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Demonstration of pseudo-ductility in high performance glass/epoxy composites by hybridisation with thin-ply carbon prepreg

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ABSTRACT

A new approach and material architecture is presented in order to overcome the inherent brittleness and unstable failure characteristic of conventional high performance composites. The concept is the use of thin-ply hybrid laminates. Fracture mechanics calculations were carried out to determine the critical carbon layer thickness for stable pull-out in a three layer unidirectional hybrid laminate, which can provide a pseudo-ductile failure. Unidirectional hybrid composites were fabricated by sandwiching various numbers of thin carbon prepreg plies between standard thickness glass prepreg plies and tested in tension. Specimens with one and two plies of thin carbon prepreg produced pseudo-ductile failure, whereas ones with three and four plies failed with unstable delamination. An explanation of the different failure modes is given in terms of the different energy release rates for delamination in various specimens. The observed damage characteristics agreed well with the expectations according to the estimated critical carbon layer thickness.

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1. Introduction

Conventional high performance polymer matrix composites offer high strength and stiffness combined with low density. However, a fundamental limitation of current composites is their inherent brittleness. Failure is usually sudden and catastrophic, with no significant damage or warning and little residual load-carrying capacity if any. Structures that satisfy a visual inspection, can fail suddenly at loads much lower than expected, for example due to hidden delaminations after a low velocity impact with a soft body [1]. To ensure safe operation, currently a much greater safety margin is applied for composites, than for more ductile materials. For example, maximum allowable design strains can be an order of magnitude lower than the strain to failure of the fibres for carbon composites under repeated loading. These serious design limitations not only prevent engineers and operators from exploiting the performance advantages of composites, but render them unsuitable for many applications in which loading conditions are not fully predictable, and catastrophic failure cannot be tolerated. Given these limitations of currently available high performance composites, materials that fail in a pseudo-ductile manner are of exceptional interest and could potentially offer a notable increase in the scope of applications including transportation and civil engineering fields.

A basic strategy to achieve pseudo-ductility is the incorporation of new ductile matrices and fibres, which needs extensive development and validation. Another option is modification of the structure of composite laminates made of commercially available raw materials, e.g. creating hybrids, which is much faster and more straightforward. Early work on hybrid composites [2–6] showed their potential to obtain gradual failure over a range of strains by mixing different types of fibres either by intimately mingling them [7–9] or by creating a ply-by-ply hybrid structure [10–13]. The so called hybrid effect was first shown by Hayashi et al. [10,11] in their study on unidirectional (UD) layered glass/carbon hybrid composites. They showed enhanced strains to failure of carbon fibres measured in a glass/carbon hybrid composite, compared to those measured in single fibre composite specimens. The strain to failure of the carbon fibres in a UD layered hybrid composite plate can be seen on typical tensile stress–strain graphs (e.g. Fig. 1) as a first peak followed by a sudden drop in stress, which corresponds to the unstable delamination of the failed plies due to excess strain energy. The figure shows the average graph for six specimens under load control, based on the original paper by Hayashi [11]. In displacement control, the drop in stress normally appears as a straight vertical line however, the desired behaviour would be more gradual, with no severe and sudden loss of stiffness, and failure initiation at much higher strains. Bunsell and Harris [12] also reported higher strain to carbon failure in their UD glass/carbon hybrids, than that of all carbon specimens. They

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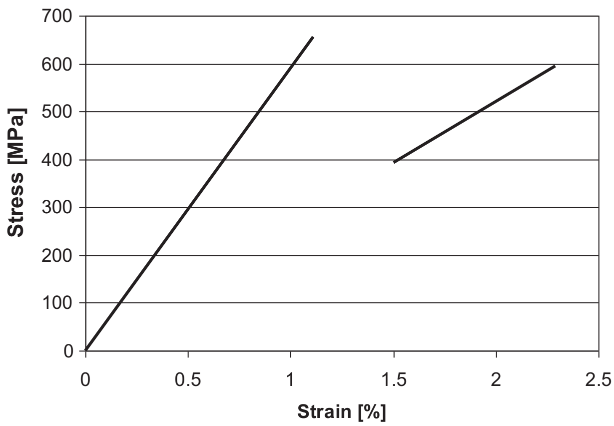

Fig. 1. Average experimental stress–strain graph of a glass/carbon hybrid composite (based on [11]).

observed gradual failure and multiple cracking of the carbon plies without unstable delamination in their hybrid composites incorporating 0.4–0.8 mm thick carbon layers well bonded to glass plies. It is close to a pseudo-ductile behaviour, but multiple carbon fracture started at strains as low as 0.28–0.53%. The reason for this gradual failure type in a standard carbon thickness hybrid was the low strain energy in the carbon plies at failure due to the very low carbon fibre failure strains, which was insufficient to drive delamination. Modern carbon fibres have strains to failure of over 1.5% and given that energy is proportional to strain squared, would not be expected to fail in the gradual manner observed in [12]. Manders and Bader [13] also observed consistently higher strains to carbon failure in their UD glass–carbon–glass sandwich laminates than in all carbon specimens. The effect of carbon layer thickness was also investigated, and it was reported that the thinner the central carbon layer, or the lower the carbon to glass ratio, the lower the extent of delaminations between the layers reinforced with different fibres. Summarising the published work it is clear that there is potential in hybrid composites to obtain a more gradual failure, and higher strains to final failure than in single fibre type composites.

More recently some authors [14–17] have reported superior mechanical properties obtained by applying thin carbon plies in composites. The authors agreed that thin plies can shift the onset of damage in quasi-isotropic composites under numerous loading conditions towards higher strains by suppressing delamination and matrix cracking, thereby potentially allowing higher design strains in critical composite structures. The reason for this effect is, that thinner plies have lower energy release rates, delaying the propagation of intra- and interlaminar cracks.

The present study focuses on thin-ply glass/carbon hybrid laminates designed to combine the benefits of both hybrid and thin-ply approaches by exploiting the full pseudo-ductility potential of these material structures. The authors also attempt to explain the key factors controlling the failure type of laminated hybrid composites. The basic concept of using a hybrid composite can provide a high initial modulus, and a residual load bearing capacity after the failure of the high modulus component. If the high modulus, lower strain to failure component can be introduced in the hybrid material in a thin-ply form, multiple fractures of the stiffer component can take place, and the delaminations between the components will be localised and made stable due to the low energy release rate of the thin plies. Therefore it should be possible to achieve a notable amount of pseudo-ductile strain during and after the failure of the stiffer component, without a large drop in stress.

The most important factor which governs the failure behaviour of ply-by-ply hybrid laminates is the mode II energy release rate, which controls delamination after the failure of the lower strain plies. For a given material combination, the energy release rate depends on the thickness of the low strain to failure layer. Using thin plies of carbon between glass plies is a good combination for several reasons:

(1) There is a significant difference in strains to failure and elastic modulus of the two different components, which allows a considerable change in properties due to hybridization.
(2) Both fibres are available in similar prepreg forms, with compatible resin systems, which enables hybrid laminate manufacturing.
(3) Carbon fibres are available in the form of thin prepregs, which are crucial for the experiments.
(4) Brittle carbon fibres are protected by more damage tolerant glass plies from any mechanical damage through manufacturing handling and testing.
(5) The translucent character of glass/epoxy composites makes delamination detection possible.

The combination of glass and carbon fibres is a suitable model system, but the concept is valid for other materials such as different grades of carbon fibres.

2. Fracture mechanics background

The present section aims at giving an overview of the possible failure types and overall damage characteristics of three layer UD glass/carbon epoxy composites under static uniaxial tensile loading. A detailed calculation is also included to estimate the critical carbon ply thickness for stable pull-out and pseudo-ductile overall failure.

2.1. Failure modes

Possible failure modes in a three layer UD hybrid laminate are highlighted in Fig. 2. In the case of standard thickness or blocked ply laminated hybrid composites, a too high ratio of carbon to glass may lead to a single crack through the whole thickness, resulting in sudden, brittle failure (Fig. 2a). The most common failure mode of ply-by-ply hybrid laminates as reported several times in the literature and in the authors' previous work [18] is shown schematically in Fig. 2b. This conventional failure type is a single fracture in the low strain layer instantaneously followed by unstable delamination, which appears on the stress–strain graphs as a significant load drop. Fig. 2c shows the desired thin-ply hybrid laminate behaviour, where delamination is suppressed, and multiple

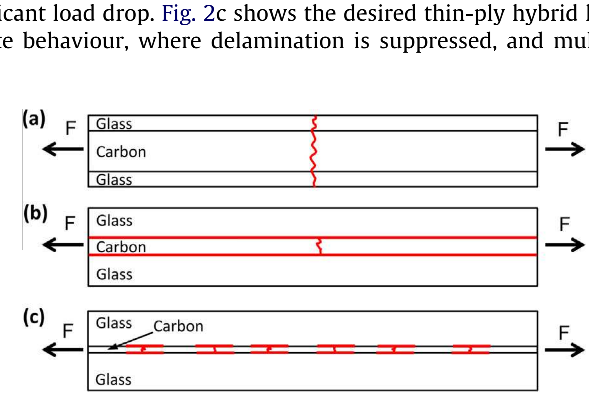

Fig. 2. Possible failure modes in a three layer UD hybrid glass/carbon laminate (red lines show fracture) (a) single crack through the whole specimen thickness (improperly sized hybrid laminate), (b) single crack in the carbon layer followed by instantaneous delamination (conventional standard ply thickness laminate), and (c) multiple fracture and localised stable pull-out of the thin carbon layer (thin-ply hybrid laminate). (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

carbon layer fractures are obtained followed by stable localised pull-out. In conventional hybrid laminates, the stress drops significantly when the low strain fibres fail and then comes a stress recovery as the high strain fibres pick up all the load. In theory a stable damage process can be achieved in a glass/carbon hybrid laminate by using thin carbon plies. Instead of a sudden drop in stress, a stable transition can be achieved between the stress peaks of the intact hybrid laminate and the delaminated layers, where only the glass plies carry load as shown in Fig. 3.

2.2. Criteria for stable pull-out

In this section preliminary fracture mechanics calculations are shown. The aim is to set up and check the criteria for stable failure of the investigated three layer hybrid laminate system. An approximate range of dimensions (especially carbon ply thicknesses) and material properties (such as required fracture toughness) which are likely to allow the carbon plies to fail in a stable manner within a glass/carbon UD ply-by-ply hybrid laminate system under tension is also given. Assuming a hybrid laminate, higher strains than the strain to failure of the more brittle component used can only be achieved without sudden loss of stiffness if these plies are pulled-out stably after they fracture, while the rest of the laminate is able to withstand the load applied. To enable the design of such a structure, two criteria should be met:

(1) When one layer fails, it must not pull out unstably. This is a function of the elastic strain energy within the ply at failure compared with the fracture energy of the interfaces on either side, and imposes a maximum layer thickness for a given delamination toughness (G_{IIC}).
(2) Failure of one layer must not lead instantly to failure of an adjacent layer due to the additional stress transferred from the failed layer.

2.2.1. Criterion for failed layer not to pull out unstably

For a laminate to fail gradually, failure of a single layer must not lead straight away to the unstable pull out of that layer. The strain energy release rate for pull out of the layer at the interfaces on either side must be less than the delamination fracture energy. Consider a three layer laminate of thickness h, subject to an overall average stress σ, made of material with modulus E_1, with an already fractured but not delaminated thin central layer of thickness t_2 and modulus E_2 embedded (see Fig. 4 for longitudinal section). The initial equivalent modulus E_{eq0} of the hybrid laminate before central layer delamination can be written as in Eq. (1), simply taking the relevant material and geometrical properties into account.

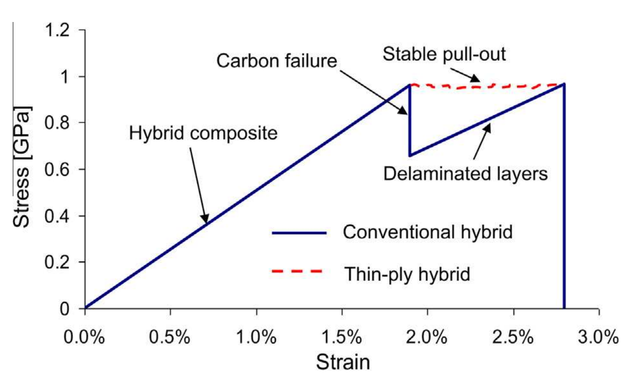

Fig. 3. Schematic of the stress–strain graph of a conventional and a thin-ply hybrid composite laminate. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

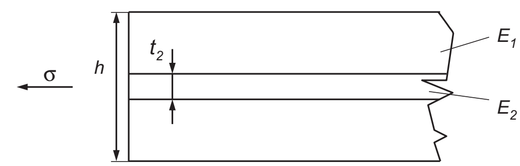

Fig. 4. Notation used in the three layer hybrid laminate calculations.

$$E_{eq0} = \frac{E_1(h - t_2) + E_2 t_2}{h} \tag{1}$$

where E_1 is the modulus of elasticity of the higher strain layer, E_2 is the modulus of elasticity of the lower strain layer, h is the overall thickness of the laminate, t_2 is the thickness of the thin central layer according to Fig. 4. The final equivalent modulus of elasticity after central layer delamination $E_{eq,f}$ is as in Eq. (2), because the broken and pulled out central ply does not contribute any more to carrying load.

$$E_{eq,f} = \frac{E_1(h - t_2)}{h} \tag{2}$$

Using the equivalent moduli of the laminate, the elastic strain energy can be written before and after pull-out of the thin central layer using the formulae in Eq. (3) which is written assuming unit length and width of the laminate.

$$U = \frac{1}{2}\sigma\varepsilon h = \frac{1}{2}\frac{\sigma^2 h}{E_{eq}} \tag{3}$$

where U is the elastic strain energy, σ is the overall average stress, ε is the overall strain and E_{eq} is the equivalent modulus of elasticity. The difference between the energy values before and after central layer pull-out assuming constant applied stress σ gives the energy release rate of G for pull out of the embedded layer. Equivalent moduli values before and after delamination (E_{eq0} and $E_{eq,f}$ respectively) are substituted in the formula of Eq. (3) yielding Eq. (4). The additional factor of two (rearranged to the right side of the equation) is present because delamination takes place on each side of the central layer and therefore the calculated energy has been used for forming two new surfaces.

$$G = \frac{\sigma^2 h^2}{4(E_1(h - t_2) + E_2 t_2)} - \frac{\sigma^2 h^2}{4E_1(h - t_2)} = \frac{\sigma^2 h^2 E_2 t_2}{4E_1(h - t_2)(E_1(h - t_2) + E_2 t_2)} \tag{4}$$

As the strain is assumed to be equal through the thickness of the laminate, the stress in the central layer σ_2 can be written in terms of the overall stress σ:

$$\sigma_2 = \frac{\sigma h E_2}{E_1(h - t_2) + E_2 t_2} \tag{5}$$

The energy release rate in terms of the central layer stress σ_2 can be written:

$$G = \frac{\sigma_2^2 t_2 (E_1(h - t_2) + E_2 t_2)}{4E_1 E_2 (h - t_2)} \tag{6}$$

It can also be written in terms of the overall strain ε:

$$G = \frac{\varepsilon^2 E_2 t_2 (E_1(h - t_2) + E_2 t_2)}{4E_1(h - t_2)} \tag{7}$$

Assuming a model laminate with critical central layer thickness t_{2c}, and central layer stress equal to the tensile strength σ_{2b}, G can be equated to G_{IIC} during delamination propagation. For a given fracture toughness, material properties (E_1, E_2, tensile strength

σ_{2b}) and overall thickness h, the maximum allowable thickness of the central layer t_{2c}, to maintain stable pull out can be calculated using Eq. (6).

Considering a glass/carbon hybrid laminate as a model system with typical material properties (E_1 = 44 GPa, E_2 = 135 GPa, σ_{2b} = 2700 MPa), mode II fracture toughness G_{IIC} = 1 N/mm and h = 1 mm overall thickness, the given critical central layer thickness is t_{2c} = 62 μm. This value implies that conventional carbon prepreg systems are too thick to achieve stable pull out, but spread tow plies or special thin prepregs may be suitable.

2.2.2. Criterion for layer failure not leading straight to overall failure

Assume that when the central layer fails all its load is immediately shed to the outer layers and that the interface will deflect cracks and prevent stress concentrations arising. Using the same terms as before, the required minimum strength $\sigma_{1b\ min}$ for the higher strain component for it not to fail immediately after the embedded lower strain layer with strength σ_{2b} fractured, is determined by the following equation:

$$\sigma_{1b\ min} = \frac{\sigma_{2b}(E_1 t_1 + E_2 t_2)}{E_2 t_1} \tag{8}$$

where t_1 is the full thickness of the two high strain layers. Assuming the same geometry and typical material properties as before, and t_2 = 62 μm equal to the critical central layer thickness, the minimum glass layer strength $\sigma_{1b\ min}$ = 1058 MPa, which implies that conventional glass fibre plies should be suitable provided their tensile strength is well above this level.

3. Experimental

The present section shows the materials, lay-up sequences, manufacturing, fabrication and test procedures applied and finally the results of thin prepreg characterisation and mechanical tests on hybrid laminates.

3.1. Materials

To meet the thickness criterion of the central layer as determined by the performed calculations, special thin carbon prepreg was used. Conventional standard thickness glass fibre prepreg was chosen for the embedding material purposes, because it fulfils the strength criterion and its strain to failure is notably higher than that of carbon plies. As an additional benefit the glass plies are translucent, allowing crack and delamination detection and tracking. Three layer hybrid specimens were manufactured using conventional E-glass reinforced epoxy matrix UD prepreg (HexPly 913G-E-5-30% supplied by Hexcel) with 0.125 mm nominal cured thickness, 192 g/m^2 glass fibre mass per unit area and 30% mass (~40% volume) cured resin content. The actual cured ply thickness of the glass/epoxy prepreg was found to be around 0.14 mm. As the stiffer component, special thin carbon prepreg commercially available from SK Chemicals (South Korea) under the trade name of SkyFlex USN020A was used. The fibres in the thin prepreg were TR 30 type carbon fibres (E = 234 GPa, strain to failure = 1.9%) supplied by Mitsubishi Rayon Co. Ltd. The corresponding matrix was SK Chemical's type K 50 epoxy resin. Both resin systems in the hybrid laminates were 120 °C cure epoxies, which were found to be compatible, although no details were provided by the suppliers on the chemical formulation of the resins. Good integrity of the ply-by-ply hybrid laminates was confirmed during testing procedures and no phase separation was observed on cross sectional micrographs. The nominal mass per unit area of the thin carbon prepreg and just the fibres are 37 g/m^2, and 22 g/m^2 respectively. The actual basic properties of the batch used in this study were measured. The resin was burnt off in an atmospheric furnace (60 min@500 °C) from ten 100 mm square prepreg samples taken from a strip across the width direction of the prepreg roll. Results of the prepreg characterisation can be found in Table 1. It can be stated, that the uncured prepreg mass per unit area is slightly higher than the factory data provided and the fibre mass fraction is low compared to standard thickness prepregs. The calculated fibre volume fraction based on the measured mass fraction of the thin carbon prepreg is v_f = 40.5%. Optical microscopy was also executed on specific cured thin prepreg hybrid laminates, and central layer thicknesses of 1, 2 and 4 plies carbon layers were 21.9 ± 5.0 μm, 48.8 ± 7.8 and 100.0 ± 11.7 μm respectively, each measured at more than 10 points along the polished cross sectional samples. This implies that the cured thickness of a single carbon ply is around 25 μm. Variation in the average thicknesses of different type cured test laminates (Tables 3 and 5) indicate carbon ply thicknesses of up to 29.3 μm, and overall thicknesses of cured 16 ply UD plates used for the carbon prepreg characterisation yield an average of 30.2 μm for the ply thickness. Discrepancies between ply thicknesses measured through microscopy and by using standard tools such as digital callipers can be the result of difficulties in microscopy to judge where the carbon/epoxy ply ends and the glass/epoxy starts on the hybrid laminate cross-sections. This usually leads to a systematic underestimation of the central layer thickness as it is easier to measure the region where there are carbon fibres and exclude the usually resin reach region at the interface between carbon and glass fibre reinforced plies. In the following sections an average value of 29 μm will be used for the cured carbon ply thickness. Mechanical properties of pure carbon/epoxy and glass/epoxy composite plates of the materials used within this study can be found in Table 2.

3.2. Lay-up sequences and manufacturing of composite specimens

Unidirectional laminates were laid up and cured using the thin carbon and the standard glass prepregs in the following sequences: $[g_2/c_n/g_2]$, where g stands for glass plies and c for carbon plies, values of n were 1, 2, 3, and 4. The manufacturing of the hybrid composite laminates was similar to the conventional process for standard prepregs. Laminates were cured in an autoclave at the recommended cure temperature and pressure cycle for the Hexcel 913 resin (60 min@125 °C, 0.7 MPa), as it was identical to the instructions given for the thin carbon prepreg. A flat aluminium tool plate and caul plates were used during the bagging and curing process. The possible effect of thermal residual stresses in the hybrid laminates was analysed for the worst case scenario of $[g_2/c_1/g_2]$ lay-up sequence. The compressive residual strain in the carbon layer was found to be 5.57×10^{-4} which is less than 3% of the strain to failure of the carbon fibres therefore it does not affect the failure process of the hybrid laminates significantly. Fabrication of the specimens was done using a diamond cutting wheel. End tabs with 1.5 mm thickness, made of glass fibre reinforced cross-ply plates, were bonded to the specimens using Redux® 810 epoxy adhesive supplied by Hexcel, cured for 60 min@70 °C.

3.3. Test procedure

Mechanical testing of the glass/carbon hybrid composite specimens was executed under uniaxial tensile loading and displacement control using a crosshead speed of 2 mm/min on a computer controlled Instron MJ6283 type 100 kN rated universal hydraulic test machine with wedge type hydraulic grips. Nominal specimen dimensions were 260/160/20/0.6 mm overall length/free length/width/thickness respectively. At least five specimens were

Table 1
Basic properties of SkyFlex USN020A thin carbon fibre prepreg.

Property	Uncured prepreg mass per unit area (g/m^2)	Fibre mass fraction (%)	Fibre mass per unit area (g/m^2)
Average	42.6	49.4	21.16
Coeff. of variation (%)	2.50	1.87	3.97

Table 2
Properties of carbon/epoxy and glass/epoxy plates of the materials used in present study.

Spec. type	Property	Initial elastic modulus (GPa)	Strain to failure (%)	Strength (MPa)	Cured ply thickness (mm)
Pure carbon composite	Average	101.7[a]	1.5[a]	1503[a]	0.029
	Coeff. of variation (%)	2.75	6.76	7.51	–
Pure glass composite	Average	40[b]	3.4[c]	–	0.14

[a] Measured on 16 ply UD laminates, specimens failed explosively at the end tabs.
[b] Based on Hexcel data (corrected in terms of cured ply thickness).
[c] From reference [19], measured on special tapered specimens to avoid end tab failure.

Table 3
Geometric properties of "1 and 2 ply carbon" type hybrid specimens.

Spec. type	Property	Width (mm)	Thickness (mm)	Free length (mm)
1 Ply carbon	Average	19.82	0.597	158.1
	Coeff. of variation (%)	0.27	1.19	0.3
2 Ply carbon	Average	19.80	0.637	157.2
	Coeff. of variation (%)	0.32	0.71	0.3

Table 4
Tensile test results of "1 and 2 ply carbon" type hybrid specimens.

Spec. type	Property	"Yield" strain (1) (%)	"Yield" stress (1) (MPa)	Strain at max. stress (%)	Maximum stress (MPa)	Final strain (2) (%)	Elastic modulus (GPa)
1 Ply carbon	Average	2.20	967.2	2.62	1047.8	2.77	44.32
	Coeff. of variation (%)	0.78	2.1	4.26	3.2	2.73	1.69
2 Ply carbon	Average	2.06	954.3	2.24	955.9	2.50	46.59
	Coeff. of variation (%)	1.02	1.8	3.73	4.1	6.39	2.49

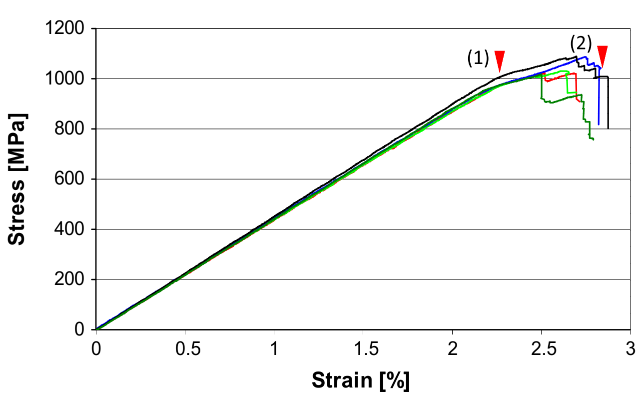

Fig. 5. Results of tensile tests on type "1 ply carbon" hybrid specimens. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

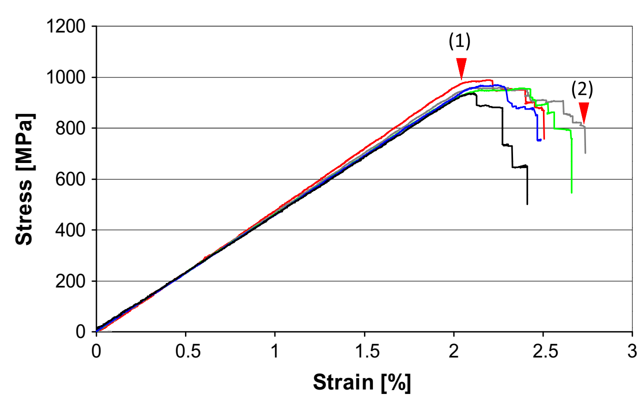

Fig. 6. Results of tensile tests on type "2 ply carbon" hybrid specimens. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

tested from each type. Strains were measured using an Imetrum video gauge system with a Sony XCD-SX910 type CCD camera at a nominal gauge length of 100 mm by tracking the speckle pattern applied on the specimen face using spray paints. Videos of the back face of the specimens were also recorded to be used for failure type and process characterisation.

3.4. Results and discussion

Specimen types were marked according to the number of carbon plies in the central layer of the laminate. Figs. 5–8 show the overall tensile stress–strain graphs obtained from the tests, based on the average stress calculated using the measured specimen

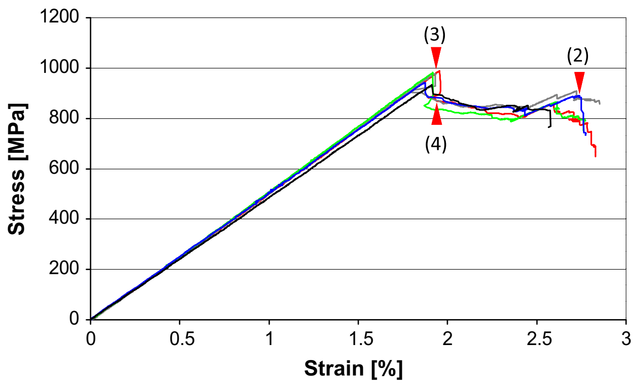

Fig. 7. Results of tensile tests on type "3 ply carbon" hybrid specimens. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

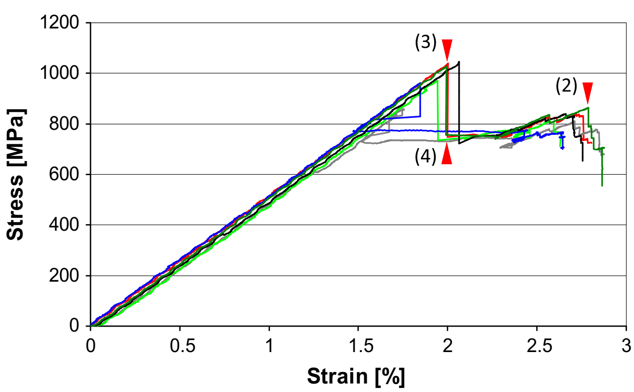

Fig. 8. Results of tensile tests on type "4 ply carbon" hybrid specimens. (Please note that instabilities on graphs are due to the optical strain measurement system.) (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

dimensions. Arrows and numbers in brackets refer to typical characteristic points on the graphs, used for data representation. The numbers in brackets will also appear in the tables summarising the test results, later in this section, to make it easier to match values with the corresponding characteristic points of the failure procedure. Point (1), where the graph deviates from linear, was determined for each specimen of types "1 and 2 ply carbon", using the intersection point of two straight lines laid on adjacent sections of the graphs. Final strain values (2) were determined at a 15% drop in stress towards the end of the failure process. This criterion was necessary in order to enable the comparison of test data of different specimen types showing various failure modes. 15% drop in stress was chosen, to exclude minor drops which are acceptable in a pseudo ductile failure procedure, but take account of more significant ones.

3.4.1. Discussion of failure types

Analysing the curves, three groups of behaviours can be distinguished: "1 and 2 ply carbon" specimens fail in the desired pseudo-ductile manner, "3 ply carbon" specimens show a small stress drop indicating an intermediate behaviour and "4 ply carbon" specimens show conventional ply-by-ply hybrid type failure with unstable delamination.

Tables 3 and 4 show the geometric properties and tensile test results of the "1 and 2 ply carbon" specimens. Please note that the term "yield" in Table 4 refers to pseudo-yielding behaviour, as none of the constituents of the hybrid composites tested show plastic deformation. "1 and 2 ply carbon" type specimens showed favourable pseudo-ductile failure types. In the case of "1 ply carbon" specimens the carbon ply fragmented progressively along the entire gauge length after the strain reached the strain to failure of the carbon fibres. This process is visible on the stress–strain graphs of the specimens as a significant change in slope. Fig. 9 shows the carbon ply cracks in the central layer of an interrupted test specimen. The cracks are visible because of the translucent nature of the glass/epoxy composite on the outside of the hybrid laminate. In the case of the "2 ply carbon" specimens, multiple cracks appeared in the carbon ply around 2% strain, in a distributed manner along the gauge length forming a striped pattern. Fig. 10 shows a specimen after an interrupted test which has localised delaminations around the carbon ply cracks. Well bonded area appears to be black, because of the translucency of the glass plies. The locally delaminated areas just around the cracks in the carbon layer are yellow, like the resin in the glass prepreg, because the rough delaminated back surface of the glass layer blocks the visibility of the carbon. The localised delaminations developed stably during further loading in parallel until almost linking up. The observed pseudo-yielding provided a stable failure process until the final failure which happened in the form of extensive localised fracturing and global splitting of the glass plies for both "1 and 2 ply carbon" type specimens.

Tables 5 and 6 show the geometric properties and tensile test results of "3 and 4 ply carbon" specimens. Please note, that the final elastic moduli in Table 6 were evaluated by fitting lines to the straight sections of the graphs of Figs. 7 and 8 after their plateau regions. These are approximate modulus values because after significant damage, the optical strain measurement system can lose full accuracy due to partial loss of the speckle pattern from the specimen surface which is to be tracked by the video gauge software. "3 and 4 ply carbon" type specimens showed conventional hybrid failure behaviour, with a significant drop in stress at the strain to failure of the carbon layer embedded. This drop was due to a single crack in the carbon layer running along the full specimen width, followed instantly by significant unstable pull out of the layer. In case of the "3 and 4 ply carbon" specimens, the extent of instant initial delamination was 20–30 and 40–50 mm each side of the carbon crack respectively. High scatter was observed in the extent of instant delamination, because it was disturbed by the end tab regions in some cases. After the instant delamination, continuous stable propagation of the delaminations was observed until the glass plies took the whole load, and the stress strain curve started to rise again after the propagation plateau (see Figs. 7 and 8 after points (4)). The load drop was more significant in the case of the "4 ply carbon" specimens, as expected, because delamination was more unstable due to the higher amount of energy released by the thicker carbon layer, and propagated further. Although the failure characteristics cannot be described as pseudo-ductile, because the stress–strain graphs have a notable sudden stress drop,

Table 5
Geometric properties of "3 and 4 ply carbon" type hybrid specimens.

Spec. type	Property	Width (mm)	Thickness (mm)	Free length (mm)
3 Ply carbon	Average	20.00	0.655	159.9
	Coeff. of variation (%)	–	1.10	0.4
4 Ply carbon	Average	19.93	0.685	157.2
	Coeff. of variation (%)	0.26	0.80	0.3

Fig. 9. Carbon-ply fragmentation in a "1 ply carbon" type specimen. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

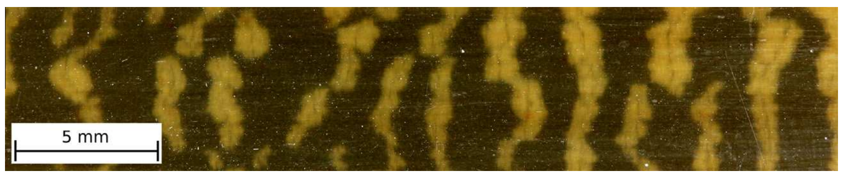

Fig. 10. Localised delaminations around multiple carbon-ply cracks in a "2 ply carbon" type specimen. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

Table 6
Tensile test results of "3 and 4 ply carbon" type hybrid specimens.

Spec. type	Property	Stress drop				Final strain (2) (%)	Initial elastic modulus (GPa)	Approx. final elastic modulus (GPa)
		Strain (3) (%)	Upper stress (3) (MPa)	Lower stress (4) (MPa)	Decrease in stress (% of upper)			
3 Ply carbon	Average	1.92	964.5	880.6	8.6	2.75	49.04	32.66
	Coeff. of variation (%)	1.63	2.6	2.5	47.7	3.71	1.67	3.57
4 Ply carbon	Average	1.93	984.5	749.8	23.5	2.75	51.41	31.91
	Coeff. of variation (%)	5.87	7.2	1.4	25.7	3.59	0.48	–

Table 7
Summary of selected test results for all specimen types.

Spec. type	Property	Carbon layer thickness (μm)	Initial elastic modulus (GPa)	Initial failure strain (%)	Final failure strain (%)	Type of pseudo-ductility if any
1 Ply carbon	Average	29	44.3	2.20	2.77	Pseudo-yielding
	Coeff. of variation (%)	–	1.69	0.78	2.73	
2 Ply carbon	Average	58	46.6	2.06	2.50	Pseudo-yielding
	Coeff. of variation (%)	–	2.49	1.02	6.39	
3 Ply carbon	Average	87	49.0	1.92	2.75	Plateau after minor stress drop
	Coeff. of variation (%)	–	1.67	1.63	3.71	
4 Ply carbon	Average	116	51.4	1.93	2.75	Short plateau and rise after notable stress drop
	Coeff. of variation (%)	–	0.48	5.87	3.59	

"3 ply carbon" specimens show a fairly stable failure type, with limited (8.6% average, 13% maximum observed) fall in stress, and a wide stable plateau up to 2.75% strain. The mentioned characteristics indicate that this specimen type has a carbon layer thickness that is close to the critical. The differences in failure characteristics of the specimens with variable central carbon layer thicknesses were caused by different energy release rates of the carbon layers. The thinner the central carbon layer, the less the energy released, and the more the delamination is suppressed.

3.4.2. Updated calculations

Using the elastic properties of the tested materials given in Table 2. (E_1 = 40 GPa, E_2 = 101.7 GPa, where index 1 refers to glass/epoxy and index 2 refers to carbon/epoxy) and the ply thicknesses, the energy release rates at carbon fibre failure initiation (kept constant at 1.93% strain) for various carbon layer thicknesses can be calculated with Eq. (7). The energy release rates (G) for lay-up sequences with 1–4 plies of thin carbon prepreg are 0.31, 0.69, 1.15, and 1.67 N/mm respectively. Comparing the energy release rates of various specimen types to the value of G_{IIC} = 1.1 N/mm which has been measured on similar hybrid specimens but with a cut through the entire carbon layer across the width, it can be stated that the calculated values are in very good agreement with the observed damage modes. Specimen types "1 and 2 ply carbon" showed stable pull-out of carbon layer, because the energy release rate in these laminates was subcritical at the initiation of carbon fibre failure. Whereas in case of specimen types "3 and 4 ply carbon" delamination took place instantly after carbon ply failure initiation, as the energy release rates for these were supercritical. The critical carbon layer thickness calculated with updated material constants and average plate thickness (h = 0.65 mm) using Eq. (6) is t_{2c} = 84 μm. It is indicated by the updated calculations that "3 ply carbon" specimen type has a carbon thickness that is just above the critical value, which is clearly reflected in the small load drop and the wide plateau on the stress–strain graphs.

3.4.3. Summary of results

Table 7 gives an overview of the most important properties valid for all the specimen types examined. According to the table it can be noted, that "3 and 4 ply carbon" specimens showed initial failures earlier than "1 and 2 ply carbon" specimens, probably because in the case of the thinner central layers the initial failure strain corresponds to multiple fibre failures sufficient to cause a significant change in stiffness rather than the first carbon failure. On the other hand the initial failure strain of the "3 and 4 ply carbon" specimens correspond to the point of delamination, which followed from the failure of the first critical cluster of carbon fibres. This also explains the high absolute value of strain of over 2% in the former case, which corresponds more to the average rather than the minimum strain of the carbon fibres. In general, hybrid specimens showed significantly higher strains at carbon layer failure than those of pure carbon/epoxy specimens, and hybrid specimens usually failed within the gauge section whereas pure carbon/epoxy showed explosive end-tab failure. A notable factor contributing to the high carbon strains to failure in all hybrid specimen types is that the glass plies acted as protective layers against grip stress concentrations. Elastic moduli of the specimens were slightly increased with the increasing carbon ratio, as expected. Final failure strains were similar, as failure was governed by the properties of the glass plies.

4. Conclusions

- Calculations have been presented to assess the mode II interlaminar fracture behaviour of a three layer ply-by-ply glass/carbon hybrid composite laminate following fracture of the carbon, and to predict the allowable central carbon layer thickness for stable pull-out.
- Stable failure to high strains has been successfully demonstrated on thin carbon prepreg reinforced glass/carbon hybrid composite materials.
- Central carbon layers $\sim$116 μm thick delaminated from the glass, the ones of $\sim$29 and $\sim$58 μm showed stable pull-out, whereas ones of $\sim$87 μm showed intermediate behaviour, consistent with the estimated transition in behaviour at about 84 μm.
- The crucial role of the central carbon ply thickness in a UD, three layer ply-by-ply glass/carbon hybrid laminate system has been proven, by observing significant changes in fracture characteristics, as a function of the carbon layer thickness.
- A novel and advantageous composite material structure has been developed that exhibits pseudo-ductile failure characteristics.

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