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Introducing ductility in hybrid carbon fibre/self-reinforced composites through control of the damage mechanisms

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ABSTRACT

Carbon fibre composites possess excellent mechanical properties, but suffer from brittleness. Hybridisation with self-reinforced polypropylene (SRPP) is a promising strategy to introduce ductility into carbon fibre-reinforced polypropylene (CFRPP). The present work demonstrates how different damage mechanisms in these hybrid composites change as a function of the carbon fibre volume fraction, the directionality of CFRPP and SRPP and their relative layer thickness. Multiple fractures of the CFRPP layers or "fragmentation" is achieved by optimising these parameters. This leads to a ductile hybrid composite with a gradual failure development.

1. Introduction

Carbon fibre-reinforced composites combine excellent mechanical properties with a low density. This makes them a preferred choice in many lightweight structural applications. Their main drawback however, is a low tensile failure strain due to the intrinsic brittleness of the reinforcing fibre. One solution is to replace the carbon fibres by fibres with a larger failure strain, such as polymer [1–3] or metal fibres [4,5]. This solution is compromised by accompanying disadvantages, such as lower strength and increased temperature sensitivity of polymer fibres and much higher density of metal fibres. There is hence a strong need for new ideas on how to improve the failure strain of fibre-reinforced composites.

The basic question is whether a brittle material can be made ductile through intelligent design. An affirmative answer to this question can be found both in naturally occurring and in man-made materials. Biological composites, such as bone and nacre, are known for their remarkable robustness against failure and sophisticated energy absorbing mechanisms [6–8]. Haversian bone, for example, is capable of undergoing high inelastic strains because of its unique microcracking process that gradually develops in its concentric lamellae [9]. Nacre's inelastic deformation is attributed to progressive sliding and stable pull-out of its platelets [10,11]. In both cases, a well-balanced interplay between microstructural parameters and constituent properties is crucial for the activated damage mechanisms [10].

Among man-made materials, ductile behaviour was successfully achieved in engineered cementitious composites (ECC), also known as bendable concrete [12]. Unlike regular concrete that fails in a brittle manner, due to a single propagating crack initiated at a pre-existing flaw, bendable concrete undergoes excessive cracking over a large volume before it fails. The ductility originates from an accurate control of the opening of these cracks by bridging them with fibres. The design requires tailoring of the fibre size, fibre strength, interfacial strength and the size of the pre-existing flaws. With the correct set of parameters, the mechanism on the tension side of a flexural test is changed from a single crack propagation to multiple cracking. The result is a failure strain improvement by two orders of magnitude, from 0.01% for standard concrete to 5% for concrete reinforced with polyvinyl alcohol fibres [12].

While the situation in fibre-reinforced polymer composites is different from that in biological composites or fibre-reinforced concrete, certain concepts are universal and can be transferred to the other materials. Two concepts that can be transferred to brittle carbon fibre composites are that (1) a more ductile fibre should be added, and (2) a gradual damage development is achieved through multiple cracking. Fibre hybridisation is a promising approach to achieve these goals. Partial replacement of carbon fibres with a more ductile fibre provides more control over the failure mechanisms. Most hybridisation studies so far have focused on the addition of glass or aramid fibres to carbon fibre composites [13–23].

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Another vital parameter in controlling the failure mechanism is the bonding strength between the layers of the hybrid composite. When the carbon fibre layers in an unbonded carbon/glass interlayer hybrid fracture, then delaminations will develop and spread over the entire length of the sample [13]. This leads to a significant vertical load drop, after which the tensile diagram of the hybrid composite resembles that of the glass fibre composite. In the case of a strong bonding however, a more gradual transition from carbon to glass failure is achieved.

Recently, multiple cracking or fragmentation was achieved by sandwiching thin carbon fibre layers between thick glass layers [14]. For thicker carbon fibre layers, the behaviour reverted back to unstable delaminations, similar to that observed by Bunsell and Harris [13]. The fragmentation case however, allowed the carbon fibre layers to break repeatedly, resulting in a sustained stress level. This allowed their hybrid composites to reach ultimate failure strains of up to 2.8% without a drastic load drop. In the same work, an analytical equation was derived to predict the maximum layer thickness that allows this fragmentation [14]. This was later extended to a more refined numerical model in [15]. The material behaviour was referred to as pseudo-ductility, which can be defined as the occurrence of ductility in an inherently brittle material through control of the damage mechanisms. So far, it has only been achieved at low volume fractions of the brittle fibre.

The failure strain improvements that can be achieved by hybridisation with aramid or glass fibres are limited by the low failure strain of these fibres. Large improvements in the ultimate failure strain are only possible through hybridisation with a much tougher fibre [24–26]. This was achieved by hybridisation of carbon fibre with self-reinforced PP (SRPP) [24]. SRPP is a tough material with a high failure strain of about 20% [1,27]. While this ultimate failure strain was also maintained in the hybrid composites, the carbon fibre failure was accompanied by a significant load drop [24]. SRPP has also shown great potential in fibre-metal laminates, where the presence of SRPP in between aluminium plies led to a more ductile response in impact [28,29]. The combination of two ductile components in fibre-metal laminates leads to a ductile tensile behaviour without a load drop prior to final failure.

This work aims to understand the parameters governing the failure development in interlayer hybrid composites of carbon fibre and self-reinforced polypropylene. Pseudo-ductility is targeted by controlling the damage mechanisms through an intelligent choice of structural and material parameters. The final purpose is to develop a new material with reasonable stiffness, but with drastically increased ultimate failure strain.

2. Materials and methods

2.1. Materials

Propex Fabrics GmbH (Germany) provided drawn polypropylene (PP) tapes, with a stiffness of 10 GPa and a strength of 500 MPa [3]. The tapes were provided on a bobbin as well as in a twill 2/2 woven fabric with an areal density of 130 g/m^2. Propex Fabrics GmbH also provided a 50 μm thick PP film for impregnating the carbon fibre weave. This film has a melting point of 163 °C and consists of the same PP grade as the drawn PP tapes.

Two types of carbon fibre preforms were used in the study. Unidirectional carbon fibre-reinforced polypropylene (CFRPP) prepregs were sourced from Toray Carbon Fibers Europe (France). These 300 μm thick T700S prepregs have a fibre volume fraction V_f of 45% (see Section 2.4).

A balanced spread tow plain weave Textreme 80PW was sourced from Oxeon AB (Sweden). The areal density was 90 g/m^2, of which 80 g/m^2 is UTS50S carbon fibre and the rest is epoxy binder. The weave was pre-impregnated in a hot press at 220 °C using a single 50 μm PP film. The pressure was alternated between 1 and 10 bar every minute for a total of 10 min, resulting in prepregs with a thickness of 104 μm and a V_f of 43%.

2.2. Composite production

Different interlayer hybrids of SRPP and CFRPP were produced (see Table 1). S and C indicates SRPP and CFRPP layers respectively, while superscripts "*w*" and "*u*" indicate woven and unidirectional preforms, respectively. The $S_xC_yS_xC_yS_x$-layups were chosen to yield sufficiently thick samples, while still having a reasonable dispersion of the carbon fibres. This dispersion is known to be important in the performance of hybrid composites [19]. The lowest carbon fibre V_f in each hybrid configuration was achieved by grouping the carbon fibre layers together in a $S_xC_yS_x$-layup. The values of "*x*" and "*y*" in these layups were chosen to yield similar thickness and V_f for the SRPP and CFRPP layers in the different configurations.

In case of UD SRPP, the tapes were wound from the bobbin onto a rectangular frame using a winding machine. The machine translates laterally, while the frame rotates. Each translation of the machine creates one S^U layer on the top and one S^U layer on the bottom of the frame. Winding was interrupted at appropriate time intervals to insert the CFRPP prepreg layers. The other layups were made by stacking of the layers.

The hybrid layups were placed in a copper mould and inserted into a preheated press at 188 °C. The materials were hot compacted for 5 min at 45 bar pressure, followed by cooling down to 40 °C in 5 min.

The CFRPP reference composite was produced using the same process parameters, but at 5 bar pressure instead of 45 bar. This lower pressure reduces material flow out of the mould and thus limits carbon fibre undulations. The higher pressure for layups with SRPP was needed to limit the intrinsic shrinkage of PP tapes during hot compaction.

Table 1
Identification of the layups, with the measured thickness and overall carbon fibre volume fraction. The carbon fibre volume fraction in the loading direction was obtained by dividing the overall fraction by 2 in case of woven CFRPP.

CFRPP	SRPP	Layup	Thickness (mm)	Carbon fibre volume fraction	
				Overall	In loading direction (%)
UD	/	C_5^u	1.38 ± 0.02	44.9 + 1.9%	44.9
Woven	/	C_{10}^w	1.04 ± 0.02	42.5 + 1.4%	22.4
/	UD	S_{20}^u	1.57 ± 0.03	0	0
Woven	/	S_{16}^w	2.35 ± 0.01	0	0
UD	Woven	$S^wC^uS^wC^uS^w$	0.90 ± 0.04	29.3 ± 1.2%	29.3
		$S_3^wC^uS_3^wC^uS_3^w$	1.79 ± 0.03	12.7 ± 1.2%	12.7
		$S_6^wC^uS_6^wC^uS_6^w$	3.12 ± 0.02	10.1 ± 2.4%	10.1
		$S_9^wC^uS_9^w$	2.97 ± 0.04	5.0 ± 2.1%	5.0
Woven	Woven	$S^wC_2^wS^wC_2^wS^w$	0.85 ± 0.03	20.2 ± 1.3%	10.1
		$S_3^wC_2^wS_3^wC_2^wS_3^w$	1.71 ± 0.03	11.0 ± 1.2%	5.5
		$S_6^wC_2^wS_6^wC_2^wS_6^w$	3.06 ± 0.06	7.2 ± 0.5%	3.6
		$S_9^wC_2^wS_9^w$	2.90 ± 0.05	4.7 ± 1.0%	2.3
Woven	UD	$S_2^uC_2^wS_2^uC_2^wS_2^u$	0.88 ± 0.02	17.6 ± 1.8%	8.8
		$S_4^uC_2^wS_4^uC_2^wS_4^u$	1.35 ± 0.03	12.2 ± 1.7%	6.1
		$S_8^uC_2^wS_8^uC_2^wS_8^u$	2.34 ± 0.06	8.4 ± 0.2%	4.2
		$S_{12}^uC_2^wS_{12}^u$	2.21 ± 0.05	5.5 ± 0.5%	2.8