



A review of the effect of loading rate on the mechanical properties of unidirectional carbon fibre reinforced polymer composites

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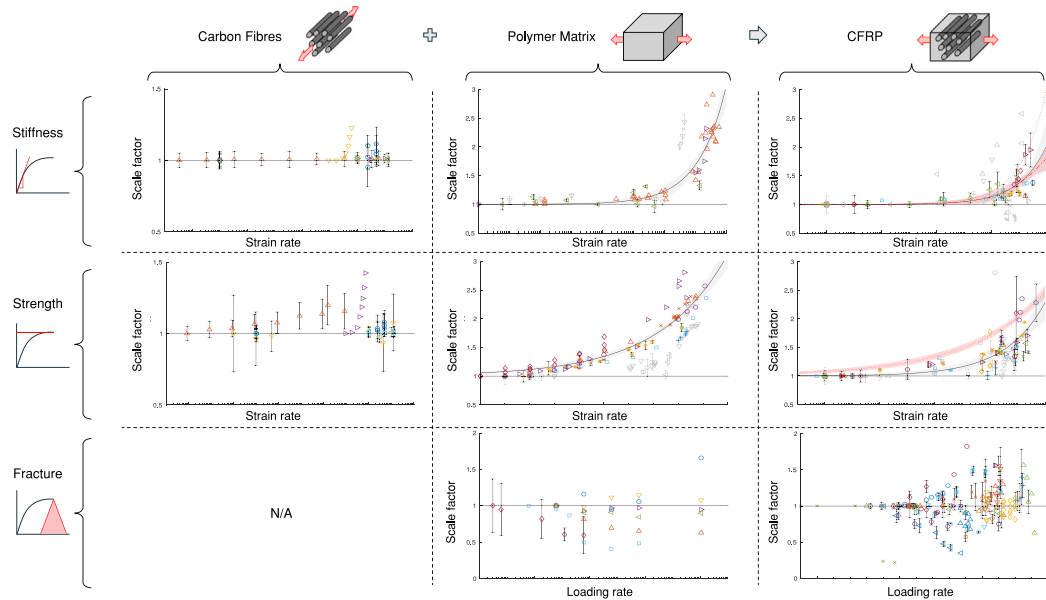
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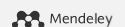
GRAPHICAL ABSTRACT

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ABSTRACT

A comprehensive literature review on the rate dependency of unidirectional CFRP composites has been carried out. Starting with the micro-scale constituents, literature data on the effect of loading rate on stiffness, strength, and fracture properties have been compared. While carbon fibres showed no significant rate effects, the rate dependency of the strength and stiffness properties was found consistent across various polymer matrix materials. In contrast, there was a general lack of consensus on the rate dependency of the matrix fracture properties.

At the macroscopic scale, the rate effects on the composite strength and stiffness properties were found to remain consistent across different combinations of constituent materials, while conflicting reports have been found on the rate dependency of the fracture properties. Finally, it has been shown that the rate dependency of composite strength and stiffness properties can be related to the constituent properties through established homogenisation theories and failure criteria.

1. Introduction

In the aerospace and automotive industry, increasingly stringent CO₂ reduction and safety requirements are driving the need for ever lighter high-performance structural components [1]. Fibre-reinforced polymer (FRP) and, in particular, carbon-fibre-reinforced polymer (CFRP) composites are key for the realisation of these new components, primarily due to their high specific properties, multi-functionality, and versatility.

Compared to conventional materials, FRPs come with a vastly extended design space encompassing its reinforcing fibre and matrix constituents, fibre architecture, stacking sequence, and other aspects. To make use of this plethora of degrees of freedom in an economically viable manner, a primarily simulation-based design and analysis process becomes necessary. Moreover, the use of FRPs in applications where the design-critical load case consists of impact or crash loading requires a thorough understanding of the composite material's behaviour across a sufficiently broad range of the loading rate spectrum. This, in turn, requires that the numerical design tools and constitutive models be able to accurately describe its entire range of rate dependent behaviour: from its elastic deformation, yield limits, and plastic deformation to its failure limits, damage evolution, and energy dissipation behaviour.

As a result, the topic of rate dependency in CFRPs has been steadily gaining attention in academic research since the 1980s, with studies covering aspects from specialised experimental techniques for high-rate characterisation to the effects of strain rate (SR) on different material properties and deformation or failure mechanisms. However, the characterisation of loading rate dependent material behaviour can be highly challenging. With the increase in available literature on the topic, there has also been an increase in conflicting or contradictory experimental results, which, in turn, can lead to confusion regarding the significance or magnitude of any strain rate effects on the different mechanical properties.

Among this growing number of publications, several reviews have also been published on the topic [2–13]. Kidane et al. [8] focused on the shear in-plane and out-of-plane strain rate sensitivity. The review paper of Singh [9] gives insight into the micro and meso scale of unidirectional (UD) carbon and glass composites. The publication of Ahmed et al. [11] presents an overview of the strain rate effect on the tensile behaviour of FRPs. Ma et al. [12] reviewed the rate dependency of FRPs with a focus on their fracture mechanisms and associated numerical models. The works from Cantwell et al. [2], Jacob et al. [5], and May [7,10] have reviewed the effect of the loading rate on mode I, mode II, and mixed-mode fracture toughness. A review of rate dependency of interlaminar fracture properties focused on cohesive zone modelling techniques was also presented by May [6]. Finally, the works by Field et al. [3] and by Perry and Walley [13] present an in-depth overview of the different experimental techniques used for

high-rate experimental characterisation in general and with a more specific focus the characterisation of FRP composites, respectively.

The present study aims to complement previous work on this topic with a more holistic, high-level review of loading rate effects on the main mechanical properties of CFRP composites, with a particular focus on their relationship to their constituent materials. By studying together the properties of the composites and their constituents, a better understanding of the source of the rate dependency for different deformation and failure mechanisms can be obtained, helping to assess current failure theories in the context of loading rate and give some clarity on previously reported conflicting experimental findings. In addition, for each of the reviewed mechanical properties, this study attempts to quantify the magnitude of their rate dependency and how well the literature data agree.

For practical reasons, the scope of this review has been narrowed to focus on UD-reinforced epoxy-matrix-based CFRP composites featuring a high volume fraction of fibre to matrix material, which represents the most common configuration for high-performance applications where dynamic loading effects are likely to play a significant role. Nevertheless, many of the findings in the present study may also be applicable to other types of UD FRP composites provided the fibre properties are significantly greater than those of the matrix. For glass-fibre-based FRPs, for example, the behaviour in matrix-dominated deformation modes can be expected to be very similar to that of the CFRP materials reviewed in the present work. However, an added loading rate effect can be expected in fibre-dominated modes, as the glass fibres are known to present rate dependent behaviour themselves [14–16]. Weaving and other reinforcement architectures have not been included in this review, as they can introduce strain rate effects of their own, which should not be conflated with the underlying behaviour of UD ply/yarn level material within. Furthermore, the range of loading rates of interest here has been limited to rates between quasi-static (QS, around 10^{-4} – 10^{-3} s⁻¹) to high-rate (HR, around 10^3 s⁻¹), thus excluding creep as well as ballistic and shock loading.

Finally, the data collected in the context of this work on the main mechanical properties of typical CFRP composites and their constituents at different loading rates have been made available in the Supplementary data section. It should be noted that in many cases, the original data were not provided explicitly and had to be extracted from graphs and figures, leaving room for possible inaccuracies in the data collection. In addition, for publications that reported individual test results, multiple individual data points from similar loading rates have been grouped and presented as the resulting mean and standard deviation values.

2. Background

This section aims to establish key concepts and terminology that are used throughout the rest of the manuscript. Section 2.1 introduces the fundamental aspects of fibre-reinforced composites, including the

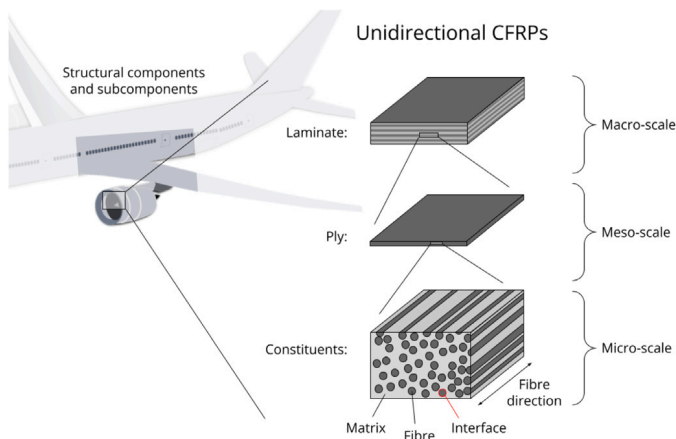


Fig. 1. Relevant length scales for a UD CFRP composite laminate. Source: Adapted from [17].

relevant length scales and nomenclature for the individual constituents and the resulting CFRP. Section 2.2 introduces the aspects related to material characterisation under high loading rates, including the loading rate spectrum covered in this work as well as typical experimental techniques for testing at different loading rate regimes. Finally, Section 2.3 introduces some of the common challenges with obtaining valid data from high-rate characterisation experiments.

2.1. Composite micro-structure and relevant length scales

The aim of combining different types of materials into a composite is to obtain improved overall properties through synergy effects. In the case of FRPs, at the microscopic or micro-scale, the material is composed of two main phases or constituents, the reinforcing fibres, which are typically in the order of $5\ \mu\text{m}$ in diameter, and the polymer matrix, Fig. 1 [17]. The ratio between the volume of these reinforcing fibres per unit of total volume of the composite is known as the fibre volume fraction, v_f . Another equally important aspect of FRPs is the bond between the fibres and the polymeric matrix. These interface, or interphase, regions between constituents are often the starting point for micro-cracks and can play a significant role in the damage and fracture of the composite. At the next length scale up from the micro-scale, so-called mesoscopic or meso-scale, a layer of FRP, typically in the order of several $10\ \text{s}$ to several 100s of μm , can be called a ply or lamina. In the case of UD FRPs, these laminae have very distinct properties in the longitudinal, or fibre reinforcement, and the transverse, or matrix-dominated, directions. Finally, at the macroscopic, or macro-scale, multiple plies are stacked to form a laminate. The orientation of the plies within a laminate can be optimised for a given application. Still, some of the more common ply lay-ups or stacking sequences include unidirectional, with all plies in the same direction, cross-ply or bi-directional, with plies offset at 90° from each other, e.g. $[(0_a/90_b)_n]_s$, or quasi-isotropic laminates, with a stacking sequence that provides reasonably similar in-plane properties for any loading direction.

2.2. High-rate experimental characterisation

Certain types of materials, including polymers and, therefore, also FRPs, are known to exhibit time-dependent behaviour, resulting in changes to their mechanical response at different loading rates. However, experimental characterisation of this behaviour can be challenging under dynamic conditions. In the following, an overview of the loading rate spectrum and associated “regimes” is given. Examples of commonly used devices and test set-ups are also presented in order to give a broader overview of the fundamental aspects of material characterisation under elevated loading rates. However, before going any

further, it should be noted that while the term “strain rate” is generally used throughout the present work, there can be significant differences between the macroscopic strains measured during the elastic, elasto-plastic, or damage evolution stages of the material response and what they really mean in terms of microscopic deformation mechanisms. Therefore, care should be taken when comparing strain rates measured at different stages in the material deformation. Also, for certain experiments, strain measures cannot be easily obtained, and loading or displacement rates are used instead. This is the case for certain fracture toughness or energy release rate (ERR) tests and, because of this, certain material properties in Section 6 only include more qualitative comparisons between loading rate regimes. Having said that, the strain rate spectrum can be divided into several ranges or regimes, shown in Fig. 2, with no clear-cut boundaries between adjacent regimes. The dynamic regime in particular is of special interest for impact and crashworthiness scenarios within the automotive and aerospace industry, where the relevant strain rates can be in the order of a few $100\ \text{s}^{-1}$ or even several $1000\ \text{s}^{-1}$ in localised damage areas [18–20].

Historically, different experimental techniques have been utilised for material characterisation at different loading rates. Some of the most relevant for the present work include screw-driven machines for strain rates between roughly 10^{-5} to $10^{-2}\ \text{s}^{-1}$, servo-hydraulic devices for strain rates between 10^{-5} to 10^1 or $10^2\ \text{s}^{-1}$, and Kolsky or split-Hopkinson bar (SHB) systems, whether tensile (SHTB) or compressive (SHPB), which are illustrated in Fig. 3 and have typically been used for dynamic testing at strain rates between 10^2 to $10^4\ \text{s}^{-1}$ [21,22]. However, the limiting factor at higher strain rates is not only the loading rate a device is physically capable of applying but also whether the results can be reliably post-processed. One of the common issues with dynamic characterisation tests is the introduction of oscillations due to the inertia and/or the resonance modes of the specimen and test set-up, which can distort the force measurements at higher loading rates. For this reason, the SHB method, with its ability to record the shapes of stress waves travelling into and out of the specimen, has become widely used for characterisation at high strain rates between 10^2 and $10^4\ \text{s}^{-1}$. However, as is discussed in more detail in Section 2.3, the SHB method can also become unreliable at high enough loading rates, leading to novel techniques, such as the image-based inertial impact (IBII), which is also shown in Fig. 3, emerging in recent years for more reliable characterisation at strain rates above $10^3\ \text{s}^{-1}$ [23–26]. There also exist several international standards for testing of polymers and/or polymer composites at high loading rates [27–31]. However, these generally focus on servo-hydraulic and impact testing and do not include some of the more specialised techniques mentioned above, which have been shown to produce more reliable results at strain rates above $10^2\ \text{s}^{-1}$. With this in mind, the results reviewed throughout the present work have also been used to assess the suitability of different experimental techniques for high-rate testing, which is discussed in more detail alongside the other conclusions in Section 7.

Other experimental techniques for dynamic loading include drop tower, pendulum impact, expanding ring, gas gun, dynamic mechanical analysis, but these will not be covered in depth in this review. A more detailed overview of the experimental characterisation set-ups and techniques can be found in [3,13,32].

Finally, in the context of polymer and polymer composite materials, experimental characterisation tests using the techniques described above have shown that many of their mechanical properties tend to increase with the loading rate following an exponential trend of the type illustrated in Fig. 4 [33], where the horizontal axis represents the base 10 logarithm of the strain rate, $\dot{\epsilon}$.

Several types of scaling functions have been shown to adequately capture this exponential trend typically observed in polymeric materials. Mayr et al. [34] proposed that the yield stress of a polymer could be described by a semi-logarithmic function of the strain rate derived from the Eyring–Polanyi equation [35,36]. The same type of equation was used in the Johnson–Cook model [37] and several authors

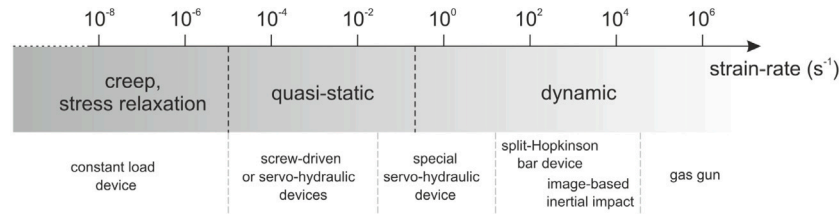


Fig. 2. Summary of different strain rate regimes from creep to high-rate loading and the experimental techniques typically associated with each.

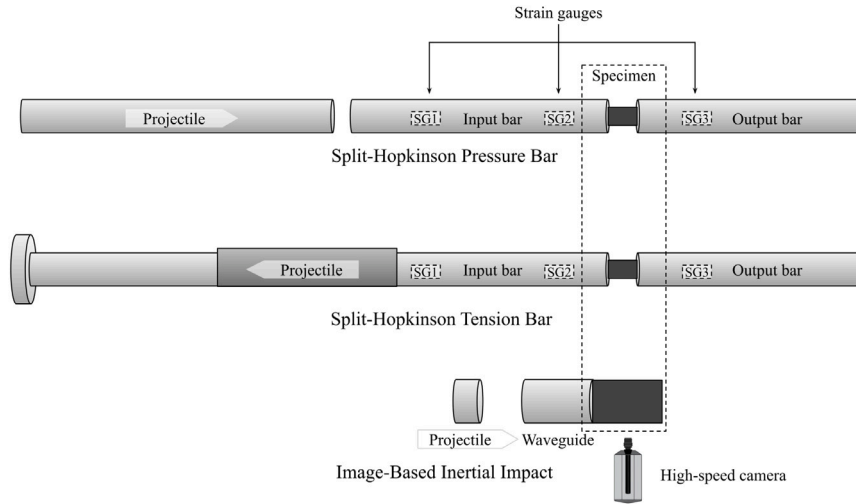


Fig. 3. Illustrated examples of a compressive and a tensile Split-Hopkinson bar set-up (SHPB and SHTB, respectively) and an Image-Based Inertial Impact (IBII) set-up [21–23]. In contrast to the latter, high-speed imaging is not a strict requirement for SHB testing.

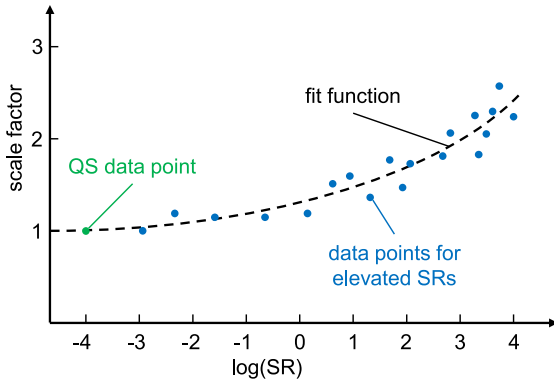


Fig. 4. Typical strain rate dependency observed for the mechanical properties of polymers and FRPs and the best-fit non-linear regression of the Cowper-Symonds exponential scaling function.

have shown it to describe the rate dependency of polymer and matrix-dominated composite strength properties [38–40]. Here, however, the rate scaling approach used in the Cowper-Symonds (CS) model [33] has been selected, Eq. (1), as it allows greater flexibility to fit experimental data that deviates from the semi-logarithmic trend at higher strain rates.

$$f(\dot{\epsilon}) = 1 + (K\dot{\epsilon})^{\frac{1}{n}} \quad (1)$$

Where $\dot{\epsilon}$ denotes the strain rate, and K and n two fitting parameters.

Throughout the following sections, a non-linear least-squares approach has been used to fit the collected literature data to the Cowper-Symonds model above as a way to quantify the rate dependency of different material properties. The best-fit parameters and their associated 95% confidence intervals are illustrated alongside the collected

data in each section and summarised in Table 1 in the conclusions, Section 7. In cases where additional models have been evaluated for a given data set, such as the rule of mixtures for the composite stiffness properties, the root mean squared error (RMSE) is used to assess their predictive capability relative to the best fit CS model. Finally, it should be noted here that, for some publications, the quasi-static strain rate value was not provided in the original source, and a typical quasi-static rate of 10^{-4} s^{-1} has been assumed. In the provided supplementary data for these cases, the strain rate is simply denoted as ‘QS’ in the ‘rate’ column.

2.3. Experimental challenges and validity of high-rate measurements

At elevated loading rates, the presence of resonance modes and/or inertia effects in the force transducer, specimen, and experimental set-up can increasingly mask the sample material’s response to the applied load when using traditional measurement techniques. In order to ensure valid test results, it is required that the forces acting on each of the specimen’s loading faces, see Fig. 5(a), be in reasonable agreement.

While, other approaches have been proposed for more specific applications, such as in the work from Medina et al. [41], which focused on the validity of high-rate double-cantilever-beam (DCB) test results, a more general measure of the degree of convergence, or “disequilibrium”, between the loading faces of the specimen can be determined according to [42]:

$$A_F = \left| \frac{F_{in} - F_{out}}{\frac{1}{2}(F_{in} + F_{out})} \right| \quad (2)$$

where F_{in} and F_{out} denote the forces acting on the specimen ends.

When this value is high, it can indicate non-uniform loading across the length of the specimen so that averaged stress and strain measurements may not be a valid representation of the actual state of the material. The valid section of the obtained stress-strain data can

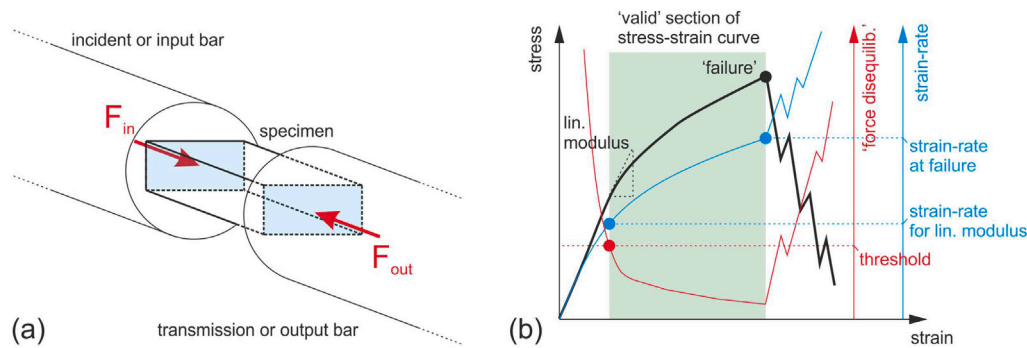


Fig. 5. (a) Forces acting on both ends of the specimen (here illustrated in the context of SHB testing), (b) stress-strain curve in conjunction with its associated force disequilibrium and strain rate data.

be identified by requiring Δ_F to fall (and remain) below a certain threshold, say, for instance, 10%, see Fig. 5(b). The initial period of non-uniform loading becomes more critical at elevated loading rates, as it renders it increasingly difficult to meet the force convergence requirement within the duration of the test. This is often the case for the Young's modulus, which is determined at the origin of the stress-strain curve, and for any strength properties associated to brittle failure modes, so that they become increasingly challenging to characterise the higher the loading rate.

In addition, under high-rate loading, even when valid stress and strain data can be obtained, the strain rate becomes more likely to change significantly throughout the test, Fig. 5(b), so that care must be taken when associating an obtained material property with a single strain rate value. In extreme cases, changes to a specimen's compliance during the course of a test can lead to strain rates at the moment of failure as much as an order of magnitude greater than the value during the initial elastic response. In the context of the present work, this may add further uncertainty when comparing data across a vast range of literature sources, as it is not always clear how the reported strain rate values have been obtained.

Throughout the following sections, any literature results that are considered unreliable for the reasons above or otherwise not directly comparable to the rest of the collected data have been identified and excluded from the curve fitting exercise to avoid misrepresenting the underlying trends.

As a final note, novel material characterisation techniques, such as the IBII introduced in the previous section, have been developed in recent years that aim to address the above limitations for very high loading rates by measuring the local stresses and strains across the specimen instead of an averaged value for the entire gauge volume.

3. Constituent properties

Due to their heterogeneity and complex micro-structure, together with the experimental challenges associated with dynamic material characterisation, the rate dependency of CFRP composites can be challenging to assess. However, by first reviewing the effects of strain rate on their constituent materials, together with the current understanding of the main deformation and failure mechanisms in unidirectionally-reinforced CFRPs, a better understanding of the macroscopic rate effects observed in the composite material may be obtained. To this end, the effects of strain rate on the properties of common polymeric matrix and carbon fibre materials are reviewed in the following sections, 3.1 and 3.2, respectively. For the matrix materials, data on the elastic, strength, and fracture properties have been collected. For the carbon fibres, only studies on the rate dependency of elastic and strength properties have been found. The constituent data reviewed in the following sections are also used in Sections 4–6 to evaluate the rate effects observed on the meso-scale properties of CFRPs and the current understanding of their deformation and failure mechanisms.

3.1. Matrix

The time-dependent behaviour of polymers is a well-known phenomenon that has been extensively studied. In amorphous glassy polymers, such as those typically used in FRP composites, this strain rate (and temperature) sensitivity can be attributed to the transition between different types of molecular motion [43–46]. As a result, the deformation mechanisms in the polymer tend to become both elastic and viscous, or time-dependent, as larger-scale molecular motion is enabled. From a solid mechanics perspective, the material's response at a given strain rate depends on the interaction between these mechanisms and can be represented in a simplified manner by rheological models as combinations of spring and dashpot elements, depending on whether the opposing force is related to the displacement (elastic) or to the velocity (viscous). At high enough accelerations, inertial forces from the mass of the testing fixtures and/or the specimen itself can also become relevant and can lead to errors in the stiffness and strength measurements if not accounted for.

3.1.1. Elastic properties

The collected data on the stiffness properties of various thermosetting matrix materials typically used in high-performance CFRPs are shown in Fig. 6 [47–51]. Keeping in mind the underlying deformation mechanisms, the data not only show a clear strain rate dependency but also a distinct change in the trend at around 10^2 s^{-1} that may be associated with the activation of viscous effects. In addition, considering different test methods and materials used, the normalised stiffness data collected shows very good agreement all the way from quasi-static to high strain rates with relatively low experimental scatter. Finally, keeping in mind the challenges discussed in Section 2.3, the data collected at higher strain rates ($> 1000 \text{ s}^{-1}$) becomes more likely to misrepresent the true stiffness of the material, which may explain the greater variation observed. Nevertheless, the normalised stiffness measurements follow a consistent trend across all data sets all the way up to the highest loading rates. The Cowper–Symonds function from Eq. (1) is also shown to fit the experimental data well with the parameters given in Table 1 in the final conclusions section.

3.1.2. Strength properties

In addition to the effects discussed above on their elastic properties, the failure mechanisms of common CFRP matrix materials also tend to transition from ductile to more brittle behaviour when going from low to high strain rates (or, in a similar manner, when going from high to low temperatures) [43–45]. This is usually accompanied by lower plastic deformation as well as strain-to-failure as the movement of polymer chains is increasingly restricted. Therefore, as the strain rate rises, this can simultaneously increase the material's stiffness while also reducing its ductility, or strain-to-failure, so that the rate dependency of the polymer's strength properties may follow a completely different trend.

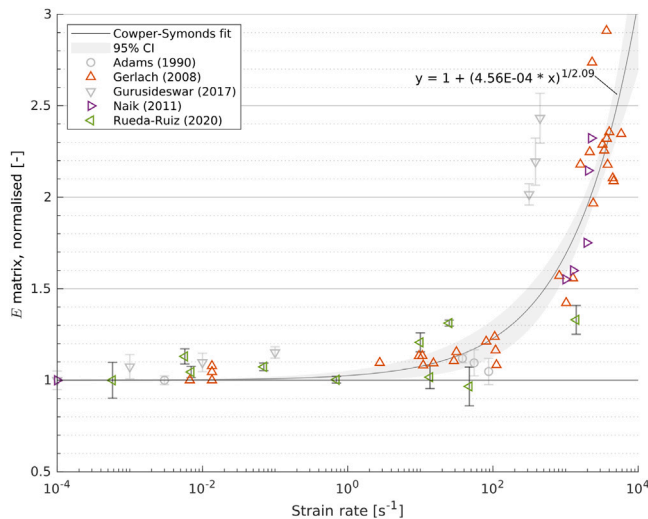


Fig. 6. Normalised Young's modulus for various polymer matrix materials from quasi-static to dynamic loading rates as reported in the literature [47–51]. The Cowper-Symonds (CS) parameters and 95% confidence interval are given in Table 1. Series excluded from the Cowper-Symonds fit are indicated with grey markers.

At the same time, the onset of failure tends to be much less repeatable than the behaviour in the elastic region and characterisation experiments are more susceptible to stress concentrations, especially for dynamic testing techniques, leading to greater scatter in experimental measurements. Because of this, data collected in Fig. 7 for strength of polymers at different strain rates show greater overall scatter than the stiffness data in Section 3.1.1 [34,47–56].

Nevertheless, the collected data are, once again, in very good agreement, with only a few exceptions. The results from Adams et al. [47] and Gurusideswar et al. [49] both showed pronounced rate dependency across different high-rate test cases but little to no strength increase from the quasi-static tests to the first high-rate case, causing the normalised properties to fall significantly below the general trend. In both studies, the high-rate tests were obtained using experimental techniques more susceptible to stress concentrations and premature failure, with Adams et al. [47] using a pendulum impact and Gurusideswar et al. [49] using a drop tower impact set-up, respectively. Therefore, it seems that the change in test set-up between quasi-static and high-rate loading may have introduced changes to the failure modes and resulting ultimate stresses, so that the quasi-static and high-rate strength measurements may not be directly comparable. In the work from Gerlach et al. [48], on the other hand, no reference quasi-static results are given below 10^{-2} s^{-1} , for which the rest of the reviewed data sets already showed a strength increase as high as 10% to 30% relative to quasi-static tests performed between 10^{-5} and 10^{-3} s^{-1} [34,49,51,53,54,56]. As a result, the data from Refs. [47–49] cannot be normalised and compared in the same way and have therefore been excluded from the curve fitting in Figs. 6 and 7.

Aside from the few exceptions discussed above, the Cowper-Symonds model can be seen to provide an excellent fit for the rate dependent scaling of the polymer strength properties, with the best-fit parameters given in Table 1 in the final conclusions section.

Therefore, regardless of the difference in materials and test methods, it appears that the rate dependency of both strength and stiffness properties is consistent across all reviewed epoxy resins, which provides a strong basis for the evaluation of the composite strength and stiffness properties in Sections 4 and 5.

3.1.3. Fracture properties

Finally, with regard to both the fracture toughness, K_{IC} , and critical energy release rate, G_{IC} , no clear increasing or decreasing trend can be

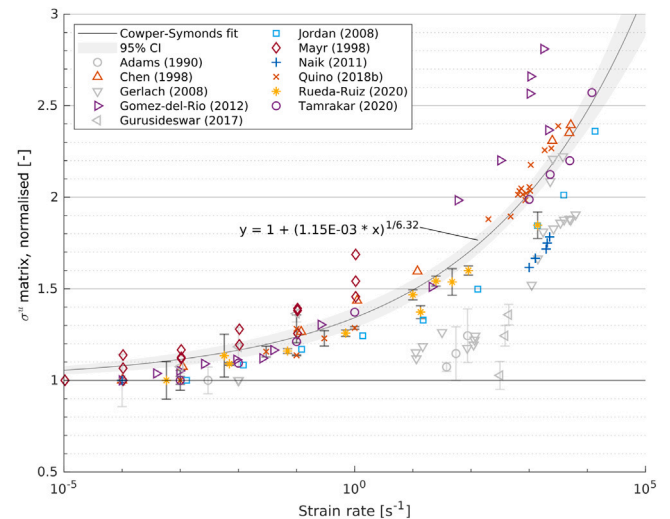


Fig. 7. Normalised strength for various polymer matrix materials from quasi-static to dynamic loading rates as reported in the literature [34,47–56]. The Cowper-Symonds (CS) parameters and 95% confidence interval are given in Table 1. Series excluded from the Cowper-Symonds curve fitting are indicated with grey markers.

found across the reviewed literature data, with different sources claiming positive, negative, or no significant strain rate effect, Fig. 8 [57–62]. Kanchanomai et al. [58] conducted single-edge notched bending (SENB) tests for the three-point bending fracture toughness of thermoset epoxy resin to determine the critical strain energy release rate, G_{IC} , at different loading rates and found that it started to decrease at loading rates above 3 – 10 mm/min. Fractography revealed the presence of shear lips and substantial deformation during fracture for the specimens tested under low loading rates, resulting in a plane stress-dominated behaviour. Pittman et al. [62] also found indications of crazing and refraction in front of the crack tip at low loading rates and noted that plastic deformation around the crack tip was due to the time-dependent deformation nature of the polymer. The presence of crack blunting indicated by localised plastic deformation consequently led to a higher load being required to propagate the crack. However, the smooth fracture surfaces of samples tested at higher rates showed brittle fracture and a plane strain-dominated condition [58]. This rate effect is also dependent on the degree of cross-linking. Pittman et al. reported an increase in dynamic fracture toughness for a highly cross-linked epoxy and hypothesised that fracture toughness could increase with loading rate as the failure mechanism switches from crazing to shear yielding due to the time-dependent nature of the crazing process [62]. Joudon et al. [63] also reported a 25% increase in mode I dynamic fracture toughness of an epoxy resin toughened with thermoplastic particles as the loading rate increased. In addition, they found clear evidence of the competition between shear yielding and crazing on the fracture surface. In contrast, in a series of compact tension tests conducted using a standard screw-driven universal testing machine, Quino found that the average G_{IC} of an epoxy resin decreased by up to 40% over almost four decades [59,64]. However, due to the data dispersion, the range of loading rates covered, and the testing methodology employed, these results should be interpreted with caution. Mishra et al. [61] used in-situ testing in a scanning electron microscope to observe the micro mechanisms involved in the fracture process zone (FPZ) ahead of the crack tip. It was found that crazing in the FPZ was dominated by chain disentanglements at low strain rates, while fracture can occur through the breaking of covalent bonds at higher strain rates, and the dominant mechanism of craze failure is expected to be through chain scission. However, the exact contribution of disentanglement crazing and scission crazing in the rate dependent fracture of epoxy resins remains an open research problem.

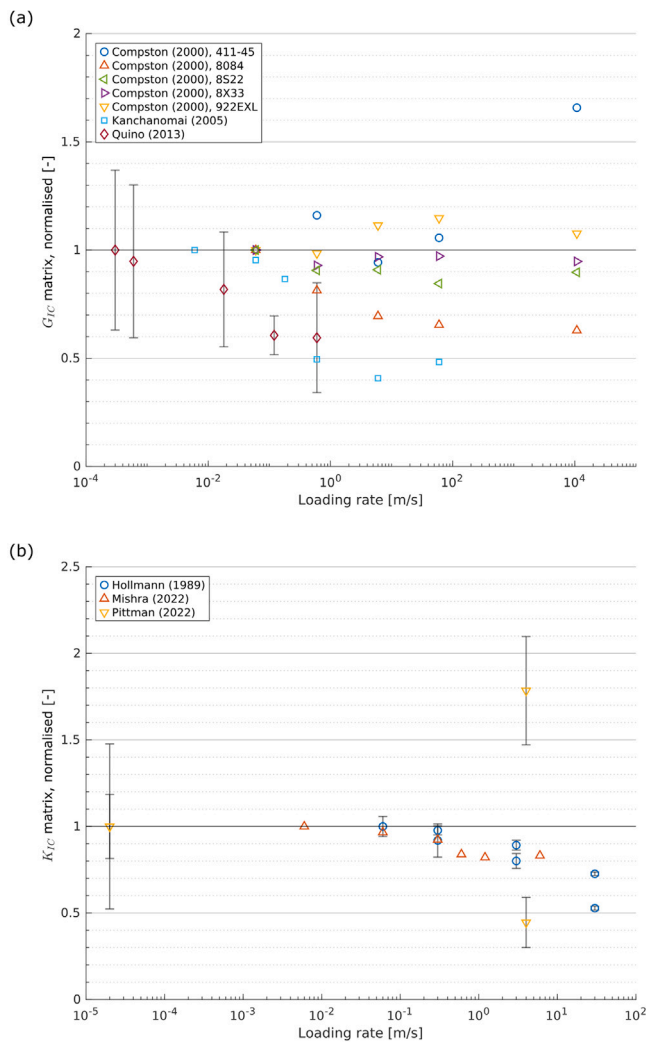


Fig. 8. Normalised fracture properties for various polymer matrix materials from quasi-static to dynamic loading rates as reported in the literature, with (a) the critical energy release rate, G_c , [57–59]; and (b) the fracture toughness, K_c , [60–62].

Overall, there have been reports of the fracture properties of epoxy resins both increasing and decreasing with the loading rate, with different possible explanations given for each, such as changes from ductile to brittle fracture or from crazing to shear yielding mechanisms. However, while there is no clear consensus, the majority of sources have reported zero or negative rate dependency, with much of the variation across loading rates falling within the high experimental scatter (as high as 40%, even at QS loading rates). Therefore, it seems reasonable to assume that, for epoxy resins commonly used in CFRPs, the rate dependency of the fracture properties is likely to be negligible or slightly negative. Nevertheless, it is hard to draw any firm conclusions based on the reported data, other than it remains a crucial area for future research to improve the reliability of fracture characterisation techniques across all loading rates.

3.2. Carbon fibre

Whereas the rate dependency of polymeric matrix materials has been clearly established, the mechanical properties of carbon fibres are typically assumed rate insensitive. However, experimental characterisation of individual fibres at dynamic loading rates is not straightforward, so that relatively few studies have been published on the rate dependency of their mechanical properties. While a modified miniaturised

Hopkinson bar has been developed to test single fibres at high strain rates for various materials [65–68], to the knowledge of the authors, only one study has used such a technique for characterising single carbon fibres [69]. Fibre properties reported in the literature have been typically measured by testing fibre bundles, either dry or impregnated with resin and cured, similar to the specimens suggested by the standard ASTM D4018 [70]. The average stress carried by individual filaments can then be estimated from the applied tensile load and the total cross-sectional area of the fibres, given by the number of filaments in a bundle and their average diameter [71] or by their volume fraction [70,72]. Although the impregnated fibre bundles may already be considered a CFRP composite, the small specimen size makes it possible to separate the behaviour of the reinforcing fibres from the surrounding polymer while containing a large enough number of fibres to give a more representative, averaged, response.

Collected data on the longitudinal tensile modulus of various carbon fibres is shown in Fig. 9 [73–78] and on the tensile strength in Fig. 10 [69,73–78]. Generally, the tests on fibre bundles were performed by adhesively bonding the bundles to a lining block through which the tensile load was applied, using universal testing machines for quasi-static loading rates, hydraulic machines for intermediate loading up to approximately 1 s^{-1} [75], and SHB for high strain rates up to approximately 1000 s^{-1} [69,73,74,77,78]. Overall, the results confirm the lack of any significant strain rate sensitivity for the carbon fibres. The only exception are the results from Younes et al. [76], which show a sharp increase in both strength and stiffness at strain rates above 10 s^{-1} . However, all tests showed a similar strain-to-failure and the dynamic tests were carried out using a drop tower set-up, which is more susceptible to overestimation of force measurements compared to the SHB tests from Refs. [69,73,74,77,78]. It is also worth noting that the results from Huang et al. [74] showed a slight increase in the measured stiffness and Naito et al. [75] showed a slight increasing trend in the strength measurements. However, the stress-strain results in [74] showed significant non-linearity for carbon fibres and the authors acknowledged the possible inaccuracy of the high-rate Young's modulus measurement. In addition, the slight increase in average bundle strength observed in [75] was attributed to an increase in the experimental scatter (decrease in Weibull modulus) due to the increase in strength and stiffness of the surrounding matrix, which allowed for better load transfer between individual fibres at higher strain rates. Therefore, the strength of the carbon fibres themselves was still deemed to be rate insensitive, as the observed increase in strength was associated with the composite nature of the fibre-matrix bundles.

3.3. Fibre-matrix interface

An additional aspect at the micro-constituent scale that can exert a significant influence on the CFRP properties is the fibre-matrix interface's strength and fracture energy. Huang et al. [79] presented a detailed review of interface characterisation techniques for FRP composites, such as microdroplet, single fibre fragmentation, fibre push-out and fibre pull-out tests. However, those techniques mostly focused on quasi-static loading rates and only a few studies have attempted to characterise the interface properties under dynamic loading. Tanoglu et al. [80] and Morlin et al. [81] characterised the interface properties for loading rates up to several mm/s and reported an increase of the interface's shear strength with strain rate of as much as 50%, but while they may technically be considered in the dynamic range, those rates were still relatively low within the context of the present work. There has also been some work done at higher loading rates but the question of rate dependency of the interface properties is not yet conclusively answered. Greenfield et al. [82] performed a single fibre fragmentation test to determine the interface strength at a loading rate above 1 m/s but did not measure the effect on the fracture energy. Tamrakar et al. [83,84] proposed a SHTB microdroplet test method for a glass fibre epoxy composite interface at a loading rate of approximately

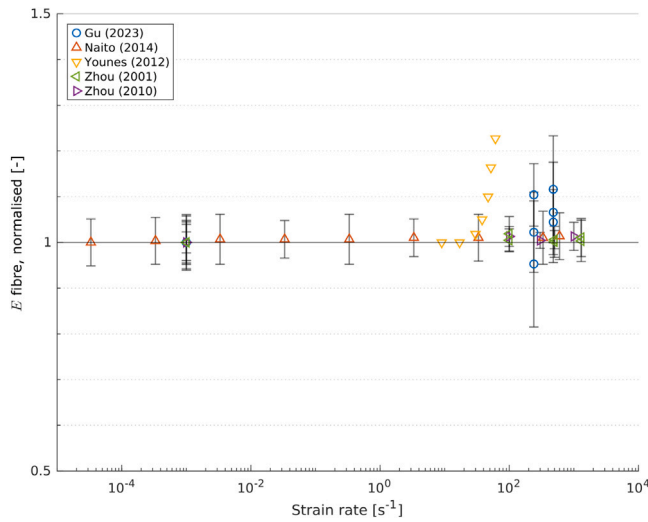


Fig. 9. Normalised Young's modulus for various carbon fibre materials from quasi-static to dynamic loading rates as reported in the literature [73–78].

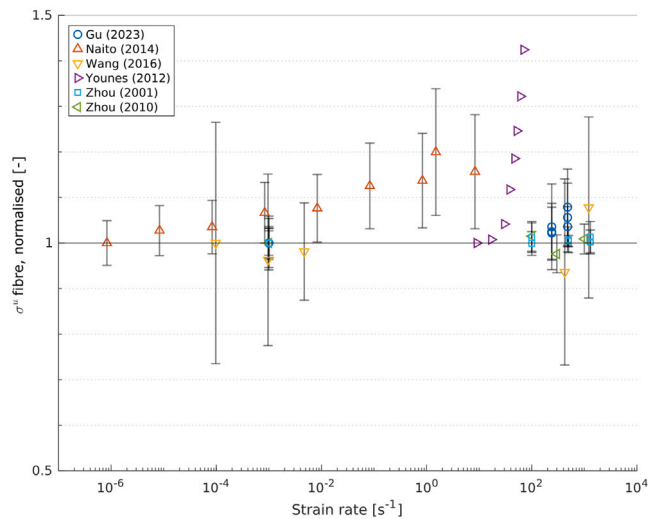


Fig. 10. Normalised tensile strength for various carbon fibre materials from quasi-static to dynamic loading rates as reported in the literature [69,73–78].

1 m/s, showing a significant increase in both strength and fracture energy. More recently, Rueda-Ruiz [51] attempted to use a high-rate nano-indentation technique at impact loading rates but the quality of the extracted force–displacement data was not good enough to determine the interface properties. Instead, the same authors proposed a micro-mechanical modelling approach as an alternative to estimate potential strain rate effects on the fibre/matrix interface's strength and stiffness in [85].

Overall, while there has been some promising work in this area, there is a general lack of consensus and very few of the experimental studies have reported results at loading rates above a few mm/s. In addition, many techniques also relied heavily on numerical modelling with many simplifying assumptions to post-process the experimental results, which may compromise the validity of any determined interface properties if any of the relevant physics is not accounted for. Therefore, no firm conclusions can be drawn at this time regarding the rate dependency of the fibre-matrix interface properties, making this topic another very interesting area for future research.

4. Composite stiffness properties

In the previous section, the strain rate dependency of typical high-performance FRP constituents has been reviewed, covering various types of carbon fibres and polymer matrix materials. At the ply or laminate scale, the behaviour of the composite is ultimately the result of the interaction between fibres and matrix, including the interface between the two. Therefore, strain rate dependency can generally be expected in the composite for any deformation modes where the resulting behaviour depends on any constituent properties known to be rate dependent themselves. Starting with the elastic moduli, the following sections contain a review of strain rate dependency of unidirectional CFRPs and how they relate to the rate dependency of their constituents. Generally, the elastic response of a unidirectional ply/lamina can be assumed transversely isotropic, with the behaviour in the longitudinal direction, E_{11} , being dominated by that of the reinforcing fibres, and the behaviour in the transverse and shear directions, E_{22} , E_{33} and G_{12} , G_{13} , G_{23} , by that of the matrix. These same subscripts, following the ply-scale material coordinate system illustrated in Fig. 1, are used throughout the following sections to indicate the different components of the reviewed mechanical properties.

The effects of strain rate on the longitudinal and transverse moduli are discussed separately in the following subsections, including comparisons to the expected rate dependency, based on the rule of mixtures approach described below and the properties of the constituents reviewed in Section 3. Assuming transverse isotropy, the composite modulus in the longitudinal, $E_l = E_{11}$, and transverse, $E_t = E_{22} = E_{33}$, directions can be approximated from the properties of the constituents by using the rule of mixtures, Eqs. (3) and (4) [86,87]:

$$E_l = v_f E_{l,f} + (1 - v_f) E_m \quad (3)$$

$$E_t = \left(\frac{v_f}{E_{t,f}} + \frac{1 - v_f}{E_m} \right)^{-1} \quad (4)$$

Where v_f is the volume fraction of the fibres within the composite, $E_{l,f}$ and $E_{t,f}$ are the longitudinal and transverse moduli of the fibre, respectively, and E_m is the modulus of the matrix. Therefore, having found the rate dependency of most matrix materials to follow a similar trend in Section 3.1, the expected rate dependency of the composite properties can be determined. For the laminates investigated in the present work, representative values of $v_f = 0.6$, $E_{l,f}/E_m = 3$, and $E_{t,f}/E_m = 55$ have been used to compare the trends extracted from the collected matrix data in Section 3 to the CFRP results, based on typical properties for IM7/8552 [88,89].

4.1. Transverse and shear moduli

As mentioned above, under transverse loading the elastic response of UD FRPs is dominated by the matrix properties, which have been shown in Section 3 to be strongly rate dependent. Therefore, the transverse and shear moduli of the composite (E_{22} , E_{33} and G_{12} , G_{13} , G_{23}) can also be expected to exhibit strong rate dependency. Using the rule of mixtures, the rate dependency of the composite's transverse and shear moduli can be assumed to behave like a combination of springs in series and can be calculated from the properties of its constituents according to Eq. (4). However, when comparing the rate dependency of the matrix and the composite, it should also be noted that, for a UD CFRP under transverse loading, the average local strain rate in the matrix will generally be higher than the average macroscopic strain rate measured across the composite [90], which could result in a slight shift between the two trends.

The expected rate dependency for a typical UD CFRP based on the rule of mixture and the data from Section 3 is shown in Fig. 11, indicated by the dashed red line, alongside the collected literature data for the transverse stiffness properties of various composite materials at different strain rates [23–26,91–100].

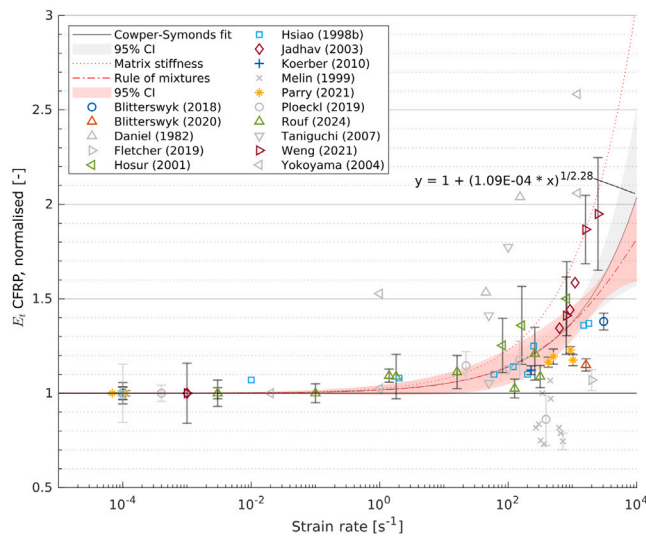


Fig. 11. Normalised transverse and shear moduli for various UD CFRP materials from quasi-static to dynamic loading rates as reported in the literature [23–26,91,93–100]. The Cowper–Symonds (CS) parameters and 95% confidence interval are given in Table 1. Series excluded from the Cowper–Symonds fit are indicated with grey markers. The expected trend based on the rule of mixtures and assumed micro-mechanical parameters is included in red for reference.

With the exception of the results from Melin et al. [96], who used a Moiré-based approach that appeared to overestimated the specimen strains (approximately 2x the strain gauge measurements), the collected literature data all show an increasing trend for the stiffness under elevated strain rates. Excluding this and a few other sources discussed below, the Cowper–Symonds equation with the best-fit parameters given in Table 1 appears to describe the trend in the collected data reasonably well. Furthermore, the 95% confidence intervals for the rate dependency of the transverse stiffness and the expected trend based on the rule of mixtures can be seen to overlap across the entire range of covered strain rates. In addition, the error between the rule of mixtures trend (RMSE = 0.138) and the best fit CS trend (RMSE = 0.135) is relatively small considering the uncertainties involved. Therefore, it can be concluded that the rule of mixtures can provide a reasonable estimate for the high-rate stiffness of a composite from known rate dependent properties of its constituents.

While the majority of the reviewed matrix data appear to follow a very similar trend, a number of sources reported stiffness increases significantly greater than what could be attributed to the polymers reviewed in Section 3 [99,101,102]. To get a better understanding of these results, a closer look at the experimental techniques used in each case is needed, keeping in mind the challenges with high-rate characterisation testing discussed in Section 2.3. While most of the presented dynamic test results were obtained by SHB, there are a few exceptions that used devices more poorly suited for high-rate measurements. The results from Daniel et al. [101] were obtained by measuring the deformation of thin rings under internal pressure, using explosives to generate the high-rate pressure waves with no way to verify the force equilibrium throughout the specimen, which may explain the unexpectedly high stiffness measurements.

Although such inertia-related issues are most often observed in larger, more massive test set-ups such as drop-tower and servo-hydraulic machines, even the SHB technique can suffer from inertia effects and poor convergence between incoming and transmitted forces at high enough strain rates, as described in Section 2.3. This was given as the likely explanation by Ploekl for the inconsistent transverse compression stiffness measurements recorded at strain rates above 200 s⁻¹ in [97]. As an alternative to SHB testing, IBII has more recently been proposed to investigate the strength and stiffness of materials at very

high strain rates [23–26]. This technique is designed to provide more reliable measurements at strain rates above 1000 s⁻¹, for which the influence of inertial effects would only allow for the measurement of an apparent modulus with the SHB. Several studies have investigated the transverse and shear moduli of UD FRPs using this technique in recent years [23–26], and the results are generally lower than the SHB data obtained at similar strain rates, such as the results from Weng et al. and Yokoyama et al. [100,102]. However, some of the earlier IBII results from Fletcher et al. [24], may have also underestimated the high-rate elastic moduli due to onset of micro-damage and non-linearity. In more recent work by Parry et al. [26], special care was taken to exclude the post-damage non-linear response when fitting the transverse modulus E_{22} , which resulted in a higher value, closer to the calculated rule of mixtures trend in Fig. 11.

With regard to the out-of-plane transverse modulus, E_{33} , data from Yokoyama [102], part of the explanation for the exceptionally high increase observed at higher strain rates may lie in the fact that the SHB technique was used at such high strain rates. However, it is also possible that the reference QS stiffness for the T700s/Toray-2501 material was underestimated since it was significantly lower than the QS value for the similar T700/2500 reported in Ref. [99]. Furthermore, the quasi-static and intermediate-rate stress–strain curves reported in [102] showed some signs of initial slack in the test set-up, which, if unaccounted for, could result in an underestimation of the initial elastic modulus.

Finally, the in-plane shear modulus, G_{12} , data from Taniguchi et al. [99] also appeared to overestimate the strain rate effect beyond what can be expected from the matrix data reviewed in Section 3. However, this could be explained by the fact that the shear modulus was extracted by post-processing 45° off-axis test results, which is a common approach also employed in [26,103] but can result in an overestimation of the shear strains if fibre rotation effects are not properly accounted for, as discussed in [104].

4.2. Longitudinal modulus

In contrast to the transverse stiffness, the elastic response in the longitudinal or fibre direction is dominated by the reinforcing fibres, whose modulus is usually orders of magnitude greater than that of the matrix (in the order of 200–300 GPa vs. 3–5 GPa). Because of this, the rate dependency of the matrix is expected to have little effect on the E_{11} elastic modulus of unidirectional FRP composites. Therefore, no significant rate dependency should be expected for the longitudinal elastic modulus of unidirectional CFRPs. A few experimental studies have aimed to characterise the longitudinal stiffness of CFRPs at different strain rates, Fig. 12, confirming, in general, the absence of any significant strain rate effects [47,98,99,105–109]. Some sources, however, did claim to have found a significant strain rate effect in the longitudinal modulus to varying degrees, [105,109], but both used test methods poorly suited for high strain rates, as discussed in Section 2.3. Al-Zubaidy et al. [105] used a drop-weight set-up for high-rate testing between 10 and 100 s⁻¹ a stiffness increase of up to 20% but did not provide any means to rule out the influence of inertial forces and/or resonance modes in the experimental set-up. Meanwhile, Zhang et al. [109], who also reported a stiffness increase of over 10%, used a servo-hydraulic test machine for loading rates up to 20 m/s, which produced increasingly noisy stress and strain measurements from which the longitudinal modulus was evaluated. Therefore, in spite of these conflicting reports, it seems reasonable to assume that the longitudinal elastic modulus of unidirectional CFRPs is largely rate insensitive. In addition, the longitudinal compression tests from Hosur et al. [110] have also been excluded from the present study as the results reported were very inconsistent, with the stiffness in the quasi-static reference tests measuring less than half the expected value and the dynamic tests showing decreasing stiffness measurements at increasing strain rates.

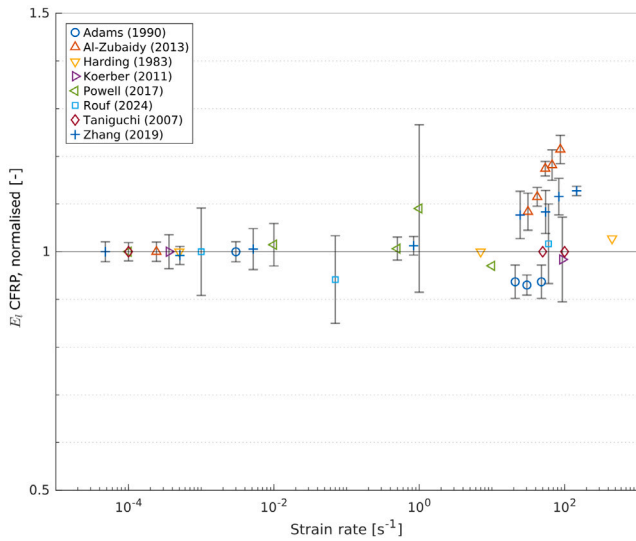


Fig. 12. Normalised longitudinal modulus for various UD CFRP materials from quasi-static to dynamic loading rates as reported in the literature [47,98,99,105–109].

Finally, even though it does not fall within the scope of the current work, it should be noted that, while unidirectional CFRPs show no strain rate dependency, many woven CFRP laminates have been shown to exhibit significant rate effects due to the role of the matrix in the non-linear elastic response of wavy fibre composites [111–114].

5. Composite strength properties

Failure of UD CFRPs can be grouped into three distinct modes: the matrix-dominated inter-fibre failure mode (IFF), covered in Section 5.1; the fibre-dominated longitudinal tensile failure mode, Section 5.2; and the longitudinal compressive, or fibre kinking, failure mode, which is influenced by both fibre and matrix properties, covered in Section 5.3.

Unlike the stiffness properties discussed in the previous section, the composite strength properties cannot be so easily derived from those of its constituents due to the complex mechanisms involved. Nevertheless, in some cases there do exist well-established failure theories that can help to relate the composite and constituent properties. These are discussed and evaluated alongside the collected literature data in the following sections. In addition, even in cases where the composite strength cannot be determined from the constituent properties, it may still be possible to relate the rate dependency of the two.

Finally, it should be noted that the yield stress and non-linear behaviour of unidirectional CFRPs are not covered in this study. However, previous work has indicated that the rate dependency of the yield stress may be similar to that of the matrix-dominated IFF strength properties [95,115].

5.1. Inter-fibre failure

Due to the significant difference in strength properties between the typical fibre and matrix materials, failure in UD FRPs under transverse loads tends to occur in the matrix and fibre-matrix interfaces in what is known as IFF. Being a matrix-dominated failure mode, the IFF strength properties can also be expected to show significant strain rate effects. However, the damage and failure mechanisms in the composite are more complex than those of the neat resins. Therefore, the relation between the composite and constituent strain rate effects may not be as simple as it was when comparing stiffness properties. Nevertheless, simplified micro-mechanics-based theories typically assume the transverse strength properties of a UD CFRP as being proportional to the strength of the matrix, with a knockdown factor to account for the effect of stress concentrations around the fibres [116].

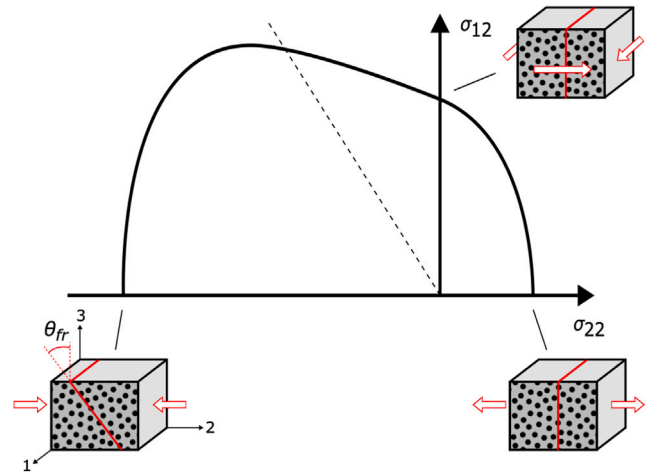


Fig. 13. Illustration of the failure envelope for IFF of a typical UD CFRP, with the effects of pressure shown by the asymmetrical shape of the envelope and the changes in the fracture plane.

Furthermore, due to Mohr–Coulomb type frictional and pressure effects on the matrix and fibre-matrix interface, IFF in UD CFRPs typically presents an asymmetrical failure envelope, illustrated in Fig. 13, where the strength tends to increase under greater transverse compression. In addition, under tensile loads, the composite is more sensitive to stress redistribution from the opening of interface cracks that would tend to remain closed in compression. As a result, transverse tension failure tends to be more brittle and may be more strongly influenced by the properties of the fibre-matrix interface regions, potentially introducing differences in the strain rate dependency between tension and compression modes.

The collected literature data on IFF strength of UD CFRPs at different loading rates are shown in Fig. 14, including transverse tension, transverse compression, and shear strength properties [23,24,39,40,92–102,104,108,117–122]. From these results, the normalised transverse tension strength properties were often slightly lower than the rest, see Gilat et al. [121], Taniguchi et al. [99] or Fletcher et al. [24]. To test the significance of any such difference in rate dependency, the CS model was fit independently to the tensile data and the compression and shear data. The results, included in Table 1, indicate that the different IFF modes may indeed follow slightly different trends, as the quality of the fit was somewhat improved when fitting the two data sets independently (RMSE = 0.148 for tension and 0.174 for compression and shear) compared to fitting all of the IFF data together (RMSE = 0.179). However, the three trends only begin to diverge significantly at strain rates above 10^3 s^{-1} and, considering the variance in the experimental data and the greater possibility of unreliable test results at higher strain rates, it is difficult to firmly conclude any inherent difference in rate dependency between the different IFF modes.

Regardless of any potential differences between transverse tension and compression strengths, the collected IFF data clearly do not remain proportional to the matrix strength reviewed in Section 3.1 as the strain rate increases, with the IFF data deviating significantly from the matrix strength trend (RMSE = 0.339).

Assuming the collected results are mostly reliable, there are several reasons why the predictions based on simplified micro-mechanics-based theories may not be valid and more sophisticated models may be necessary, taking into account the non-linear visco-plastic response of the matrix or the presence of voids and defects, for example. In addition, the effect of strain rate on the fibre-matrix interface properties is not well known and may play a significant role in the resulting composite strength under dynamic loading.

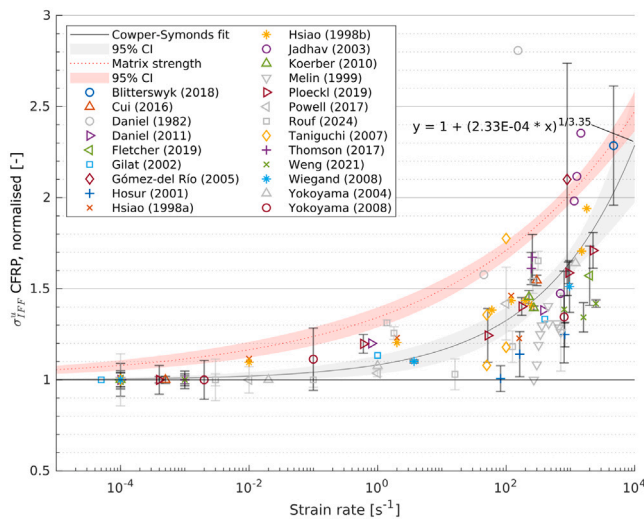


Fig. 14. Normalised IFF strength properties (transverse tension, transverse compression, and shear) for various UD CFRP materials from quasi-static to dynamic loading rates as reported in the literature [23,24,39,40,92–102,104,108,117–122]. The Cowper-Symonds (CS) parameters and 95% confidence interval are given in Table 1. Series excluded from the Cowper-Symonds fit are indicated with grey markers. The expected trend based on the matrix strength from Section 3.1 is included in red for reference.

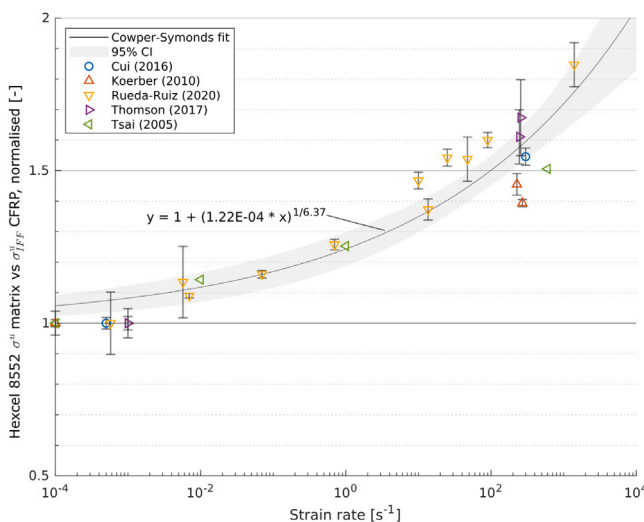


Fig. 15. Normalised IFF strength properties for Hexcel 8552 based UD FRP composites from quasi-static to dynamic loading rates as reported in the literature [38,40,95,104,118] compared to the normalised strength of the neat Hexcel 8552 matrix from Rueda-Ruiz et al. [51]. The Cowper-Symonds (CS) parameters and 95% confidence interval are given in Table 1.

Interestingly, however, when focusing on Hexcel 8552 based composite systems, for which there have been many experimental investigations in recent years, the difference between matrix and composite IFF strengths appears to be greatly reduced. Fig. 15 includes test results on neat Hexcel 8552 epoxy from Rueda-Ruiz et al. [51] and several compression and shear test results from Hexcel 8552 based composites [38,40,95,104,118]. In contrast to the previous observations from Fig. 14, here, all results appear to follow a similar trend with very good agreement (RMSE = 0.083).

Without further research on the topic, the reason for this discrepancy can only be speculated but one possible explanation may lie in the effect of defects, such as voids or disbanded fibres, which could become more critical at higher strain rates, especially under tensile loading, due to the increasingly brittle behaviour of the matrix. With the Hexcel

8552 composites being modern high-performance, toughened-epoxy prepreps, it is possible that both the presence of defects and their effect on the IFF strength are greatly reduced, allowing the matrix within the composite to reach stress levels closer to those achieved in the neat resin. Aside from additional experimental characterisation work, another promising avenue for the investigation of this potential interaction between rate dependency and material imperfections could be in the combination of rate dependent micro-mechanical modelling with the seeding of microdefects, similar to the recent work in Refs. [85, 123].

Finally, as in previous sections, it should be noted that some of the collected strength data have been excluded from the trend-fitting exercises above for various reasons. As in Section 4.1, the results from Daniel et al. [101] and Melin et al. [96] are also excluded here for the same reasons. In addition, despite presenting perfectly valid results, the transverse compression data from Yokoyama et al. [102], and the transverse tension data from Powell et al. [108] and Rouf et al. [98] have also been excluded from the non-linear regression due to the relatively high strain rate of 10^{-2} s^{-1} at which the reference quasi-static tests were performed at strain rates above. As discussed for results from Gerlach et al. [48] in Section 3.1.2, the normalised data from different sources can only be fitted to a common trend as long as the reference strain rates are reasonably close. Finally, with regard to dynamic shear strength from Cui et al. [104], which was obtained from post-processing of $\pm 45^\circ$ tension and compression test results, the strain rate from the elastic loading region, around 300 s^{-1} , has been used instead of the originally reported strain rate, around 1300 s^{-1} , which was measured after significant non-linear deformation and damage. It is possible that a similar phenomenon could have also taken place in the experiments by Ledford et al. [40], Rouf et al. [98], Taniguchi et al. [99] and Weng et al. [100], who all carried out similar $\pm 45^\circ$ tension experiments at higher strain rates. However, the reported strain rate values are assumed to be correct in the present work, since (i) no strain rate history data were reported to indicate otherwise and (ii) even if a similar correction were applied it would not drastically change the position on a semi-logarithmic plot, Figs. 14 and 15.

5.2. Fibre tensile failure

The effect of strain rate on the longitudinal tensile strength is often assumed to be practically negligible or non-existent due to the well-known rate insensitivity in the carbon fibres, see Section 3.2. However, with the thickness of a typical UD CFRP ply being hundreds of times the typical fibre diameter, the ultimate tensile stress of the composite will depend not only on the statistical distribution of individual fibre strengths but also on the ability of the polymer matrix to redistribute the stress from failed fibres to their neighbours until the remaining material can no longer carry the applied load [124]. Therefore, it would not be unreasonable to think that the rate sensitivity of the matrix may have an impact on the longitudinal tensile strength of the composite by shifting the point in the Weibull distribution of fibre strengths at which the stress from failed fibres can no longer be redistributed to the remaining intact material.

Nevertheless, most experimental studies over the years have shown no such strain rate sensitivity in the fibre tensile strength of UD CFRPs for strain rates above 1000 s^{-1} , Fig. 16 [47,99,105,106,108,109,122, 125,126]. While a few sources have reported a significant increase in fibre tensile strength at strain rates above $10 - 100 \text{ s}^{-1}$ [105,109,125], they all have in common that they used servo-hydraulic or drop-weight test set-ups, which are susceptible to inertial effects at high strain rates. Meanwhile, experimental studies using the SHB method reported strength values in line with the quasi-static values for similar and even higher strain rates [99,106,108]. Therefore, it seems most likely that the strength increase observed in Refs. [105,109,125] resulted from failing to consider dynamic effects in the force measurements at high strain rates. Because of the stark difference between the results

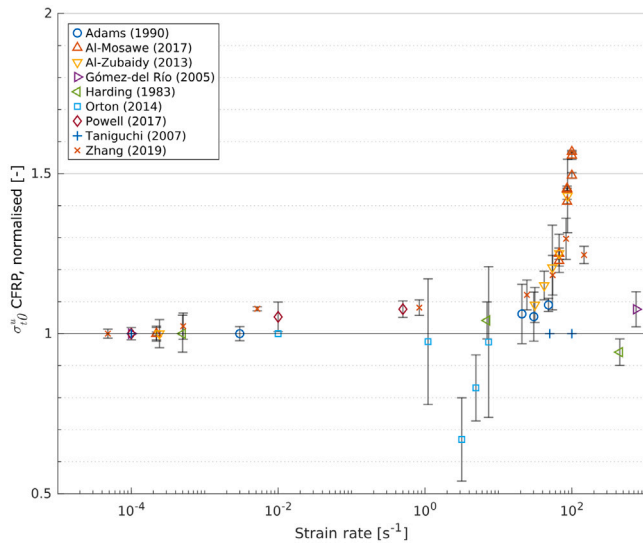


Fig. 16. Normalised fibre tensile strength properties for various UD CFRP materials from quasi-static to dynamic loading rates as reported in the literature [47,99,105,106,108,109,122,125,126].

obtained from different test set-ups, this failure mode serves as a good benchmark to justify the use of the SHB method for strain rates above 100 s^{-1} . The reduction in tensile strength observed for certain loading rates in the results of Orton et al. [126] can also be ignored, as they were attributed to experimental scatter from the hand lay-up laminate manufacture and stress concentrations at the specimen fixtures.

5.3. Fibre compressive failure

In contrast to the longitudinal tension mode, under fibre compression, the strength of unidirectional CFRPs has been shown to increase significantly with the loading rate. This may seem to contradict the lack of strain rate effects on the carbon fibre properties, as discussed in Sections 3.2 and 5.2. However, the failure mechanisms of the fibre compression mode also strongly depend on the behaviour of the polymer matrix. For longitudinal compression loads, the failure mode tends to be in the form of kinking or micro buckling of the reinforcing fibres, triggered by initial fibre waviness or misalignment. This initial misalignment causes local shear stresses within the material even under uniaxial compression, which drives further fibre rotation up to the point of local instability. As a consequence, the behaviour of the composite in shear, which is rate dependent due to the polymer matrix, strongly influences the maximum load under longitudinal compression.

While longitudinal compression failure can also occur in the form of fibre crushing instead of kinking, this usually only happens when the strength of the fibres is lower than the buckling instability load, which is not the case for the high-strength carbon fibre composites investigated in the present work, or if the material is over-constrained [127].

Because of the various complex mechanisms involved, adequate test design for the investigation of fibre buckling or kinking failure is a non-trivial matter even at QS loading rates, as the longitudinal compression specimens can be extremely sensitive to stress concentrations and even very minor specimen misalignment. On the other hand, if the specimen is not sufficiently constrained, it may split apart or rotate out of place before the buckling load can be reached. For dynamic loading, this becomes even more challenging due to the additional constraints on the mass of the test fixtures and the size of the specimens.

As an example, the results from Yokoyama et al. show a drop in strength from QS to medium rate before increasing again at high strain rates [120]. However, while the fibre kinking failure mode was achieved under QS loading, a splaying failure mode was observed in

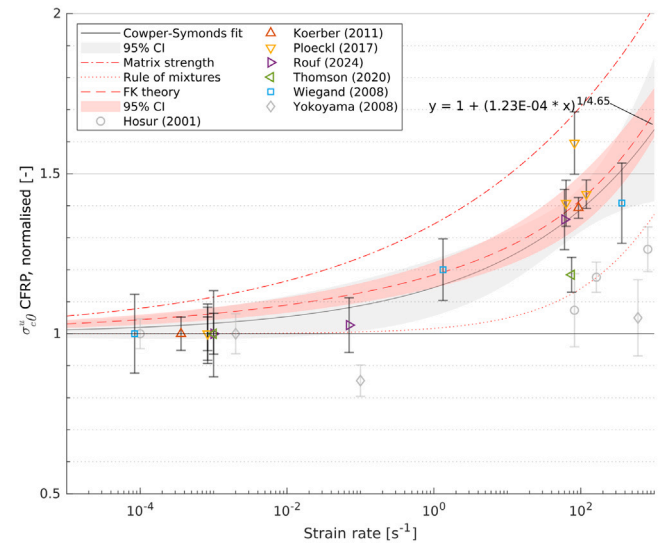


Fig. 17. Normalised fibre compression strength properties for various UD CFRP materials from quasi-static to dynamic loading rates as reported in the literature [98,107,110,119,120,128,129]. The Cowper-Symonds (CS) parameters and 95% confidence interval are given in Table 1. Series excluded from the Cowper-Symonds fit are indicated with grey markers. The expected trend based on the Budiansky and Fleck Fibre Kinking theory and assumed micromechanical parameters is included in red for reference.

the dynamic test specimens. Therefore, the strengths reported under quasi-static and dynamic loads may not be directly comparable. Several studies have reported similar difficulties in reproducing the desired kinking failure mode in UD CFRP specimens, especially under high-rate loading. To address this issue, Ploekl et al. [128] and Thomson et al. [129] proposed the use of multi-axial lay-ups, such as $[(0_a/90_b)_n]_s$ cross-ply or quasi-isotropic, to help constrain the longitudinal plies under dynamic loading conditions and showed that kinking failure could be produced more reliably. Similar issues with the failure mode changing between splitting and crushing modes were reported in the longitudinal compression tests from Hosur et al. [110], which were also excluded from the analysis in Section 4.2 for the inconsistent stiffness results presented.

Excluding these two sources where the failure mode was not consistently reproduced across quasi-static and dynamic loading rates, a clearly increasing trend in the longitudinal compressive strength can be observed, Fig. 17 [98,107,110,119,120,128,129].

This can be explained by the fibre kinking theory originally proposed by Budiansky and Fleck [127] and further developed by Pimenta et al. [130,131] using advanced micro-mechanical modelling. While there exist different analytical formulations of the theory, they all essentially describe the critical fibre buckling load as a function of the initial fibre waviness/misalignment and the non-linear shear properties of the composite in a similar form to the original from [127]:

$$\sigma_{c0}^u = \frac{G}{1 + nk^{1/n} \left(\frac{\bar{\phi}/\gamma_Y}{n-1} \right)^{(n-1)/n}} \quad (5)$$

Where G is the shear modulus of the composite, $\bar{\phi}$ is the initial fibre misalignment angle, $\gamma_Y = \tau_Y/G$ is the yield strain, and n and k are the Ramberg-Osgood parameters describing the non-linear shear response of the composite [132]:

$$\gamma = \frac{\tau}{G} \left[1 + k \left(\frac{\tau}{\tau_Y} \right)^{(n-1)} \right] \quad (6)$$

As a simplification, in order to consider the strain rate effects on the non-linear stress-strain curve and the fibre kinking strength, the n

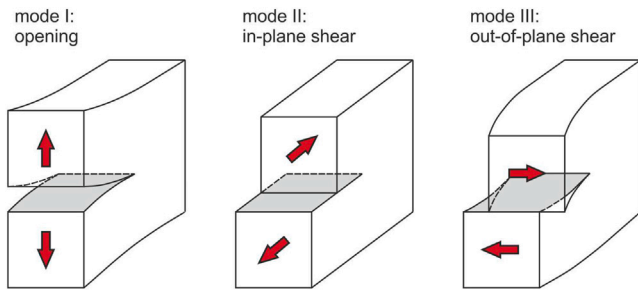


Fig. 18. Set of fracture modes to describe crack propagation within the framework of fracture mechanics.

Source: Adapted from [42].

and k parameters in the Ramberg–Osgood relation can be assumed to remain constant, while G and τ_Y are considered rate dependent [129].

Therefore, according to the fibre kinking theory, the fibre compression strength of the composite under dynamic loading becomes a function of the rate dependency of the shear properties of the laminate, G and τ_Y , which can be related to the rate dependent behaviour of the matrix, as discussed in Sections 4.1 and 5.1, respectively.

Using the above expressions and representative values of $n = 5.8$, $k = 0.43$ and $\bar{\phi} = 0.003$ from Ref. [129], the fibre kinking failure of a typical CFRP composite can be related to the strain rate dependency of the matrix strength from Section 3 and the composite stiffness calculated with the rule of mixtures in Section 4.1.

The resulting prediction, illustrated in Fig. 17 by the dashed red line, is in remarkably good agreement (RMSE = 0.097) with the best fitting CS model (RMSE = 0.083), which should give confidence in the fibre kinking theory as a reasonable explanation for the rate dependency of the longitudinal compression strength of UD CFRPs despite the low amount of reliable experimental data available.

6. Composite fracture properties

At the meso-scale, failure of a composite structure consisting of UD-reinforced plies is associated with four distinct failure modes. Those are IFF as well as fibre tensile and compressive failure within a lamina, which have been reviewed in Section 5, and delamination between adjacent plies. Each of these failure modes, in turn, can be associated with three different modes of fracture, see Fig. 18 [42].

The high-rate behaviour of each of those failure and fracture mode combinations have been investigated to vastly different degrees. Overall, the rate dependency of delamination has been studied for several decades now, whereas similar studies focusing on the critical ERRs in fibre direction only started to emerge comparatively recently. Furthermore, to the authors' best knowledge at the time of this review, neither of the mode II ERRs in the fibre direction, the Mode III ERRs for any of the failure modes, nor any of the modes pertaining to IFF have been investigated at all under high-rate loading. Consequently, this section covers the rate dependency of the ERRs associated with delamination under mode I and II in Section 6.1, and the mode I ERRs for failure in fibre tensile (G_{IC}^{f+}) and compressive (G_{IC}^{f-}) direction in Sections 6.2 and 6.3, respectively. Depending on the micro-mechanisms involved in the fracture process, a particular fracture mode may exhibit a more or less pronounced R-curve effect. Any potential such effect, however, has been left out of the scope of the present work and only steady-state propagation values have been reviewed. Furthermore, any interactions between fracture modes have also been considered beyond the scope of this review.

In order to provide a rough sense of orientation, it can be stated that the ERRs involving fibre failure, also known as translamina failure, are typically several orders of magnitude larger than those which do not, e.g. $G_{IC}^{f+} = 147 \text{ kJ/m}^2$ [133] or $G_{IC}^{f-} = 61 \text{ kJ/m}^2$ [134] compared to

$G_{IC}^{delam} = 0.196 \text{ kJ/m}^2$ or $G_{IC}^{IFF} = 0.188 \text{ kJ/m}^2$ [135] for IM7/8552 under QS loading. Furthermore, the initiation ERRs pertaining to IFF tend to be fairly similar to their delamination counterparts [135,136]. However, the differing extents to which certain toughening mechanisms develop at the micro-scale, such as fibre bridging, may cause the ERRs of those two failure modes to exhibit differing levels of the R-curve effect for crack propagation [135].

6.1. Delamination

Historically, extensive research has been carried out on the out-of-plane failure behaviour of composites, with many studies also addressing its loading rate sensitivity. Obtaining measures of strain rate or crack tip opening rates in the context of ERR characterisation tests requires employing extensive analytical calculations [137,138]. Therefore, most of the available studies present the rate dependent ERR changes in terms of loading rate as applied via the bulkhead or impactor instead of providing a measure of the associated strain rate. Hence, only sources where the loading rate was given for each experiment have been included in the present study, even though the respective experimental set-ups and metrics may vary from case to case.

The collected data for pure mode I and pure mode II loading have been grouped and reviewed separately in the following two subsections. Combined loading effects have not been covered here to the same extent due to the added difficulty in comparing data with different degrees of mode mixity and considering the lack of consensus already observed in the pure mode I and mode II results. However, a few exemplary publications that investigated the influence of strain rate on the ERR under mixed-mode (I/II) loading are mentioned for the sake of completeness [139–142].

Mode I

While the characterisation of the interlaminar mode I fracture toughness is standardised for quasi-static loading [143,144], there are no such standards for its counterpart under dynamic loading. This has resulted in a range of different test configurations as well as specimen geometries being employed across published studies. A comprehensive overview of those has been compiled by May [7].

Following the quasi-static test standard, many studies employed a double-cantilever-beam (DCB) specimen design for testing at elevated loading rates on screw-driven or servo-hydraulic testing machines [137, 145–149]. The classical DCB test configuration, however, comes with the inherent drawback of the crack opening losing its symmetry due to inertial effects under highly dynamic loading conditions. De Verdiere et al. [150] confirmed this by testing DCB specimens employing a drop tower set-up. For loading rates above 3 m/s, the opening mode changed from mode I to mixed mode (I/II) loading. To overcome this limitation, a wedge-loaded DCB (WDCB) configuration has been employed in several studies. With this set-up, the load is introduced into the specimen by pushing a wedge-shaped impactor between the sample's arms, which can be conveniently realised employing a drop tower or modified Hopkinson bar set-up [140,146,151]. One drawback of the WDCB testing configuration, however, is the necessity of having to account for the friction between wedge and specimen arms when deducing the ERR based on the experimentally obtained data [7].

A wide range of approaches to data acquisition and reduction can be found in the literature. The classical quasi-static analysis recommended in the ASTM test standard uses an optical measurement of the crack length and the load cell's force signal to then obtain the ERR based on beam theory [143]. Employing this approach in a high-rate loading context, however, has several drawbacks, such as the need for a high-speed camera for the already challenging optical measurement of crack propagation and possible inertial effects in the load cell's force signal. In recent publications, the authors overcame this latter issue by replacing the direct measurement of force with its calculated equivalent based on the optical measurement of the specimen's deflection [140,151].

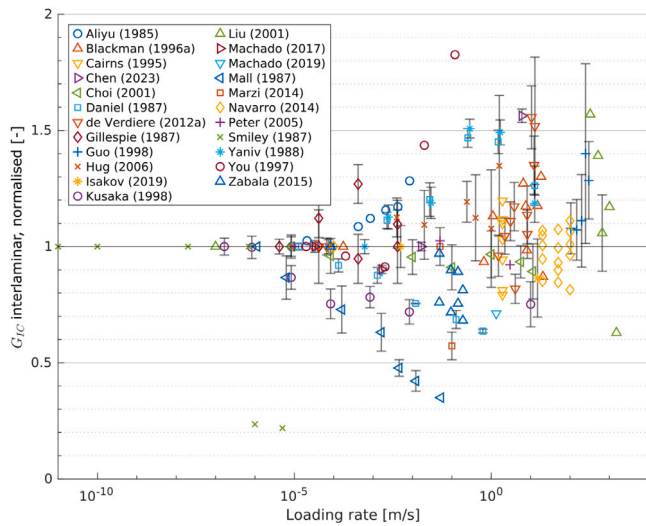


Fig. 19. Normalised mode I ERRs for various UD-reinforced CFRP materials from quasi-static to dynamic loading rates as reported in the literature [41,139,140,145–149,151,153–161].

In addition, other recent contributions have proposed using a guided fixture with adequate data reduction schemes or considering crack tip root rotations to decrease experimental variability [41,152].

Fig. 19 presents the collected data for mode I ERR of UD-reinforced CFRP material systems from sources where the loading rate metric was available in terms of cross-head displacement [41,139,140,145–149,151,153–161]. Due to the lack of a clear trend, no fit to the Cowper–Symonds equation is provided.

Mode II

Similar to mode I, the rate dependency of the mode II ERR has been studied for various combinations of CF types and epoxy resin systems. All quasi-static and medium-rate tests up to loading rates in the range of 0.1 m/s were carried out on servo-hydraulic or electro-mechanical testing machines. High-rate tests were either performed on servo-hydraulic, drop weight testing machines or SHB set-ups employing loading rates of up to approximately 10 m/s [138]. Most of the studies used an end-notched flexure (ENF) specimen geometry based on the ASTM standard for quasi-static mode II delamination testing [162,163], two employed an end-loaded-split (ELS) configuration [139,150], and one applied a Transverse Crack Tension (TCT) configuration [164]. Similar to May's review of mode I ERR testing, May et al. [10] published an extensive overview on the high-rate characterisation of the mode II ERR focused on the different experimental techniques and recommended the TCT test as one of the more promising approaches thanks to the simplicity of the test set-up and post-processing of the force measurements. Using the high-rate TCT test, Hahn et al. [164] found the mode II fracture toughness to be at least 50% higher than the quasi-static value. However, the mechanisms behind the reported increase in fracture toughness remain unclear. Recently, attempts were also made to improve the experimental accuracy of high-rate mode II ERR characterisation solely acquiring data optically or taking into account rotations of the crack tip root [140,165].

The review from May et al. [10] only gave a qualitative comparison of the rate dependency trends in the earlier literature. Here, the data from previously reviewed publications, as well as more recent studies, have been collected and normalised to allow a more quantitative comparison of the influence of strain rate. The collection of literature data for critical ERR under this failure mode, based on CF/epoxy systems, is shown in Fig. 20 [138–140,150,154,164,166–172].

For both mode I and II, there is a clear lack of consensus on the effects of loading rate on the interlaminar ERR, with different sources

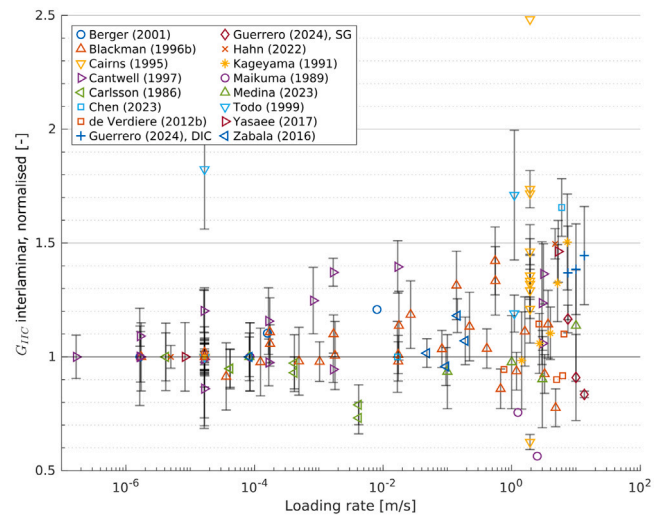


Fig. 20. Normalised mode II ERR for various UD-reinforced CFRP materials from quasi-static to dynamic loading rates as reported in the literature [138–141,150,154,164,166–172].

reporting anywhere from strongly positive to strongly negative trends. However, for mode I there appear to be a greater number of reports of significantly negative rate dependency than for mode II, which could be attributed to changes to ductility and size of the fracture process zone being more pronounced under tension than shear. A similar lack of consensus and reports of negative rate dependency due to reduced ductility in the mode I ERR have also been seen for the neat matrix in Section 3.1. As with the matrix ERR, some of the differing reports here may also be simply related to material differences, such as damage, plastic deformation, or other energy dissipation mechanisms away from the crack [13]. At the same time, it is also likely that some of these discrepancies stem from inconsistencies in the experimental techniques and/or data reduction methods, which can lead to significant uncertainty in the interpretation of the results. For instance, in the work from Guerrero et al. [141], the rate dependency of the mode II ERR changed from a positive to a negative trend depending on whether a digital image correlation based or a strain-gauge based data reduction method was employed. In addition, the average quasi-static mode I ERR reported for UD AS4/3501-6 varied between 0.15–0.20 kJ/m² [137,145,153,156,173,174] and the average quasi-static mode II ERR for UD AS4/PEEK varied between 1.81–2.42 kJ/m² [139,166–168,170]. This variation across quasi-static ERR values is comparable to the variation between quasi-static and high-rate ERR values for the same materials and highlights the need for standardisation of ERR characterisation methods, regardless of the loading rate, as well as deeper investigation into other possible contributing factors beyond the loading rate alone.

6.2. Fibre tension loading

The first study into the rate dependency of G_{IC}^{f+} was undertaken by McCarroll [175], who characterised IM7/8552 on a servo-hydraulic testing machine employing a compact tension (CT) specimen design. Using compliance calibration, G_{IC}^{f+} was found to be practically rate insensitive ($G_{IC,QS}^{f+} = 74.9$ kJ/m², $G_{IC,HR}^{f+} = 67.8$ kJ/m²). Hoffmann et al. [176] characterised the same material system, adopting the same specimen design. High-rate testing was performed on a modified SHTB system. Based on the area method G_{IC}^{f+} was observed to exhibit a significant drop under an elevated loading rate ($G_{IC,QS}^{f+} = 196.2$ kJ/m², $G_{IC,HR}^{f+} = 82.0$ kJ/m²). Kuhn et al. [177] employed a range of double-edge notched tensile (DENT) specimens to test IM7/8552 on an SHTB set-up. Using the size-effect law, G_{IC}^{f+} was found to increase under

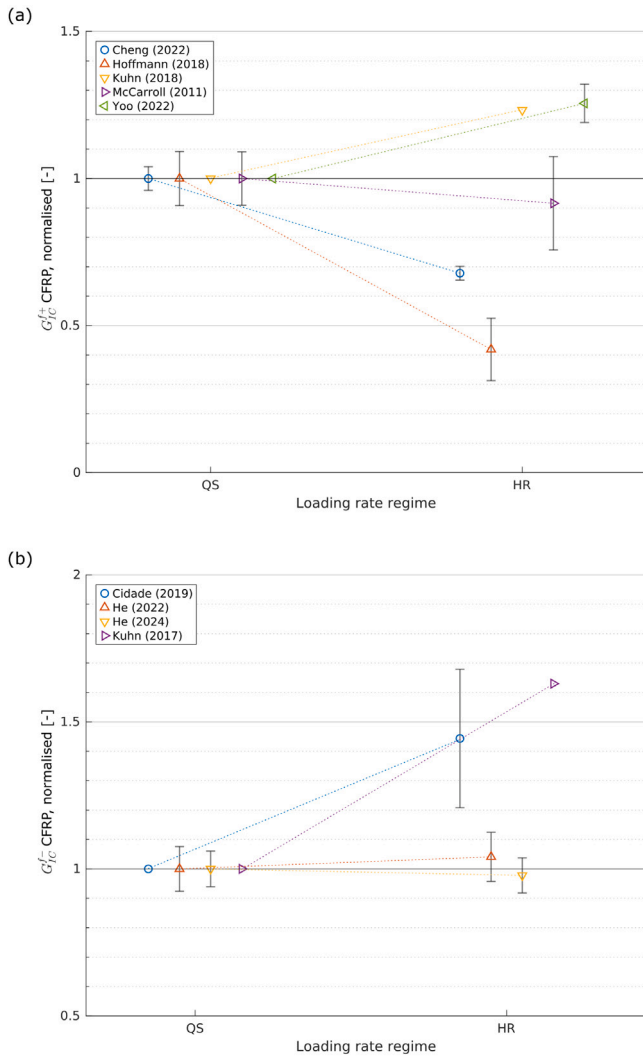


Fig. 21. Normalised longitudinal ERR for various UD CFRP materials from quasi-static to dynamic loading rates as reported in the literature, with (a) the ERR in longitudinal tension, G_{IC}^{f+} , [175–179], and (b) the normalised ERR in longitudinal compression, G_{IC}^{f-} , [180–183].

high-rate loading ($G_{IC,QS}^{f+} = 195.5 \text{ kJ/m}^2$, $G_{IC,HR}^{f+} = 241.0 \text{ kJ/m}^2$). Yoo et al. [178] also tested a range of DENT specimens made of IM7/8552, employing a servo-hydraulic testing machine in conjunction with a slack adapter for high-rate testing. Data reduction was based on the size-effect law, finding G_{IC}^{f+} to increase under an elevated loading rate ($G_{IC,QS}^{f+} = 281.5 \text{ kJ/m}^2$, $G_{IC,HR}^{f+} = 371.8 \text{ kJ/m}^2$). Cheng et al. [179] characterised T700/LT03 A using a CT specimen design. High-rate testing was performed on a bi-directional electro-magnetic SHTB system. Obtaining the J-integral via full-field strain analysis, G_{IC}^{f+} was observed to exhibit a negative loading rate trend ($G_{IC,QS}^{f+} = 166.5 \text{ kJ/m}^2$, $G_{IC,HR}^{f+} = 112.9 \text{ kJ/m}^2$). As is evident from this brief review, there is currently no consensus regarding the loading rate trend of G_{IC}^{f+} . A representation of the diverging trends is given in Fig. 21(a) [175–179]. As there is no consistent loading rate metric across the literature sources, the individual data points are simply grouped according to their loading rate regime.

While there is a clear discrepancy in the reported effects of loading rate on the critical ERR under longitudinal tension, it must be noted that discrepancies also exist for the seemingly straightforward and well-established characterisation in the quasi-static loading domain.

For instance, the range of values for $G_{IC,QS}^{f+}$ for the popular material system IM7/8552 across the sources discussed above stretches from 74.9 kJ/m^2 [175] to 281.5 kJ/m^2 [178]. Consulting additional sources [184–189] having characterised G_{IC}^{f+} for the identical material system exclusively under QS loading also does not aid in painting a clearer picture. Moreover, none of those sources suffered from excessive experimental scatter. This obvious lack of reproducibility under seemingly straightforward loading conditions clearly indicates that key influences and mechanisms have not yet been fully understood.

Beyond the loading rate trend and reproducibility issue, it must be noted that the obtained value for G_{IC}^{f+} exhibits further dependencies. Most notably, studies have shown the value of G_{IC}^{f+} to increase with ply thickness [190,191], which also extends to ply blocks, i.e. placing plies of identical orientation next to each other, thus effectively creating a thicker ply [192]. It is, however, currently still debated if this increase can be mainly attributed to mechanisms associated with the actual macroscopic fracture plane [190,192], or to other failure modes accompanying the translaminal fracture [191,193]. In addition, the value of G_{IC}^{f+} has also been shown to depend, albeit to a much lesser extent, on the total laminate thickness [192], data reduction methodology [194], and notch tip radius [187].

Recent numerical modelling work from Sommer et al. [195] also showed that the change in G_{IC}^{f+} between quasi-static and dynamic loading reported by Hoffman et al. [176] could be attributed in large part to secondary damage mechanisms away from the main translaminal crack, such as non-linear deformation and matrix damage, which are matrix-dominated thus more sensitive to the applied loading rate. In a similar manner, it is possible that at least some of the observed discrepancies across different literature sources above may also be linked to unaccounted differences in the extent of secondary dissipation mechanisms away from the macroscopic crack. However, further research is required to substantiate this hypothesis.

6.3. Fibre compression loading

The first investigation into the rate dependency of G_{IC}^{f-} was undertaken by Kuhn et al. [183], who characterised a range of double-edge notched compression (DENC) specimens made of IM7/8552, employing an SHPB set-up for testing under an elevated loading rate. Using the size-effect law, G_{IC}^{f-} was found to exhibit an increase under high-rate loading ($G_{IC,QS}^{f-} = 101.6 \text{ kJ/m}^2$, $G_{IC,HR}^{f-} = 165.6 \text{ kJ/m}^2$). Using a similar specimen design, material, and high-rate testing set-up, Cidade et al. [180] measured the J-integral to determine G_{IC}^{f-} under dynamic loading, reporting its value to fall in the range of $120 - 180 \text{ kJ/m}^2$. He et al. [181,182] characterised T700/LT03 A employing a compact compression (CC) specimen design. High-rate testing was performed on a bi-directional electro-magnetic SHPB system. Based on obtaining the J-integral, the initiation value of G_{IC}^{f-} was found to be practically rate insensitive ($G_{IC,QS}^{f-,init} = 122.2 \text{ kJ/m}^2$, $G_{IC,HR}^{f-,init} = 127.2 \text{ kJ/m}^2$ in Ref. [181] and $G_{IC,QS}^{f-,init} = 117.4 \text{ kJ/m}^2$, $G_{IC,HR}^{f-,init} = 114.8 \text{ kJ/m}^2$ in Ref. [182]), whereas the ERR during crack propagation remained higher under QS loading than under HR loading.

A direct comparison of these rate dependency trends is presented in Fig. 21(b) [180–183], where it can be seen, even though there are few sources available, that the reported effects of loading rate vary significantly. As in the previous section, there is no consistent loading rate metric across the literature sources, so the horizontal axis only indicates the loading rate regime, either quasi-static or high-rate, for each data point. It should be noted however, that for Cidade et al. [180], the associated quasi-static value was not provided and has been assumed to be identical to the value from Kuhn et al. [183], who used the same material and specimen design. In addition, the trend data from He et al. [181,182] refer to the initiation value of G_{IC}^{f-} .

Analogously to G_{IC}^{f+} , the characterisation of G_{IC}^{f-} suffers from poor reproducibility even under seemingly straightforward quasi-static loading conditions. Complementing the data detailed above with sources

having characterised IM7/8552 only quasi-statically, a remarkably broad range of values for G_{IC}^{f-} for this material system becomes apparent: $G_{IC}^{f-.init} = 25.9 \text{ kJ/m}^2$ [196], $G_{IC}^{f-.init} = 47.5 \text{ kJ/m}^2$ [186], $G_{IC}^{f-.prop} = 61.0 \text{ kJ/m}^2$ [197], $G_{IC}^{f-.prop} = 101.6 \text{ kJ/m}^2$ [198], $G_{IC}^{f-.init} = 106.3 \text{ kJ/m}^2$ [184]. Again, analogously to G_{IC}^{f+} , further research is required to shed light on the root causes behind these discrepancies, not only in relation to the effects of loading rate but also, more fundamentally, in relation to the significant variation reported under similar quasi-static loading conditions.

7. Conclusions

A comprehensive review of publicly available research on the strain rate dependency of high-performance unidirectional CFRPs and their constituent materials has been presented. The study reveals common trends in the rate dependency of stiffness and strength properties of UD CFRPs and explores their relationship to the properties of typical carbon fibres and polymer matrix materials. The main findings are summarised below.

First, the behaviour of typical CFRP constituents has been reviewed in Section 3. For carbon fibres, no significant rate dependency has been found for either the stiffness or strength properties in Section 3.2. Meanwhile, in Section 3.1, the elastic moduli and strength properties of various polymer matrix materials have been shown to increase with strain rates, following an exponential trend that can be satisfactorily captured by the Cowper–Symonds equation [33]. In addition, the scaling of both strength and stiffness properties with the strain rate was found to be reasonably consistent across the various polymer matrix materials in the literature. The critical energy release rates, on the other hand, showed no consistent trend, with many conflicting reports found in the literature. Moreover, the characterisation experiments tended to suffer from poor repeatability, and, in many cases, any potential difference between quasi-static and dynamic test results was overshadowed by the considerable experimental scatter. While the reasons for this are not fully understood, the fact that brittle fracture can often become a dynamic process even under quasi-static loading conditions may be an important factor.

Second, the effects of strain rate on the elastic, strength, and fracture properties of various UD CFRP composites have been reviewed in Sections 4, 5, and 6, respectively. For each case, the observed trends, have been compared to those of the fibre and matrix materials from Section 3 to assess the relationship between the rate dependency at the constituent and composite scale.

In the matrix-dominated transverse and shear directions presented in Section 4.1, the elastic moduli of the various reviewed UD CFRPs showed a similar exponential trend across different materials and sources. In addition, the reviewed data are shown to be in good agreement with predictions based on the rule of mixtures when assuming rate dependency of the matrix properties to follow the trend obtained in Section 3.1. As with the individual fibre properties, the literature data on the fibre-dominated longitudinal stiffness properties of UD CFRPs in Section 4.2 were generally in agreement in terms of the lack of any significant strain rate effect, with a few exceptions attributed to the test methods used.

In Section 5, the rate dependency of the strength properties associated with IFF, fibre tension, and fibre compression failure modes have been reviewed. In Section 5.1, the IFF strengths showed a relatively consistent trend across the reviewed CFRP materials, which, again, was shown to fit the Cowper–Symonds exponential model. In addition, while simplified micromechanics theories tend to assume the transverse strength properties to be proportional to the strength of the matrix, at higher strain rates, the reviewed composite IFF strength did not appear to remain proportional to the matrix strengths from Section 3.1. Although further research on this topic is required, some possible reasons for this include the influence of the fibre-matrix interface properties or the presence of material defects and imperfections leading

to earlier failure at higher loading rates. Interestingly, when narrowing down the scope to more recent studies on the Hexcel 8552 epoxy and its composites, this difference between the rate effect on the matrix strength and the composite IFF strength almost entirely disappeared. If the above hypotheses hold true, this may point to the greater reliability of modern prepreg laminates and high-rate testing methods.

The effect of strain rate on the fibre-dominated fibre tension and fibre compression failure modes have been reviewed in Sections 5.2 and 5.3, respectively. As expected, the fibre tensile strength generally showed no significant rate dependency. However, a few sources did report a pronounced increase in the fibre tensile strength under dynamic loading, which has been attributed to the experimental techniques used for high-rate testing. Conversely, under longitudinal compression, the strength of UD CFRPs is shown to be significantly rate dependent across multiple sources and material systems in the literature. In addition, taking into account the difficulties in dynamic experimental characterisation of this failure mode, the results from various sources obtained using different material systems and specimen designs appear to be relatively consistent. The well-established fibre kinking theory introduced by Budiansky and Fleck, which relates the failure in longitudinal compression to the role of the matrix in supporting the reinforcing fibres, is shown to capture this rate dependency of the fibre compressive failure as a function of the rate dependent matrix properties with reasonable accuracy considering the many uncertainties involved.

Finally, in Section 6, a review of the strain rate dependency of the fracture toughness and critical energy release rate properties associated with delamination, fibre tension and fibre compression failure modes has been presented. However, in all cases, there was a general lack of consensus across the collected literature data, including very poor reproducibility even for the same material systems tested under similar quasi-static loading conditions. This may indicate a need to consider other contributing factors besides the loading rate, such as additional dissipation mechanisms, in the experiment design and data-reduction processes.

A summary of the findings for each of the mechanical properties reviewed throughout the manuscript, including the Cowper–Symonds fit parameters where applicable, is given in Table 1. Aside from providing a concise summary of the present findings, the parameters in Table 1 may also be used to model the rate dependency of UD CFRP properties in cases where a full characterisation of the material rate dependency is not available.

Overall, a few main conclusions can be drawn regarding the rate dependency of both, the constituents and their corresponding UD CFRP composites:

1. The rate dependency of strength and stiffness properties appears to be consistent across different sources and material systems. In addition, various homogenisation theories and failure criteria, i.e. the rule of mixtures or the fibre kinking theory, have been shown capable of predicting the rate dependency of the corresponding composite properties with reasonable accuracy.
2. Both at the constituent and the composite scales, there is a lack of consensus regarding the rate dependency of fracture properties, which highlights the need for standardisation of experimental characterisation techniques, not only for dynamic but also for quasi-static loading rates.
3. It has been found that, even in recent years, high-rate stiffness and strength characterisation studies still frequently used test methods ill-suited for high-rate loading, often leading to incorrect conclusions. This, again, highlights the need for standardisation of high-rate experimental techniques for material characterisation. Based on the reviewed data, the following recommendations can be made:

- (a) Servo-hydraulic machines, although capable of greater loading rates, should be used with special care for the

Table 1

Summary of reviewed strain rate dependency on different mechanical properties of UD CFRPs and constituents. Where applicable, the best fit Cowper–Symonds parameters and their respective confidence intervals are given.

Property	Trend	Cowper–Symonds parameters, Eq. (1)	
		K (95% CI)	n (95% CI)
Matrix Stiffness	Positive	4.56E–04 (3.69E–04, 5.43E–04)	2.09 (1.53, 2.66)
Matrix Strength	Positive	1.15E–03 (7.72E–04, 1.54E–03)	6.32 (5.44, 7.19)
Matrix Mode I Energy Release Rate	No consensus	–	–
Matrix Mode I Fracture Toughness	No consensus	–	–
Fibre Stiffness	None	–	–
Fibre Strength	None	–	–
CFRP Transverse Stiffness	Positive	1.09E–04 (5.54E–07, 2.17E–04)	2.28 (1.27, 3.30)
CFRP Longitudinal Stiffness	None	–	–
CFRP IFF Strength (All)	Positive	2.33E–04 (9.39E–05, 3.71E–04)	3.35 (2.36, 4.34)
CFRP IFF Strength (Compression & Shear)	Positive	1.07E–04 (–2.19E–05, 2.36E–04)	4.82 (3.00, 6.64)
CFRP IFF Strength (Tension)	Positive	3.58E–04 (1.93E–04, 5.22E–04)	2.31 (1.43, 3.19)
Hexcel 8552 Matrix and CFRP IFF Strength	Positive	1.22E–04 (–2.59E–05, 2.70E–04)	6.37 (4.71, 8.03)
CFRP Fibre Tension Strength	None	–	–
CFRP Fibre Compression Strength	Positive	1.23E–04 (–2.20E–04, 4.67E–04)	4.65 (1.87, 7.43)
CFRP Mode I Energy Release Rate	No consensus	–	–
CFRP Mode II Energy Release Rate	No consensus	–	–

characterisation of CFRPs at high loading rates, e.g. above 10 s^{-1} , unless the reliability of the force measurements can be verified. Some techniques that can be used to obtain valid results at higher strain rates include pre-accelerating and/or minimising the mass of any moving parts, such as specimen fixtures or load cells, and inertia compensation in post-processing. However, dynamic equilibrium should always be ensured when doing so, as discussed in Section 2.3 [28].

- (b) Even when using specially designed equipment, such as SHB systems, issues related to inertia and lack of force convergence throughout the specimen can become unavoidable at high enough strain rates, so special care should always be taken when testing at strain rates above a few hundred s^{-1} .
- (c) As a result, novel experimental techniques, such as the Image-Based Inertial Impact technique, may be more suitable for material characterisation at very high strain rates.

In addition, a few key areas for future research have been identified that may help to resolve some of the unanswered questions on the topic of rate dependency of UD CFRP composites:

1. Combined with micro-mechanical modelling, reliable characterisation of the fibre-matrix interface properties, especially under elevated loading rates, could prove particularly useful to explain the rate dependency of matrix-dominated composite strength properties.
2. The effect of loading rate on the fracture properties remains unclear for both, neat matrix and composite failure modes. However, based on the poor reproducibility even under quasi-static loading, it is possible that some of the conflicting reports may be influenced by other contributing factors beyond the loading rate alone, such as energy dissipation from secondary deformation and/or damage mechanisms.
3. Lastly, it has been discussed that even SHB tests may be not be well suited for characterisation of certain material properties at strain rates above a few hundred s^{-1} . However, much of the reviewed strength and stiffness data for strain rates over 1000 s^{-1} was obtained using the SHB technique. Therefore, it may be worth revisiting some of these results in the future using alternative experimental techniques, such as the IBII.

Finally, although the scope of this work was limited to high volume fraction, epoxy matrix UD CFRP materials, the findings on the relationship between the composite and constituent properties may also be applicable to other material combinations, such as glass-fibre

reinforced polymer composites, or other environmental effects, such as temperature, in the sense that the composite properties may be reliably derived from the properties of their constituents.

CRediT authorship contribution statement

D. Thomson: Writing – review & editing, Writing – original draft, Visualization, Project administration, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **M. Ploeckl:** Writing – review & editing, Writing – original draft, Visualization, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **J. Hoffmann:** Writing – review & editing, Writing – original draft, Visualization, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **M. Lißner:** Writing – review & editing, Writing – original draft, Visualization, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **C. Pohl:** Writing – review & editing, Writing – original draft, Visualization, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **G. Quino:** Writing – review & editing, Writing – original draft, Visualization, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **K. Ramakrishnan:** Writing – review & editing, Writing – original draft, Visualization, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **M. Toenjes:** Writing – review & editing, Writing – original draft, Visualization, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **H. Cui:** Writing – review & editing, Supervision, Conceptualization. **N. Petrinic:** Writing – review & editing, Supervision, Conceptualization.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Dedicated to the memory of Prof. Nik Petrinic (1964–2025).

Appendix A. Supplementary data

Supplementary material related to this article can be found online at <https://doi.org/10.1016/j.compositesa.2025.108773>.

Data availability

The data used in this article has been made available in the following online repositories, along with interactive HTML versions of all scatter plots [199,200]:

A review of the effect of loading rate on the mechanical properties of unidirectional carbon fibre reinforced polymer composites (Reference data) (Mendeley Data)

RDBC (reference data and interactive plots) (GitHub)

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