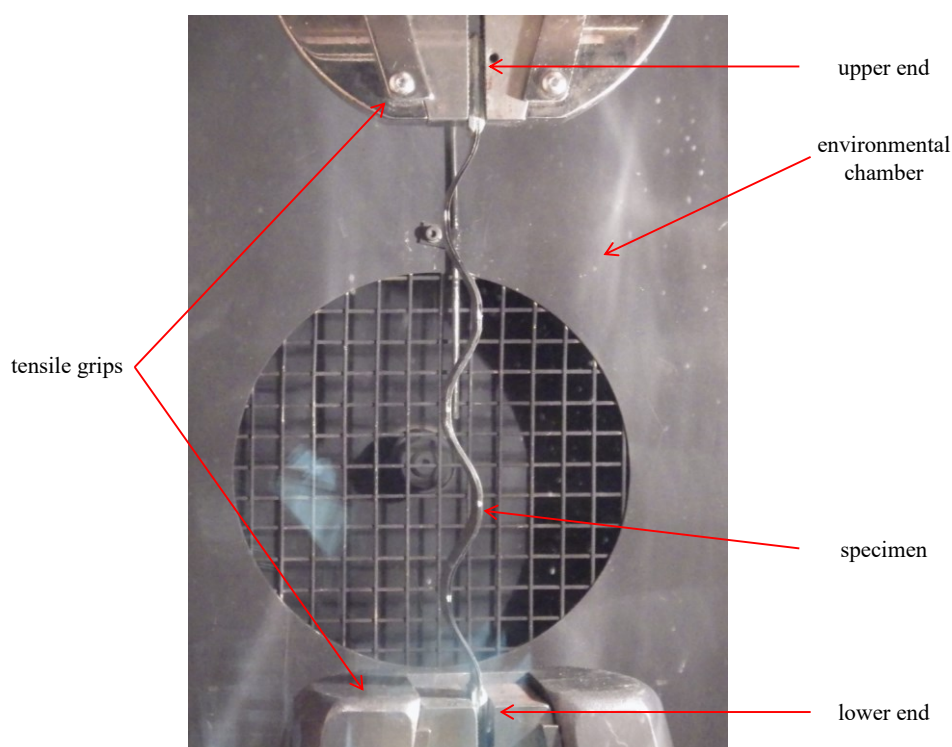


Figure 51. Experimental setup for the re-shaping test of the interleaved composite actuator.

To investigate the shape recovery capability, the re-shaped actuator was returned to the chamber and gripped at its upper end, while the lower end was unconstrained. The specimen was then re-heated to 120°C and the shape recovery process was videoed using a Pentax camera until no further significant shape change was observed. The specimen was then cooled down to room temperature. After cooling, the gauge length of the specimen was measured using a calliper.

3.2.3.3 Results of the linear actuator experimental investigation

Figure 52 shows the tensile load-displacement response of a specimen at elevated temperature. The figure shows that the load was initially small while the specimen was extending (approximately 50 N for the displacement of around 12 mm) and then significantly increased as the specimen became fully straightened. Figure 52 also shows still images from the re-shaping video of the specimen.

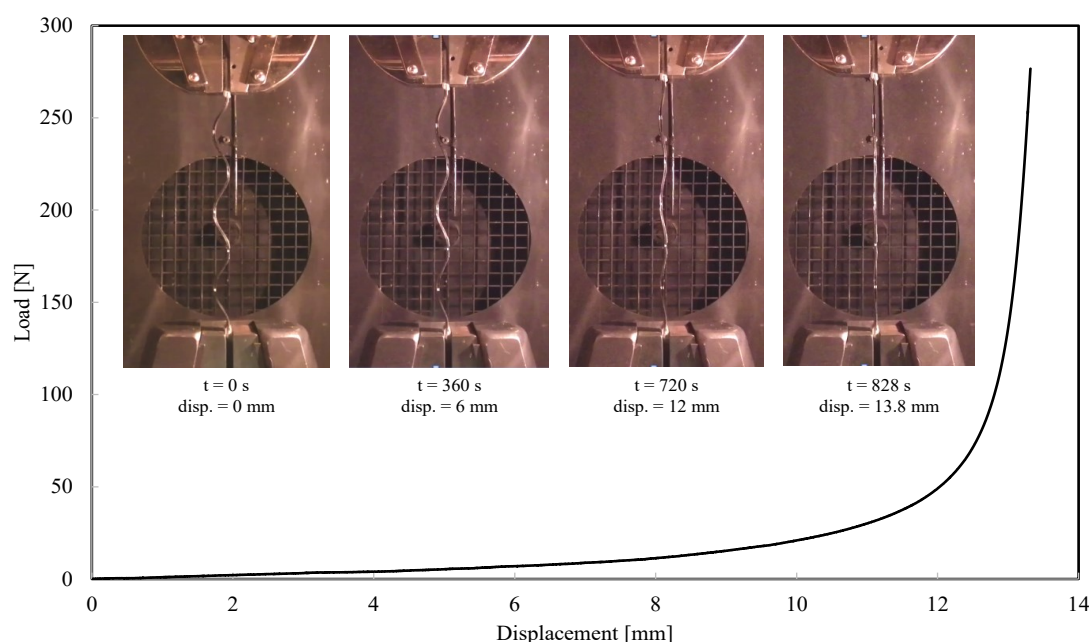


Figure 52. The load-displacement response of an interleaved composite actuator in tension at 120°C.

Table 3 summarises the gauge length of the specimen measured after curing, after re-shaping while mounted in the grips, after removing from the grips and after shape recovery. The gauge length of the as-cured actuator was 180.2 mm. After re-shaping while mounting on the grips, the gauge length was 193.3 mm. After removing from the grips, a small springback of 0.1% was observed. Upon re-heating, the specimen almost fully recovered to the initial, corrugated shape. The recovered gauge length was 0.8% greater than the initial measurement.

Table 3. The measured gauge length of the interleaved actuator after curing, re-shaping and shape recovery.

State	Measured gauge length (mm)
As-cured	180.2
Re-shaped to straight shape before removed from grips	193.3
Removed from grips	193.1
After shape recovery	181.6

Figure 53 shows a series of frames extracted from the shape recovery video of the actuator from the straight configuration. The specimen almost fully recovered its initial shape in around 2 minutes after returning to the chamber. (It is difficult to precisely determine the shape recovery time because the chamber was recovering the temperature at the early stage of the shape recovery video.)

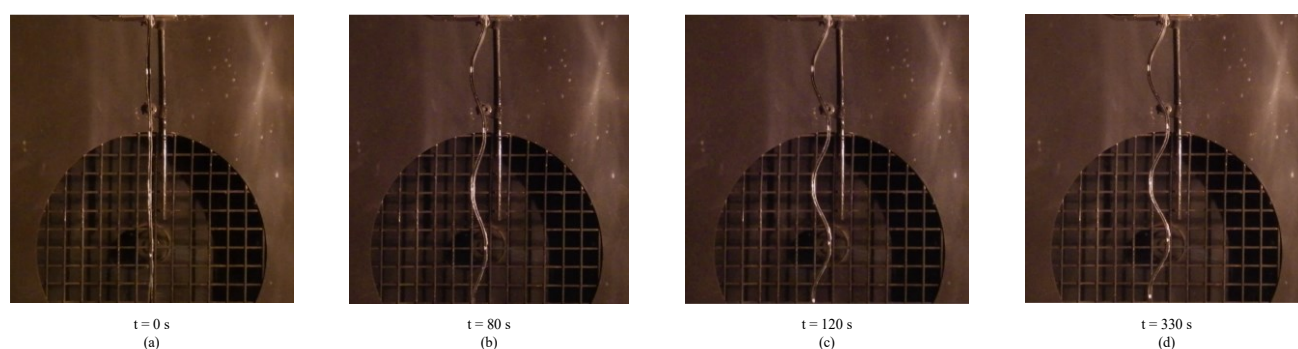


Figure 53. Still frames extracted from the shape recovery video of the specimen.

Figure 54 plots the recovered wavelength of one wave of the actuator as a function of time. (It was intended to measure the overall gauge length of the specimen, but the specimen was not fully visible at the beginning of the video.) The wavelength was measured between crests of a wave of the actuator. (Two points were marked on the adjacent crests of a wave and the distance between the two marks was measured from the video using commercial designing software [Creo 3.0, PTC software]. To ensure measurement accuracy, a scale factor was determined by comparing the measured fan mesh length with the actual distance measured by a calliper.) The figure shows that the specimen

recovered rapidly in the first 100 seconds, but as the stored elastic stresses within the composite plies were released, the recovery rate reduced.

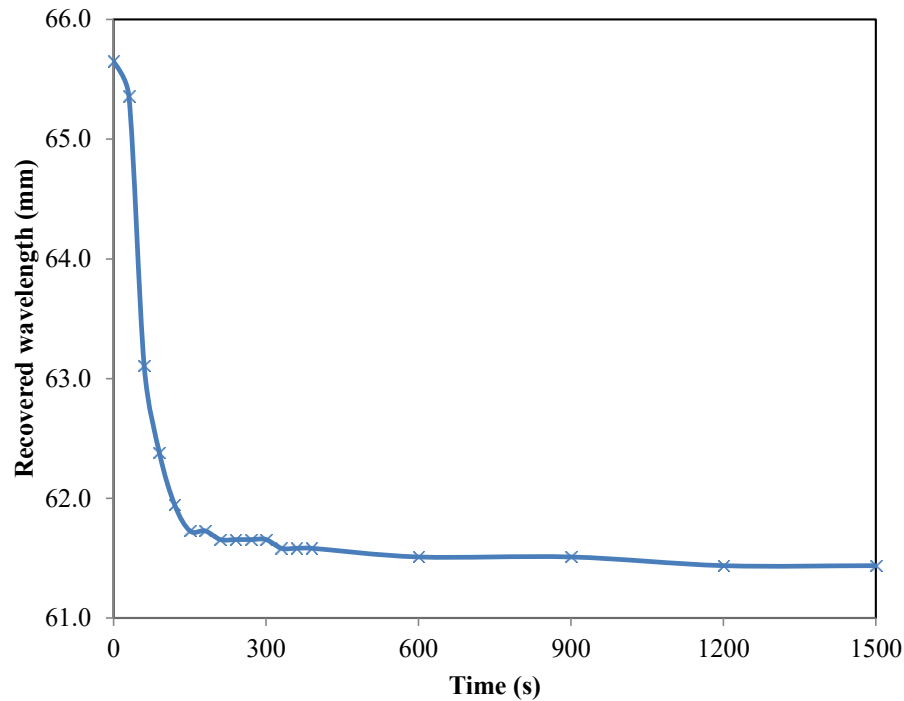


Figure 54. Recovered wavelength-time plot of a linear actuator specimen at 120°C.

Figure 55a shows the polished cross section from microscope inspection of the interleaved zone of the specimen after heating in the hot press. It can be seen that the PS interleaves are present between the carbon fibre/epoxy laminae in the gauge length region. On re-shaping to the straight configuration, Figure 55b shows that separation of the plies occurred at various positions in the specimen.

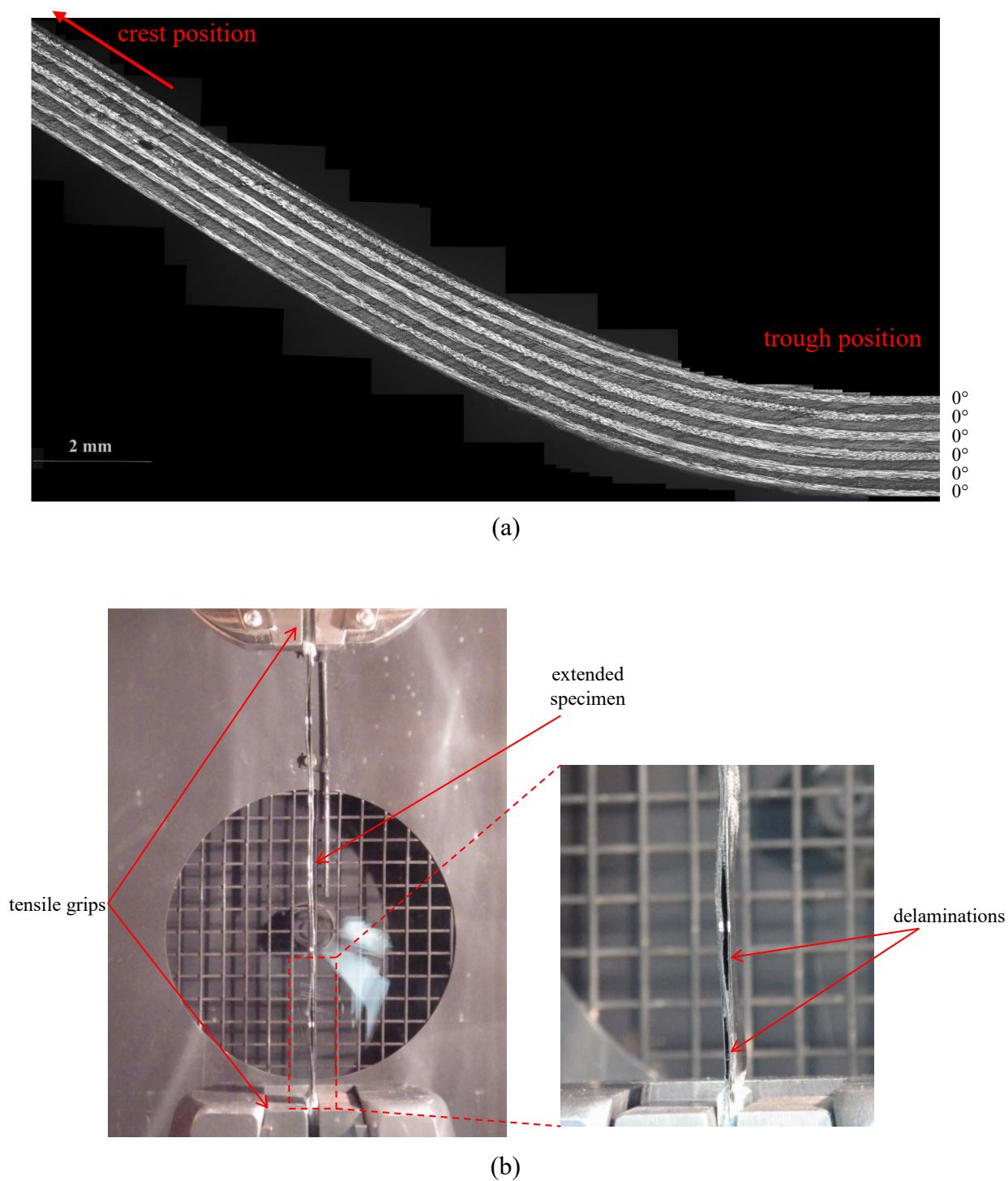


Figure 55. (a) An image of the cross section of the as-manufactured specimen and (b) images to show delamination in the specimen after re-shaping.

3.2.3.4 Discussion and summary of the interleaved linear actuator study

It has been shown that a linear actuator made of PS-interleaved carbon fibre/epoxy composites can be straightened when subjected to a tensile test at elevated temperature, without the need of ply gaps/cuts to permit internal composite ply sliding. However, examination of a straightened specimen showed that there was delamination at a number of positions. Delamination initiated due to through-thickness tensile stresses produced during the straightening of the waves. The delamination may also have been exacerbated by the small difference in the length of the carbon fibre/epoxy plies (see Figure 48). The shortest composite lamina layer was straightened earlier than the other, longer, ones, resulting in increasing through-thickness tensile stresses. The unequal lengths of carbon fibre/epoxy laminae were due to a manufacturing error.

Despite the delamination, the linear actuator recovered to its original corrugated shape upon reheating. The shape recovery rate became slower as the recovery progressed due to the viscoelastic behaviour of PS interleaves.

3.2.4 An expanded composite mesh with shape contraction capability

3.2.4.1 Development of mesh concept

It has been seen that delamination occurred in the linear sine wave actuator when the actuator was tested under uniaxial tension at elevated temperature. To reduce the tendency to delaminate, the modified interleaved composite configuration, as shown in Figure 56, could be used. Here the wavy configuration again consists of continuous carbon fibre/epoxy laminae and PS interleaves, but at the crest and trough positions of the sine waves a non-interleaved composite is used (identified as ‘composite block’ in Figure 56).

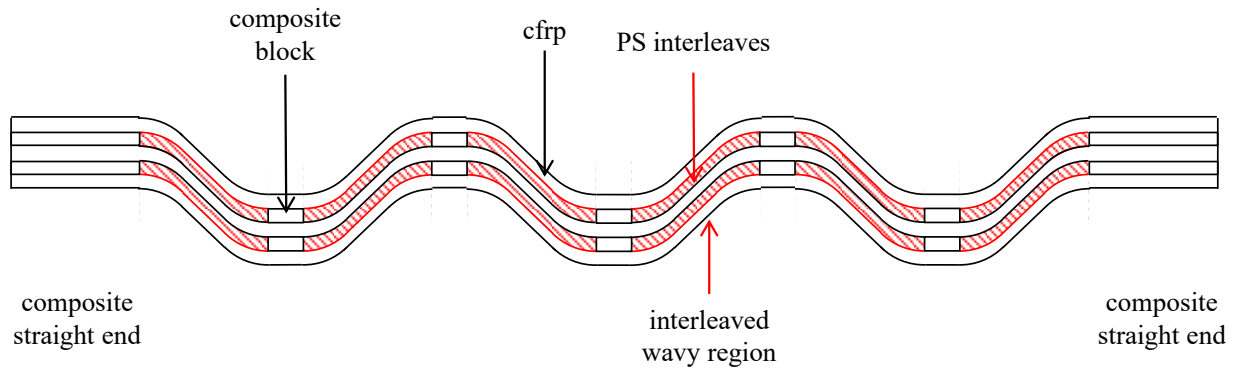


Figure 56. A sketch of a modified interleaved composite actuator. (White layers are carbon fibre/epoxy composites and red layers represent PS interleaves.)

Actually, it is difficult to manufacture this composite using the previous method mentioned in section 3.2.3.2 and ensure all the carbon fibre/epoxy laminae are of the same length. To address this problem, one approach would be to manufacture this composite in a flat shape (see Figure 57) and then re-shape this at elevated temperature into the wave configuration.

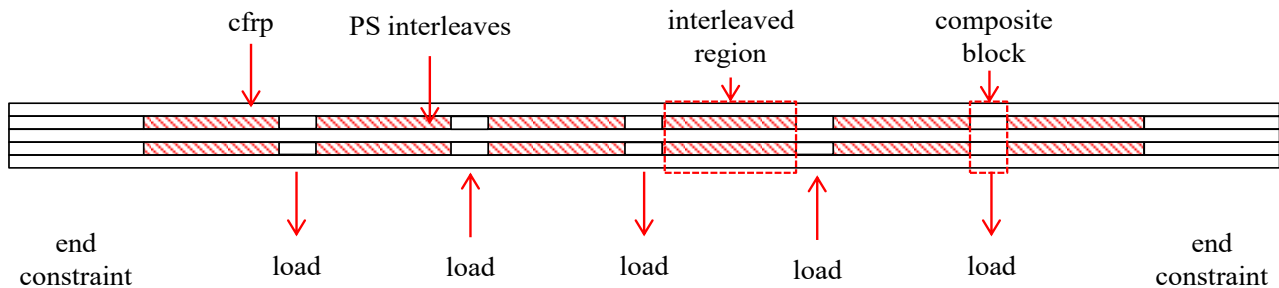


Figure 57. Flat layup for wavy actuators to be re-shaped to the wavy profile.

If two sets of the flat layup of Figure 57 are laid up with suitably positioned PTFE release films as shown in Figure 58, then, after curing, this could be re-shaped at elevated temperature to form the mesh structure shown in the lower diagram of Figure 58.

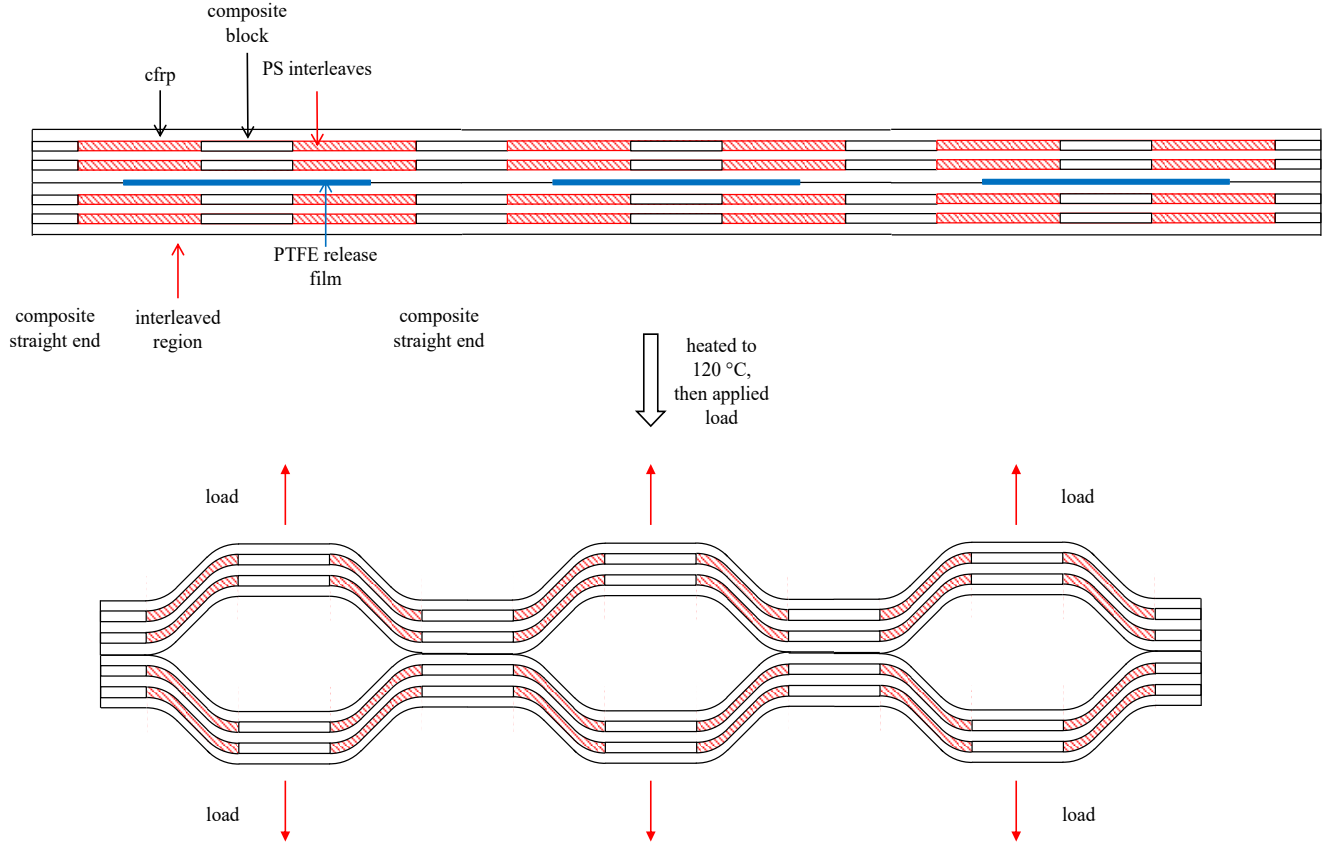


Figure 58. A sketch of the expanded mesh structure at as-cured and expanded stages.

3.2.4.2 Displacement analysis of the mesh

To determine the maximum possible expansion of the mesh and to estimate the corresponding load a simple beam analysis was performed. Taking advantage of the symmetry, consider one half of a unit cell of the mesh as shown in Figure 59a. This shows the mesh in the expanded state and shows the force, $P/2$, required to achieve this at the elevated temperature. If the non-interleaved portions are considered rigid then the analysis for the cell expansion need only consider the interleaved portion (see Figure 59b). Simple beam theory gives

$$\frac{\delta_{max}}{2} = \frac{PL^3}{24(EI)_{HT}} \quad (7)$$

where δ_{max} is the maximum possible displacement, L is the length of the interleaved region and $(EI)_{HT}$ is the flexural stiffness of the interleaved composite at elevated temperature.

The maximum change in curvature, $1/R_{max}$, occurs at the ends of the interleaved portion and is given by

$$\frac{1}{R_{max}} = \frac{PL}{4(EI)_{HT}} \quad (8)$$

Combining equations (7) and (8) gives

$$\delta_{max} = \frac{L^2}{3R_{max}} \quad (9)$$

Noting the maximum strain, ε_{max} , in a composite ply, of thickness t , is given by

$$\varepsilon_{max} = \frac{\sigma_{max}}{E} = \frac{y}{R_{max}} = \frac{t}{2R_{max}} \quad (10)$$

where E is the elastic modulus of the beam and y is the distance to the centroid of the beam.

Then the maximum possible displacement (δ_{max}) can be expressed in terms of the maximum strain (ε_{max}) as

$$\delta_{max} = \frac{2\varepsilon_{max}L^2}{3t} \quad (11)$$

In this study, the length of the interleaved region (i.e. L) was determined to be 45 mm. Taking the maximum allowable strain (i.e. ε_{max}) as 1×10^{-2} and the thickness of T300/914 composite (i.e. t) as 0.125 mm then the maximum possible displacement of the unit cell mesh is 108 mm.

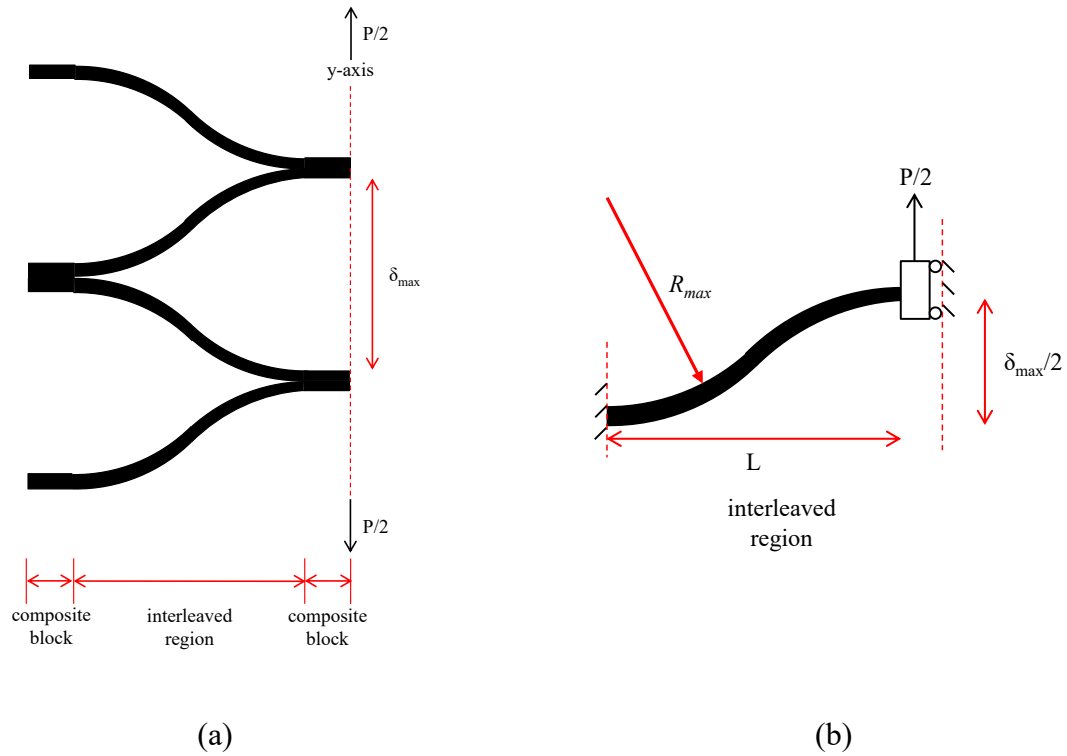


Figure 59. (a) A sketch of an expanded mesh structure with definitions of terms and (b) a simplified sketch.

3.2.4.3 Experimental investigation

A flat composite panel of 270 mm length and 150 mm width, as shown in Figure 60, was manufactured by hand layup. The layup configurations of interleaved regions and composite blocks are also shown in Figure 60.

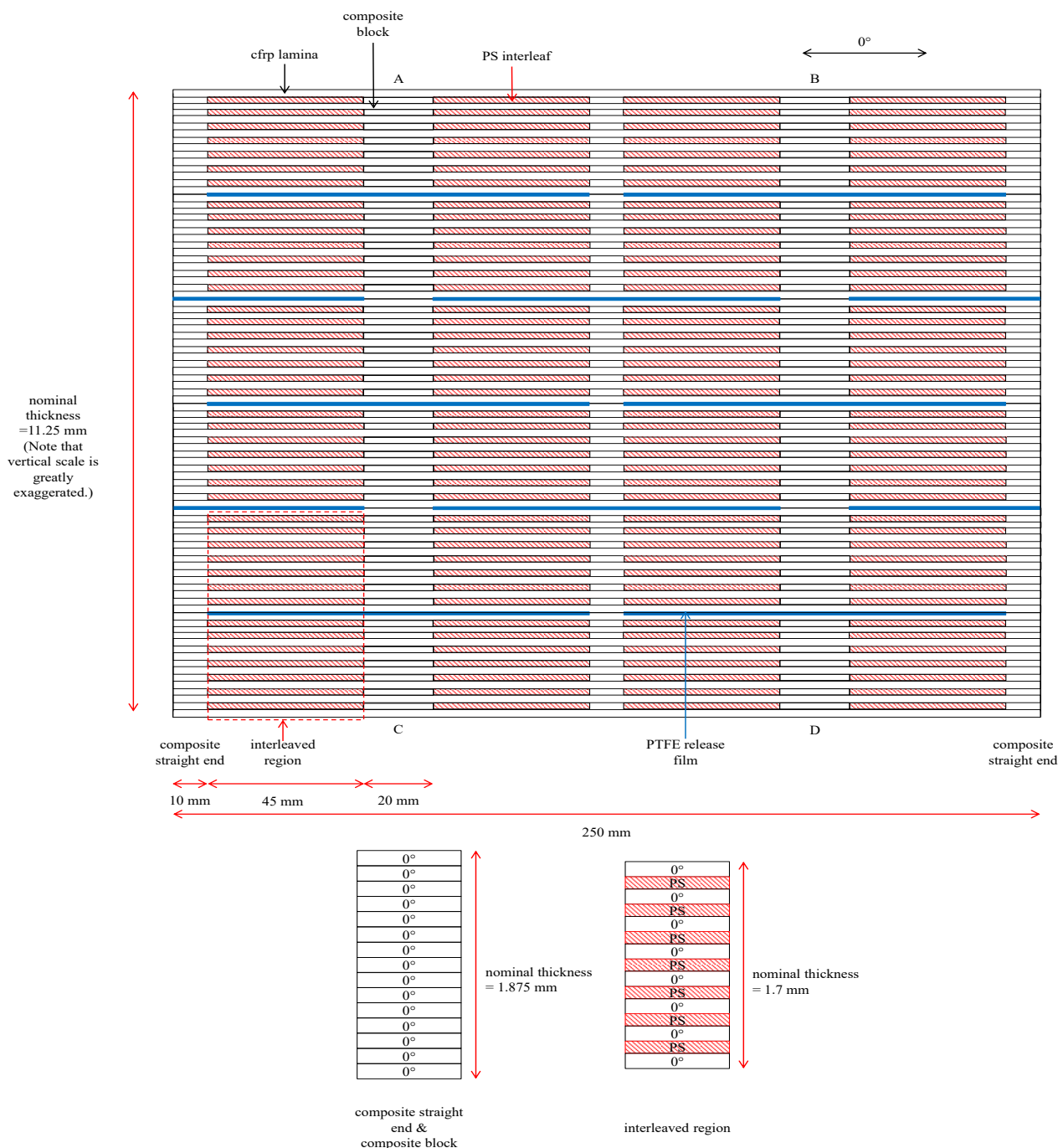


Figure 60. A longitudinal section of the mesh and layup details.

The panel was cured in an autoclave [Autoclave, LBBC, UK] at 175°C and a 100 psi pressure for 1 hour, according to the manufacturer's instructions. Specimens of 250 mm length and 20 mm width were prepared from the cured panel using the dry saw cutting machine. The 0° fibre direction was along the longer side of these specimens. They were sand blasted [Hand Blast Cabinet, West London

Shotblast centre, Ltd, UK] at the composite block segments A, B, C and D and then bonded to aluminium blocks (20 mm × 20 mm square and 10 mm in height) using a high temperature resistant adhesive [LOCTITE EA9514, Henkel, UK]. The assembly was placed in an oven and then heated at 120°C with vacuum for 1 hour to cure the adhesive. Figure 61 shows a specimen with aluminium blocks to be tested. It can be seen that the upper surface of the mesh was not flat; the composite blocks were slightly thicker than the interleaved region. This is due to the difference in thickness of the PS interleaves (0.1 mm) and carbon fibre/epoxy laminae (0.125 mm).

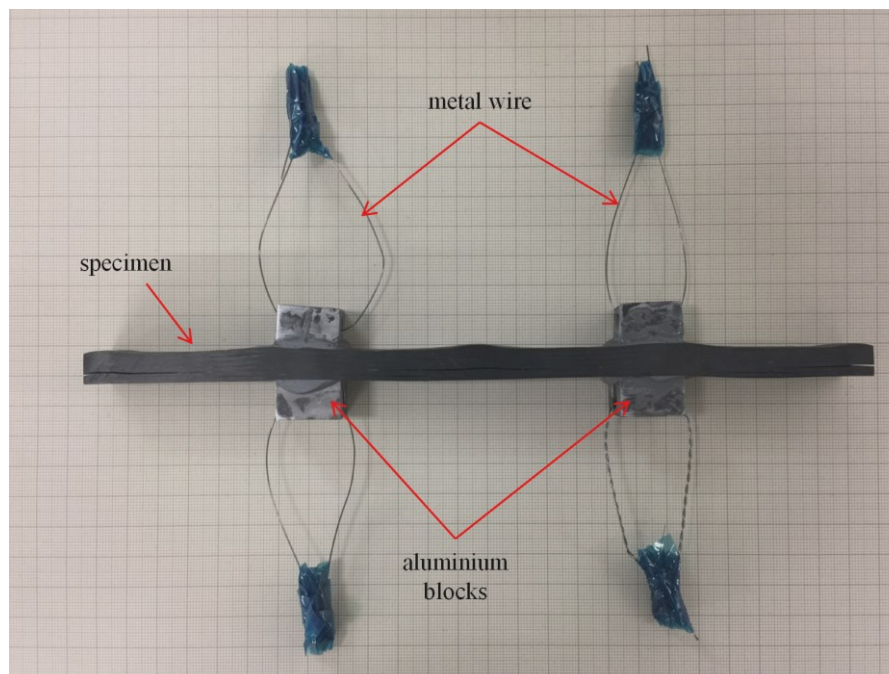


Figure 61. An expandable mesh specimen with aluminium blocks and metal wires to be tested.

As mentioned earlier, the flat specimens were re-shaped to the expanded mesh configuration. It was decided to vertically expand each unit cell by 60 mm, i.e. the mesh should be expanded by 180 mm in the vertical direction. To achieve this, two approaches were performed as follows.

In the first approach, a specimen was subjected to a uniaxial tension test on a universal mechanical testing machine [Instron 5985, Buck, UK] fitted with a 10 kN load cell, within an environmental chamber [SFL, Phoenix Calibration & Services Ltd, West Midland, UK]. The specimen was

connected to the test machine by using wire loops which passed through the aluminium blocks and around horizontal bars mounted on the machine loading arms (see Figure 62). A small pre-load (approximately 0.5 N) was applied to retain the specimen in tension. The specimen was heated to 120°C and kept for 15 minutes at this temperature. (A thermocouple was used to monitor the temperature of the specimen.) A tension load was then applied to expand the mesh at a constant crosshead displacement speed of 5 mm/min to the intended displacement of 180 mm. (However, due to the height limit of the test machine and the chamber capacity, a crosshead displacement of 84 mm was actually applied.) The expansion process of the specimen was recorded using the Pentax camera. The specimen was then cooled down to room temperature while maintaining the applied displacement, removed from the test machine and scanned in the flatbed scanner to determine the vertical extension.

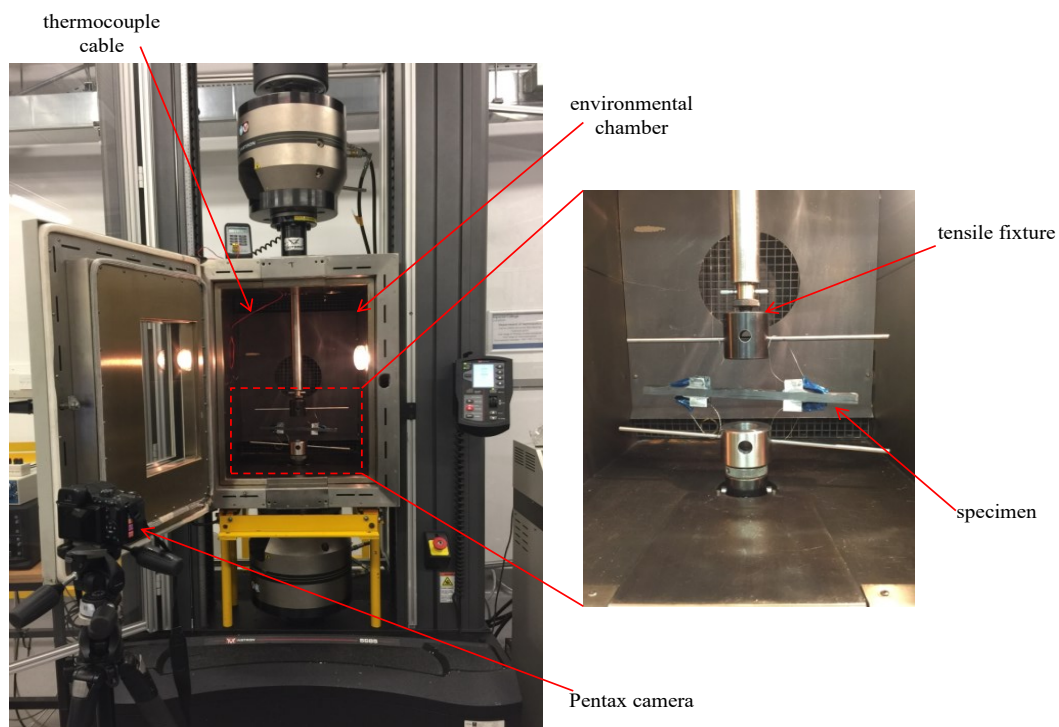


Figure 62. Experimental setup of the expansion test in tension mode.

In the second approach, a specimen was heated to 120°C in the chamber and kept at this temperature for 15 minutes. Each cell of the mesh was manually opened up and aluminium blocks (20 mm long, 10 mm wide and 54 mm height) were inserted into each open cell. The specimen was then cooled down to room temperature with the aluminium blocks to keep the mesh in the open configuration. Again, the cooled specimen was scanned on the flatbed scanner to measure the vertical extension after cooling.

To investigate the contraction capability, the expanded specimen was returned to the chamber previously heated to 120°C in an unconstrained state. At this temperature, the specimen would contract and recover towards the initial flat shape. This process was recorded using the Pentax camera until no significant shape change of the specimen was observed.

3.2.4.4 Results of mesh experimental investigation

Figure 63 shows the load-displacement response of a mesh specimen subjected to tensile loading at 120°C (i.e. the first approach). The figure shows that the load linearly increased when the actuator was expanding, with a number of small drops in load due to the slipping of the metal wires during the test. For the applied crosshead displacement of 84 mm, the measured load was around 20 N. No load-displacement relationship of a mesh specimen expanded by hand (i.e. the second approach) was measured.

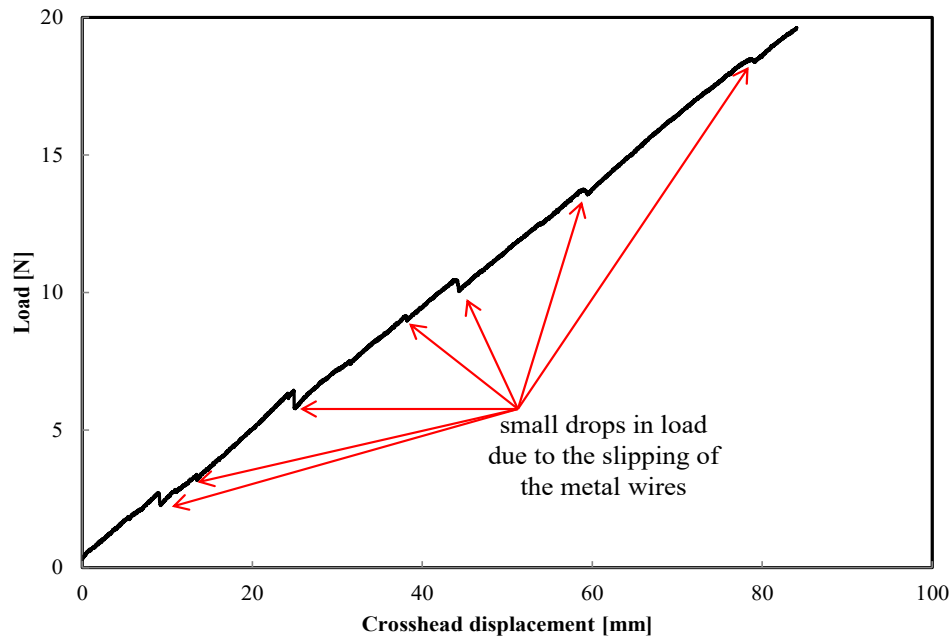


Figure 63. Load-displacement curve of an interleaved composite mesh subjected to a tensile test at 120°C.

Figure 64 shows expanded mesh specimens resulting from the two expansion approaches. The thickness of a unit cell (comprised of two interleaved laminates separated by PTFE releases film) was on average 3.2 mm and the overall thickness of the as-cured mesh was 11.11 mm at the composite blocks and 9.08 mm at the interleaved region. After re-shaping, for the mesh expanded using the first approach, each mesh cell opening was around 25 mm and the overall vertical extension of the mesh was 74.2 mm. For the specimen opened using aluminium blocks (i.e. the second approach), each mesh cell was opened by 54 mm and the overall vertical extension of the mesh was 162 mm.

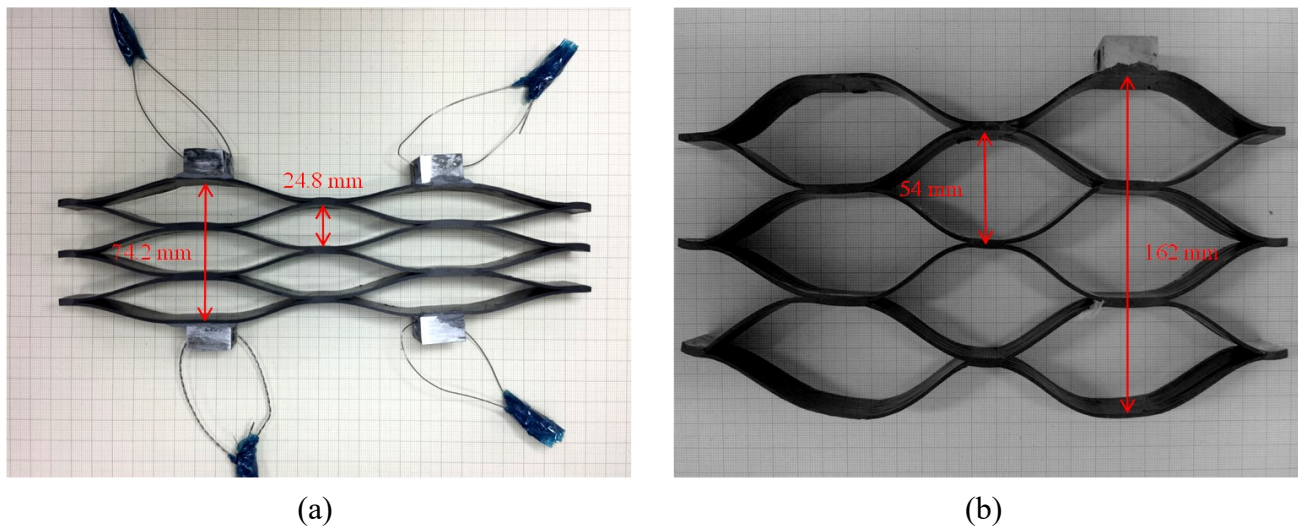


Figure 64. Images of interleaved composite mesh specimens expanded by (a) tensile testing and (b) aluminium blocks.

Figure 65 shows a series of frames extracted from the shape contraction video of two specimens from the expanded configuration. The specimen expanded using the first approach (see Figure 65a) took around 6 minutes to the original, flat shape after returning the specimen to the chamber at 120°C. (The shape recovery time was counted after the chamber door was closed. Therefore, at the beginning of this period, the chamber was recovering the temperature and the actual shape memory time was different to determine.) The specimen expanded using the second approach (see Figure 65b) also recovered in the similar manner.

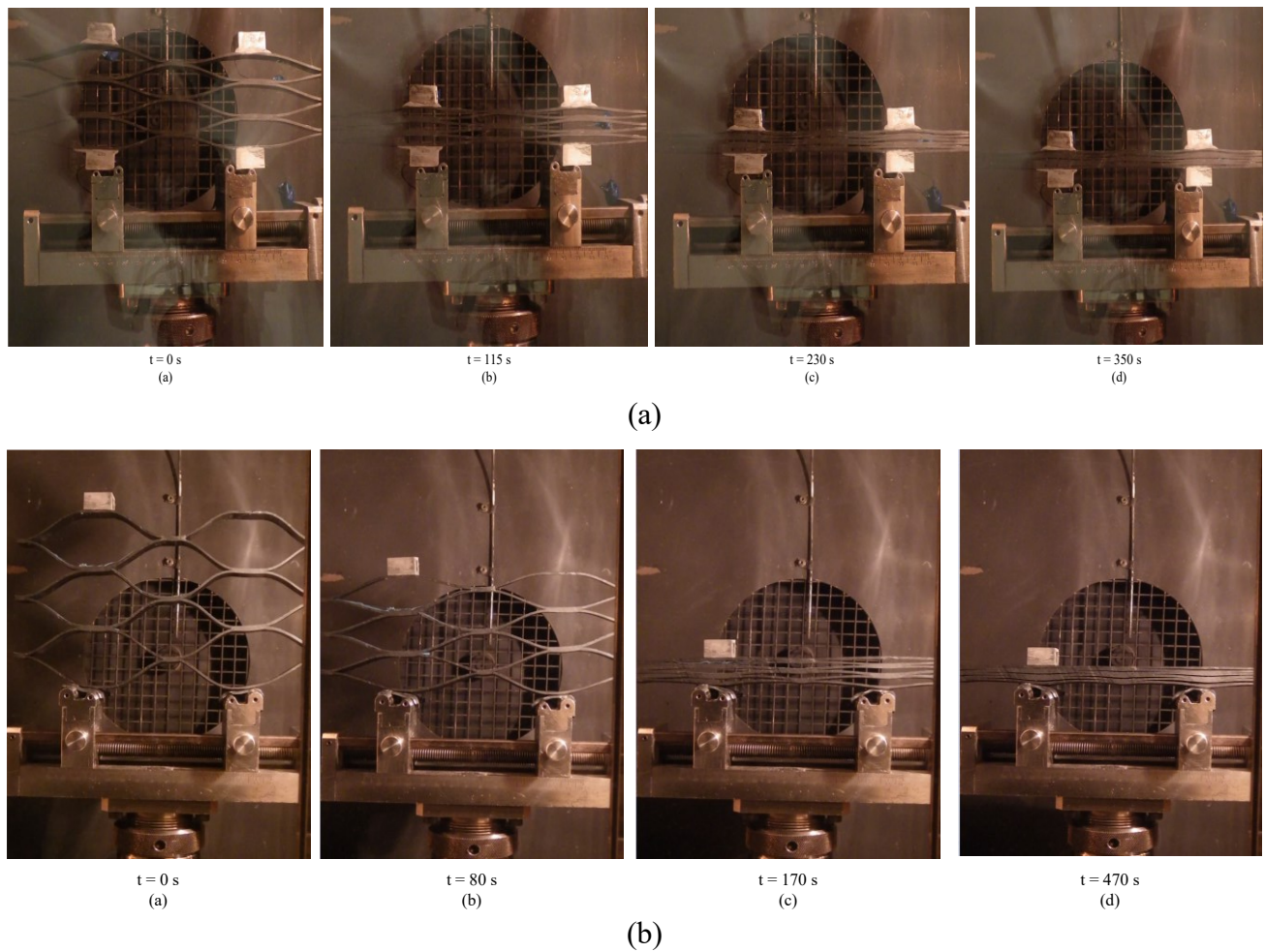


Figure 65. Still images extracted from the shape memory video of the expanded mesh using (a) test machine and (b) aluminium blocks.

Figure 66 plots the recovered vertical expansion versus time for the specimen expanded using the second approach on contracting to the original, flat shape. It can be seen that the contraction rate is initially rapid due to the release of the stored elastic stresses within the composite plies, but becomes slower, as the specimen slowly and continuously recovered to its initial shape.

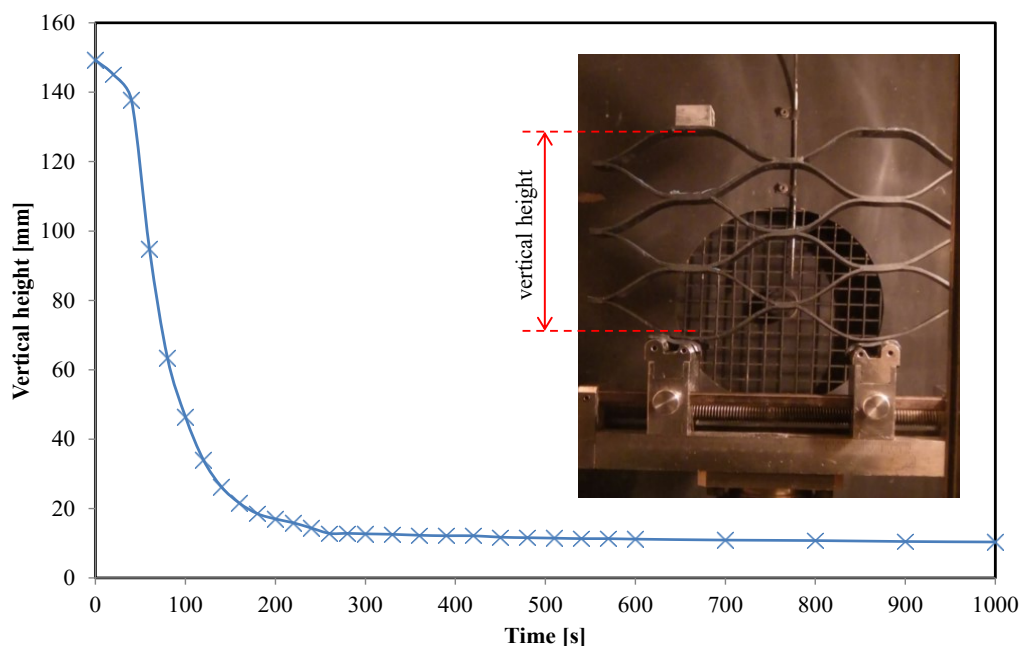


Figure 66. Vertical height versus time plot of the mesh specimen opened using the second approach at 120°C.

3.2.4.5 Discussion and summary of an interleaved composites mesh structure

It has been seen that a mesh structure made of interleaved carbon fibre/epoxy composites was re-shaped from a flat shape to an expanded mesh configuration when heated at elevated temperature. The manufacture of the mesh structure in the flat shape not only reduced the complexity of manufacture, but also ensured that the continuous carbon fibre/epoxy laminae were of the same length. No delamination in the mesh was observed after examination of expanded specimens. The high-temperature tensile load-displacement relationship (see Figure 63) showed that the mesh structure can be readily expanded by a low actuation force (the maximum force applied was 19.8 N). The crosshead displacement in this figure includes the extension of the wire loops. However, the actual opening after cooling is 74.2 mm minus 6.4 mm for the thickness of the interior solid composite blocks. The load, P_{180} , to cause the intended opening of 60 mm in each cell can be estimated by linear scaling to be 52.6 N. This is in good agreement with the load predicted from the simple beam theory of 55.4 N.

The expanded mesh structure recovered to its initial flat shape upon re-heating. Again, a similar conclusion can be drawn that the shape recovery is rate dependent due to the viscoelastic behaviour of the PS interleaves.

3.3 Summary

Thermoplastic-interleaved carbon fibre/epoxy composites have been used in three deployable configurations. In all cases, these interleaved composites exhibited excellent re-shaping and shape recovery capabilities. Very little springback was observed. However, delamination in the PS interleaves occurred in an integrated hinge specimen probably due to the over-tightening of the clamping mechanism which stopped internal ply sliding. Delamination was also found in a linear sine-wave actuator specimen due to through-thickness tensile stresses generated during the straightening of the waves. The third configuration was an initially flat interleaved configuration which could be pulled open at elevated temperature to form a mesh structure. In this configuration no delamination was observed.

In all three deployable configurations, the shape recovery rate became slower as the recovered shape was approached. This was due to the viscoelastic behaviour of PS interleaves. The viscoelastic behaviour of the PS interleaf will also influence the stresses developed in re-shaping. It is therefore important to understand the effect of viscoelastic behaviour on shape memory capability of these interleaved composite.

Chapter 4 An investigation of the viscoelastic behaviour of thermoplastic-interleaved composites

4.1 Introduction

This chapter describes the investigation of the viscoelastic behaviour at elevated temperature of polystyrene (PS) interleaf and of PS-interleaved carbon fibre/epoxy composites subjected to three-point flexure tests.

It has been observed that interior off-axis reinforced plies of 90°-bend PS-interleaved composite specimens showed damage after re-shaping at 120°C to 180°-bend or flat configurations (127). This damage may be due to the viscoelastic behaviour of the PS interleaf. A preliminary investigation was performed of the rate of re-shaping of a 90°-bend specimen to the 180°-bend configuration at elevated temperature (187). In this case the re-shaping was performed by placing the 90°-bend specimen on a 180°-bend mould, enclosing the whole assembly in a vacuum bag, and then after heating to 120°C, applying vacuum pressure at a controlled rate. It was observed that no interior composite ply damage occurred if the vacuum was slowly applied. This confirmed that the PS-interleaved composite exhibited a significant rate-dependence. Knowledge of the viscoelastic behaviour of the interleaved composite is important as it will enable determination of how rapidly an interleaved composite can be re-shaped without damage and how long a re-shaped component will take to deploy on re-heating.

4.2 Experimental investigation

4.2.1 Materials

A unidirectional carbon fibre/epoxy composite (HexPly, 914C-TS-5-34%, from Hexcel UK, with a cured ply thickness of 0.125 mm and a curing temperature of 175°C) was used in the current work. Polystyrene (Styrolution PS 124N/L, supplied as 0.1 mm thick films from TCKT, Austria) was selected as the interleaf material. Basic mechanical properties at room temperature of these materials are shown in Table 4.

Table 4. The room-temperature properties of the materials used in this study.

Materials	Property	Value
HexPly carbon fibre/epoxy composite	E_{11} , GPa	120 ^a
	E_{22} , GPa	8.5 ^a
	ν_{12}	0.32 ^b
	G_{12} , GPa	5.27 ^b
Styrolution 124N/L PS film (25°C)	E , GPa	3.2 ^c
	ν	0.35 ^d

^a both E_{11} and E_{22} of carbon fibre/epoxy composite are from manufacturer's datasheet

^b ν_{12} and G_{12} of carbon fibre/epoxy composite are from (185)

^c E of PS film is from manufacturer's datasheet

^d ν of PS film is from the website (<http://polymerdatabase.com/polymer%20physics/Poisson%20Table.html>) [Accessed date: 12.03.2018]

4.2.2 DMTA testing of the PS interleaf

Viscoelastic properties of the PS interleaf at elevated temperature were measured using a dynamic mechanical thermal analyser (DMTA) machine [RSA-G2, TA Instruments, UK]. (DMTA tests are commonly used to measure the temperature-mechanical properties of polymers and their composites in both static and dynamic loading modes (188, 189).) In this study, a temperature sweep test in tension mode was performed to measure the temperature-dependent elastic modulus of the material and an elevated-temperature frequency sweep test in shear mode was used to obtain the frequency-dependent shear modulus.

For the temperature sweep test, 0.1 mm thick rectangular PS films of 40 mm length and 6 mm width were prepared. Figure 67 shows the experimental setup of the temperature sweep test. A specimen was clamped by two stainless steel grips (with a gauge length between the grips of approximately 20 mm) and then the assembly was placed inside the built-in chamber. The specimen was subjected to a tensile test at a frequency of 6.28 rad/s (1 Hz) over a temperature range of 25°C to 120°C at a constant heating rate of 3 °C/min in air. A small pre-load (0.5 N) was applied to ensure that the specimen was initially in tension. The applied strain was cycled between 0% and a maximum of 0.1%. Both force and displacement were recorded by the test machine to determine the storage and loss elastic modulus.

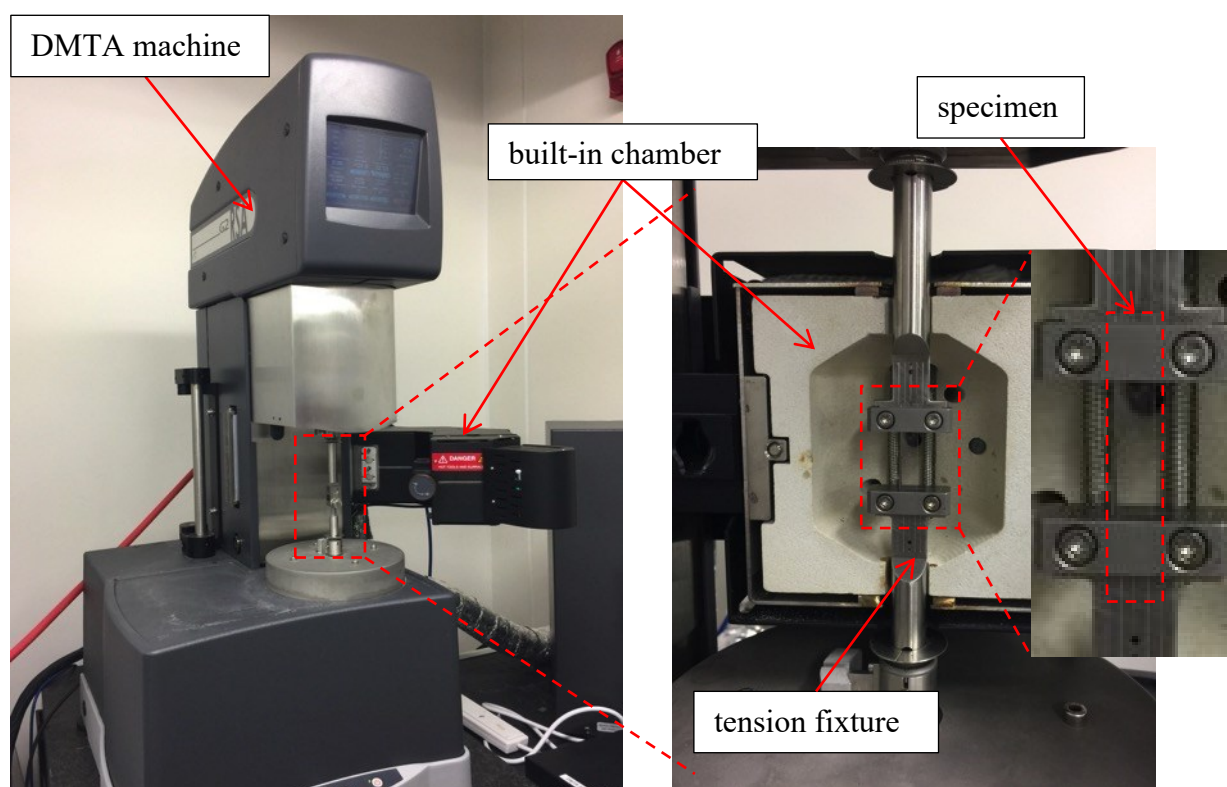


Figure 67. The experimental setup of temperature sweep test in tension mode.

For the elevated-temperature frequency sweep shear test, ten 0.1 mm thick polystyrene films (100 mm × 100 mm square) were stacked together and then placed on a metal plate in a vacuum bag together with release films and breather layer. The assembly was placed in an oven [Forced

convection oven, OF-21E, Lab Companion, Jeio Tech, Korea] at 120°C for 15 minutes with vacuum pressure to bond these films.

Rectangular PS specimens of 12 mm long, 16 mm wide and 1 mm thick (required by the DMTA test machine) were prepared. These specimens were tested in a lap joint configuration test as shown in Figure 68. Two PS specimens were gripped in a shear-sandwich gripping fixture fitted to the test machine. A screw was used to retain the specimens in between the central steel plate and the outer support plates. (The gap between each outer plate and the central plate was 1 mm, as required by the test machine.) The assembly was placed within the built-in chamber. During the test, the central steel plate driven in a sinusoidal manner with the intended frequency. Displacement and corresponding force were measured to determine the storage and loss shear modulus. The test was conducted at 120°C and the applied shear strain was cycled between 0% and a maximum of 1%. The applied frequency was in the range from 0.09 rad/s (0.014 Hz) to 11.9 rad/s (1.894 Hz). (These values were derived to be similar to the maximum shear strain rate predicted for the three-point bending tests described in the next section.)

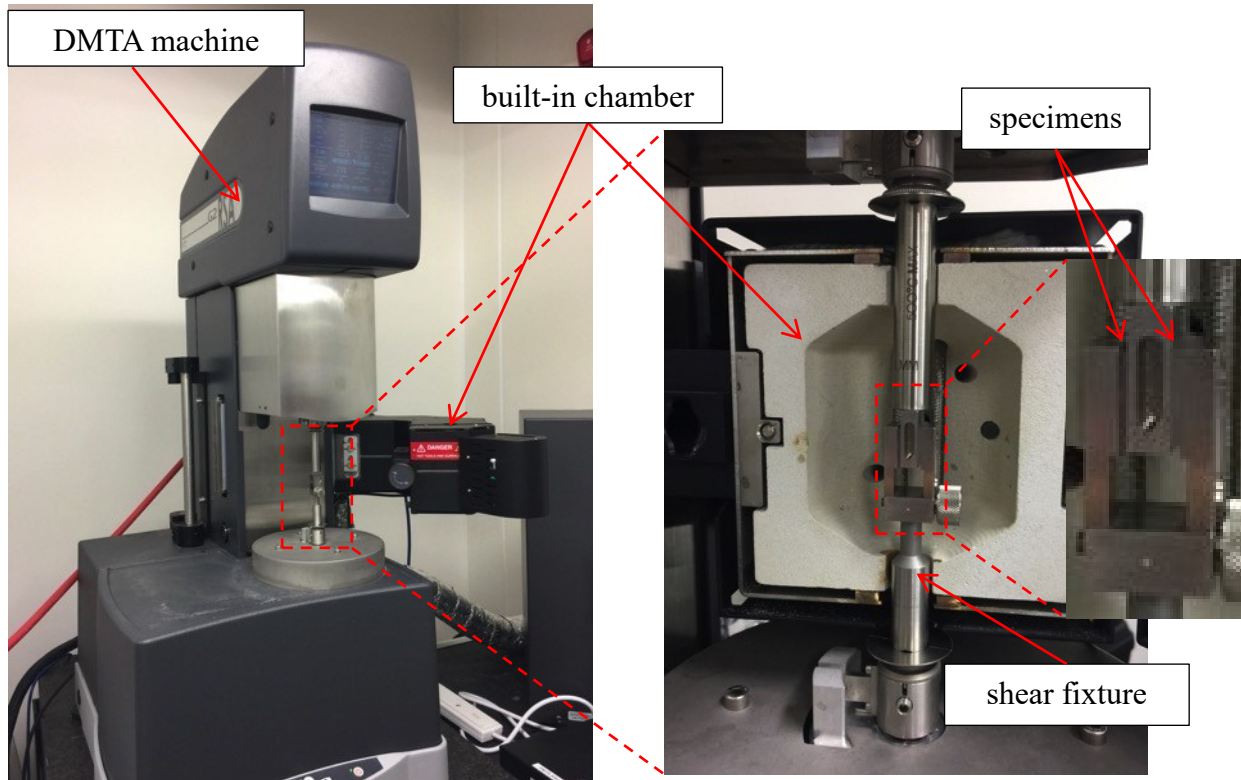


Figure 68. The experimental setup of frequency sweep test in shear mode at 120°C.

4.2.3 Flexure testing

4.2.3.1 Specimen design, manufacture and inspection

Laminates consisting of 8 plies of carbon fibre/epoxy laminae and 7 polystyrene interleaves in an alternating sequence were manufactured by hand layup. Three layup configurations, as shown in Figure 69, were prepared. These panels were 300 mm long in the 0° fibre direction and 150 mm wide.

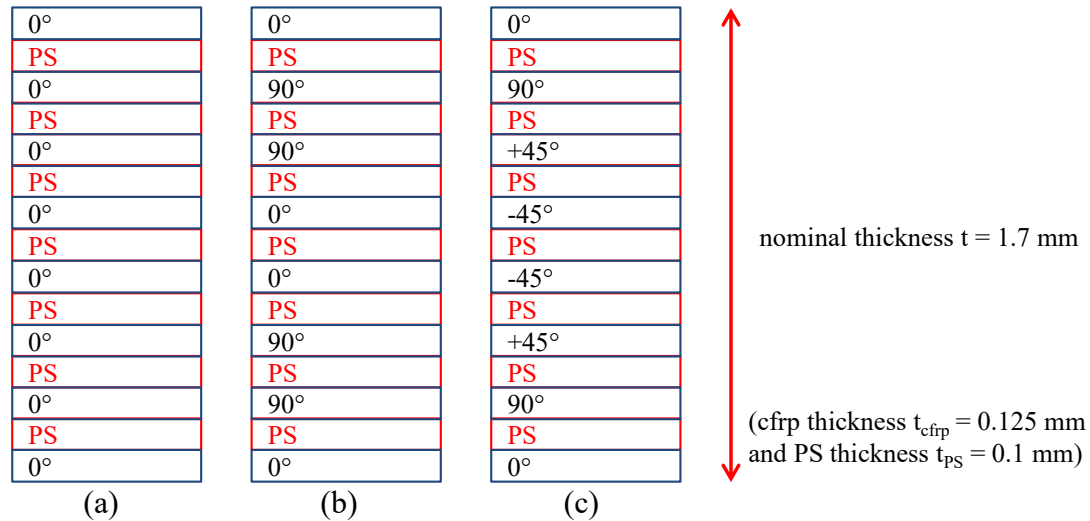


Figure 69. A sketch of the interleaved composite in (a) UD, (b) cross-ply and (c) quasi-isotropic layup configurations where blue layers are carbon fibre/epoxy laminae and red layers are PS interleaves.

These laminates were processed in an autoclave [Autoclave, LBBC, UK] at 175°C and a 100 psi pressure for 1 hour. After curing, these laminates were c-scanned ultrasonically in a water tank using a 10 MHz transducer [Triton 1700TT Immersion Tank, TecniTest]. Pulse-echo method was applied for the scanning.

To examine the cross sections of these laminates, small sections of these laminates were mounted in epoxy resin [Epoxicure, Buehler, UK] with a hardener [Kleer set hardener, 11 10 89, MetPrep, Buehler, UK] and cured for 8 hours at room temperature in air. The mounted sections were then polished in a grinder-polisher machine [MetaServ/MetaPol, Buehler, UK] with a SiC grinding paper in the grit level of P600 and 3 dispersions of diamond (9 μm), alumina (0.6 μm) and silco (0.03 μm) and then examined under an optical microscope [Zeiss, Axio Imager M2m, Germany].

Flexure specimens of 80 mm long and 10 mm wide were prepared from the cured laminates (see Figure 70) using a circular dry-saw cutting machine [Circular dry saw, Startrite, UK]. For these specimens, the 0° fibre direction was parallel to the longer edge of the specimens.

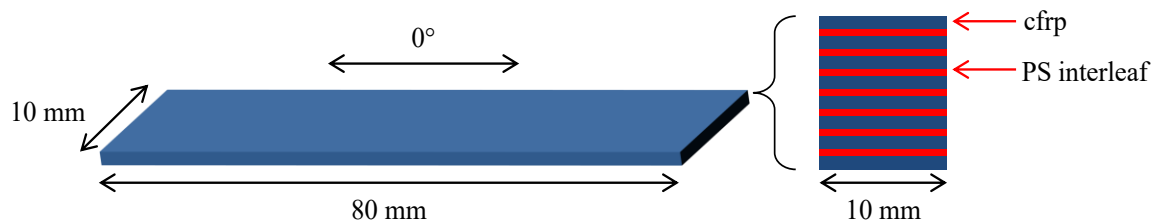


Figure 70. A sketch of an interleaved flexure specimen with the cross-section configuration.

4.2.3.2 Quasi-static flexure tests

Flexure testing was conducted to investigate the viscoelastic behaviour but first the quasi-static flexural stiffness change of the interleaved composite on heating to 120°C was measured. A PS-interleaved specimen was subjected to three-point bending tests according to the ASTM D7264 standard. It was placed on a three-point bending fixture at a support span of 64 mm. Rollers of 6.4 mm in diameter were used. The test was performed on an Instron universal test machine [Instron 5985, Buck, UK] fitted with a 1 kN load cell within an environmental chamber [SFL, Phoenix Calibration & Services Ltd, West Midland, UK]. A thermocouple was used to measure the temperature of the flexure specimens. An iMetrum video gauge system [iMetrum, Bristol, UK] was used to monitor the central vertical displacement of the flexure specimens during testing. Figure 71 shows the experimental setup of the three-point flexure test.

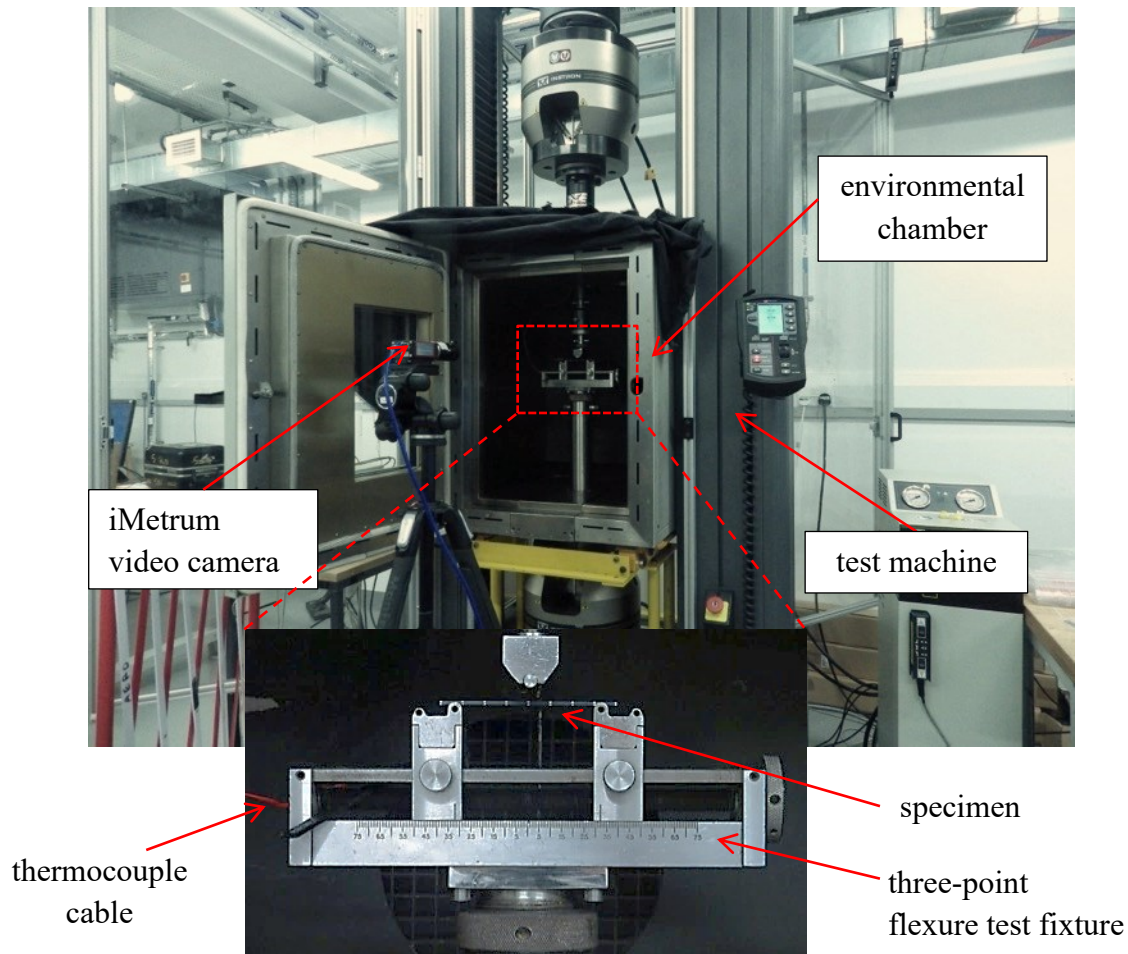


Figure 71. The experimental setup of the three-point flexure test.

For the observation of the flexural stiffness change of PS-interleaved composites, flexure specimens were initially loaded at room temperature to a 1.5 mm displacement at 1 mm/min and then unloaded at the same rate. This speed was chosen according to the ASTM D7264 standard. The specimens were then heated to 120°C in the environmental chamber and kept at this temperature for 15 minutes. The heated specimens were loaded to a 5.0 mm displacement at 1 mm/min and then unloaded at the same displacement rate. After this, the specimens were cooled down to room temperature, re-heated to 120°C and re-loaded to the 5.0 mm displacement at 1 mm/min and finally unloaded at the same speed. The test procedure is sketched in Figure 72. Load was recorded by the Instron software and vertical central displacement was monitored by the iMetrum video gauge system. Five specimens of each layup configuration were tested.

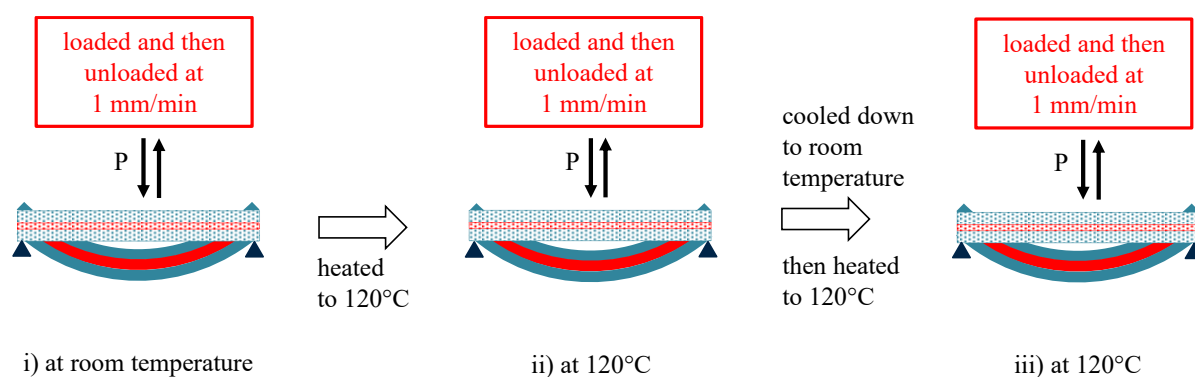


Figure 72. A sketch of the test procedure to observe the stiffness change of PS-interleaved specimens subjected to flexure test at room temperature and 120°C.

4.2.3.3 Rate-effect flexure tests

Three-point flexure tests were also performed to investigate of the viscoelastic behaviour of the PS-interleaved composite at elevated temperature. In this case, flexure specimens were initially heated to 120°C in the environmental chamber and kept at this temperature for 15 minutes. Each specimen was loaded to a 5.0 mm displacement at a crosshead speed of 1 mm/min which was obtained from the ASTM D7264 standard. The specimen was held at this displacement for 10 minutes and any change in applied load was recorded by the Instron software. The specimen was then rapidly unloaded at a crosshead displacement rate of 400 mm/min. Upon unloading, the specimens recovered towards the original straight configuration. The iMetrum video gauge system was used to monitor the central displacement of the specimen throughout the test. Figure 73 summarises the procedure of the rate-effect test and identifies the different test phases. The whole test was then repeated for a sequence of increased loading crosshead speeds of 10, 100 and 200 mm/min. Finally, the whole test was repeated at the initial loading rate of 1 mm/min. Five specimens of each layup configuration were tested.

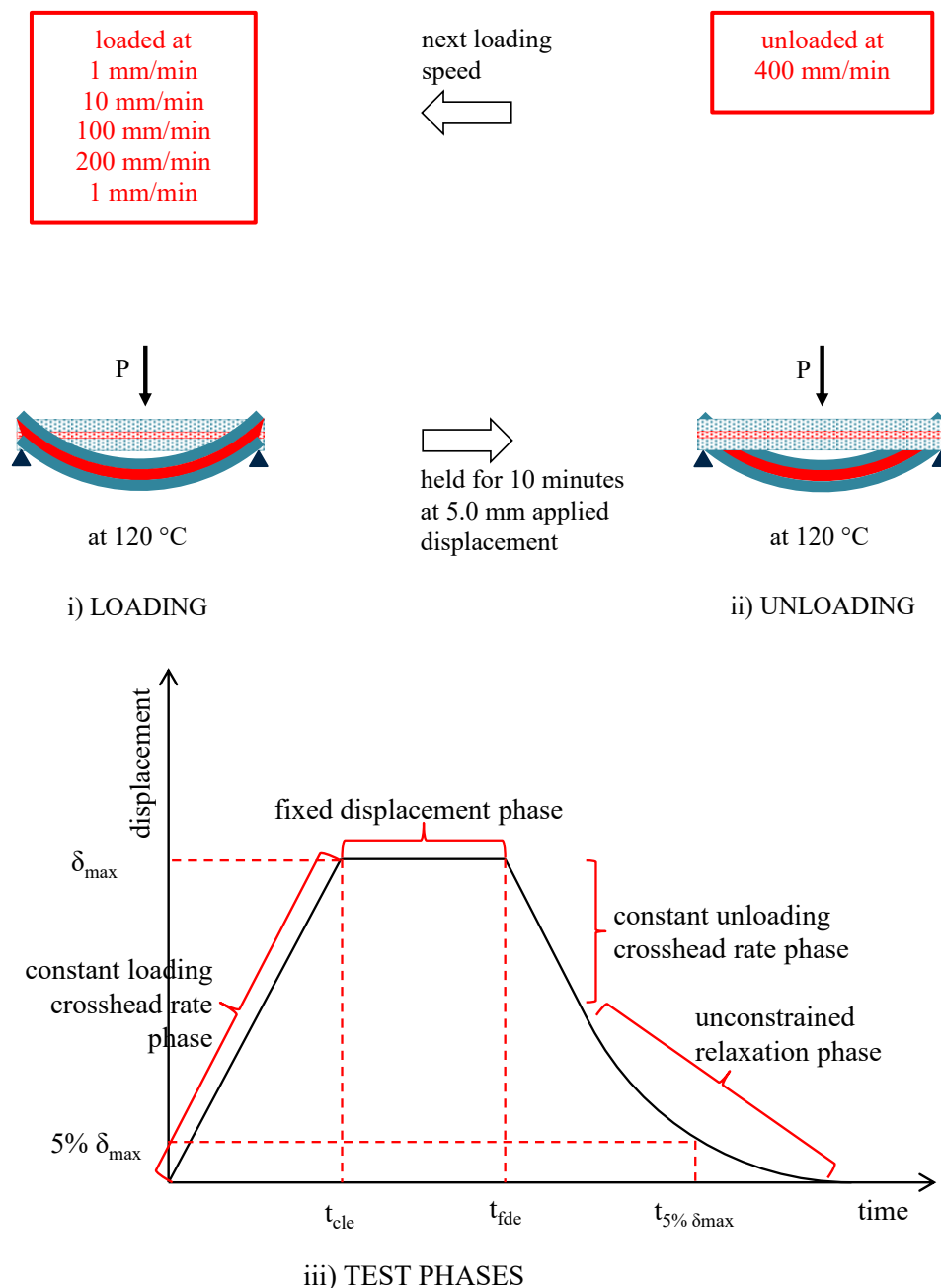


Figure 73. A sketch of the test procedure to investigate the rate effect of PS-interleaved specimens subjected to flexure test at 120°C.

4.3 Results and discussion

4.3.1 Results of DMTA tests

Figure 74 shows the measured storage and loss elastic moduli versus temperature plot of a PS specimen subjected to a tensile test at a constant frequency of 6.28 rad/s (1 Hz) in DMTA. It can be

seen that the storage modulus (E') is around 3 GPa at 25°C, gradually reduced to 2.5 GPa at 60°C, but undergoes increasingly rapid reduction as the temperature is further increased. At temperatures higher than 110°C the storage modulus is less than 1 MPa (i.e. a reduction of over 99% in storage modulus from the low temperature value). This indicates that the PS interleaf possesses the stiffness variation capability on heating. The figure also shows that the glass transition temperature (T_g), taken as the temperature value at the peak point of the loss modulus (E'') curve, of the PS interleaf is around 85°C. This is in good agreement with the value measured from Differential scanning calorimetry (DSC) of the PS interleaf at 87°C (190).

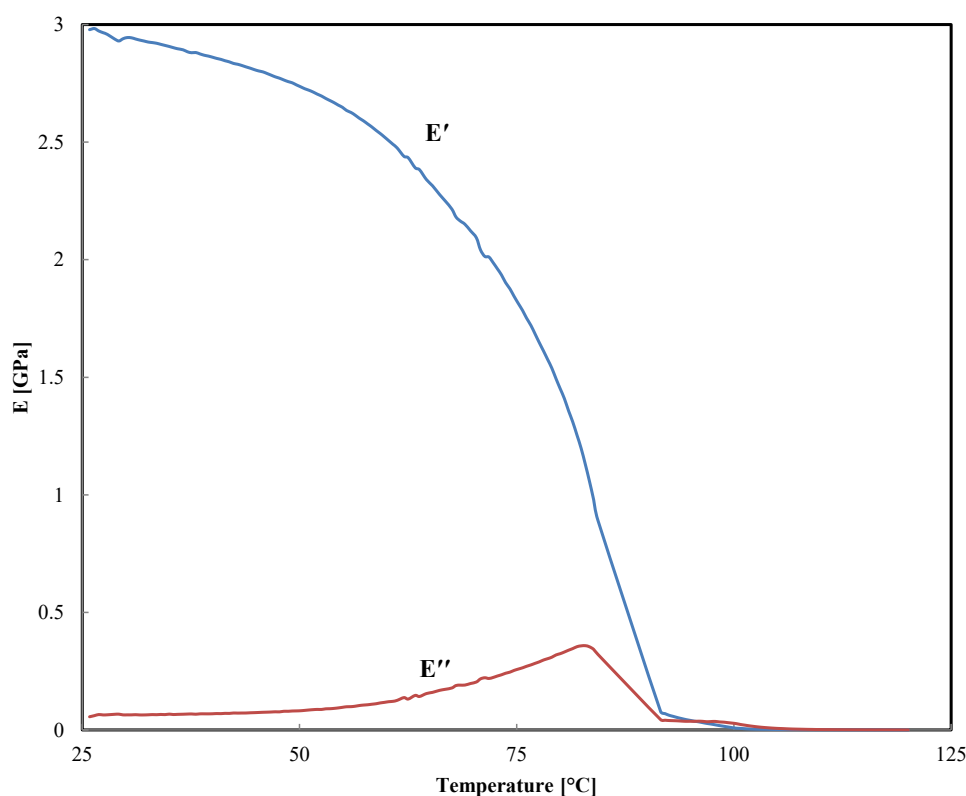


Figure 74. Typical storage and loss elastic modulus versus temperature plot of a PS interleaf specimen subjected to tensile test.

Table 5 summarises the storage modulus at 25°C and 120°C for all PS interleaf specimens. The PS interleaf shows an average room temperature storage modulus of 2.14 GPa. A large scatter of the storage modulus results can be seen in the table. The reason for this deviation of the elastic modulus

is unknown and further work is required to investigate this. At 120°C, all specimens show a reduction in the storage modulus over 99%. This indicates that the PS interleaf possesses the stiffness variation capability on heating.

Table 5. Storage modulus of PS interleaf specimens at 25°C and 120°C measured from DMTA tension tests.

Specimen number	Storage modulus E' (GPa)		% loss*
	25°C	120°C	
1	1.75	0.000286	99.98
2	1.61	0.0016	99.90
3	1.88	0.000239	99.99
4	2.50	0.000624	99.98
5	2.97	0.000535	99.98
Average	2.14	0.000657	99.97

$$* \% \text{ loss} = \frac{E'_{120^\circ\text{C}}}{E'_{25^\circ\text{C}}} \times 100$$

Figure 75 plots the storage and loss shear moduli versus applied frequency for a PS specimen loaded in shear at 120°C. It can be seen that both storage and loss shear modulus (G' and G'' respectively) increase with increasing frequency. The material exhibits a solid-like, elastic dominant behaviour, with less work done to specimen dissipated. The loss shear modulus exceeds the storage shear modulus at around 12 rad/s (1.91 Hz). This implies that the PS interleaf transforms to the viscous region at this frequency. Beyond this crossover point, the material behaves in a predominantly viscous manner and more work done to specimen are dissipated as heat.

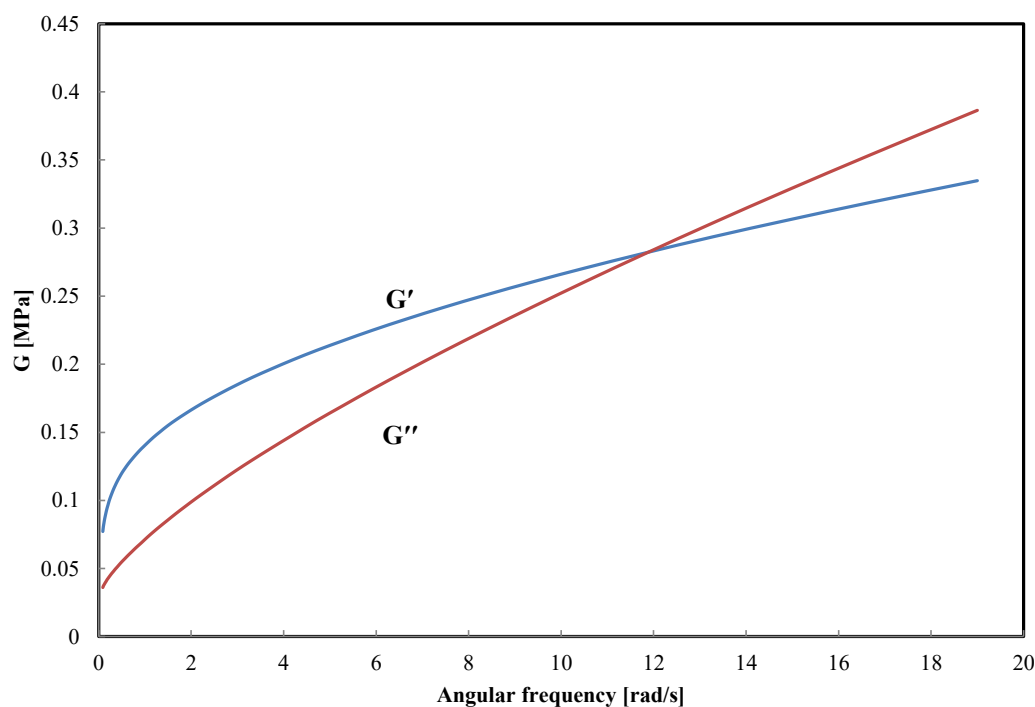


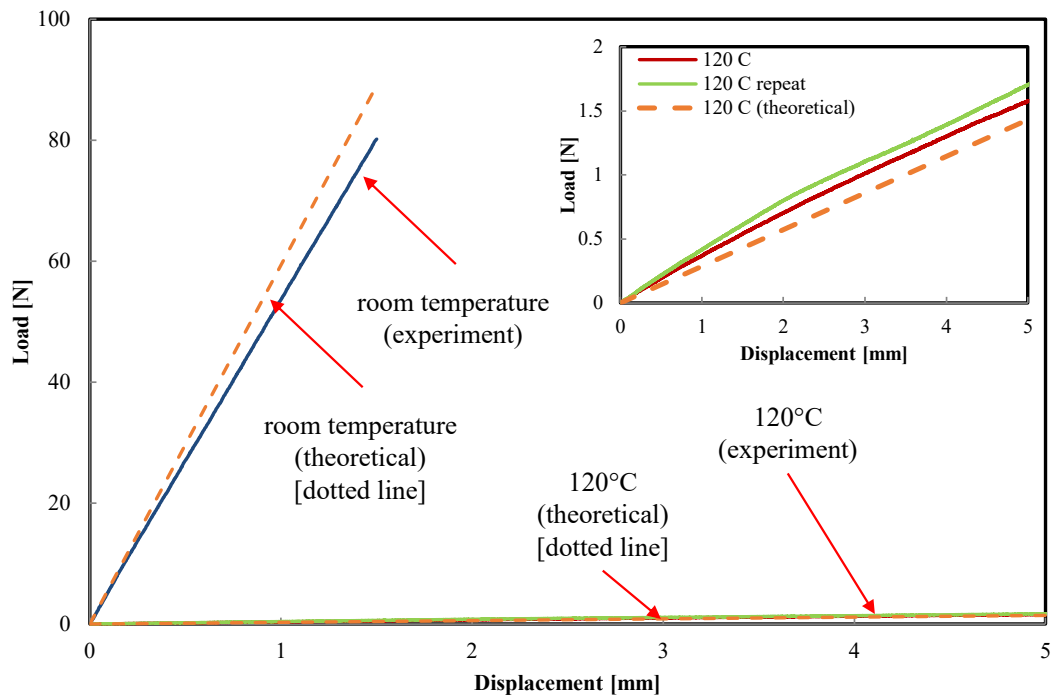
Figure 75. Typical storage and loss shear modulus versus frequency response of a PS interleaf specimen subjected to shear test at 120°C.

4.3.2 Results of flexure test

4.3.2.1 Quasi-static tests

Figure 76 shows the experimentally observed typical load-displacement responses of the interleaved specimen subjected to three-point bending tests at a constant crosshead speed of 1 mm/min at room temperature and at 120°C. Table 6 summarises the results from all the flexure specimens tested. For all three layup configurations, the results show that on heating, the effective flexural modulus reduces by over 99%. This is consistent with previous work (5). The repeat elevated-temperature test shows that this large loss in stiffness is again achieved after cooling to room temperature and re-heating to 120°C. The experimentally measured effective flexural moduli are not in perfect agreement with the theoretical predictions which are also included in Table 6. (Theoretical predictions of the flexural modulus of interleaved composites at room temperature and 120°C are shown in Appendix A.1) At room temperature, the average effective flexure modulus for each layup

configuration is lower than the predicted value. This may be due to additional displacement resulting from shear distortion of the PS interleaves not included in the theoretical prediction. Whereas at 120°C, the measured flexure moduli are in good agreement with the theoretical prediction shown in Appendix A.1.



(a)

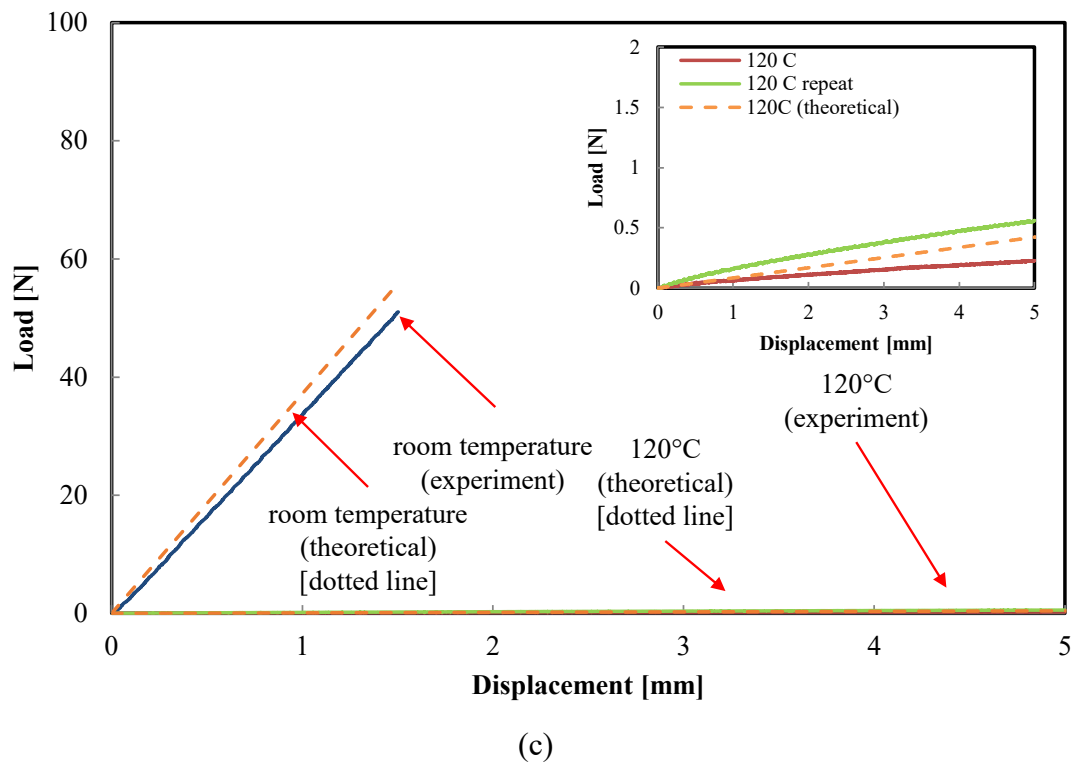
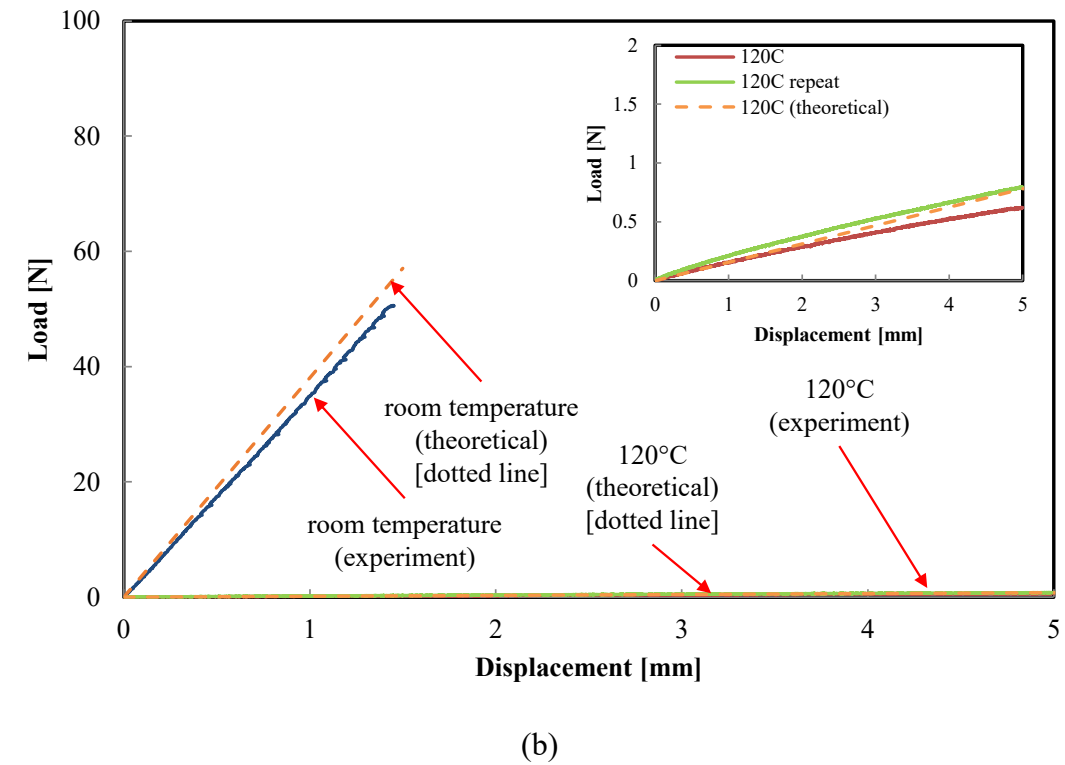


Figure 76. The load-displacement response of a PS-interleaved specimen in (a) UD, (b) cross-ply and (c) quasi-isotropic layup configurations loaded in three-point bending at 1 mm/min. (The sub-figures show the elevated-temperature load-displacement response with a magnified load scale.)

Table 6. The effective flexural moduli at room temperature and 120°C for the interleaved composite of three layup configurations.

	UD		Cross-ply		Quasi-isotropic	
	Room temperature	120°C	Room temperature	120°C	Room temperature	120°C
^a Average E (GPa)	70.8	0.39	48.3	0.18	46.2	0.09
Standard deviation	4.1	0.04	5.8	0.05	2.8	0.03
^a Theoretical E _f (GPa)	79.1	0.38	49.6	0.20	50.6	0.12
^b % loss of room temperature E _f		99.5		99.6		99.8

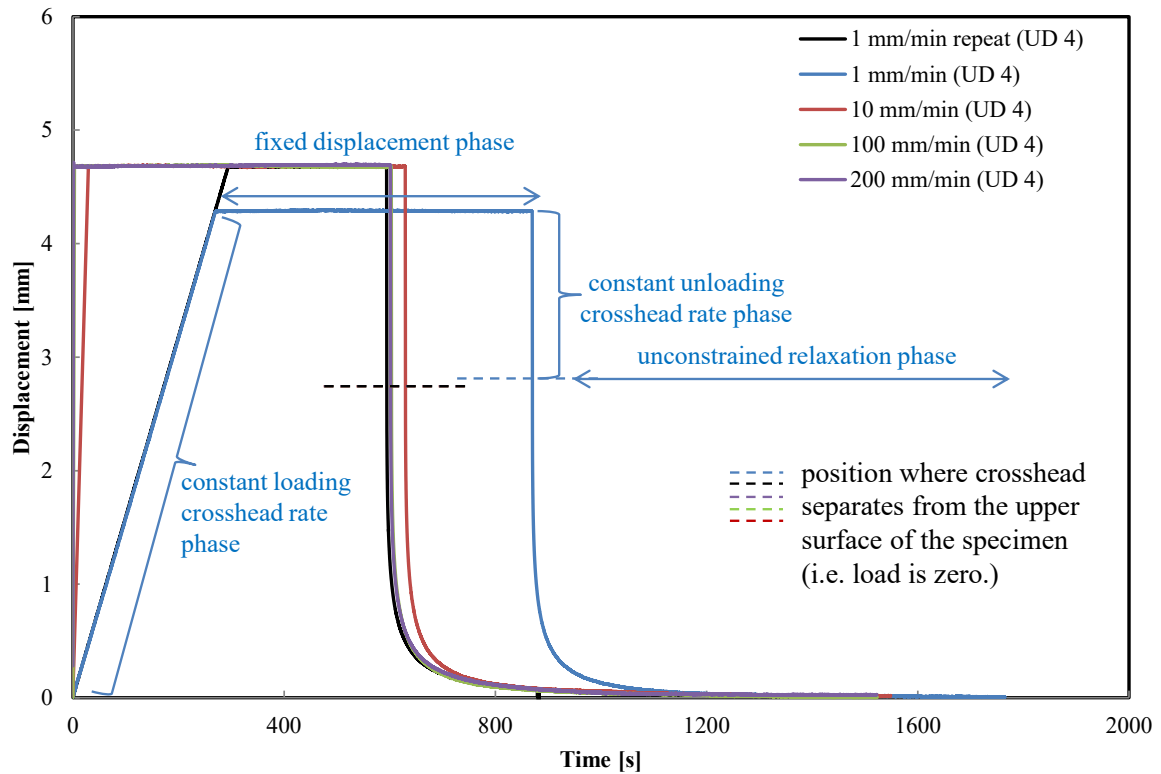
^a the calculation of both measured E and theoretical E_f is shown in appendix A.1

^b % loss of room temperature E_f was calculated using the following equation

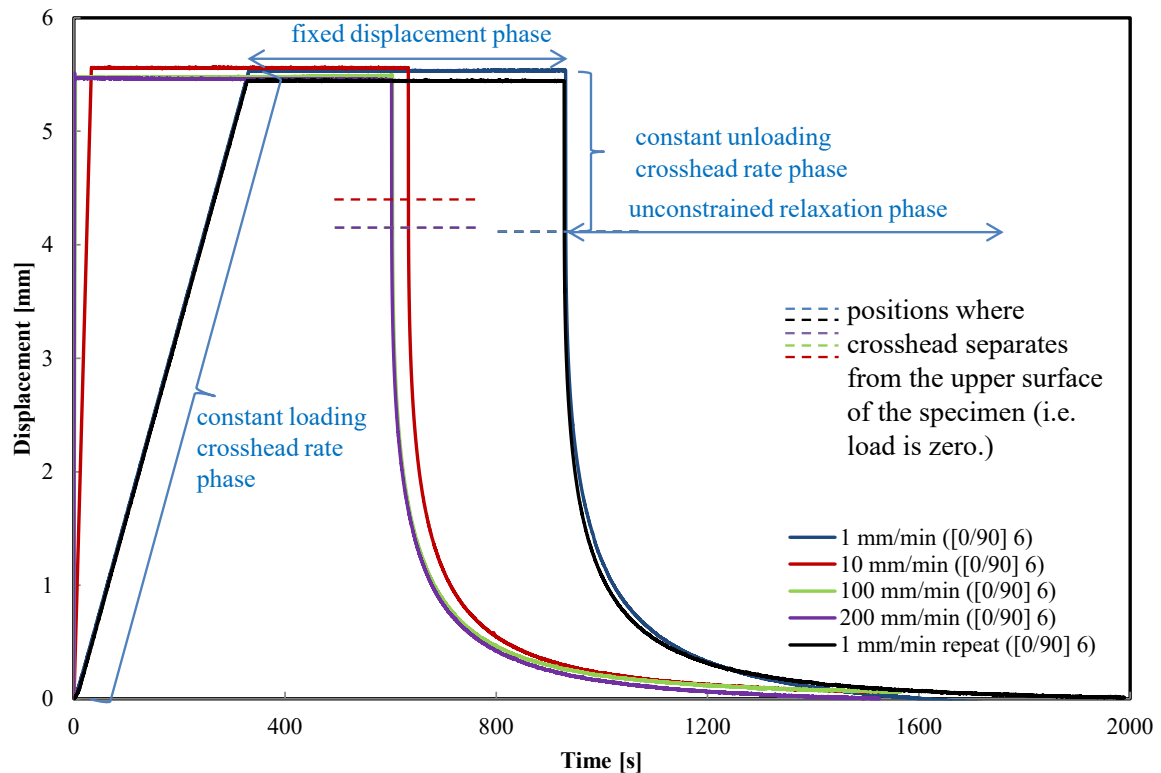
$$\% \text{ loss of room temperature } E_f = \frac{\text{Average } E_f \text{ at } 120^\circ\text{C}}{\text{Average } E_f \text{ at room temperature}}$$

4.3.2.2 Rate-effect tests

Figure 77 shows sample displacement versus time plots from one specimen of each layup loaded at 120°C and at different crosshead speeds. The phases of the rate-effect flexure test (i.e. constant loading crosshead rate, fixed displacement, constant unloading crosshead rate and unconstrained relaxation – see Figure 73) are readily identified (these are marked for the 1 mm/min loading case for each of the layups in Figure 77 as an example). The difference in the maximum applied displacement of the plots shown in Figure 77 (particularly evident for the UD specimen) was due to the slightly different starting points for the crosshead above the upper surface of each specimen. (The applied 5.0 mm displacement was programmed to start from the initial position of the crosshead. The zero displacement displayed in Figure 77 and other figures in this section, corresponds to the displacement when the crosshead contacted the specimens.) Figure 77 includes, for each plot, the position of the point during the unloading phase when the crosshead loading pin separates from the specimen (i.e. when the applied load returns to zero). Beyond this position the specimen is in the unconstrained relaxation phase.



(a)



(b)

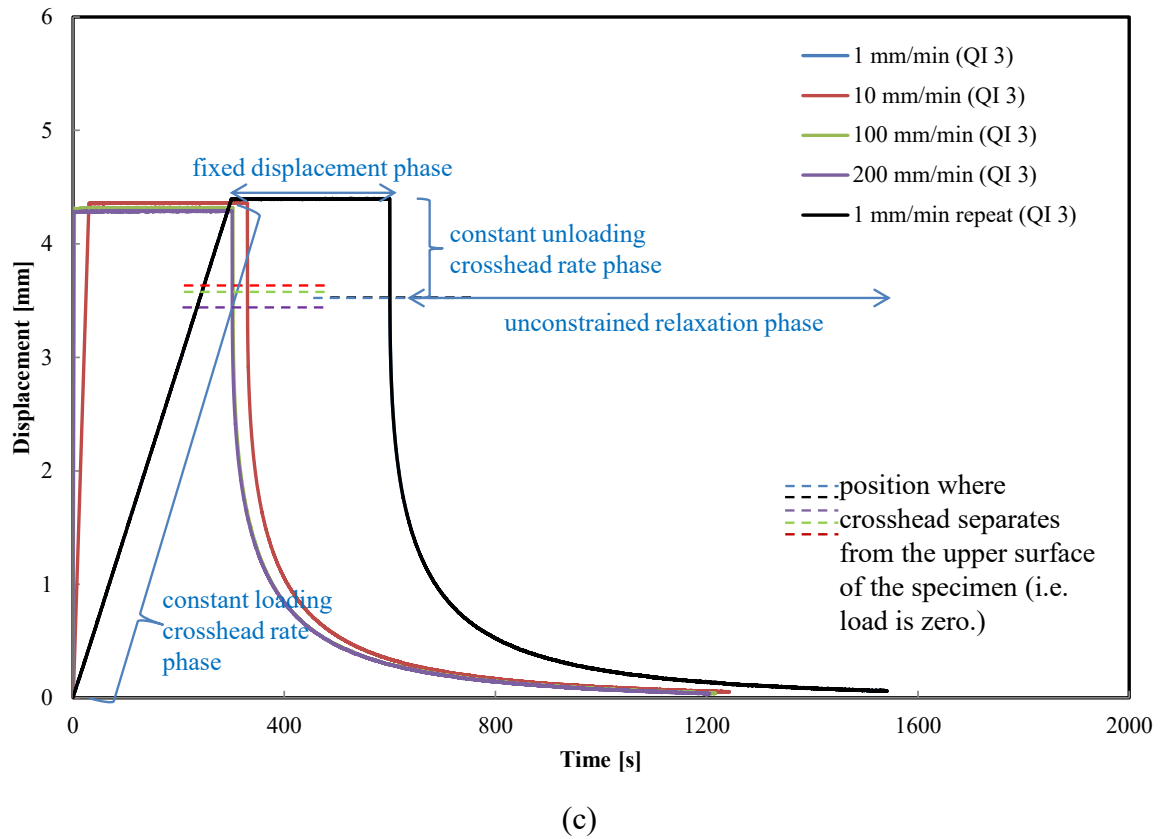
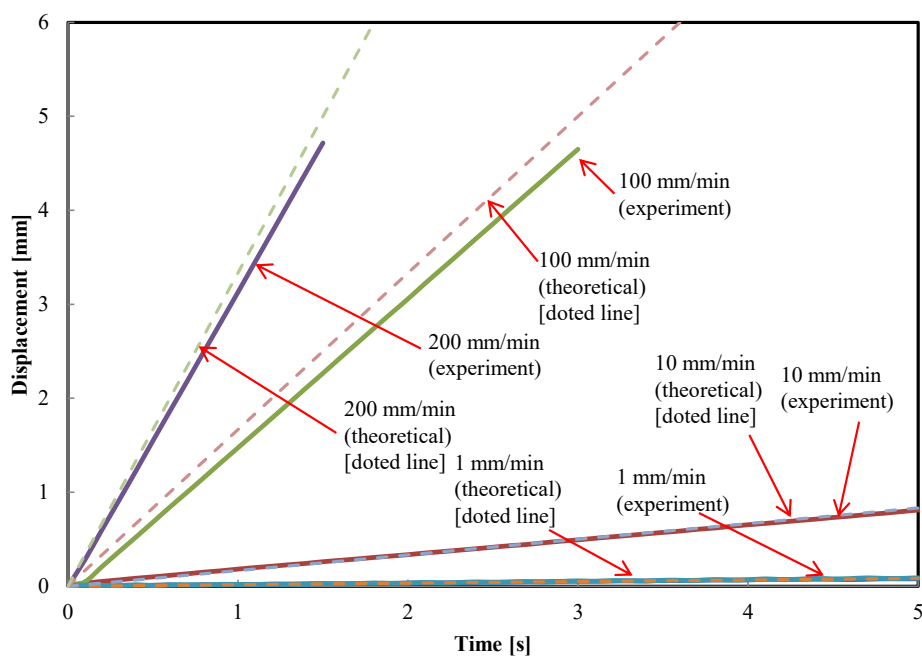
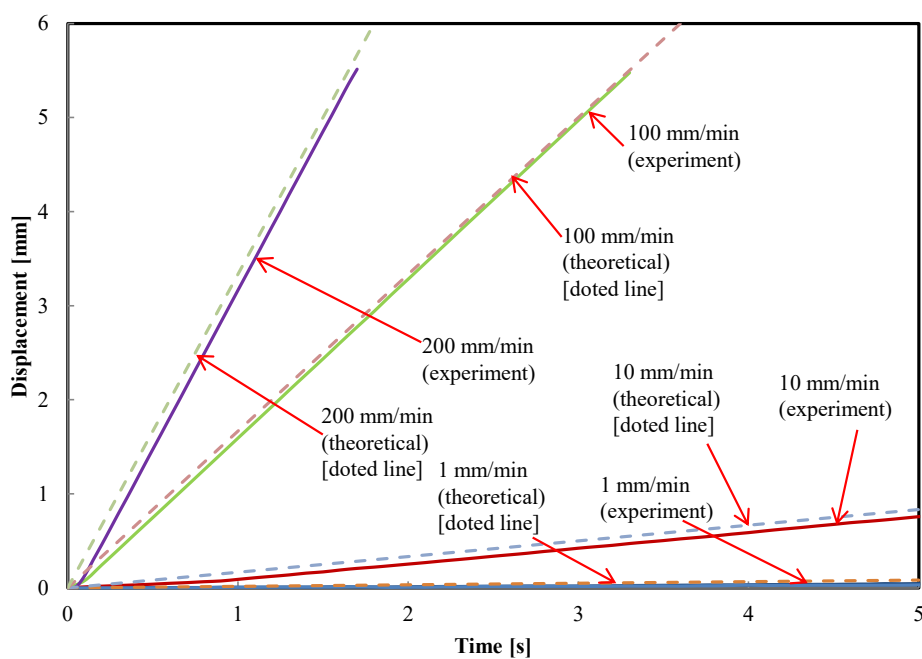


Figure 77. Displacement-time curves of one interleaved specimen in (a) UD, (b) cross-ply and (c) quasi-isotropic layup configuration loaded at 120°C and at different loading speeds.

Figure 78 shows displacement-time plots with expanded time scales to check on the linearity (or otherwise) during the initial few seconds of the displacement-controlled loading phases. For each plot the theoretical (i.e. intended) displacement-time line is also shown. The plots show in all cases excellent linearity and good agreement with the theoretical lines.



(a)



(b)

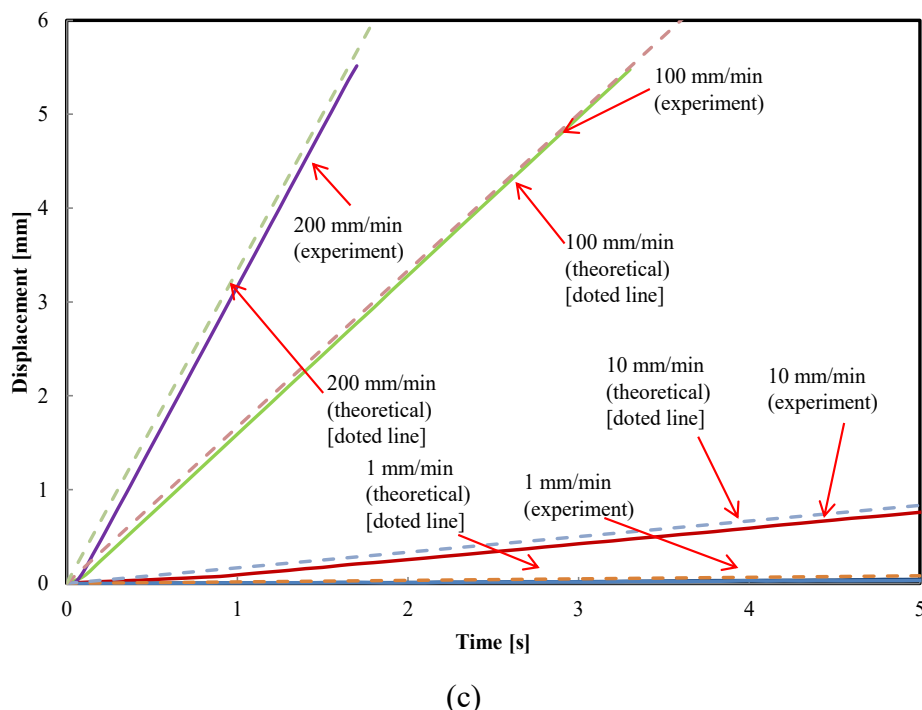
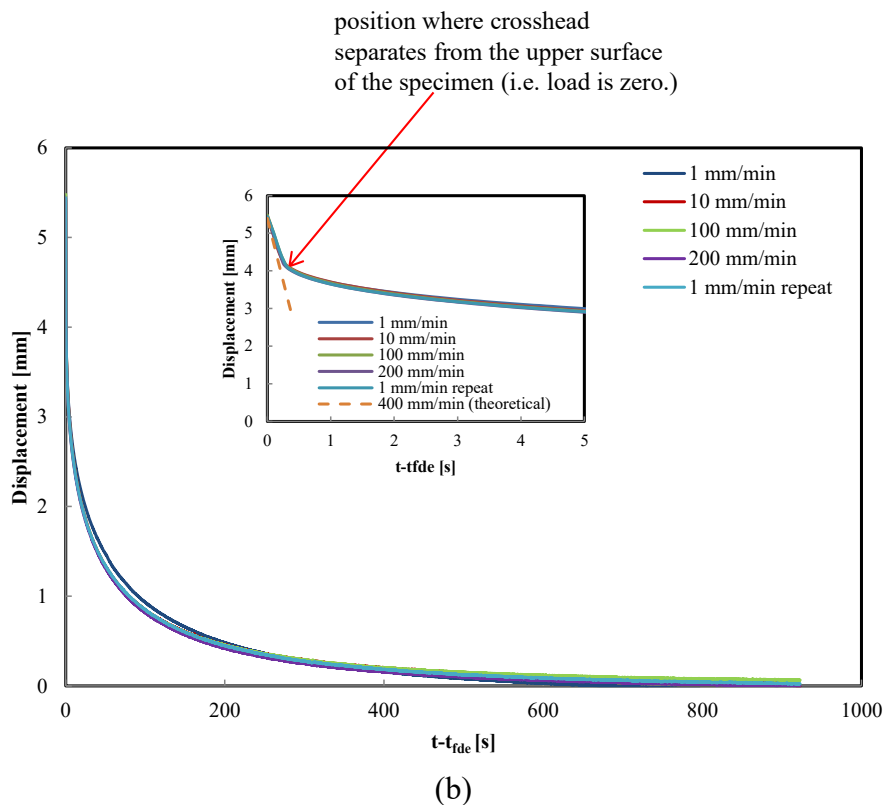
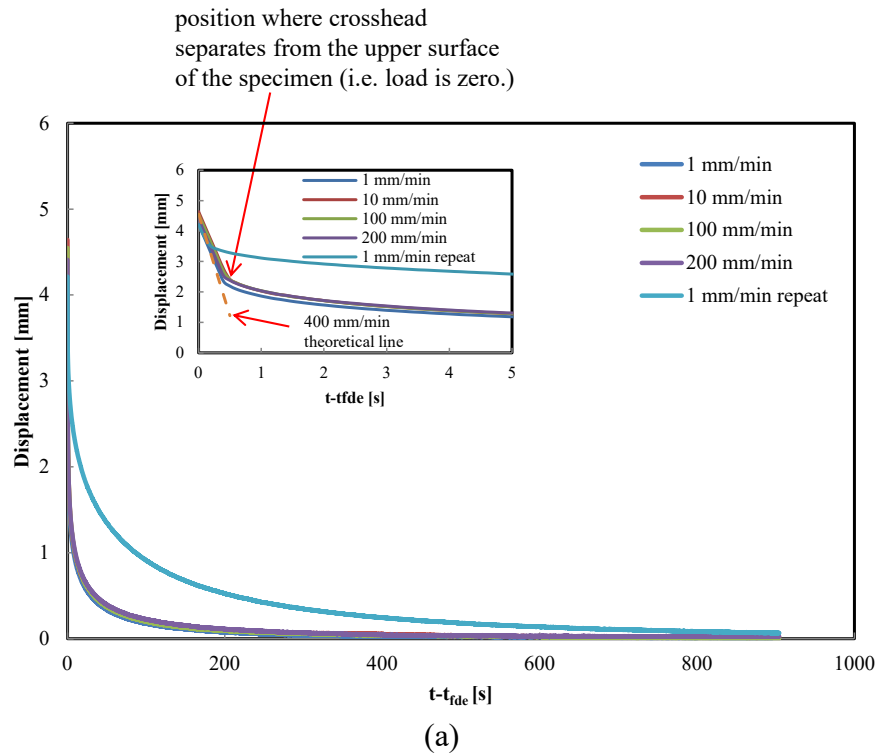


Figure 78. Plotted displacement-time response of an interleaved specimen in (a) UD, (b) cross-ply and (c) quasi-isotropic layup configuration for the initial loading and unloading phases to show the actual crosshead speeds.

Figure 79 shows the displacement during the unloading and unconstrained relaxation phases plotted against the time after the end of the fixed displacement phase (i.e. plotted against $t - t_{fde}$ where t is time measured from when the loading pin first contacts the specimen and t_{fde} is the time at the end of the fixed displacement phase). It can be seen that, for a given layup, the response during the unloading and unconstrained relaxation phases are almost identical for all loading speeds. (Only the repeated 1 mm/min UD specimen test is different and it is suspected that the specimen before unloading was not in the equilibrium state as the specimen was kept at the fixed displacement for 5 minutes.) This indicates that the fixed displacement phase was of a sufficiently long duration to ensure that all specimens had returned to the same stress state. The sub-figures in these plots clearly show the initial constant unloading rate phase and these are all in good agreement with the intended 400 mm/min unloading rate (again, except for the repeated 1 mm/min UD specimen). These plots also confirm that in the final phase, the specimens recover towards the original shape for all layup configurations, and exhibits a typical overdamped response.



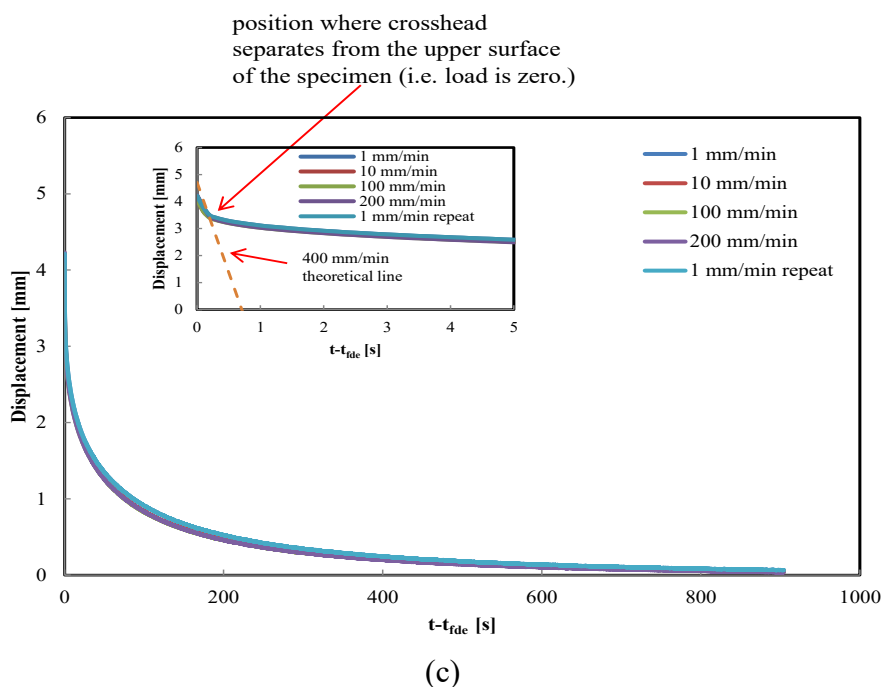
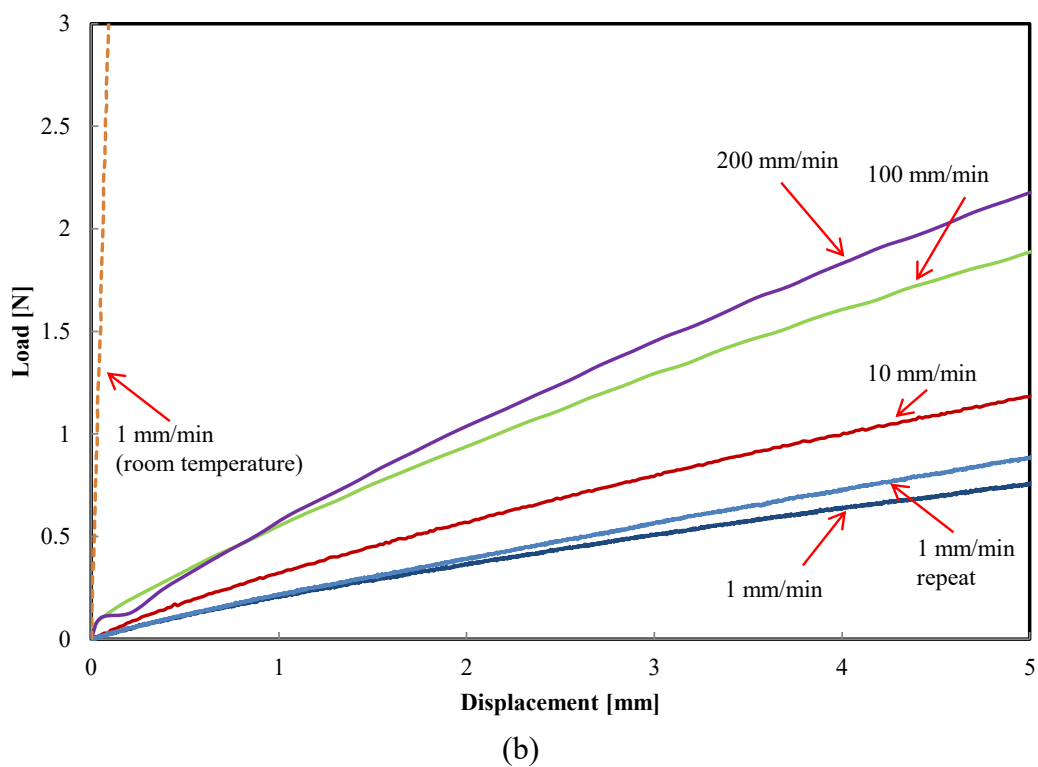
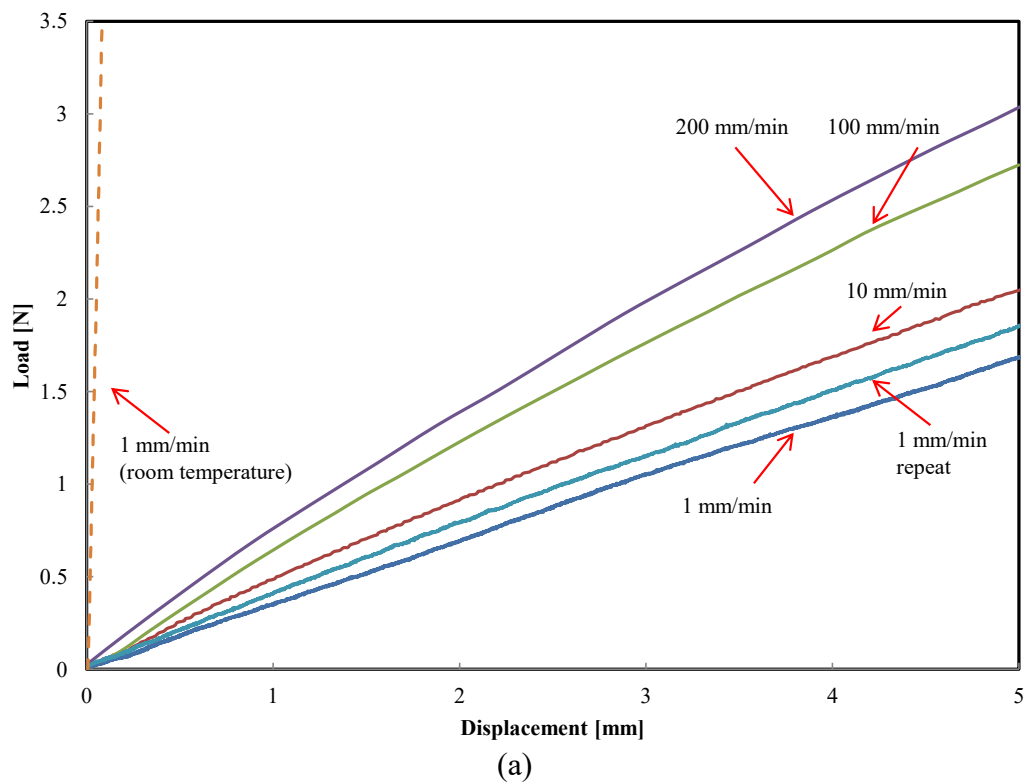


Figure 79. Typical displacement-time plot of an interleaved specimen in (a) UD, (b) cross-ply and (c) quasi-isotropic layup configuration during the unloading and unconstrained relaxation phases (The time is the $t-t_{fde}$ region and starts from zero.)

Figure 80 shows the load-displacement curves determined for the loading phases of the flexure tests. These curves confirm that there is a rate dependence due to the viscoelastic behaviour of the polystyrene interleaf (the initial slope increases with increasing displacement rate). Such rate dependence indicates that the force required to re-shape such an interleaved composite and the stress developed within the composite will depend on the speed of the re-shaping process. (A higher re-shaping speed will lead to greater stresses developed in the laminate.) Figure 81 plots the effective flexural modulus of the specimens as a function of crosshead speed (The effective flexure modulus was determined using the slope of the best-fit straight line between 1 mm and 2 mm applied displacement.). This figure clearly indicates that the flexural modulus increases with increasing loading speeds.



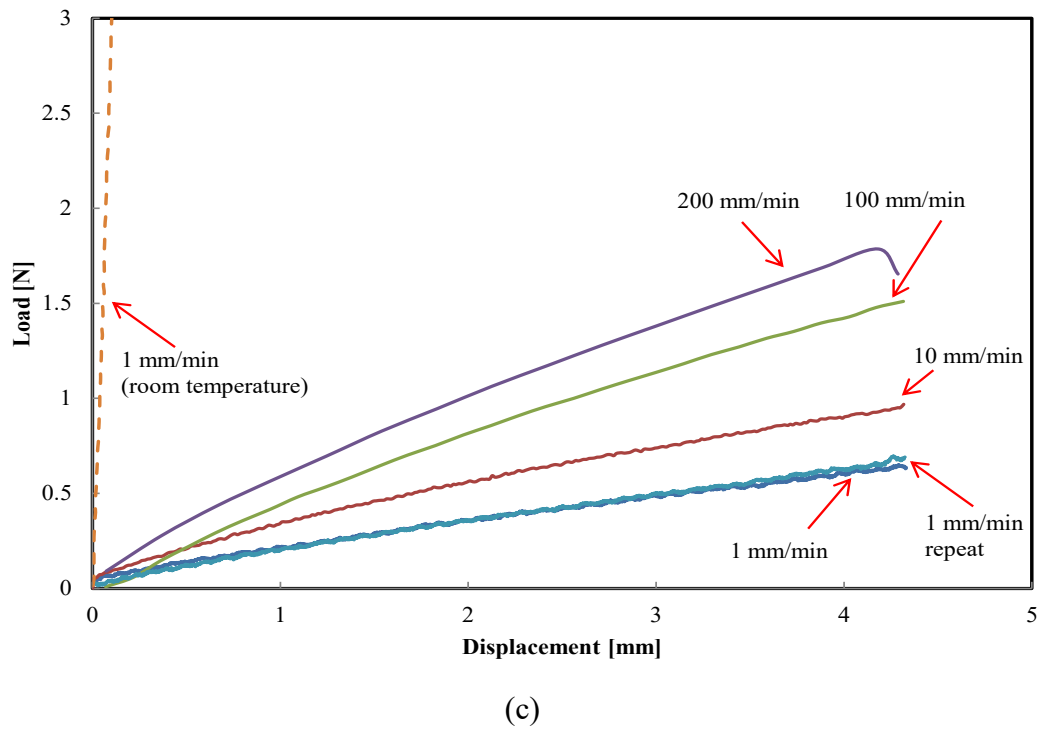


Figure 80. Typical load-displacement curves of an interleaved flexure specimen in (a) UD, (b) cross-ply and (c) quasi-isotropic layup configuration.

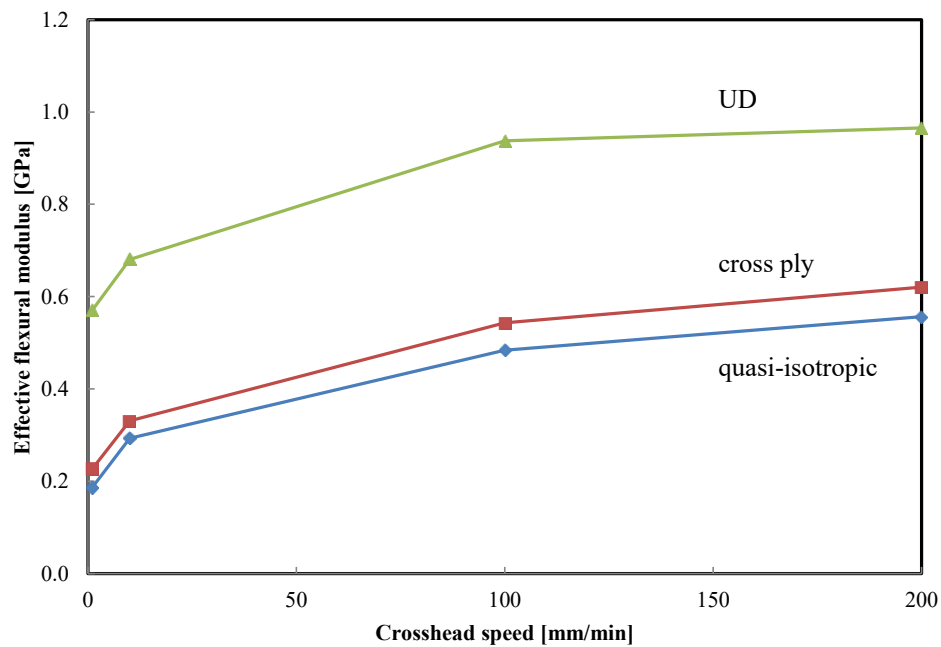
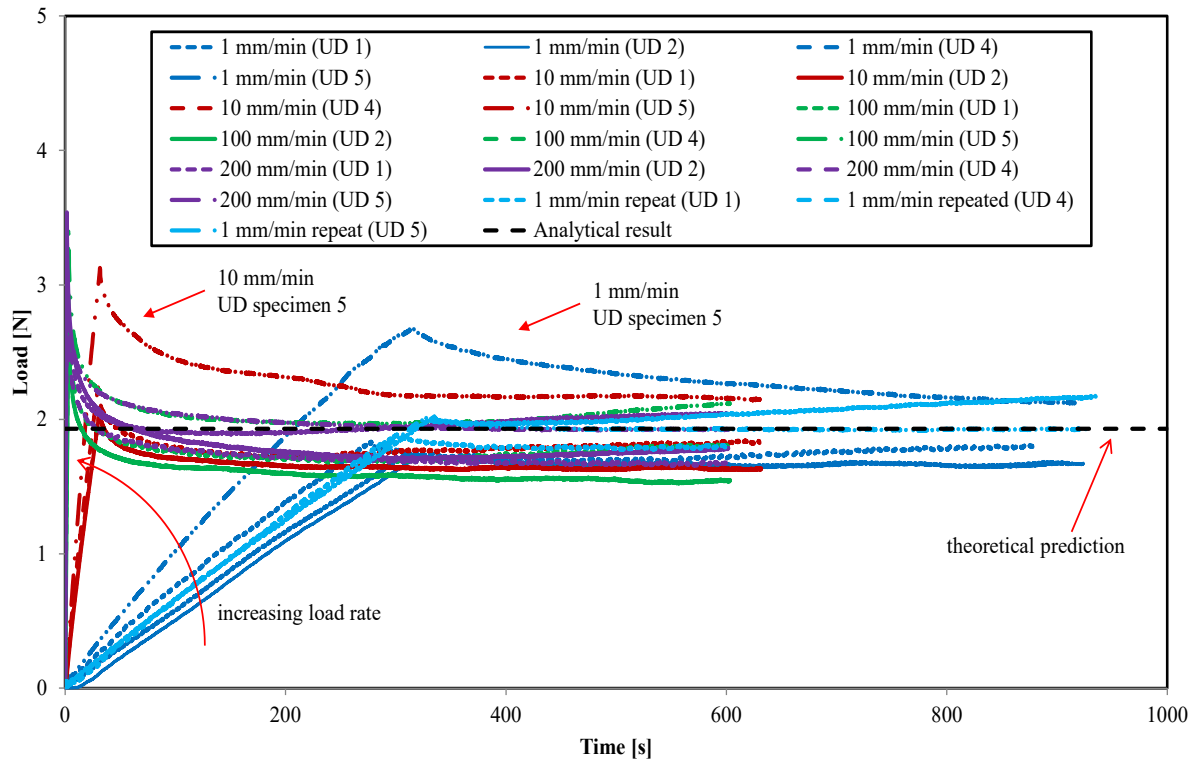
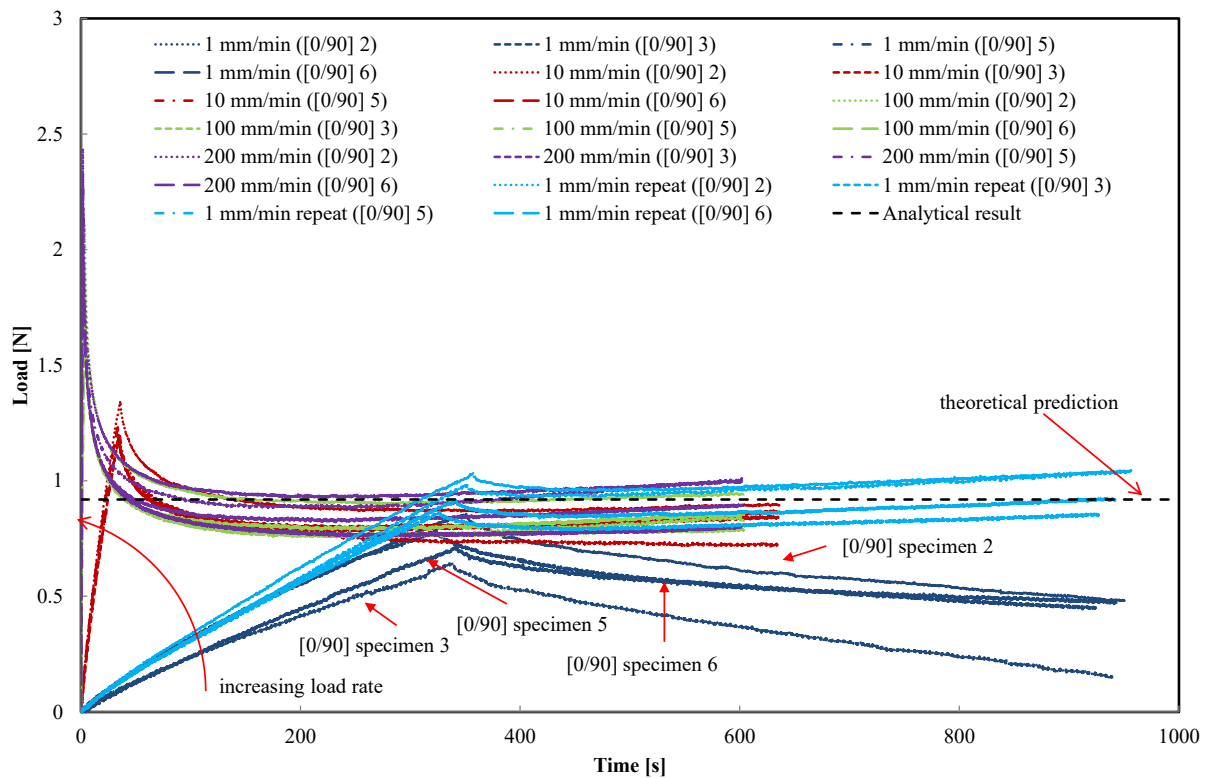


Figure 81. Effective flexural modulus of the specimens in the three layup configurations when subjected to various crosshead loading speeds.

Figure 82 plots the force versus time response of the interleaved specimens for the constant crosshead rate loading and fixed displacement phases for the range of crosshead loading rates used. In all cases, the higher crosshead speed leads to a higher force during the loading phase. The force relaxes due to the stress relaxation of the polystyrene interleaf during the fixed displacement phase and all the different loading speed cases tend to the same approximate load towards the end of this phase. Particularly, for the cross-ply interleaved flexure specimens, the 1 mm/min loading curves show that the specimen in the fixed displacement phase continued to relax with the given time (i.e. the specimen is not in equilibrium at the end of the fixed displacement phase). This may be due to the thermal stress induced by the three-point bend fixture which was not sufficiently heated to 120°C before testing. (These specimens were tested at the 1 mm/min loading speed at 120°C, but both the chamber and flexure fixture were still heated up from room temperature. Later at the higher loading speeds of these specimens, the stiffness did not continue to relax over the time. When tested at the repeated 1 mm/min loading speed, these specimens did not exhibit relaxation.) Also, the stiffness of the 1 mm/min loading curves of the specimen 3 and 5 is lower than the other cross-ply specimens. The stiffness of these two specimens in the subsequent tests of 10, 100, 200 and repeated 1 mm/min is also less than the other cross-ply specimens. This may be due to specimen faulty during specimen manufacture. For the UD specimens, the 1 and 10 mm/min loading cases of the specimen 5 show a higher stiffness than the other UD specimens. This could be due to metal support fixture (and the specimen) not being fully heated before these tests took place and so the end of the specimen would have been at a temperature less than 120°C. Later in the tests of this specimen (when the fixture would have been remained the required temperature) the stiffness matches that of the other UD specimens.



(a)



(b)

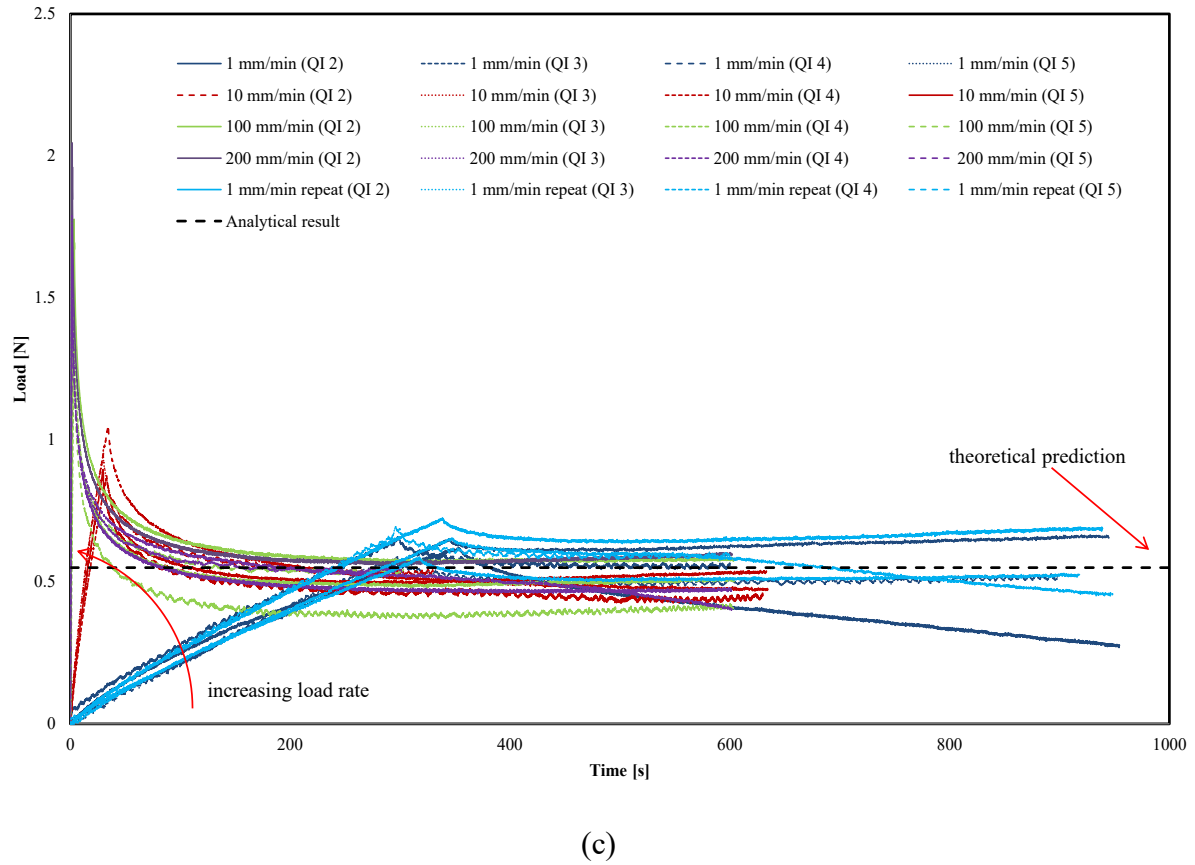
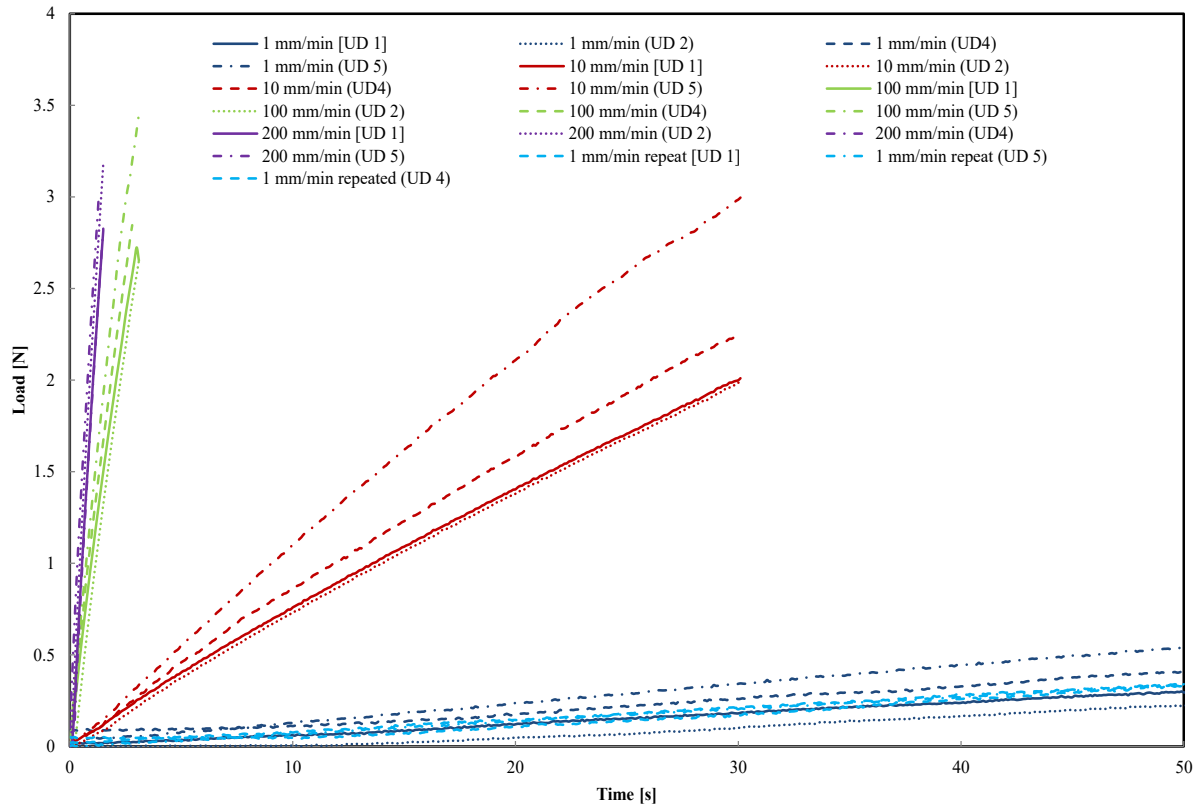
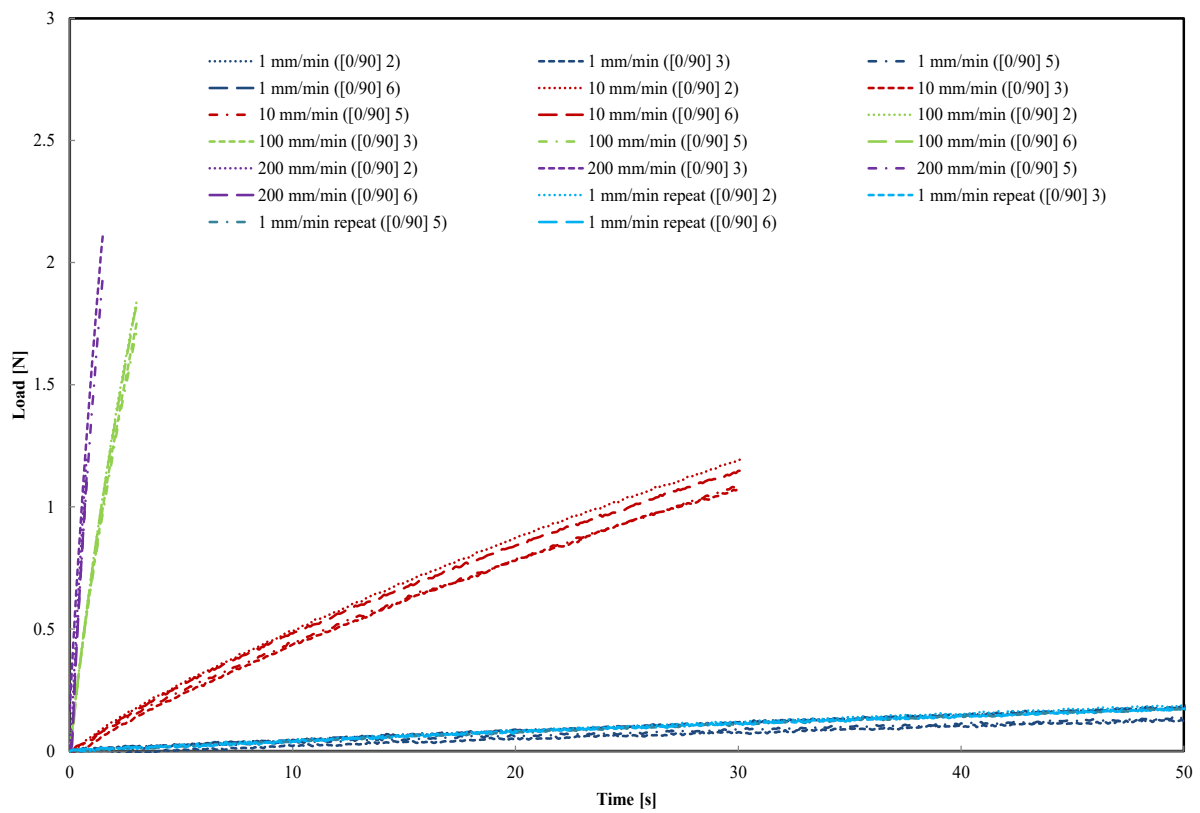


Figure 82. Load-time curves of the interleaved composite specimens in (a) UD, (b) cross-ply and (c) quasi-isotropic layup configuration loaded at different crosshead speeds at 120°C.

Figure 83 plots load-time responses with expanded time scales to check on the linearity (or otherwise) during the initial 50 seconds of the constant loading crosshead rate phase. This figure confirms that the higher loading crosshead speeds can cause greater load for all layup configurations.



(a)



(b)

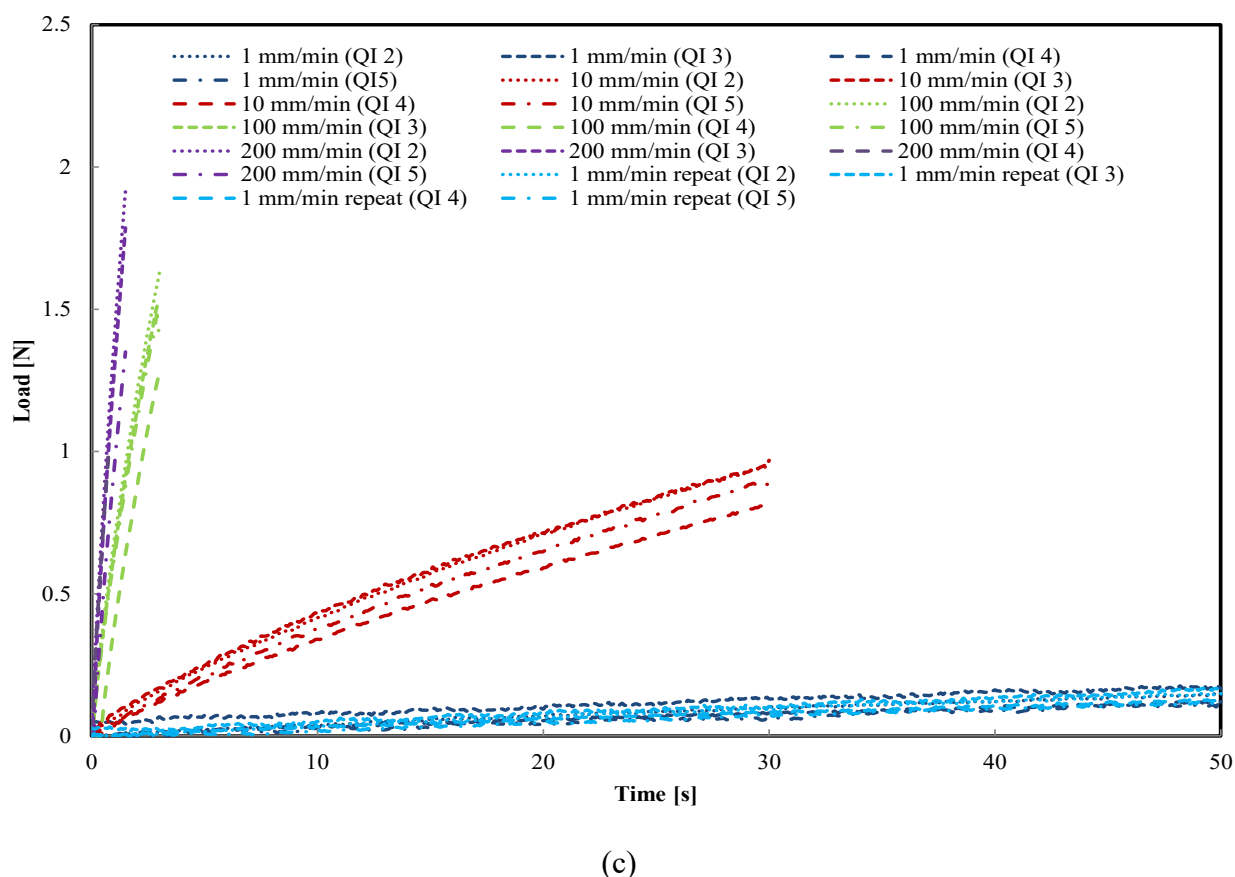
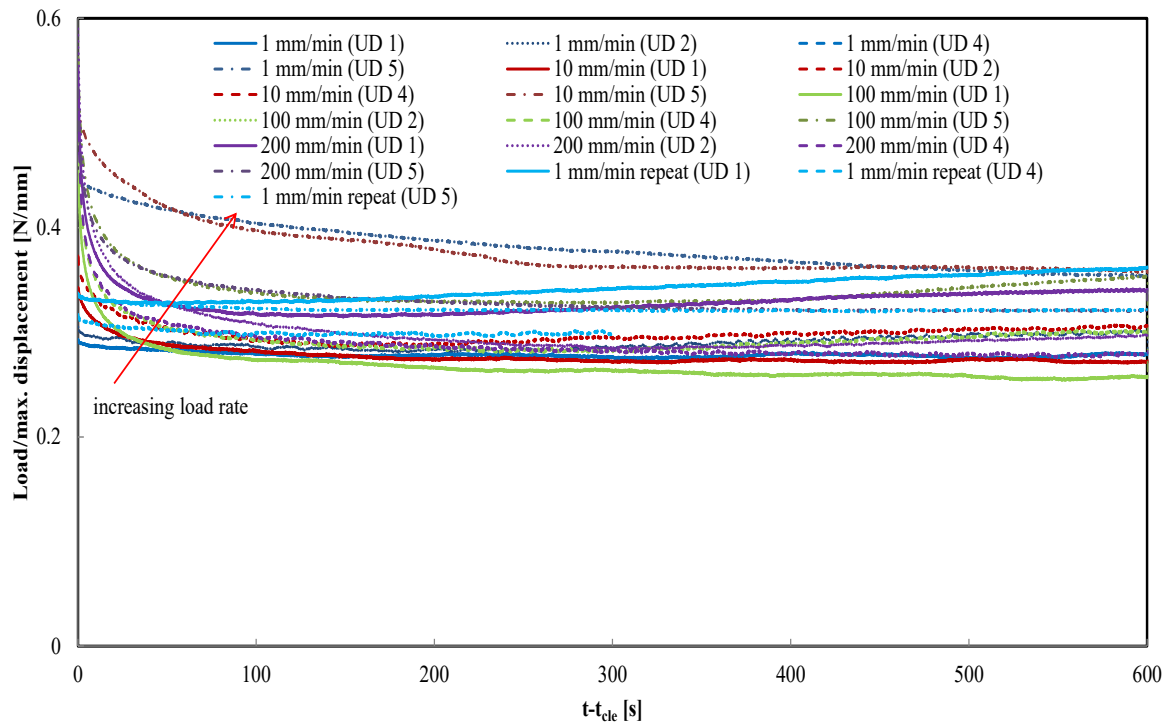
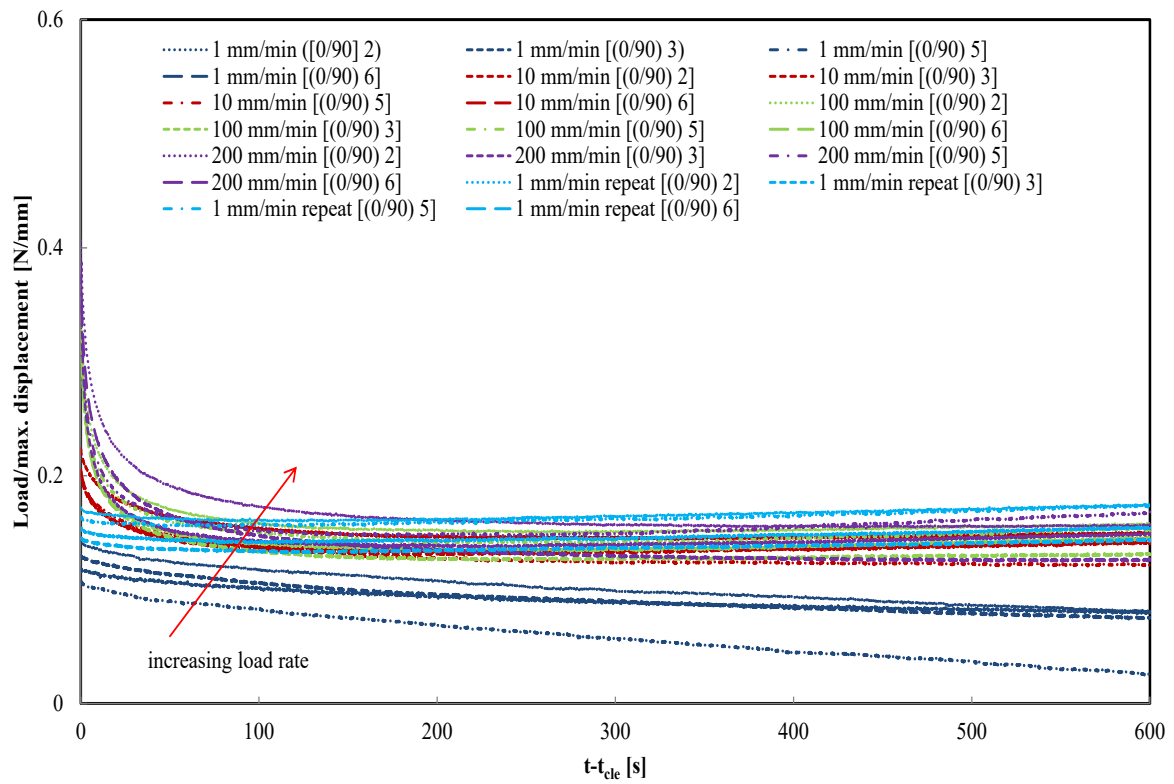


Figure 83. Plotted load-time response of an interleaved specimen in (a) UD, (b) cross-ply and (c) quasi-isotropic layout configuration for the initial loading phases.

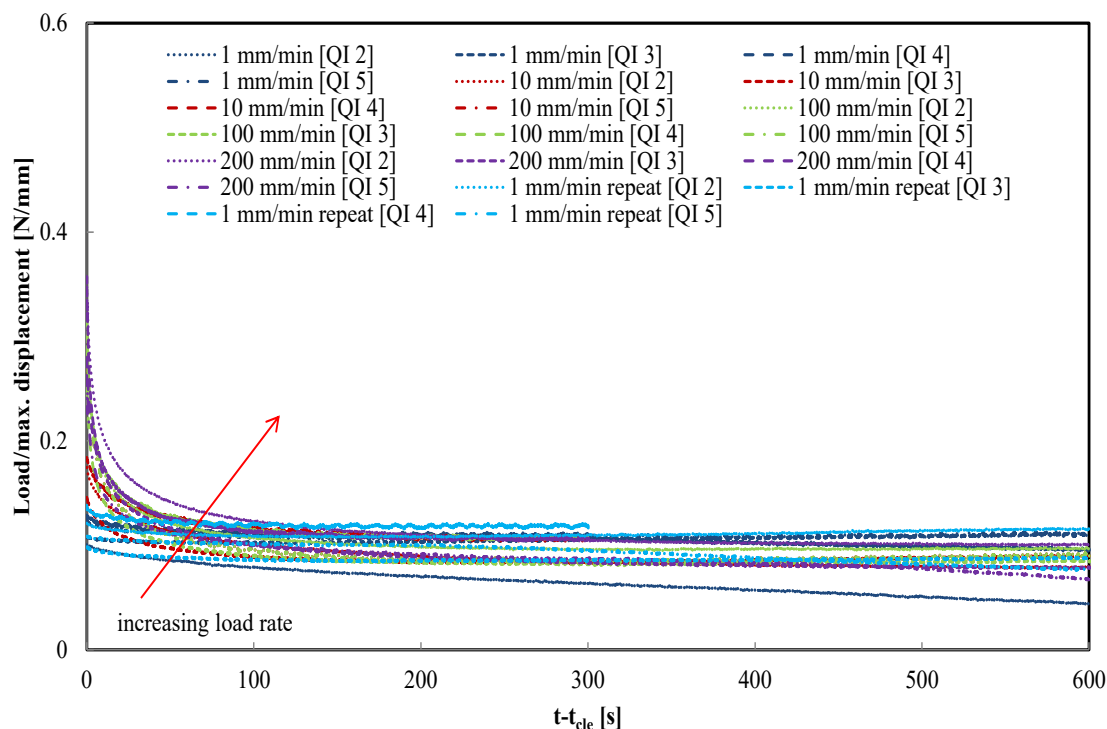
Figure 84 shows the load/maximum displacement during the fixed displacement phase plotted against the time after the end of the constant loading crosshead rate phase (i.e. plotted against $t - t_{cle}$ where t is time measured from when the loading pin first contacts the specimen and t_{cle} is the time at the end of the constant loading crosshead rate phase). It can be seen that, for a given layup, the response during the fixed displacement phase are generally similar for all loading speeds. (There are scatters on the 1 mm/min loading cross-ply specimens and the UD specimen 5 loaded at 1 mm/min and 10 mm/min. It is suspected that the three-point bending test fixture was not properly heated before testing which cooled down the specimens.) This figure confirms that specimens loaded at higher crosshead rates relaxed and tended towards the constant load level till the end of this phase.



(a)



(b)



(c)

Figure 84. Load/max. displacement-time plots of PS-interleaved specimens in (a) UD, (b) cross-ply and (c) quasi-isotropic configuration during fixed displacement phase.

4.3.3 Microscope examination

Figure 85 shows micrographs of the cross section of the interleaved composites at as-cured stage. It can be seen that after curing, PS interleaves are uniformly distributed between the carbon fibre/epoxy laminae for the UD and quasi-isotropic laminates. For the cross-ply laminate, the micrograph shows that the content of PS in terms of thickness between the interior 0° laminae and between the lower interior 0° lamina and adjacent 90° lamina is approximately 50% less than the other PS layers of the laminate and is around 50% of the PS layers of the UD laminate. This could be that PS was squeezed out the composite during curing, or the thickness of that typical PS layer was accidentally smaller than the other PS layers. The latter explanation is less likely since in later micrographs (see Figure 86

and Figure 87) the thickness of the PS layers is much more uniform throughout the cross-ply laminates.

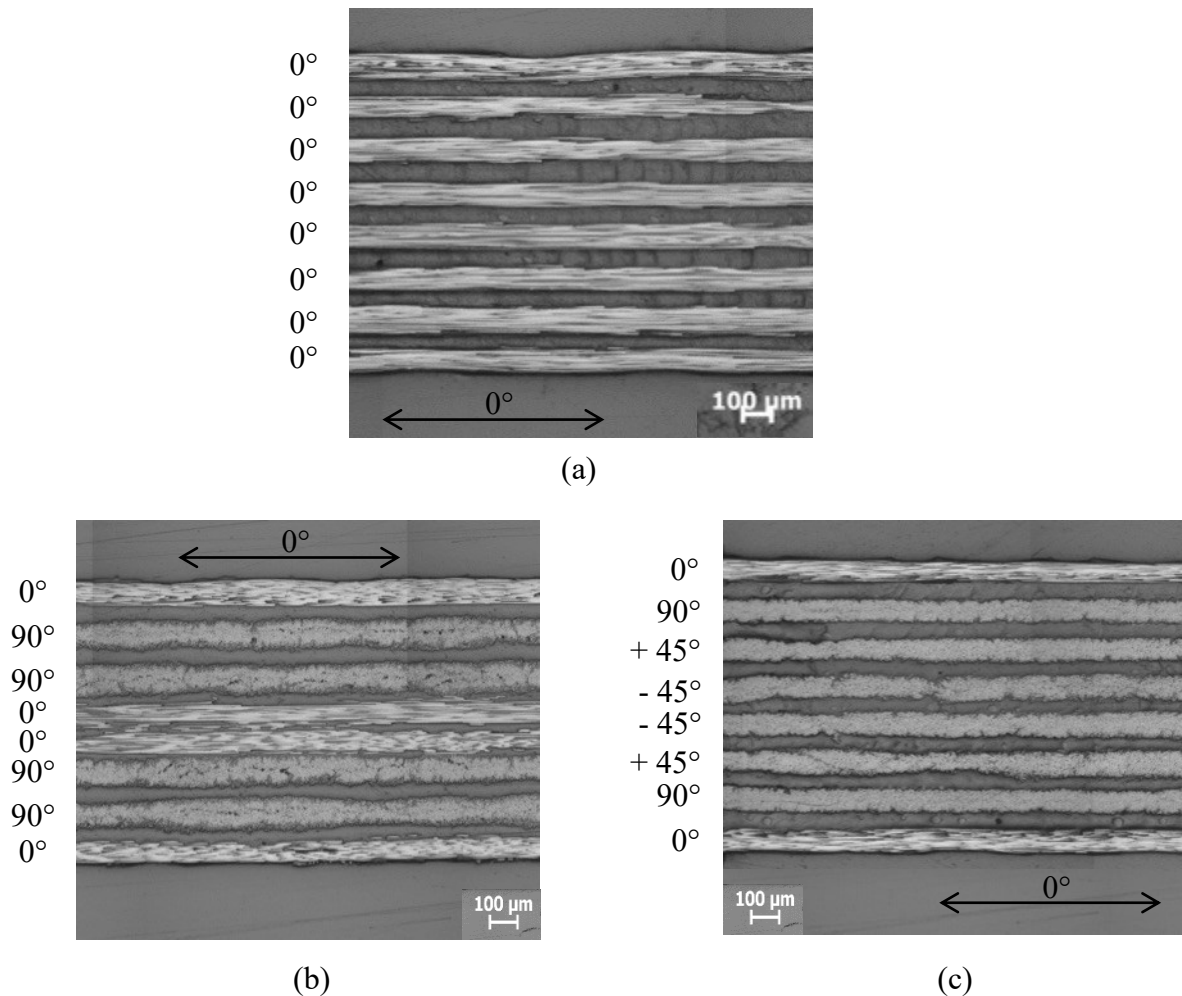


Figure 85. Microscope images of the as-cured cross section of a PS-interleaved specimen in (a) UD, (b) cross-ply and (c) quasi-isotropic layup configuration.

Figure 86 shows the cross section from microscope inspection of the interleaved specimen after loaded at the crosshead rate of 1 mm/min at 120°C. This figure shows that for all layup configurations, no interior composite ply damage can be observed. Voids can also be observed in these specimens due to the outgassing of the PS interleaves at elevated temperature.

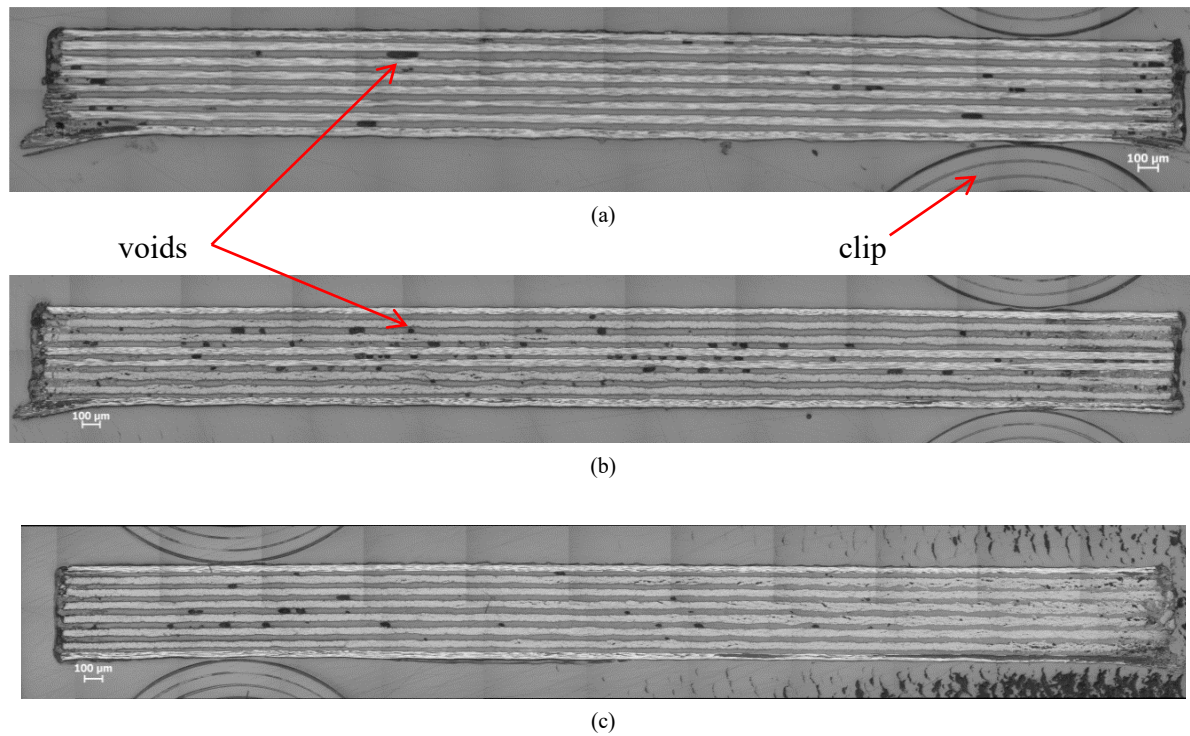


Figure 86. Microscope images of the cross section of a PS-interleaved specimen in (a) UD, (b) cross-ply and (c) quasi-isotropic layup configuration after tested at the 1 mm/min crosshead rate at 120°C.

Figure 87 shows the cross section from microscope inspection of the interleaved specimen after loaded in sequence of the loading rates at 120°C. It can be seen that the interior ply damage can be seen at the 90° plies of the cross-ply interleaved specimen and at the 45° plies of the quasi-isotropic interleaved specimen loaded at the highest cross speed. This may be due to the viscoelastic behaviour of the polystyrene interleaf. This indicates that with higher loading crosshead rate, re-shaping of the specimen can result in interior off-axis ply fracture. The micrographs also show that voids (the black area in the figures) are formed in the interleaved composites after testing. (This phenomenon has been observed in previous work on 90°-bend specimens which were re-shaped to a 180°-bend and a flat configuration (127).)

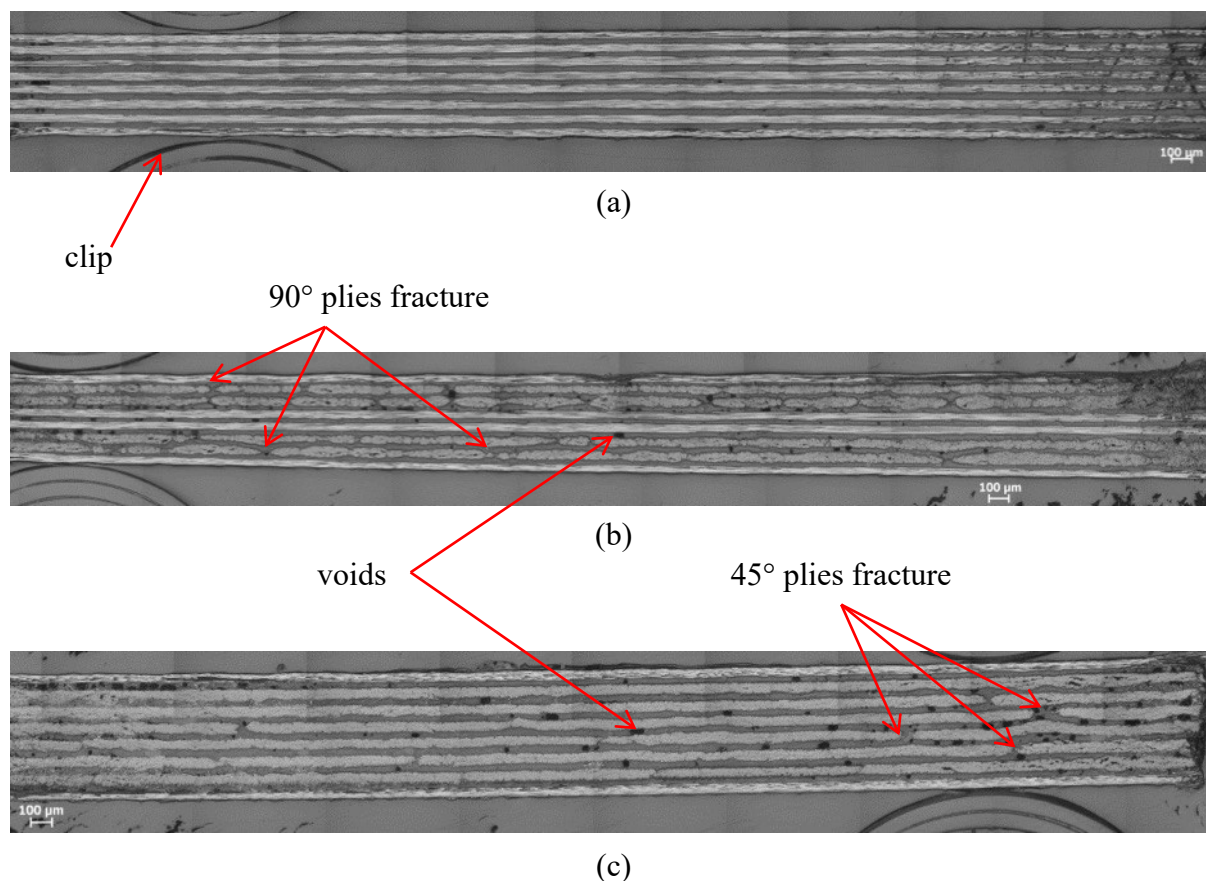


Figure 87. Microscope images of the cross section of a PS-interleaved specimen in (a) UD, (b) cross-ply and (c) quasi-isotropic layup configuration after testing at a series of crosshead loading speeds at 120°C.

4.4 Summary

The viscoelastic properties of PS interleaf were examined. DMTA results confirmed that the PS interleaf exhibits a significant reduction in storage modulus when heated from room temperature to 120°C. The frequency-sweep test at 120°C showed that the shear modulus of the PS interleaf is frequency-dependent.

The stiffness change of the interleaved composite was also investigated. It showed that the stiffness can reduce by over 99% on heating and can be restored when cooled down. Moreover, such stiffness change was repeatable on re-heating.

The viscoelastic behaviour of a PS-interleaved composite manufactured in UD, cross-ply and quasi-isotropic layup configurations was also investigated. Upon loading at different speeds, the

interleaved composite exhibited the rate-dependent load-displacement response, due to the viscoelastic behaviour of the PS interleaf. The interior ply damage observed at higher loading speed may be due to this viscoelastic behaviour. In the fixed displacement phase, the load of the higher loading crosshead rate cases relaxed and tended towards to a constant value. When removing the displacement, the interleaved composite recovered to the original, flat shape. The shape recovery rate was fastest at the beginning, but became slower due to the viscoelastic properties of the polystyrene interleaf.

Chapter 5 Finite element investigation of the shape memory capability of interleaved composites

5.1 Introduction

This chapter describes an investigation into the finite element (FE) modelling of the shape memory capability of interleaved composites. It reports the FE modelling strategy developed for a polystyrene (PS)-interleaved composite arc specimen as it is flattened and then allowed to recover its original shape at elevated temperature. In a second study, the viscoelastic properties of the polystyrene are incorporated in the FE modelling of three-point flexure tests and the predicted response is compared to the experimentally observed behaviour reported in Chapter 4.

5.2 FE modelling of flattening a PS-interleaved composite arc structure

5.2.1 Introduction

An interleaved composite, consisting of carbon fibre/epoxy laminae and polystyrene (PS) interleaf layers, has been developed which exhibits controllable stiffness and shape memory capability (5, 127). Trials have been previously performed to experimentally investigate the shape memory capability of these interleaved composites (127). A 90°-bend specimen, see Figure 88a, in which the bend is formed of a circular arc with an inner radius of 12 mm was used in the investigation. The layup of the specimen was $[0^\circ/\text{PS}/90^\circ/\text{PS}/90^\circ/\text{PS}/0^\circ]$ where PS is a polystyrene interleaf and the other plies are unidirectional carbon fibre/epoxy laminae. To investigate the re-shaping capability, this specimen was heated to 120°C and re-shaped manually to form a flat shape under a vacuum bag (Figure 88b). A vacuum was then applied to retain the specimen in the flat shape. The specimen was then cooled to room temperature, while maintaining the vacuum, and removed from the mould

(Figure 88c). To examine the shape recovery capability of the re-shaped specimen, it was re-heated to 120°C and the extent to which it returned to the original shape was observed (Figure 88d). Experimental results showed that the specimen almost fully recovered to its original bend configuration.

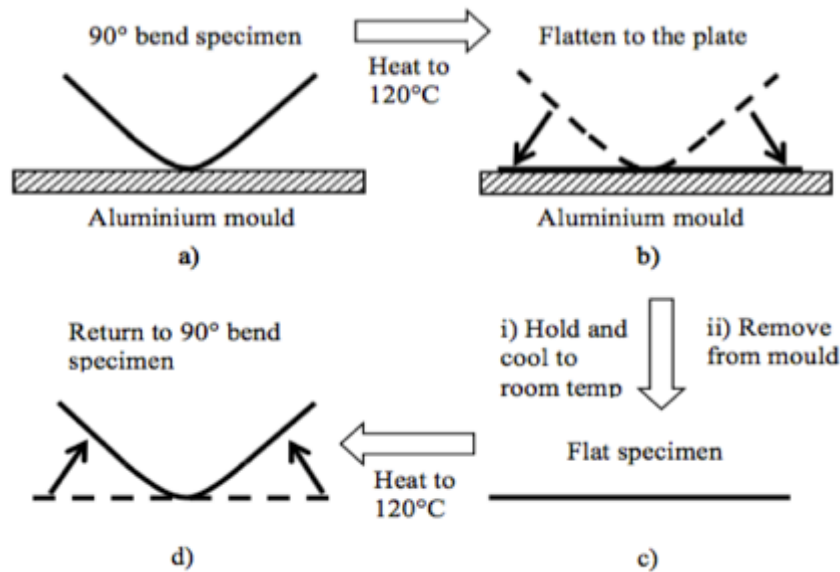


Figure 88. Experimental procedure to assess shape memory capability (a) initial state of 90°-bend specimen (b)-(c) re-shaping for production of flat configuration and (d) shape recovery (127).

Finite element modelling can be used to investigate the shape memory capability of interleaved composites. In this section, the aim was to explore how FE modelling would be used to model the re-shaping and shape recovery stages of an interleaved composite arc structure including the prediction of the stresses in the re-shaped state and the magnitude of any springback after cooling.

5.2.2 Development of the FE model

5.2.2.1 Materials

A unidirectional carbon fibre/epoxy composite [HexPly, 914C-TS-5-34% with a cured ply thickness of 0.125 mm, made by Hexcel, UK] and polystyrene (PS) interleaves [Styrolution PS 124N/L with a

thickness of 0.1 mm, supplied as films from TCKT, Austria] were used in the previous work (127) and so were employed in this modelling study. Basic mechanical properties of these materials are shown in Table 7. The elevated-temperature elastic modulus of PS was measured from DMTA temperature sweep tests in tension mode. The details of the tests can be seen in section 4.2.2.

Table 7. The mechanical properties of the materials used in this study.

Materials	Property	Value
HexPly carbon fibre/epoxy composite	E_{11} , GPa	135 ^a
	E_{22} , GPa	8.5 ^a
	ν_{12}	0.32 ^b
	G_{12} , GPa	5.27 ^b
Styrolution 124N/L PS film (25°C)	E , GPa	3.2 ^c
	ν	0.35 ^d
Styrolution 124N/L PS film (120°C)	E , GPa	0.001
	ν	0.35 ^d

^a both E_{11} and E_{22} of carbon fibre/epoxy composite were from Hexcel datasheet

^b ν_{12} and G_{12} of carbon fibre/epoxy composite were from (185)

^c E of PS film at room temperature was from Styrolution datasheet

^d ν of PS film was from the website (<http://polymerdatabase.com/polymer%20physics/Poisson%20Table.html>) [Accessed date:12.03.2018]

5.2.2.2 Outline of modelling problem

A FE model of the re-shaping and shape recovery of a simple curved interleaved laminate shown in Figure 89 was investigated. The laminate consisted of two 0° unidirectional carbon fibre/epoxy composite plies separated by a PS interleaf. The plan was to develop the modelling procedure for the re-shaping of this interleaved laminate to a flat configuration and for the shape recovery from the flat shape.

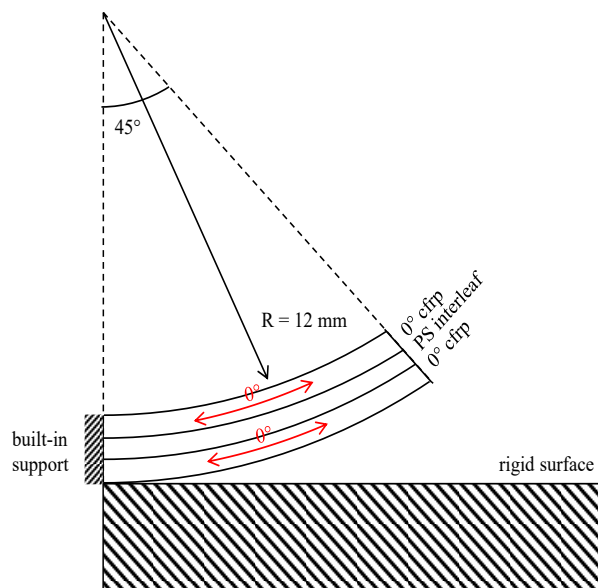


Figure 89. A sketch of the interleaved composite to be modelled.

In its initial state the radius of the upper surface (R) is 12 mm. The laminate has a subtended angle of 45° and rests on a frictionless rigid surface which is tangential to the left hand end of the lower surface which is built-in. Figure 90 shows the modelling procedure for the interleaved curved laminate. The steps to be modelled are as follows.

The 1st step is to re-shape the laminate to a straight form (step i, Figure 90). In this step, the PS interleaf is in its low stiffness state (i.e. at high temperature). The laminate is to be flattened onto the horizontal rigid surface using a uniformly distributed pressure loading on the upper surface of the laminate.

The 2nd step is to cool down the laminate while holding it in the flat configuration (step ii a, Figure 90). In this step, the stiffness of the PS interleaf is reset to its room temperature value (i.e. at its full stiffness) while the pressure loading is maintained. This step must include the stresses which are developed in the composite plies in step i. After this, the pressure loading is removed and any springback is to be determined (step ii b, Figure 90).

The 3rd step is to evaluate the shape recovery capability of the flat laminate (step iii, Figure 90). In this step, the stiffness of the PS interleaf is re-assigned the high-temperature value with no pressure

loading applied. The recovery of the specimen towards its original curved configuration is to be modelled and the extent of the recovery to be determined.

It is important to emphasise that although the modelling investigates the re-shaping of the interleaved laminate at elevated temperature (120°C for PS-interleaved composites), springback at room temperature and shape recovery, thermal stresses induced during heating/cooling are not considered in the current model.

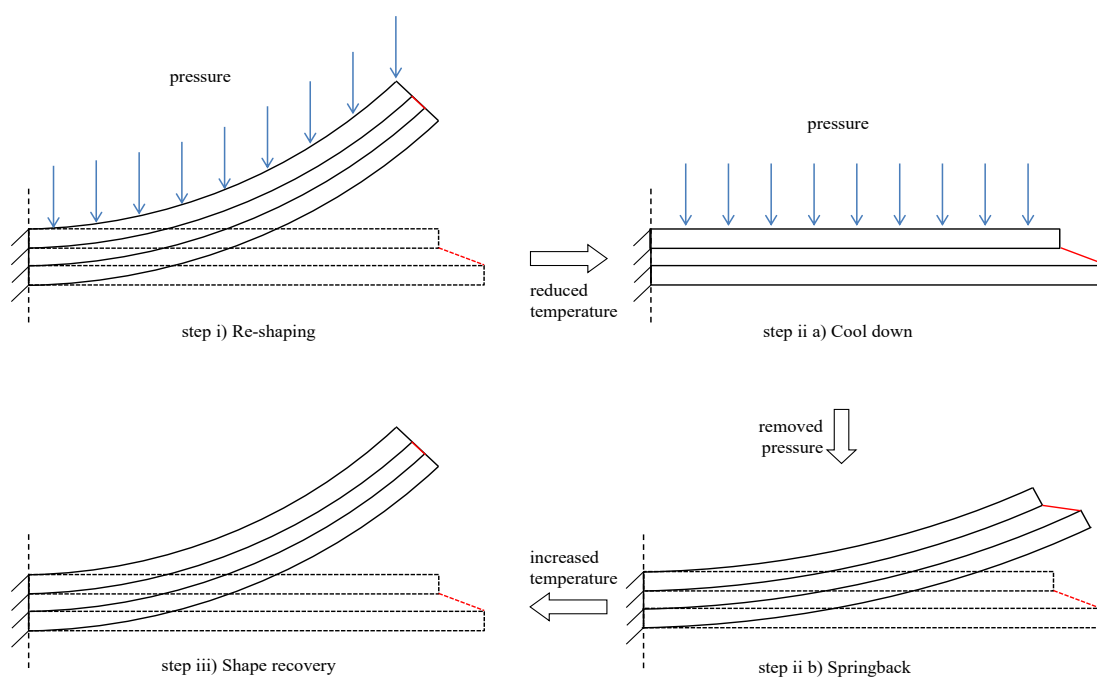
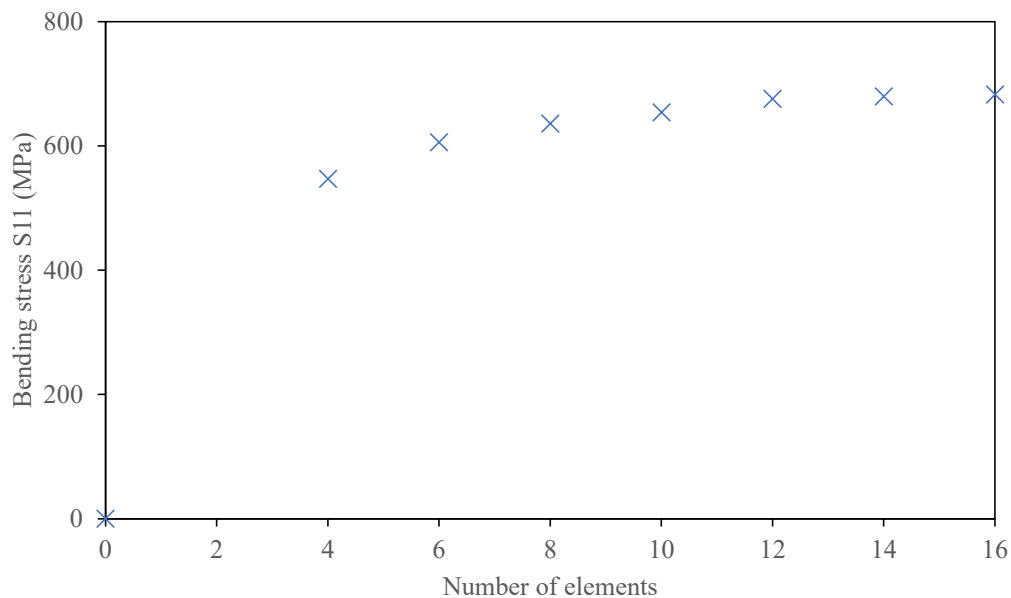


Figure 90. A sketch of the re-shaping and shape recovery cycle of a curved interleaved composite.

5.2.3 Details of the finite element model

The modelling was performed with the ABAQUS finite element software (6.14 version). Two-dimensional, 4-node, plane strain elements with reduced integration (CPE4R) were used to model the specimen. In order to determine the element size for the composite plies, a mesh convergence study was conducted. Various element sizes were used in the composite plies along the through thickness direction. Along the longer side of the composite ply, element of 0.5 mm in length was defined. Figure 91a shows the bending stress of the tenth node away from the built-in end versus the element

number along the through thickness direction of the composite ply. It can be seen that the bending stress increases upto 12 elements through the composite ply thickness, then becomes stable as element number further increases. Therefore, in this study, a fine mesh was used with the smallest elements of 0.02 mm in width (i.e. 12 elements through the composite ply thickness) and 0.5 mm in length for the composite ply (see Figure 91b). For the polystyrene ply, a mesh being square with a side length of 0.1 mm was used. This is because the PS is unlikely to significantly affect the overall mechanical behaviour of the laminate due to the low stiffness of the material, especially at elevated temperature (127).



(a)

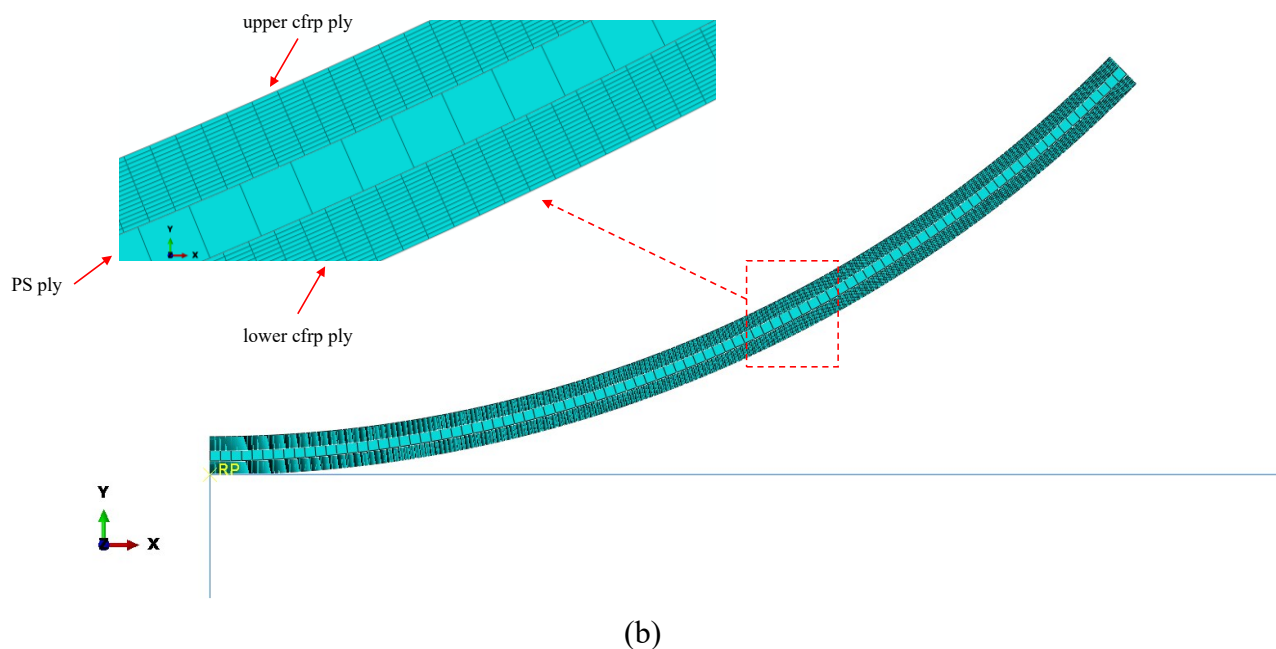


Figure 91. (a) The plot of the bending stress of the composite ply along the fibre direction versus the element number and (b) an image of the element size used in the modelling.

The composite plies were assigned with the orthotropic properties shown in Table 7 for both high and low temperature states. (The 0° fibre direction is along the longer side of these plies.) For the polystyrene ply, at room temperature the isotropic properties shown in Table 7 were used. For elevated temperature, the polystyrene was assigned with the high-temperature isotropic properties which contains a very low elastic modulus measured from DMTA (see Table 7).

Constraints were introduced to achieve the built-in end condition for the left end of the specimen. Tie constraints were used to connect the composite plies to the PS layer. An analytical rigid body was modelled to provide the horizontal surface below the specimen. A surface-to-surface, frictionless penalty contact was used between the lower surface of the laminate (as slave surface) and the horizontal rigid surface (as master surface).

A dynamic, explicit solution including non-linear geometry was used. Figure 92 plots the overall kinetic energy (ALLKE) and the overall strain energy (ALLSE) as a function of time of the modelling during the re-shaping stage (i.e. step i in Figure 90). It can be seen that the overall kinetic

energy is significantly small compared with the overall strain energy. This shows that the laminate is subjected to a quasi-static loading.

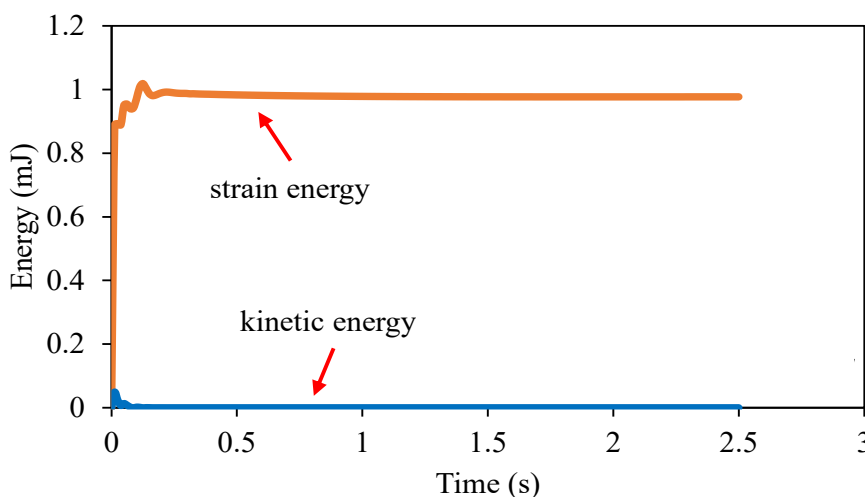


Figure 92. Overall strain and kinetic energy versus time plot of the composite arc modelling during the re-shaping stage.

The modelling of the individual processing steps was performed as follows. In the 1st step of re-shaping the laminate (step i in Figure 90), the specimen was assigned with the high-temperature stiffness properties and was loaded with a uniform downward pressure loading to drive the specimen down on to the horizontal rigid surface. The pressure loading was initially increased from zero to 0.4 MPa in 1 second and then held constant. As mentioned earlier, kinetic and strain energies of the model were monitored to confirm static equilibrium was achieved.

In the 2nd step (step ii in Figure 90), the laminate was initially cooled down while maintaining the pressure loading. To achieve this in the model, a restart analysis was performed from the equilibrium state achieved at the end of step i. The deformed geometry of the interleaved laminate (i.e. in the ‘flattened’ configuration) was imported together with the stresses in the upper and lower carbon fibre/epoxy composite plies. Stresses in the PS interleaf were set to zero and its stiffness properties were reset to the room temperature values (i.e. to the high stiffness state). The pressure loading was held constant at 0.4 MPa (the peak value applied in step i) for 2 seconds to achieve the condition at

the end of step ii a in Figure 90. Then the pressure loading was linearly reduced to zero in 2 seconds and any springback was observed in step ii b in Figure 90.

In the 3rd step (step iii in Figure 90), the laminate recovered to its original curved shape at elevated temperature. To model this, a second restart analysis was performed from the equilibrium state achieved at the end of step ii. Again, the deformed geometry of the laminate was imported from the previous analysis together with the stresses in the upper and lower composite plies. Similarly, stresses in the PS interleaf were set to zero and its stiffness properties were now returned to the high temperature values (i.e. the low stiffness state).

5.2.4 Results and discussion

Table 8 summarises the bending stress in the carbon fibre/epoxy laminae at 4 mm away from the left end of the model after re-shaping (end of step i), cooling down (end of step ii a) and pressure loading removal (end of step ii b), and shape recovery (end of step iii). The theoretical predictions are also provided in brackets in this table.

Table 8. Bending stresses of composite plies at re-shaped, cooled, springback and re-heated stages.

		Bending stress (MPa)			
		after re-shaping	after cooling	after pressure loading removal	after re-heating
top cfrp	upper surface	637.5 (703.1)	598.6	565.2 (569.1)	-53.3 (0)
	lower surface	-605.7 (-703.1)	-600.7	-605.3 (-677.4)	26.5 (0)
bottom cfrp	upper surface	615.5 (703.1)	646.1	649.5 (677.4)	71.9 (0)
	lower surface	-636.9 (-703.1)	-616.8	-561.8 (-569.1)	-29.8 (0)

It can be seen that after re-shaping, the maximum bending stresses on the upper and lower surface of the top and bottom composite plies are 637.5, -605.7, 615.5 and -636.9 MPa. These FE results are of right magnitude but are somewhat smaller than the simple beam theory prediction. Also, the maximum bending stress from the FE prediction is less than the compressive strength of the composite plies (970 MPa) at 120°C from the manufacturer's datasheet and thus no damage of the

composite plies should be seen. The shear stress in the polystyrene varies almost linearly from zero at the left hand end to 0.53 MPa at the right hand end.

After cooling down, there is a small change in the stress state of the carbon fibre/epoxy laminae. After pressure removal, the bending stresses of the composite laminae from the FE modelling are of right magnitude but are smaller than the values predicted from the beam theory. The magnitude of the maximum shear stress in the PS layer significantly increased to 8.6 MPa which occurred at the left hand end of the PS layer. This is expected as the laminate behaves as an integral structure at room temperature.

Finally, after shape recovery, the stresses in the composite plies are now small (maximum value of approximately 71.9 MPa) and the stresses in the polystyrene are again very low (maximum value of around 0.39 MPa).

The stress difference compared to the beam theory was due to the non-zero value of the elastic modulus used in the FE model for the PS in the high temperature, low stiffness state. (The beam theory predictions assume that the PS has zero stiffness in the high temperature state.)

Figure 93 shows the deflected geometry of the specimen at the end of the re-shaping process. The specimen is not completely flattened to the rigid surface; a small portion at the right hand end does not reach the horizontal surface. This is as expected. The bending moment, M^* , required to flatten the laminate is given by

$$M^* = \frac{1}{R} \sum_{for\ all\ laminae} EI \quad (12)$$

This gives the bending moment M^* as 4.39 N mm per unit width.

The maximum bending moment per unit length, M , at the built-in end due to the pressure loading of 0.4 MPa acting as a cantilever of length L mm is given by

$$M = \frac{0.4 \times L^2}{2} \quad (13)$$

Combining these two equations gives the length that will remain unflattened as

$$L = \sqrt{\frac{2 \times 4.39}{0.4}} = 4.13 \text{ mm} \quad (14)$$

This agrees with the FE-predicted length of the unflattened portion (see Figure 93).

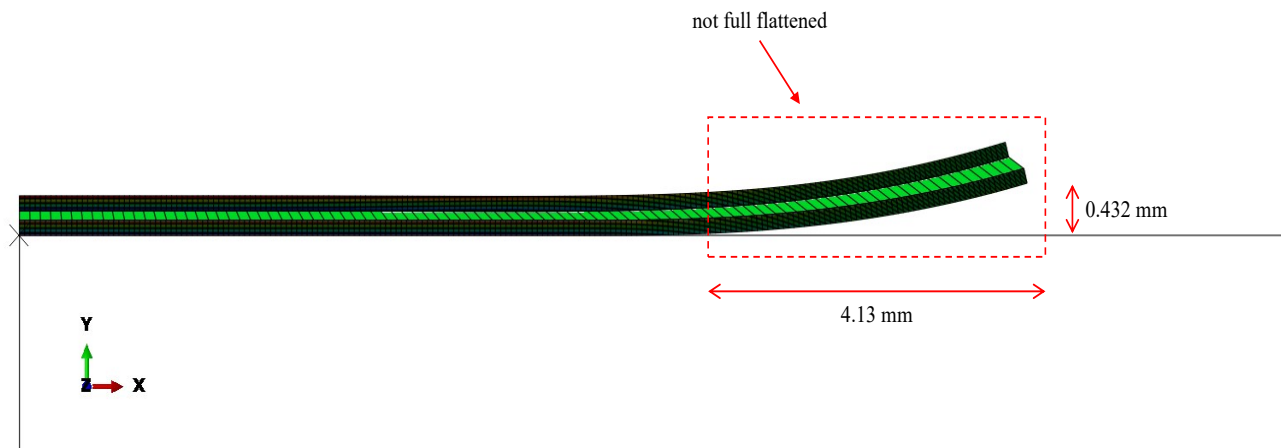


Figure 93. The re-shaped geometry of the interleaved laminate at the end of the reshaping process.

Figure 94 shows the geometry of the laminate after cooling down and removing the pressure load. The FE analysis predicted a small springback of 0.23 mm at the end of step ii.

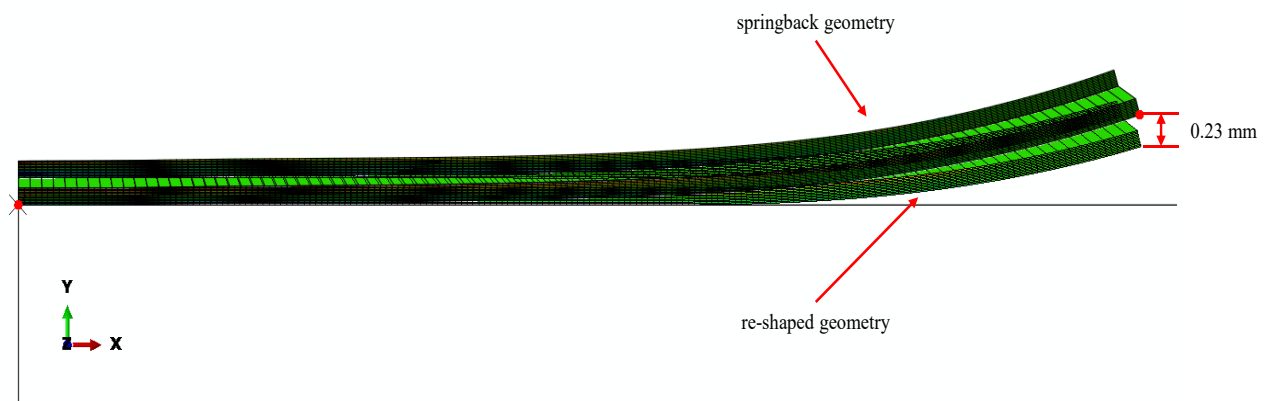


Figure 94. The springback geometry of the interleaved laminate at the end of the springback process.

Figure 95 shows the recovered geometry of the laminate which is the predicted final state achieved at the end of step iii. Compared to the initial shape of the specimen, it can be seen that the laminate almost recovered to the original shape (a difference of 0.46 mm).

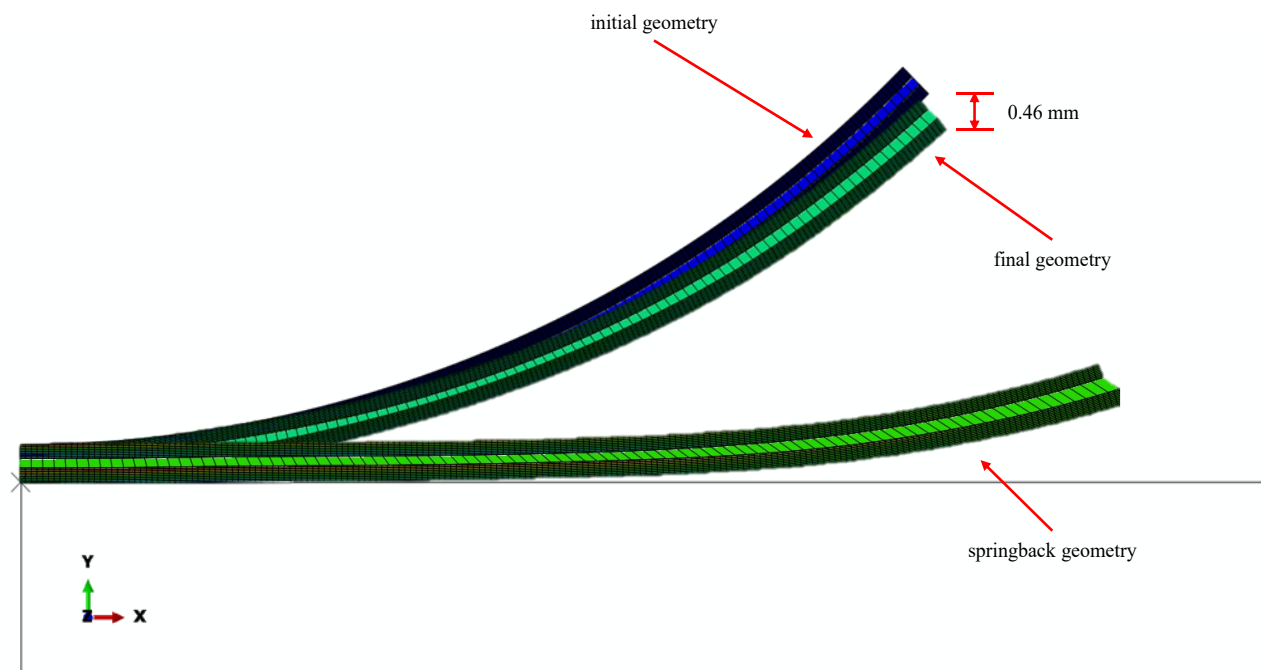


Figure 95. The final geometry of the interleaved laminate at the end of the shape recovery process.

5.2.5 Summary of composite arc modelling

FE modelling has been performed to investigate the reshaping, springback and shape recovery of a circular arc laminate made of interleaved composite. The modelling used the ‘Restart’ facility in Abaqus and analysed the development of bending stresses in the carbon fibre/epoxy laminae upon reshaping using a uniform applied pressure and evaluated the springback of the laminate at room temperature after removing the pressure. However, further validation (e.g. of the springback deflection) is needed.

5.3 FE modelling of interleaved composites under three-point flexure

5.3.1 Introduction

This section reports the FE investigation of the viscoelastic behaviour of interleaved composites subjected to three-point flexure tests at elevated temperature. As shown in Chapter 4, polystyrene (PS)-interleaved composites exhibit a significant rate dependence when loaded at higher crosshead

speeds in three-point flexure tests due to the viscoelastic behaviour of the PS interleaf. When the composites were held at a constant applied displacement, the force of the higher crosshead speed cases relaxed towards that of the lower crosshead speed cases. Upon unloading and shape recovery, the recovery rate was rapid initially, but became slower, due to the viscoelastic behaviour of the PS interleaves. The aim of this study is to investigate the viscoelastic behaviour of the interleaved composites subjected to three-point flexure tests using FE modelling. This work was conducted in conjunction with T. Kanzleiter, an MSc project student. The current author assisted in the supervision and helped in the FE modelling development. The work was reported in an MSc Project report (191).

5.3.2 Viscoelastic properties of the PS interleaf

Frequency-dependent viscoelastic properties at 120°C of the PS interleaf measured from DMTA tests, as shown in Chapter 4, were also included in the current modelling. According to the ABAQUS documentation (6.14 version), the linear viscoelastic behaviour of a material can be determined using a Prony series equation given by

$$g_R(t) = 1 - \sum_{i=1}^N \bar{g}_i^P (1 - e^{-t/\tau_i^G}) \quad (15)$$

where $g_R(t)$ is the time-dependent shear modulus in the non-dimensional form and is defined as the ratio of time-dependent shear modulus $G_R(t)$ and the instantaneous shear modulus G_0 ($g_R(t) = \frac{G_R(t)}{G_0}$), N is the number of terms in the Prony series which should be half of the number of data point pairs given, and $\bar{g}_i^P$ and τ_i^G are material constants.

The angular frequency-dependent storage shear modulus $G'(\omega)$ and loss shear modulus $G''(\omega)$ can be derived from the above Prony series equation by converting the time-dependent domain to the frequency-dependent domain via the Fourier transformation and are given by

$$G'(\omega) = G_0 \left[1 - \sum_{i=1}^N g_i \right] + G_0 \sum_{i=1}^N \frac{g_i \tau_i^2 \omega^2}{1 + \tau_i^2 \omega^2} \quad (16)$$

$$G''(\omega) = G_0 \sum_{i=1}^N \frac{g_i \tau_i \omega}{1 + \tau_i^2 \omega^2} \quad (17)$$

where ω is the angular frequency and G_0 is the instantaneous shear modulus.

By giving the experimental storage and loss shear modulus and angular frequency results in the tabular form, a nonlinear least-square fitting is performed by ABAQUS to yield the Prony series material constants N , $\bar{g}_i^P$ and τ_i^G . Table 9 shows these material constants.

Table 9. The Prony series material constants used in the modelling.

N	$\bar{g}_i^P$	τ_i^G
1	8.98×10^{-1}	8.82×10^{-4}
2	4.00×10^{-2}	1.32×10^{-2}
3	1.91×10^{-2}	6.08×10^{-2}
4	1.25×10^{-2}	2.92×10^{-1}
5	1.46×10^{-2}	1.83×10^0

5.3.3 Details of the FE model

The modelling of an interleaved composite subjected to three-point flexure tests was conducted in the ABAQUS software (version 6.14). The composite consists of eight composite plies and seven

polystyrene (PS) interleaf layers (see Figure 96). Due to the symmetry, only half of the flexure specimen was modelled. In this modelling, specimens in unidirectional (UD) layup (see Figure 69 for the stacking sequence) were investigated.

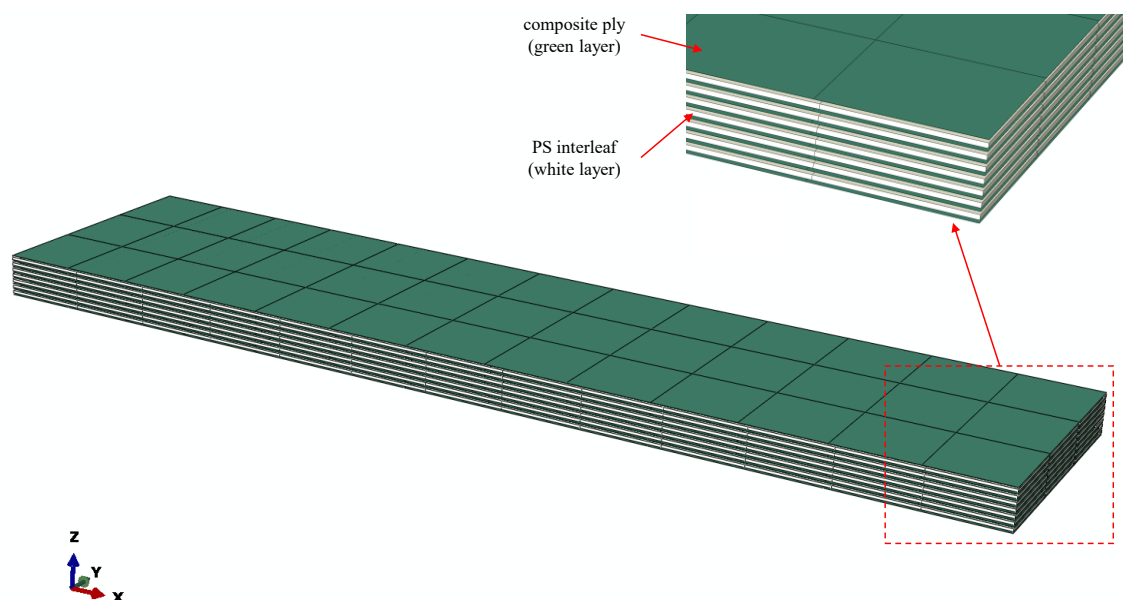


Figure 96. The three-point bending model showing the composite plies (green layers) and PS interleaves (white layers).

Three-dimensional, 4-node, standard shell elements with reduced integration (S4R) were used for the composite plies. (A preliminary study showed that shell elements can significantly reduce the computational time.) The composite plies were assigned with the composite material properties in Table 4. (The 0° fibre direction is along the longer side of the specimen.) A mesh size study was conducted to determine the element size for the composite plies. Specimens in element size of 0.8, 1.6 and 3.2 mm with 7 integration points throughout the shell thickness were developed and tested at the crosshead speed of 100 mm/min. Force at the support of the specimens of each element size was recorded. Figure 97a plots the maximum force of the specimen in the three element sizes. It shows that the maximum forces of the specimens are similar and are close to the average experimental result (dashed line). Moreover, mass scaling technique was used in this model to save computational

time. Specimens in the element size of 0.8 and 3.2 mm was conducted with mass scaling factors of 10, 100 and 1000. Figure 97b shows the mass scaling study results. It can be seen that a scaling factor of 10 can reduce computational time by approximately 95%, while remaining the error level below 3% compared to the result of the specimens without mass scaling. Therefore, the shell element of 3.2 mm in square side length with a mass scaling factor of 10 was selected for the composite plies. (However, the number of integration points of the shell element was not considered in the current modelling. Further investigation on the number of integration points is required to ensure the result accuracy.)

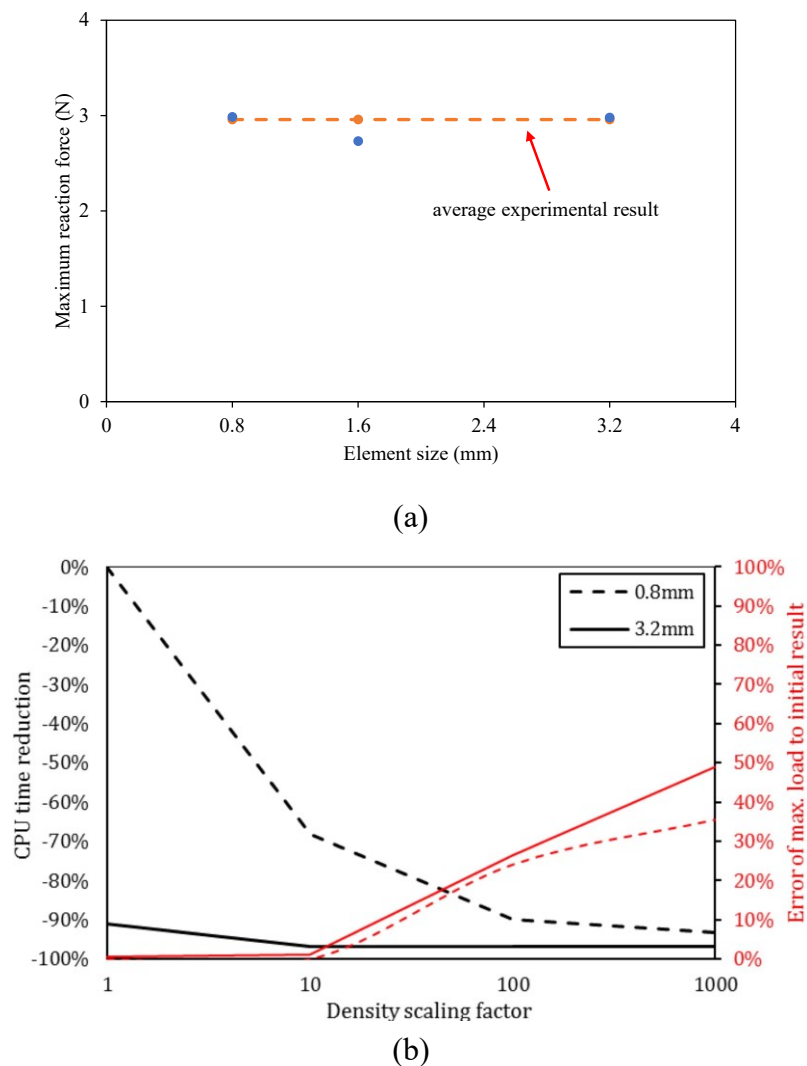


Figure 97. The mesh study results of the current modelling demonstrating the effect of (a) element size and (b) mass scaling technique (191).

Three-dimensional, eight-node continuum elements with reduced integration (C3D8R) were used for the PS interleaves. Only one element through the thickness of the PS layer was used, as it is unlikely to affect the overall behaviour of the specimen due to its low stiffness because the shear stress would be uniform across the thickness except very close to the free end of the specimen. The PS interleaves were assigned with the isotropic PS properties shown in Table 7. The viscoelastic properties derived from the Prony series approximation, as shown in section 5.3.2, were included. A rigid steel roller of 6 mm in diameter was modelled using four-node 3D bilinear and quadrilateral (R3D4) elements.

Tie constraints were introduced between the composite shell plies and the PS layer surface to prevent separation between the layers. X-axis symmetry constraint was introduced at the mid-length of the specimen. The y-axis and z-axis movements at the support position were constrained. Displacement-control boundary condition with amplitudes equivalent to the experimental crosshead displacement data was applied on the roller to achieve compression onto the interleaved composite specimen. A surface-to-surface, frictionless hard contact was used between the upper surface of the specimen (as slave surface) and the rigid roller surface (as master surface).

A dynamic, explicit solution with non-linear geometry was used. The overall strain and kinetic energies were monitored to ensure the quasi-static behaviour of the modelling. Figure 98 plots the overall strain and kinetic energies as a function of time. It shows that the overall kinematic energy is below 1% of the strain energy for all cases. This indicates that the current modelling is under a quasi-static loading.

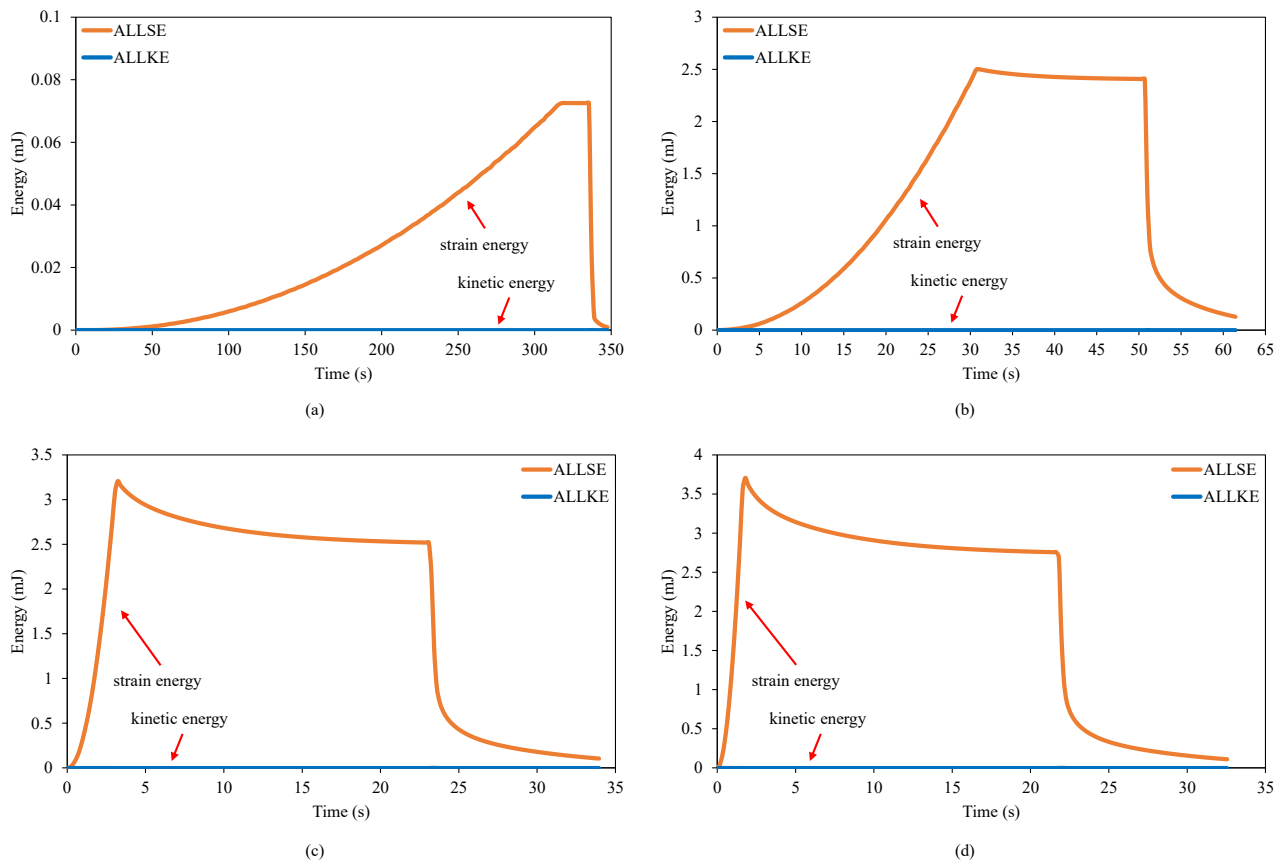


Figure 98. The strain and kinematic energies versus time plots of the modelling specimen under three-point bending loaded at (a) 1 mm/min, (b) 10 mm/min, (c) 100 mm/min and (d) 200 mm/min.

The modelling was performed as follows. An interleaved composite specimen, assigned with the high-temperature material properties, was subjected to three-point flexure tests to the deflection of 5.0 mm by pulling down the rigid steel roller at the crosshead speeds of 1 mm/min. The specimen was then kept at this displacement for 20 seconds. (This was different from the experiment in which specimens were kept at this displacement for 10 minutes, due to the very long computational running time. The aim here was to observe the onset of stress relaxation of specimens loaded at higher crosshead speeds, as shown in Chapter 4.) Finally, the roller was removed from the specimen in 0.5 seconds, and the specimen allowed to recover to its original shape. Load and displacement throughout the whole process were recorded. The process was then repeated for crosshead speeds of 10, 100 and 200 mm/min.

5.3.4 Results and discussion

Figure 99 shows the load from FE prediction and experiment plotted against time of a specimen loaded at different crosshead loading speeds for constant loading and fixed displacement phases. As mentioned earlier, the modelling specimen was kept at the intended displacement for 20 seconds. The experimental results (dashed curves) are shown for each test speed. The figure shows that in all cases, the higher crosshead speed leads to a higher load value due to the rate dependence of the PS interleaves. The gradients of the load curves from modelling are generally consistent with that of the experimental values. (The UD 5 specimen exhibited higher loads at speeds of 1 and 10 mm/min due to the insufficient heating time, as reported in Chapter 4.) The load relaxation due to the relaxation of the PS interleaves is observed and is in a good agreement with the experimentally observed behaviour. This indicates that the Prony series parameters used in the current model are correct.

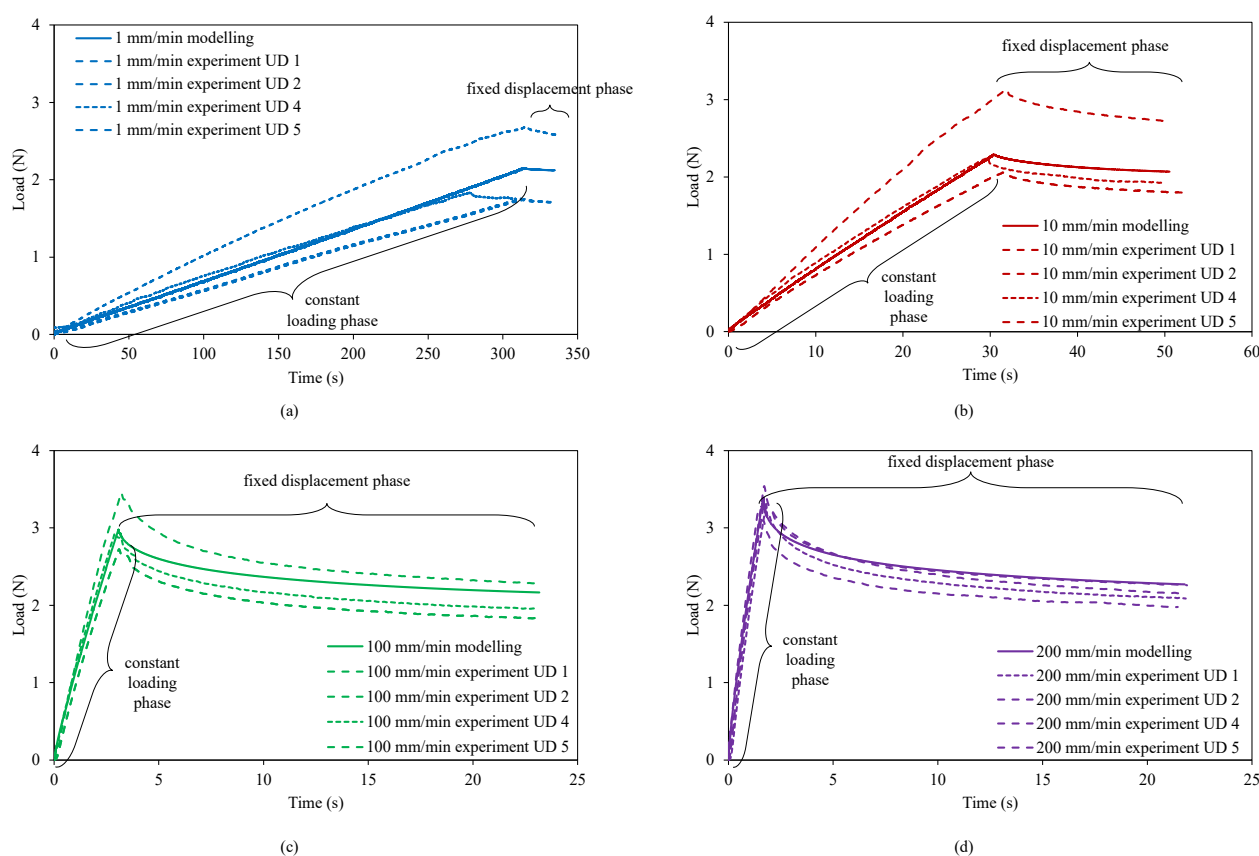


Figure 99. The load versus time plots of an interleaved specimen from modelling (solid curves) and experiment (dashed curves) subjected to three-point flexure tests at the crosshead speeds of (a) 1 mm/min, (b) 10 mm/min, (c) 100 mm/min and (d) 200 mm/min.

Figure 100 shows the load-displacement relationships of specimens tested at various crosshead speeds in the constant crosshead loading phase. The prediction (solid curves) and the experimental results (dashed curves) are shown for each test speed. In all cases, the FE model predicts a higher load value when tested at higher crosshead speeds, due to the rate dependence of the PS interleaves. This is in agreement with the experimental results. Also, the stiffness of the FE specimen increases with increasing loading speeds, which is consistent with the experimentally observed behaviour. (Again, the stiffness of the UD 5 specimen at 1 and 10 mm/min crosshead speeds is higher than the other UD specimens and the FE prediction.) This indicates that the interleaved composite is rate-dependent upon re-shaping and higher stresses can be developed within the composite plies when re-shaped at a fast speed.

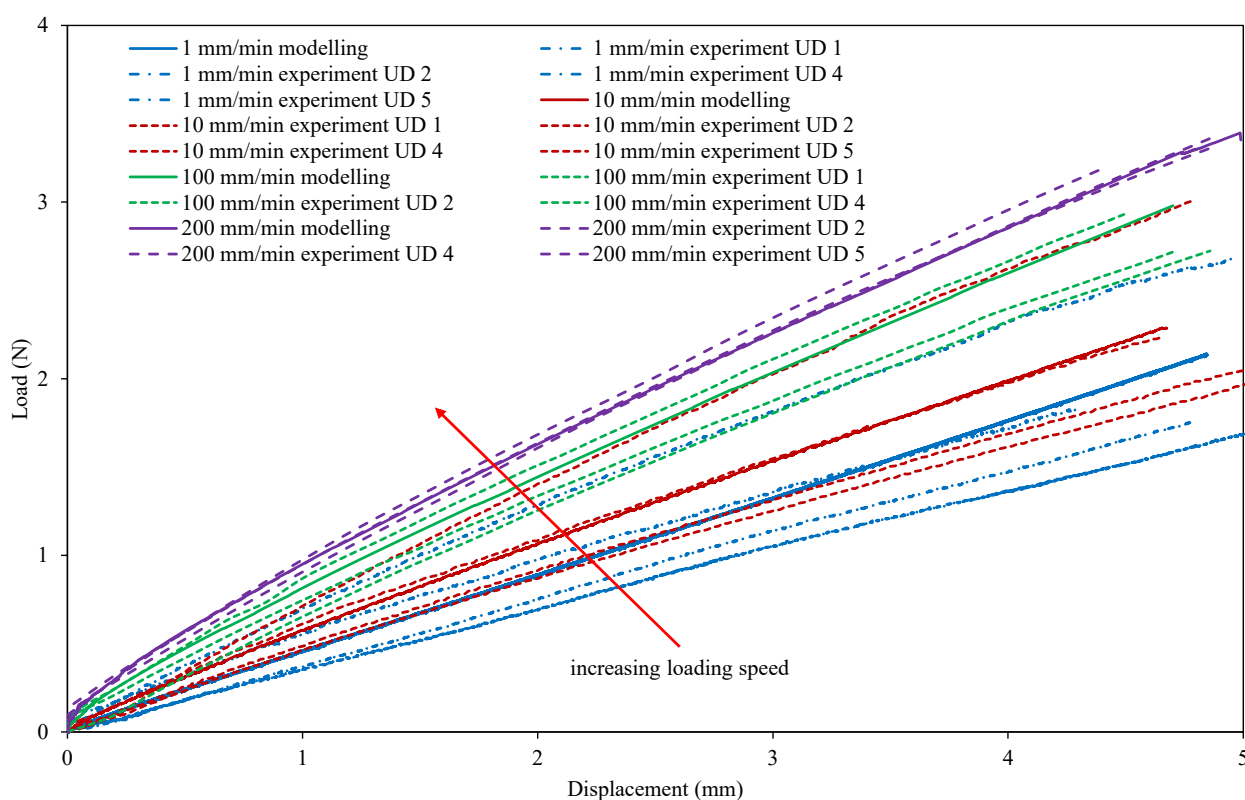


Figure 100. The load-displacement relationship from modelling of the specimen subjected to three-point bending at various crosshead speeds.

Figure 101 plots displacement against time from modelling and experiment of a specimen at the constant unloading and unconstrained relaxation phases. The prediction (solid curves) and the experimental results (dashed curves) are shown for each test speed. The prediction curves exhibit a fast recovery at the beginning of the unloading phase, followed by a slow, gradual recovery as time proceeded. This is consistent with the experimental curves as shown. (The UD 1 specimen was loaded from a different zero position than the other UD specimens for all loading cases. It detached from the roller at a different displacement position in the constant unloading phase at the 400 mm/min unloading speed.) The figure shows that the shape recovery rate is fastest at the beginning due to the release of the stored elastic stresses within the composite plies, but becomes slower as the PS interleaves gradually recovered towards the original shape which provided sufficient damping to ensure the response is overdamped. This indicates that shape recovery of the PS-interleaved composites is affected by the viscoelastic properties of the PS interleaves. Also, the FE prediction shows that the roller detached the specimens slightly earlier than the experimentally observed behaviour. This is because the unloading speed for the FE specimens (600 mm/min) was higher than the 400 mm/min in experiment due to a FE model development error. However, the FE specimens may not be in the equilibrium state before unloading as they were kept only 20 seconds in the fixed displacement phase. (In experiment, the specimens were held in the fixed displacement phase for 10 minutes to ensure they are in the equilibrium state.) Further development of longer duration in the fixed displacement phase is therefore required to ensure the FE specimens are in the same stress state (i.e. in equilibrium) before unloading.

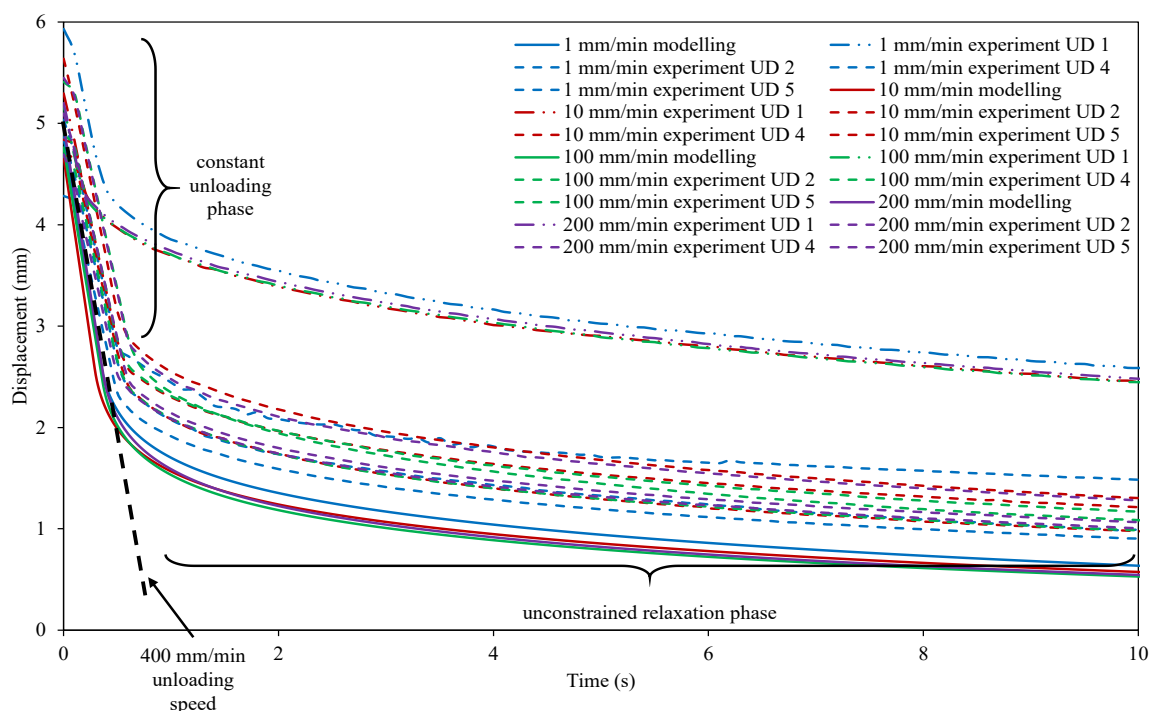


Figure 101. The displacement plotted against time of a specimen from modelling (solid curves) and experiment (dashed curves) during the constant unloading and unconstrained relaxation stages.

5.3.5 Summary of three-point flexure test modelling

FE analysis has been performed to predict the behaviour of interleaved composites subjected to three-point flexure tests at different crosshead loading speeds. The predicted rate-dependent load-displacement response during the constant crosshead loading rate phase and the load relaxation during the fixed displacement phase of the specimen were consistent with the experimental results. When removing the applied displacement, the specimen recovered towards its initial shape. The shape recovery rate from the modelling was fast initially, then became slower as the PS interleaves slowly recovered towards the initial shape. Further development (e.g. longer load relaxation time and longer shape recovery time) and investigation (e.g. the interior ply damage of the interleaved composites with the cross-ply and quasi-isotropic layups) could be performed.

5.4 Summary

Finite element modelling has been used to predict the shape memory capability of polystyrene-interleaved composites. The re-shaping, springback and shape recovery stages of the material have been successfully modelled.

FE analysis has also been used to model viscoelastic behaviour of the PS-interleaved composites subjected to three-point flexure tests. The modelling used the viscoelastic properties of the PS interleaf which were measured in DMTA tests. The modelling predicted the loading rate dependence, load relaxation and shape recovery rate of the PS-interleaved composites and these were found to be consistent with the experimental results reported in Chapter 4.

Chapter 6 Interleaved carbon fibre/epoxy composites for easy repair of interlaminar damage

6.1 Introduction

This chapter describes the investigation of interleaved carbon fibre/epoxy composites to enable easy repair of interlaminar damage by using a ‘weak link’ strategy. In mechanical systems, a weak link is a replaceable sacrificial element that is designed to fail preferentially in the event of an overload condition and so protect other parts that are more difficult and/or expensive to repair or replace. For a composite laminate, it is not possible to introduce an element which can be readily replaced in the event of damage. Instead, it is proposed that a readily repairable weak link in the form of a thermally bondable interleaf can be incorporated within the composite laminate. The interleaf material will need to possess the following characteristics.

- i) Have a strength that is lower than the interlaminar shear strength (ILSS) of the parent composite laminate (so that the interleaf becomes the weak link) but not so low that the interleaved composite is too susceptible to interlaminar damage.
- ii) Be capable of thermal bonding to itself at a temperature which will not degrade the parent composite laminate.
- iii) Be able to withstand the cure conditions of the parent composite and remain as a continuous, discrete layer in the cured composite.
- iv) Possess a good bond to the parent composite so that failure will occur within the interleaf rather than at the composite-interleaf interface (unless the interleaf can thermally re-bond to the parent composite in the event of failure at the interface).

To illustrate this concept, a schematic diagram of the damage-repair cycle of the weak link interleaved composite subjected to impact is shown in Figure 102. On impact the interleaved

composite would preferentially fail within the interleaf layers due to impact assuming the impact is not so great as to cause in-plane failure of the plies due to bending/membrane stresses. To repair the laminate, the laminate would then be heated and subjected to external pressure to achieve bonding within the interleaf.

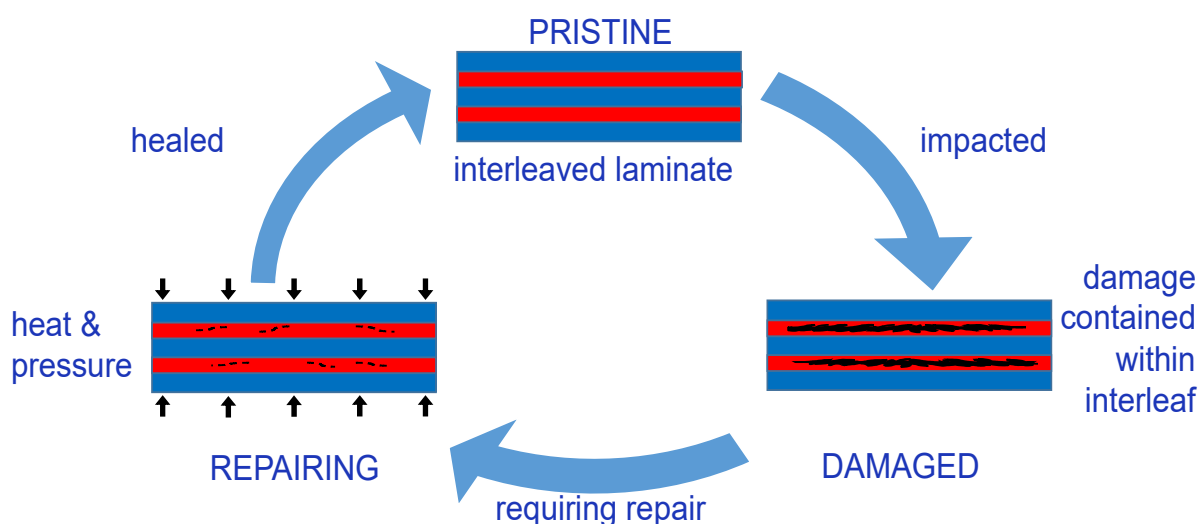


Figure 102. A sketch of the damage-repair process of an interleaved composite subjected to impact.

For this study, two interleaf polymers were used. Polylactide (PLA) film was selected as the interleaf material for the initial investigation. Preliminary characterisation of the PLA had shown that it possessed a shear strength of 24 MPa. This was measured in short beam shear tests manufactured according to the ASTM D2344M on composite specimens containing a single mid-plane PLA interleaf. This shear strength is significantly smaller than the ILSS value of the parent composite and so would certainly act as a weak link. PLA was therefore suitable for demonstrating the concept but in practice an interleaf material with a shear strength rather closer to the composite ILSS would be required.

To investigate the effectiveness of the concept with an interleaf with a higher shear strength, a polycarbonate (PC) film was selected for a second set of experimental tests. (No shear strength data was available in the polycarbonate material datasheet, but it has been reported that the room-temperature shear strength of polycarbonates is around 50 MPa measured from Iosipescu test (192).)

6.2 Details of experimental investigation

6.2.1 Materials

A carbon fibre/epoxy system (HexPly, 914C-TS-5-34%) was selected as the parent composite for this investigation (although the concept is equally applicable to other thermoset and thermoplastic matrix composites). This was produced by Hexcel (Cambridge, UK). The composite has a cure temperature of 175°C, a cured ply thickness of 0.125 mm and an interlaminar shear strength (ILSS) of 78 MPa according to the Hexcel datasheet. Polylactide film (Ingeo™ 2003D PLA film with a thickness of 0.1 mm and a melting temperature of 190°C) was supplied by TCKT from Austria and polycarbonate film (Lexan 8B35 PC film with a thickness of 0.175 mm, and T_g of 153°C) was purchased from SABIC, UK.

6.2.2 Manufacture of interleaved panels

Three types of interleaved panels with carbon fibre/epoxy layers were manufactured. The interleaved panels A consisted of 16 plies of unidirectional carbon fibre/epoxy laminates and one polymer interleaf film at the mid-plane. Interleaved panels B consisted of 24 plies of unidirectional carbon fibre/epoxy laminates and one polymer interleaf film at the mid-plane. Detailed layup configurations are shown in Figure 103a and Figure 103b respectively. These A and B panels were 200 mm long in the 0° fibre direction and 120 mm wide. Flexure test specimens were prepared from panels A and B. Interleaved panels C consisted of 8 plies of carbon fibre/epoxy laminates and 7 plies of thermoplastic interleaves (TI) in an alternating quasi-isotropic sequence $[0^\circ/\text{TI}/90^\circ/\text{TI}/+45^\circ/\text{TI}/-45^\circ/\text{TI}/-45^\circ/\text{TI}/+45^\circ/\text{TI}/90^\circ/\text{TI}/0^\circ]$. These panels were 160 mm in the 0° fibre direction and 140 mm wide. The panels C were manufactured for static indentation tests. The layup configuration of this interleaved panel is shown in Figure 103c.

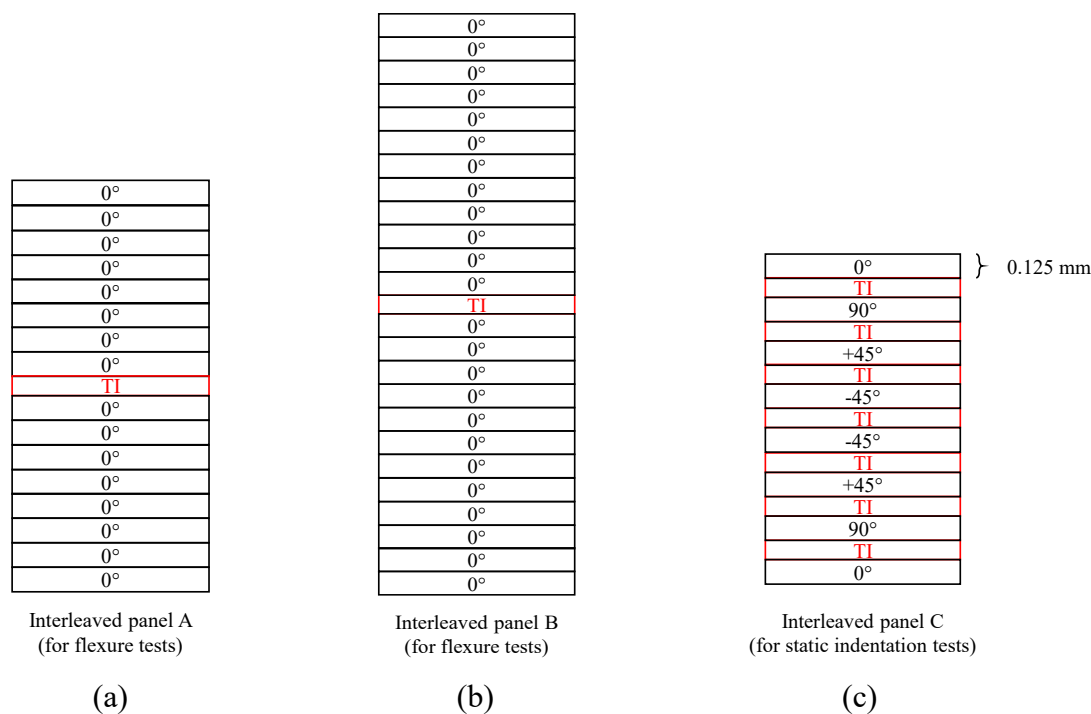


Figure 103. The layup configurations of the three interleaved panels in this investigation. (White layers are carbon fibre/epoxy laminae and red layers are interleaves.)

For the PLA-interleaved panels for flexure testing (interleaved panels A and B), three batches of panels were manufactured. Batch I (using interleaved panel A) contained PLA films without any surface treatment and was cured in an autoclave [Autoclave, LBBC, UK] at 175°C and a 100 psi pressure for 1 hour, as recommended by Hexcel. (This interleaved composite was designated as ‘standard PLA’.) Batch II (using interleaved panel B) contained PLA films which were treated on both surfaces in oxygen plasma at room temperature and low pressure for 5 minutes (This interleaved composite was designated as ‘plasma-treated PLA’). Batch II was cured in a heated press [Carver 4122GE, Carver, INC., USA] at 175°C and a sufficient pressure for 1 hour. (Unfortunately, the pressure was not recorded. However, it was sufficient to press the laminae together during hot press. The average thickness of the plasma-treated PLA after hot press curing was 3.18 mm.) Batch III (again using interleaved panel B) also contained plasma-treated PLA films. Batch III was cured in the heated press at 175°C and a sufficient pressure for 1 hour, followed by a post-curing in an oven [Forced convection oven, OF-21E, Lab Companion, Jeio Tech, Korea] at 190°C under vacuum

pressure for 5 minutes (This interleaved composite was designated as ‘post-cured PLA’). Such post curing was to remove the trapped air/voids at the PLA/composite interfaces. Test results in the following section have been shown that the post-curing process can significantly improve the shear strength of the plasma-treated PLA interleaf.)

For the PLA-interleaved panels to be subjected to static indentation tests (interleaved panel C), the plasma treatment, curing and post curing process were the same as those used for the Batch III of the interleaved panel A.

For the PC-interleaved panels (interleaved panels B and C), no plasma treatment was applied onto PC films. The panels were cured in the autoclave at 175°C and a 100 psi pressure for 1 hour. No post-curing process was applied. This work was conducted in conjunction with Y.Bai, an MSc project student. The author assisted in the supervision and helped in the manufacture and experiments. The work was reported in an MSc Project report (193).

After curing (or post-curing for some PLA-interleaved panels as mentioned), these panels were c-scanned ultrasonically using a 10 MHz transducer [Triton 1700TT Immersion Tank, TecniTest]. Pulse-echo method was applied for the scanning. After c-scan, these panels were dried at room temperature in air.

To examine the cross sections of these laminates, small sections of each panel were mounted in epoxy resin [Epoxicure, Buehler, UK] with a hardener [Kleer set hardener, 11 10 89, MetPrep, Buehler, UK] and cured for 8 hours at room temperature. The mounted sections were then polished in a grinder-polisher machine [MetaServ/MetaPol, Buehler, UK] with a SiC grinding paper in the grit level of P600 and 3 dispersions of diamond (9 µm), alumina (0.6 µm) and silco (0.03 µm) and then examined using an optical microscope [Zeiss, Axio Imager M2m, Germany].

6.2.3 Preparation of flexural and static indentation test specimens

Flexure specimens, 80 mm long and 10 mm wide (see Figure 104), were prepared from the interleaved panels A & B containing PLA and PC interleaves using a circular dry-saw cutting machine [Circular dry saw, Startrite, UK]. For these specimens, the 0° fibre direction was parallel to the longer edge of the specimens.

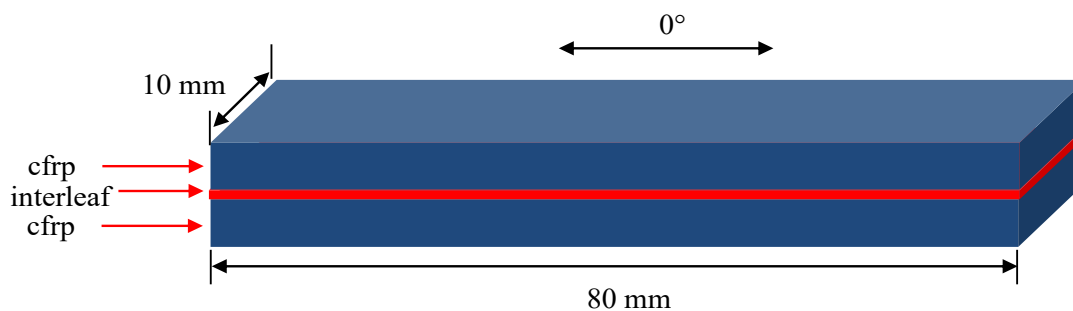


Figure 104. A sketch of the geometry of an interleaved composite test specimen in flexure.

The interleaved panel C (140 mm long and 120 mm wide), containing PLA and PC interleaves, were used for static indentation tests (see Figure 105). The 0° fibre direction was parallel to the longer edge of the panels.

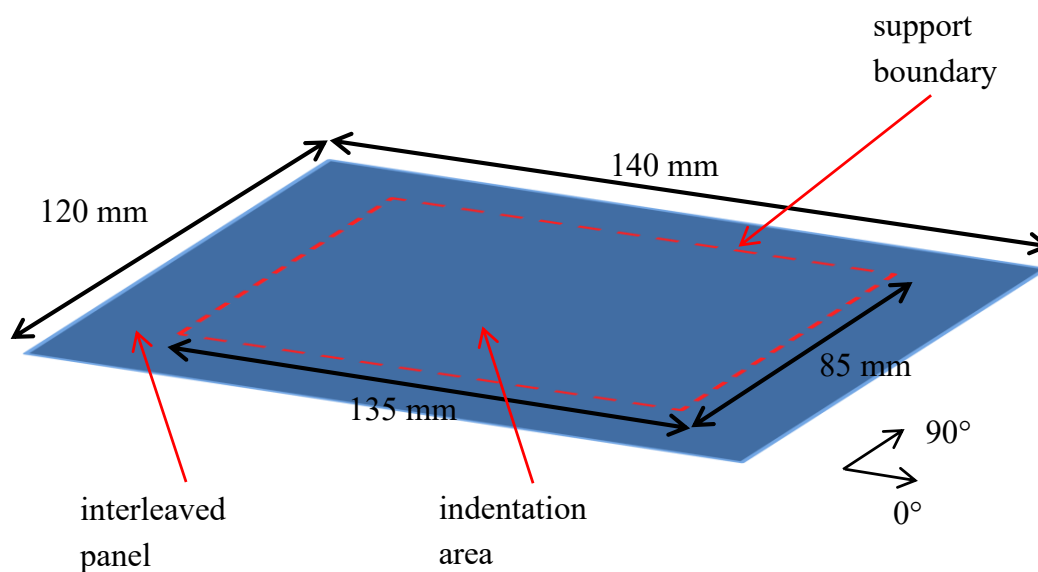


Figure 105. A sketch of the geometry of an interleaved composite indentation panel.

6.2.4 Evaluation of thermal properties of thermoplastic interleaf materials

The thermal properties, including glass transition (T_g) and melting (T_m) temperatures, of the PLA and PC interleaves were measured using a differential scanning calorimeters (DSC) [Discovery, DSC1-0438, TA instruments, UK]. PLA specimens of around 6 mg by weight were heated at a constant heating rate of 10 °C/min over a temperature range of 0°C to 200°C, then cooled down to 0°C and finally re-heated to 200°C at the same heating rate. The whole process was operated in an inert nitrogen environment. PC specimens of around 2 mg by weight were heated at a constant heating rate of 10 °C/min over a temperature range of 20°C to 250°C, then cooled down to 20°C and finally re-heated to 250°C at the same heating rate. The T_g and T_m of the PLA and PC interleaves were measured from the re-heating process. Five PLA specimens and two PC specimens were tested.

6.2.5 Mechanical testing

6.2.5.1 Flexure testing

The flexure test was performed in a three-point bending fixture according to ASTM D7264M standard. Each specimen was loaded in a universal test machine [Instron 5969, Bucks, UK] with a 50 kN load capacity. For PLA-interleaved specimens, a support span of 64 mm was used and for PC-interleaved specimens, a support of 50 mm was used. (Smaller span would have been investigated if required to generate a shear failure mode but in the event this was not necessary for the interleaf materials under consideration.) The diameter of the support and loading rollers was 6.4 mm. The experimental setup is shown in Figure 106. Flexure tests were conducted at a crosshead speed of 1 mm/min at room temperature until each specimen completely failed in shear (In the case of PLA interleaved specimens, they became separated at the mid-plane after testing. In the case of PC specimens, the specimens only separated over one half of the length to the mid-span position.). After testing, the fracture surfaces of the two halves were examined using an optical microscope. At least five specimens of each panel were tested.

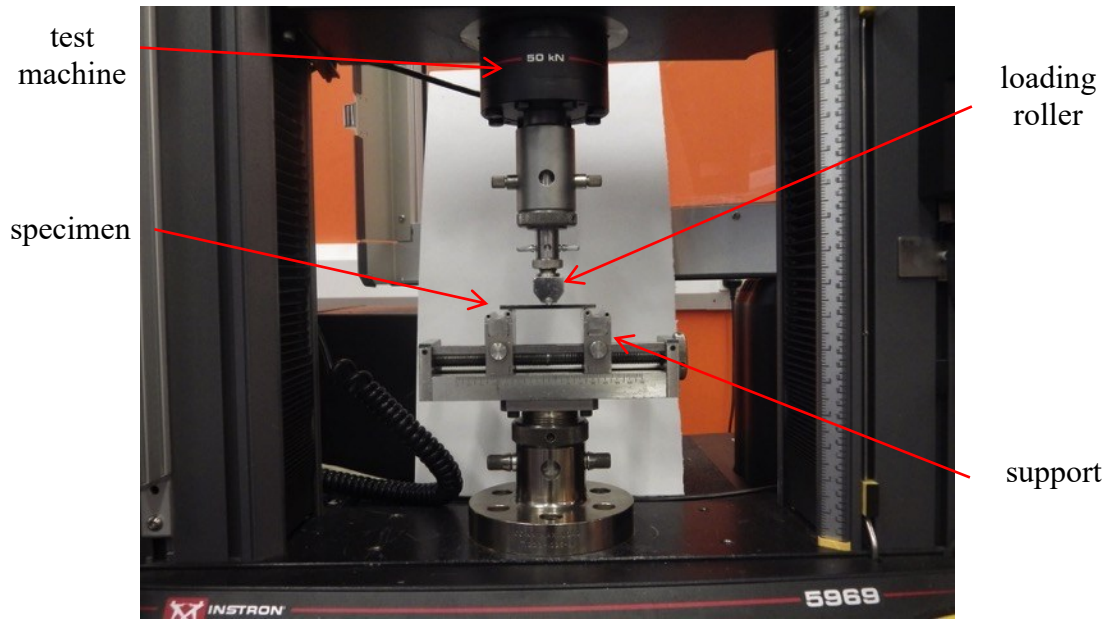


Figure 106. The test setup of the three-point bending test.

For each specimen (of thickness t and width w), the load causing shear failure (P_c) was measured and the associated the shear strength (τ^u) was calculated from

$$\tau^u = \frac{3 \times P_c}{4 \times t \times w} \quad (18)$$

6.2.5.2 Static indentation test

Each interleaved panel was subjected to an indentation test at a crosshead speed of 1.25 mm/min according to the ASTM D6264M standard. The specimen was supported on a 135 mm $\times$ 85 mm rectangle formed by semi-circular section, 10 mm diameter steel bars mounted on a steel plate (250 mm $\times$ 250 mm square, 25 mm thick) containing an aperture of 115 mm $\times$ 65 mm. A similar steel plate, again with a rectangular arrangement of semi-circular section bars, was placed on the top surface so that the specimen panel was effectively supported between the semi-circular section bars in contact with the upper and lower surfaces of the panel. Four bolts, tightened to a torque of 30 Nm, were used to connect the upper steel plate to the lower one. A hemispherical indenter of 6 mm in diameter was used to apply the load at the centre of the interleaved panel. The experimental setup is

shown in Figure 107. The test was performed on an Instron universal test machine [Instron 5969, Bucks, UK] with a 50 kN load capacity, at a crosshead speed of 1.25 mm/min. The displacement was applied to cause damage initiation (indicated by a sudden change in the load-displacement curve) and then increased further to ensure a delamination of approximately 20-30 mm across. (Initially, trials were necessary to establish the required applied displacement to achieve this delamination size.) The critical load associated with damage initiation was recorded. After each test, the interleaved panel was c-scanned to confirm delamination damage had occurred. Two PLA interleaved panels and five PC-interleaved panels were tested.

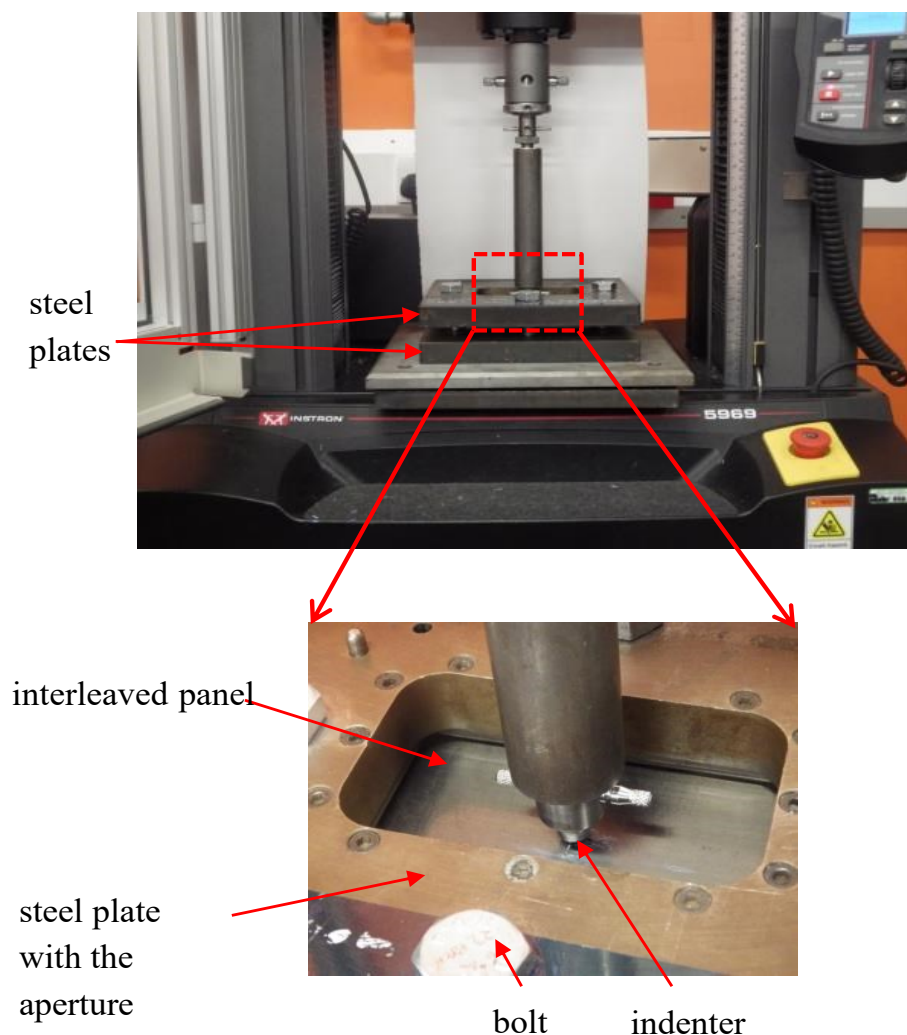


Figure 107. The test setup of the indentation test.

6.2.6 Repair process

For each flexure specimen, the two halves were stacked back together ensuring precise alignment, placed on an aluminium plate, enclosed in a vacuum bag together with release films and air breather cloth as shown in Figure 108. (For PC specimens, the alignment procedure was not required as they were not separated.) The assembly was then heated to the required repair temperature in an oven [Forced convection oven, OF-21E, Lab Companion, Jeio Tech, Korea] and kept at this temperature for 5 minutes, followed by a further 15 minutes at this temperature under vacuum pressure. The repair temperature was 190°C and 175°C for the PLA and PC-interleaved laminates respectively. The repaired specimens were re-tested in three-point flexure tests according to the mentioned procedure to determine the recovery of shear strength measured. The PLA-interleaved flexure specimens were subjected to two damage-repair cycles, except the standard PLA-interleaved flexure specimens underwent one cycle. The PC-interleaved flexure specimens were subjected to three damage-repair cycles.

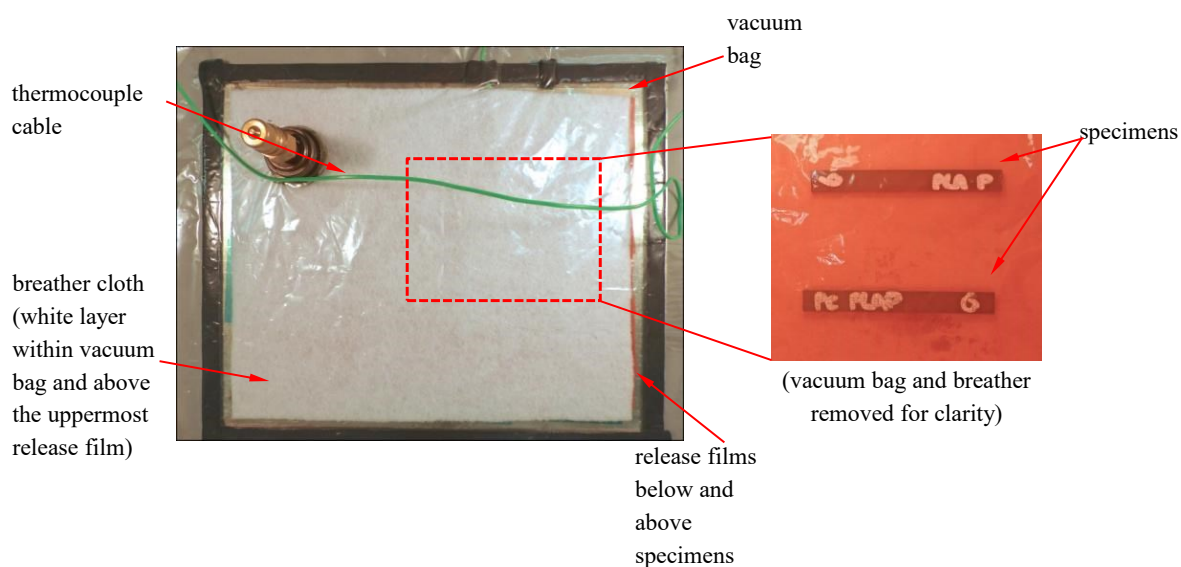


Figure 108. An image of the setup of a vacuum bag for repair in a vacuum bag for the repair.

For the indentation specimens, there was no requirement to align separated sub-laminates (they had only been delaminated in a central zone). For two PLA indentation panels the repair process was identical to that of the flexure specimens. For the PC-interleaved indentation panels, the repair process was performed in a heated press [Press HLP40, Hofer, Germany] so that a pressure of 25 psi was applied during the repair.

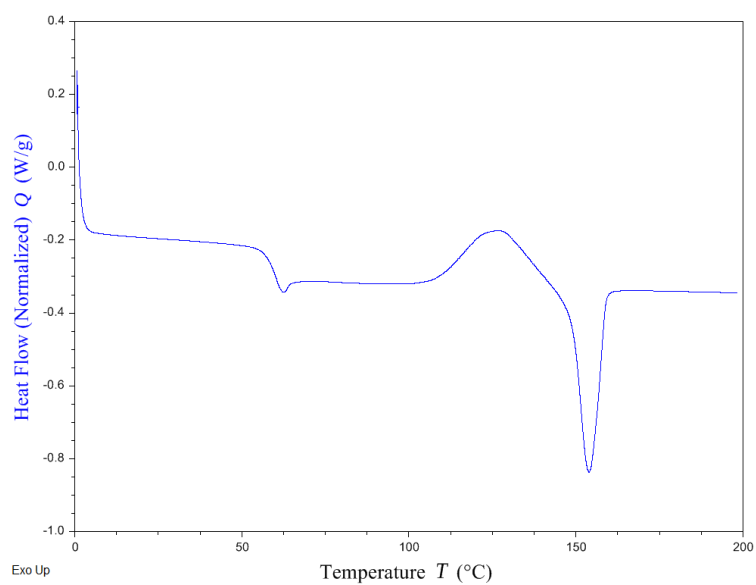
The repaired specimens were c-scanned and then re-tested according to the procedures detailed previously to determine recovery of the critical load for the indentation tests. PLA-interleaved panels and PC-interleaved panels were subjected to two indentation-repair cycles.

6.3 Results and discussion

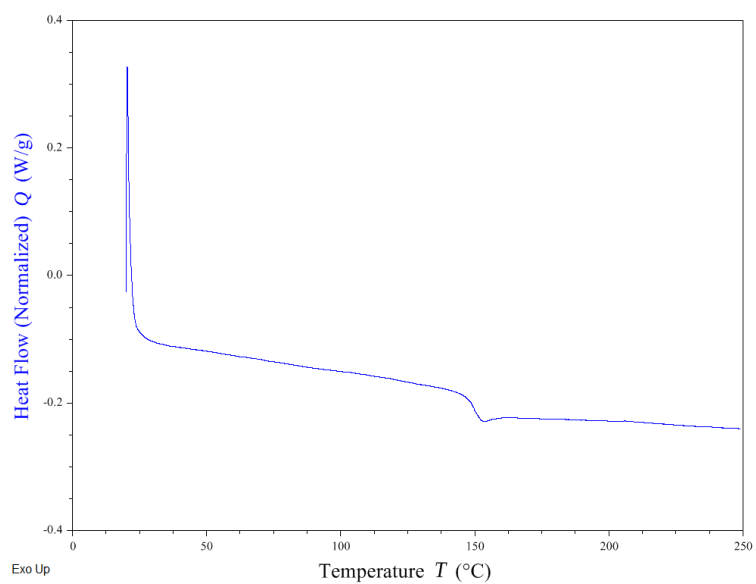
6.3.1 Thermal properties of PLA and PC interleaves

Figure 109 shows typical DSC results of a PLA and a PC specimen following the procedure described in section 6.2.4 and Table 10 summarises the T_g and T_m of the PLA and PC specimens. As mentioned earlier, both T_g and T_m of the PLA and PC interleaves were measured from the re-heating process. T_g is taken from the mid-height of the first drop and T_m is from the transition point of the final drop.

It can be seen that the PLA possesses an average T_g of 58.6°C and an average T_m of 153.8°C, while the PC shows an average T_g of 149.3°C, but no T_m recorded in the intended temperature range.



(a)



(b)

Figure 109. Typical DSC results of (a) a PLA and (b) a PC interleaf specimen for the re-heating process.

Table 10. DSC results of PLA and PC interleaves.

(a) PLA			
	Specimen weight (mg)	Glass transition temperature (T_g) (°C)	Melting temperature (T_m) (°C)
1	4.9	58.7	153.7
2	6.1	58.7	153.8
3	6.5	58.7	153.8
4	5.7	58.7	153.8
5	6.4	58.4	153.9
average	5.9	58.6	153.8

(b) PC			
	Specimen weight (mg)	Glass transition temperature (T_g) (°C)	Melting temperature (T_m) (°C)
1	1.3	149.6	-
2	2.2	148.9	-
average	1.8	149.3	-

6.3.2 Shear strength recovery from flexure tests

The failure mode in all the flexure tests was a shear failure in or adjacent to the interleaf region of the specimens. The upper and lower parts of each specimen fully separated at failure. The associated shear strength was therefore determined from the maximum load causing shear failure using equation (18).

6.3.2.1 PLA-interleaved specimens

Figure 110 shows the typical load-displacement responses of a standard PLA–interleaved specimen (actually that of specimen 3 is drawn) subjected to three-point bend tests before and after repair. The stiffness of the specimen after repair is less than that of the pristine specimen (approximately 7.3% loss in stiffness). The maximum applied load (associated with the onset of interlaminar failure) of the pristine specimen is 366.6 N and after repair, the maximum applied load is 279.5 N (i.e. a 76.2% recovery in interlaminar shear strength). This shows that the maximum applied load is significantly, but not fully, restored.

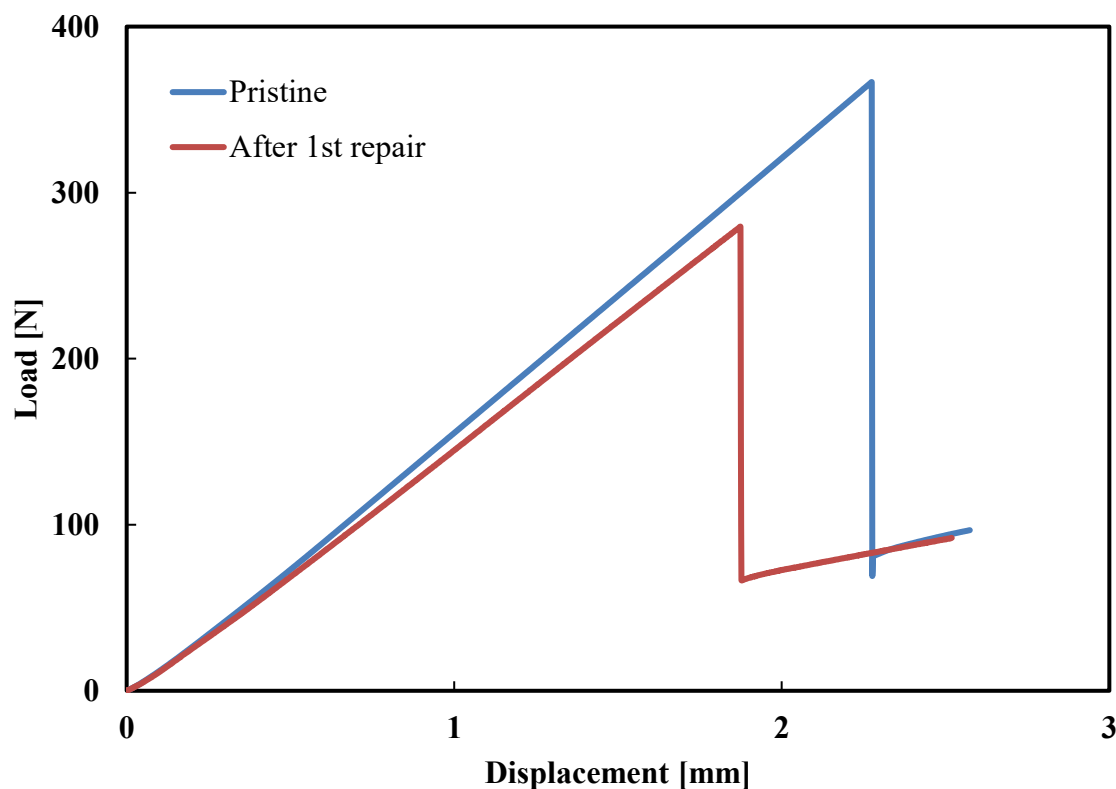


Figure 110. Typical load-displacement responses from three-point bend test of a standard PLA-interleaved specimen before and after repair.

Figure 111 shows the typical load-displacement responses of a plasma-treated PLA-interleaved specimen (actually that of specimen 1 is drawn) subjected to three-point flexure tests in two damage-repair cycles. The plasma-treated PLA-interleaved specimens were manufactured to improve the bond of the PLA interleaf to the composite laminates through the plasma treatment. It can be seen that the stiffness of the specimen at its pristine, after the 1st repair and after the 2nd repair are almost identical. The maximum applied load (associated with the onset of interlaminar failure) of the pristine specimen was 67.8 N. The low maximum applied load could be due to the voids formed within the plastic-treated PLA layer during press curing. Surprisingly, after the 1st repair, the maximum applied load increased to 454.4 N. This might be because the trapped air/voids were removed during the 1st repair and thus the plasma-treated PLA interleaf regained its full shear strength. After the 2nd repair, the load was reduced to 402.5 N.

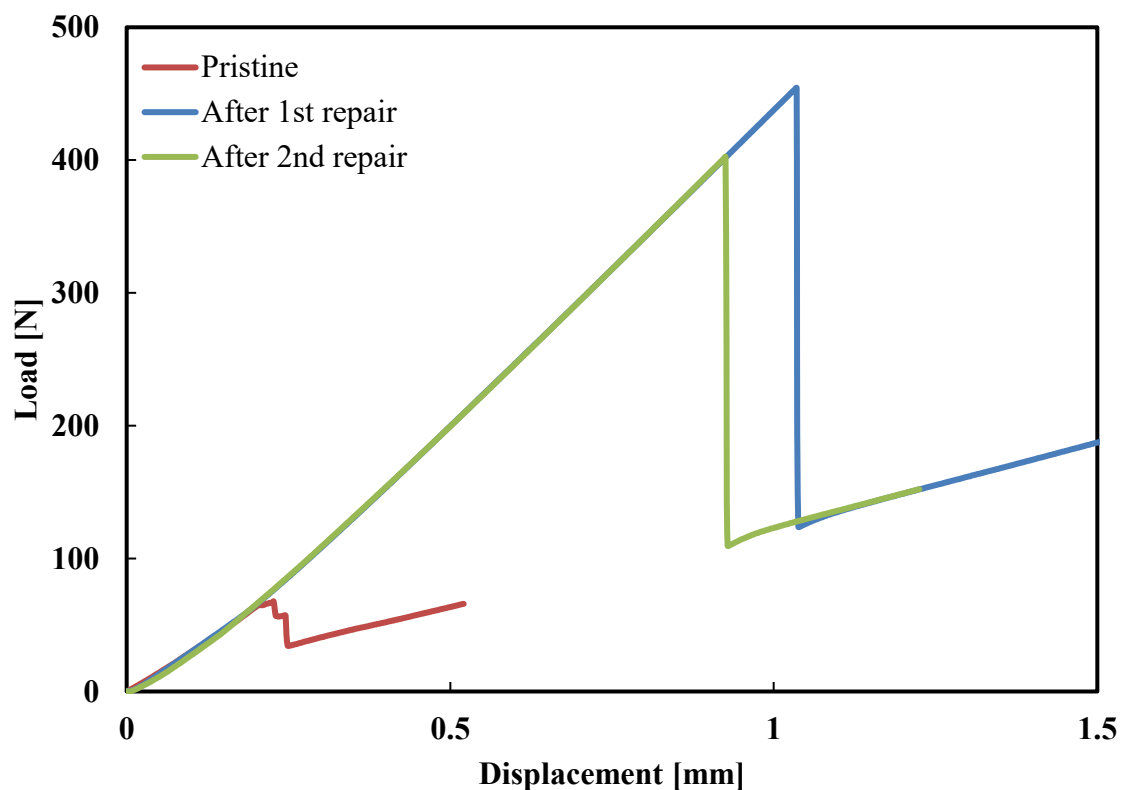


Figure 111. Typical load-displacement responses from three-point bend test of a plasma-treated PLA-interleaved specimen in multiple damage-repair cycles.

Figure 112 shows the typical load-displacement responses of a post-cured PLA-interleaved specimen (specimen 1) subjected to three-point flexure tests in two damage-repair cycles. The post-cured PLA-interleaved specimens were manufactured because of the observed increase in strength of the plasma-treated PLA-interleaved specimens after the 1st repair. It can be seen that the stiffness of the specimen is almost unchanged after each repair. The maximum applied load (associated with the onset of interlaminar failure) is 438.8 N for the pristine specimen, 368.3 N after the 1st repair and 343.2 N after the 2nd repair. This may be attributed to the squeezing-out of the PLA during each repair with the use of pressure.

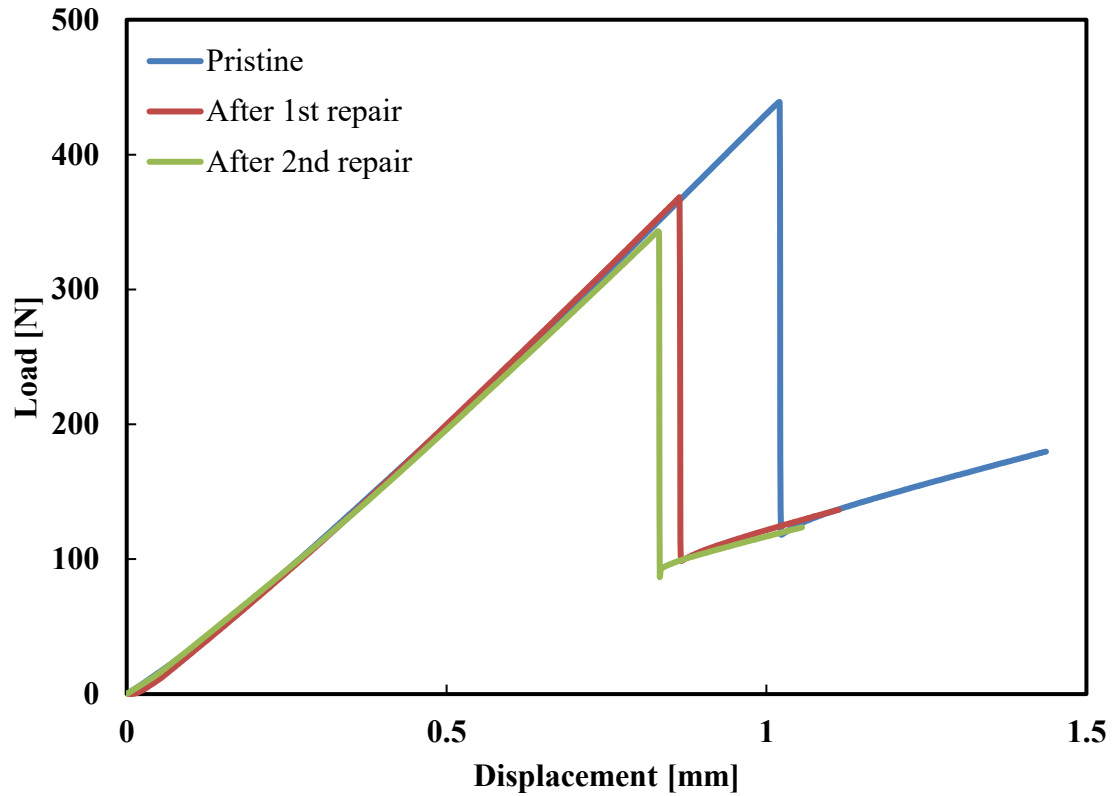


Figure 112. Typical load-displacement responses from three-point bend test of a plasma-treated PLA-interleaved specimen in multiple damage-repair cycles.

Table 11 shows the shear strengths determined from equation (18) for the three batches of PLA-interleaved specimens in multiple flexure test-repair cycles. The percentage recovery of the shear strength achieved by a repair is calculated from

$$Recovery (\%) = \frac{\text{strength after a repair}}{\text{strength before that repair}} \times 100 \quad (19)$$

The overall percentage recovery of the shear strength after a series of repairs is determined using equation (20).

$$\text{overall recovery (\%)} = \frac{\text{strength after final repair}}{\text{pristine strength}} \times 100 \quad (20)$$

Table 11. The shear strength recovery of the three batches of PLA-interleaved specimens measured in different flexure test-repair cycles.**(a) Standard PLA**

Specimen No.	Pristine	After 1 st repair	
	Shear strength (MPa)	Shear strength (MPa)	Recovery (%)
1	9.0	6.9	76.2
2	10.2	3.7	36.1
3	9.7	6.8	69.9
4	8.5	3.4	39.9
Average	9.3	5.2	55.5
Standard deviation	0.7	1.9	20.5

(Test results of specimen 5 were lost.)

(b) Plasma-treated PLA

Specimen No.	Pristine	After 1 st repair		After 2 nd repair		Overall recovery (%)
	Shear strength (MPa)	Shear strength (MPa)	Recovery (%)	Shear strength (MPa)	Recovery (%)	
1	3.4	11.4	335.3	10.1	88.6	297.1
2	1.7	8.4	494.1	8.7	103.6	511.8
3	3	6.7	223.3	6.7	100.0	223.3
4	1.7	9.6	564.7	-	-	-
5	4.9	11.5	234.7	10.5	91.3	214.3
6	3	8.9	296.7	-	-	-
Average	3.0	9.4	358.1	9.0	95.9	311.6
Standard deviation	1.2	1.8	140.6	1.7	7.1	138.5

(Test results of specimen 4 and 6 after 2nd repair were not available as they were used for microscope examination.)**(c) Post-cured PLA**

Specimen No.	Pristine	After 1 st repair		After 2 nd repair		Overall recovery (%)
	Shear strength (MPa)	Shear strength (MPa)	Recovery (%)	Shear strength (MPa)	Recovery (%)	
1	11	9.2	83.6	8.6	93.5	78.2
2	12.2	9.2	75.4	8.2	89.1	67.2
3	13	10.8	83.1	3.7	34.3	28.5
4	12	8.1	67.5	7.2	88.9	60
5	6.8	7.5	110.3	6.8	90.7	100
6	9	5.4	60	5.5	101.9	61.1
Average	10.6	8.4	80	6.7	83	65.8
Standard deviation	2.3	1.8	17.4	1.8	24.4	23.6

Figure 113 plots the average shear strength measured in the flexure test-repair cycles on the standard, plasma-treated and post cured PLA-interleaved specimens. It can be seen that the average shear strength of the pristine standard PLA-interleaved specimens is 9.3 MPa. After a repair, the average shear strength is 5.2 MPa. The shear strength recovery is on average 55.5%. But, as shown in Table 11a, there is a large variability in the percentage recovery.

As mentioned earlier, the plasma-treated PLA specimens were manufactured in order to improve the bond of the PLA to the epoxy matrix by the plasma treatment. However, Figure 113 shows that initial shear strength is low after the plasma treatment. This indicates that initially the plasma-treated PLA possesses a significantly low shear strength or a low interlaminar shear strength with the epoxy matrix. Upon the 1st repair, the shear strength is remarkably increased by almost 360%, to 9.4 MPa. This can be due to the removal of trapped air and/or voids in the PLA layer during the 1st repair. The average recovery of shear strength is about 96% after the 2nd repair, which is very good.

A further batch of the plasma-treated PLA specimens was prepared and post-cured as described earlier because after the 1st repair the shear strength of the plasma-treated specimens was significantly improved. The results shown in Figure 113 indicate that the pristine post-cured PLA-interleaved specimens have an average shear strength of 10.6 MPa. This is substantially larger (by over 350%) than that of the plasma-treated PLA specimens and is consistent with the results after the 1st repair of those specimens. For the post-cured PLA specimens, the average shear strength after the 1st repair is about 80% of the initial value and after a 2nd repair, the overall recovery reduces to 66% on average of the initial value. Again, there was significant scatter in the test result. (One specimen showed a very low shear strength overall recovery of 28.5% as shown in Table 11c).

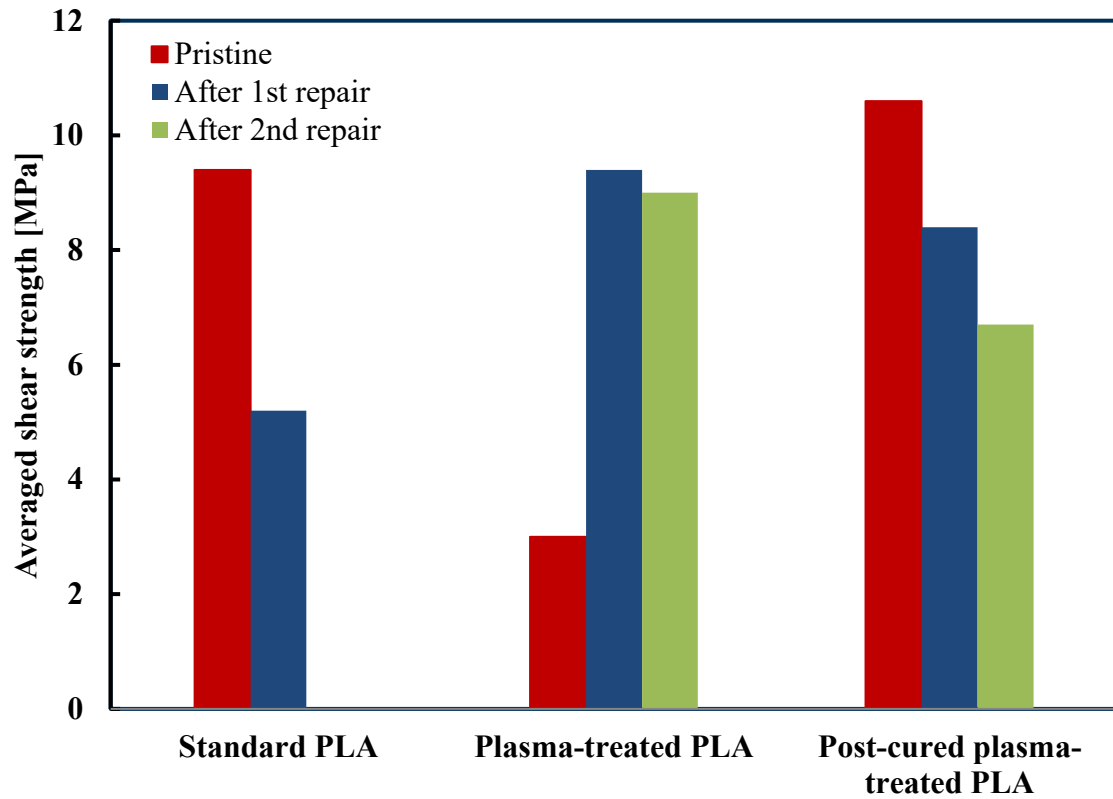


Figure 113. The plot of the average shear strength of the three batches of PLA-interleaved specimens subjected to three-point bend tests.

6.3.2.2 PC-interleaved specimens

Figure 114 shows typical load-displacement curves measured in three-point bending tests on a PC-interleaved specimen (specimen 2) in the pristine state and after several damage-repair cycles. It can be seen that there is a small loss (around 7.5%) in the flexural stiffness after the 1st repair but the stiffness then remains unchanged after subsequent repairs. The loss of the stiffness could be due to damage at the PC/composite interface which cannot be repaired subsequently. As for the PC-interleaved specimens, there is significant, but partial, recovery of the maximum applied load. For the specimen represented in Figure 114, the maximum load drops by 23.3% after the 1st repair and by a further 4% after the 2nd repair. After the 3rd repair this specimen shows a further drop in the maximum load of 9.2%. The reduction in maximum load is because the PC was squeezed out from specimen edge during repair.

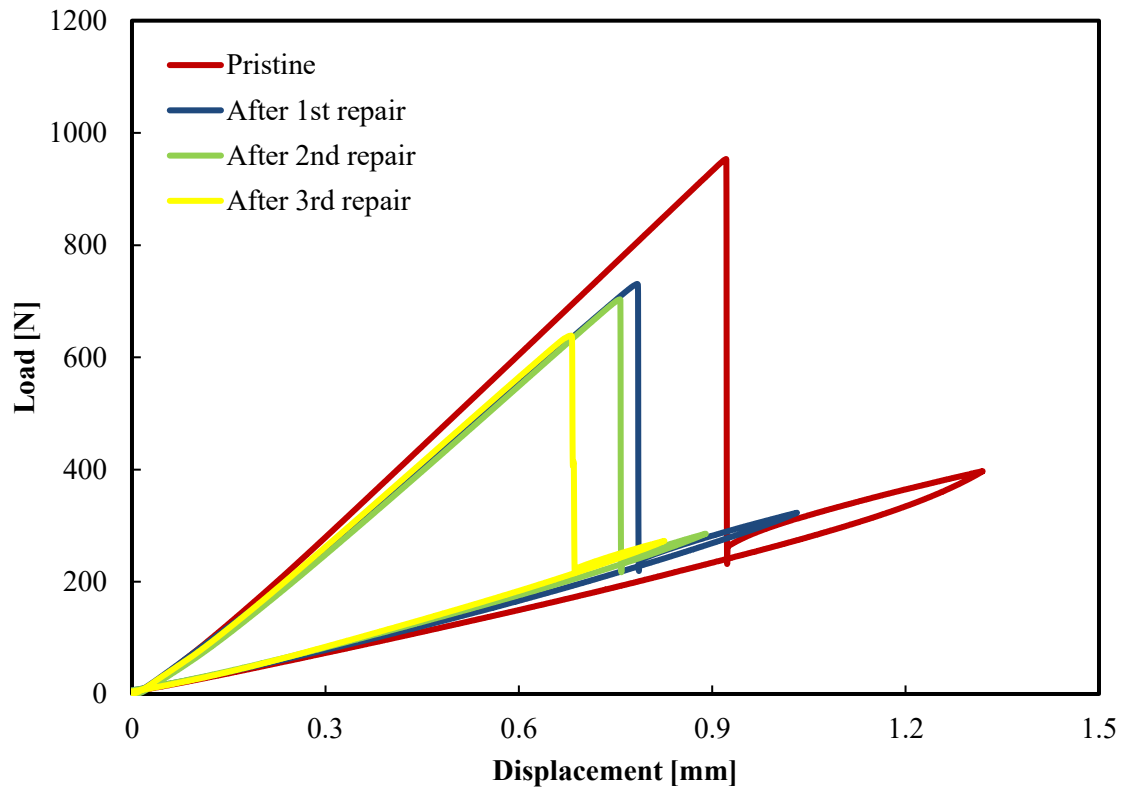


Figure 114. Typical load-displacement responses from three-point bend tests of PC specimens in multiple damage-repair cycles.

Table 12 summarises the shear strength measured in the PC-interleaved specimens for the pristine state and after each of the subsequent repairs and Figure 115 plots these values. Table 12 also includes values for the percentage recovery and overall recovery as calculated from equations (19) and (20).

Table 12. The shear strength of the PC specimens measured in different flexure-repair cycles.

	Pristine	After 1 st repair		After 2 nd repair		After 3 rd repair		
Specimen No.	Shear strength (MPa)	Shear strength (MPa)	Recovery (%)	Shear strength (MPa)	Recovery (%)	Shear strength (MPa)	Recovery (%)	Overall recovery (%)
1	24.8	20.0	80.7	16.4	81.7	17.0	103.9	68.4
2	22.4	17.4	77.7	16.9	96.9	15.3	90.7	68.3
3	23.8	18.2	76.6	17.9	98.1	15.9	88.9	66.9
4	24.1	18.5	76.6	17.5	95.1	17.4	99.0	72.1
5	24.5	18.5	75.4	16.7	90.0	16.3	98.1	66.5
Average	23.9	18.5	77.4	17.1	92.4	16.4	96.1	68.4
Standard deviation	0.9	0.9	2.0	0.6	6.7	0.8	6.2	2.2

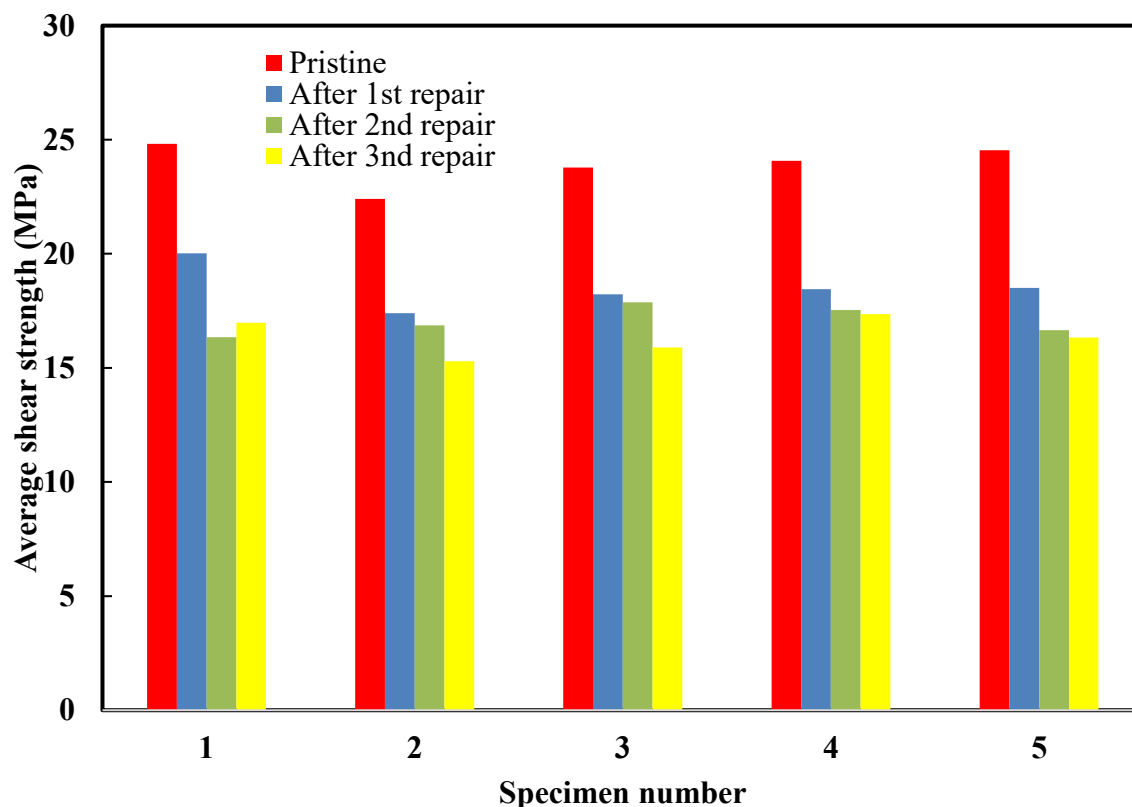


Figure 115. The plot of the average shear strength of the PC specimens subjected to three-point bend tests in multiple damage-repair cycles.

The figure shows that the average pristine shear strength of the PC-interleaved specimens is 23.9 MPa and the shear strength is restored to 18.5 MPa (on average 77.4% of the initial value) after the first repair. The average recovery of shear strength after the second repair is 92.4% and after the third repair is 96.1%. The average overall recovery of shear strength is 68.4%. Also, the standard deviations of the results of the PC-interleaved specimens are smaller than those of the PLA-interleaved specimens, which indicates that the PC is a more reliable interleaf material than PLA for the easy repair concept.

6.3.3 Static indentation test results

6.3.3.1 Post-cured PLA-interleaved panels

Figure 116 shows the load-displacement curves of a post-cured PLA-interleaved panel (the test panel 2) in two indentation-repair cycles. It is showed that after the 1st repair, the structural stiffness of the panel is fully recovered. However, the recovery of the critical applied load associated with the onset of interlaminar failure is rather low (54.9% of the pristine value). After the 2nd repair, the structural stiffness is again fully restored, and the recovery of the critical applied load is 82.1%. The overall recovery of the critical applied load is 45.1%.

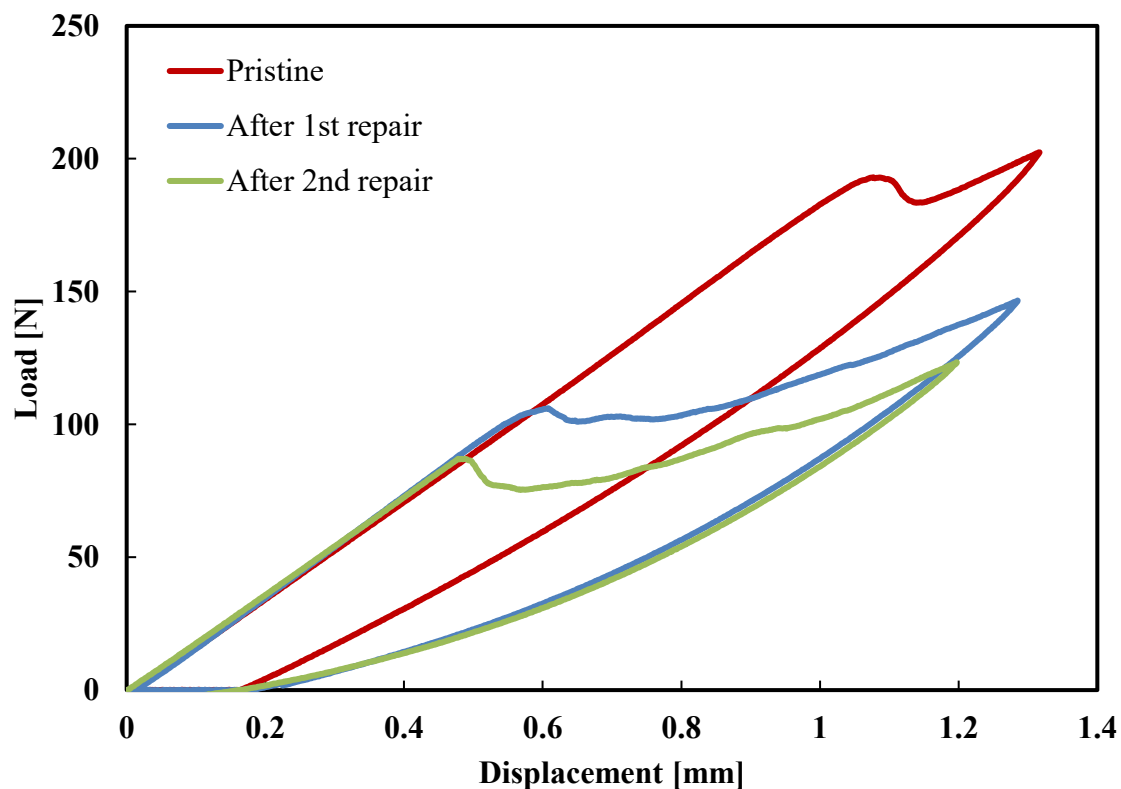


Figure 116. The load-displacement curves from static indentation tests a post-cured PLA-interleaved indentation panel in two indentation-repair cycles.

Table 13 summarises the load causing interlaminar failure of the two panels tested and Figure 117 plots the load causing interlaminar failure measured in static indentation tests of the panels for the

pristine state and after each of the subsequent repairs. It can be seen that the repair of these two panels are quite different and this probably reflects the difference on the PLA content within the panels.

Table 13. The interlaminar failure load measured in static indentation tests of the post-cured PLA-interleaved panels.

Specimen No.	Pristine	After 1 st repair		After 2 nd repair		Overall recovery (%)
	Maximum applied load (N)	Maximum applied load (N)	Recovery (%)	Maximum applied load (N)	Recovery (%)	
1	114	90	79.8	59	65.6	51.8
2	193	106	54.9	87	82.1	45.1
Average	153.5	98	67.4	17.1	73.9	48.5

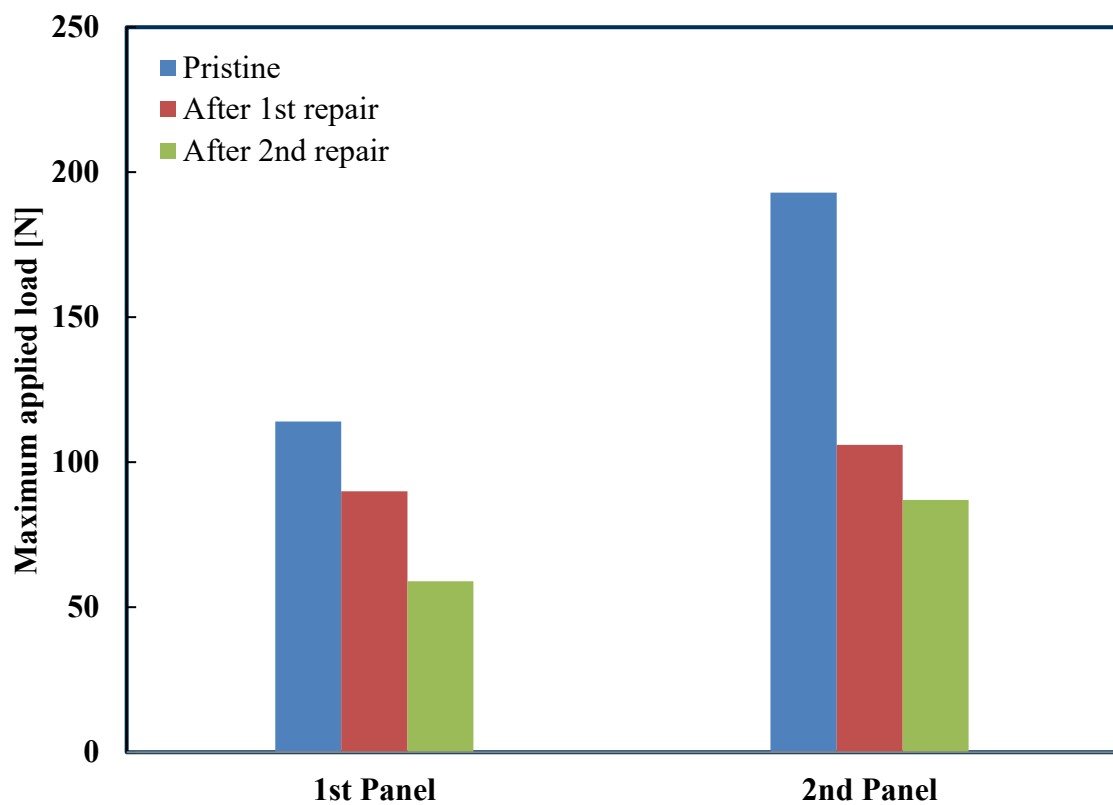


Figure 117. The plot of the maximum applied load measured in indentation tests of the post cured PLA-interleaved panels in multiple indent-repair cycles.

The post-cured PLA-interleaved panels for indentation tests were c-scanned after manufacture as described earlier in section 6.2.2. Figure 118a shows the c-scan attenuation image of a post-cured interleaved panel (actually that of panel 2 is shown) using the ultrasound signals reflected from the

glass plate positioned below the panel in the immersion tank. It is showed that the PLA interleaves are uniform after the post-curing, as no signals are reflected or scattered when travelling throughout the panel due to defects or cracks. However, some sections of the panel have less PLA (the black areas) due to the leaking of the PLA during the post-curing. After the 1st indentation test, the c-scan image (see Figure 118b) shows that the damage occurred at the indentation area. (A significant change in colour, i.e. the ultrasound signal attenuation level, can be observed due to the use of a transducer of 5 MHz frequency. This is because signal attenuates greater at higher frequency.) Additional damage near to the indentation damage area can also be observed. This may have been the result of the less content of PLA at that area. After the 1st repair, this damage disappeared, but a slender void inclined at 45° appeared (see Figure 118c) probably due to leakage of the PLA during heating for the repair. (Again, the transducer of 5 MHz frequency was used.) The indentation damage due to the 2nd test is clearly visible in the C-scan (see Figure 118d). After a 2nd repair, the damage is partially repaired, but a significant amount of small 'green area' appears on the ultrasound image. This is presumably due to voids created by further PLA leakage. After the 3rd indentation test, the c-scan image (see Figure 118f) shows that the damage occurred at the indentation area again. Additional damage around to the indentation damage area can also be observed. This may have been the result of the less content of PLA at that area.

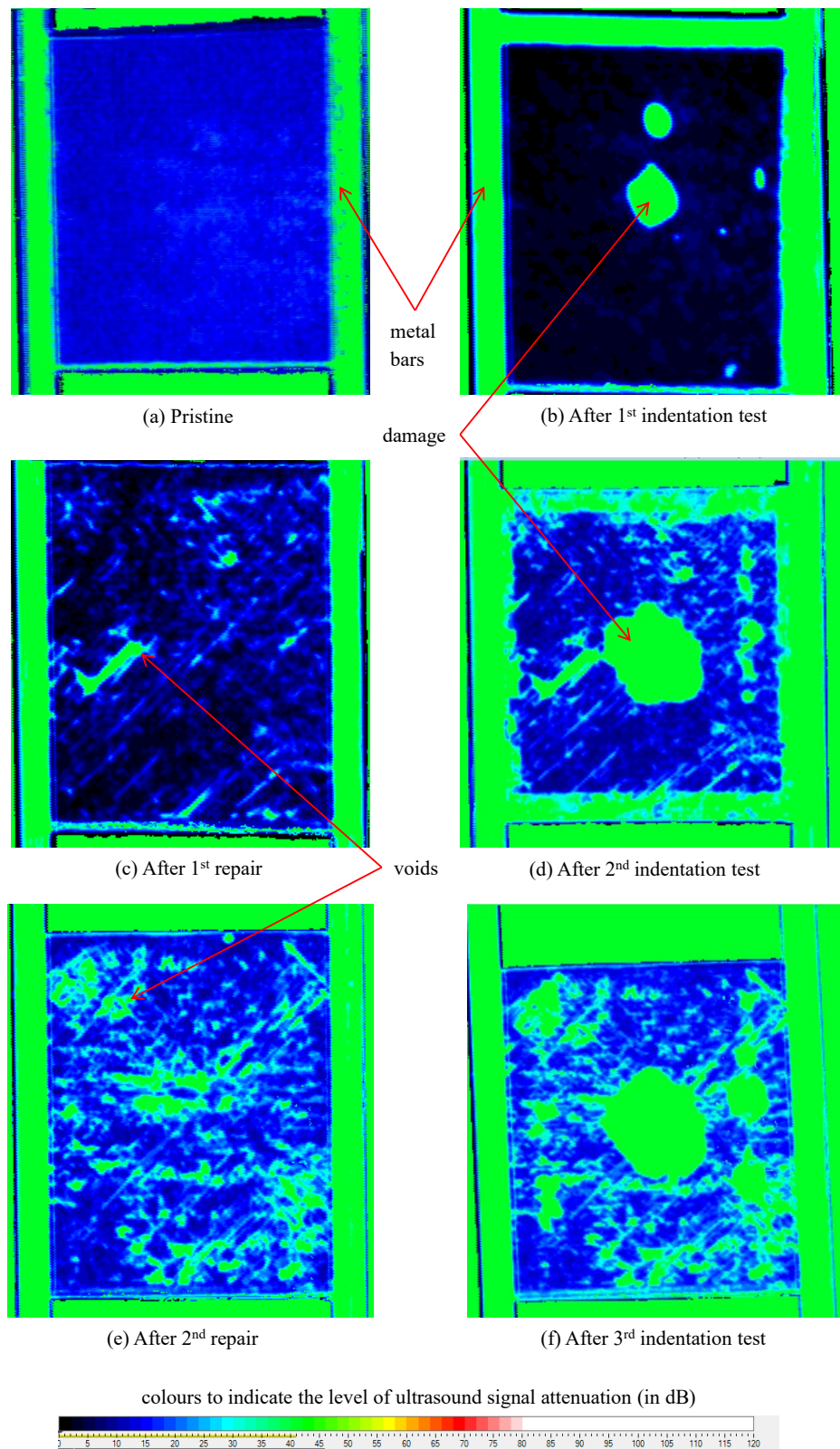


Figure 118. The c-scan image of a post-cured PLA interleaved panel (panel 2) at (a) as-cured, (b) 1st indented, (c) 1st repaired, (d) 2nd indented and (e) 2nd repaired stages.

6.3.3.2 PC-interleaved panels

Figure 119 shows the load-displacement curves of a typical PC-interleaved panel in the pristine state and after two repairs. It can be seen that there is a loss in stiffness (of approximately 19%) after the first repair and that the stiffness after the second repair is very similar to that after the first repair. These curves also show a significant reduction in interlaminar failure load after the repairs.

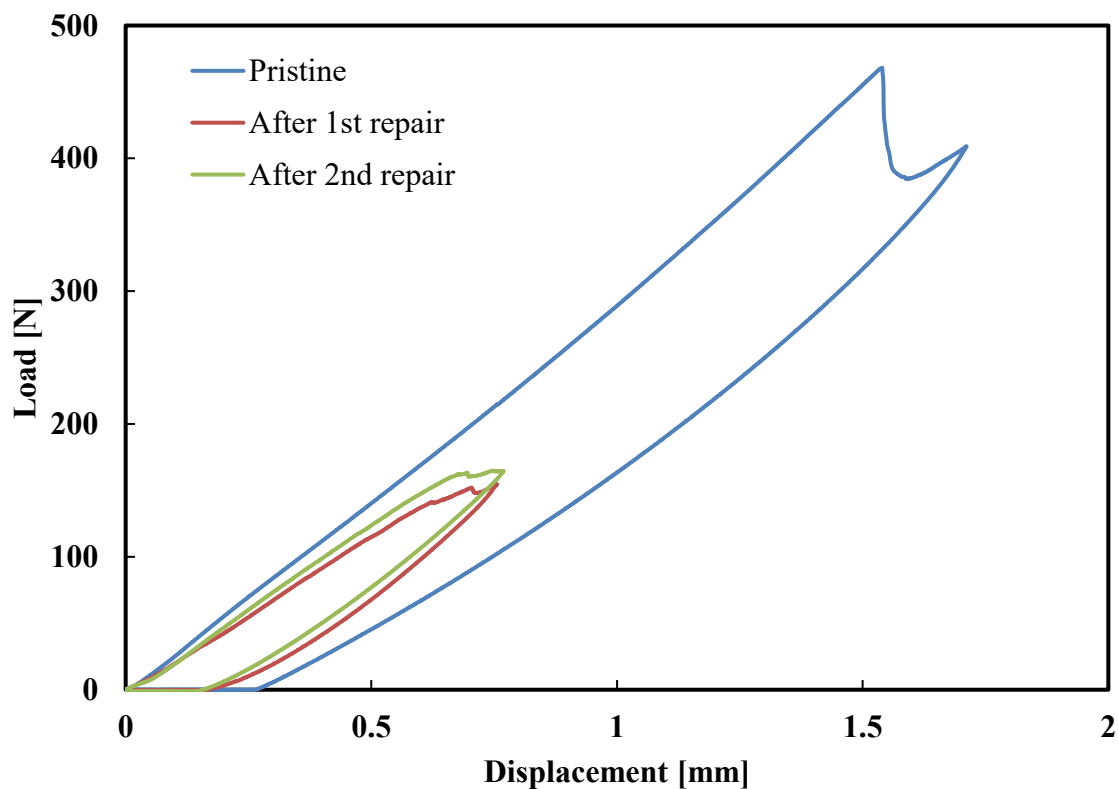


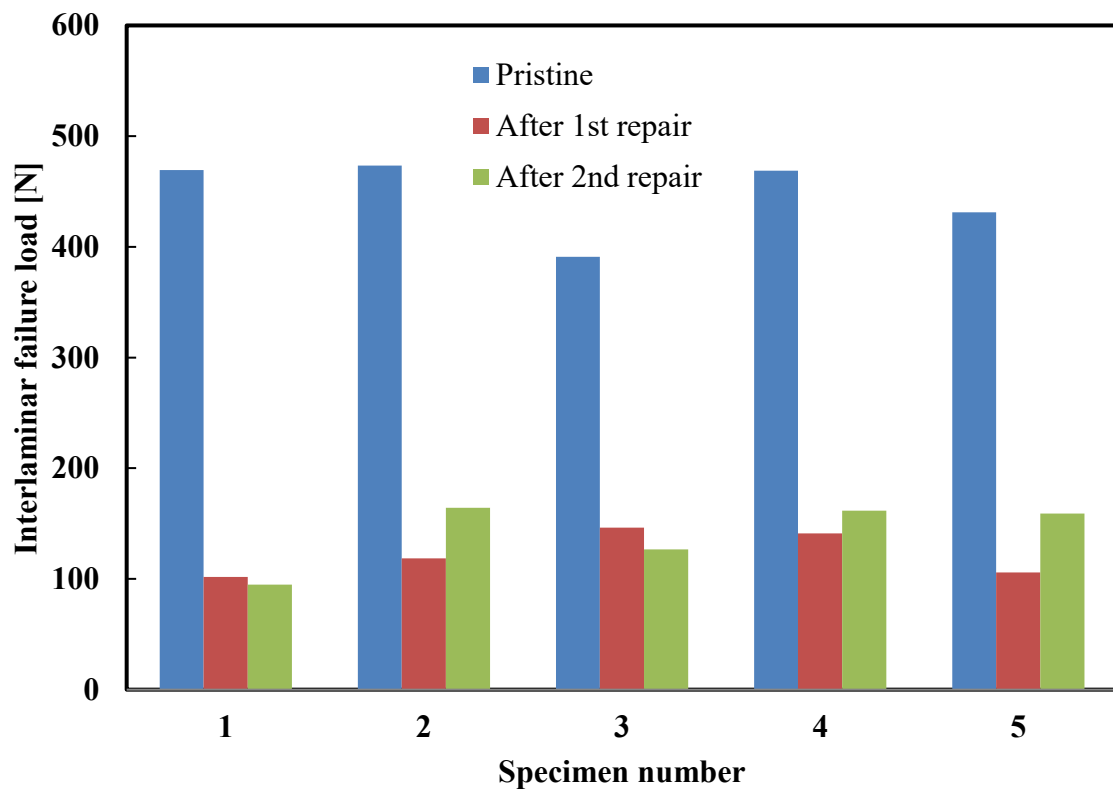
Figure 119. The load-displacement curves of a PC-interleaved panel in the indentation-repair cycle.

Table 14 summarises the interlaminar failure loads and recovery percentages measured in the PC-interleaved specimens in the pristine and repaired states and these values are plotted in Figure 120.

Table 14. The maximum applied load measured in indentation tests of the PC-interleaved panels.

Specimen No.	Pristine	After 1 st repair		After 2 nd repair		Overall recovery (%)
	Maximum applied load (N)	Maximum applied load (N)	Recovery (%)	Maximum applied load (N)	Recovery (%)	
1	469	102	21.7	95	93.1	20.3
2	473	118	24.9	164	139.0	34.7
3	391	146	37.3	127	87.0	32.5
4	469	141	30.1	162	114.9	34.5
5	431	106	24.6	159	150.0	36.9
Average	446.6	122.6	27.7	141.4	116.8	31.8
Standard deviation	35.5	20.0	6.1	30.0	27.6	6.6

It can be seen that the average interlaminar failure load of the PC-interleaved specimens is 446.6 N in the pristine state with a standard variation of 35.5 N. After the 1st repair, the average interlaminar failure load is 28% of the pristine value. After the 2nd repair, the average overall recovery of the interlaminar failure load slightly increased to almost 32%.

**Figure 120.** The plot of the maximum applied load measured in indentation tests of the interleaved panels in multiple indent-repair cycles.

The interleaved panels for indentation tests were c-scanned after the autoclave curing as mentioned before. Figure 121a shows the c-scan attenuation image of a PC-interleaved panel using the ultrasound signals reflected from the glass platform. It is shown that the PC interleaves are generally uniform after the curing, but there are small black areas in the PC interleaves (in these areas, the signal attenuation is lower than the others). This may be the voids formed during the autoclave curing of the interleaved panels. After the 1st indentation test, the ultrasound image of the panel (see Figure 121b) indicates the damage due to the indentation test. After the 1st repair, the damage disappeared, as shown in Figure 121c. However, a large number of black areas occurred. This can be due to the formation of voids during the repair which affects the re-bonding of the PC interleaves. The ultrasound image taken after the 2nd indentation test (see Figure 121d) shows the presence of damage as pointed. After the 2nd repair, the damage is no longer visible in the ultrasound image (Figure 121e), but there is a significant amount of voids formed.

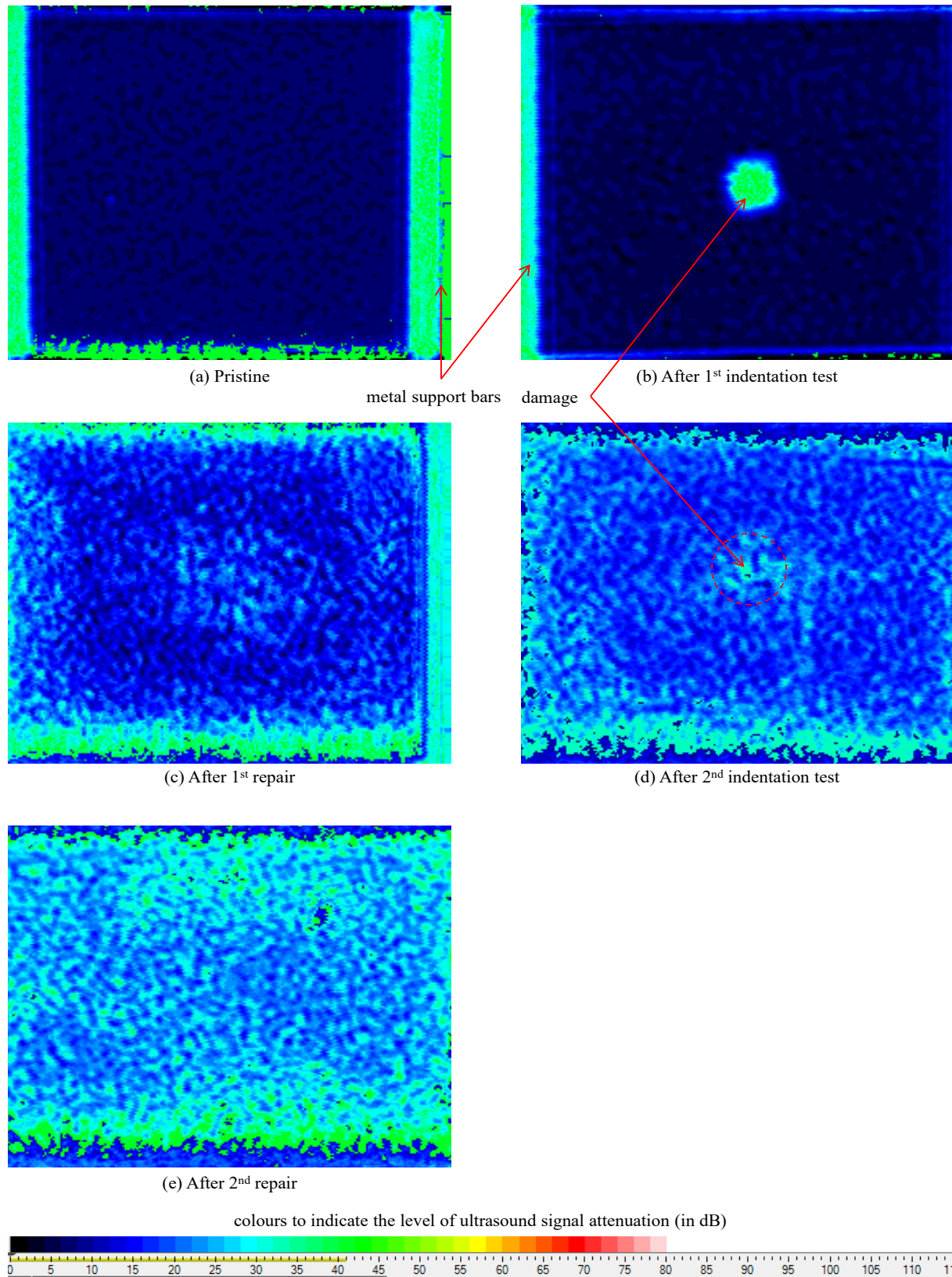


Figure 121. The c-scan images of a typical PC interleaved panel at (a) as-cured, (b) 1st indented and (c) 1st repaired, (d) 2nd indented and (e) 2nd repaired stages.

6.3.4 Microscope examination

6.3.4.1 PLA-interleaved specimens

Figure 122 shows the polished cross section of a PLA-interleaved flexure specimen from each of the three batches, i.e. standard, plasma-treated and post-cured. It can be seen that there is a distinct PLA layer at the mid-plane of all specimens after curing.

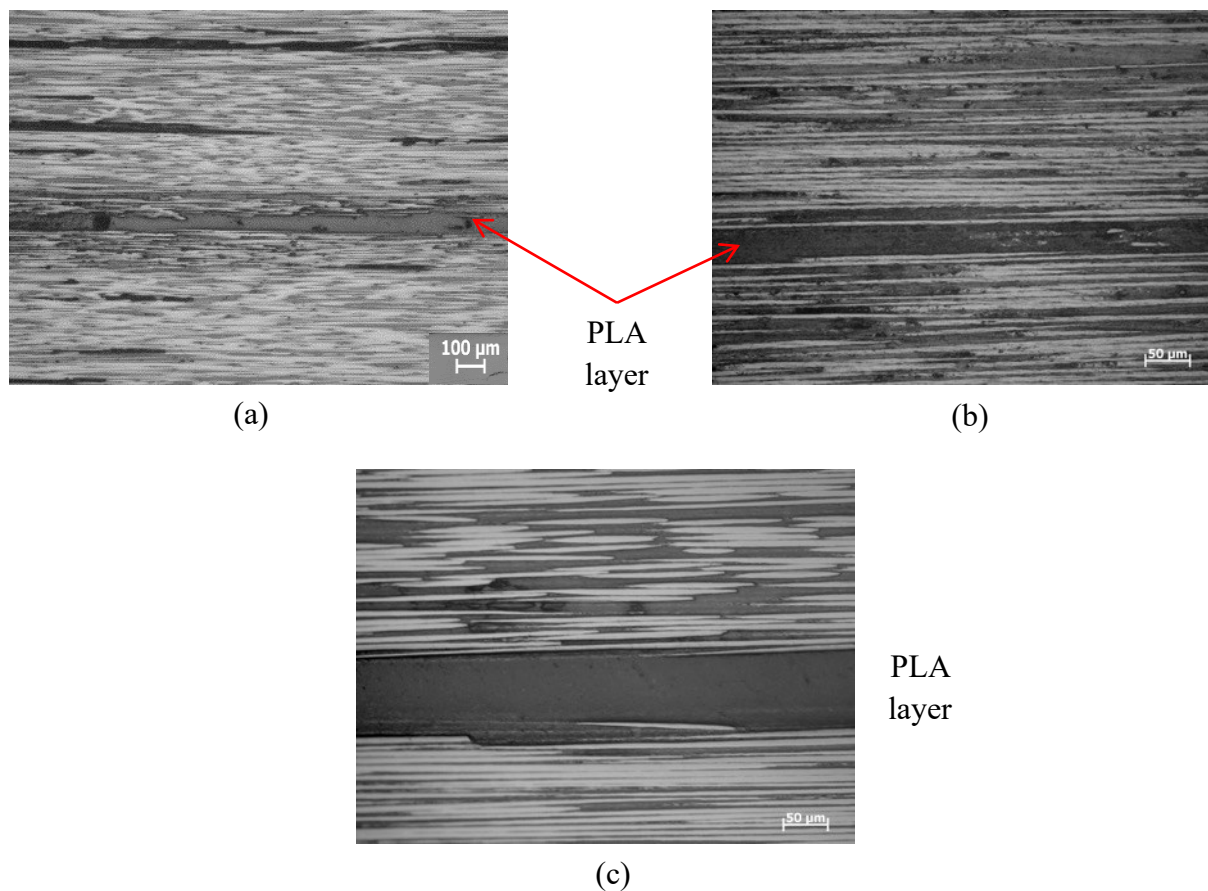


Figure 122. Microscope images of the typical polished cross-sectional area of (a) standard PLA, (b) plasma-treated PLA and (c) post-cured PLA-interleaved flexure specimens.

Figure 123 shows the fracture surface of the PLA-interleaved flexure specimens after the 1st flexure test. For the standard PLA specimens, failure at the interface between the PLA layer and carbon fibre/epoxy composite is observed (see Figure 123a). This accounts for the low strength recovery as the PLA cannot re-bond to the cured composite surface. This also indicates that the bond of the

standard PLA and the composite is weak. For the plasma-treated PLA specimens, Figure 123b has revealed that the PLA had remained well on the composite plies. It can be also seen that the PLA, on both fracture surfaces, leaks out significantly (neat carbon fibre/epoxy surface can be clearly seen). For the post-cured PLA specimens, the PLA can be seen on both fracture surfaces (Figure 123c).

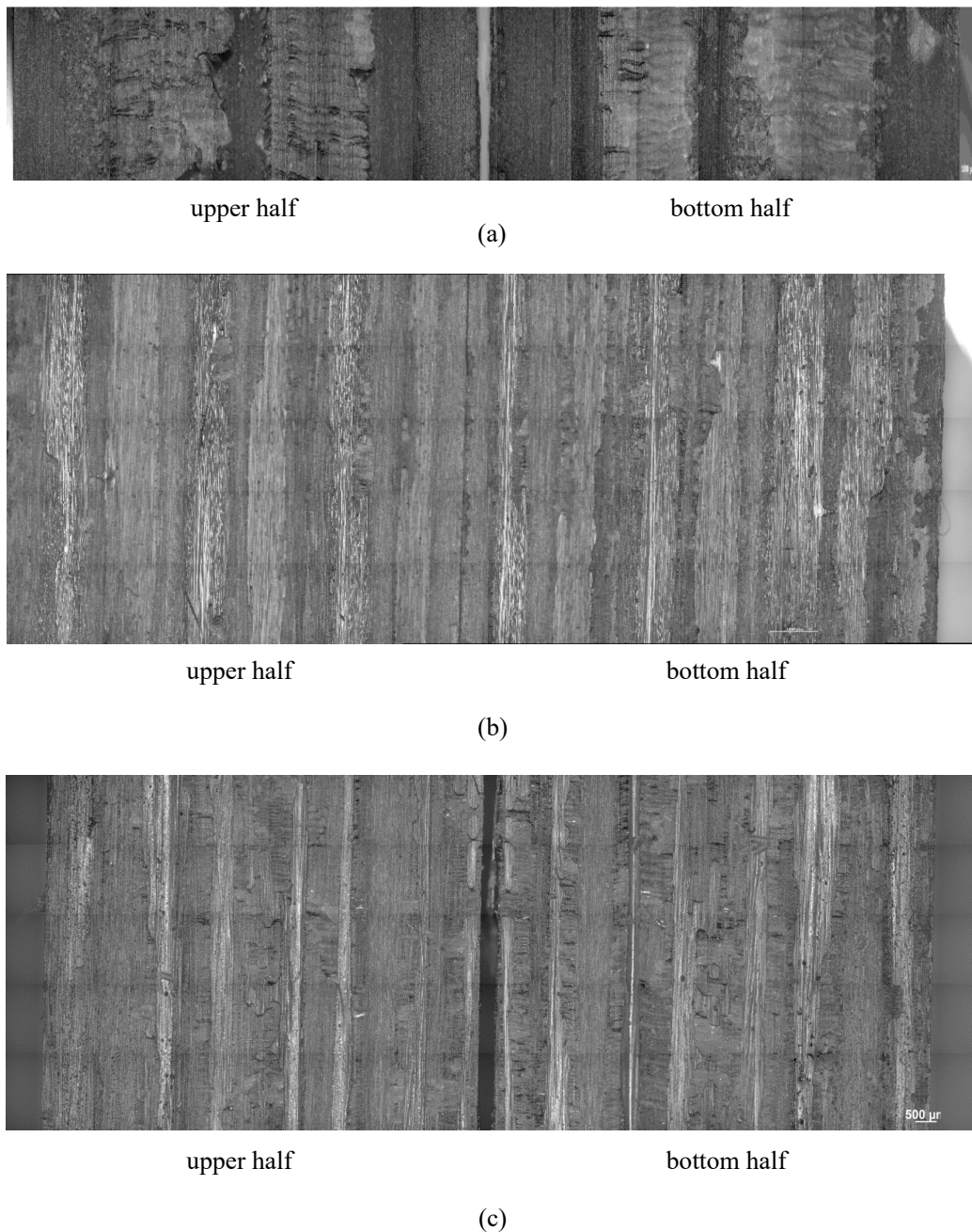


Figure 123. Microscope images of the fracture surface of (a) standard PLA, (b) plasma-treated PLA and (c) post-cured PLA-interleaved flexure specimens.

Figure 124 shows the images of the fracture surface of the PLA-interleaved flexure specimens after the 1st flexure tests. The carbon fibre/epoxy composite regions are highlighted in these images. These regions, unfortunately, cannot be repaired. This may account for the loss in strength recovery of the PLA-interleaved composites.

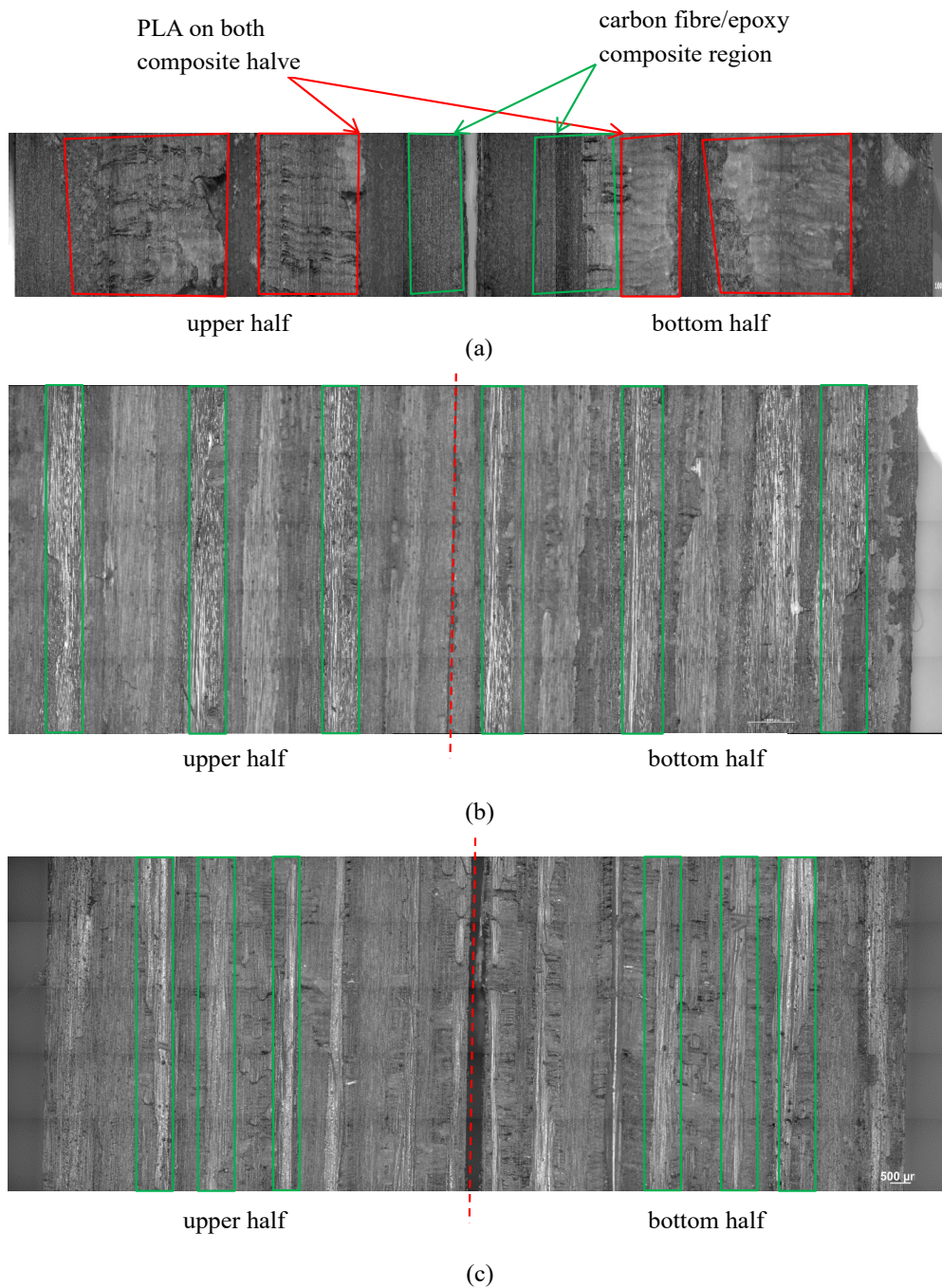


Figure 124. Microscope images of the fracture surface of (a) standard PLA, (b) plasma-treated PLA and (c) post-cured PLA-interleaved flexure specimens with the highlighted carbon fibre/epoxy composite regions.

Figure 125 shows the cross section of a PLA-interleaved indentation test panel after curing and after the 2nd damage-repair cycles. Figure 125a shows that the PLA layer at every interface can be seen, but there is considerable variation in the thickness of the PLA layers. This may be due to the squeezing-out of the PLA during the post curing at 190°C (the PLA is molten at this temperature) but it could also be due to leakage of the PLA during curing. It can be also seen that after a repair, a significant leaking of the PLA occurred. (Some PLA can be seen on the outermost unidirectional composite plies.) Interior off-axis ply damage was also observed.

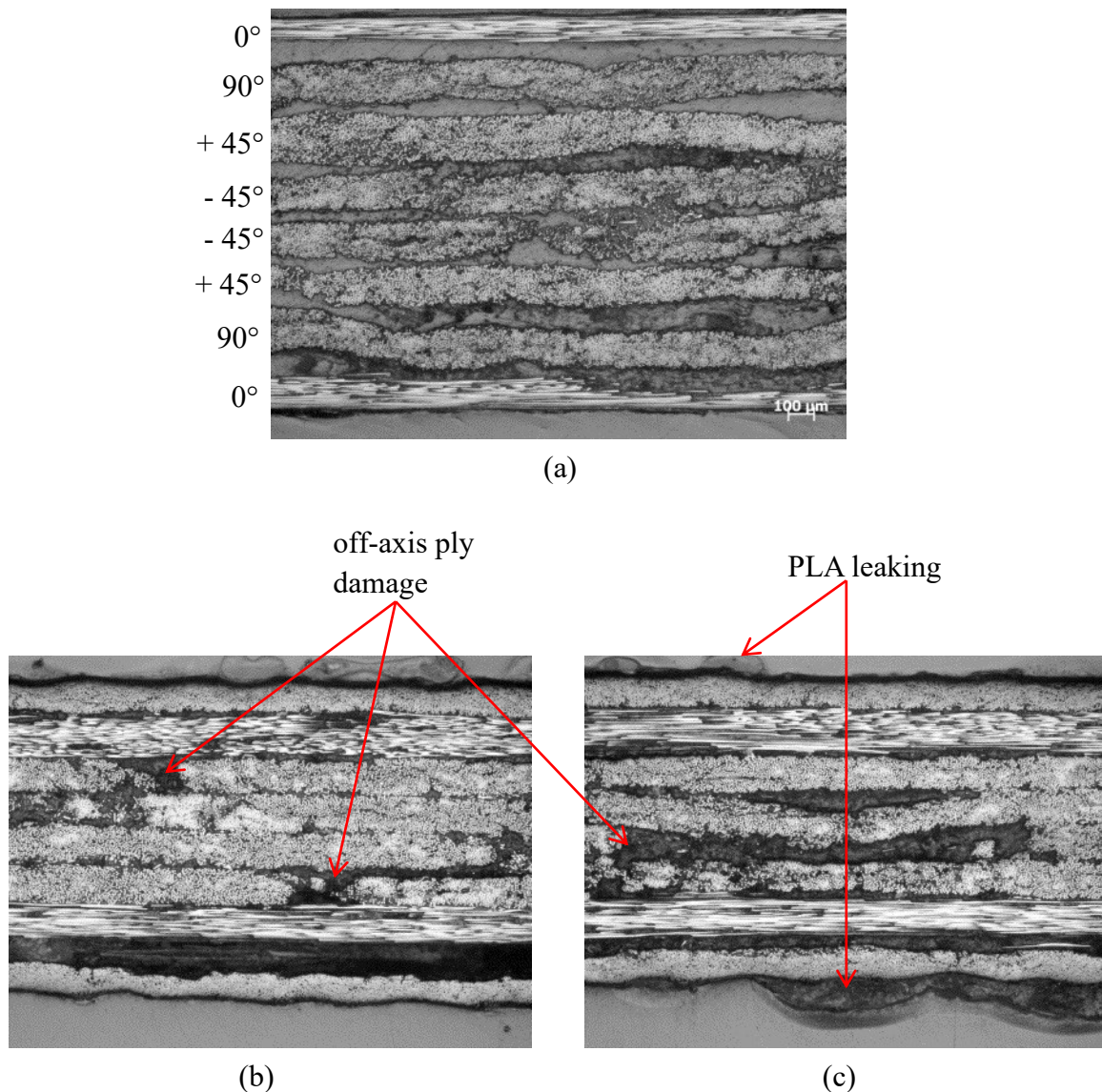


Figure 125. Microscope images of the cross-sectional area of a PLA-interleaved panel subjected to static indentation test.

6.3.4.2 PC-interleaved specimens

Figure 126 shows the polished cross-sectional images of a PC-interleaved flexure specimen after curing and after the 1st indentation test. It is showed that a distinct PC interleaf layer is at the mid-plane of the specimen after curing. After testing, the fracture surface of the specimen shows that the PC had remained well on both surfaces (see the highlighted area in red of Figure 126b). The microscope image also shows that the shear damage is within the PC interleaf layer.

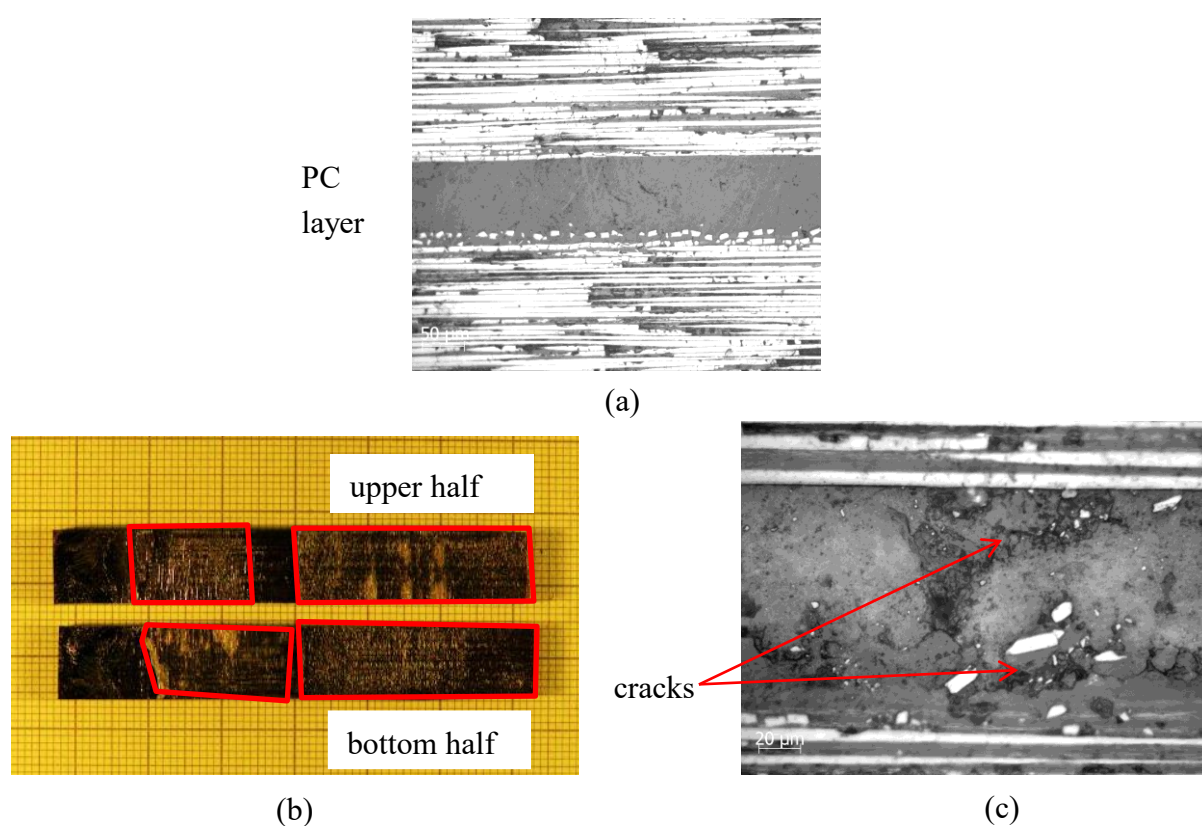


Figure 126. Microscope image of (a) a PC-interleaved flexure specimen after curing and a PC-interleaved flexure specimen with (b) an image of the fracture surface and (c) a microscopic image to show the shear damage within the PC interleaf layer.

Figure 127 shows the polished cross section of a PC-interleaved panel for static indentation tests after curing and after indentation. It can be seen that after curing, distinct PC layers can be seen at interfaces with composite plies (Figure 127a). However, after indentation, interlaminar damage can be seen (Figure 127b). Such damage cannot be repaired as the PC cannot re-bond to cured composite

plies. This results in low repair efficiency. Another observation is that large voids were formed after repair (Figure 127c). This is due to the moisture trapped within the PC layers which was vaporised during repair and gave rise to a significant increase in thickness. The voids were also detected in the c-scan results (the ‘black area’) as shown in Figure 127.

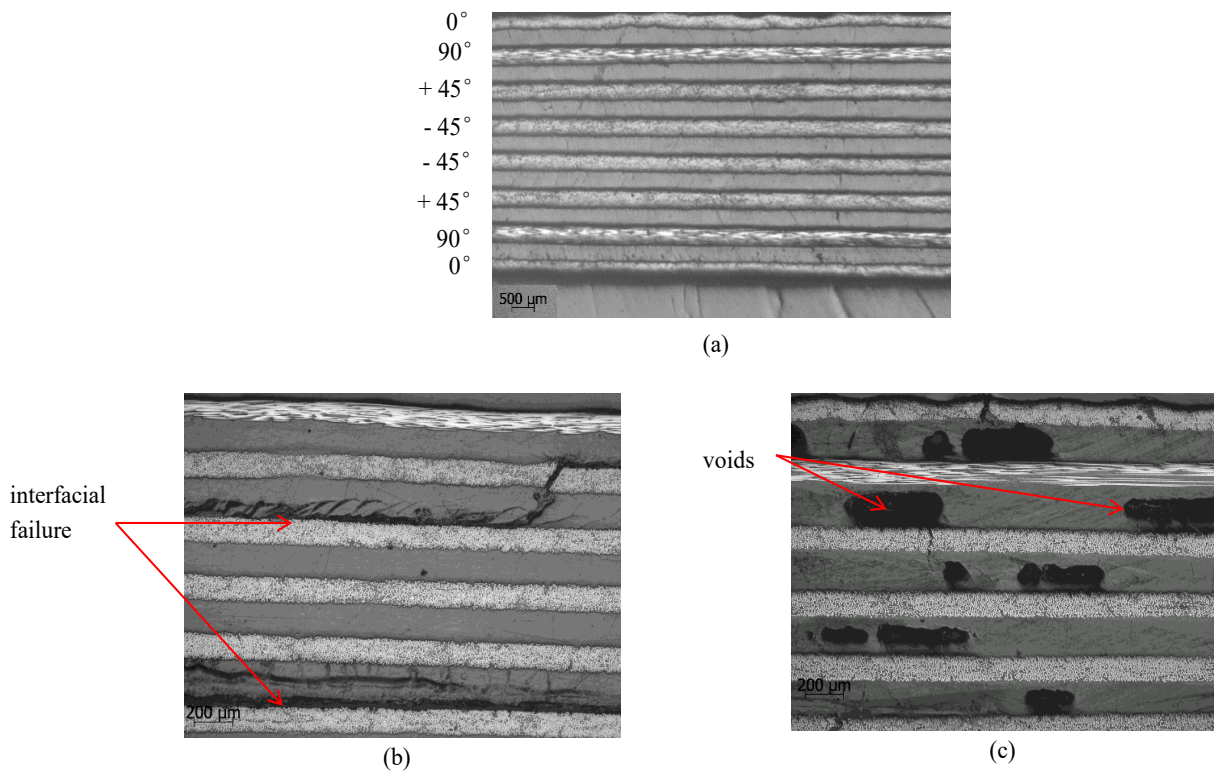


Figure 127. Microscope images of a PC-interleaved panel (a) after curing, (b) after indentation and (c) after repair to show the formation of large voids in the PC interleaves.

6.4 Discussion and summary

PLA-interleaved and PC-interleaved carbon fibre/epoxy composites were manufactured and subjected to three-point flexure and static indentation tests to investigate whether shear damage can preferentially occur within the thermoplastic interleaf and this damage can be repaired by heating the specimens to a suitable temperature under vacuum pressure.

Three batches of PLA-interleaved specimens were examined. The standard PLA-interleaved specimens (i.e. before any repairs) had an average interlaminar shear strength of 9.3 MPa as measured in three-point flexure tests. (This was significantly lower than the value of 24 MPa which have been measured in the ASTM short beam shear tests but the short beam shear test is widely regarded as a qualitative rather than quantitative method (194). Other methods, such as interlaminar shear device tests (195) and thick adherent shear tests (196), could be used to measure the shear strength of the interleaf materials.) After a repair the average shear strength recovery was 55.5%. This may be due to the fracture at PLA-composite interface and such fracture cannot be repaired. The plasma-treated PLA-interleaved specimens had an average pristine interlaminar shear strength of 3.0 MPa measured in three-point flexure tests. After the 1st repair cycle the average shear strength recovery was over 350% and after the 2nd repair cycle the overall recovery of the shear strength of 311.6% was achieved. Such high recovery is due to the significantly low initial shear strength. The post-cured PLA-interleaved specimens had an average pristine interlaminar shear strength of over 10.6 MPa measured in three-point flexure tests. After the 1st repair cycle the average shear strength recovery was 80% and after the 2nd repair cycle the overall recovery of the shear strength of 65.8% was achieved. A possible explanation for this reduction is the loss of PLA after each repair. The strength recovery in static indentation tests of two specimens was lower at 45% and 52%, again after two repair cycles. The main reason for not achieving better strength recovery figures was that a lot of the PLA was observed to leak out of the composite during the repair process.

As expected, the PC-interleaved specimens did have a better interlaminar shear strength with an average of just over 24 MPa measured in the three-point flexure tests on the pristine material. A small loss in the flexure stiffness after the 1st repair was observed. This may be due to the presence of interlaminar damage at PC-composite interface which cannot be subsequently repaired. Another possible explanation may be that the separated parts of a specimen, when stacked together for a repair, were not in good alignment to each other. Moreover, after the 1st repair, the average shear

strength recovery was 77% and after the 3rd repair cycles these specimens showed an average overall strength recovery of 68%. This may be due to the loss of PC after each repair. Unfortunately the performance in the static indentation tests was rather disappointing with an average overall strength recovery of 32%. The problem observed with the PC-interleaved static indentation test specimens was the development of voids during the repair process which significantly weakened the interleaf regions. The voids were due to moisture absorbed by the PC interleaves (perhaps during immersion in the C-scan tank) which was vaporised during the repair at 175°C.

The work reported here has demonstrated the potential of using an interleaving strategy for polymer matrix composite laminates to enable the easy repair of interlaminar damage. Both interleaf materials examined here achieved strength recovery but failure at the interleaf-composite interface, leakage (in the case of PLA) and the development of voids (in the case of PC) prevented full strength recovery. Alternative interleaf materials, which overcome these problems and have shear strengths closer to the interlaminar shear strength of the polymer composite, can be investigated to make easy-repair interleaved composites that are suitable for practical applications.

Chapter 7 Conclusions and future work

suggestions

7.1 Conclusions

Thermoplastic-interleaved carbon fibre/epoxy composites for controllable stiffness, shape memory and easy repair capabilities were investigated in this PhD study. The following sections summarise the findings and key conclusions from this study.

7.1.1 The investigation of polystyrene (PS)-interleaved composites with shape memory capability for deployable structures

- PS-interleaved carbon fibre/epoxy composites were used in three deployable configurations. In all cases, specimens showed excellent re-shaping and shape recovery on heating. This indicates that PS-interleaved composites have the potential for the use in deployable structure applications.

7.1.1.1 Integrated PS-interleaved composite hinges

- A challenge in the integrated PS-interleaved composite hinge design was to ensure composite plies could slide relative to each other, when folded or deployed at elevated temperature. Two designs to allow this sliding were examined. In one design, all plies except the innermost one in the interleaved zone were cut so that during folding these cuts opened so permitting sliding, and during deployment they closed. In the second design all plies except the outermost one contained gaps so that during folding these gaps closed and during deployment they opened. Specimens of both designs showed internal ply sliding and so re-shaping capability, although

for the first design specimens did not fully closed during deployment due to the PS which migrated into the cuts and stopped ply sliding. This indicates that conventional composite laminates integrated with interleaved composites, with appropriate arrangement, can exhibit shape memory capability.

- Simple beam theory can be used to assess the amount of springback and recovery of the integrated hinges. The measured springback and recovery angles of the integrated hinges were in good agreement with the predicted values.
- Delamination occurred in one integrated hinge specimen during the shape recovery process. It was suspected that the specimen was over-tightened by the clamping mechanism which constrained the sliding of internal composite plies.
- Although integrated hinges exhibited shape memory capabilities with cuts or gaps, these increased the complexity of manufacture and reduced the structural integrity of the interleaved laminates.

7.1.1.2 An interleaved, sine-wave actuator for the linear actuation capability

- The actuator was straightened under tension at elevated temperature with a small applied force. This indicates that if interleaved composites were used for deployable structures, only small actuation forces would be required.
- This structure achieved internal ply sliding without the need of cuts or gaps.
- The success of the linear actuation has shown that PS-interleaved composites have the potential of the use for deployable trusses.
- However, delamination occurred in the actuator during the straightening process due to through-thickness tensile stresses and the small length differences between the carbon fibre/epoxy laminae. The design of the actuator would need to be improved to address this problem.

7.1.1.3 An expanded composite mesh

- The mesh was manufactured in a flat shape and then re-shaped to an expanded configuration. This greatly reduced the manufacturing complexity, compared with the manufacture of the wave actuator.
- The success of expansion indicates that interleaved composites have the potential for shape forming manufacture from simple flat sheets.
- Zones of non-interleaved composite were induced at the crest and trough positions of the wave form of the laminates within the expanded mesh. This prevented the initiation of delamination during the expansion process.

7.1.2 Viscoelastic investigation of PS-interleaved composites

- Results from repeated flexure tests at room temperature and 120°C showed that PS-interleaved composites were capable of changing the flexural stiffness repeatedly on heating. This is important if multiple shape changes of interleaved composite are required.
- Simple beam theory can be used to predict the flexural modulus of PS-interleaved composites and showed reasonable agreement with the measured values at room and elevated temperatures.
- Significant loading-rate dependence and stress relaxation were observed in three-point flexure testing of PS-interleaved composites at 120°C. In this testing of multidirectional laminates, failure was observed at off-axis plies at higher loading rates.
- Results from the flexure tests and from DMTA shear tests performed on the polystyrene were used in the development and validation of finite element modelling for the prediction of the rate-dependent behaviour of PS-interleaved composites.

7.1.3 FE modelling of interleaved composites with shape memory capability

- A finite element approach to model the re-shaping and shape recovery of a PS-interleaved circular arc specimen was developed using the Abaqus 'Restart' facility.
- The bending stresses of the composite laminae in the re-shaped state were of right magnitude but were smaller than the simple beam theory prediction. This is due to the low stiffness used in the FE model for the PS interleaf in the high temperature, low stiffness state, rather than the zero value assumed in the beam theory analysis.
- With further validation (e.g. for springback), this FE approach could be used in the design of complex interleaved composite structures with shape memory capability.
- Finite element analysis was also used to model the behaviour of a PS-interleaved composite specimen subjected to three-point flexure tests at 120°C at a range of loading rates. The model included the viscoelastic properties of the PS measured in DMTA tests. The model predicted that to achieve a given displacement, a higher force is required when loaded at higher crosshead speed. This was in agreement with experimental results that showed the re-shaping of the PS-interleaved composite is rate-dependent. The modelling also showed that the applied force for a specimen loaded at higher crosshead speeds relaxed towards to the lower crosshead speed cases when fixed at a given displacement. Finally, the modelling showed that upon shape recovery, the specimen initially exhibited the fastest recovery speed, but the recovery rate became slower in a manner of an overdamped system.
- Further investigation on the PS-interleaved composites with interior off-axis composite plies should be performed to investigate the damage of the interior off-axis composite plies observed in experiments.

7.1.4 Interleaved carbon fibre/epoxy composites for easy repair of interlaminar damage

- The easy repair concept for interlaminar damage using thermoplastic-interleaved carbon

fibre/epoxy composites has been demonstrated. Both polylactide (PLA)-interleaved and polycarbonate (PC)-interleaved carbon fibre/epoxy composites subjected to three-point flexure tests have shown that shear damage could be confined to occur within the thermoplastic interleaf and could be readily repaired by the application of heat and pressure.

- Both PLA and PC interleaf materials are not suitable for practical applications; they showed low shear strengths and significant leakage during repairs. Alternative interleaf materials should be investigated.

7.2 Suggestion for future work

- Development of a deployable mesh structure

An expanded mesh made of PS-interleaved composites was developed. It showed that the mesh could be expanded from the as-manufactured flat state with external loading at elevated temperature and would retain this expanded state on cooling. However, it would be beneficial if the mesh, still manufactured in the flat shape, could deploy to the expanded shape on heating. Three techniques could be investigated.

In the 1st strategy, a mesh is manufactured in a flat shape and re-shaped to the expanded configuration on heating, just as described in this thesis. A technique should be explored to relax the elastic stresses in the mesh while maintaining the expanded form. If this can be achieved, then the mesh can be flattened at 120°C, cooled and deploy on re-heating.

In the 2nd strategy, the use of an interleaved composite with a two-stage shape memory should be explored. With such a composite the mesh can again be manufactured in the flat form, expanded at an elevated temperature and then cooled, just as described in this thesis. However, with a two-stage shape memory composite, it can be re-heated to an elevated temperature, lower than the first, and the mesh flattened and then cooled to retain the flat shape. On re-heating to the lower elevated temperature, the mesh should deploy to the expanded form.

Finally, in the 3rd strategy, the self-expansion of the mesh by using residual stress could be explored. A mesh of an interleaved carbon fibre/epoxy laminates using a type of cross-ply configuration will be manufactured in a flat shape. The aim will be to explore how the residual stresses, developed with the cross-ply configuration, can be used to drive the deployment where the laminate is re-heated to above the T_g of the polystyrene.

- FE analysis of the shape memory capability of interleaved composites including viscoelastic properties

FE analysis was used to model the shape memory capability of PS-interleaved composite and to predict rate effect during re-shaping and shape recovery. The next stage is to apply the current FE approach to the design of complex interleaved composite structures with shape memory capability and to validate this with experimental observations.

- Alternative interleaf materials for easy repair investigation

Both PLA- and PC-interleaved carbon fibre/epoxy composites showed good strength recovery when subjected to several damage-repair cycles in three-point bending tests. However, they did not exhibit acceptable strength recovery when tested in static indentation tests. Also, the shear strength of both interleaf materials was not sufficiently high for practical applications. Therefore, alternative interleaf materials should be considered. Polymers such as polysulfone (PSU), polyether ether ketone (PEEK), poly(ether sulfone) (PES) and phenoxy should be investigated. If leaking occurs on these interleaf materials when repairing, glass microspheres could be added into interleaf materials to increase the viscosity.

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Appendix

A.1. Calculation of effective flexural modulus at room temperature and 120°C

This section describes the calculation of the measured and theoretical flexural modulus of the PS-interleaved flexure specimens tested at room temperature and 120°C.

A.1.1. Apparent flexural modulus at room temperature and 120°C

The measured apparent flexural modulus (E) of the PS-interleaved specimens at room temperature and 120°C can be calculated from

$$E = \frac{L^3 m}{4bh^3} \quad (21)$$

where L is the span, m is the slope from load-displacement curve, b is the specimen width and h is the overall thickness of the laminate.

A.1.2. Effective flexural modulus at room temperature and 120°C

The effective flexural modulus (E_f) of PS-interleaved specimens can be calculated according to simple beam theory. Figure 128 shows a PS-interleaved specimen in a UD layup for the calculation deduction.

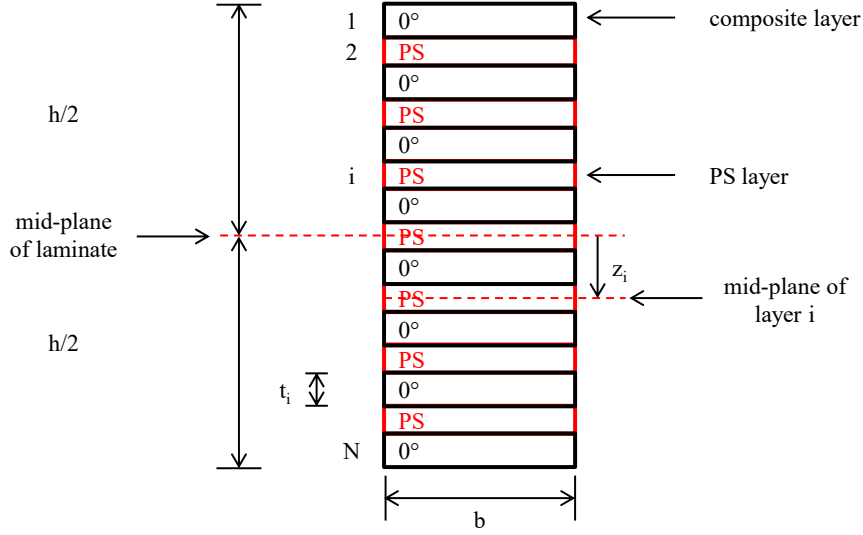


Figure 128. A PS-interleaved flexure test specimen with definitions of terms.

Two cases are analysed. At room temperature, the laminate behaves as an integral beam element (i.e. the flexural modulus of PS at room temperature is assumed to be sufficiently large, although this is very small compared with the stiffness of the composite layer in the longitudinal direction (in this case the fibre direction)). Also, it is assumed that sections remain plane and normal to the axis of the laminate, when the laminate is deflected. Therefore, the effective flexural modulus of PS-interleaved specimen at room temperature (E_f^{RT}) can be calculated from

$$E_f^{RT} = \frac{12E_i}{h^3} \sum_{i=1}^N \left(\frac{t_i^3}{12} + t_i z_i^2 \right) \quad (22)$$

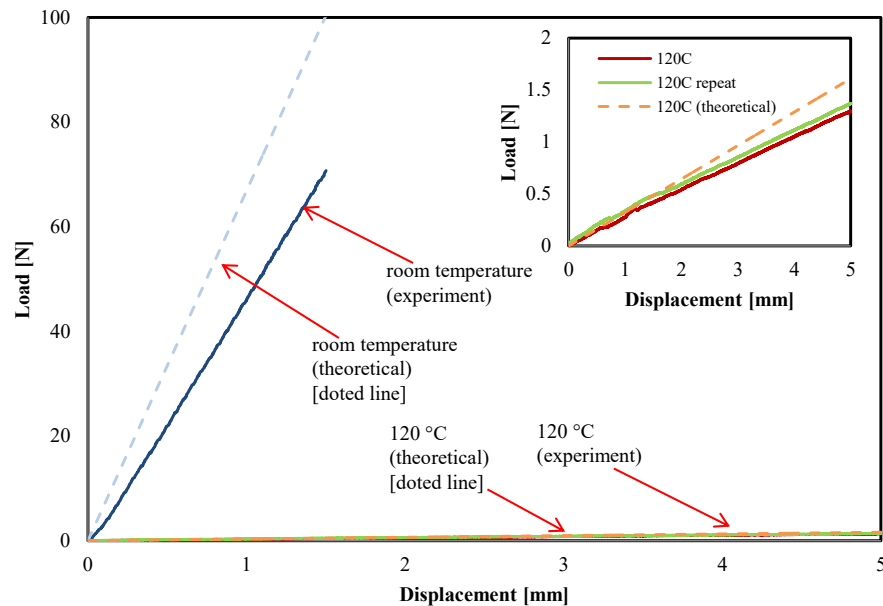
where E_i is the stiffness of unidirectional composite layer in the longitudinal direction.

At 120°C, the stiffness of PS layers is negligible and the composite layers can slide relatively to each other. The sections now are no longer plane. The effective modulus of PS-interleaved specimen at 120°C (E_f^{HT}) is given by

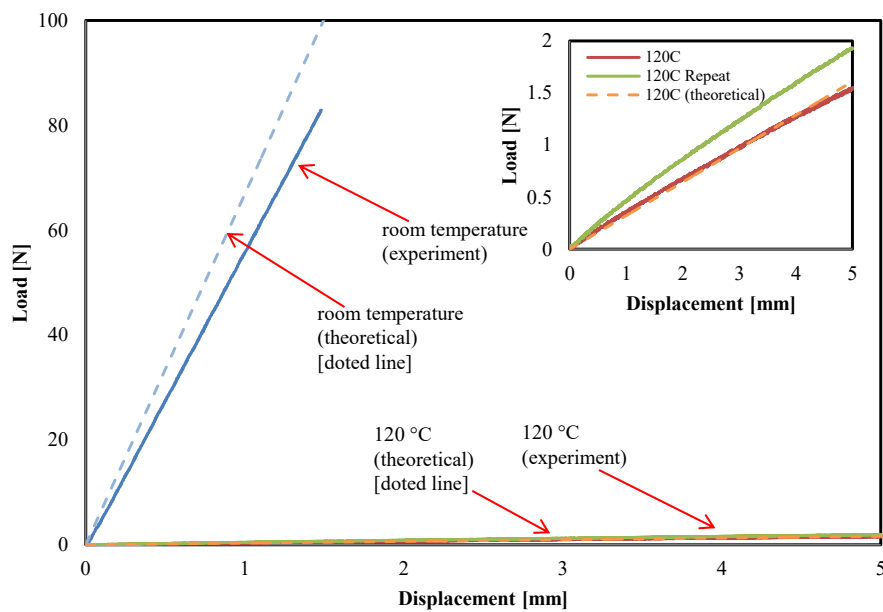
$$E_f^{HT} = \frac{E_i \sum_i^N t_i^3}{h^3} \quad \text{for composite layers only} \quad (23)$$

A.2. Test results of polystyrene-interleaved composites subjected to three-point flexure tests for viscoelastic behaviour investigation

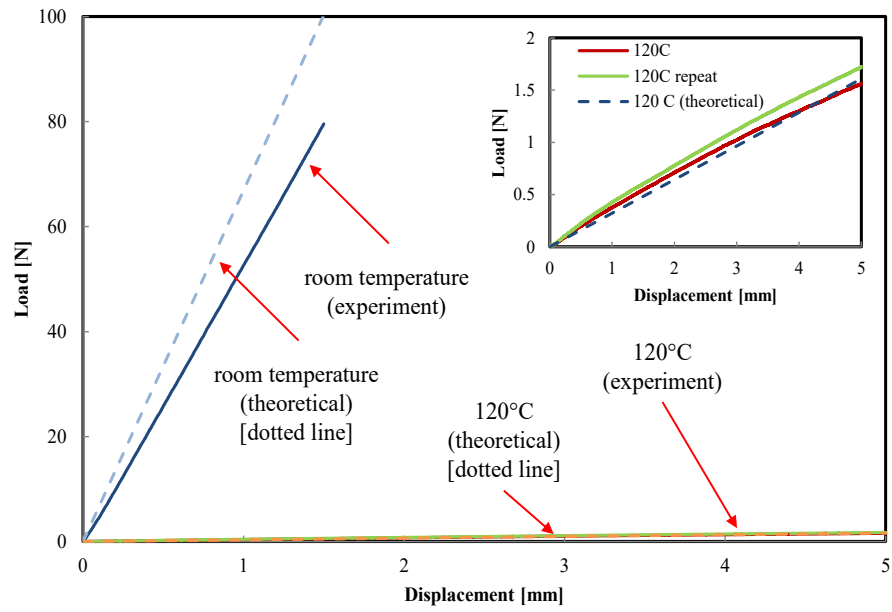
A.2.1. Stiffness variation tests at room temperature and 120°C



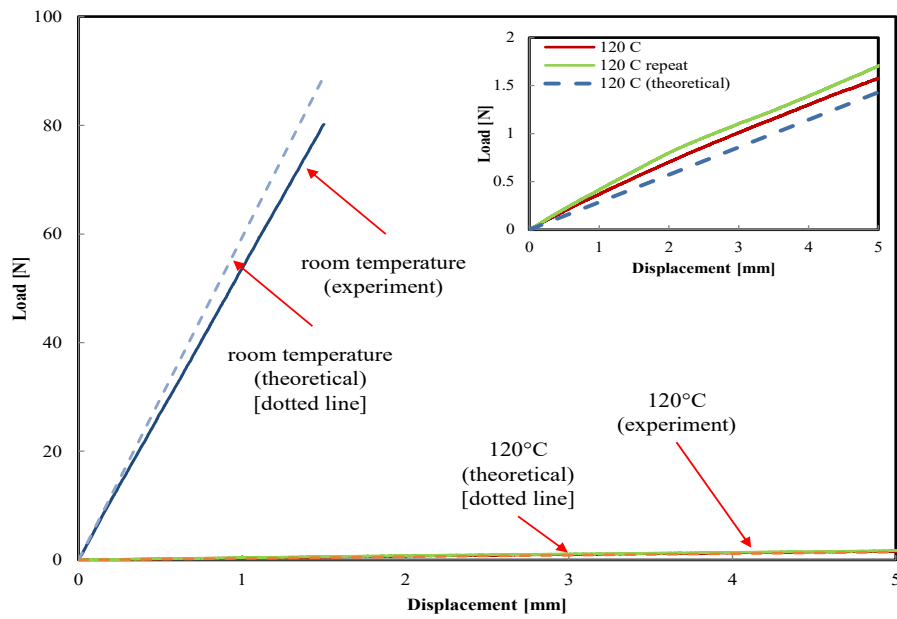
(a) UD specimen 1



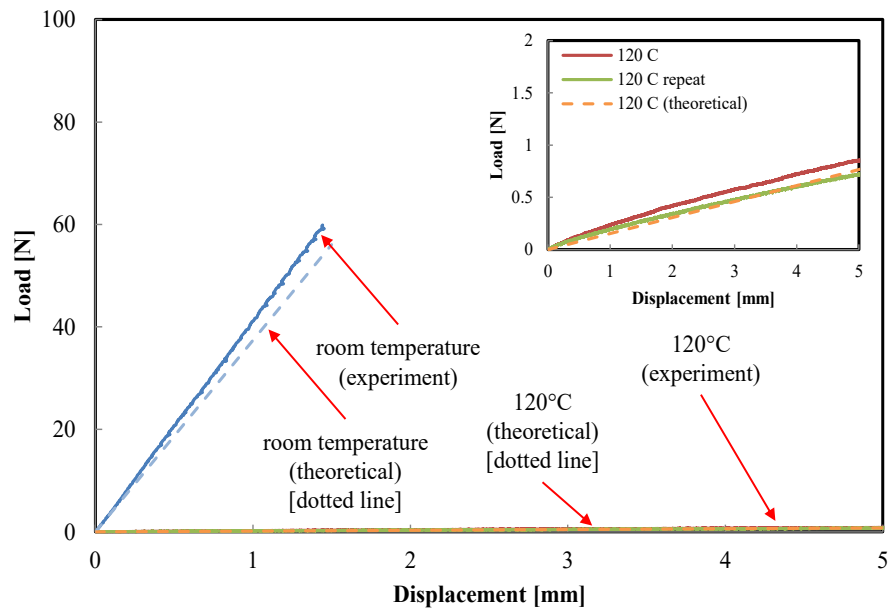
(b) UD specimen 3



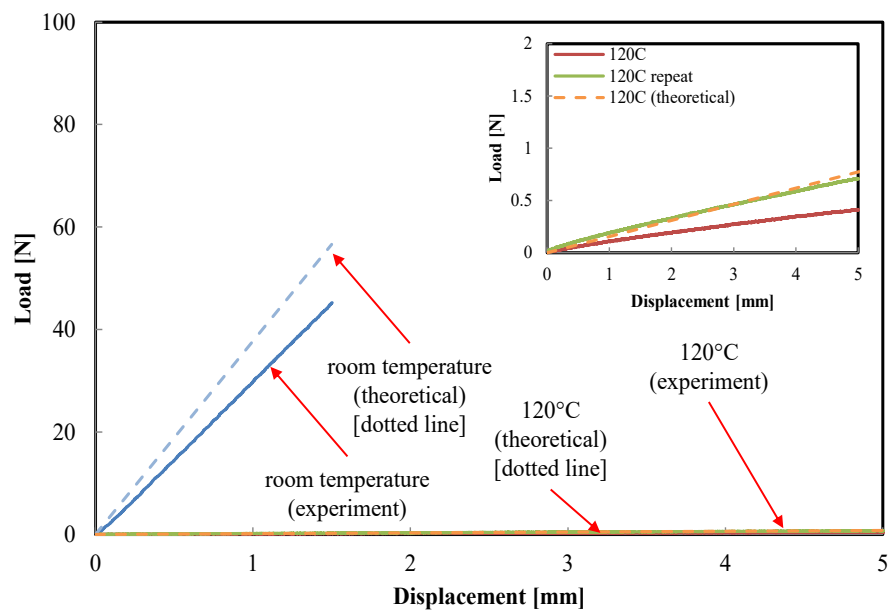
(c) UD specimen 4



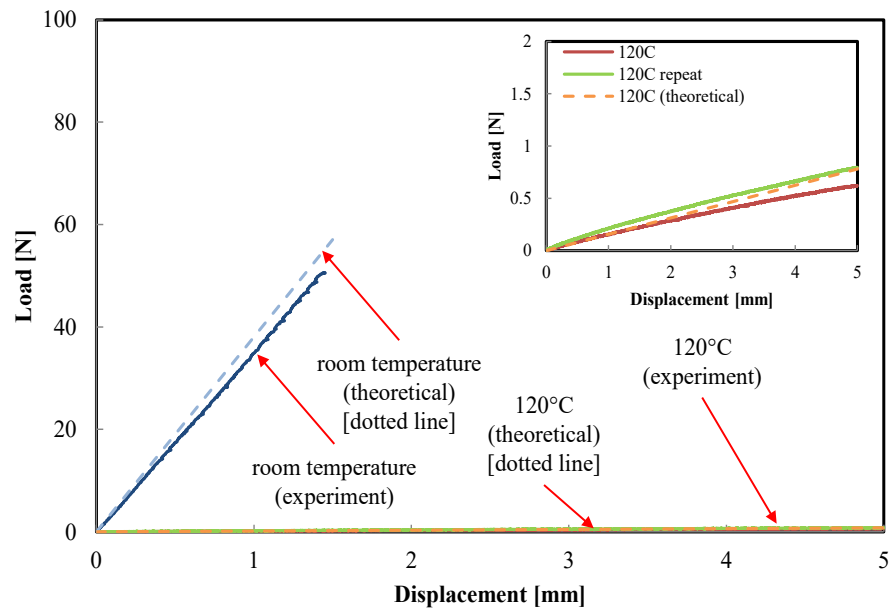
(d) UD specimen 5



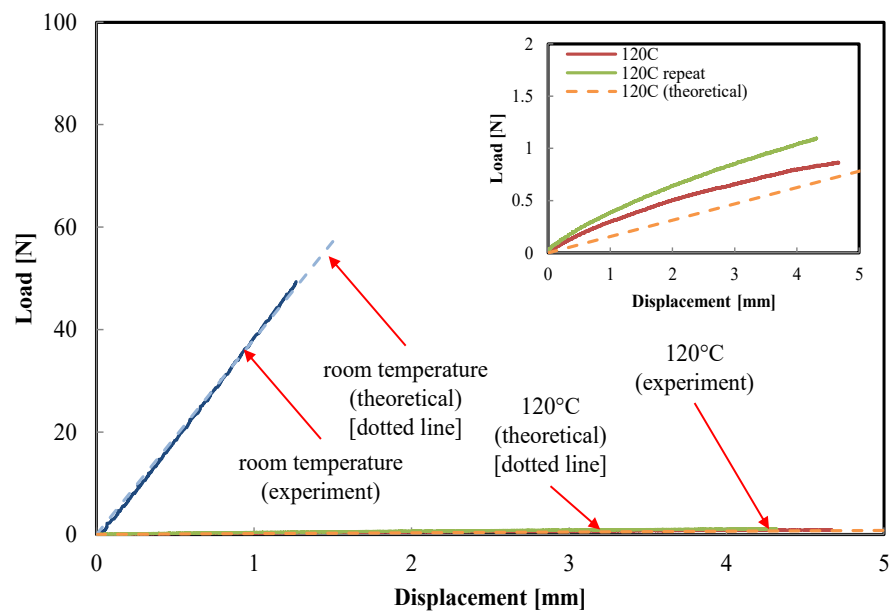
(e) [0/90] specimen 1



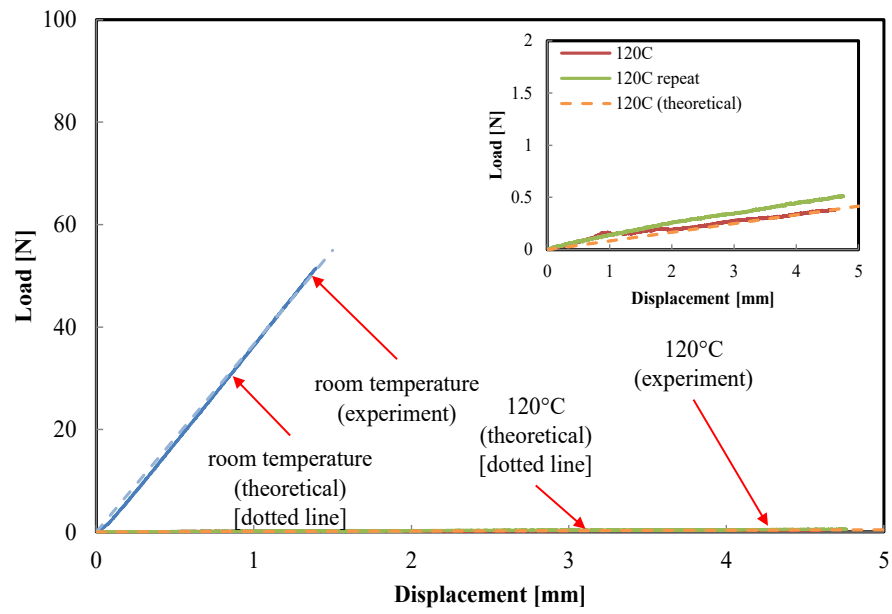
(f) [0/90] specimen 3



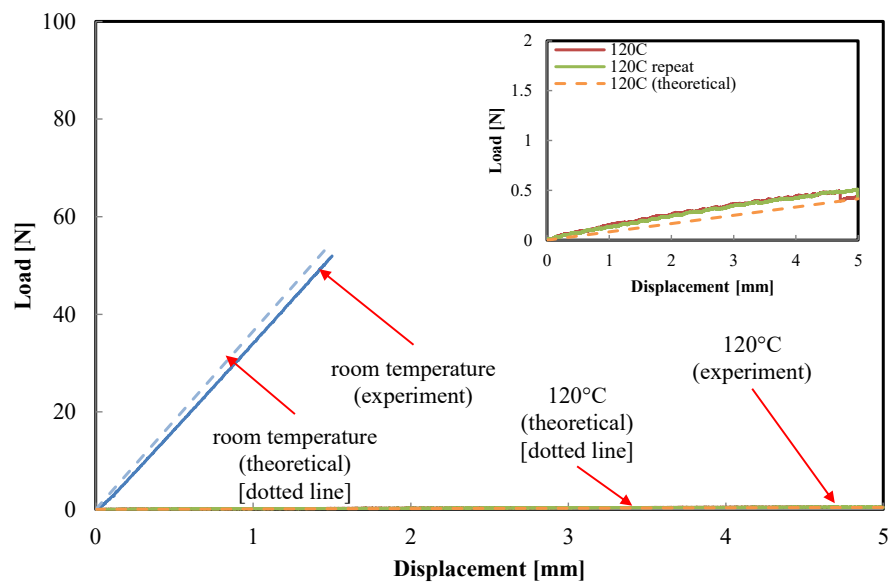
(g) [0/90] specimen 4



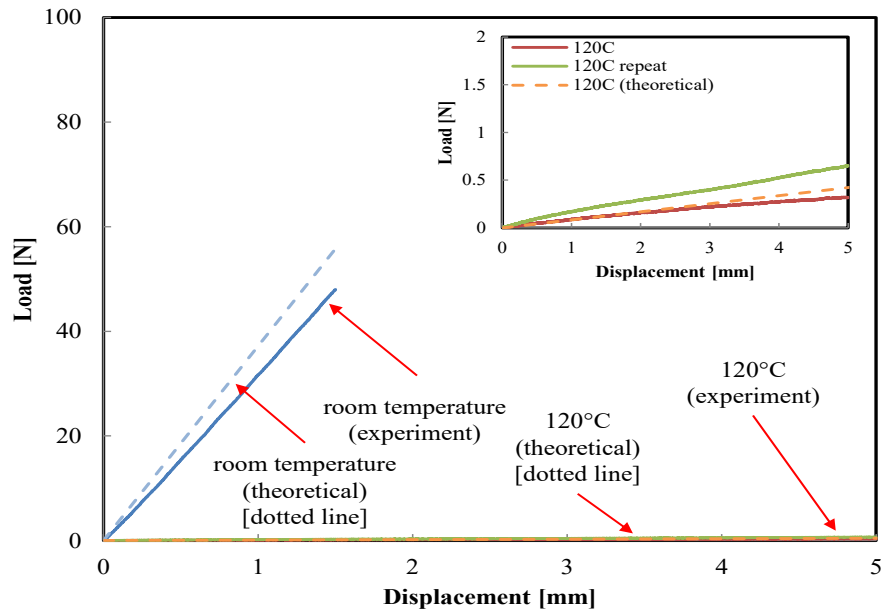
(h) [0/90] specimen 5



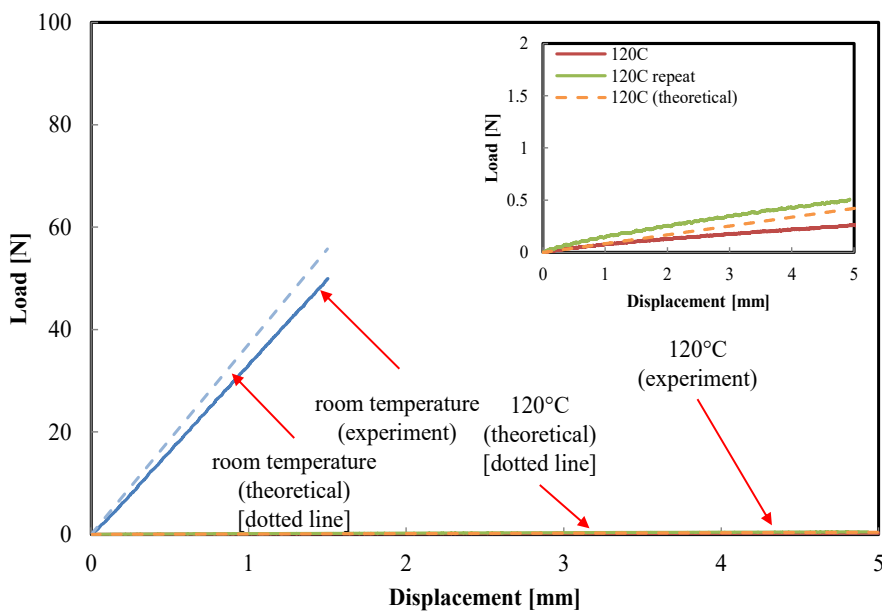
(i) QI specimen 1



(j) QI specimen 2

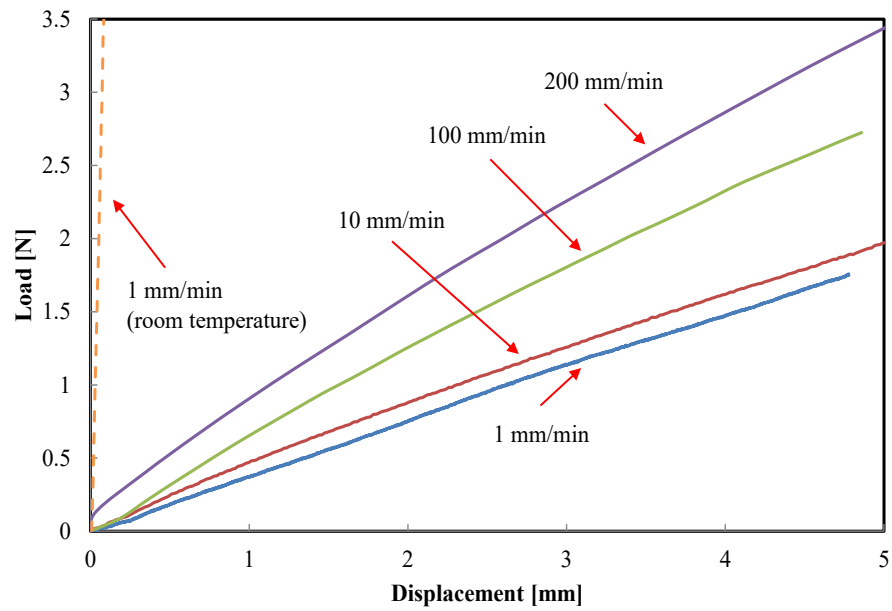


(k) QI specimen 3

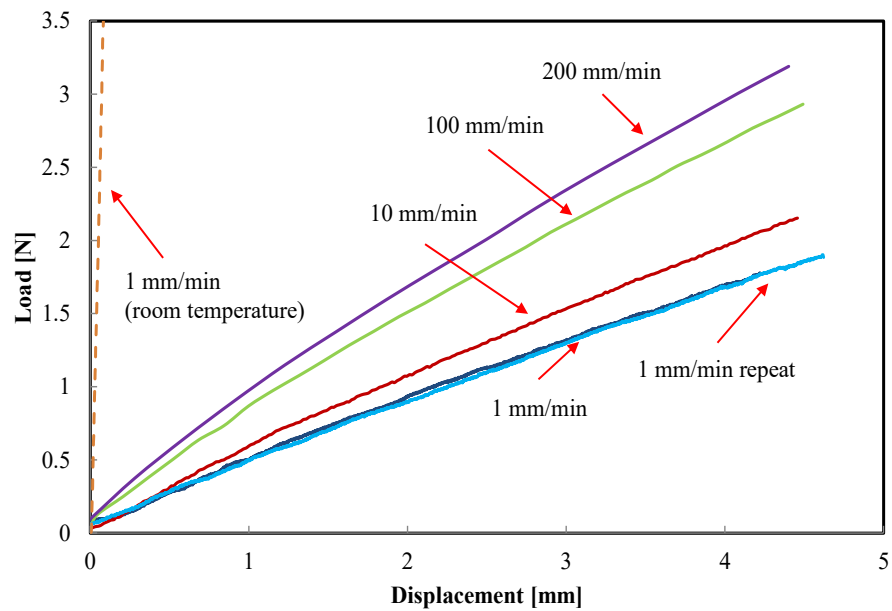


(l) QI specimen 5

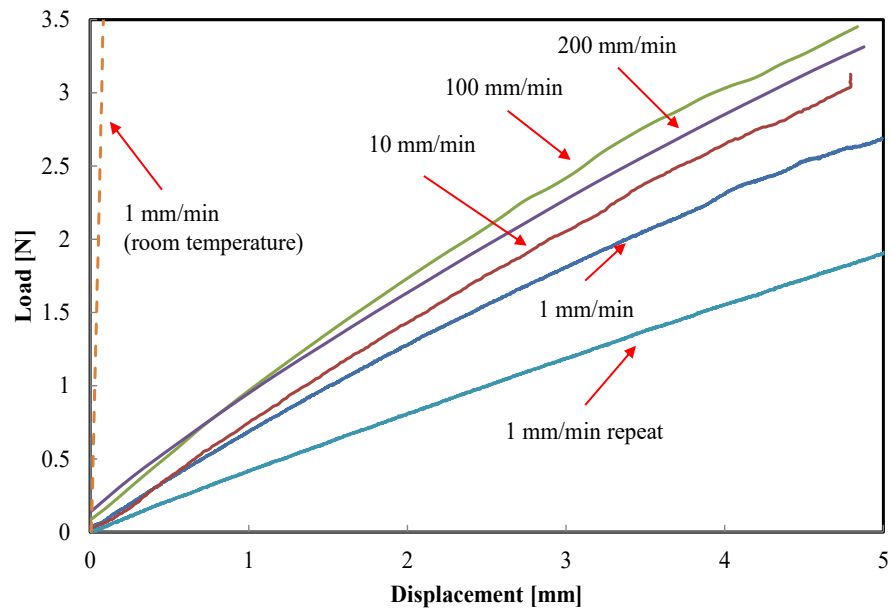
A.2.2. Rate-dependence tests



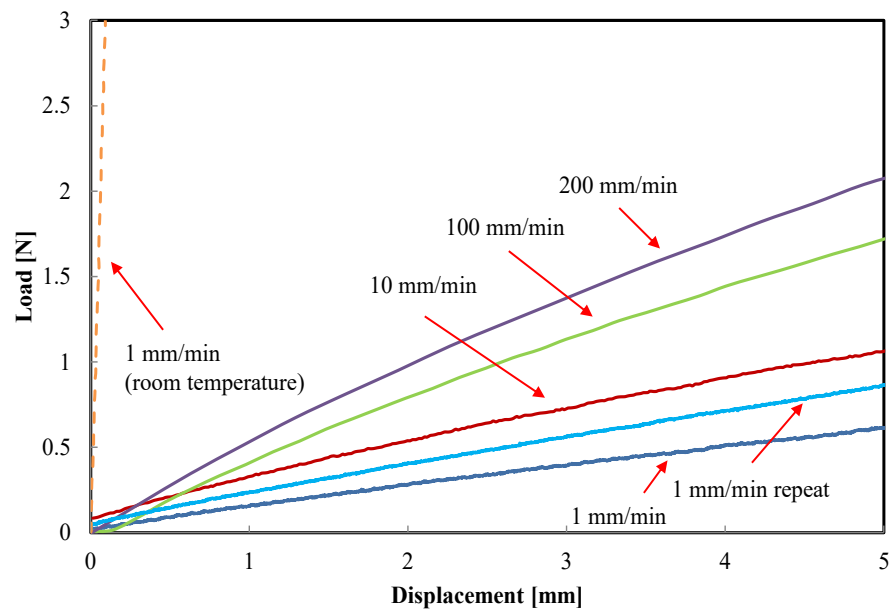
(a) UD specimen 2



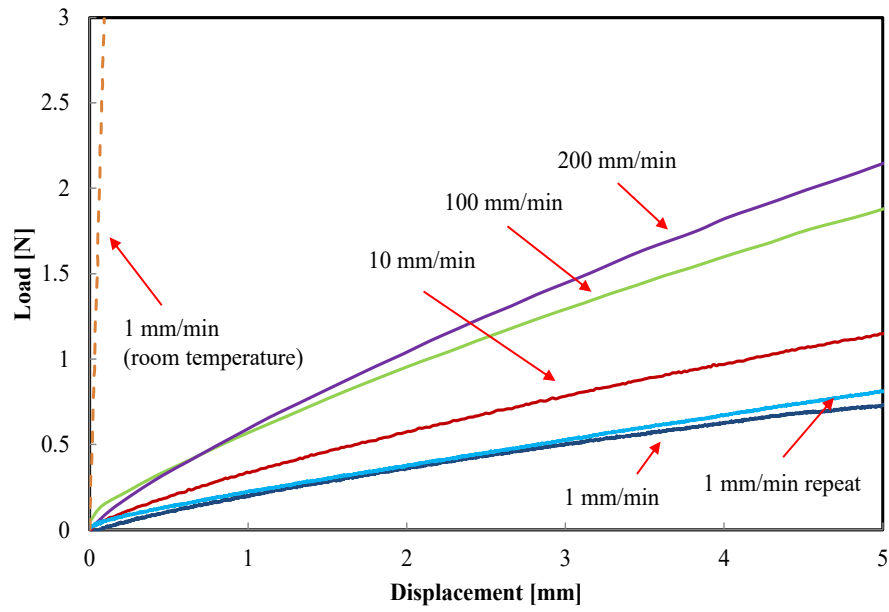
(b) UD specimen 4



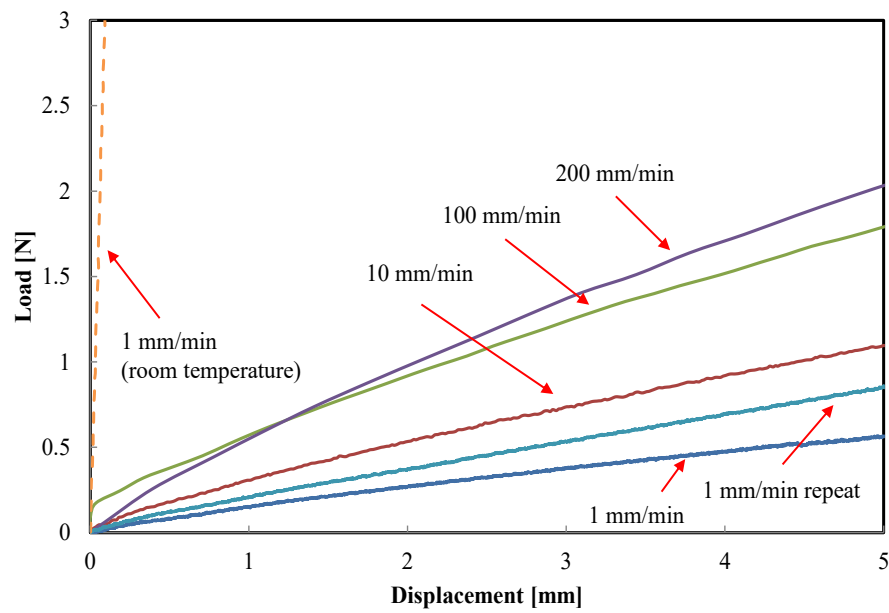
(c) UD specimen 5



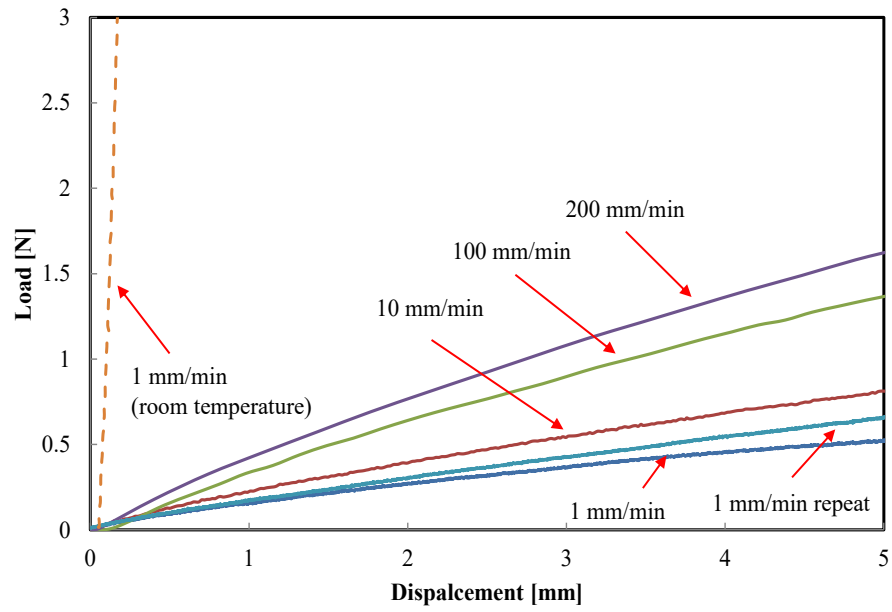
(d) [0/90] specimen 3



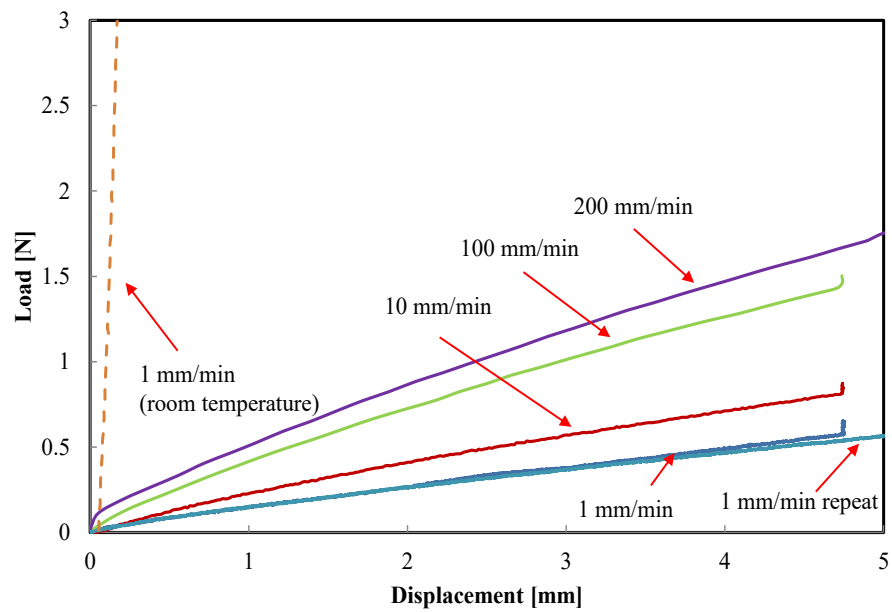
(e) [0/90] specimen 4



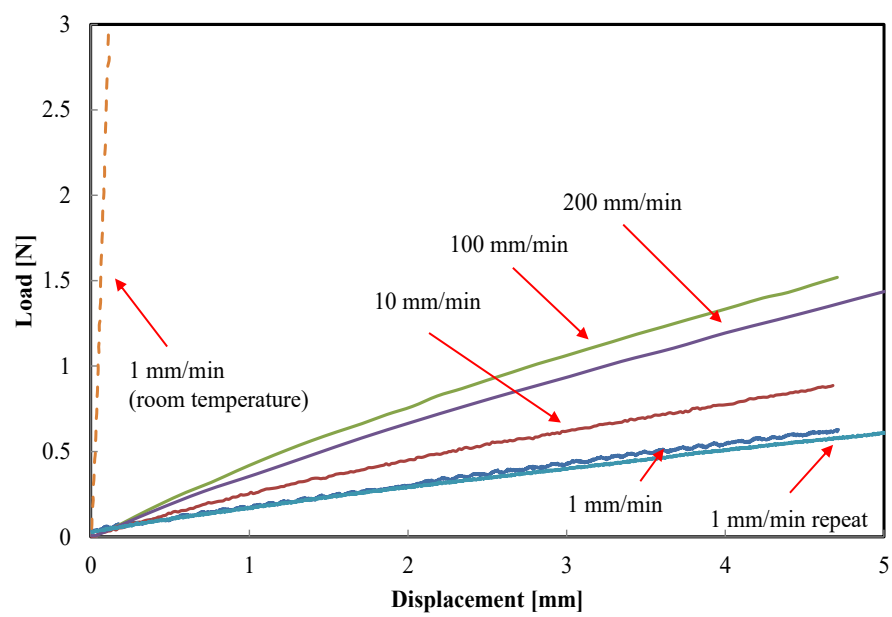
(f) [0/90] specimen 5



(g) QI specimen 1



(h) QI specimen 4



(i) QI specimen 5