



Novel Photomechanics Approaches to Determine the Intra- & Inter-laminar Properties of Composites at High Rates of Strain

Fabrice Pierron
UNIVERSITY OF SOUTHAMPTON

04/23/2019
Final Report

DISTRIBUTION A: Distribution approved for public release.

Air Force Research Laboratory
AF Office Of Scientific Research (AFOSR)/ IOE
Arlington, Virginia 22203
Air Force Materiel Command

REPORT DOCUMENTATION PAGE				Form Approved OMB No. 0704-0188	
The public reporting burden for this collection of information is estimated to average 1 hour per response, including the time for reviewing instructions, searching existing data sources, gathering and maintaining the data needed, and completing and reviewing the collection of information. Send comments regarding this burden estimate or any other aspect of this collection of information, including suggestions for reducing the burden, to Department of Defense, Executive Services, Directorate (0704-0188). Respondents should be aware that notwithstanding any other provision of law, no person shall be subject to any penalty for failing to comply with a collection of information if it does not display a currently valid OMB control number. PLEASE DO NOT RETURN YOUR FORM TO THE ABOVE ORGANIZATION.					
1. REPORT DATE (DD-MM-YYYY) 03-06-2019		2. REPORT TYPE Final		3. DATES COVERED (From - To) 15 Jan 2017 to 14 Jan 2019	
4. TITLE AND SUBTITLE Novel Photomechanics Approaches to Determine the Intra- & Inter-laminar Properties of Composites at High Rates of Strain				5a. CONTRACT NUMBER	
				5b. GRANT NUMBER FA9550-17-1-0133	
				5c. PROGRAM ELEMENT NUMBER 61102F	
6. AUTHOR(S) Fabrice Pierron				5d. PROJECT NUMBER	
				5e. TASK NUMBER	
				5f. WORK UNIT NUMBER	
7. PERFORMING ORGANIZATION NAME(S) AND ADDRESS(ES) UNIVERSITY OF SOUTHAMPTON UNIVERSITY RD SOUTHAMPTON, SO17 1BJ GB				8. PERFORMING ORGANIZATION REPORT NUMBER	
9. SPONSORING/MONITORING AGENCY NAME(S) AND ADDRESS(ES) EOARD Unit 4515 APO AE 09421-4515				10. SPONSOR/MONITOR'S ACRONYM(S) AFRL/AFOSR IOE	
				11. SPONSOR/MONITOR'S REPORT NUMBER(S) AFRL-AFOSR-UK-TR-2019-0031	
12. DISTRIBUTION/AVAILABILITY STATEMENT A DISTRIBUTION UNLIMITED: PB Public Release					
13. SUPPLEMENTARY NOTES					
14. ABSTRACT This project aims at exploring new ways of testing composite materials at high rates. In particular, the project focuses on transverse tension and shear, both intra and interlaminar. It is well known that such properties are difficult to measure using Kolsky bars because of the relatively low wave speeds and low strains to failure, preventing quasi-static equilibrium to be achieved before failure. As a consequence, reliable available data only span up to a few 100s of s⁻¹. Recently, progress in ultra-high speed imaging has allowed a new methodology to emerge. The underpinning idea is to use acceleration fields obtained by ultra-high speed imaging of grid or speckle patterns to provide stress information through mechanical equilibrium. Moreover, the implementation of inertial impact enables for free edge reflection of the impeding compression wave to be turned into tension, providing tensile properties (stiffness and strength) without the need for any gripping of the test specimen. This methodology has been termed Image-Based Inertial Impact (IBII). In this project, the IBII methodology has been adapted so far to the transverse tensile properties of carbon/epoxy UD composites, as well as to interlaminar tension. Strain rates in the range of 1000s of s⁻¹ were achieved with exceptional quality data to measure the stiffness and strength. Future work will extend it to intra and interlaminar shear, and the combination of transverse tension and shear.					
15. SUBJECT TERMS EOARD, Materials, Interlaminar properties, High strain rate composites, intralaminar properties					
16. SECURITY CLASSIFICATION OF:			17. LIMITATION OF ABSTRACT SAR	18. NUMBER OF PAGES	19a. NAME OF RESPONSIBLE PERSON GARNER, DAVID
a. REPORT Unclassified	b. ABSTRACT Unclassified	c. THIS PAGE Unclassified			19b. TELEPHONE NUMBER (Include area code) 011-44-1895-616021

Report on Grant FA9550-17-1-0133

***Novel photomechanics approaches to determine the
intra- and inter-laminar properties of composite
materials at high rates of strain***

European Office of Aerospace Research and
Development (EOARD)

Principal investigator: Professor Fabrice Pierron

Date: Friday April 12th 2019

Abstract

This project aims at exploring new ways of testing composite materials at high rates. In particular, the project focuses on transverse tension and shear, both intra and interlaminar. It is well known that such properties are difficult to measure using Kolsky bars because of the relatively low wave speeds and low strains to failure, preventing quasi-static equilibrium to be achieved before failure. As a consequence, reliable available data only span up to a few 100s of s^{-1} .

Recently, progress in ultra-high speed imaging has allowed a new methodology to emerge. The underpinning idea is to use acceleration fields obtained by ultra-high speed imaging of grid or speckle patterns to provide stress information through mechanical equilibrium. Moreover, the implementation of inertial impact enables for free edge reflection of the impeding compression wave to be turned into tension, providing tensile properties (stiffness and strength) without the need for any gripping of the test specimen. This methodology has been termed Image-Based Inertial Impact (IBII).

In this project, the IBII methodology has been adapted so far to the transverse tensile properties of carbon/epoxy UD composites, as well as to interlaminar tension. Strain rates in the range of 1000s of s^{-1} were achieved with exceptional quality data to measure the stiffness and strength. Future work will extend it to intra and interlaminar shear, and the combination of transverse tension and shear.

Keywords: Virtual Fields Method, ultra-high speed imaging, grid method, inverse identification, inertial impact, IBII test

1. Introduction

This project aims at exploring new ways of testing composite materials at high rates. In particular, the project focuses on transverse tension and shear, both intra and interlaminar. It is well known that such properties are difficult to measure using Kolsky bars because of the relatively low wave speeds and low strains to failure, preventing quasi-static equilibrium to be achieved before failure. As a consequence, reliable available data only span up to a few 100s of s^{-1} .

In opposition to this, the present approach aims at taking advantage of the inertial effects as acceleration fields, derived from full-field time-resolved displacement measurements, can be used as a volume-distributed load cell. This first seminal idea was published in 2011 [1] for a purely elastic loading. It was further progressed through the use of purely inertial loadings [2, 3] and the present project aims at extending the idea to the transverse tensile and shear intra and interlaminar properties of UD composites.

Researchers associated to the project:

- Mr Jared Van Blitterswyk, PhD student funded by the University of Southampton
- Dr Lloyd Fletcher, Post-doctoral researcher, partly funded by the present project and the Photodyn programme¹
- Dr Fabrice Pierron, Professor and PI, funded by the Photodyn programme¹

2. Methodology

The methodology employed for this project involved both numerical and experimental tools. The work was organized as follows.

- Literature review on high strain rate tests for composites, with a focus on interlaminar tensile and shear properties (matrix-dominated). A review paper has been published on the back of this literature review.
- Explicit finite element simulations of the image-based inertial impact (IBII) tests to design the test configuration, in particular the impact pulse (impactor length and material, impactor speed).
- Based on these finite element simulations, synthetic grid deformation is employed [4] to explore the main sources of error for the test. In particular, edge data reconstruction and optimal smoothing parameters are critical.
- New 'rigid-body-like' virtual fields have been developed leading to the so-called 'linear stress-gauge approach' [5] for refined strength identification.
- Extension of the above for off-axis IBII tests [6].
- A series of experimental campaigns has been implemented to determine modulus and strength for both intra and interlaminar tension.
- Finite element modelling is currently being used to design new test configurations for shear properties. Off-axis IBII is considered for intralaminar shear and combination with transverse tension, while direct IBII shear is being explored for shear-dominated properties.

¹ www.photodyn.org

3. Principal conclusions of the project

- The IBII test is well adapted to test 90° UD test specimens in tension [7]. The initial compression wave does not damage the material and the use of a small impactor provides a narrow band of tensile stress that fractures the specimen, enabling the strength to be measured.
- The IBII test is well adapted to test through-thickness UD test specimens in tension [8]. The specimens are much smaller than the in-plane ones as their length is the plate thickness. This requires smaller grid pitches and thinner impactors.
- Grids are now printed directly onto the test specimen with a flatbed printer². The minimum achievable pitch is of the order of 0.33 mm, which is adequate for interlaminar tension specimens down to about 15 mm in length (cut from 15 mm thick plates).
- Both modulus and strength are identified remarkably consistently with coefficients of variations lower than 5% for the modulus, and lower than 20% for the strength. This has been successful for both intra and interlaminar configurations.
- The achievable strain rates easily reach above 1000 s⁻¹.
- The off-axis IBII test allows for simultaneous identification of transverse tension and shear response³. A journal article is in preparation.

4. Future work

This project has led to significant progress in the development of IBII tests for composites, for both inter and intra-laminar properties.

The subject will be pursued by a new PhD student funded by the Faculty of Engineering and Physical Sciences, started in Jan. 2018, Mr Sam Parry. Off-axis IBII tests will be developed further. We will also be looking at extending these tests to multi-directional laminates. A new PhD student should be recruited this year to pursue this.

Extension of the work to [IBUS tests](#) is underway and a journal article is being finalized.

5. Impact

Following Prof. Pierron's keynote presentation at CompTest2018 in Leuven, Belgium in April 2017, Solvay Brussels showed great interest for the IBII methodology. He was invited to provide a 2h workshop on the method at Solvay's HQ in Brussels on June 30th 2017. The workshop combined IBII for composites and Image-Based DMTA for polymers (grant FA9550-15-1-0293). Solvay then visited our group in Southampton with their R&D director on 29th September 2017, and Prof. Pierron was invited to give a presentation at the Solvay Symposium on damage & fracture mechanics in polymers and composites on November 28th 2017.

As part of the Photodyn programme, an industrial engagement event was organised on September 6-7 2017 in Southampton⁴. At this event, our IBII results on composites were presented and raised a lot of interest. DSTL will provide us with samples to test to

² http://photodyn.org/wp-content/uploads/2016/03/Report_UniversityPrintCentre.pdf

³ http://photodyn.org/Presentations/LF_BSSM2018_IBII_CFIP45_Final/LF_BSSM2018_IBII_CFIP45_Final.html

⁴ <http://photodyn.org/events/past-events-industrial-engagement-day>

demonstrate feasibility on a material system they are interested in (an MTA has been put in place).

Several informal collaborations have also been established to test materials for which extensive data exists on quasi-static and/or moderate strain rates:

- Professor Takashi Yokoyama provided us with a 10 mm thick plate of carbon/epoxy UD to test at higher strain rates in the Interlaminar plane. Data up to 100 s^{-1} exists in his group. This will however need more test design to show that such small thicknesses can be tested (up to now, the specimens were about 20 mm thick).
- Dr Devlin Hayduke from Materials Sciences Corp., Horsham, PA, USA sent us an 18 mm thick panel of CYTEC MTM45-1/ 12K AS4 material for interlaminar properties. This material has been the object of extensive testing and data is freely available⁵.
- Dr Julien Berthe from ONERA, France, provided us with a series of thin plates from T700GC/M21 prepreg. ONERA has extensive data for this material up to moderate strain rates [9] and we will explore further the strain rate dependence.
- Dr Leslie Lamberson is currently adopting our methodology and we published a conference paper together:
Cadiot X., Fletcher L., Pierron F., Lamberson L., Identification of Dynamic Damage Evolution in Composites using the Virtual Fields Method, ASC 33rd Annual Technical Conference, 18th US-Japan Conference on Composite Materials, ASTM D30, 24-26 September in Seattle, USA, 2018

Awards

- Jared Van Blitterswyk has reached [second place at the 2018 SEM student paper competition](#)
- Jared Van Blitterswyk has won the [2018 BSSM Young Stress Analyst competition](#)

Public engagement

- Presentation on high speed imaging and demonstration of the IBII test to general public audiences during the 2019 SOTSEF festival (about 100 people in total).

6. Publications

Journal articles:

- Van Blitterswyk J., Fletcher L., Pierron F.
Characterisation of the interlaminar properties of composites at high strain rate: a review.
Advanced Experimental Mechanics (JSEM Publisher), Vol. 2, pp. 3-28, 2017.
http://dx.doi.org/10.11395/aem.2.0_3.
- Fletcher L., Van Blitterswyk J., Pierron F.
A novel image-based inertial impact (IBII) test for the transverse properties of composites at high strain rates.
Journal of Dynamic Behaviour of Materials, vol. 5, n° 1, pp. 65-92, 2019.
<https://doi.org/10.1007/s40870-019-00186-y>
- Van Blitterswyk J., Fletcher L., Pierron F.
Image-based inertial impact test for composite interlaminar tensile properties.

⁵ http://www.niar.wichita.edu/coe/NCAMP_Documents/Cytec5320-1/NCP-RP-2008-004 Rev NC MTM45-1 AS4-145 UNI Qual Stat report 3-25-16.pdf

Journal of Dynamic Behaviour of Materials, vol. 4, n° 4, pp. 543-572, 2018.

<http://dx.doi.org/10.1007/s40870-018-0175-1>

- Pierron F., Fletcher L.
Generalized stress-strain curves for IBII tests on isotropic and orthotropic materials.
Journal of Dynamic Behaviour of Materials, in revision, 2019.

Two more journal articles are being drafted to complete the PhD thesis of Jared Van Blitterswyk. To be submitted in July 2019.

Invited plenary or keynote presentations

- Pierron F., Fletcher L., Siviour C.R., Zhu H., New opportunities in high strain rate testing of composite materials, 8th International Conference on Composites Testing and Model Identification (CompTest), 5-7 April 2017 in Leuven, Belgium. Invited keynote presentation.
- Pierron F., Davis F., Fletcher L., Seghir R.
Current trends in image-based high strain rate testing
Annual SEM Conference (Society for Experimental Mechanics), 12-15 June in Indianapolis, IN, USA, 2017. Invited keynote presentation.

Conferences articles:

- Fletcher L., Pierron F.
Inertial impact method to measure the transverse and through-thickness tensile strengths of laminated composites at high strain rates.
Annual SEM Conference (Society for Experimental Mechanics), 12-15 June in Indianapolis, IN, USA, 2017.
- Seghir R., Fletcher L., Pierron F.
Image-based inertial ultrasonic tests to measure the transverse stiffness and strength of UD composites at high strain rates.
Annual SEM Conference (Society for Experimental Mechanics), 12-15 June in Indianapolis, IN, USA, 2017.
- Van Blitterswyk J., Fletcher L., Pierron F.
High strain rate through-thickness testing of composites.
BSSM annual conference, 29-31 August 2017, Sheffield, UK.
- Van Blitterswyk J., Fletcher L., Pierron F.
High strain rate interlaminar tensile properties of fibre-reinforced polymer composites using a novel image-based inertial impact (IBII) test.
Annual SEM Conference (Society for Experimental Mechanics), 4-7 June in Greenville, SC, USA, 2018.
- Fletcher L., Van Blitterswyk J., Pierron F., Combined shear/tension testing of fibre composites at high strain rates using an image-based inertial impact test.
DYMAT conference, 10-14 September 2018 in Arcachon, France.
- Fletcher L., Van Blitterswyk J., Pierron F.
Image-based impact (IBI) tests for high-rate behaviour of composites.

16th European Mechanics of Materials Conference, 26-28 March in Nantes (France), 2018.

- Van Blitterswyk J., Fletcher L., Pierron F.
High strain rate interlaminar tensile properties of fibre-reinforced polymer composites using a novel image-based inertial impact (IBII) test.
18th European Conference on Composite Materials (ECCM18), 24-26 June 2018.
- Van Blitterswyk J., Fletcher L., Pierron F., An image-based inertial impact test for high strain rate interlaminar shear properties of fibre-reinforced polymer matrix composites.
BSSM annual conference, 29-31 August 2018, Southampton, UK.
- Van Blitterswyk J., Fletcher L., Pierron F.
Off-axis testing of fibre composites at high strain rates using an image-based inertial impact test
BSSM annual conference, 29-31 August 2018, Southampton, UK.

7. Review meetings

- June 27th 2017, in Southampton
Participants: Dr Matt Snyder (EOARD), Dr Jason Foley (AFRL), Dr Lloyd Fletcher, Dr Rian Seghir, Dr Frances Davis (UoS, post-doctoral researchers), Mr Jared Van Blitterswyk (UoS, PhD student), Prof. Fabrice Pierron (PI, UoS)

8. Appendix:

- Van Blitterswyk J., Fletcher L., Pierron F.
Characterisation of the interlaminar properties of composites at high strain rate: a review
Advanced Experimental Mechanics (JSEM Publisher), Vol. 2, pp. 3-28, 2017.
http://dx.doi.org/10.11395/aem.2.0_3
- Fletcher L., Van Blitterswyk J., Pierron F.
A novel image-based inertial impact (IBII) test for the transverse properties of composites at high strain rates.
Journal of Dynamic Behaviour of Materials, vol. 5, n° 1, pp. 65-92, 2019.
<https://doi.org/10.1007/s40870-019-00186-y>
- Van Blitterswyk J., Fletcher L., Pierron F.
Image-based inertial impact test for composite interlaminar tensile properties.
Journal of Dynamic Behaviour of Materials, vol. 4, n° 4, pp. 543-572, 2018.
<http://dx.doi.org/10.1007/s40870-018-0175-1>

Online presentations:

- http://photodyn.org/Presentations/Pierron_2017_CompTest_keynote/Pierron_2017_CompTest_keynote.html
- http://photodyn.org/Presentations/LF_BSSM2018_IBII_CFIP45_Final/LF_BSSM2018_IBII_CFIP45_Final.html

- http://photodyn.org/Presentations/BSSM18_JVB_InterlaminarShear/BSSM18_JVB_InterlaminarShear.html
- http://photodyn.org/Presentations/BSSM18_JVB_InterlaminarTension/BSSM18_JVB_InterlaminarTension.html

References

1. Moulart, R., et al., *Full-field strain measurement and identification of composites moduli at high strain rate with the virtual fields method*. Experimental Mechanics, 2011. **51**(4): p. 509-536.
2. Pierron, F. and P. Forquin, *Ultra high speed full-field deformation measurements on concrete spalling specimens and stiffness identification with the Virtual Fields Method*. Strain, 2012. **28**(5): p. 388-405.
3. Pierron, F., H. Zhu, and C. Siviour, *Beyond Hopkinson's bar*. Philosophical Transactions of the Royal Society A: Mathematical, Physical and Engineering Sciences, 2014. **372**(2023): p. 20130195.
4. Rossi, M., et al., *Effect of DIC spatial resolution, noise and interpolation error on identification results with the VFM*. Strain, 2015. **51**(3): p. 206-222.
5. Fletcher, L., J. Van Blitterswyk, and F. Pierron, *Novel image-based spalling test for the transverse properties of unidirectional composites at high strain rates*. Composites Part A: Applied Science and Manufacturing, 2017. **Submitted**.
6. Pierron, F. and L. Fletcher, *Generalized stress-strain curves for IBII tests on isotropic and orthotropic materials*. Journal of Dynamic Behaviour of Materials, 2019.
7. Fletcher, L., J. Van-Blitterswyk, and F. Pierron, *A Novel Image-Based Inertial Impact Test (IBII) for the Transverse Properties of Composites at High Strain Rates*. Journal of Dynamic Behavior of Materials, 2019.
8. Van Blitterswyk, J., L. Fletcher, and F. Pierron, *Image-based inertial impact test for composite interlaminar tensile properties*. Journal of Dynamic Behavior of Materials, 2018. **4**: p. 543-572.
9. Berthe, J., et al., *Consistent identification of CFRP viscoelastic models from creep to dynamic loadings*. Strain, 2013. **49**(3): p. 257-266.

Characterisation of the Interlaminar Properties of Composites at High Strain Rates: A Review

Jared Van BLITTERSWYK¹, Lloyd FLETCHER² and Fabrice PIERRON³

^{1,2,3} Faculty of Engineering and the Environment, University of Southampton, Southampton SO17 1BJ, UK

(Received 30 June 2017; accepted 30 June 2017)

Abstract: This paper provides a review of the strain rate effect on the through-thickness mechanical properties of fibre-reinforced polymer composites. The challenges and limitations associated with existing methods for through-thickness testing are discussed. Specific attention is given to the split Hopkinson pressure bar, and implications of the intrinsic limitations for measuring mechanical properties. The influence of strain rate on elastic modulus, ultimate strength and ultimate strain is provided for compressive, tensile and shear loading conditions. The review is concluded with a discussion on the direction of future research focussed around full-field imaging techniques.

Keywords: High strain rate, Interlaminar properties, Fibre-reinforced polymer composites, Test methods

1. Introduction

Fibre-reinforced polymer (FRP) composite materials have become popular in many sectors such as aerospace, naval, automotive and infrastructure. The incorporation of these materials in design has resulted in necessarily thicker sections to support complex three-dimensional stress states. Often, these structures are subjected to high-rate loading that can generate significant through-thickness stresses. For example: bird and foreign object strikes, vehicle collisions, ballistic impacts, blast, etc. Due to the relatively lower strength in the through-thickness direction, these stresses can lead to micro cracking or delaminations and premature failure of the material. Advances in computational modelling power have enabled designers to simulate such complex loading conditions with the objective of designing safer structures. However, achieving the full potential of numerical simulations hinges on the accuracy of constitutive models for the through-thickness response of the material. The works by Daniel et al. [1] and Gillespie et al. [2] are among the few efforts to calibrate/modify existing failure models to account for strain rate effects in the through-thickness direction. The effects of strain rate are still not well understood, largely due to the lack of standard testing procedures, experimental data, and limitations inherent to existing test methods. Existing studies suggest a strain rate dependency on the through-thickness elastic stiffness and strength properties [3–6]. Unfortunately, high levels of scatter, between and within studies, has prevented establishment of reliable material models.

The development of test methods for measuring through-thickness parameters at quasi-static and high strain rates has lagged behind in-plane testing. This is partly due to the strong influence of the aerospace industry, where many composite sections could be adequately characterized by their in-plane properties. Now, the desire to use thicker composite structures in the aerospace industry (*i.e.*: brackets, wing spar box, etc.) is driving the need for reliable through thickness properties. At high strain rates, numerous factors make

testing materials particularly challenging for both in-plane and the through-thickness directions. One factor is the relatively small through-thickness dimension of a typical thin laminate, which does not facilitate adaptation of traditional in-plane tests. The small through-thickness dimension introduces complications with gripping and alignment of the specimen [7–9]. The common solution is to increase the size of the specimen to enable traditional coupon designs to be used. However, this must be done carefully to ensure properties obtained from the test specimen is representative of a typical laminate. For woven fibre reinforcements, the minimum dimensions are governed by the size of a representative volume element [7, 10], while for laminates, higher void concentrations may introduce a significant volume effect. Apart from geometrical considerations, additional difficulties are encountered when testing composites at high strain rates.

The management of inertial effects becomes critical when testing at strain rates above 100 s^{-1} (Fig. 1 in [11]). This tends to be particularly problematic for many existing techniques, which make use of limited information and rely on a number of assumptions to quantify the material's response. The best example of this is the popular split Hopkinson pressure bar (SHPB), which acts as a dynamic load cell under very specific conditions (in particular, one-dimensional wave propagation, elastic bar deformation, quasi-static equilibrium conditions). Quasi-static equilibrium can be quite difficult to achieve, particularly for materials with low wave speeds. This significantly restricts achievable strain rates to approximately $1\text{--}2 \times 10^3 \text{ s}^{-1}$ for failure properties in compression [4, 12] and an order of magnitude lower in tension [6].

The purpose of this paper is to provide an overview of existing techniques for measuring properties of FRP composites in the through-thickness direction at intermediate and high strain rates. A brief summary of challenges associated with quasi-static through-thickness testing is first provided,

with particular attention to those issues also relevant to high strain rate testing (Section 2). This is followed by a review of existing approaches for through-thickness testing at high strain rates in Section 3. Further attention is given to the SHPB and implications of the intrinsic assumptions for testing composites. Section 4 offers a review of the state-of-the-art understanding of the strain rate effect on elastic modulus, strength and ultimate strain under tension, compression and shear loading. Finally, Section 5 discusses the use of full-field optical measurements with particular attention to the potential advancement and development of new and existing test methods.

2. Challenges Common to Quasi-Static and High Strain Rate Tests

Obtaining material properties for composites in the through-thickness direction is a long outstanding problem. The challenges faced by experimentalists for quasi-static loading are first reviewed here as many apply to high strain rate testing. For instance, introducing a tensile load generally requires gripping of the specimen. Often this causes premature failure due to eccentric loading (bending stresses) or stress concentrations at the grips. This is especially problematic for through-thickness testing due to the inherently smaller specimen dimensions. Manufacturing and machining quality can also have significant influence on the failure behaviour. For thick specimens formed by bonding several smaller laminates, failure can occur near the bond interface, resulting in failure strengths that are not representative of an original laminate [7]. Composites also show a strong volume effect. Void content from manufacturing increases with volume causing a degradation of matrix properties and reduced tensile and shear strengths [13, 14]. Pre-defects are especially critical as failure can initiate prematurely from voids or micro cracks cause by the machining process. All of these issues have created a notable sensitivity of material properties to test method [15–17]. For the interested reader, a review of available quasi-static test methods can be found in references [7, 8, 13]. In addition to the aforementioned challenges, several others present themselves at high strain rates, which will be discussed in the next section.

3. Through-Thickness Test Methods for Intermediate and High Strain Rates

This section provides a brief summary of available methods to test composites in the through-thickness direction at intermediate and high strain rates. The reader is referred to the review by Field *et al.* [11] for a more general overview of testing methods applicable to both in-plane and through-thickness directions.

One of the key challenges for testing at high strain rates is the accurate measurement of strain and force. At intermediate strain rates, high speed load frames are commonly used. Such systems allow the crosshead to be accelerated to the desired speed prior to engaging the grip mechanism. Unfortunately, force measurement may be unreliable at high loading speeds due to inertial effects and ‘ringing’ in the

load cell [18–20]. As a result, high speed load frames are typically limited to strain rates well below 100 s^{-1} .

Inertia-induced ringing is also problematic for contact strain instruments such as mechanical extensometers. Strain gauges have been used extensively for dynamic testing due to their simplicity and sufficiently high dynamic response, which can be sampled with available data acquisition systems (typically sampled at MHz rates [2]). Non-contact approaches, such as high speed optical extensometers [19, 20] or laser Doppler velocimetry [20, 21], have been used to measure global deformation. Unfortunately, these techniques are not well-suited to characterise heterogeneous deformations resulting from inertia. With advancements in computing power and high speed imaging technology, full-field measurements offer a viable alternative to traditional strain gauges for high strain rate testing [22–26].

Full-field optical techniques, particularly digital image correlation (DIC), has seen widespread use for quasi-static testing, but limited use at high strain rates [18, 19, 22, 26, 27]. ‘Full-field’ refers to the large number of measurement points obtained via automated image processing. At the necessary frame rates for high strain rate testing, the resolution of traditional high speed cameras is inadequate to perform meaningful full-field measurements in the case of heterogeneous states of deformation. This is a result of the memory readout structure, which requires spatial resolution to be sacrificed for temporal resolution. Limited spatial resolution is problematic for DIC, which requires correlation over a relatively large subset of pixels. A new generation of high speed cameras, known as ‘ultra-high speed cameras’ (frame rates in excess of 1 Mfps) provides much greater spatial resolution at the cost of record length [28]. With these cameras, DIC is more feasible; however, the grid method offers a better compromise between spatial and measurement resolution [29, 30]. The grid method, is similar to DIC but relies on tracking of a regular grid rather than a random pattern. This approach is particularly well suited for measuring small strains (*i.e.*: ultimate strains for composites in the through-thickness direction), which is a challenge for DIC. The combination of ultra high speed imaging and the grid method has shown promise in recent literature for obtaining full-field measurements of displacement and strain at high strain rates [23–25]. In these studies, the measurement resolution is high enough that inverse identification procedures, such as the virtual fields method (VFM), can be used to identify constitutive properties. However, this technique is still in its early stages of development and has yet to be used for high strain rate testing of composite materials in the through-thickness direction. Further discussion surrounding the potential of this approach for high strain rate testing will be presented in Section 5.

To extract material properties from the limited information provided by typical instruments (*i.e.*: strain gauges) requires a number of assumptions about the material response. These assumptions impose constraints on the complexity of the test. This is a commonality in existing techniques, in that

they are sufficiently simple to enable limited amounts of information to be used to extract constitutive parameters governing the dynamic behaviour. For example, the Taylor impact test [31], uses post-impact measurements of plastic deformation to infer dynamic flow stresses in the material. However, the Taylor impact test has limited application for thermoset fibre-reinforced composites, which exhibit small strains to failure and minimal plastic deformation. Similarly, a drop weight test [11] attempts to use the impactor as a load cell. Similar to high speed load frames, inertial ringing within the impactor may mask the underlying mechanical response. The Kolsky pressure bar [32], also known as the Split Hopkinson Pressure Bar (SHPB) (Fig. 1), operates similarly in that the loading bars are used as a dynamic load cell. This arrangement is only valid for a very specific set of conditions and relies on numerous assumptions about the dynamic response of the material. [33].

The key is that all available techniques make use of limited information. The test data is generally provided by a few point measurements typically provided by strain gauges. As a result, all current test methods suffer from many intrinsic limitations, a number of which are particularly challenging to overcome when testing more compliant materials like unidirectional composites in transverse tension or shear. Following discussions will focus on the SHPB due to its overwhelming popularity for high strain rate testing of composites. Particular attention is given to explaining the variations for tensile, compression and shear testing, and the intrinsic limitations most relevant to through-thickness testing of FRP composites.

3.1 Split-Hopkinson pressure bar (SHPB) testing

The SHPB has played an invaluable role in obtaining information on the high strain rate response of FRP composites in the through-thickness direction [2, 4–6, 10]. With this technique, a specimen is generally subjected to a compressive loading pulse while being sandwiched between two elastic bars, denoted as the incident and transmitter bars (Fig. 1a). In this configuration, a striker is used to impact the incident bar and induce a compressive pulse. The specimen may also be loaded by direct impact (Fig. 1b); however, this approach is less common due to issues with alignment and pulse transmission. In the two bar configuration, the pulse from the striker propagates towards the specimen, with the input pulse recorded via the strain gauge on the bar. Once the pulse reaches the specimen, some of the compressive pulse is transmitted through the specimen into the transmitter bar, and some is reflected back through the incident bar at the specimen interface. The amount of the pulse which is reflected or transmitted is dependent on the impedance mismatch between the bars and the specimen. Under specific conditions, the pulses measured on the incident and transmitter bars can be used to deduce the stress state within the specimen using one-dimensional wave theory. In these cases, the portion of the wave transmitted through the specimen describes the stress in the specimen, while the reflected

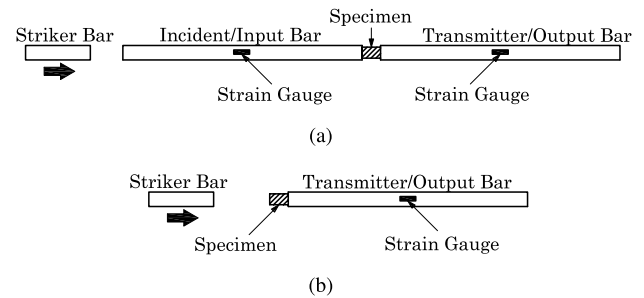


Fig. 1 Schematic of two forms of a pressure bar apparatus: a) Split-Hopkinson Pressure Bar, and b) direct impact pressure bar. Adapted from Gama *et al.* [34]

pulse describes the strain rate [5]. Most efforts have focused on obtaining the dynamic compressive properties of composites in the through-thickness direction due to the relative simplicity of implementing the test [11]. The compression SHPB test has been adapted to permit high rate testing of materials in tension, shear, torsion, bending, and combined load states [2, 5, 34, 35].

The principles of a tensile SHPB are similar to compression. In the case of tensile loading, the main differences lie in the means of generating the pulse, typical specimen geometries, and gripping/attachment to the input and transmitter bars [33]. Tensile pulses are most commonly applied using a direct impact on a flange attached to the incident bar, or through the release of a static tensile stress in the incident bar [20]. Tensile pulses may also be applied using a top hat specimen and hollow transmitter bars, or using the reflected pulse from the transmitter bar with a collar used to protect the specimen in compression. The reader is referred to reference [20] for further details. U-shaped striker bars have also been considered as a means of generating longer pulses and extending the range of achievable strain rates [36].

An alternative approach to the two-bar reflected pulse system is the single bar spalling test [37]. In this case, the specimen is attached on one side to the end of the input bar and the transmitter bar is removed so one edge of the specimen is free. The specimen is loaded in tension after the compressive pulse reflects off of the free edge and becomes tensile. Strain gauge measurements and point-wise laser Doppler velocimetry may be used to infer the tensile strength of the material. This is accomplished using an analytical solution for the specimen stress state based on the assumption of one-dimensional wave propagation [37]. The spall test is particularly useful for loading the specimen with a single pulse, as long as the compressive strength largely exceeds the tensile strength. This enables the input pulse to be tailored such that no damage is caused during compressive loading, but the reflected tensile wave causes failure.

For direct tensile loading, the specimen is generally attached to the incident and transmitted bars using threaded inserts or adhesives [38]. Mechanical inserts have the disadvantage of

creating additional reflections and dispersion of the wave, affecting the accuracy of the measurement using SHPB theory [19, 20]. The grips can also introduce stress concentrations and heterogeneous deformation. Specimen geometry is an important consideration for tensile testing. A waisted cylindrical specimen is commonly used to ensure failure occurs within the gauge section [22, 33]. However, due to the contribution of the non-waisted section to the response of the material, calculation of stress and strain are more complicated using standard one-dimensional wave theory [22]. Further, the specimens must be designed such that the gauge section undergoes uniform axial stress. The appropriate geometry is dependent on the material [22].

The SHPB has also been used to measure the interlaminar shear properties. There are two common configurations. A single or double lap shear joint loaded in compression using a SHPB was common in early studies. A shortcoming of the lap joint is high normal stresses at the ends of the overlap, which typically dominates the initiation of failure [39, 40]. An alternative approach is to load thin-walled tubular specimens in shear using a torsional SHPB [5, 41]. This arrangement is very similar to the compression SHPB, with the exception that a shear wave is induced by applying a torque pre-load on the incident bar.

3.1.1 Assumptions and limitations associated with the SHPB

A number of assumptions are required for strain gauge measurements on the bars to be used to infer the stress and strain states in the specimen for a SHPB test. The assumptions are briefly summarised below, and the interested reader is referred to the paper by Gama *et al.* [34] for a more in-depth review. The assumptions to be satisfied for a valid test include:

1. one-dimensional stress wave propagation in the incident and transmission bars (*i.e.*: negligible wave dispersion);
2. interfaces between the specimen and bars remain planar at all times;
3. the specimen is in a condition of quasi-static stress equilibrium (*i.e.*, the forces exerted at both specimen ends are equal in magnitude and opposite in sign), and;
4. friction effects at the specimen-bar interfaces can be neglected.

Tests suffering from high amounts of dispersion (violating assumption 1) exhibit a non-linear initial region of the stress-strain curve, followed by oscillations about the straight line that would exist for a bar free of distortion [34]. An example is shown in the work by Gerlach *et al.* [10], where the stress-strain response at high strain rate exhibits a clear oscillatory behaviour (Fig. 2). A number of corrective approaches have been developed [34], but, the effects of dispersion are never completely removed. Aside from minimising misalignment at the bar-specimen interfaces, ‘pulse shaping’ is the only other approach to reduce the amount of

dispersion by smoothing and limiting high frequency content in the compressive pulse [34, 42]. Pulse shaping is generally achieved by modifying the shape of the impactor or placing a thin layer of low impedance material, or plastically deforming metal, between the striker and input bar. This however limits the strain rate that the tested specimen will see.

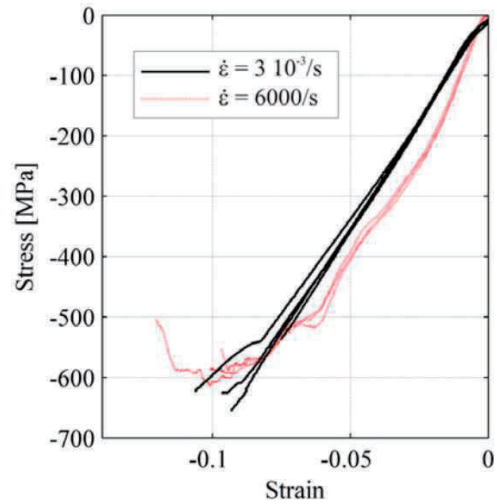


Fig. 2 Compression stress-strain response highlighting the influence of wave dispersion effects on the linear response measured using a SHPB at a strain rate of $6,000 \text{ s}^{-1}$. Cubic specimens, 10 mm thick, carbon/epoxy 3D weave. Taken from Gerlach *et al.* [10]

The second assumption is generally satisfied for FRP composites in the through-thickness direction due to the relatively low acoustic impedance compared to the bars. This condition is more problematic for harder materials that may cause local deformation of the bars [34]. The issue of stress equilibrium (assumption 3) for SHPB testing has been studied extensively in the literature [34, 43, 44]. A schematic diagram of an anvil type specimen of constant thickness e , in a SHPB test is shown in Fig. 3.

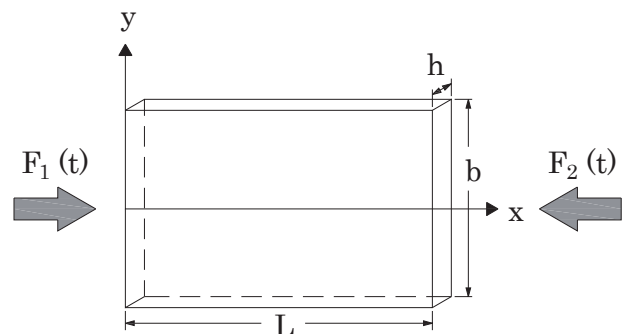


Fig. 3 Schematic of anvil-type specimen subjected to arbitrary end loads

The specimen is subjected to two time varying end loads from the reactions at the incident and transmitter bar interfaces. Assuming the loads are applied normal to the specimen end faces, the local equilibrium for this specimen has

the form of Eq. (1),

$$F_1(t) + F_2(t) = \rho e \int_0^L \int_{-b/2}^{b/2} a_x(x, y, t) dx dy \quad (1)$$

where $F_1(t)$ and $F_2(t)$, denote the forces at the incident and transmitter bar interfaces, respectively, b and L denote the specimen width and length, respectively, ρ is the density of the material (assumed constant here), and a_x is the local acceleration in the x direction. Stress equilibrium assumes that the force at the incident bar-specimen interface is equivalent to that at the transmitter bar-specimen interface (*i.e.*: acceleration is neglected). When the pulse reaches the specimen the input force generates a stress wave in the specimen (*i.e.*: inertial effects caused by transient acceleration). This stress wave travels through the specimen to the output bar where some is transmitted and some reflects and reverberates in the specimen. When the stress wave initially travels through the specimen, a heterogeneous stress state results, violating the assumption of quasi-static equilibrium (condition 3). In the case of softer materials, the stresses arising from inertial effects are large and mask the true mechanical response until the waves eventually damp out after multiple reverberations [38, 43]. The contributions of inertia are difficult to identify as no obvious effects are created in the apparent stress-strain behaviour. In a state of quasi-static equilibrium, the sum of incident and reflected pulses will equal the transmitted pulse ($\epsilon_I + \epsilon_R = \epsilon_T$) [38, 45]. For through-thickness tests, this condition is not satisfied until late in the test, as illustrated by the work of Gama *et al.* [45] (Fig. 4). A general criterion is that quasi-static stress equilibrium occurs after approximately three or four reverberations of the pulse through the specimen [43, 44]. Rather than counting reverberations, Gillespie *et al.* [2] used an 'R criterion', based on the difference between incident and transmitted signals, to assess the validity of a test. Regardless of the technique used to determine when quasi-static stress equilibrium conditions have been achieved, this assumption is always a source of discussion [34]. Therefore, it is generally accepted that the SHPB is simply inadequate to reliably measure initial material stiffness [4, 33, 34, 43].

The low wave speed in the through-thickness direction of FRP composites, coupled with the low strains to failure, impose significant restrictions on the achievable strain rates (typically less than 10^3 s^{-1} [35]). The problem with inertia for the SHPB is exacerbated by the quasi-brittle nature of the material, the low wave speed, and the low signal to noise ratio [41, 46]. Low strains at failure means that a state of quasi-static stress equilibrium may not be achieved before the test is complete. Short specimens and pulse shaping may be used to reduce the time to achieve quasi-static equilibrium and improve the uniformity of strain rate [34, 43, 47]; however, quasi-static equilibrium still may not be achieved prior to failure and short specimens may suffer from more significant end friction effects [38]. Notwithstanding the fact that for tension, the specimens cannot be very short. Gorham [48] showed that no aspect ratio effectively removes inertial effects. Softer materials also suffer from low signal-to-noise levels in the reflected pulse [38].

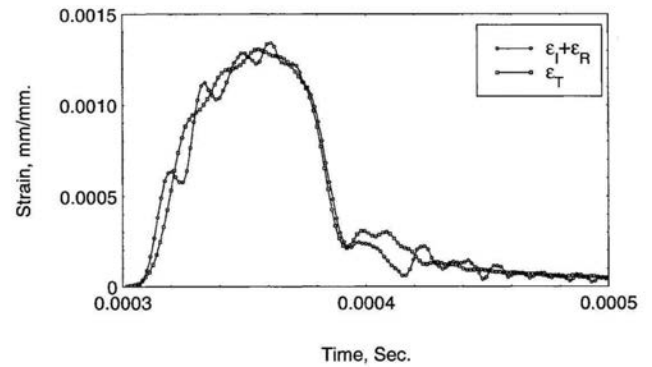


Fig. 4 Verification of quasi-static stress equilibrium for a through-thickness compression test at an average strain rate of $1,125 \text{ s}^{-1}$. Cubic specimens, 12.7 mm thick, plain weave S-2 glass/vinyl ester composite. Taken from Gama *et al.* [45]

Efforts to improve this include reducing the cross-section and stiffness of the input bars, using more sensitive strain gauges, using polymer bars [49], or direct force measurement [38]. Unfortunately, these approaches generally suffer from higher levels of wave dispersion and inertial ringing, making force measurement more unreliable.

4. Review of Interlaminar Properties at High Strain Rates

A review of current literature reporting on the sensitivity of through-thickness properties to strain rate will now be discussed. A compilation of references for compression, tension and shear properties is presented in Appendix A in Tables 1, 2 and 3, respectively. It must be noted that the following discussion focusses on relative trends between quasi-static and high strain rate properties. This is a result of the inconsistency in the literature regarding material composition (fibre and matrix materials, fibre volume fraction, reinforcement architecture, etc.), which makes direct comparison of quoted properties uninformative. Also, failure strain rate is defined inconsistently in the literature with some reporting an average strain rate and others reporting instantaneous strain rates. Defining a failure strain rate using a SHPB is difficult and not necessarily reliable due to the dependency on damage progression in composite materials. Therefore, no attempt is made by the present authors to convert reported strain rates to an equivalent metric across all studies.

4.1 Strain rate effects on compressive properties

4.1.1 Compressive elastic modulus

As discussed in the previous section, effects from inertia tend to mask the initial behaviour of the material under dynamic loading using a SHPB. Therefore, measurements of the modulus using the SHPB can only be regarded as 'apparent'. The degree to which inertia influences the measurement of strain is difficult to resolve using strain measurements on the incident and transmitter bars. This may contribute to the lack of general agreement in the literature regarding the sensitivity of the dynamic modulus to strain rate.

A summary of relative change in compressive elastic modulus as a function of strain rate is shown in Fig. 5. The majority of studies report a general increase in apparent modulus with increasing strain rate [4, 47, 50–53]. This behaviour is expected for a matrix-dominated behaviour as reported for thermoset resins tested without reinforcing fibres [46]. The relative increase in modulus is highly variable among studies, with increases ranging between 60 % [47, 53] and 150 % at strain rates around 10^3 s^{-1} [4]. According to the tests performed by Yokoyama and Nakai [4], the level of sensitivity appears to be more dependent on the reinforcement architecture (*i.e.*: cross-ply versus plain weave) compared to reinforcement material (glass versus carbon fibres). Despite having a lower compressive strength, the cross-ply laminates absorbed more energy compared to the plain weave composites. No physical explanation is provided for this behaviour. This may be a result of the increased void content for the plain weave architecture. In the work by Hosur *et al.* [54], the only other study to consider cross-ply laminates, the compressive modulus was higher compared to quasi-static values, but decreased with increasing strain rate. No physical explanation was offered by the authors for this trend.

The influence of fibre architecture is unclear, as very few studies consider pre-preg laminates, as illustrated in Fig. 5. Considering only studies that analyse plain weave architectures [9, 10, 12, 27, 45, 47, 51–53, 55], the magnitude of strain rate sensitivity is difficult to discern due to large scatter in reported measurements (see Table 1). For example, Song *et al.* [47] and Akil *et al.* [52] report increases in stiffness ranging from 75 % to 115 % (for strain rates near $1,000 \text{ s}^{-1}$), where as Shen *et al.* [51] report a mean increase of 350 % at $1,200 \text{ s}^{-1}$. The unrealistically high effect of strain rate on the modulus measured by Shen *et al.* [51] suggests that specimens are unlikely to be in stress equilibrium. Alternatively, a number of studies on plain weave composites [9, 10, 12, 27] show negligible variations of the modulus over a similar range of strain rates.

It is not surprising that many studies using the SHPB report an increase in compressive elastic modulus. In the early stages of a test, the reaction force on the input bar does not equal the force on the transmitter bar due to a significant contribution from acceleration (see Eq. (1)). As a result, the strains measured by the input bar are lower, and stress in the specimen, computed using the force on the transmitter bar, is higher. This leads to a perceived stiffening of the material. This effect is likely to increase with increasing strain rate as a greater portion of the impact force is expended in accelerating the material. The degree to which inertia affects a test is dependent on several factors, contributing to the scatter in the relative increases in stiffness across the literature.

In general, a review of the literature indicates that the compressive elastic modulus is matrix dominated and probably increases with increasing strain rate. The magnitude of this sensitivity is uncertain due to high scatter in the literature.

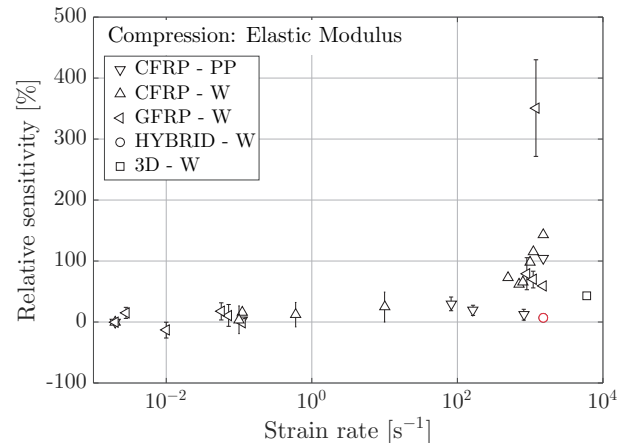


Fig. 5 Summary of relative strain rate sensitivity for compressive modulus from the literature. ‘PP’ and ‘W’ denote pre-preg and plain weave reinforcement, respectively. Red symbol outline denotes that values are quoted with respect to modulus at lowest strain rate considered ($1,275 \text{ s}^{-1}$ for [12]). Error bars denote the range of reported sensitivity and not standard deviation. Data taken from [4, 10, 12, 47, 50, 54]

4.1.2 Compressive strength

A summary of relative change in compressive strength as a function of strain rate is shown in Fig. 6. From Fig. 6 it is clear that there is high variability associated with the magnitude of change in strength with increasing strain rate. The trend at intermediate strain rates appears much more defined, showing a near linear positive sensitivity. There is a small group of studies that show good consistency at higher strain rates, reporting a moderate and positive increase in strength for strain rates below $1,000 \text{ s}^{-1}$ [4, 9, 50–52]. Reported increases vary between 6 % and 23 % for both CFRP and GFRP. The remaining studies report large relative increases in strength compared to quasi-static values: 33 % [45] (reaching a near asymptotic value at 700 s^{-1}), 46 % [12], 56 % [53], 80 % (between $1,000$ and $1,940 \text{ s}^{-1}$) [55] and 180 % [27]. The large increase in strength at high strain rates reported by Woo *et al.* [55] (80 %) was attributed to high energy absorption by the kevlar fibres. The maximum strain rate achievable by Woo *et al.* [55] was limited by low pulse transmission (as low as 10 %) through the specimen. This creates higher signal-to-noise ratio for strain measurement on the transmission bar, which is used to infer stress in the specimen. Quasi-static values were also not provided for comparison. In the study by Pankow *et al.* [27], the stress-strain response was highly non-linear and was heavily contaminated by dispersion. This introduces uncertainty when attempting to determine ultimate properties. In contrast to these results, Gerlach *et al.* [10] reported that failure strength remains approximately constant with increasing strain rate (quoted up to $6,000 \text{ s}^{-1}$).

A small number of studies report mixed trends for strength at high strain rates [4, 47, 54]. Hosur *et al.* [54] reported a positive sensitivity to strain rate, but peak stresses that are below quasi-static values. The specimens were loaded to failure under quasi-static conditions; however, at 82 s^{-1} and 164 s^{-1} the compressive input pulse was not strong enough to damage the specimens. Therefore, comparisons

made by the authors between peak stress at these strain rates and quasi-static strength are not equivalent. Therefore, only measurements collected at 817 s^{-1} can be used to assess the effect of strain rate on compressive strength. From this case, strain rate causes a decrease in strength by 37 %. The authors attributed this to a progressive change in failure mechanism from splitting and crushing at 163 s^{-1} to crushing and shearing at 817 s^{-1} . The instability of compressive loading tends to cause the specimen to fail in shear according to the strength of the matrix. Therefore, the change in failure mode is unlikely to be an intrinsic property of the material and more a result of the experimental setup (structural behaviour). Song *et al.* [47] reported lower strengths at strain rates up to 800 s^{-1} compared to quasi-static values. At higher strain rates ($> 1,000 \text{ s}^{-1}$), the dynamic strength exceeds the quasi-static one. The authors offer no explanation for this behaviour. Similar to the compression modulus, Yokoyama & Nakai [4] found that compressive strength was sensitive to reinforcement architecture. The woven glass-epoxy laminate exhibited a positive strain rate sensitivity, whereas the carbon/epoxy cross-ply pre-preg and plain weave laminates had a negative sensitivity. This was attributed to the properties of the fibres, however, the carbon/epoxy laminates had 15-20% higher fibre volume fractions. The potential dependency on reinforcement architecture (i.e: pre-preg., weave, etc.) is difficult to ascertain since the majority of studies focus on one type of reinforcement and have large scatter.

Other factors contributing to scatter include specimen geometry and uncertainty associated with quasi-static values. All studies in the literature use cubic or cylindrical specimens. A study by Tagarielli *et al.* [56] showed that the compressive strength and strain is highly sensitive to specimen geometry. For the same contact surface area, cylindrical specimens were found to fail at lower stress and strain compared to cubic specimens. As the volume of the cylindrical specimens was larger, this may indicate a volume effect. In any case, there appears to be a sensitivity to geometry and/or volume, which differs between studies in the literature. Another source of difficulty for comparing studies is that some work do not report quasi-static values at the same strain rates. Considering relative changes in strength over the high strain rate tests only, the values range between -5 % and +30 % compared to the range of -40 % to +40 %, when compared against quasi-static values, as shown in Fig. 6. This does not necessarily imply better accuracy, and instead suggests that high strain rate tests are reasonably repeatable under different conditions.

Compressive strength is heavily dominated by structural failure due to the inherent instability of the loading condition. Specimens will tend to fail in shear, which is determined by the strength of the resin. Some authors report a change in failure mode, which is attributed to strain rate sensitivity [27, 50, 51]. This highlights the challenge associated with measuring compressive strength and is likely a strong contributor to scatter in the literature. The intrinsic instability of the test makes the measured strength highly

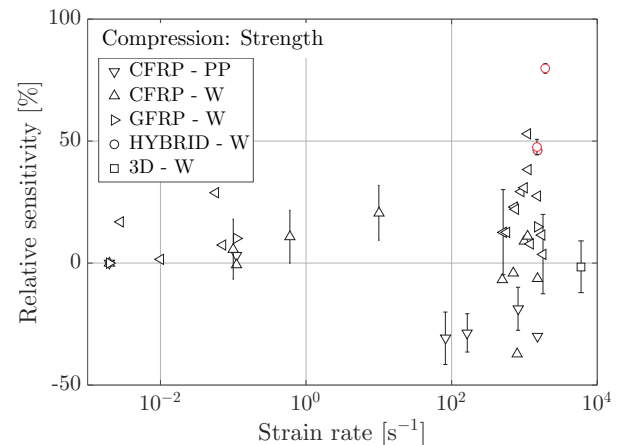


Fig. 6 Summary of relative strain rate sensitivity for compressive strength from the literature. 'PP' and 'W' denote pre-preg and plain weave reinforcement, respectively. Red symbol outline denotes that values are quoted with respect to strength at lowest strain rate considered ($1,275 \text{ s}^{-1}$ for [12]). Error bars denote the range of reported sensitivity and not standard deviation. Data taken from [4, 9, 10, 12, 27, 45, 47, 50–55]

sensitive to variations in manufacturing, fibre volume fraction, defects, void content, specimen geometry, specimen preparation (flat contact surfaces) and alignment.

In summary, the literature suggests that there is likely a positive sensitivity of compression strength to strain rate but the magnitude is uncertain. The sensitivity to fibre material and architecture is also difficult to determine as a consequence of limited studies and high scatter.

4.1.3 Ultimate compressive strain

As with strength and stiffness, the influence of strain rate on failure strain is uncertain. Many authors report a slight decrease (less than 14 %) or negligible influence of strain rate on failure strains as shown in Fig. 7 [4, 9, 27, 50, 52, 53]. Song *et al.* [47] reported a negative sensitivity, but one that is much stronger than reported by most (-62 %). Pankow *et al.* [27], Shah Khan *et al.* [50] and Gama *et al.* [45] reported a negligible sensitivity to strain rate. In the study by Shah Khan *et al.* [50], only strain rates up to 10 s^{-1} were considered. The results from Gama *et al.* [45] and Pankow *et al.* [27] illustrate the effects of dispersion. Both studies show oscillations in the initial portion of the stress-strain curve. Further, strain measurements with 2D DIC by Pankow *et al.* [27] showed significant heterogeneity throughout the entire test. This was attributed to the coarse textile architecture and local variations in wave speed within the material. However, the reliability of this conclusion is questionable due to the poor spatial and temporal resolution of the measurements.

The studies mentioned above report an opposite effect of strain rate to the positive sensitivity reported by Gerlach *et al.* [46] for an RTM-6 resin tested without reinforcement. A positive sensitivity was also established by Naik *et al.* [12] (+34 % between $1,275 \text{ s}^{-1}$ and $1,503 \text{ s}^{-1}$) and Gama *et al.* [45] (up to +108 % at $1,125 \text{ s}^{-1}$). The notably higher sensitivity from Gama *et al.* [45] is likely unreliable

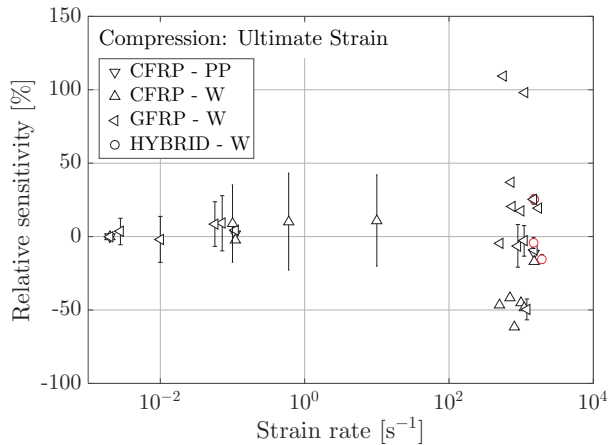


Fig. 7 Summary of relative strain rate sensitivity for ultimate compressive strain from the literature. 'PP' and 'W' denote pre-preg and plain weave reinforcement, respectively. Red symbol outline denotes that values are quoted with respect to strain at lowest strain rate considered ($1,275 \text{ s}^{-1}$ for [12] and $1,007 \text{ s}^{-1}$ for [55]). Error bars denote the range of reported sensitivity and not standard deviation. Data taken from [4, 9, 12, 27, 45, 47, 50–55]

due to significant contamination from dispersion. However, the magnitude of this sensitivity is similar to that reported by Kapoor *et al.* [57] for a Kevlar 29/polypropylene (PP) 2D woven composite (+134 % up to $4,300 \text{ s}^{-1}$). A larger magnitude of sensitivity is to be expected for the through-thickness behaviour that is dominated by a ductile thermoplastic matrix due to the stronger molecule mobility. Hosur *et al.* [54] found that failure strains increased with increasing strain rate, but were lower than quasi-static values. As discussed in Section 4.1.2, this was attributed to a change in failure mode. Further, the strain rate sensitivity cannot be assessed for intermediate strain rates in that study since the specimens did not fail. Song *et al.* [47] reported a decrease in ultimate strain between 500 s^{-1} and 800 s^{-1} , followed by an increase at strain rates above $1,000 \text{ s}^{-1}$. No explanation was provided by the authors for this behaviour.

In general, the effect of strain rate on ultimate strains has yet to be established. A number of studies reveal that the effects of dispersion may still have significant influence on the measurement of ultimate properties, which may contribute to the high levels of disparity of conclusions reported within the literature.

4.2 Strain rate effects on tensile properties

Comparatively fewer studies are available on the tensile strain rate dependency of composites in the through-thickness direction. This is because testing in tension is more complicated than it is in compression. As described in Section 2, additional challenges are introduced with specimen gripping, alignment and sensitivity to stress concentrations. Studies on resin strength in tension are also very limited and the majority of available studies are of little benefit since tensile specimens loaded with a SHPB commonly fail outside of the gauge region and thus have high scatter [46]. All of these issues tend to impose additional restrictions on the attainable strain rates using a tensile SHPB.

4.2.1 Tensile elastic modulus

Very few studies have reported measurements of high strain rate elastic modulus [6, 9, 35, 46, 58, 59]. A summary of findings from the literature is provided in Fig. 8. Lifshitz and Leber [58] used a tensile SHPB to test carbon/epoxy pre-preg and glass/epoxy woven composites. They reported a greater increase in modulus for the pre-preg (+40%) compared to the woven composite (+18 %) for strain rates up to 195 s^{-1} . Similar trends were reported by Medina and Harding [59] at strain rates up to 950 s^{-1} (+31 % for carbon/epoxy pre-preg, -13 % for glass/epoxy weave). At such high strain rates, it is likely that inertia has significant influence on the apparent modulus. Dispersion may also act to mask the true mechanical response. This is shown in longitudinal and transverse strain measurements within the specimen, which exhibit oscillations throughout the duration of the test (Fig. 3 in [59]). Comparison of carbon and glass epoxy weaves show that the tensile modulus appears to be insensitive to fibre material. Nakai and Yokoyama [6, 35] noted a substantial increase in the 'apparent' modulus, which was thought to be a result of the viscoelastic modulus of the resin. This explanation is in agreement with studies on the tensile properties of epoxy resins, which show a marked increase in apparent modulus at high strain rates [46].

Hufenbach *et al.* [60] used direct tensile loading in a two-bar SHPB configuration to test two glass/polypropylene woven composites. A very large sensitivity to strain rate (up to +500 %) and considerable scatter was observed. The large scatter led to the conclusion that the SHPB technique was unsuitable for studying coarse reinforcement architectures [60]. Hufenbach *et al.* [60] also comment on the influence of manufacturing on the measured tensile modulus. Compaction levels were reported to be much higher for L-shaped beam specimens compared to the dog bone specimens used with the SHPB. This resulted in a much higher modulus for the L-shaped beams.

In summary, the tensile modulus is a parameter that is scarcely investigated in the literature. There is some indication of a positive sensitivity to strain rate but high scatter and the lack of studies makes this difficult to say with any certainty.

4.2.2 Tensile strength

The majority of studies which focus on tensile strength utilise some form of a tensile SHPB, with strain rates generally lower than 400 s^{-1} [6, 18, 19, 35, 58, 60, 61]. A summary of relative strain rate sensitivity reported in the literature is provided in Fig. 9. While most studies report an increase in strength at high strain rates, the magnitude is unclear due to high inter and intra study scatter. Some report moderate increases in strength of around 30 % compared to quasi-static values [58, 59], while others report much more significant increases of up to several hundred percent [6, 35, 61]. In the case of Naik *et al.* [61], it appears that the reaction forces on the specimen are not equal for much of the test (not in quasi-static stress equilibrium). This is

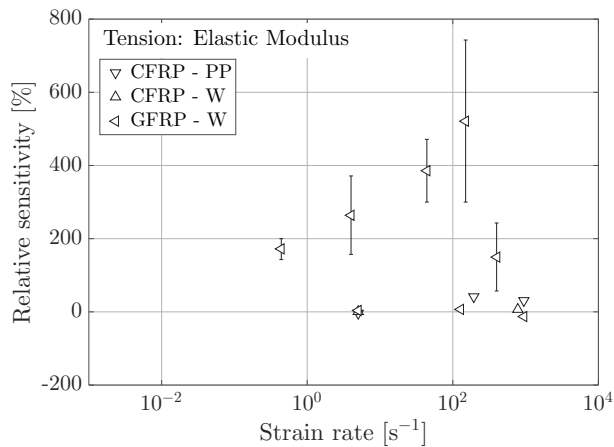


Fig. 8 Summary of relative strain rate sensitivity for tensile modulus from the literature. 'PP' and 'W' denote pre-preg and plain weave reinforcement, respectively. Error bars denote the range of reported sensitivity and not standard deviation. Data taken from [58–60]

difficult to confirm due to poor sampling of strain gauge signals from the incident and transmitter bars. In the studies by Nakai and Yokoyama [6, 35], where much lower strain rates are considered, it appears that quasi-static stress equilibrium is obtained. Those studies report large increases in strength, which is expected for a matrix-dominated property based on studies of an unreinforced epoxy resin [46].

Studies by Gerlach *et al.* [10] and Govender *et al.* [9] report higher strain rates up to $11,000 \text{ s}^{-1}$ and $1,800 \text{ s}^{-1}$, respectively. In the case of Gerlach *et al.* [10], tensile loading was induced using a customised fixture to load an overlapping joint, or 'cross' specimen, using a compressive SHPB. In that study, it is unclear how strain rate was defined, or how the effects of the custom loading fixture were accounted for. Govender *et al.* [9] used a spall test configuration to measure the tensile failure strength of a woven glass fibre-vinyl ester composite. Pulse time-shifting was employed to estimate forces in the specimen at failure. This approach removes the requirement for quasi-static stress equilibrium, allowing much higher strain rates to be achieved. However, the approach requires corrections for dispersion and the assumption of one-dimensional wave propagation through the bar and specimen. The failure location was determined post-mortem with the strength estimated using the maximum computed stress seen by that position. Some failed specimens exhibited substantial residual strength after a crack had initiated. Failure surfaces showed signs of fibre bridging, which suggests that the interlaminar failure is not purely brittle [9]. The results from this study could not be used to assess the effect of strain rate as only one strain rate was considered and no quasi-static values were reported for comparison. Quasi-static testing was unsuccessful due to consistent failure within the grips. Instead, the authors compare strength to the manufacturer's quoted resin strength. While the spall test arrangement appears promising for higher strain rate characterisation, more information must be collected during the test in order to remove the limiting assumptions of the SHPB that remain.

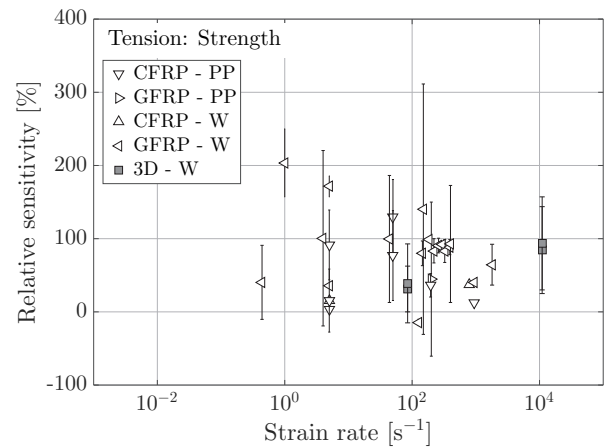


Fig. 9 Summary of relative strain rate sensitivity for tensile strength from the literature. 'PP' and 'W' denote pre-preg and plain weave reinforcement, respectively. Error bars denote the range of reported sensitivity and not standard deviation. Data taken from [6, 9, 10, 18, 19, 35, 58–61]. For reference [9] (shaded symbols) values are reported relative to the matrix properties

The scatter in reported strengths have been attributed to a number of issues associated with tensile testing. Machining defects prevented Lifshitz and Leber [58] from obtaining reliable strength measurements on their carbon/epoxy specimens. For tensile testing in the through-thickness direction, machining defects are particularly problematic as cracks can easily propagate between plies. Specimen geometry was shown to have a significant influence on the dynamic response. Using full-field measurements, Gilat *et al.* [22] showed that less than half of the gauge region on a waisted specimen was subjected to uniform stress. This results in overestimation of strain and strain rate using SHPB theory. Lifshitz and Leber [58] also had issues obtaining consistent bonds between specimen halves (machined in two pieces). This is also expected to have an effect on those studies where the specimens are directly bonded to the input bar [9, 58, 61]. Others have chosen to introduce the load by bonding the specimen onto threaded inserts [10, 35, 59, 60]. Slight misalignments between the specimen and loading axis will introduce bending stresses in the specimen, resulting in considerable scatter in measured strength values. Similar issues have been reported for high strain rate tensile testing of epoxy resins [41, 46]. Variability in the technique used to introduce the load contributes to scatter and is a result of the lack of test standards for high strain rates.

In summary, scatter across the literature is too large to conclude that there is any influence of strain rate on tensile strength.

4.2.3 Ultimate tensile strain

Similar to tensile strength, it is difficult to obtain reliable strain measurements with existing techniques. In studies performed by Gerlach *et al.*, the gauge region dimensions [19], or customised loading fixtures [10] prevented measurements of strain to be made. Some studies report values of ultimate strain at high strain rates, but do not supply

quasi-static values for comparison [58, 61]. In other studies, the focus was on obtaining measurements of strength and stiffness, and ultimate strains were not reported [9, 18, 52]. Strain gauges have been used to measure strain directly from the specimen [58, 59]. Medina and Harding [59] measured a 22 % and 65 % increase in failure strain at 950 s^{-1} for carbon/epoxy pre-preg and woven composites, respectively. This study highlights two challenges with using strain gauges. In their study, it was difficult to place the strain gauge appropriately so that it was positioned on the failure plane. A similar challenge was reported by Liftshitz and Leber [58]. An additional issue experienced by Medina and Harding [59] was premature failure of the strain gauges. The ultimate strains measured by Medina and Harding [59] at such high strain rates (950 s^{-1}) are likely to be contaminated by inertia effects. Therefore, the trends reported by Medina and Harding [59] are unlikely to be reliable representations of the true effect of strain rate on ultimate strain.

There are a few studies that report failure strains based on SHPB theory [6, 35, 60], but the reported trends are inconclusive. A summary of the failure strains presented in the literature is shown in Fig. 10. Nakai and Yokoyama performed two studies on carbon/epoxy pre-preg composites [6, 35]. They reported a general increase in failure strain for strain rates up to approximately 100 s^{-1} . The reported magnitude is unreliable due to the high scatter in their measurements (strain rate sensitivity ranges from -50 % to +90 % within scatter). Hufenbach *et al.* [60] reported that failure strain has little sensitivity to strain rate between quasi-static and strain rates up to 400 s^{-1} .

In summary, obtaining reliable strain measurements at high strain rates has proven to be a challenge. As a result, very few studies report ultimate strains. From those that attempt to measure ultimate strains, the influence of strain rate cannot be determined with certainty.

The limited available literature suggests there is much outstanding work to be done to develop testing procedures to yield reliable through-thickness tensile properties. Tensile tests are highly sensitive to gripping, alignment and stress concentrations. This does not facilitate reliable measurement of material properties using existing techniques, which rely on a number of assumptions about the material response. Further, not only is there significant scatter in the measured high strain rate properties, but there is high scatter in the quasi-static measurements to which they are compared. A potential alternative to strain gauges is optical full-field measurement techniques. Full-field measurements provide far more information about the specimen response, which is required to alleviate assumptions and advance high strain rate tensile characterisation.

4.3 Strain rate effects on interlaminar shear properties

4.3.1 Shear modulus

The combined stress states induced in many existing shear tests complicates the matter of obtaining a true estimate for

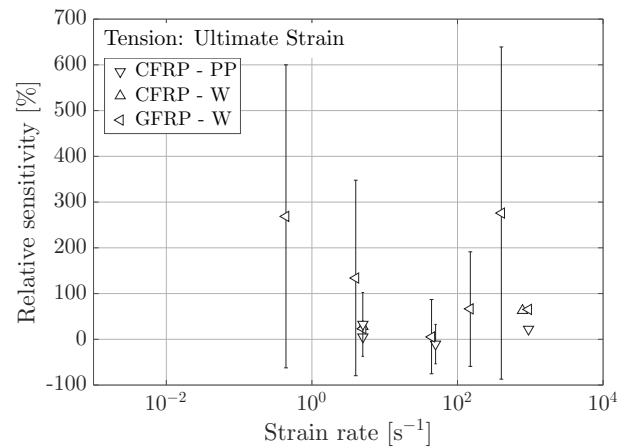


Fig. 10 Summary of relative strain rate sensitivity for ultimate tensile strain from the literature. 'PP' and 'W' denote pre-preg and plain weave reinforcement, respectively. Error bars denote the range of reported sensitivity and not standard deviation. Data taken from [6, 35, 59, 60]

the shear modulus. Therefore, far fewer studies attempt to extract the interlaminar shear modulus, compared to the interlaminar shear strength (ILSS). A summary of relative strain rate sensitivity for shear modulus is presented in Fig. 11.

Bouette *et al.* [39] reported no appreciable variation in shear modulus measured using double and single lap joints. The modulus was estimated using failure stress and strain assuming linear elastic behaviour to failure. However, strain gauge measurements by Bouette *et al.* [39] reveal a non-linear behaviour near failure, and would act to reduce the 'apparent modulus'. Hallett *et al.* [40] also used single lap specimens and reported an average decrease in stiffness with increasing strain rate. The magnitude of sensitivity is difficult to determine as their measurements had significant variance. The authors suggested that normal stresses at the notches were responsible for the scatter. Using thin-walled tubular specimens, Naik *et al.* [5] reported a net increase in apparent shear modulus (+33 % for glass/epoxy and +41 % for carbon epoxy) up to $1,000 \text{ s}^{-1}$. Similar to the study by Bouette *et al.* [39], the shear response was found to be non-linear by Naik *et al.* [5]. This makes it difficult to obtain an estimate of the true shear modulus. Naik *et al.* [5] provided an estimate on the lower bound of shear modulus by fitting a line between the origin and a point on the curve near peak stress. Naik *et al.* [5] also claimed that the values obtained with the tubular specimens are more representative of the true behaviour since notch effects are eliminated. Interestingly, this did not seem to have a significant effect on the level of scatter in their measurements. Scatter in properties measured using tubular specimens may be attributed to micro cracks/damage induced from machining [5], specimen geometry (fillet radius, wall thickness) [5, 62], and layup orientation [62].

In summary, the effect of strain rate on shear modulus is not clearly understood. This is primarily due to mixed stress states in the specimen leading to biased estimates of shear modulus, as well as inherent limitations of the Hopkinson

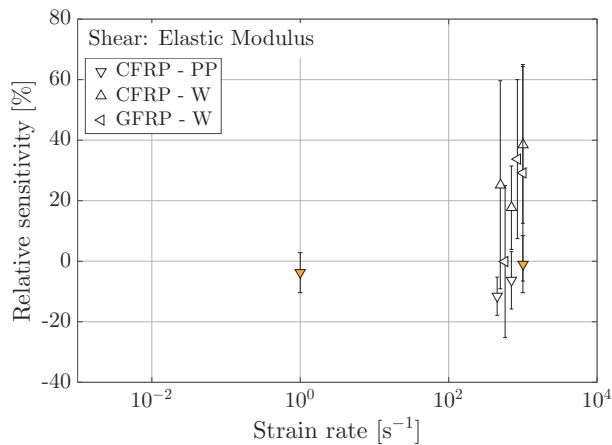


Fig. 11 Summary of relative strain rate sensitivity for shear modulus strain from the literature. 'PP' and 'W' denote pre-preg and plain weave reinforcement, respectively. Orange and purple symbols denote testing in the 1-3 and 2-3 directions, respectively. White denotes that direction is not specified. Error bars denote the range of reported sensitivity and not standard deviation. Data taken from [5, 39, 40]

bar analysis as for the tensile modulus.

4.3.2 Shear strength

The through-thickness response of composites is particularly sensitive to combined states of tension and shear. When tensile stresses are present, the interlaminar shear strength has shown to decrease significantly [3, 63]. It is very difficult to achieve the desired state of pure shear in quasi-static and high strain rate testing and therefore, most studies are limited to reporting 'apparent shear strength' [2, 7, 16]. The influence of strain rate on shear strength, based on studies from the literature, is illustrated in Fig. 12.

Single and double lap specimens were initially the most popular choices for high strain rate shear testing since they could be tested using a compressive SHPB. Bouette *et al.* [39], Harding and Li [64] and Harding and Dong [3] used double lap specimens to extract the ILSS. These studies found that lap joint specimens are unreliable since failure initiates near the ends of the overlap under a state of combined shear and normal stresses. The same was found for single-lap specimens [3, 39, 40]. The work with single lap specimens shows that the amplitude of stress concentration can be reduced if the overlap is kept small [39, 40]. While gauge region geometry can be altered to reduce the combined stress state it cannot be eliminated. This is a likely contributor to the high levels of experimental scatter on reported shear modulus and failure stresses.

Other approaches to measuring the interlaminar shear properties include the use of the double V-notch shear test (Iosipescu) [21], the out-of-plane off-axis tests [2], the double notched shear test in dynamic compression [65], short beam shear tests [65], and thin-walled tubular specimens loaded using a torsional SHPB [5, 62]. Yokoyama and Nakai [65] found that strengths obtained with the double-notched shear specimens compared well with the short-beam shear tests. Further, measured ILSS values agree well with sim-

ulated stress levels in the centre of the specimens. This lead them to conclude that the influences of the stress concentrations from the notches were negligible. Further, the compressive normal stresses at the notches acted to suppress delamination. The result was much lower levels of scatter compared to previous studies using a single or double lap specimen. Hufenbach *et al.* [21] used a lightweight Iosipescu fixture for testing at intermediate strain rates up to approximately 60 s^{-1} [21]. This appeared to work reasonably well for testing in the 2-3 plane, but yielded unacceptable levels of scatter in strength measurements for the 1-3 plane and thus, the strain rate sensitivity could not be determined. This was attributed to low stress and strain levels and measurement resolution of the load cell. Naik *et al.* [5] compared thin-walled tubular specimens, loaded with a torsional SHPB, to single-lap specimens loaded in dynamic compression using a SHPB. Tubular specimens were selected for detailed analysis as they created a purer state of shear stress. The level of scatter from single lap specimens was similar but produced slightly lower strength values. This is likely a result of the combined tension-shear stress state. The opposite was found by Gowtham *et al.* [62] who reported lower strengths measured using a torsional SHPB compared to a single lap shear joint. This was attributed to stress concentrations from the weave reinforcement in the failure plane of the tubular specimens. The weave reinforcement was shown to create variations in stiffness and stress along the radial and circumferential directions, acting as stress concentrations. This violates one of the primary assumptions that stress is uniform throughout the thickness.

A common characteristic to many of these studies is the high levels of scatter, which prohibits definitive claims from being made about the strain rate sensitivity [10, 21, 40]. Qualitative trends may still be useful, and there appears to be more of a general agreement between studies in these trends. Many studies report that the interlaminar shear strength increases moderately with increasing strain rate [2, 3, 5, 62, 65]. The level of strain rate sensitivity has yet to be determined reliably, with reports of strength increases ranging from 15% [62] to 200% [62]. There is also a collection of studies that report a constant strength [39, 60, 65].

In summary, there is some indication of a positive influence of strain rate on interlaminar shear strength. However, the presence of combined tension and shear stresses, and stress concentrations have prevented reliable characterisation of the high strain rate behaviour.

4.3.3 Ultimate shear strain

In the majority of cases, it appears that strain was either not reliably measured [21], or the focus of the study was on shear strength and thus, strains at failure were not reported [2, 62, 65]. Some fixtures and specimen geometries make shear strain measurements challenging or impossible [10]. Of the remaining studies, there is very little agreement as to the effect of strain rate on failure strain. The variation in strain rate sensitivity across the literature is presented in

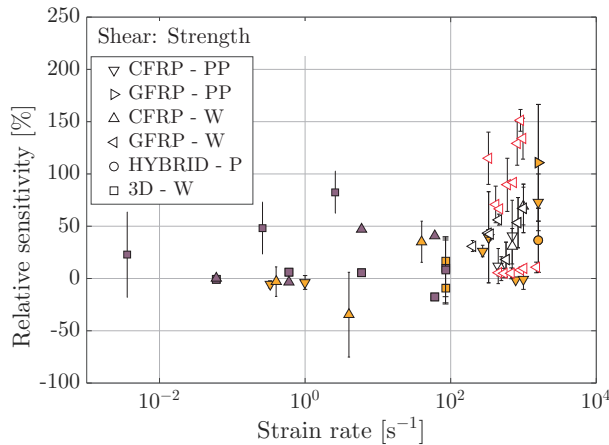


Fig. 12 Summary of relative strain rate sensitivity for shear strength from the literature. ‘PP’ and ‘W’ denote pre-preg and plain weave reinforcement, respectively. Red symbol outline denotes that values are quoted with respect to strain at lowest strain rate considered (260 s^{-1} for [2], 0.04 s^{-1} for [21] and 300 s^{-1} for [62]). Orange and purple symbols denote testing in the 1-3 and 2-3 directions, respectively. White denotes that direction is not specified. Error bars denote the range of reported sensitivity and not standard deviation. Data taken from [2, 3, 5, 10, 21, 39, 40, 60, 62, 64, 65]

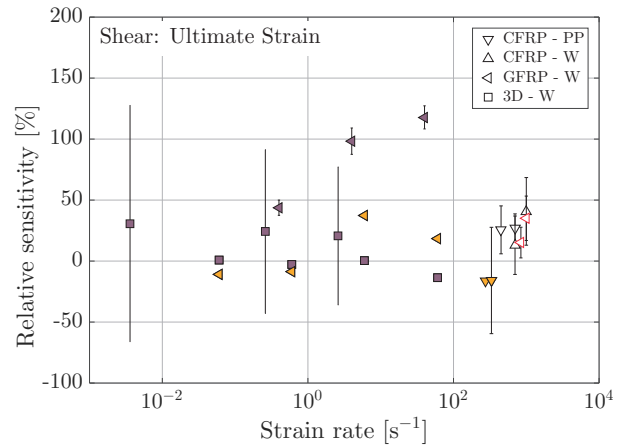


Fig. 13 Summary of relative strain rate sensitivity for ultimate shear strain from the literature. ‘PP’ and ‘W’ denote pre-preg and plain weave reinforcement, respectively. Red symbol outline denotes that values are quoted with respect to strain at lowest strain rate considered (496 s^{-1} and 576 s^{-1} for [5], and 0.04 s^{-1} for [21]). Orange and purple symbols denote testing in the 1-3 and 2-3 directions, respectively. White denotes that direction is not specified. Error bars denote the range of reported sensitivity and not standard deviation. Data taken from [3, 5, 21, 40, 60]

Fig. 13. Two studies by Bouette *et al.* [39] and Hufenbach *et al.* [60] report that shear strain at failure is independent of strain rate, up to $1,000 \text{ s}^{-1}$ and 400 s^{-1} , respectively. The lack of sensitivity to strain rate may be a shortcoming of lap specimens with failure being heavily influenced by normal stresses at the ends of the overlap, similar to Hallett *et al.* [40]. In contrast, Gillespie *et al.* [2] reported an increase in peak strain up to a strain rate of 582 s^{-1} , followed by a reduction at higher strain rates. These higher strain rates approach the upper limit for equilibrium, set by the ‘R criterion’, and are likely contaminated by inertia effects. Harding and Li [64] found a significant increase (approx. 250 %) in failure strains with impact speed. Unfortunately, the strain rates at failure were not reported. Naik *et al.* [5] measured a 38 % increase in failure strain between strain rates of $576 - 1,000 \text{ s}^{-1}$, but did not provide quasi-static reference values for comparison.

Much like the cases of shear modulus and shear strength, ultimate shear strains presented in the literature are heavily influenced by mixed stress states and stress concentrations. As a result the current literature cannot be used to determine the influence of strain rate on ultimate shear strain.

5. Advanced Testing Approaches Using High Speed, Full-Field Measurements

A review of the literature highlights several key limitations with existing test methods that hinder the advancement of material characterisation for composites in the through-thickness direction. The assumptions required for a SHPB test are particularly restrictive to the maximum strain rate that can be obtained. In compression, strain rates are limited generally to less than $1,500 \text{ s}^{-1}$, and to an even greater extent in tension (typically less than 100 s^{-1}). This is due to the assumption that the specimen is subjected to a state of uniform stress. This is not the case, especially in early stages

of a test where inertia effects create heterogeneous deformation [22, 27]. For example, Pankow *et al.* [27] used high speed imaging in an attempt to resolve full-field strain profiles on a specimen subjected to compression with a SHPB. Despite poor quality measurements and lack of temporal resolution, they found that the strain field was highly non-uniform. This was also found by Gilat *et al.* [22]. Gilat *et al.* [22] and Nakai and Yokoyama [35] note that SHPB theory overestimates strain for waisted specimens due to heterogeneous deformation. Strain gauge measurements also become highly sensitive to gauge position and are generally unreliable for obtaining measurements of ultimate strain [58, 59].

Govender *et al.* [9] configured a Hopkinson bar into a spall test in an attempt to remove some of the restrictive assumptions of the SHPB and measure tensile strength. The spall test, commonly used for concrete [37], is used sparingly in the literature for through-thickness testing. The spall test approach was also used by Gerlach *et al.* [46] for high strain rate tensile testing of epoxy resins. The approach of Govender *et al.* [9] utilised phase shifting of the waves measured on the bar to infer the stress in the material at failure. High speed cameras were required to record the approximate time and location of the failure. The major shortcoming of the spall test approach used by Govender *et al.* [9] is that the input stress is inferred from measurements on the incident bar. Tensile strength is also inferred based on measurements of the global response (reflected pulse measured in the input bar), which is affected by dispersion.

In the studies by Govender *et al.* [9], Pankow *et al.* [27] and Gilat *et al.* [22], the temporal resolution is insufficient to properly resolve the initial response of the material. In these cases, common high speed cameras were used, which can achieve frame rates on the order of a 1.5×10^5 frames per second. The advantage of these cameras is that they offer

relatively long record times, at the expense of frame rates (see Fig. 1 in [66]¹). Moreover, as frame rate increases, the spatial resolution decreases due to memory read out limitations. Therefore, to achieve the necessary frame rates for quantitative imaging at high strain rates, ultra high speed cameras are required. These cameras use different strategies to overcome memory read out issues. An example of this is the 'in-situ storage charge coupled device' (IS-CCD), in which the memory for each pixel is located on the sensor [28]. These cameras offer significant opportunity for dynamic material characterisation as they are simple to operate and trigger, can be used for stereo imaging, and offer framing rates up to 5×10^6 frames per second.

An issue with testing brittle materials (*i.e.*: through-thickness tension for a composite) is small strains to failure. This requires high spatial/temporal resolution and low noise. With the development of ultra high speed cameras and full-field measurement techniques, such as DIC [67] or the grid method [68], some of the fundamental assumptions attached to current test methods may be alleviated. This offers great potential to improve current test methods or develop new techniques with kinematic fields that are not necessarily uniform, or are intentionally non-uniform, as also proposed in [22]. With the ability to resolve the temporal evolution in surface displacement, and hence acceleration, the specimen's acceleration field may be used as an embedded load cell (see Eq. (1)). These full-field maps may be processed using an inverse identification technique, such as the Virtual Fields Method (VFM) [24, 25, 69] to reconstruct stress and extract material properties without need for measurement of the external force. This removes the requirement for stress equilibrium and uniform uniaxial strain states attached to SHPB testing. In fact, the presence of a heterogeneous strain field during impact may be beneficial as multiple constitutive properties could be extracted from a single test using the VFM. This may be used to overcome some of the limitations of current test methods for shear characterisation as the combined stress state could be characterised and potentially used for robust and accurate stiffness and strength identification.

Very encouraging results have been obtained recently, demonstrating the potential of this approach for identifying the in-plane strength and stiffness of composite laminates at very high strain rates ($> 2,000 \text{ s}^{-1}$) [23, 29]. Moulart *et al.* [23] used the SHPB to generate a pulse with full-field maps processed using the VFM to identify Young's modulus and Poisson's ratio for a quasi-isotropic laminate. Zhu [29], Pierron *et al.* [25], and Pierron and Forquin [24] have demonstrated that this approach can be extended to spall test configurations to extract the stiffness and strength of brittle materials. This approach opens up a wide design space to design innovative tests to accurately determine the mechanical response of materials at strain rates not achievable with any current technique, and in particular, offers a

very promising alternative for high strain rate testing in the through-thickness direction for polymer matrix fibre composites.

6. Conclusions

Understanding the influence of strain rate on the through-thickness mechanical properties of polymer matrix fibre composites is critical for the design of thick structures, or structures subjected to dynamic loading in the through-thickness direction. The split Hopkinson pressure bar has served as the primary tool for measuring high strain rate material properties. However, this approach relies on limited experimental information and suffers from a number of strong inherent assumptions. The result is poor consistency across the literature regarding the high strain rate response of fibre-reinforced polymer composites in the through-thickness direction. Part of this can be attributed to inconsistency in the literature regarding material composition (fibre and matrix materials, fibre volume fraction, reinforcement architecture, etc.), and the way that strain rate is reported.

Available studies on the through-thickness properties suggest that a positive sensitivity to strain rate may be exhibited by the compressive modulus, compressive ultimate strength, and tensile modulus. However, the magnitude of sensitivity is uncertain due to large amounts of scatter between and within studies. Further, the degree to which inertia influences the stress and strain measurements is difficult to determine and therefore, measurements of the modulus using the SHPB can only be regarded as 'apparent'. Literature shows that the influence of strain rate cannot be determined with certainty for ultimate strains in tension, compression and shear. In the case of shear properties, reported values are heavily affected by combined states of tension and shear stress. The effect of tensile normal stresses are amplified by manufacturing defects and geometric features, resulting in unreliable strength and strain values. Comparatively, greater uncertainty surrounds the influence of strain rate on tensile properties. In this case, high scatter is problematic for both high strain rate and quasi-static measurements. These tests are also particularly sensitive to load alignment, gripping technique and stress concentrations resulting from geometrical features of the specimen or machining defects.

The development of high speed cameras and full-field measurement techniques, such as DIC or the grid method, offer a promising outlook for high strain rate testing. Full-field measurements enable many fundamental assumptions attached to current test methods to be alleviated. This creates opportunity to improve current test methods and/or develop new techniques, which induce heterogeneous strains, to identify mechanical properties using inverse identification procedures. The authors believe that use of full-field measurements, in current or new test methods, will lead to a more consistent understanding of the effects of strain rate on the through-thickness properties of composites.

¹updated image can be found at <http://photodyn.org/tools/ultra-high-speed-camera>

Acknowledgements

This material is based on research sponsored by the Air Force Research Laboratory, under agreement number FA9550-17-1-0133. The U.S. Government is authorized to reproduce and distribute reprints for Governmental purposes notwithstanding any copyright notation thereon. The views and conclusions contained herein are those of the authors and should not be interpreted as necessarily representing the official policies or endorsements, either expressed or implied, of the Air Force Research Laboratory or the U.S. Government.

Mr Jared Van Blitterswyk acknowledges the support of EP-SRC for funding through a Doctoral Training Grant. Dr Lloyd Fletcher and Prof. Fabrice Pierron acknowledge support from EPSRC through grant EP/L026910/1. The authors are also grateful to the grant programme manager, Dr David Garner from EOARD/AFOSR.

References

- [1] Daniel, I. M., Werner, B. T. and Fenner, J. S.: Strain-rate-dependent failure criteria for composites, *Composites Science and Technology*, **71**-3 (2011), 357–364.
- [2] Gillespie, J. W., Gama, B. A., Cichanowski, C. E. and Xiao, J. R.: Interlaminar shear strength of plain weave S2-glass/SC79 composites subjected to out-of-plane high strain rate compressive loadings, *Composites Science and Technology*, **65**-11-12 (2005), 1891–1908.
- [3] Harding, J. and Dong, L.: Effect of strain rate on the interlaminar shear strength of carbon-fibre-reinforced laminates, *Composites Science and Technology*, **51**-3 (1994), 347–358.
- [4] Yokoyama, T. and Nakai, K.: High strain-rate compressive characteristics of laminated composites in the through-thickness direction, *CD-ROM Proc. SEM X International Congress & Exposition on Experimental & Applied Mechanics*, (2004).
- [5] Naik, N. K., Asmelash, A., Kavala, V. R. and Ch, V.: Interlaminar shear properties of polymer matrix composites: strain rate effect, *Mechanics of Materials*, **39**-12 (2007), 1043–1052.
- [6] Nakai, K. and Yokoyama, T.: Through-thickness tensile strength of carbon/epoxy laminated composites under impact loading, *CD-ROM Proc. 16th International Conference on Experimental Mechanics*, (2014).
- [7] Olsson, R.: A survey of test methods for multiaxial and out-of-plane strength of composite laminates, *Composites Science and Technology*, **71**-6 (2011), 773–783.
- [8] Mespoulet, S.: Through-thickness test methods for laminated composite materials, PhD thesis, Imperial College of Science, Technology and Medicine, London, UK, (1998).
- [9] Govender, R. A., Louca, L. A., Pullen, A., Fallah, A. S. and Nurick, G. N.: Determining the through-thickness properties of thick glass fiber reinforced polymers at high strain rates, *Journal of Composite Materials*, **46**-10 (2012), 1219–1228.
- [10] Gerlach, R., Siviour, C. R., Wiegand, J. and Petrinic, N.: In-plane and through-thickness properties, failure modes, damage and delamination in 3D woven carbon fibre composites subjected to impact loading, *Composites Science and Technology*, **72**-3 (2012), 397–411.
- [11] Field, J. E., Walley, S. M., Proud, W. G., Goldrein, H. T. and Siviour, C. R.: Review of experimental techniques for high rate deformation and shock studies, *International Journal of Impact Engineering*, **30**-7 (2004), 725–775.
- [12] Naik, N. K., Ch, V. and Kavala, V. R.: Hybrid composites under high strain rate compressive loading, *Materials Science and Engineering A*, **498**-1-2 (2008), 87–99.
- [13] Broughton, W. R.: Through-thickness testing, *Mechanical Testing of Advanced Fibre Composites* (Hodgkinson, J. M. ed.), Woodhead Publishing Limited (2000), chapter 8.
- [14] He, Y., Makeev, A. and Shonkwiler, B.: Characterization of nonlinear shear properties for composite materials using digital image correlation and finite element analysis, *Composites Science and Technology*, **73** (2012), 64–71.
- [15] Cui, W., Liu, T., Len, J. and Ruo, R.: Interlaminar tensile strength (ILTS) measurement of woven glass/polyester laminates using four-point curved beam specimen, *Composites Part A: Applied Science and Manufacturing*, **27**-11 (1996), 1097–1105.
- [16] Makeev, A., Carpentier, P. and Shonkwiler, B.: Methods to measure interlaminar tensile modulus of composites, *Composites Part A: Applied Science and Manufacturing*, **56** (2014), 256–261.
- [17] Charrier, J. S., Laurin, F., Carrere, N. and Mahdi, S.: Determination of the out-of-plane tensile strength using four-point bending tests on laminated L-angle specimens with different stacking sequences and total thicknesses, *Composites Part A: Applied Science and Manufacturing*, **81** (2016), 243–253.
- [18] Hufenbach, W., Hornig, A., Zhou, B., Langkamp, A. and Gude, M.: Determination of strain rate dependent through-thickness tensile properties of textile reinforced thermoplastic composites using L-shaped beam specimens, *Composites Science and Technology*, **71**-8 (2011), 1110–1116.
- [19] Gerlach, R., Siviour, C. R., Wiegand, J. and Petrinic, N.: The strain rate dependent material behavior of S-GFRP extracted from GLARE, *Mechanics of Advanced Materials and Structures*, **20**-7 (2013), 505–514.
- [20] Davis, J. R.: High strain rate tensile testing, *Tensile Testing* (2nd ed.), ASM International (2004), 251–263.
- [21] Hufenbach, W., Langkamp, A., Hornig, A. and Ebert, C.: Experimental determination of the strain rate dependent out-of-plane shear properties of textile-reinforced composites, *CD-ROM Proc. 17th European Conference on Composite Materials*, (2009).
- [22] Gilat, A., Schmidt, T. E. and Walker, A. L.: Full field strain measurement in compression and tensile split Hopkinson bar experiments, *Experimental Mechanics*, **49**-2 (2009), 291–302.

- [23] Moulart, R., Pierron, F., Hallett, S. R. and Wisnom, M. R.: Full-field strain measurement and identification of composites moduli at high strain rate with the virtual fields method, *Experimental Mechanics*, **51**-4 (2011), 509–536.
- [24] Pierron, F. and Forquin, P.: Ultra-high-speed full-field deformation measurements on concrete spalling specimens and stiffness identification with the virtual fields method, *Strain*, **48**-5 (2012), 388–405.
- [25] Pierron, F., Zhu, H. and Siviour, C. R.: Beyond Hopkinson's bar, *Philosophical Transactions of the Royal Society A*, **372** (2014), 20130195.
- [26] Koeber, H., Xavier, J. and Camanho, P. P.: High strain rate characterisation of unidirectional carbon-epoxy IM7-8552 in transverse compression and in-plane shear using digital image correlation, *Mechanics of Materials*, **42**-11 (2010), 1004–1019.
- [27] Pankow, M., Salvi, A., Waas, A. M., Yen, C. F. and Ghiorse, S.: Split Hopkinson pressure bar testing of 3D woven composites, *Composites Science and Technology*, **71**-9 (2011), 1196–1208.
- [28] Etoh, T. G. and Mutoh, H.: An image sensor of 1 Mfps with photon counting sensitivity, *Proc. 26th International Congress on High-Speed Photography and Photonics* (2005), 301–307.
- [29] Zhu, H.: A novel methodology for high strain rate testing using full-field measurements and the virtual fields methods, PhD thesis, University of Technology of Troyes, France, (2015).
- [30] Grédiac, M., Blaysat, B. and Sur, F.: A critical comparison of some metrological parameters characterizing local digital image correlation and grid method, *Experimental Mechanics*, **57**-6 (2017), 871–903.
- [31] Taylor, G. I.: The testing of materials at high rates of Loading, *Journal of the Institution of Civil Engineers*, **26**-8 (1946), 486–519.
- [32] Kolsky, H.: An investigation of the mechanical properties of materials at very high rates of loading, *Proceedings of the Physical Society B*, **62**-11 (1949), 676–700.
- [33] Gray, G. T. III: Mechanical testing and evaluation, *ASM Handbook*, **8**, ASM International (2000), 462–476.
- [34] Gama, B. A., Lopatnikov, S. L. and Gillespie, J. W.: Hopkinson bar experimental technique: a critical review, *Applied Mechanics Reviews*, **57**-4 (2004), 223–250.
- [35] Nakai, K. and Yokoyama, T.: Dynamic stress-strain properties of carbon/epoxy laminated composites in through-thickness direction: tension and compression, *Proc. 9th International Symposium on Impact Engineering*, (2016).
- [36] Gerlach, R., Kettenbeil, C. and Petrinic, N.: A new split Hopkinson tensile bar design, *International Journal of Impact Engineering*, **50** (2012), 63–67.
- [37] Novikov, S. A. and Chernov, A. V.: Determination of the spall strength from measured values of the specimen free surface velocity, *Journal of Applied Mechanics and Technical Physics*, **23**-5 (1982), 703–705.
- [38] Chen, W. W.: Experimental methods for characterizing dynamic response of soft materials, *Journal of Dynamic Behavior of Materials*, **2**-1 (2016), 2–14.
- [39] Bouette, B., Cazeneuve, C. and Oytana, C.: Effect of strain rate on interlaminar shear properties of carbon/epoxy composites, *Composites Science and Technology*, **45**-4 (1992), 313–321.
- [40] Hallett, S. R., Ruiz, C. and Harding, J.: The effect of strain rate on the interlaminar shear strength of a carbon/epoxy cross-ply laminate: comparison between experiment and numerical prediction, *Composites Science and Technology*, **59**-5 (1999), 749–758.
- [41] Gilat, A., Goldberg, R. K. and Roberts, G. D.: Strain rate sensitivity of epoxy resin in tensile and shear loading, *Journal of Aerospace Engineering*, **20**-2 (2007), 75–89.
- [42] Vecchio, K. S. and Jiang, F.: Improved pulse shaping to achieve constant strain rate and stress equilibrium in split-Hopkinson pressure bar testing, *Metallurgical and Materials Transactions A: Physical Metallurgy and Materials Science*, **38**-11 (2007), 2655–2665.
- [43] Li, Z. and Lambros, J.: Determination of the dynamic response of brittle composites by the use of the split Hopkinson pressure bar, *Composites Science and Technology*, **59**-7 (1999), 1097–1107.
- [44] Ravichandran, R. and Ghatuparthi, S.: Critical appraisal of limiting strain rates for compression testing of ceramics in a split Hopkinson pressure bar, *Journal of the American Ceramic Society*, **77**-1 (1994), 263–267.
- [45] Gama, B. A., Gillespie, J. W., Mahfuz, H., Raines, R. P., Haque, A., Jeelani, S., Bogetti, T. A. and Fink, B. K.: High strain-rate behavior of plain-weave S-2 glass/vinyl ester composites, *Journal of Composite Materials*, **35**-13 (2001), 1201–1228.
- [46] Gerlach, R., Siviour, C. R., Petrinic, N. and Wiegand, J.: Experimental characterisation and constitutive modelling of RTM-6 resin under impact loading, *Polymer*, **49**-11 (2008), 2728–2737.
- [47] Song, Z., Wang, Z., Ma, H. and Xuan, H.: Mechanical behavior and failure mode of woven carbon/epoxy laminate composites under dynamic compressive loading, *Composites: Part B*, **60** (2014), 531–536.
- [48] Gorham, D. A., Griffiths, L. J., Martin, D. J., Billington, E. W., Brissenden, C., Signoret, C., Pouyet, J. M. and Lataillade, J.-L.: Specimen inertia in high strain-rate compression, *Journal of Physics D: Applied Physics*, **22**-12 (1989), 1888–1893.
- [49] Zhao, H. and Gary, G.: On the use of SHPB techniques to determine the dynamic behavior of materials in the range of small strains, *International Journal of Solids and Structures*, **33**-23 (1996), 3363–3375.
- [50] Shah Khan, M. Z. and Simpson, G.: Mechanical properties of a glass reinforced plastic naval composite material under increasing compressive strain rates, *Materials Letters*, **45**-3-4 (2000), 167–174.
- [51] Shen, L., Li, Y. and Wang, Z.: Experimental

- investigation of the effect of strain rate on the compression behavior of 3D E-glass fiber-reinforced composites, *Applied Mechanics and Materials*, **174-177** (2012), 1528–1532.
- [52] Akil, Ö., Yıldırım, U., Güden, M. and Hall, I. W.: Effect of strain rate on the compression behaviour of a woven fabric S2-glass fiber reinforced vinyl ester composite, *Polymer Testing*, **22-8** (2003), 883–887.
- [53] Güden, M., Yıldırım, U. and Hall, I. W.: Effect of strain rate on the compression behavior of a woven glass fiber/SC-15 composite, *Polymer Testing*, **23-6** (2004), 719–725.
- [54] Hosur, M. V., Alexander, J., Vaidya, U. K. and Jeelani, S.: High strain rate compression response of carbon/epoxy laminate composites, *Composite Structures*, **52-3-4** (2001), 405–417.
- [55] Woo, S.-C. and Kim, T.-W.: High strain-rate failure in carbon/Kevlar hybrid woven composites via a novel SHPB-AE coupled test, *Composites Part B*, **97** (2016), 317–328.
- [56] Tagarielli, V. L., Minisgallo, G., Mcmillan, A. J. and Petrinic, N.: The response of a multi-directional composite laminate to through-thickness loading, *Composites Science and Technology*, **70-13** (2010), 1950–1957.
- [57] Kapoor, R., Pangen, L., Bandaru, A. K., Ahmad, S. and Bhatnagar, N.: High strain rate compression response of woven Kevlar reinforced polypropylene composites, *Composites Part B*, **89** (2016), 374–382.
- [58] Lifshitz, J. M. and Leber, H.: Response of fiber-reinforced polymers to high strain-rate loading in interlaminar tension and combined tension/shear, *Composites Science and Technology*, **58-6** (1998), 987–996.
- [59] Medina, J. L. and Harding, J.: The effect of strain rate on the through-thickness tensile stiffness and strength properties of fibre-reinforced epoxy composites, *Journal de Physique IV France*, **10** (2000), 275–280.
- [60] Hufenbach, W., Langkamp, A., Gude, M., Ebert, C., Hornig, A., Nitschke, S. and Böhm, H.: Characterisation of strain rate dependent material properties of textile reinforced thermoplastics for crash and impact analysis, *Procedia Materials Science*, **2** (2013), 204–211.
- [61] Naik, N. K., Yernamma, P., Thoram, N. M., Gadipatri, R. and Kavala, V. R.: High strain rate tensile behavior of woven fabric E-glass/epoxy composite, *Polymer Testing*, **29-1** (2010), 14–22.
- [62] Gowtham, H. L., Pothnis, J. R., Ravikumar, G. and Naik, N. K.: Dependency of dynamic interlaminar shear strength of composites on test technique used, *Polymer Testing*, **42** (2015), 151–159.
- [63] Cui, G. Y. and Ruiz, C.: Through-thickness failure of laminated carbon/epoxy composites under combined stresses, *Composites Science and Technology*, **53-3** (1995), 253–258.
- [64] Harding, J. and Li, Y. L.: Determination of interlaminar shear strength for glass/epoxy and carbon/epoxy laminates at impact rates of strain, *Composites Science and Technology*, **45-2** (1992), 161–171.
- [65] Yokoyama, T. and Nakai, K.: Evaluation of interlaminar shear strength of a unidirectional carbon/epoxy laminated composite under impact loading, *Journal de Physique IV France*, **134** (2006), 797–804.
- [66] Reu, P. L. and Miller, T. J.: The application of high-speed digital image correlation, *The Journal of Strain Analysis for Engineering Design*, **43-8** (2008), 673–688.
- [67] Sutton, M. A., Orteu, J.-J. and Schreier, H. W.: *Image Correlation for Shape, Motion and Deformation Measurements*, Springer (2009).
- [68] Grédiac, M., Sur, F. and Blaysat, B.: The grid method for in-plane displacement and strain measurement: a review and analysis, *Strain*, **52-3** (2016), 205–243.
- [69] Pierron, F. and Grédiac, M.: *The Virtual Fields Method: Extracting Constitutive Mechanical Parameters from Full-Field Deformation Measurements*, Springer (2012).

Appendix A: Literature Survey Tables

Table 1 Summary of published studies on strain rate effects of interlaminar compressive properties of fibre-reinforced polymer composites. Notes: Constituent materials listed in the format: 'fibre/matrix'. Material labels are included in '()' for association with Fig. 5 - Fig. 7. Quasi-static: italic typeface; high strain rate: regular typeface; test method: electro-mechanical load frame (HMF), hydraulic load frame (HLF), high-speed load frame (HSLF), split Hopkinson pressure bar (SHPB); specimen geometry (in '[]'): R = rectangular/cubic specimens, C = cylindrical specimens

Reference	Material	Test [Specimen]	Strain Rate (s^{-1})	Notes
Yokoyama T. & Nakai K. (2004) [4]	carbon/epoxy pre-preg. (10/90) _s : T700S/2500 (CFRP-PP) carbon/epoxy 2D plain weave: T300B/2500 (CFRP-W) glass/epoxy 2D plain weave: E-2/2500 (GFRP-W)	<i>EMLF</i> (<i>Instron 4505</i>) [C-stack] Compression SHPB [C]	0.002 – 1,510	Dynamic stress equilibrium is not achieved during the early stages of the test. Authors study influence of strain rate on secant modulus instead. Reinforcement architecture has a greater influence on the compressive properties with exception to compressive strength. Strain rate sensitivity attributed to viscoelastic nature of the epoxy resin. carbon/epoxy pre-preg: increase in elastic modulus (+105%), decrease in failure strength (-30%), and failure strain (-11%) at 1,510 s^{-1} ; carbon/epoxy 2D weave: increase in elastic modulus (+143%), decrease in strength (-6%), and decrease in ultimate strain (-17%) at 1,510 s^{-1} ; glass/epoxy 2D weave: increase in elastic modulus (+60%), increase in strength (+15%), and decrease in ultimate strain (-9%) at 1,510 s^{-1} .
Song Z. <i>et al.</i> (2014) [47]	carbon/epoxy 2D satin weave: T300-3/- (GFRP-W)	<i>HLF (MTS 810)</i> [C] Compression SHPB [C]	500 – 1,100	High strain rate stress-strain curves are affected by dispersion. Consistency of test questionable as strength decreases significantly only at 800 s^{-1} . No explanation for this behaviour is provided. Elastic modulus increases (qualitatively) up to 1,100 s^{-1} , and strength decreases up to 800 s^{-1} (-37%) before increasing at 1,100 s^{-1} (+11%). Ultimate strains decrease between -62% at 800 s^{-1} and -42% at 1,100 s^{-1} .
Naik N.K. <i>et al.</i> (2008) [12]	carbon/epoxy 2D satin weave: -/- (CFRP-W) E-glass/epoxy 2D plain weave: -/- (GFRP-W) E-glass/epoxy 2D satin weave: -/- (GFRP-W) carbon 2D satin weave & E-glass 2D plain weave/epoxy: -/- (HYBRID-W)	Compression SHPB [C]	1,275 & 1,503	Quasi-static properties not provided. Stress wave attenuation attributed to offset in force between input and transmitter bar. Inertia effects present during entire loading sequence. Transmitter bar peak force used as conservative estimate for computing stress in specimen. Elastic modulus estimated using strain and stress at peak stress. Slight increase in elastic modulus (+7%), strength increases (+46%), and ultimate strain increases (+25%) at 1,503 s^{-1} relative to 1,275 s^{-1} .
Hosur M.V. <i>et al.</i> (2001) [54]	carbon/epoxy pre-preg.(10/90) _s : PANEX 33/APCM LLC (DA 4518) (CFRP-PP)	<i>LF not spec.</i> [R] Compression SHPB [R]	82, 164, 817	Quasi-static strain estimated using crosshead displacement corrected for compliance. Higher quasi-static ultimate strains attributed to greater time for stress redistribution. Incident pulse generated by SHPB is highly non-uniform (significant dispersion suspected). Samples at 82 s^{-1} did not fail with SHPB, therefore, strength at quasi-static conditions cannot be compared. Elastic modulus increases at 82 s^{-1} (+30%), followed by a decrease with increasing strain rate to 817 s^{-1} (+12%). Strength increases (+19% at 817 s^{-1} relative to 82 s^{-1}), but lower compared to quasi-static values. Failure strain increases (+19% at 817 s^{-1} relative to 82 s^{-1}), but are lower than quasi-static values.

Woo S.-C. & Kim T.-W. (2016) [55]	carbon & kevlar/epoxy 2D twill weave: T300/Kevlar49/- (HYBRID-W)	Compression [C]	SHPB	1,007, 1,485, 1,941	Low transmission of the input pulse through the specimen ($10\text{--}12\%$ at $1,485\text{ s}^{-1}$ and $1,941\text{ s}^{-1}$). Therefore, low signal-to-noise ratio on stress measurement (noise in stress-strain curves). Material behaves in a more brittle manner at high strain rates. Acoustic emission signals were analysed to identify the onset of failure, and damage progression within the specimen. Effect of strain rate on modulus was not reported. Strength increases ($+80\%$ at $1,941\text{ s}^{-1}$), and ultimate strain decrease (-15% at $1,941\text{ s}^{-1}$).
Kapoor R. <i>et al.</i> (2016) [57]	kevlar/polypropylene 2D plain weave: Kevlar 29/MAG-PP (Not included in figures)	Compression [C]	SHPB	1,370, 2,005, 2,538, 3,239, 3,440, 4,264	Possibly the first paper to report on the high strain rate through-thickness properties for kevlar/thermoplastic resin composites. Specimens have very low aspect ratio ($L/D = 0.1\text{--}0.3$). Reliable measurements were not possible with the thinnest specimens. Elastic modulus increases ($+245\%$ at $4,264\text{ s}^{-1}$ relative to $1,370\text{ s}^{-1}$), strength increases ($+196\%$ at $4,264\text{ s}^{-1}$ relative to $1,370\text{ s}^{-1}$), and ultimate strain increases ($+134\%$). Significant increase in ultimate strain due to ductile behaviour of thermoplastic matrix. Toughness increases in a non-linear manner ($+808\%$ at $4,264\text{ s}^{-1}$ relative to $1,370\text{ s}^{-1}$).
Gama B. <i>et al.</i> (2001) [45]	S-2 glass/vinyl ester 2D plain weave: -/- (GFRP-W)	<i>LSLF (Instron - not spec.)</i> [R] Compression [R]		200 - 1600	Strains measured directly from specimens using strain gauges. Input pulse from SHPB highly non-uniform. Strain measured by strain gauges mounted on the specimen is used until unloading, then SHPB theory used. High levels of dispersion make stress-strain curves uninformative. 'Non-linear' strains are defined to account for parasitic strains from dispersion. Specimens may not reach quasi-static stress equilibrium prior to failure. Effect of strain rate on modulus not reported. Strength increases ($+38\%$ at $1,125\text{ s}^{-1}$) before reaching an approximately asymptotic value at higher strain rates. Ultimate strain increases ($+98\%$ at $1,125\text{ s}^{-1}$), but with high uncertainty.
Govender R. <i>et al.</i> (2011) [9]	glass/vinyl ester 2D plain weave: E-glass 24 oz./Derakane 8084 (GFRP-W)	<i>EMLF (Zwick Universal)</i> [C] Compression [C]		510	Cross head displacement used to estimate strain due to concerns about strain gauge alignment. Small specimens prevented the use of extensometers. Effect of strain rate on modulus not reported as quasi-static stress equilibrium was not achieved until late in the test. Strength initially increases ($+13\%$ at 510 s^{-1}) but decreases with increasing strain rate ($+4\%$ at $1,800\text{ s}^{-1}$). Slight reduction in ultimate strain (-5%).
Gerlach R. <i>et al.</i> (2012) [10]	carbon/epoxy 3D weave: (Tenax HTS/HTA)/RTM-6 (3D-W)	<i>EMLF (not spec.)</i> <i>HLF (Im. s^{-1})</i> [R] Compression [R]		0.004 - 6,000	Strain was measured using a laser extensometer for quasi-static tests. At intermediate strains digital speckle photography was used. Modified direct impact SHPB used for high strain rate testing. Compression stress-strain response at high strain rates shows some waviness due to dispersion. Strength measured as maximum stress before unstable failure. Tangent modulus (measured at 300 MPa) increases ($+43\%$ at $6,000\text{ s}^{-1}$). Negligible strain rate effect on strength.

Shah Khan M.Z. & Simpson G. (2000) [50]	carbon/epoxy 2D plain weave: DF1400/Synolite 0288-T 1 resin (CFRP-W)	HLF [R]	0.001 - 10	Quasi-static strains estimated using crosshead displacement corrected for compliance. Hydraulic load frame used to load at intermediate strain rates. Strength and ultimate strain taken at point of maximum stress. Inconsistent failure modes thought to contribute to scatter in measured ultimate parameters. Elastic modulus increases (+25% at 10 s^{-1}), strength increases (+21% at 10 s^{-1}), and strain increases (+11% at 10 s^{-1}). Significant scatter in all measured parameters (up to $\pm 25\%$, 12% and 33% for elastic modulus, strength and ultimate strain, respectively).
Shen <i>et al.</i> (2012) [51]	glass/vinyl ester 3D weave: E-glass/ML-506 (3D-W)	HLF (MTS 810) [R] Compression SHPB [R]	0.001 - 1,200	Hydraulic load frame used to test at quasi-static and intermediate strain rates. Stress strain curves are very non-linear and exhibit very large increase in the tangent modulus (defined above 0.025 strain). Unclear how many specimens were tested. Unrealistically high effect of strain rate on the tangent modulus suggests specimens are not in a state of stress equilibrium. Tangent modulus increases (+350% at $1,200 \text{ s}^{-1}$) with very high uncertainty ($\pm 79\%$). Strength increases (+8% at $1,200 \text{ s}^{-1}$) and ultimate strain decreases (-50% $1,200 \text{ s}^{-1}$).
Akil <i>et al.</i> (2003) [52]	glass/vinyl ester 2D weave: S-2 glass/- (GFRP-W)	HLF (<i>not spec.</i>) [R, C] Compression SHPB [C]	0.001 - 900	Tangent modulus, fitted to later region of 'linear' response and used to estimate elastic modulus. Strain rate sensitive behaviour attributed to viscoelastic properties of the matrix. Consistent shear failure mode between quasi-static and high strain rate tests. Tangent modulus increases (+79% at 900 s^{-1}) with high uncertainty ($\pm 26\%$), strength increases (+29% at 900 s^{-1}), and ultimate strain decreases slightly (-6% at 900 s^{-1}).
Pankow <i>et al.</i> (2011) [27]	glass/epoxy 3D weave: Z-fibre/SC-15	HLF (<i>not spec.</i>) [R] Compression SHPB [C]	QS - 1750	Used 2D DIC performed with high speed camera to measure strain on the specimen. Poor spatial and temporal resolution results in high error ($> 5\%$ strain). No details provided on the DIC setup. High strain rate tests suffer from high levels of dispersion (oscillations in stress-strain response). This introduces uncertainty in strength and ultimate strain measurements. Effect of strain rate on modulus not reported quantitatively. Strain rate appears to have negligible effect on elastic modulus. Strength increases relative to quasi-static values (+11% at $1,750 \text{ s}^{-1}$) but not strain rate sensitive (within scatter). Ultimate strains increase relative to quasi-static values (+20% at $1,750 \text{ s}^{-1}$) but also not strain rate sensitive (within scatter).
Guden <i>et al.</i> (2004) [53]	glass/epoxy 2D weave: S-2 glass/SC-15	EMLF (Shimadzu AG-D) [C] Compression SHPB [C]	0.0001 - 1,100	Compressive modulus computed using linear region of stress-strain curve at 4% strain. Strain rate limited to $1,100 \text{ s}^{-1}$, above which quasi-static stress equilibrium could not be achieved prior failure. Consistent shear failure mode between quasi-static and high strain rate tests. Elastic modulus increases (+70% at $1,100 \text{ s}^{-1}$), strength increases (+53% at $1,100 \text{ s}^{-1}$), ultimate strain decreases slightly with increasing strain rate (-3% at $1,100 \text{ s}^{-1}$).

Table 2 Summary of published studies on strain rate effects of interlaminar tensile properties of fibre-reinforced polymer composites. Notes: Constituent materials listed in the format: 'fibre/matrix'. Material labels are included in '()' for association with Fig. 8 - Fig. 10. Quasi-static: italic typeface; high strain rate: regular typeface; test method: electro-mechanical load frame (EMLF), hydraulic load frame (HLF), high-speed load frame (HSLF), split Hopkinson pressure bar (SHPB); specimen geometry (in ' '): W = waisted, D = dog-bone, L = L-shaped/curve beam, O = off-axis, C = cylindrical specimens

Reference	Material	Test [Specimen]	Strain Rate (s^{-1})	Notes
Nakai K. & Yokoyama T. (2014), (2016) [6, 35]	carbon/epoxy pre-preg. ([0], [0/90] _s): T700S/2521, T700S/2500 (CFRP-PP)	<i>EMLF</i> (<i>Instron 5500R</i>) [W] Tension SHPB [C-W]	0.02 – 60	Waisted specimens were bonded to magnesium alloy end caps using DP-460 epoxy. Strains under quasi-static and high strain rate loading were measured using strain gauges. Specimen appears to be in a state of quasi-static stress equilibrium for much of the test. Strains computed using SHPB overestimate true strains due to non-uniform deformation of the gauge region. Effect of strain rate on elastic modulus not reported. Strength increases more for the cross-ply layup compared to the unidirectional layup (+130% at 50 s^{-1} ([0/90] _s), +77% at 50 s^{-1} ([0] _s), and ultimate strain increases more for the cross-ply layup compared to the unidirectional layup (+31% at 50 s^{-1} ([0/90] _s), -10% at 50 s^{-1} ([0] _s). Large uncertainty in strength (up to $\pm 51\%$ for [0/90] _s and $\pm 18\%$ for [0] _s at 50 s^{-1}) and ultimate strains (up to $\pm 61\%$ for [0/90] _s and $\pm 43\%$ for [0] _s at 50 s^{-1}).
Lifshitz J. & Leber H. (1998) [58]	carbon/epoxy pre-preg.: AS4/3502 ([0] _s) (CFRP-PP) glass/epoxy 2D plain weave: NEMA/ASTM G-10 (GFRP-W)	Tension SHPB [C-W-O]	127-195	Tension, shear and combined tension-shear behaviour was studied using two types of specimens. Waisted specimens were used for tension, and off-axis waisted specimens were used for tension/shear loading. Off-axis specimens were formed by bonding two halves together. Bonded specimens very difficult to machine from CFRP and the results had too much scatter to be meaningful. Specimens bonded to incident and transmitter bars (Hyscon adhesive). Oscillations in stress-strain response and erratic shape close to failure thought to be a result of micro crack formation. Strains measured using strain gauges. No quasi-static ultimate strains reported. Elastic modulus increases (+41% for carbon/epoxy at 195 s^{-1} , +7% for glass/epoxy at 127 s^{-1}). Strength increases for the carbon/epoxy laminate (+36% at 195 s^{-1}), but decreases for the glass/epoxy laminate (-14% at 127 s^{-1}).
Naik N.K. <i>et al.</i> (2010) [61]	glass/epoxy 2D plain weave: E-glass/LY556 (GFRP-W)	<i>LF not spec.</i> [C-W] Tension SHPB [C-W]	140 – 400	Specimens bonded to end tabs which fit inside incident and transmitter bars. Authors claim specimen reach quasi-static stress equilibrium; however, plots of reaction forces do not appear equal during loading. Strength increases significantly when compared to quasi-static values (+88% at 390 s^{-1}). The influence of strain rate on strength is much lower within the range of high strain rates considered (+11% at 390 s^{-1} relative to 145 s^{-1}).
Gerlach R. <i>et al.</i> (2013) [19]	S2-glass/epoxy pre-preg.: -/FM94 (cut from GLARE sheets) (GFRP-PP)	<i>EMLF (not spec.)</i> [C-W] Tension SHPB [C-W]	5 x 10 ⁻⁴ , 10, 200	Attempts to measure interlaminar strength from GLARE plates. Specimens bonded directly to incident and transmitter bars. Due to the small effective gauge length (2 mm), no meaningful strain measurements could be performed using digital speckle photography. Large scatter in through-thickness direction attributed to inconsistent failure modes and variation in manufacturing the GLARE plates. Effect of strain rate on elastic modulus and ultimate strain not reported. Strength increased (+45% ($\pm 105\%$)) at 200 s^{-1} .

Hufenbach W. <i>et al.</i> (2013) [60]	glass/epoxy 2D weave: (MKF & Twintex) (GFRP)	Tension SHPB [D.L.]	5×10^{-4} - 400	Large scatter in SHPB tension tests attributed to reinforcement architecture. Authors conclude that SHPB is not suitable for testing coarse textile architectures. Varying levels of compaction in L-shaped beam specimens resulted in unacceptable scatter (results not reported). Effects of strain rate on modulus not reported. Increasing strength (+93% at 400 s^{-1}) with approximately constant strain at failure (within scatter: -75% – +87% at 44 s^{-1} to -87% – +639% at 400 s^{-1}).
Govender R. <i>et al.</i> (2011) [9]	glass/vinyl ester 2D plain weave: E-glass 24 oz./Derakane 8084 (GFRP-W)	Compression SHPB-spall [C]	1,800	No quasi-static values reported due to consistent failure within the grips. A spall test was used to measure tensile strength. Pulse time-shifting used to estimate forces in the specimen at failure. A high speed camera was used to qualitatively monitor the failure. The effect of strain rate on the elastic modulus and ultimate strain were not reported. The average strength was 125 MPa (std. dev. = 21.2 MPa). Authors compare this to the strength quoted by the epoxy manufacturer (76 MPa) to conclude that the material likely exhibits a significant sensitivity to strain rate.
Gerlach R. <i>et al.</i> (2012) [10]	carbon/epoxy 3D weave (Tenax HTS/HTA)/RTM-6 (3D-W)	<i>EMLF (not spec.)</i> [CROSS] <i>HLLF (ltn. s^{-1})</i> [CROSS] Compression SHPB [CROSS]	0.004 - 11,000	Cross-specimens used to introduce load without edge effects for 3D reinforcement. Specimens loaded in compression with a SHPB using U-shaped fixture. Specimen geometry and fixture made non-contact measurements prohibitive. Stress state not uniform due to small fillets between cross arms. Influence of strain rate on elastic modulus and ultimate strain not reported. Strength generally increases (+84% – +94%) with increasing strain rate, but with high scatter ($\pm 60\%$).
Hufenbach W. (2011) [18]	glass/polypropylene 2D plain weave: Twintex TPP 60 745 (GFRP-W)	<i>EMLF (ZWICK Z250)</i> [L] HSLF (INSTRON VHS 160/20) [L]	10^{-4} - 7	3D DIC used on L-shaped beam specimens. Significant scatter in the optical measurements attributed to the coarse textile surface pattern. Heterogeneous strain fields through thickness. Strains averaged over subsets. Limited measurement resolution at intermediate strain rates (HS camera resolution). Effect of strain rate on modulus and ultimate strain not reported. Strength increases significantly (+204% at 1 s^{-1} and +171% at 7 s^{-1}).

Medina J. & Harding J. (2000) [59]	carbon/epoxy pre-preg.: T300/924 (CFRP-PP) carbon/epoxy 2D plain weave: Fibredux 924C/833 (CFRP-W) R-glass/epoxy 2D plain weave: Fibredux 924G/20982 (GFRP-W)	LF not spec. [W-C] Tension SHPB [W-C]	5 - 940	Waisted specimens bonded to steel end caps. Strain gauges occasionally failed before the specimen, or were positioned off of the failure plane. Longitudinal strain gauge shows persistent oscillations in strain. Results show that reinforcement architecture has larger influence than fibre material. Poisson's ratio doubled when glass fibres used compared to carbon fibres. carbon/epoxy pre-preg: Tensile modulus increases with strain rate (+31%), tensile strength increases (+12%), and tensile strain also increases (+22%); carbon/epoxy 2D plain weave: Tensile modulus increases with strain rate (+7%), tensile strength increases (+37%), and tensile strain also increases (+63%); glass/epoxy 2D plain weave: Tensile modulus decreases with strain rate (-13%), tensile strength increases (+40%), and tensile strain also increases (+65%).
---------------------------------------	---	--	---------	--

Table 3 Summary of published studies on strain rate effects of interlaminar shear properties of fibre-reinforced polymer composites. Notes: Constituent materials listed in the format: 'fibre/matrix'. Material labels are included in '()' for association with Fig. 11 - Fig. 13. Quasi-static: italic typeface; high strain rate: regular typeface; test method: electro-mechanical load frame (EMLF), hydraulic load frame (HLF), high-speed load frame (HSLF), split Hopkinson pressure bar (SHPB); specimen geometry (in ' '): SBS = short beam shear (3 point bend), I = notched/unnotched shear test, R = rectangular/cubic specimens, O = off-axis, T = thin-walled tube, SL = single lap-shear joint, DL = double lap-shear joint

Reference	Material	Test [Specimen]	Strain Rate (s^{-1})	Notes
Naik N.K. <i>et al.</i> (2007) [5]	carbon/epoxy 2D plain-weave: <i>—/—</i> (CFRP-W), E-glass/epoxy 2D plain weave: <i>—/—</i> (GFRP-W)	<i>LF not spec.</i> [SL] Torsion SHPB [T] Compression SHPB [SL]	496 – 1,000	Single lap and tubular specimens considered at high strain rates. Viscoelastic behaviour of matrix and less time for damage propagation responsible for increase in shear strength. Post failure microscopy of tubular specimens reveals features associated with a pure shear stress state. Shear stress may not be constant due to the fabric and different properties within the specimen thickness. Wall thickness set at 3 mm since thinner walls gave inconsistent results (thought to be a result of manufacturing defects). No quasi-static ultimate strains provided. Modulus estimated by stress and strain at point when quasi-static stress equilibrium was thought to be achieved. <u>carbon/epoxy 2D plain weave:</u> Increase in modulus (+38%), increase in strength (+69%) and increase in ultimate strain (+41%) at $1,000 s^{-1}$ relative to $496 s^{-1}$; <u>glass/epoxy 2D plain weave:</u> Increase in modulus (+29%), increase in strength (+67%) and increase in ultimate strain (+35%) at $1,000 s^{-1}$ relative to $576 s^{-1}$.
Harding J. & Dong L. (1994) [3]	carbon/epoxy pre-preg.: T800/924 ([0], [0/90], [± 45]) (CFRP-PP)	<i>LF not spec.</i> [DL] Tension SHPB [DL]	275 – 332	Scatter is so large that the experiments cannot be considered as proof of a strain rate dependence. Improved specimen design required if strength values are to be accepted with confidence. Normal stress plays a significant role in the failure process as shown by micrographs of failure surfaces. Influence of strain rate on shear modulus not reported. <u>carbon/epoxy pre-preg. [0] (1-3):</u> shear strength increases (+26%) and ultimate shear strain decreases (-16%) at $275 s^{-1}$. <u>carbon/epoxy pre-preg. [0/90] (2-3):</u> shear strength increases (+39%) and ultimate shear strain decreases (-16%) at $332 s^{-1}$. <u>carbon/epoxy pre-preg. [± 45] (1-3):</u> shear strength increases (+38%) and ultimate shear strain decreases (-22%) at $281 s^{-1}$.

Hufenbach W. <i>et al.</i> (2009) [21]	glass/polypropylene hybrid yarn: MKF 3D weave (3D-W) & Twintex 2D weave (GFRP-W)	<i>HSLF (INSTRON VHS 16020)</i> [I]	6×10^{-3} - 60	Lightweight Iosipescu fixture used to reduce the effects of frame inertial. High speed cameras used for visual analysis of damage and failure processes. Approximately equal interlaminar shear strength for both composites. Low failure stress and strain levels (Twintex in 1-3 plane) make the measurements noisy and unreliable.
Influence of strain rate on shear modulus not reported.				
MKF (2-3): shear strength increases (+6%) up to 6 s^{-1} , followed by a decrease (-18%) at 60 s^{-1} , and ultimate shear strain decreases (-13%) at 60 s^{-1} .				
Twintex (2-3): shear strength increases (+41%) and ultimate shear strain increases (+18%) at 60 s^{-1} .				
Twintex (1-3): shear strength trends unclear (range between -35% at 6 s^{-1} and +35% at 60 s^{-1}) and ultimate shear strain increases (+118%) at 60 s^{-1} .				
Hufenbach W. <i>et al.</i> (2013) [60]	glass/polypropylene hybrid yarn: MKF 3D weave (3D-W)	<i>HSLF (INSTRON VHS 16020)</i> [I]	5×10^{-4} - 60	Lightweight Iosipescu fixture used to reduce the effects of frame inertial. Used previously up to 60 s^{-1} [21]. Highly non-linear shear response at all strain rates.
Influence of strain rate on shear modulus not reported. Shear strength increases (+83% at 3 s^{-1}), and negligible effect on ultimate shear strain (within scatter: $+21\% \pm 57\%$).				
Gowtham H.L. <i>et al.</i> (2015) [62]	E-glass/epoxy 2D plain weave: -/LY556(HY951 Hardener) (GFRP-W)	Torsion SHPB [T] Compression SHPB [SL]	192 - 457 (T) - 1500 (SL)	300 300
Quasi-static testing performed only for torsion specimens (details not reported). Quasi-static torsional simulations performed to assess possible stress concentrations in tubular specimens. Single-lap and tubular specimens compared using two SHPBs. Lower strength measured with thin-walled specimens thought to be a result of variations in stiffness and stress along the radial, and circumferential directions.				
Effect of strain rate on shear modulus and ultimate shear strain not reported. Interlaminar shear strength increases (+56% at 457 s^{-1}) for tubular specimens and increases (+11% at $1,500 \text{ s}^{-1}$) for single-lap specimens.				
Gerlach R. <i>et al.</i> (2012) [10]	carbon/epoxy 3D weave: (Tenax HTS/HTA)/RTM-6 (3D-W)	<i>EMLF (not spec.)</i> [I - <i>dbl notch</i>] Compression SHPB [I - <i>dbl notch</i>]	0.004 - 11,000	Two binder volume fractions considered (3% and 6%). Tubular specimens not practical for 3D weaves due to the arrangement and wide spacing of reinforcement. A double-notch shear specimen was adopted. Average shear strength is higher in the 2-3 plane compared to the 1-3 plane. Inadequate resolution and accuracy of optical shear strain measurements (digital speckle photography). Large scatter in strength measurement attributed to stress concentrations at notches.
Influence of strain rate on shear modulus and ultimate strain not reported. Shear strength increases in the 1-3 plane (+52% (3 % binder) and +37% (6 % binder)) and increases in the 2-3 plane (+34% (3 % binder) and +31% (6 % binder)) at $11,000 \text{ s}^{-1}$.				

Yokoyama T. & Nakai K. (2006) [65]	carbon/epoxy pre-preg.: T700/2521 (I0)	<i>EMLF (Instron 5500R) [I, SBS]</i> Compression SHPB [I - dbl. notch]	0.02 - 780	Similar results are obtained when the notched shear specimens are loaded in tension and compression. Strength values from notched specimens agree well with the short beam shear (SBS) tests. Compressive normal stresses around notches thought to reduce effect of stress concentrations. Effect of strain rate on shear modulus and ultimate shear strain not reported. Negligible effect of strain rate on interlaminar shear strength up to 780 s^{-1} .
Bouette B. <i>et al.</i> (1992) [39]	carbon/epoxy pre-preg.: T300/5208 (I0)(CFRP-PP)	<i>EMLF (LF not spec.) [SL]</i> <i>HLF (LF not spec.) (Im. s^{-1}) [SL]</i> Tension SHPB [SL, DL]	0.001 - 1,000	Two specimens were designed using FEA: one permitting the determination of the shear modulus, and the other (double lap shear specimen) permitting measurement of the shear strength (1-3 plane). Only values for single-lap specimens reported. Strain measured using strain gauges on the specimen. Authors emphasize that strength values must be interpreted with caution due to tensile normal stresses ends of the overlap. Maintaining shorter overlap length reduces peel stresses. Effect of strain rate on ultimate shear strain not reported. Negligible change in shear modulus or interlaminar shear strength up to $1,000 \text{ s}^{-1}$ (within experimental scatter).
Hallett S.R. <i>et al.</i> (1999) [40]	carbon/epoxy pre-preg.: T300/914 (I0/90) _B	<i>LF not spec. [SL (Z)]</i> Compression SHPB [SL (Z)]	5×10^{-4} - 800	Z-shaped single-lap specimens used. Shear strain measured directly using $\pm 45^\circ$ rosette strain gauge. High speed photography (Cordin) used to observe failure mechanisms. Linear regression fit used to estimate shear modulus from stress-strain curve (noise and oscillations in curve). Too much scatter on shear strength values to resolve strain rate dependency. Failure found to initiate near the notches, suggesting that the stress concentration may dominate the failure. Shear modulus increases (+41% at 700 s^{-1}) but with high scatter ($\pm 34\%$). Negligible effect on shear strength (within experimental scatter) at 450 s^{-1} and 700 s^{-1} . Ultimate shear strain increases (+27% at 700 s^{-1}) but with high scatter ($\pm 20\%$ at 450 s^{-1} and $\pm 12\%$ at 700 s^{-1}).
Gillespie J. <i>et al.</i> (2005) [2]	S-2 glass/epoxy 2D weave (15x15 and 5x5) /SC79 (GFRP-W)	<i>LF not spec. [R-O]</i> Compression SHPB [R-O]	QS - 1,000	Out-of-plane off-axis specimens loaded in compression to obtain shear strength. Off-axis angles considered include: 0° , 15° , 30° , 45° , 60° , 75° , and 90° . An 'R-value' criterion (based on relative difference between reaction forces) was to determine which tests were likely contaminated by inertial effect (limits strain rate to $< 600 \text{ s}^{-1}$). A strain rate dependent failure criterion was identified based on experimental results. Effect of strain rate on shear modulus and ultimate shear strain not reported. Interlaminar shear strength found to increase (+134%) at 985 s^{-1} .

Harding J. & Li Y.L. (1992) [64]	carbon/epoxy pre-preg. (I0): T300-3000A/Ciba-Geigy XD 927 (CFRP-PP) E-glass/epoxy 2D plain weave: -/Ciba-Geigy XD 927 (CFRP-W) hybrid carbon-glass/epoxy: T300-3000A/Ciba-Geigy XD 927, -/Ciba-Geigy XD 927 (GFRP-W)	LF not spec. (Instron) [DL] Tension SHPB [DL]	QS - 1,600	Shear strains and shear modulus were not measured. Load cell displacement used to infer an apparent shear strain. No sensitivity to fibre volume fraction or ply layout for the same material. Problems arose with hybrid specimens as a result of a discontinuity in elastic properties (different types of reinforcing plies) on either side of the failure plane. Finite element simulations show significant normal stresses at the ends of the overlap. Stress concentrations thought to dominate failure initiation in the specimens. carbon/epoxy pre-preg: increase in failure strength (+73%) at 1,600 s⁻¹ but with high scatter (± 27%); glass/epoxy 2D weave: increase in strength (+111%) at 1,600 s⁻¹ but with high scatter (± 56%); hybrid glass/epoxy: increase in strength (+37%) at 1,600 s⁻¹ but with high scatter (± 31%).
----------------------------------	--	---	------------	--



Image-Based Inertial Impact Test for Composite Interlaminar Tensile Properties

J. Van Blitterswyk¹ · L. Fletcher¹ · F. Pierron¹

Received: 22 May 2018 / Accepted: 1 September 2018 / Published online: 11 September 2018
© The Author(s) 2018

Abstract

In this work an image-based inertial impact test is proposed to measure the interlaminar tensile stiffness and strength of fibre-reinforced polymer composite materials at high strain rates. The principle is to combine ultra-high-speed imaging and full-field measurements to capture the dynamic kinematic fields and exploit the inertial effects generated under high strain rate loading. The kinematic fields are processed using the virtual fields method to reconstruct stress averages from maps of acceleration. In this way, the specimen acts like a dynamic load cell, with no gripping or external force measurement required. Stress averages are combined with strain measurements to construct stress–strain curves and identify the interlaminar stiffness and tensile strength. Special optimised virtual fields are also implemented to identify interlaminar stiffness parameters from complete maps of strain and acceleration. Interlaminar stiffness and tensile strength are successfully identified at average, peak strain rates on the order 3500 s^{-1} and 5000 s^{-1} , respectively. Results show an increase in stiffness between 30 and 35%, and an increase in strength of 125% compared to quasi-static values.

Keywords High strain rate · Interlaminar tension · Fibre-reinforced polymer composites · Dynamic test methods · Ultra-high speed imaging · Virtual Fields Method

Introduction

There are many engineering applications where fibre-reinforced polymer (FRP) composite structures are subjected to dynamic loading (e.g., impact, blast, crash, etc.). In order to develop useful numerical simulations for composite structures subjected to these loading cases, a thorough understanding of the constitutive behaviour over a wide range of strain rates must first be established. Since interlaminar properties are matrix-dominated, literature suggests that the interlaminar stiffness and strength should exhibit a strain rate dependency [1, 2]. However, the number of studies attempting to measure high strain rate properties in the interlaminar direction are relatively scarce and inconsistent [2].

Studies on the strain rate dependency of interlaminar tensile properties are comparatively fewer compared to compression and shear. The effect of strain rate is not well understood, as exemplified by highly scattered measurements in available studies [2]. For example, measured changes in stiffness for glass/epoxy woven composites range from + 10% at 125 s^{-1} [3], to + 500% at 400 s^{-1} [4], and – 15% at 950 s^{-1} [5], compared to quasi-static values. A similar situation exists for tensile strength measurements at high strain rates. Most studies report moderate increases in strength of around 30% compared to quasi-static values [3, 5], while others report much more significant increases of up to several hundred percent [6–8]. Therefore, much ambiguity remains regarding the effect of strain rate on interlaminar stiffness and tensile strength. Some of the inconsistency in the literature stems from variations in material composition (fibre and matrix materials, fibre volume fraction, reinforcement architecture). Additionally, scatter is likely amplified by other factors complicating tensile tests, such as gripping, alignment and geometry (stress concentrations).

Limitations of the Split-Hopkinson pressure bar (SHPB) technique are thought to be the primary source of inconsistency in measurements above $200\text{--}300\text{ s}^{-1}$. At such

✉ J. Van Blitterswyk
j.van-blitterswyk@soton.ac.uk

L. Fletcher
l.c.fletcher@soton.ac.uk

F. Pierron
f.pierron@soton.ac.uk

¹ Mechanical Engineering, University of Southampton,
Southampton SO17 1BJ, UK

strain rates, inertial effects violate the necessary assumption that the specimen is in a state of quasi-static equilibrium. As a result, measurements of strain on the input and output bars cannot be used to directly infer the stress–state in the material until specimen acceleration dampens out. The low wave speed and small ultimate strain typical of composite materials in the through-thickness direction means that a state of quasi-static equilibrium may not be achieved before failure occurs. The lower wave speed also tends to reduce the signal-to-noise ratio of the material response transmitted to the output bar [1, 9]. Even if inertial effects dampen out prior to failure, it is generally agreed that the SHPB cannot produce reliable measures of the material stiffness [10–13].

More recently, full-field measurement techniques, combined with ultra-high-speed imaging, have shown to be a promising alternative for high strain rate testing. Ultra-high-speed imaging refers to framing rates upwards of 1 Mfps, according to [14]. Having access to dynamic full-field measurements of displacement alleviates some of the fundamental assumptions attached to test methods such as the SHPB. Specifically, the Virtual Fields Method (VFM) can be used to process dynamic maps of acceleration to exploit the inertial, heterogeneous deformation and reconstruct stress-averages in the material. Moulart et al. [15] showed that acceleration could be used in practice to identify stiffness parameters at high strain rates. The ‘Image-Based Inertial Impact’ (IBII) test first emerged from the study by Pierron and Forquin [16], and was later formalised in [17]. The concept of using acceleration to reconstruct stress-averages and identify constitutive material properties has since been successfully demonstrated in a number of experimental studies [16–25].

This work presents an extension of the IBII test method to measure the interlaminar stiffness and tensile strength of a unidirectional, carbon/epoxy composite. Both interlaminar material planes are considered (1–3 and 2–3). The IBII test is advantageous as it eliminates the need to grip the specimen or measure force externally; both of which are particularly problematic for testing interlaminar specimens in tension. This paper will begin by describing the test concept and design approach in Sect. 2. A parametric finite element model is used to design the experimental parameters (e.g., projectile length and impact velocity) in Sect. 3. In Sect. 4 the displacement fields from the finite element model are used to synthetically deform images. These images are used to systematically optimise the post-processing parameters and estimate the error on identified stiffness parameters. A description of the experimental setup is presented in Sect. 5. Experimental results are presented and discussed in Sect. 6. Finally, the limitations of the proposed test, and future developments for the work are discussed in Sect. 7.

Image-Based Inertial Impact (IBII) Test Concept and Design

In the IBII test, the specimen is loaded dynamically by impacting one side of the specimen (represented by $F(t)$ in Fig. 1a). The impact induces a compressive wave, which travels through the specimen towards the free edge. When the wave reaches the free edge, it reflects back towards the impact edge as a tensile pulse. The idea is to tailor the experimental parameters to ensure that the reflected tensile pulse is sufficient to cause specimen failure. Ultra-high-speed imaging is combined with the grid method to capture dynamic full-field displacements. The underlying constitutive properties are encoded in these maps, and are extracted using the VFM. The initial compressive loading is used to identify the elastic modulus, and the ‘linear stress-gauge’ equation [20] is used to estimate the tensile strength of the specimen.

The Virtual Fields Method (VFM)

The VFM is used to directly identify constitutive properties from full-field measurements of strain and acceleration. The specific application of the VFM to the IBII test is outlined in [20] and therefore, only the specific aspects that apply to the interlaminar test are included here. The reader is referred to [26] for a comprehensive description of the VFM.

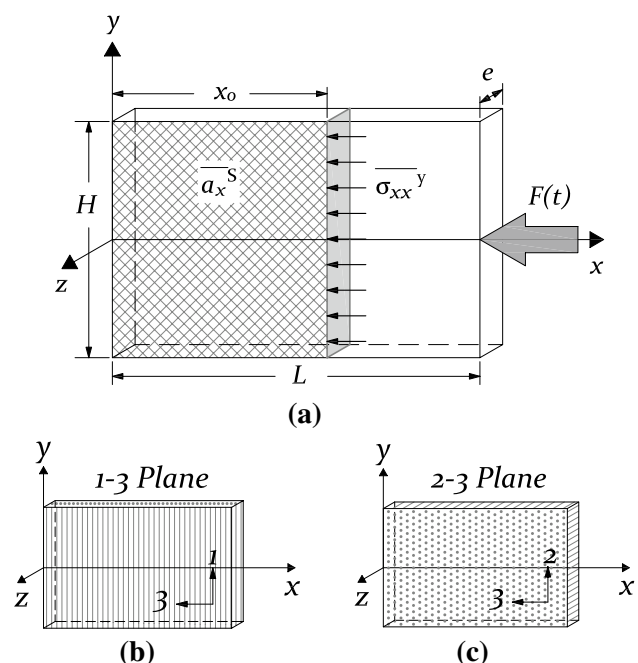


Fig. 1 **a** Schematic of an impacted interlaminar specimen. Fibres are either, **b** parallel to the y axis (1–3 interlaminar plane) or **c** parallel to the z axis (2–3 interlaminar plane)

The principle of virtual work is satisfied by any virtual displacement fields that are continuous and piecewise differentiable. In this work, the following additional assumptions are made: (1) constant density, thickness and stiffness in space; (2) kinematic fields are uniform in the thickness dimension of the sample; and (3) the specimen can be considered to be in a state of plane stress. Under these conditions, the principle of virtual work takes the following form:

$$-\int_S \boldsymbol{\sigma} : \boldsymbol{\epsilon}^* dS + \int_{\delta l} \bar{\mathbf{T}} \cdot \mathbf{u}^* dl = \rho \int_S \mathbf{a} \cdot \mathbf{u}^* dS \quad (1)$$

where S denotes the surface of the region of interest, $\bar{\mathbf{T}}$ is the Cauchy stress vector, which is applied to the in-plane boundaries denoted by l , $\mathbf{u}^*$ is the virtual displacement field, $\boldsymbol{\epsilon}^*$ the virtual strain field and $\boldsymbol{\sigma}$ the Cauchy stress tensor. Note that $:$ and $\cdot$ denote the dot product in matrix and vector forms, respectively. The left hand side of Eq. (1) represent the internal and external virtual work, respectively. The term on the right hand side of the equation represents the virtual work of inertial forces. In order to use Eq. (1), it is necessary to define a constitutive model. Here, a linear elastic, orthotropic constitutive model is used. The material response in the 1–3 plane is very similar to the interlaminar response considered in [20]. For that case, using a uniaxial virtual strain field, and neglecting transverse strains from the very small Poisson effect, the VFM equation is reduced to:

$$-E_{33} \int_S \epsilon_{xx}^* \epsilon_{xx} dS + \int_{\delta l} T_x u_x^* dl = \rho \int_S a_x u_x^* dS \quad (2)$$

In the case of the 2–3 interlaminar plane, an isotropic constitutive model is more appropriate, where Q_{33} and Q_{23} are directly identified. Therefore, Eq. (1) becomes:

$$\begin{aligned} & -Q_{33} \int_S (\epsilon_{xx}^* \epsilon_{xx} + \epsilon_{yy}^* \epsilon_{yy} + \frac{1}{2} \epsilon_{xy}^* \epsilon_{xy}) dS \\ & -Q_{23} \int_S (\epsilon_{xx}^* \epsilon_{yy} + \epsilon_{yy}^* \epsilon_{xx} - \frac{1}{2} \epsilon_{xy}^* \epsilon_{xy}) dS \\ & + \int_{\delta l} (T_x u_x^* + T_y u_y^*) dS = \rho \int_S (a_x u_x^* + a_y u_y^*) dS \end{aligned} \quad (3)$$

In both cases, virtual displacements at the impact edge are set to zero to exclude the virtual work of the unknown stress distribution along the boundary.

Stiffness Identification

Special Optimised Virtual Fields An infinite number of virtual kinematic fields can be selected to use Eqs. (2) and (3). If strain and acceleration fields are exact, all virtual fields will provide the same identification. In practice, measurements inevitably contain noise and therefore, each set of

virtual fields will provide different identifications. If one selects the virtual fields intuitively, it is impossible to ensure that the chosen ones are optimal. Special optimised virtual fields are adopted here, as developed in [27]. This procedure automates the selection of virtual fields for direct identification of stiffness parameters, such that the sensitivity to strain noise is minimised. For the 1–3 plane, the uniaxial virtual displacement field results in the direct identification of E_{33} . For the 2–3 plane, the isotropic, special optimised virtual field procedure is used to directly identify Q_{33} and Q_{23} [17]. From the identified stiffness parameters, one can calculate ν_{23} ($\nu_{23} = Q_{23}/Q_{33}$) and E_{33} ($E_{33} = Q_{33}(1 - \nu_{23}^2)$). This is performed at each time step over the whole field. Full details on the implementation of special optimised virtual fields are described in [26].

Stress-Gauge Equation Following from Fletcher et al. [20], the ‘stress-gauge’ (SG) equation is derived using the following rigid-body virtual field: $u_x^* = 1$, $u_y^* = 0$. Approximating integrals with discrete sums, one obtains a direct relationship between acceleration and average stress:

$$\overline{\sigma_{xx}}^y = \rho x_o \overline{a_x}^S \quad (4)$$

where superscript y coupled with an overline denotes the width average at x_o , and superscript S coupled with an overline denotes the surface average between the free edge and x_o (see Fig. 1a). The stress-gauge equation is used to plot $\overline{\sigma_{xx}}^y$ as a function of the average axial strain at each cross-section. This is used to verify the linearity of the response and spatial uniformity of the stiffness. The slope of the line provides a measure for E_{33} (1–3 plane) or Q_{33} (2–3 plane). For 2–3 plane specimens, E_{33} is calculated using ν_{23} from the special optimised identification routine.

Strength Identification

By considering additional rigid body virtual fields ($u_x^* = 0$, $u_y^* = 1$; $u_x^* = y$, $u_y^* = -x$), a linear approximation of the stress distribution may be reconstructed across the width of the specimen. This is referred to as the ‘linear stress-gauge’ (LSG) equation (Eq. 5),

$$\sigma_{xx}(LSG) = \rho x_o \overline{a_x}^S + \frac{12\rho x_o y}{H^2} (\overline{a_{xy}}^S - \overline{a_y x}^S + x_o \overline{a_y}^S) \quad (5)$$

where H denotes the total specimen width (Fig. 1a). The full derivation of Eq. (5) is detailed in [20]. It is important to highlight that the stress-gauge equations are valid regardless of the constitutive behaviour. However, the linear stress-gauge equation only provides the first moment of stress, and does not fully resolve σ_{xx} across the width of the specimen. This means that the stress-gauge only provides an estimate of the stress at the crack location. In this work, the linear stress-gauge equation is used for estimating interlaminar

tensile strength because it accounts for possible asymmetry in the stress field (two-dimensional wave propagation).

An additional, post-hoc way to reconstruct stress is to use the identified stiffness parameters to convert strain maps to stress maps. Divergence of the different stress measures (from acceleration and from strain) provides information about the damage process, indicating when the assumptions of the constitutive law break down (i.e., plasticity, damage, failure, etc.). This can also be used to validate the constant and linear approximations for the stress field as reconstructed with the stress-gauge equations.

Numerical Design and Optimisation

The following section describes the implementation of the numerical simulations and parametric design sweeps to select test parameters.

Model Configuration and Parametric Design Sweep

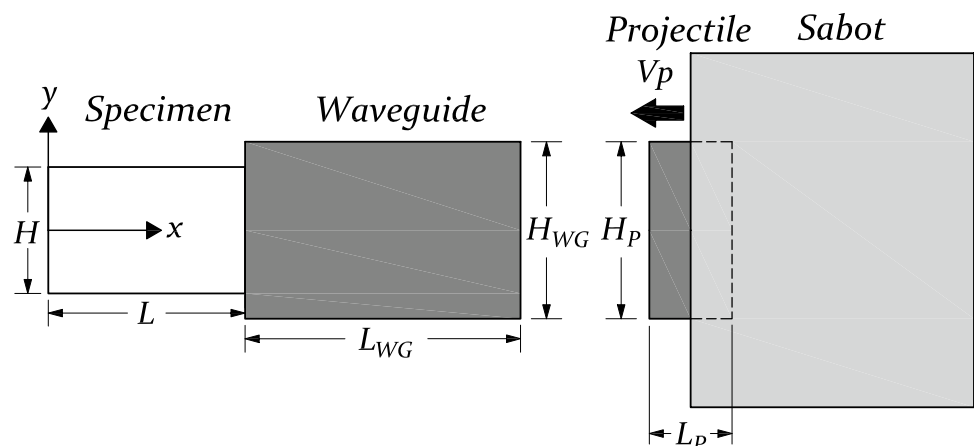
The IBII test configuration for a generic interlaminar specimen is shown in Fig. 2. Fibres are either parallel to the y-axis (1–3 plane), or parallel to the z-axis (2–3 plane), as shown in Fig. 1b and c, respectively. It is desirable to tailor the experimental parameters such that the reflected tensile stress is sufficiently high to cause failure. It is also desirable to maximise the ratio of reflected tensile stress to input compressive stress. This reduces the risk of damage during compressive loading. Numerical simulations are used to establish a design envelope such that both requirements are satisfied.

There are some important differences between the test design for interlaminar and in-plane properties. For direct imaging of the specimen, the geometry of an interlaminar test specimen is dependent on the thickness of the laminate. In this work, a thicker laminate (18 mm) is considered for practical reasons. Specifically, this enables plate-like

specimen to be machined and accurate grids to be deposited for full-field measurement purposes. To maximise measurement spatial resolution, a smaller grid pitch is required (on the order of 0.3 mm compared to 0.9 mm in [20]). With a smaller length compared to the transverse tension specimens, the wave transit time is also shortened. This requires a higher framing rate to ensure sufficient temporal resolution of the kinematic fields (the Shimadzu HPV-X camera allows for frame rates up to 5 Mfps at full resolution).

Separate design sweeps were performed for each interlaminar plane. The length of specimens from both material planes was fixed at the nominal plate thickness of 18 mm. A height of 12 mm was selected to maximise the spatial resolution of the camera (Shimadzu HPV-X, 400×250 pixels), including approximately 2 mm at the free edge of the specimen to account for rigid body motion. The material used in this study is a unidirectional carbon/epoxy composite, AS4-145/MTM45-1. The properties of this material were characterised by the National Center for Advanced Material Performance (NCAMP) as summarised in [28]. Unfortunately, only quasi-static interlaminar tensile strength was measured in that campaign ($\sigma_T^{ult} = 50.4$ MPa). Therefore, for test design it was assumed that reported values for the in-plane transverse Young's modulus, shear modulus, and Poisson's ratio were representative of that for the interlaminar planes. The quasi-static, in-plane transverse compressive strength ($\sigma_C^{ult} \approx 290$ MPa) was assumed as a conservative limit for allowable compressive stress. To the authors' knowledge, the high strain rate properties for this material have not been measured. Therefore, it was also assumed that the material strength will exhibit a similar strain rate sensitivity to that reported in [20] measured using the same IBII test (+ 57% increase in strength at strain rates on the order of 2000 s^{-1}). Even though reference [20] focussed on the in-plane transverse properties, the reported strain rate sensitivity was expected to be reasonably representative of the interlaminar behaviour as a matrix dominated property. Therefore, the interlaminar tensile strength at high strain

Fig. 2 Schematic of the IBII test configuration for interlaminar specimens



rates is estimated to be 80 MPa. For the IBII interlaminar test the design space is defined by the tensile and compressive strengths as: $-290 \text{ MPa} < \overline{\sigma_{xx}}^y < 80 \text{ MPa}$.

The design sweep in [20] showed that the experimental parameters that primarily influence the stress state in the material are projectile length and impact velocity. Therefore, these are the only parameters considered in the design sweep of the IBII test here. In the design sweep the height of the waveguide, H_{WG} , and projectile, H_p , were fixed at 25 mm. Having a larger projectile and waveguide improves contact alignment and stability of the sample on the test stand. This also prevents the sabot from striking the alignment stand. A separate simulation study showed that the impactor height has little influence on the test provided that it does not exceed three times the specimen width. The height of the sabot is also fixed at the barrel diameter of the purpose-built gas gun (50 mm). The projectile length and projectile speed were selected using a parametric design sweep. The range of simulated values for these variables are listed in Table 1.

The sabot length was variable such that the total length of projectile-sabot assembly was a constant 50 mm. A maximum projectile length of 20 mm was set to avoid creating an input pulse length that exceeds the specimen length. This would result in a superposition of the input and reflected waves, reducing the maximum tensile stress in the specimen. Since the waveguide and projectile are made of the same material, the waveguide length must be at least twice the length of the projectile to avoid clipping the pulse. Therefore, the waveguide length, L_{WG} , is fixed at 50 mm.

Finite Element Implementation

All simulations were performed in ABAQUS/Explicit. Plane stress CPS4R elements (2D, 4 node, reduced integration) were used in all simulations. The mesh size, and stiffness-proportional damping coefficient, β , were first selected using a separate parametric sweep. Some numerical damping is required to control the artificial high frequency oscillations that occur in the explicit dynamic simulations. The criterion for selection was minimisation of the error between the reconstructed stress averages, using the stress-gauge equation (Eq. 4), and simulated stress averages, over an entire wave reflection. This sweep resulted in a mesh size of 0.1 mm and β coefficient of 7×10^{-7} ms. For the remaining

components in the simulation a mesh size of 0.2 mm was used to maintain a similar mesh density. The time step incrementation was not fixed, however, the data output step was set to match that of the camera used for the experiments (0.2 μs inter-frame time). Isotropic linear elasticity was assumed for the projectile, waveguide, sabot and 2–3 plane interlaminar specimens. For 1–3 plane specimens a transverse isotropic, linear elastic material model was used. The simulated geometries and material properties for the interlaminar IBII test are listed in Table 2.

Parametric Sweep Results

The results from the parametric sweep for 1–3 and 2–3 plane specimens are shown in Fig. 3a and b, respectively. Based on the design space defined in Sect. 3.1, it is possible to select a range of experimental parameters that satisfy the design

Table 2 Simulation parameters and material properties for the interlaminar tension IBII test

Specimen (AS4-145/MTM45-1)	
E_{11} (GPa)	129 ^{a*}
E_{22} (GPa)	7.9 ^{a*}
E_{33} (GPa)	7.9 ^{a*}
G_{13} (GPa)	3.65 ^{a**}
G_{23} (GPa)	3.65 ^{a**}
ν_{13}	0.015 ^a
ν_{23}	0.225 ^a
Specimen length (mm)	18
Specimen height (mm)	12
ρ (kg m ⁻³)	1605 ^b
Mesh size (mm)	0.1
Waveguide and projectile (aluminium 6061-T6)	
E (GPa)	70
ν	0.3
ρ (kg m ⁻³)	2700
Waveguide length (mm)	50
Waveguide height (mm)	25
Projectile height (mm)	25
Mesh size (mm)	0.2
Sabot (nylon 6–6)	
E (GPa)	3.45
ν	0.4
ρ (kg m ⁻³)	1140
Sabot length (mm)	50
Sabot height (mm)	25
Mesh size (mm)	0.2

^aFrom [28]

^bMeasured using a micro balance and water immersion

*Average transverse modulus

**Average in-plane shear modulus

Table 1 Summary of simulated values used in parametric sweep for test design

Parameter	Min	Max	Increment
Projectile length, L_p (mm)	2.5	20	2.5 ($L_p \leq 5$) 5 ($L_p \geq 5$)
Impact speed, V_p (m s ⁻¹)	20	50	10

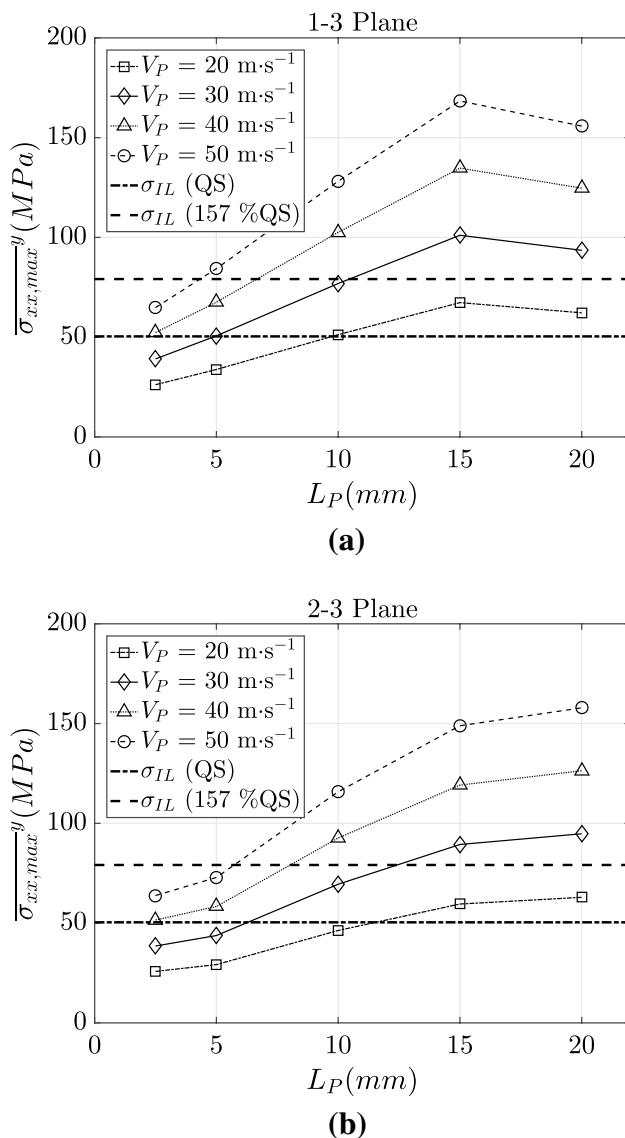


Fig. 3 Maximum reflected tensile stress, $\bar{\sigma}_{xx,max}^y$ for interlaminar IBII specimens as a function of projectile length and velocity: **a** 1–3 plane specimens, **b** 2–3 plane specimens

requirements. For both specimens, the design envelope that satisfies the requirements is given by: projectile length [$10 \text{ mm} < L_P < 20 \text{ mm}$] and impact velocity [$V_P > 40 \text{ m s}^{-1}$]. From this, the test configuration selected for both experiments is as follows: $L_{WG} = 50 \text{ mm}$, $L_P = 10 \text{ mm}$, and $V_P = 50 \text{ m s}^{-1}$. Based on simulated stress fields, the design strength of 80 MPa is expected to be reached at approximately $x = 10 \text{ mm}$ from the free edge.

Material and Experimental Setup

Specimen Manufacturing

Material properties of AS4-145/MTM45-1 are provided in Table 2. The density of the plate was measured using a micro balance and water immersion to be $1605 \pm 20 \text{ kg m}^{-3}$. The plate had an average measured cured thickness of 17.9 mm (est. 128 layers, 0.14 mm cured ply thickness [28]). Twenty interlaminar specimens were cut ($10 \times 1-3$ material plane and $10 \times 2-3$ material plane). The specimens were first rough cut from the plate using a large tile saw with a diamond cutting wheel. The specimen faces were then cut using a Streurs E0D15 diamond saw. The automated stage was set to a low feed rate of 0.1 mm s^{-1} to reduce the likelihood of inducing machining defects. For the 1–3 plane, specimen dimensions ($L \times H \times e$) were measured to be $17.9 \text{ mm} \times 12.1 \text{ mm} \times 2.6 \text{ mm}$ (SD $\pm 0.2 \text{ mm}$, $\pm 0.2 \text{ mm}$, $\pm 0.6 \text{ mm}$). Similarly, 2–3 plane specimen dimensions were measured to be $18.2 \text{ mm} \times 12.0 \text{ mm} \times 2.6 \text{ mm}$ (SD $\pm 0.1 \text{ mm}$, $\pm 0.3 \text{ mm}$, $\pm 0.4 \text{ mm}$). Note that reported thickness measurements include the grid deposited on the surface.

Grid Deposition Techniques

Grids with a pitch p of 0.3 mm were bonded to ten specimens (5 for each interlaminar plane), using the process outlined in [29]. The epoxy layer had a typical thickness of approximately $225 \mu\text{m}$. While this deposition procedure worked quite well in [20] for in-plane specimens, the smaller specimens in this study were more susceptible to grid defects from air bubbles in the underlying resin layer. Since grid defects are detrimental to the inverse identification procedures, a second grid deposition process was explored for the remaining ten specimens (5 for each interlaminar plane). A thin coat of white rubber paint (Rust-Oleum Peel Coat) was first applied to the specimen. The paint layer had a typical thickness of approximately $20 \mu\text{m}$. A series of black squares were then printed onto the painted surface with a Canon Océ Arizona 1260 XT flat bed printer. This formed a white grid with an average pitch of 0.337 mm. Trial prints of uniform grids were found to contain periodic defects every 80 mm in the horizontal direction. This was used to define the true print resolution and adjust the grid pitch when printing on specimens. As grids were defined according to a constant ‘points-per-pitch’ ratio (6:7 (x:y) closely matches that of the true resolution), the actual pitch in the vertical and horizontal directions is 0.338 mm and 0.336 mm, respectively. More information is available online [30, 31]. This

required processing of the grid images using the iterative procedure described in [32]. The limited printer resolution reduces the spatial resolution compared to the bonded grids. However, this was a manageable compromise given the simplicity of the deposition process and significant reduction in grid defects.

Specimen Naming Convention

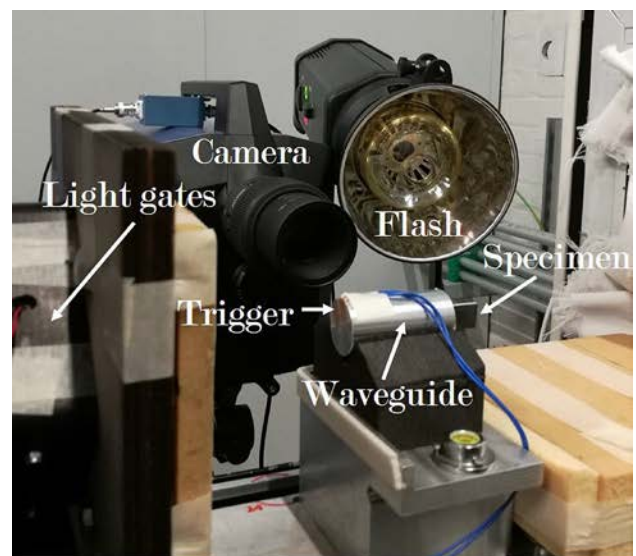
Specimens will be referred here by specimen number followed by a dash and a letter specifying the grid type; ‘P’ denotes a printed grid ($p = 0.337$ mm), and ‘B’ denotes a bonded grid ($p = 0.3$ mm), respectively. The interlaminar plane is specified in square brackets. For example, specimen #1 from the 1–3 plane with a 0.337 mm printed grid pitch is referred to as: ‘#1-P[1–3]’.

Experimental Setup

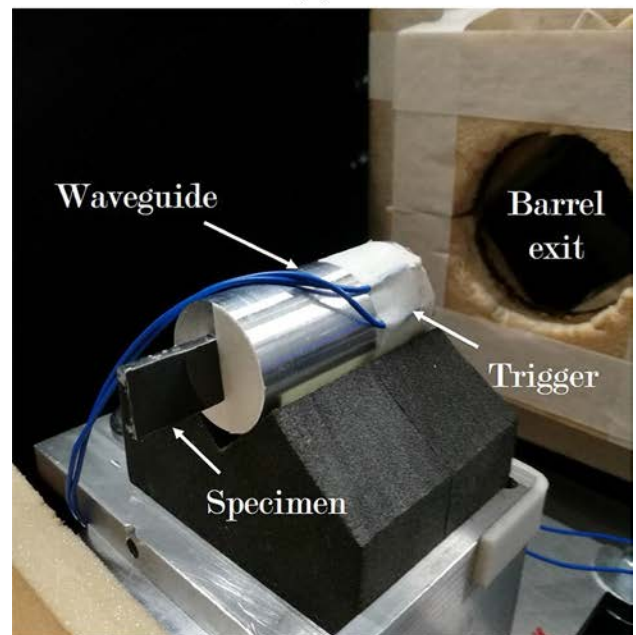
All tests were performed using the compressed air impact rig described in [20]. The gas gun reservoir pressure was set for a nominal impact velocity between 50 and 55 m s⁻¹. Each specimen was bonded to the back of a 6061-T6 aluminium waveguide (50 mm length, 25 mm diameter) using a thin layer of cyanoacrylate glue. A copper contact trigger on the front of the waveguide was used to trigger the camera. A 10 μ s delay was programmed between the trigger event and image capture to account for the traverse time of the wave through the waveguide. All images were captured using the Shimadzu HPV-X camera (frame rate = 5 Mfps) with a Sigma 105 macro lens. The optical setup and a mounted specimen are shown in Fig. 4a and b, respectively. Further details about the optical measurement system are provided in Table 3. The low fill factor of the Shimadzu HPV-X required intentional blurring of the images to avoid aliasing of the grid and parasitic fringes in the strain maps [33]. An out-of-plane movement test (2 mm) was performed prior to each test to minimise the strain fringes (below the noise threshold) and thus, ensure that the images were sufficiently blurred. The camera stand-off distance was adjusted according to the grid being imaged. To maximise the spatial resolution of the Shimadzu HPV-X camera, the 0.3 mm grids were sampled at 6 pixels per pitch, whereas the 0.337 mm grids were sampled at 7 pixels per pitch.

Image Processing and Identification of Material Properties

The full image processing procedure is described in the following sections to explain how material properties can be extracted from deformed images of a specimen with a grid on its surface. The key steps are summarised in a flow chart shown in Fig. 5.



(a)



(b)

Fig. 4 Experimental setup used for all interlaminar tests: **a** camera and flash arrangement around the test chamber, and **b** a mounted specimen supported on a test stand in the test chamber

Obtaining Displacement Fields from Deformed Grid Images

The Shimadzu HPV-X camera is used to collect a set of deformed grid images. These grid images are processed using the grid method to obtain phase maps, using a windowed discrete Fourier transform. For the bonded grids only, the phase maps were corrected for air bubble defects using a three-step procedure. (1) Each ϕ_x phase map was fitted with a mesh of linear finite elements (8×5 elements (x,y)) to capture gradients in the phase fields. The phase values at

Table 3 Imaging and measurement performance for the IBII tests with the grid method

Optical setup	
Camera	Shimadzu HPV-X
Pixel array size	400 × 250
Sensor	FTCMOS
Inter-frame time	0.2 μs
Integration (shutter) time	110 ns
Number of images	128
Lens	Sigma 105 mm
Flash	Bowens Gemini 1000 Pro
Grid method	
Grid pitch (mm)	0.3, 0.337
Sampling (pixels/period)	6, 7
Field of view (mm)	20 × 12.5, 19.25 × 12.04
Displacement computation	Iterative [32]

the nodal positions were determined using a least-squares regression fit and linear shape functions were used to interpolate the phase within each element. The regression plane fit was then subtracted from the raw phase field to obtain a map of residuals. Grid defects were first characterised by regions with phases values exceeding a 2σ threshold on the residual. (2) A second linear regression plane fitting was performed to the phase maps, with the defects identified in (1) removed. The full extent of the defect was characterised by again using a 2σ threshold. These defect maps were then used to remove defects in the ϕ_y maps. (3) A sliding square window of seven pitches in length was used to linearly interpolate the phase information over the defective regions identified in (2). The displacement fields were then computed from the ‘corrected’ phase maps using the iterative approach described in [32]. The iterative approach accounts for initial phase modulations in the grid (e.g., remaining small grid defects, slight grid spacing irregularities). The phase maps contain discontinuous jumps when the grid displaces more than one pitch. These jumps were corrected using spatial and temporal unwrapping. Spatial unwrapping was performed using the procedure described in [34]. Temporal unwrapping was performed using an in-house MATLAB routine. In this procedure, the spatial mean of the unwrapped phase is plotted against time and the 2π mean phase jumps in time are corrected to obtain a monotonic increase of the longitudinal displacement representative of the rigid body translation in the impact direction.

Obtaining Strain and Acceleration Fields

One pitch of information is corrupted on the border of the phase maps due to edge effects from the windowed Fourier transform. Rather than discarding this data, previous studies have shown that identifications using the virtual

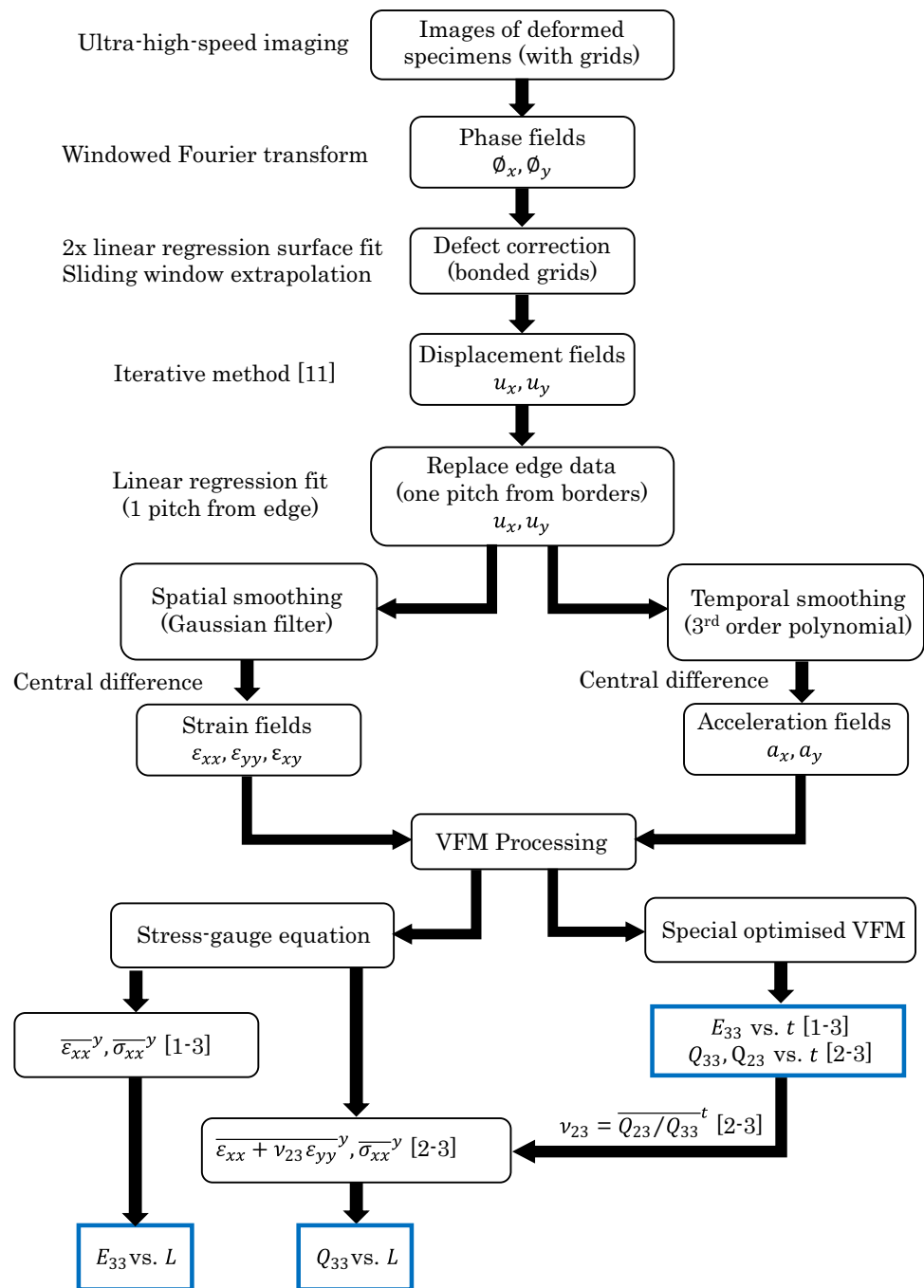
fields method were drastically improved when this data was recovered using some sort of extrapolation [35, 36]. In this work, the corrupted displacement data was replaced using a linear regression fitting based on the data over one pitch (6 pixels (0.3 mm grids) or 7 pixels (0.337 mm grids)) inwards from the corrupted region. The extrapolation was performed independently for each row of pixels (u_x fields), or column of pixels (u_y fields). This approach was found to be better at rejecting noise compared to the approach in [20], where data was recovered using a linear extrapolation based on two points inward from the corrupted region. The displacement maps were then processed in two ways to obtain acceleration and strain fields (Fig. 5). Displacement maps were smoothed temporally using a third order Savitsky–Golay filter, and then differentiated twice in time to obtain acceleration maps. Displacement fields were padded in time by one half of the kernel size (in frames) to minimise edge effects from the filter. Raw displacement maps were also smoothed spatially, using a Gaussian filter, before differentiating to obtain strain maps. Both temporal and spatial differentiations were computed using a central difference. Strain rate maps were computed from the smoothed strain maps using a central difference, except for computing strain rate at fracture, which was performed using a backward difference based on the raw strain maps to avoid temporal leakage from unrealistic strains computed after crack initiation. To reduce edge effects from spatial smoothing, the displacement fields were first padded out by 3 smoothing kernels using a linear extrapolation. The fields were smoothed and then cropped back to the original size. The details of the corrections and smoothing can be consulted in the Matlab program provided as supplementary material with this article.

Identifying Material Properties from Kinematic Fields

The special optimised virtual fields methods presented in Sect. 2.1 were used to process acceleration and strain fields to identify stiffness parameters. Material properties were identified from acceleration and strain maps using two VFM approaches: (1) using special optimised virtual fields, and (2) using reconstructed stress averages to compute stress–strain curves at each position along the specimen length.

Special Optimised Virtual Fields The special optimised virtual fields approach provides an identification of each stiffness parameter for each time step. For 1–3 plane specimens, the reduced approach was used to process a_x and ϵ_{xx} fields to directly identify E_{33} . For 2–3 plane specimens, the a_y , ϵ_{yy} and ϵ_{xy} fields were also included in the general isotropic formulation of the special optimised virtual fields approach. In this case, Q_{33} and Q_{23} were identified, from which E_{33} and ν_{23} were determined. The value of each identified stiffness parameter for the test was taken as the average over all time steps where the identification is stable. The

Fig. 5 Flow chart of processing procedure to identify material stiffness parameters from deformed grid images. Note that the exact same procedure is used for processing the experimental data and the image deformation simulations



identification is generally poor during the first few frames of the test due to low strains as the wave enters the specimen. Stability is also challenged as the wave reflects from the free-edge, as the incoming and reflected waves superimpose, resulting in temporary low strain and acceleration signal. When the material cracks in tension, non-physical strains corrupt the identification. Therefore, optimal conditions for identification (high strain and acceleration signal) generally occurs during the first compressive loading after the stress wave has entered the specimen, but before it reflects at the

free edge. The optimized virtual fields were expanded using a basis of piecewise functions (finite elements), as proposed initially in [37]. A virtual mesh refinement study was performed on the image deformation data (described in Sect. 5) and the results showed that the identification converged at a virtual mesh of 5×1 elements (x,y) and 5×4 elements (x,y) for the reduced and isotropic special optimised routines, respectively. Data is discarded within one pitch plus one spatial smoothing kernel at the impact edge to reduce smoothing filter edge effects on the identification.

Reconstructed Stress–Strain Curves Here, the stress–gauge equation (Eq. 4) was used to calculate stress averages ($\overline{\sigma_{xx}^y}$) from a_x fields for all specimens. Using these stress averages, combined with axial strain ($\overline{\epsilon_{xx}^y}$ for 1–3 plane specimens, and $\overline{\epsilon_{xx} + \nu_{23}\epsilon_{yy}^y}$ for 2–3 plane specimens), stress–strain curves were generated along the length of the specimen. For the case of 2–3 plane specimens, the identified value of ν_{23} from the special optimised virtual fields procedure was used. The interlaminar stiffness (E_{33} for 1–3 plane specimens, and Q_{33} for 2–3 plane specimens) was identified using a linear regression fit to the stress–strain curve up to the maximum compressive stress. This is henceforth referred to as the ‘stress–strain curve’ approach. E_{33} was calculated for 2–3 plane specimens using the identified value of ν_{23} from the special optimised virtual fields approach. The identification of E_{33} tends to be poor near the free and impact edges due to extrapolated data at the edges of the specimen, and edge effects from spatial smoothing. Therefore, the value of E_{33} for the test was taken as the average of identified values over the middle 50% of the specimen. The stress-average reconstructed at the first crack location using linear-stress gauge equation was used to estimate the interlaminar tensile strength.

Clearly, the selection of spatial and temporal smoothing parameters will influence the identification procedures. The smoothing parameters were selected using an image deformation simulation procedure similar to [36, 38, 39], as described in the following section.

Smoothing Parameter Selection and Error Quantification

Generating Synthetic Images

The purpose of this section is to select optimal smoothing parameters to be used for the experiments and estimate the experimental error using an image deformation procedure. The general concepts are described here, but the reader is referred to [36, 38, 39] for further details. A sequence of ‘static’ synthetic images were generated for both types of grids used in the experiments using an analytical function to describe the light intensity, $s(x, y)$. For the white-on-black grids, with a 0.3 mm pitch, the intensity at any position was described as:

$$s(x, y) = A + \gamma \cdot \left(1 + \frac{1}{2} (1 + \cos(2\pi x/p)) \cdot (1 + \cos(2\pi y/p)) \right) \quad (6)$$

while the intensity at any position for the black-on-white grids, with a 0.337 mm pitch, was defined as:

$$s(x, y) = A + \gamma \cdot \left(1 - \frac{1}{2} (1 + \cos(2\pi x/p)) \cdot (1 + \cos(2\pi y/p)) \right) \quad (7)$$

where A is the average grey level illumination, γ is the pattern contrast amplitude (between 0 and 1), and p is the grid pitch. Displacement fields from finite element simulations were used to create a set of deformed images using super-sampling interpolation. Up-sampled images were generated, and then sub-sampled by pixel averaging to simulate the resolution of the Shimadzu HPV-X camera (400×250 pixels). Specifically, synthetic images were generated for the bonded, 0.3 mm, white-on-black grids, and the printed, 0.337 mm, black-on-white grids, using contrast values measured from experimental static images. The parameters used to generate each set of synthetic images are listed in Table 4. Magnified views of the synthetic grid images are compared with experimental grids in Fig. 6. The lighting gradient along the specimen length is more pronounced on the black-on-white grids. Therefore, the synthetic image is based on average intensity and contrast of the experimental grids over the length of the specimen. This explains the slight differences in contrast between synthetic and experimental 0.337 mm pitch grids.

The deformed synthetic images were then processed using the same procedure as the experimental images (Sect. 4.5). Different combinations of spatial and temporal smoothing were used to quantify the effect of processing parameters on the identification of stiffness parameters.

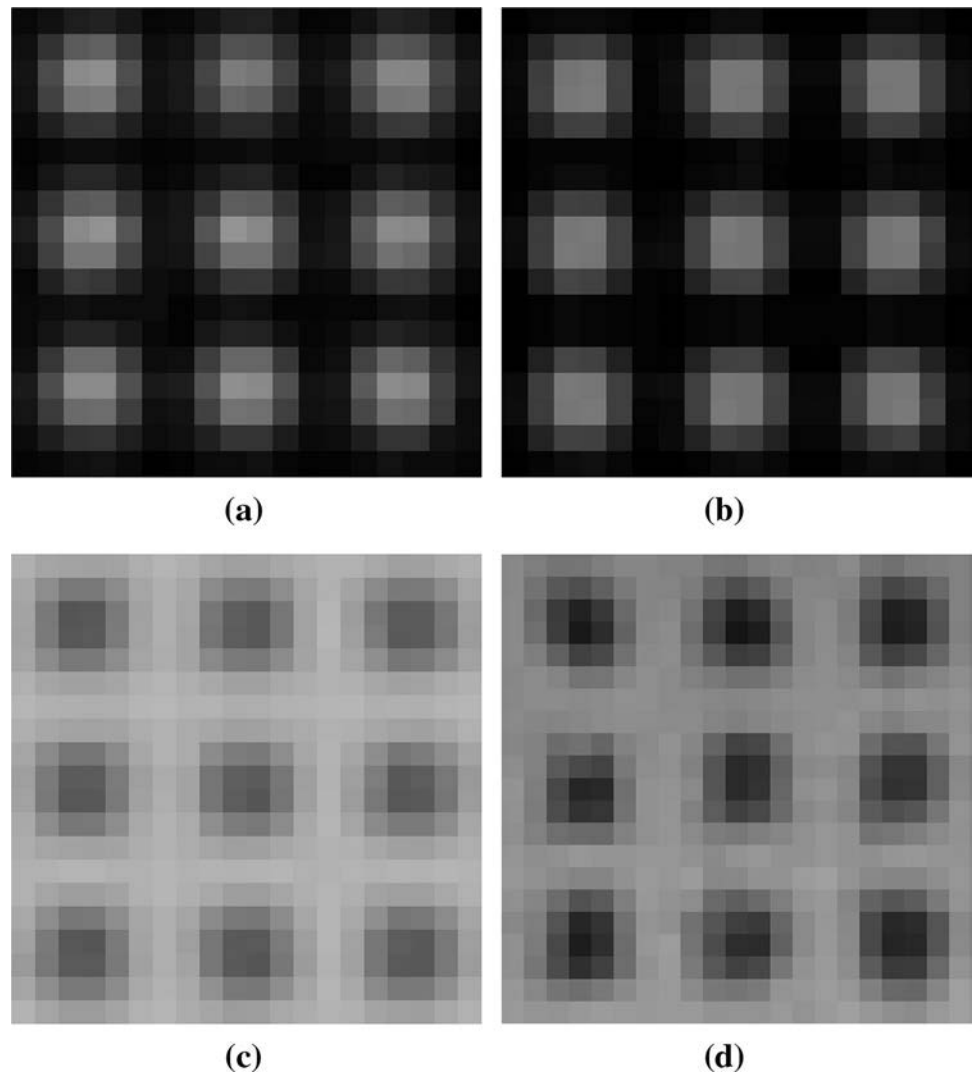
Identification Sensitivity to Smoothing Parameters

Each combination of smoothing kernels was used to process synthetic images with 30 copies of noise. Gaussian white noise with a uniform standard deviation was used to approximate experimental noise. In reality, noise is dependent on grey level intensity, which could be accounted for in the future. The standard deviation was set to that measured from a series of static grid images captured with the Shimadzu HPV-X camera (Table 4). The sensitivity to smoothing parameters was assessed using the maximum total error (e_T) between the reference stiffness value and the value identified from the processed synthetic images. The total error was defined as the absolute value of the systematic error e_S , plus or minus two times the random error, e_R ($e_T = |e_S \pm 2e_R|$). The systematic error (e_S) was considered as the difference between the mean identified

Table 4 Summary of parameters used to generate synthetic images for processing parameter optimisation

Image parameter	Printed grids	Bonded grids
Grid pitch (mm)	0.3	0.337
Mean grey level (% dyn. range)	50	40
Grid contrast amp. (% dyn. range)	20	25
Noise amplitude (% dyn. range)	0.4	0.25
Pixel sampling (pixels/period)	7	6

Fig. 6 Magnified views of grid images: **a** 0.3 mm synthetic grid (6 pixels/period), **b** 0.3 mm experimental grid (6 pixels/period), **c** 0.337 mm synthetic grid (7 pixels/period), and **d** 0.337 mm synthetic grid (7 pixels/period)



stiffness parameter and the reference stiffness, normalised by the reference stiffness as,

$$e_{S,ij} = \frac{1}{Q_{FE,ij}} \left(\frac{1}{N} \sum_{k=1}^N Q_{ID,ij}^k - Q_{FE,ij} \right) \quad (8)$$

for $i = 2, 3, j = 3$

where $Q_{ID,ij}^k$ is the identified stiffness parameter from the image deformation simulations for noise iteration k , N is the number of noise copies ($N = 30$), and $Q_{FE,ij}$ is the reference stiffness value used to generate the simulated displacement fields and deformed images. Random error (e_R) was defined as the standard deviation of the identified stiffness over the 30 copies of noise normalised by the reference stiffness as,

$$e_{R,ij} = \frac{1}{Q_{FE,ij}} \sqrt{\frac{\sum_{k=1}^N (Q_{ID,ij}^k - \overline{Q_{ID,ij}})^2}{N}} \quad (9)$$

for $i = 2, 3; j = 3$

where $\overline{Q_{ID,ij}}$ is the mean identified stiffness parameter from the image deformation simulations over all copies of noise. The reader is encouraged to recall Sect. 4.5.3 for a description of the identification procedures.

For each combination of smoothing kernels, the systematic, random and total error were calculated. Examples of systematic and random error maps for the identification of E_{33} [1–3] with 0.3 mm grid, using the reduced special optimised virtual fields approach are shown in Fig. 7a, b, respectively. The error maps were very similar for identifications using the stress-gauge approach and therefore, only maps for the optimised VFM will be presented. The systematic and random errors follow similar trends as shown in previous works for quasi-static tests [38], with minimum systematic error and high random error when no smoothing is applied. The band of low systematic error represents a consistent trade-off between bias from smoothing acceleration fields (temporal smoothing), and

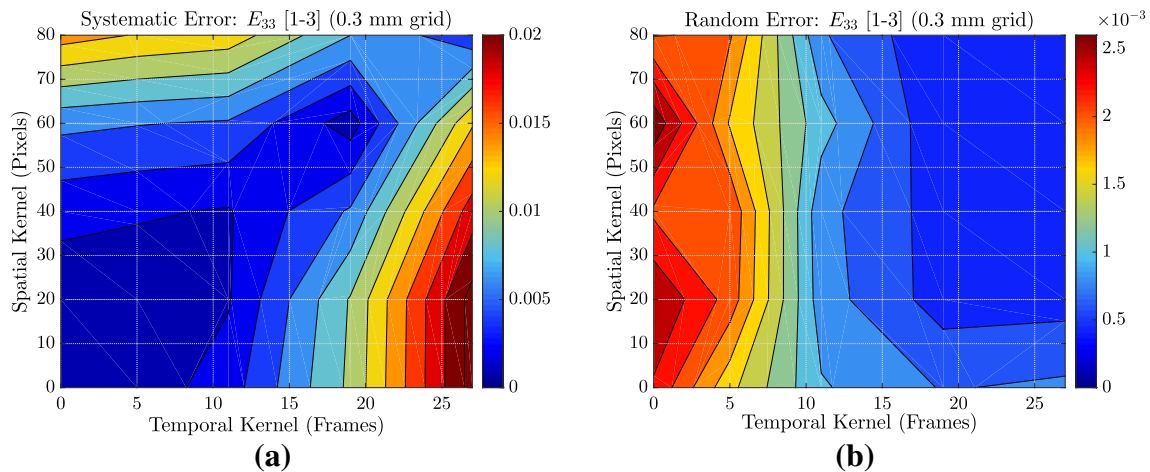


Fig. 7 Simulated identification error for E_{33} as a function of spatial and temporal smoothing kernel size using the reduced special optimised routine (1–3 plane specimen, 0.3 mm grid, 6 pixels per period sampling): **a** normalised systematic error, and **b** normalised random error

strain fields (spatial smoothing), such that the reference stiffness is most accurately identified. The magnitude of random error is low since the identification is derived from the average over several temporal frames. The random error is more strongly influenced by temporal smoothing (noise in acceleration signal) since the optimised virtual field routine are optimised to minimise strain noise, and not acceleration noise, and there is a double differentiation in time compared to a single differentiation in space.

For the 1–3 interlaminar plane specimen, the total error maps are shown in Fig. 8a (0.3 mm grid, 6 pixels/period) and b (0.337 mm, 7 pixels/period). The total error maps for 6 and 7 pixels/period are very similar in shape and magnitude. This implies that grid sampling has only a small influence on the identification in the present case. These

maps suggest that optimal levels of smoothing correspond to a spatial kernel of 41 pixels, and a temporal kernel of 11 frames. For these parameters the estimated error on E_{33} is approximately 0.5%.

The total error maps for Q_{33} and Q_{23} , for 2–3 plane specimens with 0.3 mm (6 pixels/period) and 0.337 mm (7 pixels/period) grids are shown in Figs. 9 and 10, respectively. The higher total error at low smoothing indicates a higher sensitivity to random error. This is to be expected as ϵ_{yy} and a_y have a lower signal-to-noise ratio. The inclusion of these fields into the identification increases the sensitivity to noise of both Q_{33} and Q_{23} , which are identified simultaneously. Nevertheless, the minimum total error is not significantly increased compared to the 1–3 plane identification, with the minimum occurring around a 41 pixels spatial smoothing

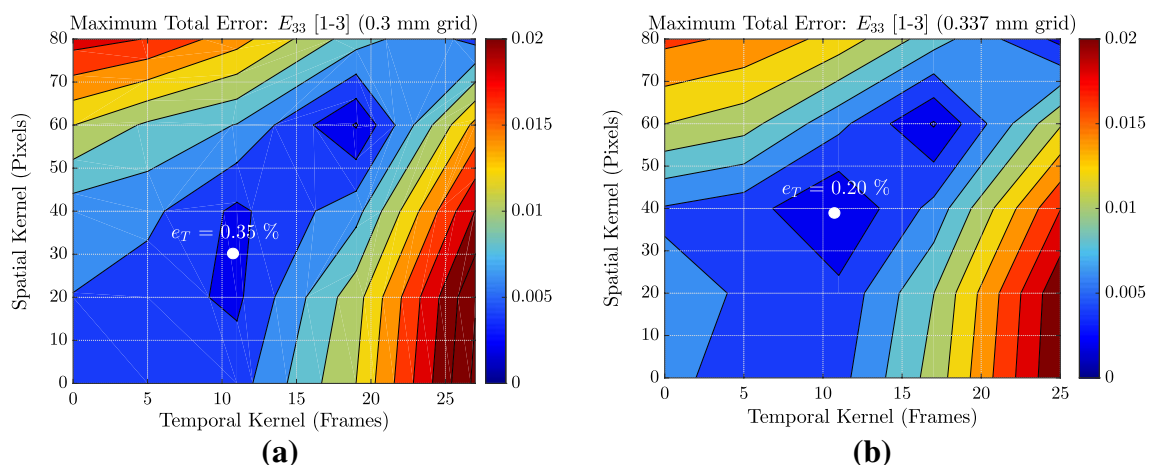


Fig. 8 Maximum, simulated identification error for E_{33} as a function of spatial and temporal smoothing kernel size using the reduced special optimised routine (1–3 plane specimen): **a** normalised total error

for 0.3 mm grid (6 pixels/period sampling), and **b** normalised total error for 0.337 mm grid (7 pixels/period sampling). Minimum error indicated by white circle

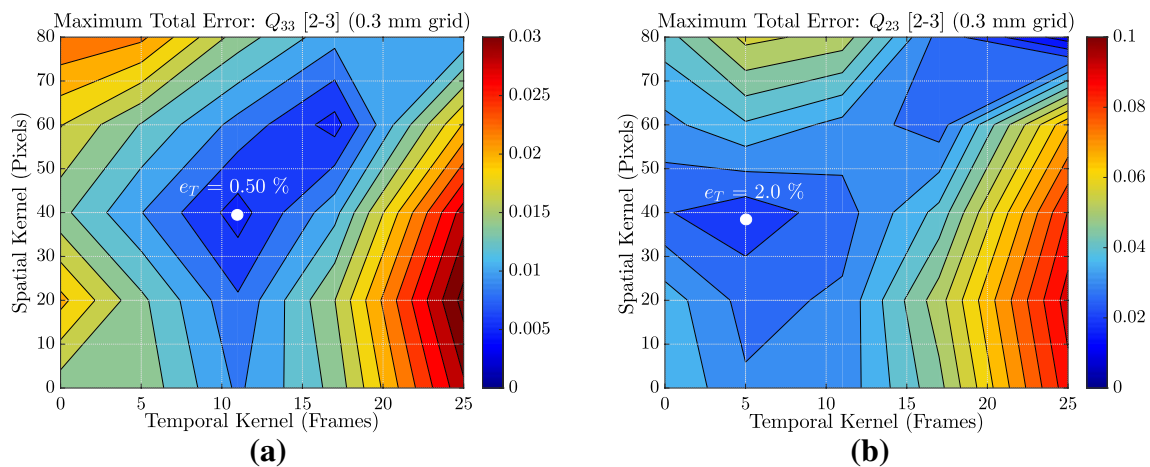


Fig. 9 Maximum, simulated identification error for Q_{33} and Q_{23} as a function of spatial and temporal smoothing kernel size using the reduced special optimised routine (2–3 plane specimen, 0.3 mm grid,

6 pixels/period sampling): **a** normalised total error for Q_{33} , and **b** normalised total error for Q_{23} . Minimum error indicated by white circle

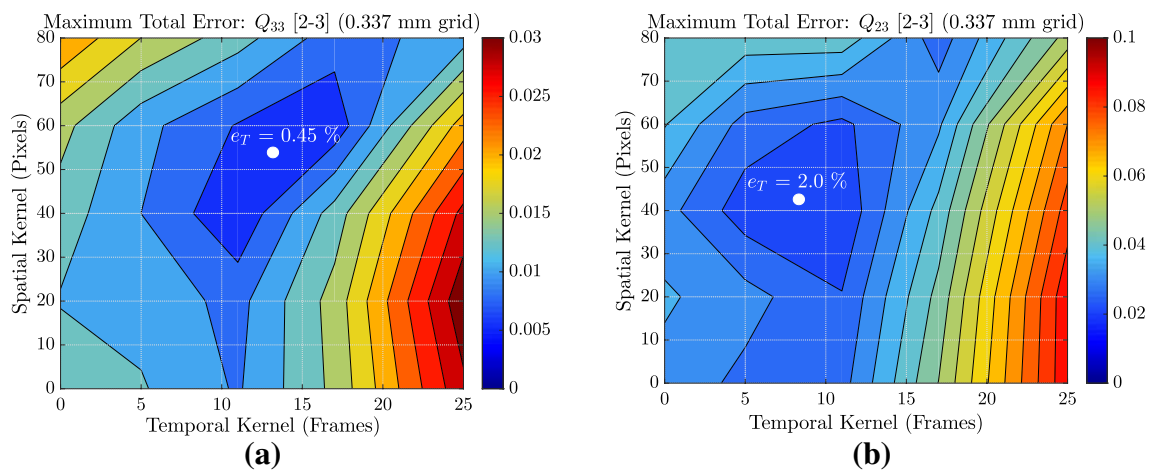


Fig. 10 Maximum, simulated identification error for Q_{33} and Q_{23} as a function of spatial and temporal smoothing kernel size using the reduced special optimised routine (2–3 plane specimen, 0.337 mm

grid, 7 pixels/period sampling): **a** normalised total error for Q_{33} , and **b** normalised total error for Q_{23} . Minimum error indicated by white circle

kernel, and 11 frames for the temporal smoothing kernel. These parameters have an associated error on the identification of Q_{33} and Q_{23} of approximately 0.5% and 2–3%, respectively.

The total error maps are extremely useful for selecting the optimal smoothing parameters in a rational way and provide an estimate of the total error associated with the experimental identifications. The optimal smoothing parameters selected for processing experimental images are listed in Table 5. Also included are the corresponding error estimates and measurement resolution (standard deviation of field) from static experimental images processed with the selected smoothing parameters. Image deformation simulations indicate that the predicted errors are very low, despite having limited spatial resolution.

Simulations also suggest that this configuration is highly robust to spatial and temporal smoothing, so long as the user does not select extreme smoothing parameters. This is a key advantage of this IBII test configuration. The user can quickly establish reasonable limits on the bounds of smoothing for a given grid pitch assuming a rough knowledge of the material in question. For example, looking at the contour maps, the temporal smoothing is the parameter that mostly influences the error. The total error sharply rises when the temporal smoothing kernel approaches the time for the wave pulse to traverse across the specimen. Therefore, so long as the user can estimate the wave speed in the material with reasonable accuracy, an upper bound on temporal smoothing can be quickly established. Likewise, simulations suggest that spatial smoothing does not

Table 5 Selected smoothing parameters for processing experimental images and corresponding measurement performance

	Grid Pitch (mm)			
	1–3 Plane		2–3 Plane	
Parameter	0.3	0.337	0.3	0.337
Spatial Kernel (pixels)	31	41	41	41
Temporal Kernel (frames)	11	11	11	11
Error Q_{33} (%)	0.4	0.2	0.5	0.5
Error Q_{23} (%)	–	–	2.8	2.1
Measurement resolution				
Displacement (μm)	0.3	0.4	0.3	0.4
(Pixel)	0.006	0.008	0.006	0.008
(Pitch)	$p/1000$	$p/850$	$p/1000$	$p/850$
Strain ($\mu\text{m m}^{-1}$)	46	56	46	56
Acceleration ($\times 10^5 \text{ m s}^{-2}$)	8.4	5.1	8.4	5.1

significantly effect the identification so long as the total kernel is less than 10 grid pitches.

Experimental Results and Discussion

Full-Field Measurement Results

In this section typical experimental kinematic fields are presented for two time steps for a 1–3 plane and 2–3 plane inter-laminar specimen. Since it is difficult to get a true appreciation for the dynamic nature of an impact test through still images, videos of all kinematic field for all specimens, as well as the raw grey level images can be found in the data repository detailed at the end of the manuscript.

Full-field maps of u_x and u_y are shown for specimen #2-P[1–3] in Fig. 11 for two time steps; the first time step corresponds to a state where the initial compressive pulse is well within the specimen ($t = 7 \mu\text{s}$), and the second corresponds to a time after the pulse has reflected but before tensile failure ($t = 17 \mu\text{s}$). Similar maps for specimen #6-B[2–3] are provided in Fig. 12. Note that the mean u_x displacement has been subtracted to remove the rigid-body displacement

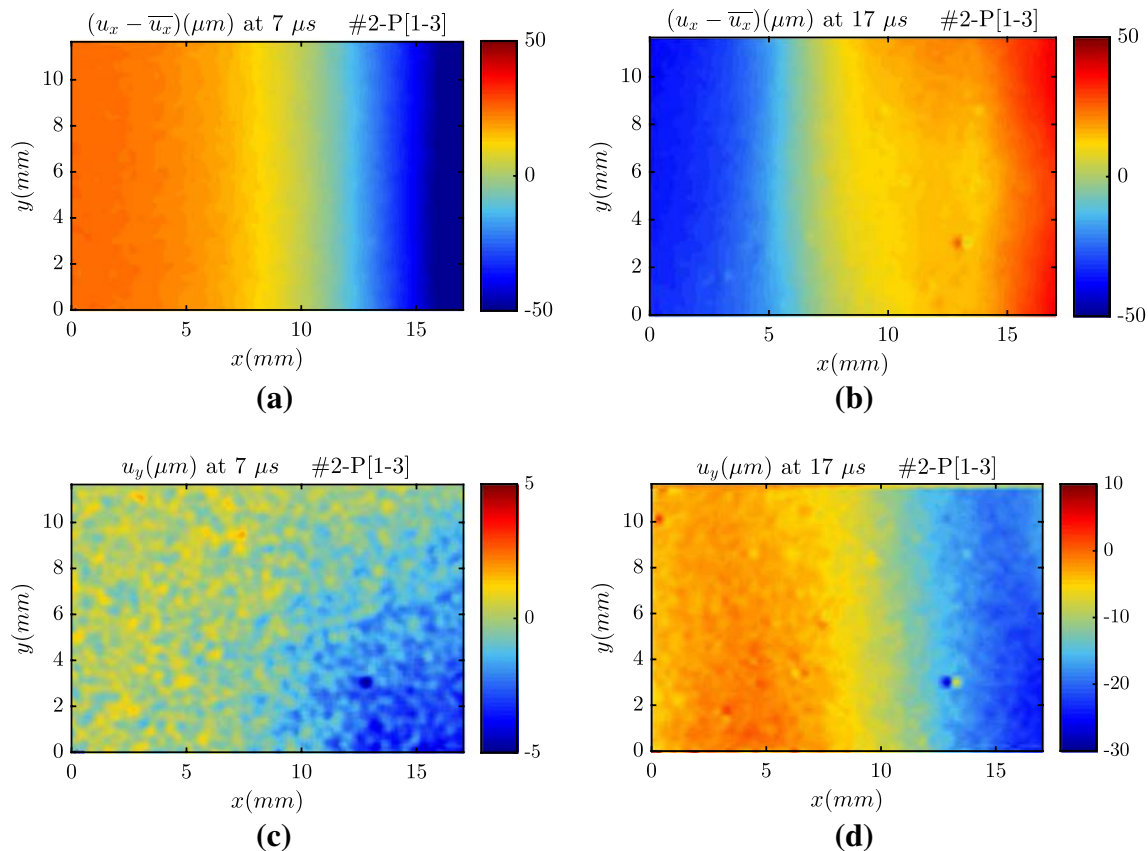


Fig. 11 Experimental displacement fields; **a, b** u_x (μm), and **c, d** u_y (μm), for specimen #2-P[1–3] at $7 \mu\text{s}$, and $17 \mu\text{s}$. Note that the mean u_x displacement has been subtracted to remove the rigid-body displacement

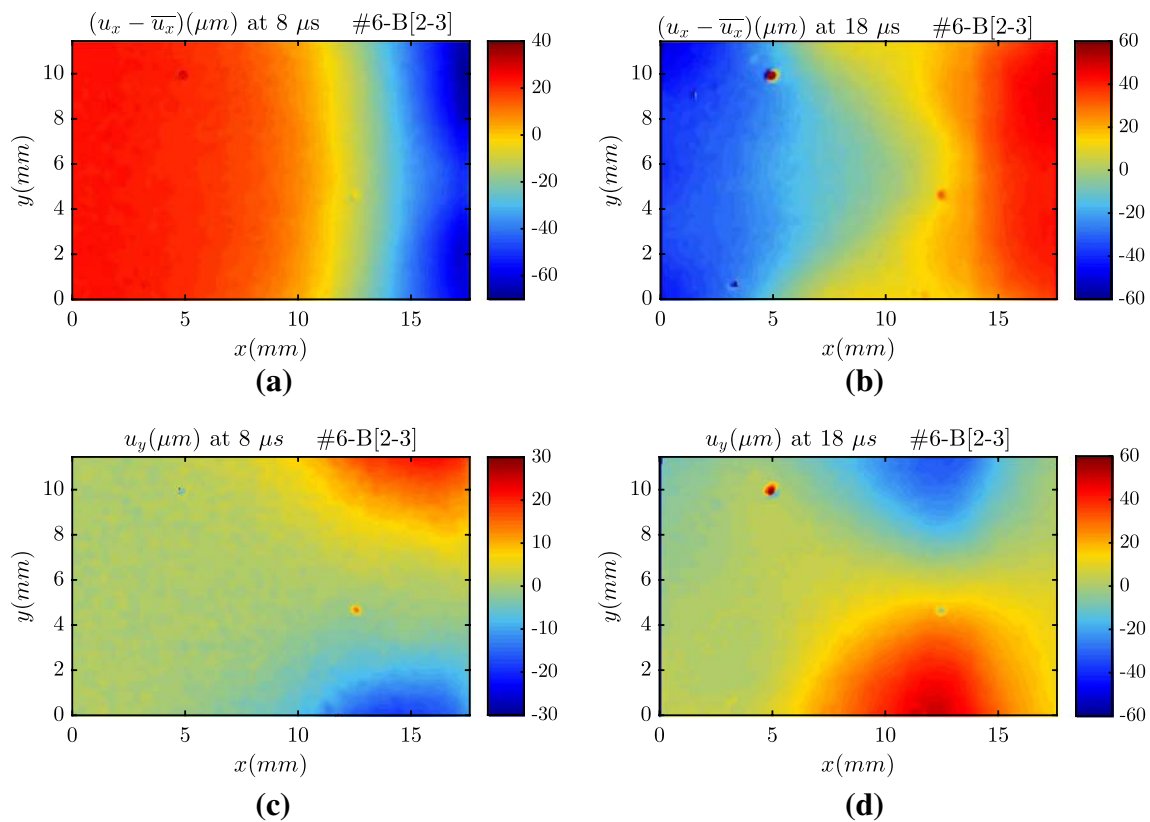


Fig. 12 Experimental displacement fields; **a, b** u_x (μm), and **c, d** u_y (μm), for specimen #6-B[2–3] at 8 μs , and 18 μs . Note that the mean u_x displacement has been subtracted to remove the rigid-body displacement

and show the deformation. Due to the high lateral stiffness from the fibre reinforcement for specimen #2-P[1–3], the magnitude of u_y is much smaller and the signal to noise ratio is much poorer compared to u_x . In the case of specimen #6-B[2–3], u_y is approximately one order of magnitude larger. The corresponding acceleration fields for the two specimens are shown in Figs. 13 and 14. The edges of the pulse are most clearly identified in a_x maps, which show that local accelerations are on the order of 10^7 m s^{-2} . This corresponds to average axial forces on the order of 4 kN, or axial stress on the order of 100 MPa (Fig. 15). Defects are difficult to remove from acceleration fields since no temporal information is considered in the correction procedure. While this provides a reasonable reconstruction in the majority of cases, the defective region is still identifiable in some frames, as shown in Fig. 14d. While defects cannot be completely removed, providing some compensation for defects is beneficial as it acts as an outlier removal. This is particularly beneficial for identifications from stress–strain curves, which rely on local strain values.

The ϵ_{xx} , ϵ_{yy} and ϵ_{xy} strain maps for the two specimens, at the same two time steps, are shown in Figs. 16 and 17. The experimental strain maps for the 1–3 interlaminar specimen confirm that the ϵ_{yy} strains are much lower than for the ϵ_{xx}

strains. Significant ϵ_{yy} strains are measured in the case of the 2–3 plane interlaminar specimen, however, the signal-to-noise ratio remains less favourable than the ϵ_{xx} strains. As a result, the edge data extrapolation procedure does not perform as well, creating some localised regions of high artificial strain. As mentioned previously, despite having high signal in the ϵ_{xx} fields, the lower signal in the ϵ_{yy} and ϵ_{xy} fields will act to reduce the identification stability since these strains are used to simultaneously identify Q_{33} and Q_{23} . Strain rate maps ($\dot{\epsilon}_{xx}$) are shown in Figs. 18 and 19 for specimen #2-P[1–3] and specimen #6-B[2–3], respectively. Local $\dot{\epsilon}_{xx}$ strain rates are on the order of 4000 s^{-1} to 7000 s^{-1} . For specimen #6-B[2–3], $\dot{\epsilon}_{yy}$ were measured on the order of 3000 s^{-1} . Strain rates in the experiments are slightly lower than that predicted from processed synthetic images based on simulated fields (peak compressive strain rates on the order of $14,000 \text{ s}^{-1}$ and peak tensile strain rates on the order of $10,000 \text{ s}^{-1}$). As previously explained, this is expected since the simulation assumes perfect, hard contact at waveguide interfaces between the projectile and specimen. Some ‘pulse-shaping’ is expected in the experiments from the thin layer of tape on the front face of the waveguide for the camera trigger, and the thin layer of glue between the waveguide and specimen.

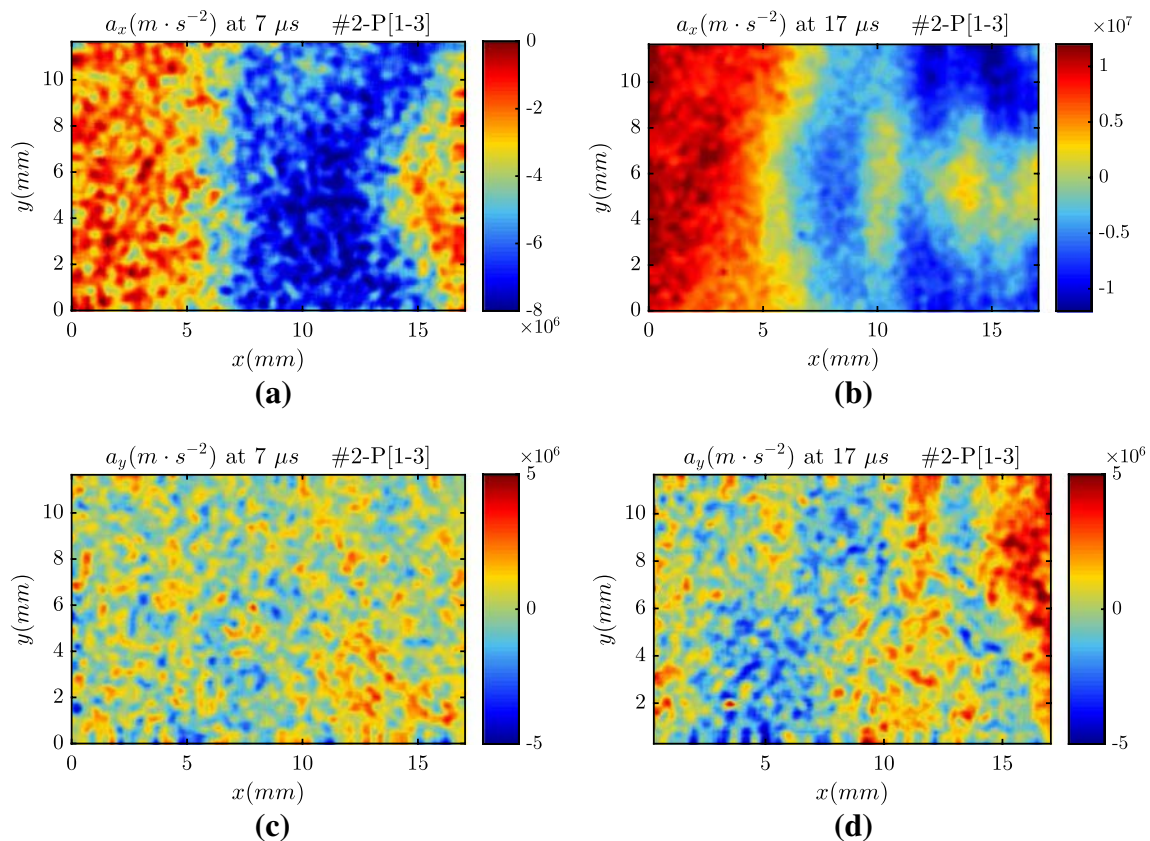


Fig. 13 Experimental acceleration fields; **a**, **b** a_x (m s^{-2}), and **c**, **d** a_y (m s^{-2}), for specimen #2-P[1–3] at 7 μs , and 17 μs

Interlaminar properties were identified from the acceleration and strain fields using the methods presented in Sect. 2 and Sect. 4.5.3. The identification of interlaminar stiffness parameters are presented in Sect. 6.2 and the identification of interlaminar tensile strength are presented in Sect. 6.3.

Stiffness Identification

Experimental measurements of interlaminar stiffness parameters are presented separately for each of the identification techniques. The identifications using the special optimised virtual fields methods are presented first in Sect. 6.2.1 followed by the stiffness identifications with the stress–strain curve approach in Sect. 6.2.2.

Special Optimised Virtual Fields Approach

Identifications of E_{33} with the reduced optimised virtual fields method for all specimens are shown in Fig. 20. Similarly, identifications of Q_{33} and Q_{23} are shown in Figs. 21 and 22, respectively. From identifications of Q_{33} and Q_{23} , ν_{23} and E_{33} can be determined (not shown). The identified value for all parameters was taken as the average over the time frames that Q_{33} was stable. Note that identifications

are stopped just prior to the wave reflection from the free edge (approx. $t = 10 \mu\text{s}$). The identifications do not recover beyond this due to data reconstruction errors at the free edge, low strains as the wave reflects, and the formation of macro cracks shortly after. A summary of identified interlaminar stiffness parameters for 1–3 plane specimens and 2–3 plane specimens is provided in Tables 6 and 7, respectively. In the case of specimens #6-B[1–3], #7-B[1–3], #3-P[2–3], #5-P[2–3] and #10-B[2–3], the trigger delay was too long and a true static reference was not captured, as the wave had partially propagated into the specimen in the first image. To determine how far the pulse had propagated into the specimen when image capture began, a reference image taken before the test was correlated with the first image captured during the test. The region where ϵ_{xx} exceeded the noise floor was used to locate the pulse front. Images were then reprocessed with this region excluded (these specimens marked with a superscript ‘ r ’ in Table 6 and Table 7). Note that this does not affect interlaminar strength measurements as the test was designed so that failure occurs in the middle of the specimen, where a reference grid was maintained.

The random error associated with identifications for 2–3 plane specimens is higher compared to the 1–3 plane. A possible explanation for this is the inclusion of a_y , ϵ_{yy}

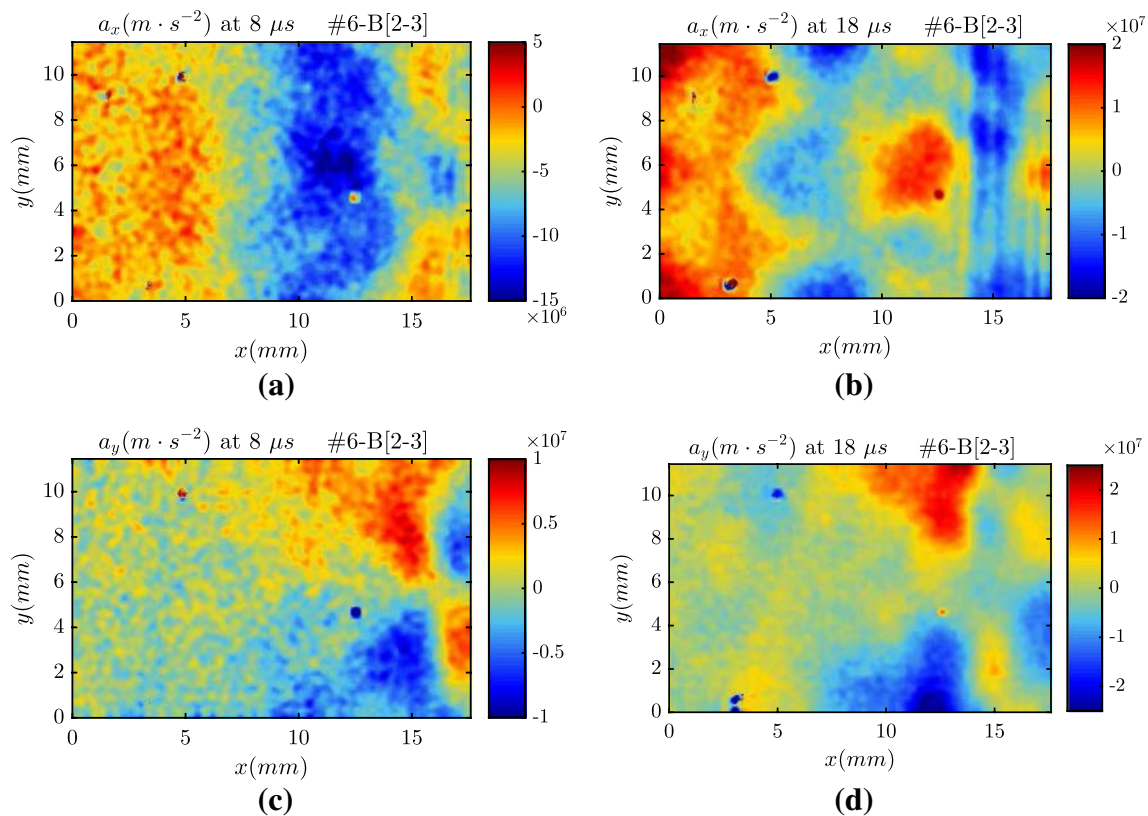


Fig. 14 Experimental acceleration fields; **a**, **b** a_x (m s^{-2}), and **c**, **d** a_y (m s^{-2}), for specimen #6-B[2–3] at 8 μs , and 18 μs

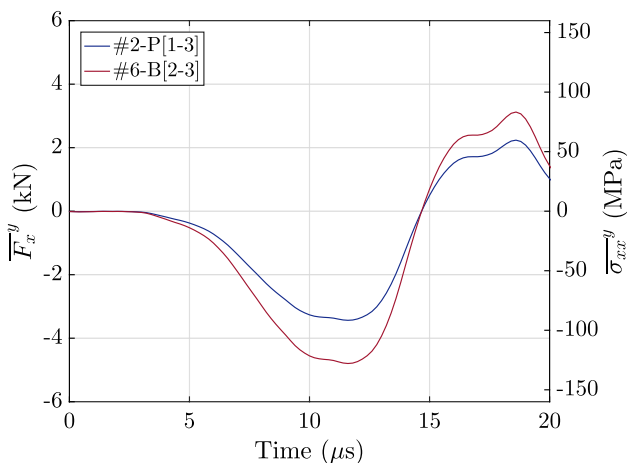


Fig. 15 Average axial force and average axial stress profiles for specimens #2-P[1–3] and #6-B[2–3]

and ϵ_{xy} fields in the identification procedure, which have a lower-to-noise ratio. The optimised virtual fields method is formulated such that each set of virtual fields results in the direct identification of each stiffness parameter. Since these parameters are identified simultaneously, the identification of a weakly activated material parameter will influence the identification of the other parameters.

The minimisation based on strain noise also makes the identifications sensitive to defects, which have a high signal-to-noise ratio, particularly in the case where the activated strain fields have low signal-to-noise-ratio. This is particularly problematic when defects occur around the edges of the specimen and interact with the edge extrapolation procedures. This will inevitably influence the identification routines and may account for the higher inter-specimen scatter in identified values for 2–3 plane specimens. To some extent, grid defects and their interaction with low signal to noise ϵ_{yy} strains will also influence the identification of Q_{33} using stress–strain curves (presented in Sect. 6.2.2). However, this is thought to be minimal since strains generated from the Poisson effect are small compared to the axial strains. The interaction of the virtual fields with defects may also explain the low-frequency oscillations in the identification of E_{33} from some 1–3 plane specimens (e.g., #1-P[1–3]), however further investigation is required to confirm this.

The average value for $E_{33}[1-3]$ was 10.9 GPa with a coefficient of variation (COV) of 3.5 %. This level of scatter is quite low and comparable to that for quasi-static testing of this material (COV = 3.6 %) [28]. $E_{33}[2-3]$ is identified from Q_{33} and ν_{23} with an average value of 10.4 GPa, and a COV of 6.1%. The slightly higher scatter in $E_{33}[2-3]$ is likely caused by the inclusion of fields with low signal-to-noise ratios into

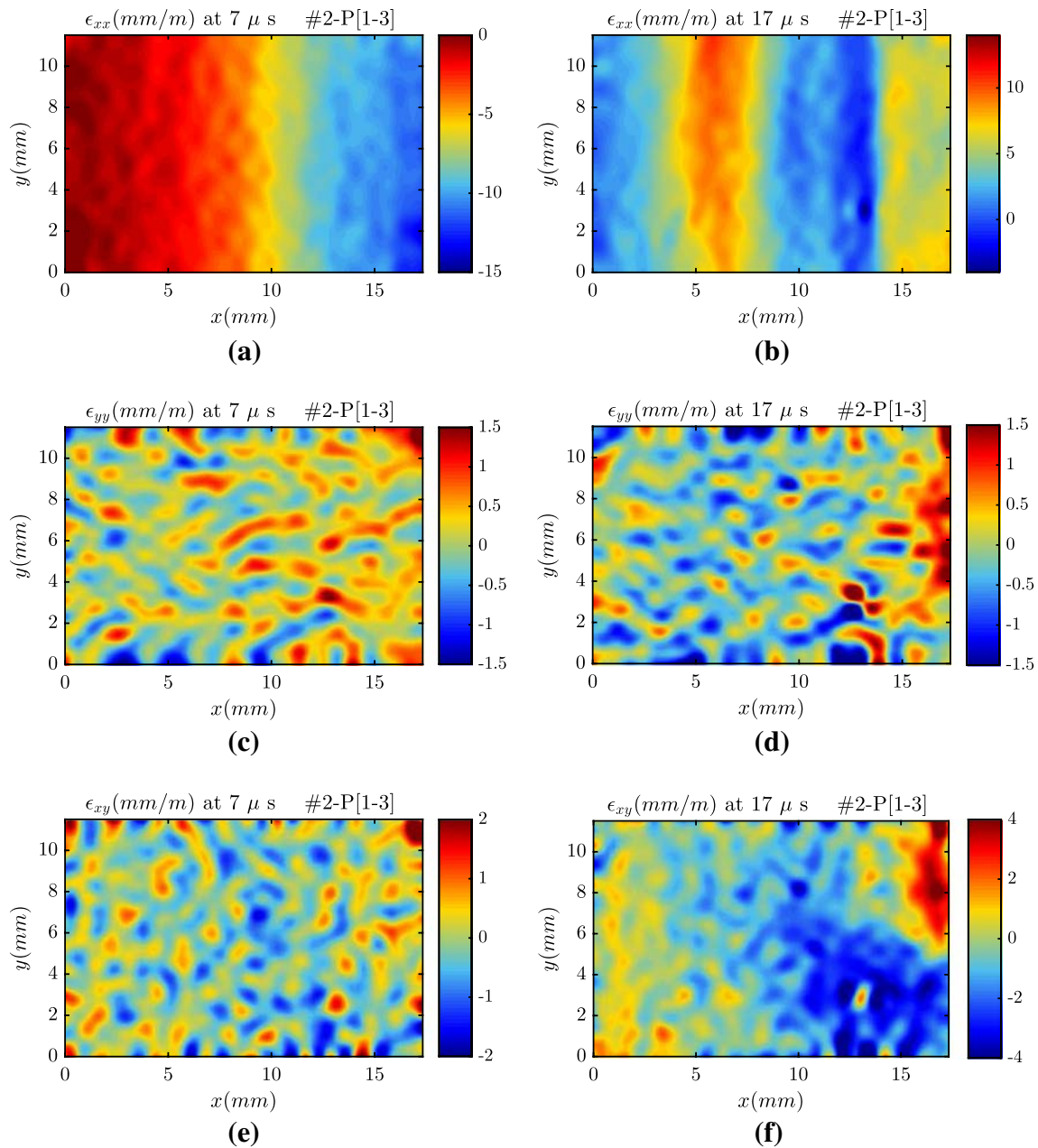


Fig. 16 Experimental strain fields; **a, b** ϵ_{xx} (mm m^{-1}); **c, d** ϵ_{yy} (mm m^{-1}), and **e, f** ϵ_{xy} (mm m^{-1}) for specimen #2-P[1-3] at 7 μs , and 17 μs

the identification routine as described previously. Therefore, the stiffness measured on the 1–3 plane specimens is thought to be more reliable. The measured interlaminar modulus of 10.9 GPa represents an increase of 38% compared to quasi-static values [28].

Overall, the measurements are quite promising considering this is the first implementation of the IBII test to obtain interlaminar properties at such high strain rates. Regarding strain rate, it is difficult to assign a single strain rate value to the measurements due to the heterogeneity of the fields. However, when axial strain is high, so too is strain rate (see

Figs. 16, 17, 18, 19). Therefore, the peak, width-average strain rate ($\dot{\epsilon}_{xx}^{\text{peak}}$) achieved during the compressive loading sequence can be considered as the limiting case for an ‘effective’ strain rate for these measurements (Table 8). It is part of future work to derive an effective strain rate using the virtual fields and the virtual strain rate fields to be able to quote a well-defined value. However, the strain rate sensitivity of this material is low enough so that this is not a critical issue and a mean or peak value is a good estimate. For most specimens, the peak compressive strain rate is on the order of 3500 s^{-1} . Obtaining stiffness measurements at such strain

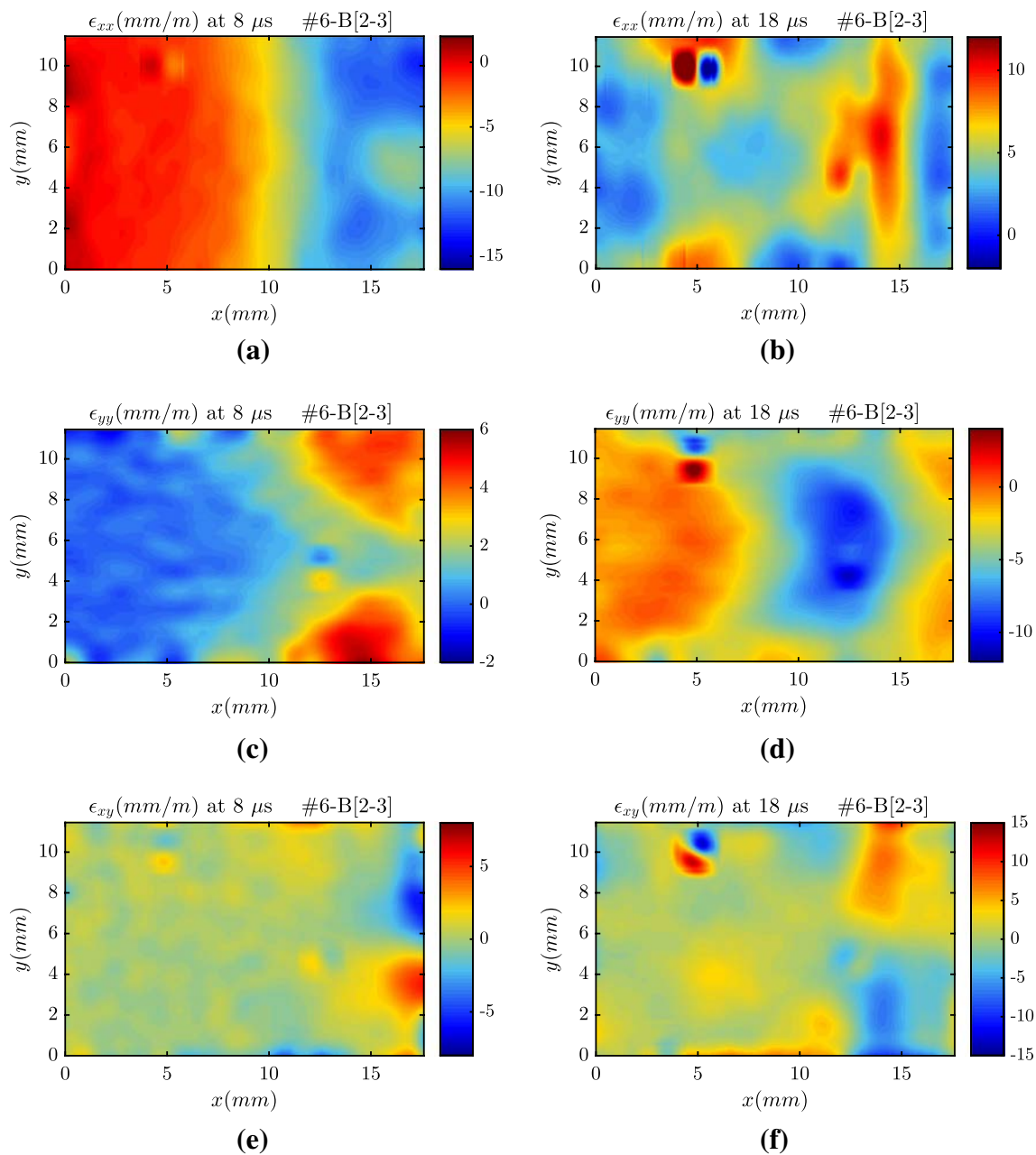


Fig. 17 Experimental strain fields; **a, b** ϵ_{xx} (mm m^{-1}); **c, d** ϵ_{yy} (mm m^{-1}), and **e, f** ϵ_{xy} (mm m^{-1}) for specimen #6-B[2–3] at 8 μs , and 18 μs

rates with the SHPB test is challenging and generally unreliable. Identifications using reconstructed stress–strain curves are presented next.

Stress–Strain Curve Approach

The stress-gauge equation is used to reconstruct $\overline{\sigma_{xx}}^y$ at each cross-section. This can be drawn as a function of $\overline{\epsilon_{xx}}^y$ (1–3 plane) or $\overline{\epsilon_{xx} + \nu_{23}\epsilon_{yy}}^y$ (2–3 plane) to generate stress–strain curves at each cross-section. Examples of stress–strain curves generated at a cross-section near the middle of the

sample are shown for specimens #2-P[1–3], #7-B[1–3], #2-P[2–3], and #6-B[2–3] in Fig. 23. The linearity of the stress–strain responses is quite remarkable considering the high strain rates at which these measurements are made. A linear regression fitting to the compressive loading region of the curve was used to identify the interlaminar stiffness at each cross-section.

The spatial identification of E_{33} is presented for all 1–3 plane specimens in Fig. 24, and for all 2–3 plane specimens in Fig. 25. Note that in the case of 2–3 plane specimens, the identified value of ν_{23} from the special optimised virtual

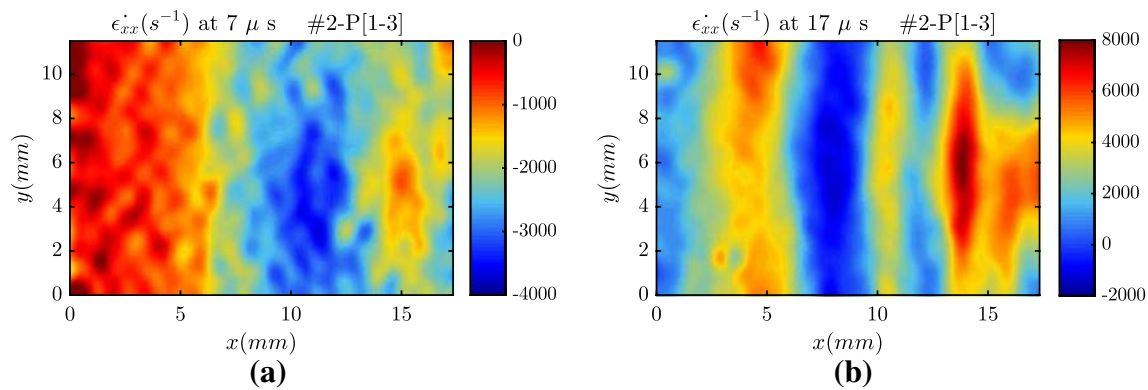


Fig. 18 Experimental strain rate fields; **a, b** ϵ_{xx} (s^{-1}) for specimen #2-P[1–3] at 7 μs , and 17 μs

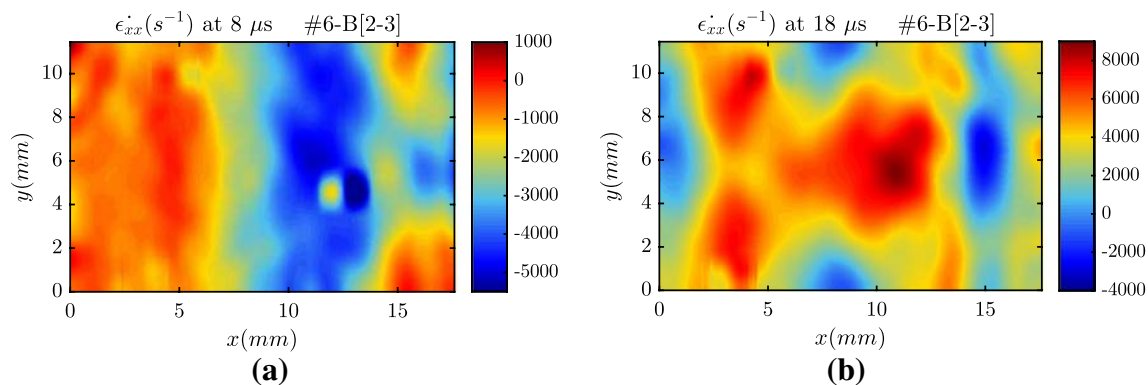
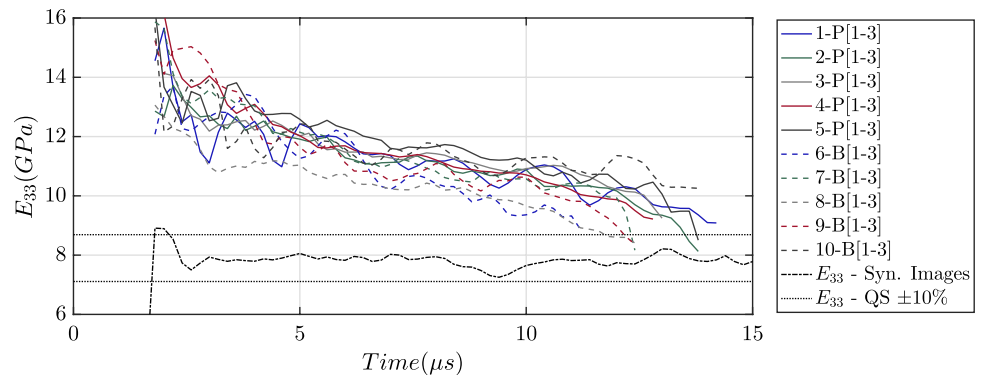


Fig. 19 Experimental strain rate fields; **a, b** ϵ_{xx} (s^{-1}) for specimen #6-B[2–3] at 8 μs , and 18 μs

Fig. 20 Interlaminar Young's modulus, E_{33} , identified for all 1–3 plane specimens using the reduced special optimised virtual fields method. Identification from image deformation simulation processed with the same smoothing parameters is provided for comparison



fields procedure is used for determining Q_{33} . Also note that one pitch, plus one smoothing kernel is excluded from the free edge and impact edge. The identification is unreliable in these regions due to smoothing edge effects and low spatial averaging near the free edge.

In general, the identifications from reconstructed stress–strain curves are quite consistent and stable over the middle portion of the specimen. A slightly lower stiffness is measured near the free edge on some samples with bonded 0.3 mm grids, particularly specimens #6-B[1–3],

#8-B[1–3], and #9-B[1–3]. Since this is not observed on specimens with printed, 0.337 mm grids, this is thought to be a result of a slight rotation of the grid with respect to the specimen. This could also be a result of some missing data at the edge from trimming the overflow epoxy during the grid de-bonding process. Both have the effect of increasing the amount of missing data at the free edge, and thus the error on reconstructed stress. This highlights another key advantage of using printed grids, as alignment is easily and consistently controlled, and no additional

Fig. 21 Interlaminar Q_{33} stiffness identified for all 2–3 plane specimens using the isotropic formulation of the special optimised virtual fields method. Identification from image deformation simulation processed with the same smoothing parameters is provided for comparison

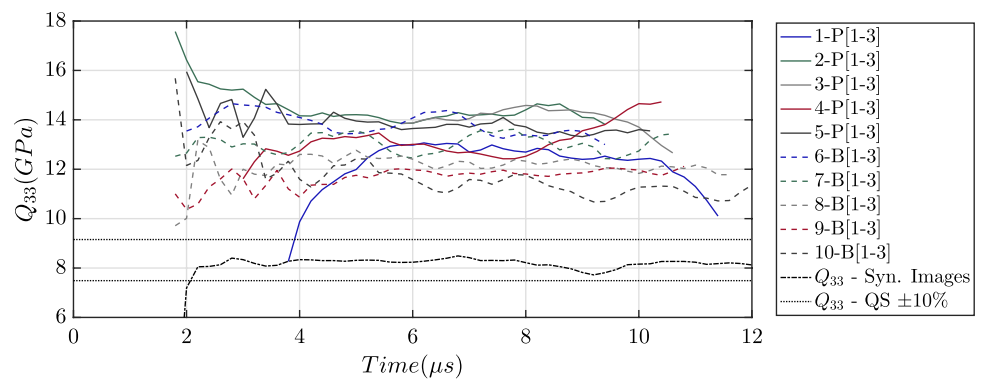


Fig. 22 Interlaminar Q_{23} stiffness identified for all 2–3 plane specimens using the isotropic formulation of the special optimised virtual fields method. Identification from image deformation simulation processed with the same smoothing parameters is provided for comparison

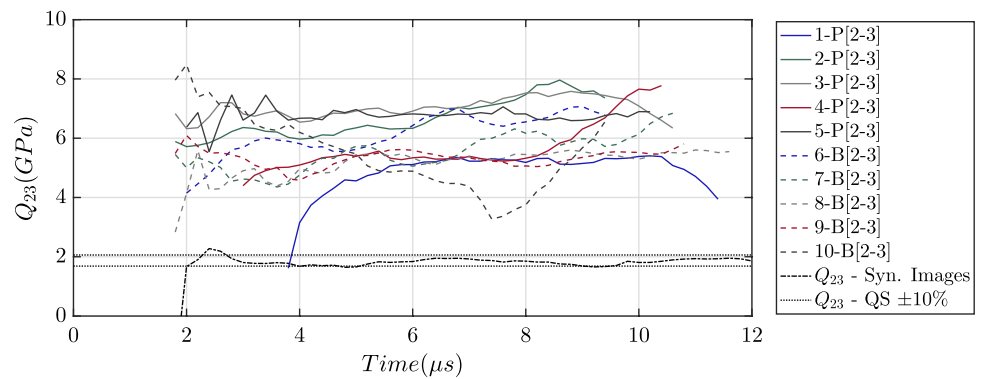


Table 6 Measured high strain rate interlaminar elastic modulus for AS4-145/MTM45-1 (1–3 plane specimens)

Specimen	1–3 plane	
	E_{33} (SS) (GPa)	E_{33} (VFM) (GPa)
1-P	10.3	11.0
2-P	10.3	10.8
3-P	10.5	11.7
4-P	10.5	11.0
5-P	10.8	11.1
6-B ^r	10.3	10.3
7-B ^r	10.3	10.7
8-B	9.8	10.3
9-B	9.9	10.9
10-B	10.9	11.3
Mean	10.4	10.9
SD	0.34	0.35
COV (%)	3.3	3.5
Diff. to Q–S (%)	+ 31	+ 38

VFM special optimised virtual fields approach, SS stress–strain curve approach

^rIdentification performed over reduced field of view

steps are required to clean up the edges of the specimen following grid application. The identification of E_{33} [1–3] and E_{33} [2–3] was measured to have an average value of 10.4 GPa, and 10.2 GPa, respectively. The coefficient of variation for both types of specimens is low ranging between 3 and 6%. Between 1–3 and 2–3 plane specimens, this represents a 30% increase in stiffness compared to quasi-static transverse stiffness measurements ($E_{33} = 7.9$ GPa) [28].

Figures 24 and 25 show that the identifications of E_{33} and Q_{33} fluctuate periodically along the length of the specimen by approximately 0.5 GPa from the mean. This pattern was also observed in the identifications from synthetic images, although with a lower magnitude (0.2 GPa). Image deformation simulations suggest that this oscillation is primarily attributed to fluctuating error on reconstructed acceleration and strain as the pulse moves through the extrapolation region at the free edge (one pitch). Extrapolation errors are highest as the high signal information within the pulse travels through the extrapolated region. It is thought that the experimental images are more sensitive to this since the pulse is smoother compared to the simulated pulse (i.e., extrapolation errors affect high signal

Table 7 Measured high strain rate interlaminar stiffness for AS4-145/MTM45-1 (2–3 plane specimens)

Specimen	2–3 plane				
	Q_{33} (SS) (GPa)	Q_{33} (VFM) (GPa)	ν_{23} (VFM)	E_{33} (SS) (GPa)	E_{33} (VFM) (GPa)
1-P	12.6	12.5	0.41	10.4	10.5
2-P ^r	14.1	14.1	0.45	11.3	11.3
3-P ^r	13.1	13.9	0.50	9.9	10.4
4-P	13.1	13.1	0.43	10.7	10.7
5-5 ^r	13.7	13.7	0.49	10.4	10.4
6-B	12.1	13.9	0.44	9.8	11.2
7-B	13.0	13.1	0.43	10.7	10.7
8-B	11.7	12.2	0.44	9.4	9.9
9-B	13.3	11.9	0.46	10.5	9.4
10-B ^r	11.8	12.15	0.44	9.4	9.5
Mean	12.8	13.0	0.45	10.2	10.4
SD	0.83	0.87	0.03	0.61	0.64
COV (%)	6.5	6.7	6.4	6.0	6.1
Diff. to Q–S (%)	–	–	–	+ 30	+ 32

VFM special optimised virtual fields approach, SS stress–strain curve approach

^rIdentification performed over reduced field of view**Table 8** Peak compressive width-average strain rate ($\overline{\epsilon_{xx}^y}$)

Specimen	1–3 plane	2–3 plane
	$\overline{\epsilon_{xx}^y}$ (s ^{−1})	$\overline{\epsilon_{xx}^y}$ (s ^{−1})
1-P	− 3000	− 2200
2-P	− 3100	− 4400
3-P	− 3700	− 3800
4-P	− 3200	− 2800
5-P	− 3700	− 1700
6-B	− 4800	− 4500
7-B	− 3900	− 5100
8-B	− 3300	− 2900
9-B	− 3300	− 5400
10-B	− 2000	− 3000

information for a longer period of time). The effect is more pronounced on the 0.337 mm grids due to lower measurement resolution and a larger extrapolation region at the free edge.

To the authors' knowledge, no interlaminar high strain rate data are available for AS4-145/MTM45-1. Furthermore, existing studies are limited to reporting an 'apparent' modulus, and scatter is so large that strain rate effects cannot be reliably extracted [2]. Therefore, it is difficult to make direct comparisons with other studies. This preliminary study shows that a consistent measurement of E_{33} can be obtained by processing measured strain and acceleration fields in two different ways. Both approaches are suitable for identifying a global stiffness value in this study since material properties

do not vary in space or time. Therefore, a comparison of the two methods provides a kind of validation of the measured values. The use of image deformation also shows that both routines can identify the reference E_{33} within 1% when smoothing parameters are chosen appropriately. However, when material properties vary in space and time preference might be given to a single identification method depending on the information desired. For example, the stress–strain curve method might be preferred in cases where a spatial variation in stiffness is of interest, since it provides a stiffness measurement for each transverse slice along the length of the specimen. Alternatively, the special optimised virtual fields might be more useful if one was interested in resolving time-dependent behaviours, as it gives a single stiffness value for each point in time. This level of information and versatility is not available with existing test methods and highlights the potential of image-based test methods for high strain rate testing.

Strength Identification

The linear stress–gauge equation (Eq. 5) is used to estimate the tensile strength of the material, as in [20]. A comparison of the stress fields reconstructed using the identified constitutive model and the linear stress–gauge equation are shown at a frame before fracture ($t = 15.0 \mu\text{s}$) in Fig. 26a and b, respectively. The agreement is excellent apart from the region close to the impact, demonstrating that for such a test, the linear representation in Eq. 6 provides a reasonable actual approximation of the stress field. Fracture initiation is

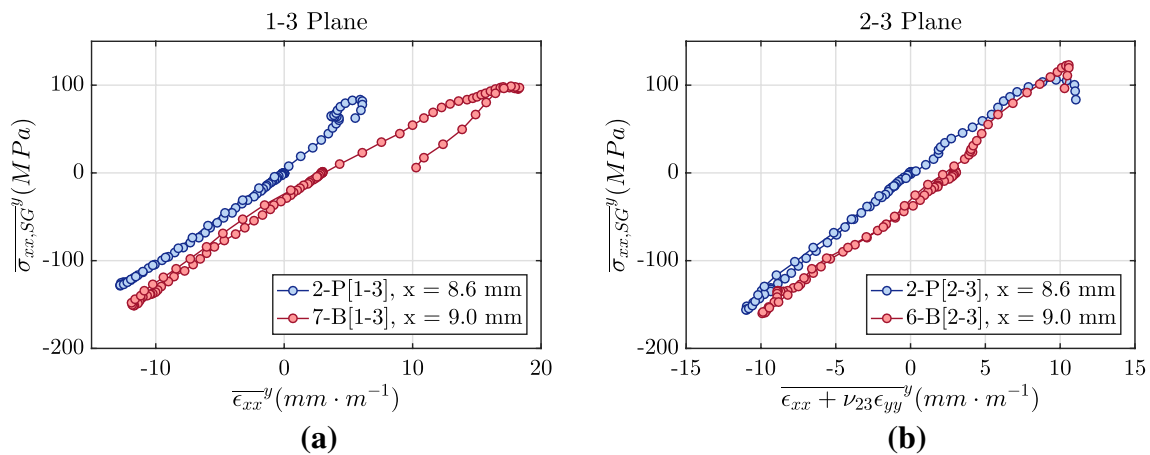


Fig. 23 Stress–strain curves generated near the middle of the sample using the stress-gauge equation for: **a** specimens #2-P[1–3] and #7-B[1–3] and **b** specimens #2-P[2–3] and #6-B[2–3]. Note that all stress–strain curves begin near the origin but have been offset by 3 mm m^{-1} for clarity

Fig. 24 Interlaminar Young's modulus, E_{33} , for all 1–3 plane specimens identified using the stress-gauge approach. Identification from image deformation simulation processed with the same smoothing parameters is provided for comparison. Note that the extrapolated data at the edges of the specimen has been removed

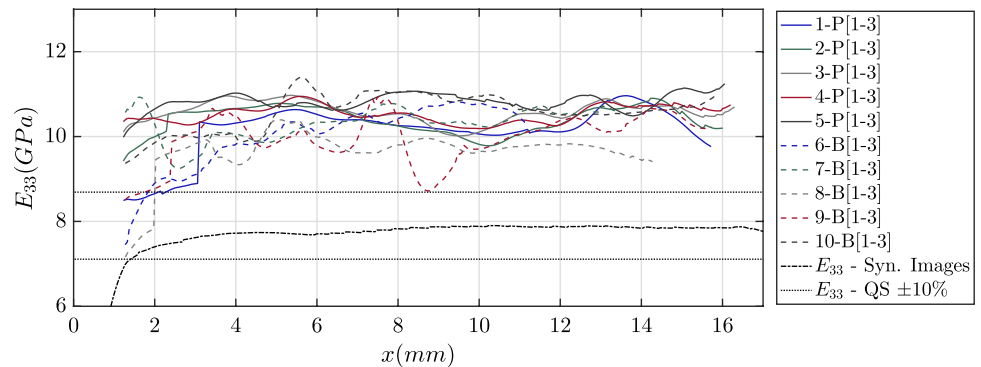
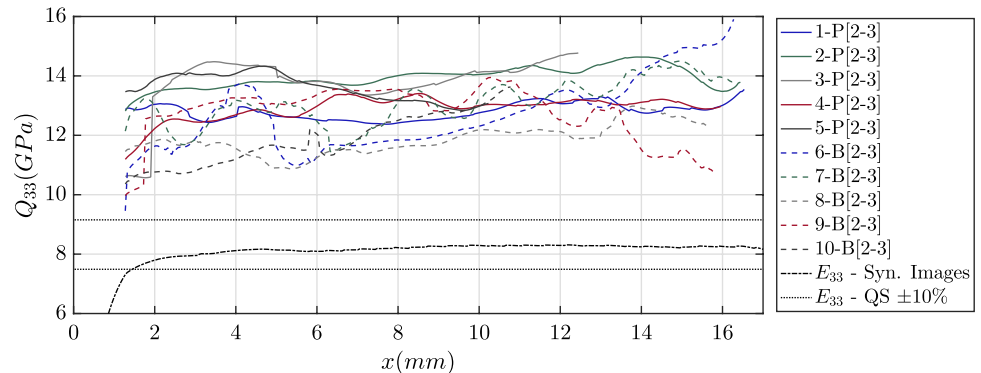


Fig. 25 Interlaminar stiffness, Q_{33} , for all 2–3 plane specimens identified using the stress-gauge approach. Identification from image deformation simulation processed with the same smoothing parameters is provided for comparison. Note that the extrapolated data at the edges of the specimen has been removed



identified using the raw, un-smoothed maps of ϵ_{xx} . A crack becomes clearly visible in the ϵ_{xx} field as a concentrated region of high (artificial) strain, as shown in Fig. 26c. The temporal variations of local stress, computed with the linear stress-gauge equation, was extracted from a $2p \times 4p$ virtual gauge region centred on the identified crack initiation site (as shown in Fig. 26c). The corresponding stress maps, and stress–strain curve over the virtual gauge region are shown in sub figures d, e and f. While not clearly

shown in the strain map in Fig. 26d a second crack had started to form at $x = 9.5 \text{ mm}$, but did not crack the paint. The crack clearly appears three frames later. At the fracture frame shown, the acceleration fields are strongly influenced by these two cracks, explaining the discrepancy between the stress field in Fig. 26d and the reconstructed field using the linear stress-gauge shown in Fig. 26e.

The interlaminar tensile strength is taken as the maximum stress over time within the gauge region. A summary

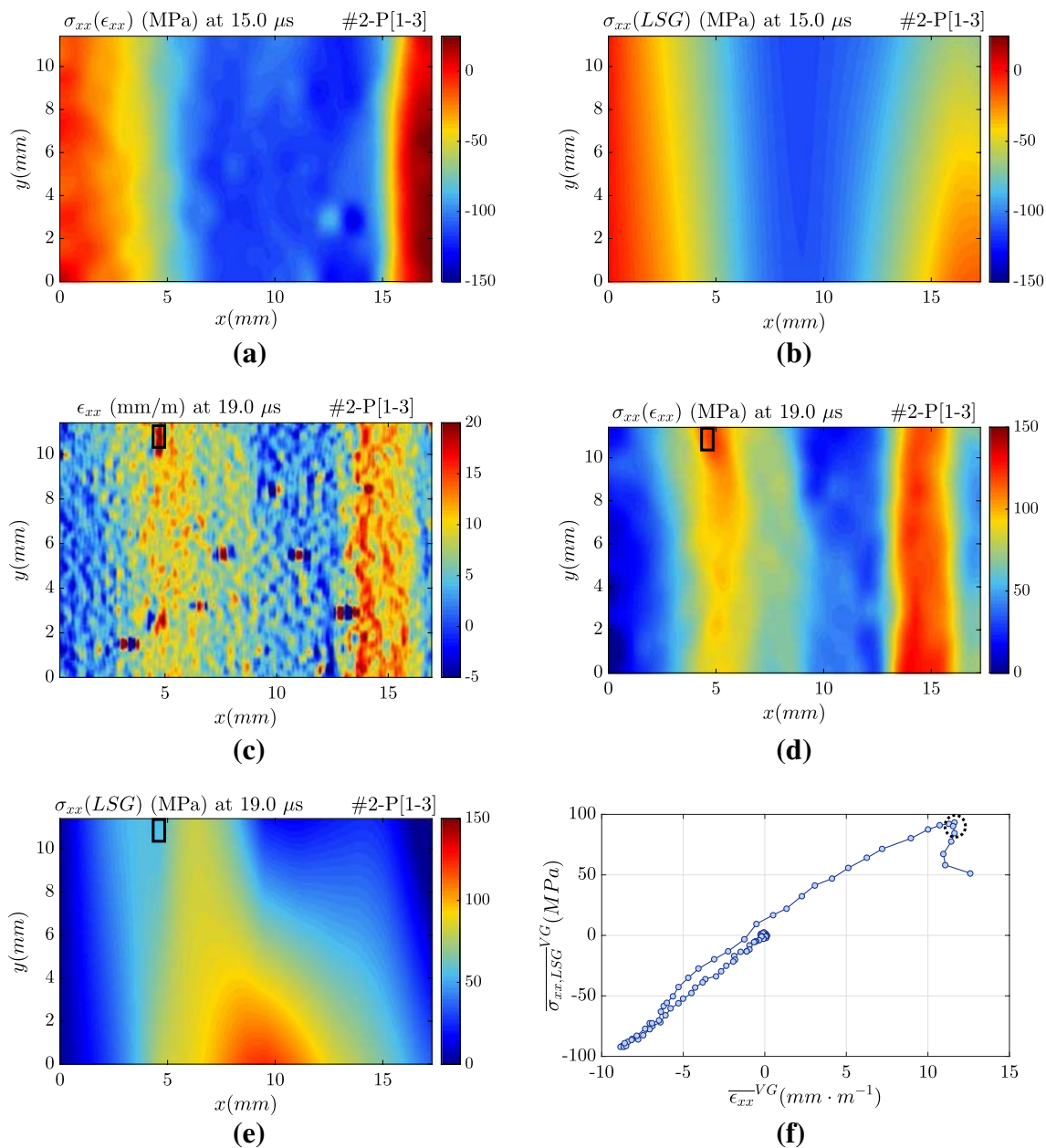


Fig. 26 Strength identification diagnostics for specimen #2-P[1–3]. Diagnostic figures before fracture ($t = 15.0 \mu\text{s}$): **a** stress field (MPa) constructed from ϵ_{xx} using identified E_{33} ($\sigma_{xx}(\epsilon_{xx})$), and **b** stress field (MPa) constructed using the linear stress-gauge equation ($\sigma_{xx}(LSG)$). Diagnostic figures for a time just after the identified fracture time ($t = 19.0 \mu\text{s}$): **c** raw, un-smoothed ϵ_{xx} strain field (mm m^{-1}), **d** $\sigma_{xx}(\epsilon_{xx})$

(MPa), **e** $\sigma_{xx}(LSG)$ (MPa), and **f** stress–strain curve generated using average $\sigma_{xx}(LSG)$ and ϵ_{xx} within the virtual gauge region. In **(c)**, **(d)** and **(e)** the virtual gauge is shown as the black rectangle. In **(f)** the dashed circle indicates the point of fracture and extracted strength using the linear stress-gauge equation

of measured tensile strength using the stress-gauge and linear stress-gauge is provided in Table 9. The tensile average strain rates from within the virtual gauge region ($\overline{\epsilon'_{xx}}^{VG}$) just prior to failure are also provided in Table 9. Some specimens fractured but the initiation of a crack was not clearly identifiable within the kinematic maps. This suggests that either a crack initiated on the back face, or the specimen fractured from impact with the back of the test chamber. For

these cases, the peak stress average is reported in Table 9 for comparison, but is excluded from the calculation of strength statistics (identified by a superscript ‘x’ in Table 9). Note that strain rate maps were computed using a backward differentiation scheme to avoid temporal leakage from non-physical forward strains caused by the crack. Strain rate at fracture was estimated by extrapolating from a regression fitting to $\overline{\epsilon'_{xx}}^{VG}$ over five frames prior to fracture. For specimens

Table 9 Measured high strain rate interlaminar tensile strength for AS4-145/MTM45-1 and peak tensile strain rate ($\overline{\epsilon_{xx}^{VG}}$) within virtual gauge at fracture location

Specimen	1–3 plane			2–3 plane		
	$\overline{\sigma_{xx}^{VG}}$ (SG) (MPa)	$\overline{\sigma_{xx}^{VG}}$ (LSG) (MPa)	$\overline{\epsilon_{xx}^{VG}}$ (s ⁻¹)	$\overline{\sigma_{xx}^{VG}}$ (SG) (MPa)	$\overline{\sigma_{xx}^{VG}}$ (LSG) (MPa)	$\overline{\epsilon_{xx}^{VG}}$ (s ⁻¹)
1-P	94.3	95.0	3100	72.6 ^x	83.9 ^x	4300 ⁺
2-P	83.2	95.7	4600	107.2	109.7	4600
3-P	90.3	94.9	5300	112.0	115.6	4100
4-P	86.1	121.3	3500	74.5	81.5	6300
5-P	107.7	135.6	4900	54.4 ^x	71.5 ^x	6800 ⁺
6-B	130.5	157.6	5300	116.4	122.9	3900
7-B	91.1	106.6	6200	136.9	143.5	4900
8-B	92.1 ^x	98.8 ^x	6000 ⁺	77.4	79.9	4700
9-B	82.4 ^x	83.7 ^x	6000 ⁺	126.7	130.8	5800
10-B	44.6 ^x	50.7 ^x	4000 ⁺	88.1 ^x	134.1 ^x	5300 ⁺
Mean	97.6	115.2	4700	107.3	112.0	4900
SD	16.5	24.2	1100	23.6	24.0	900
COV (%)	16.9	21.0	22.8	22.0	21.4	17.6
Diff. to Q–S (%)	+ 96	+ 135	–	+ 113	+ 122	–

SG stress-gauge approach, LSG linear stress-gauge approach

^xNot included in average strength calculation⁺Peak value over test

where a macro crack was not visible, the peak value of $\overline{\epsilon_{xx}^{VG}}$ is listed (identified by a superscript ‘+’ in Table 9).

The results in Table 9 suggest that strain rate has a significant influence on the interlaminar tensile strength. The interlaminar strength with the linear stress-gauge equation was measured to be 115 MPa (COV = 21%) for the [1–3] specimens and 112 MPa (COV = 21%) for the [2–3] specimens. Combining the two interlaminar planes results in an average strength of 114 MPa, with a COV = 20%. This represents an increase in strength of 125% compared to the quasi-static value of 50.4 MPa (COV = 13%) in [28]. Since the strain rate is generally high when the strain is high, the peak width-averaged strain rate at the plane of fracture offers an ‘effective’ strain rate for the measurements. This is generally on the order of 5000 s⁻¹. It is worth noting that comparisons to quasi-static values need to be interpreted with some caution because quasi-static interlaminar tensile tests are quite sensitive to experimental factors such as misalignment, gripping and volume effect. Typical scatter in the literature for quasi-static strength measurements ranges from 10 to 50% [6, 7, 40, 41]. Therefore, the reported coefficient of variation for quasi-static interlaminar strength of this material is comparatively low. The level of scatter in the current measurements at high strain rates is promising, with scatter comparable to well controlled quasi-static tests.

The time histories of stress averages at the fracture location are shown in Fig. 27 for four specimens (2 from each interlaminar plane). Also shown for comparison is the average stress in the gauge region as reconstructed using the

stress-gauge equation (Eq. 4) and the constitutive model. In the latter, the interlaminar stiffness is taken as the average values identified from stress–strain curves and the special optimised virtual fields routine. However, this stress is not used as a measure of strength due to the uncertainty in determining when strains become non-physical and contamination from grid defects as explained later. Figure 27 shows the initial compressive loading, where all three stress measures agree well. Two of the presented specimens (Fig. 27a, b) show good agreement during the unloading, until a marked drop in average stress (stress-gauge and linear stress-gauge equations) is observed between $t = 15.5$ – 19 μ s. Specimen #2-P[1–3] shows a small offset and low-amplitude oscillation in stress averages at the start of the unloading phase. It is suspected that some through-thickness wave dispersion may have occurred as the wave reflects due to a non-square free edge cut. This effect is observed over a very short duration, and is unlikely to influence strength measurements.

In some cases an offset in stress arises between the stress reconstructed from acceleration and stress computed using the constitutive model (#3-P[1–3]; 2-P[2–3], 3-P[2–3], #6-B[2–3], 7-B[2–3], 8-B[2–3] and 9-B[2–3]), as exemplified in Fig. 27c and d. This is a result of fracture occurring near a grid defect (within one smoothing kernel). Artificial strains caused by the defect biases the strain at the fracture location. The offset increases during the unloading phase when the wave reflects from the free edge. This suggests that the quality of grid defect corrections diminish as the test progresses and the kinematic fields become more complex. This is more problematic for specimens with bonded grids, which

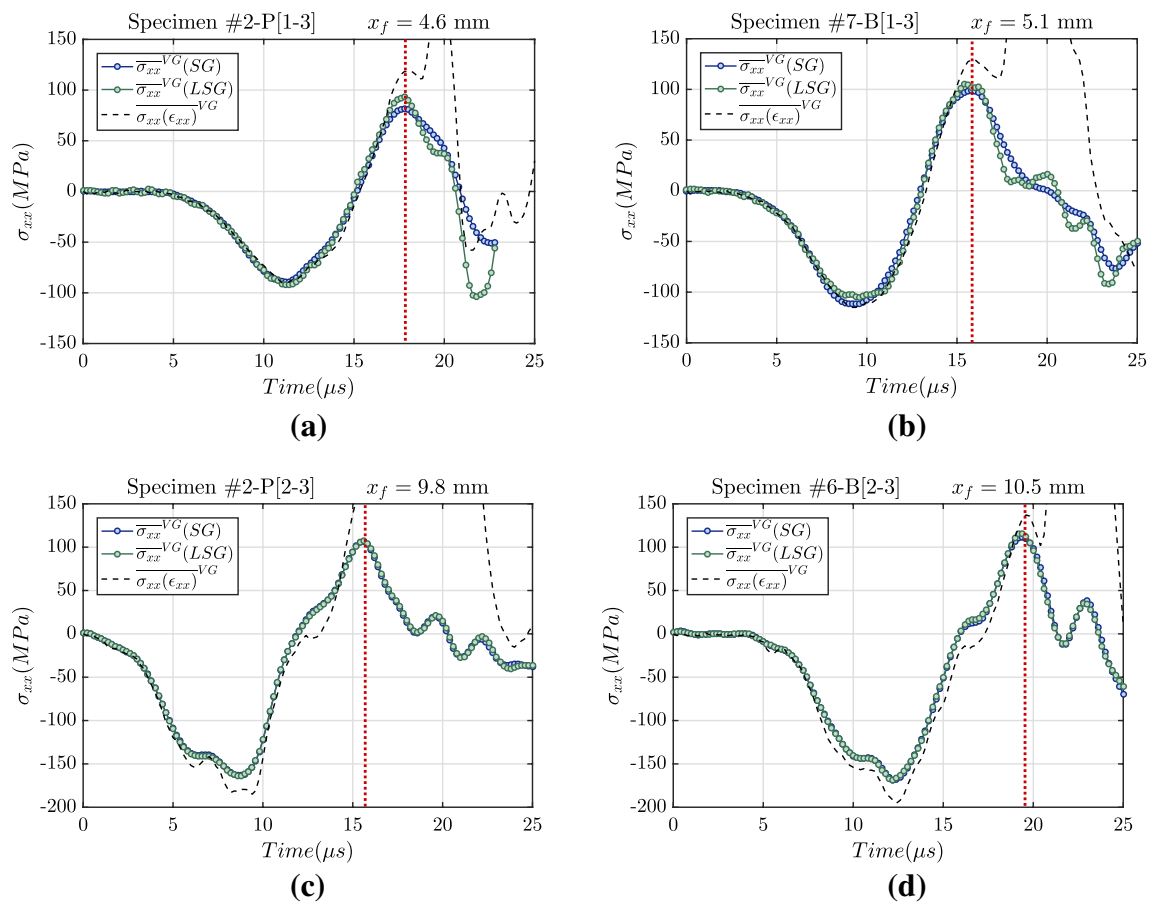


Fig. 27 Comparison of the temporal variations in average stress within the virtual gauge area at the location of fracture as reconstructed using the stress-gauge approach ($\sigma_{xx}^{VG}(SG)$), linear stress-gauge ($\sigma_{xx}^{VG}(LSG)$) and strain ($\sigma_{xx}(\epsilon_{xx})^{VG}$) for **a** specimen #2-P[1–3],

b specimen #7-B[1–3], **c** specimen #2-P[2–3], and **d** specimen #6-B[2–3]. Note that the location of failure is included in the header of each sub figure and the red dashed line indicates the time at which a macro-crack is clearly visible in the un-smoothed strain maps

suffer from high numbers of defects caused by missing grid or bubbles in the resin layer as discussed in Sect. 4.2. The number of defects are significantly reduced when grids are printed, making it the preferred grid deposition technique. This also supports the use of the stress-gauge equations for estimating tensile strength. The stress-gauge equations provide a much more robust estimate of strength since acceleration fields are not smoothed spatially, and because of the spatial averaging procedure used to reconstruct stress.

As previously mentioned, no high strain rate studies have been reported on AS4-145/MTM45-1, thus, no direct comparisons can be made. However, indirect comparisons can be made with studies on the high strain rate through-thickness properties of other composite systems. Comparison of the current measurements to those in other studies shows the robustness of the IBII test for measuring interlaminar strength. For example, a similar strain rate sensitivity was measured by Nakai & Yokoyama (+ 77%) [6, 7]. However, the scatter on measured strength was significantly higher (39% COV), and measurements were made at much lower strain rates (50 s^{-1}).

This amount of scatter is typical of most studies reporting tensile strength measurements using a SHPB at high strain rate such as [4, 8, 42]. Govender et al. [43] used pulse time-shifting to avoid the assumption of quasi-static equilibrium, and produced strength measurements of similar consistency to the current study. However, their approach relied on predicted stresses based on 1-D wave theory and no quasi-static values were provided to quantify the strain rate sensitivity. By using ultra-high speed imaging and full-field measurements, many of the assumptions tied to existing techniques are removed. The measurements reported here exemplify the potential for such techniques to be applied to obtain remarkably consistent strength measurements.

Future Work and Limitations

While further work is still required to refine the IBII test, the preliminary results are promising. This study exemplifies the potential of test methods developed around

full-field imaging to expand the range of strain rates where composite interlaminar properties can be reliably characterised. Based on this initial study, the following points describe some limitations of the test method. This is followed by a summary of possible future investigations to better understand the preliminary results proposed here, improve the test method, and possible extensions for more comprehensive material characterisation.

Assumptions for Processing the Data with the Virtual Fields Method Since only surface deformations are measured, the following assumptions are required here: (1) the specimen is in a state of plane stress, and (2) the mechanical fields need to be uniform through the thickness. Compared to the in-plane specimens in the study by Fletcher et al. [20], the specimen dimensions are less favourable here for ensuring a state of plane stress is achieved. Based on the current specimen manufacturing technique, the smallest thickness that could be achieved, while maintaining good dimensional consistency, was approximately 2 mm. Non-uniformity through the thickness may arise following wave reflection from an angled free edge, which would violate assumption (2). Further work is required to better understand the effect of specimen geometry (e.g., thickness variation and free edge perpendicularity) on the identifications of stiffness and strength. This will be explored as part of future work using three-dimensional finite element simulations and simultaneous imaging on the front and back surfaces of the specimens during experiments. Image deformation simulations will be required to quantify the effect of these geometric ‘defects’ and other experimental factors not accounted for in the current simulation. This will assist in identifying the source of discrepancies between identifications from simulated and experimental images and establish tolerances for specimen geometry and the experimental setup.

Reconstruction of Edge Data Oscillations in the stiffness identified from the stress–strain curves was primarily attributed to reconstruction errors as the pulse moves through the extrapolated data region at the free edge. This occurs since data is inferred for a temporal phenomenon, based solely on spatial information. This effect is amplified for the 0.337 mm printed grids compared to the 0.3 mm grids due to a larger reconstructed data region. A simple way to reduce extrapolation errors is to reduce the grid pitch to enable a lower grid sampling to be used (5 pixels per pitch). It is thought that the effect of reconstructed data on stiffness identifications may also depend on the pulse. A simple way to study this is to use image deformation simulations by systematically varying the pulse characteristics (e.g., rise time, duration and amplitude). The necessity to reconstruct edge data with the grid method will unavoidably introduce errors. Fortunately, this issue will become less problematic as the spatial resolution of ultra-high-speed cameras improves.

Grid Defects The results from this study suggest that grid defects have a significant influence on the identification of stiffness parameters using the isotropic special optimised virtual fields routine. Artificial signal with high signal-to-noise ratio is thought to cause bias in the optimisation of the virtual fields, due to the low signal associated with the underlying material response. The effect is amplified when defects occur near the edges of the specimen and interact with the data extrapolation procedures. Identifications of Q_{33} and Q_{23} have higher scatter and are less stable compared to the image deformation simulations, which do not take into account grid defects. There are several other factors that are also not considered in the simulations and therefore, future work is required to isolate and quantify the effect of defects on the identifications. Image deformation simulations serve as a useful tool for conducting such investigations. Systematic studies on defect position, size and density can be performed by corrupting grey-levels of synthetic images based on defect profiles from real grid images. The corrupted synthetic images may then be processed using the same procedure as experimental images to quantify the influence on identified stiffness parameters.

Considering these limitations, the following summarises possible directions of future work to develop the interlaminar IBII test.

Identification of a Strain Rate Sensitive Constitutive Law Assigning a single strain rate to the measured properties is difficult due to the heterogeneous nature of the fields. This is unavoidable when testing at high strain rates in the wave regime. While this is also problematic for existing techniques, full-field measurements create the potential to exploit the heterogeneity in strain rate fields. One approach is to formulate the unknown stiffness parameters as a function of strain rate in the optimised virtual fields routine. Provided the material is subjected to a range of strain rates, the coefficients of the model could be identified from each set of kinematic fields. The potential of this approach could be explored using a user-defined material sub-routine in ABAQUS combined with the synthetic image deformation methodology. Another approach is to use the virtual fields and the virtual strain rate fields to identify a well-defined value for the effective strain rate. This will be investigated in the future.

Extension to Thinner Specimens The drawback with the current configuration is that thick plates are required, which are expensive and may not be representative of thinner panels. It will be of particular interest in the future to see how small in length we could go, and beyond this, whether a thinner specimen sandwiched between two tabs would be feasible.

Error Quantification for Interlaminar Tensile Strength This work has demonstrated the usefulness of image deformation simulations for selecting optimal smoothing

parameters and estimating the total error on stiffness identifications. The ideal extension of this is an image deformation routine that provides an estimate on the error for interlaminar tensile strength. This requires finite element modelling with cohesive elements to simulate the formation of the crack. Displacements from these simulations can then be used to generate a set of deformed images. This will provide valuable insights into the effect of temporal and spatial smoothing on the current strength identification procedures. This could also extend the work to crack propagation and toughness measurements.

Combined Tension–Shear Testing Work is currently under way to explore the extension of the IBII concept to develop a test for obtaining interlaminar shear, and combined tension–shear properties. This would enable more complete failure envelopes to be populated at high strain rates. This information is not currently available and would be invaluable for improved high strain rate modelling of composite materials.

Conclusions

This work presented the design and experimental validation of the IBII test to measure interlaminar tensile properties at high strain rates. By combining ultra-high-speed imaging with the grid method, full-field maps of displacement, strain and acceleration are obtained. The virtual fields method is used to identify interlaminar stiffness and tensile strength from the measured kinematic fields. The key results from this study are summarised as follows:

- Despite limited spatial resolution, the current study demonstrates that measurement quality of current ultra-high speed cameras is sufficient to identify interlaminar stiffness and tensile strength from the same test.
- Stiffness and strength were found to exhibit a substantial sensitivity to strain rate. An average interlaminar elastic modulus of 10.3 GPa was identified across all 1–3 plane specimens using reconstructed stress–strain curves. For the same specimens, the reduced special optimised virtual fields approach identified an average modulus of 10.7 GPa. Stiffness measurements were made at peak average strain rates on the order 3500 s^{-1} . This represents an increase between 30 and 35% compared to quasi-static values.
- Tensile strength was found to have a higher strain rate sensitivity than stiffness. The average tensile strength over all specimens was measured to be 114 MPa at peak average strain rates on the order 4500 s^{-1} . This corresponds to an increase of approximately 125%, compared to quasi-static values.

- Image deformation simulations are a powerful diagnostic tool for characterising the errors arising from measurement resolution and noise. This enables one to robustly select optimal smoothing parameters. It is also a useful diagnostic tool for studying the effect of post-processing operations (e.g., effect of data extrapolation at the edges) and experimental factors (e.g., grid size, specimen geometry, grid contrast) on the identification of stiffness parameters.
- The use of printed grids has proven to be effective for significantly reducing the number of grid defects. Therefore, this approach is recommended for interlaminar testing where higher magnification is required. However, the current technology is limited to grid periods on the order of 0.3 mm, which will not be enough either for higher spatial resolution cameras or for higher magnification (smaller specimens). Other deposition routes like the one recently proposed by Brodnik et al. [44] could be pursued.

This work shows that new test methods based on full-field measurements are very promising for measuring interlaminar properties at high strain rates. This enables data to be collected at strain rates where current techniques are unreliable. The capability of full-field measurements will only increase as ultra-high-speed camera technology improves. This will create additional opportunities for the development of advanced tests, and a more complete characterisation of composite materials at high strain rates.

Acknowledgements This material is based on research sponsored by the Air Force Research Laboratory, under Agreement Number FA9550-17-1-0133. The authors are grateful to the grant programme manager, Dr. David Garner from EOARD/AFOSR. The U.S. Government is authorized to reproduce and distribute reprints for Governmental purposes notwithstanding any copyright notation thereon. The views and conclusions contained herein are those of the authors and should not be interpreted as necessarily representing the official policies or endorsements, either expressed or implied, of the Air Force Research Laboratory or the U.S. Government. The authors kindly acknowledge Dr. Devlin Hayduke from Material Sciences Corporation, USA for supplying the material tested in this work. The authors would also like to thank Mr. Ian Wilcox from the University of Southampton Print Centre and Dr. Yves Surret for developing the method for printing grids directly onto samples. Dr. Lloyd Fletcher and Prof. Fabrice Pierron acknowledge support from EPSRC through Grant EP/L026910/1. Jared Van Blitterswyk acknowledges Ph.D. funding support from EPSRC.

Compliance with Ethical Standards

Data Provision All data supporting this study are openly available from the University of Southampton repository at: <https://doi.org/10.5258/SOTON/D0561>.

Open Access This article is distributed under the terms of the Creative Commons Attribution 4.0 International License (<http://creativecommons.org/licenses/by/4.0/>), which permits unrestricted use, distribution, and reproduction in any medium, provided you give appropriate

credit to the original author(s) and the source, provide a link to the Creative Commons license, and indicate if changes were made.

References

- Gerlach R, Siviour CR, Petrinic N, Wiegand J (2008) Experimental characterisation and constitutive modelling of RTM-6 resin under impact loading. *Polymer* 49(11):2728–2737
- Van Blitterswyk J, Fletcher L, Pierron F (2017) Characterisation of the interlaminar properties of composites at high strain rates: a review. *Adv Exp Mech* 2(June):3–28
- Lifshitz JM, Leber H (1998) Response of fiber-reinforced polymers to high strain-rate loading in interlaminar tension and combined tension/shear. *Compos Sci Technol* 58(9):987–996
- Hufenbach W, Langkamp A, Gude M, Ebert C, Hornig A, Nitschke S, Böhm H (2013) Characterisation of strain rate dependent material properties of textile reinforced thermoplastics for crash and impact analysis. *Proc Mater Sci* 2:204–211
- Medina JL, Harding J (2000) The effect of strain rate on the through-thickness tensile stiffness and strength properties of fibre-reinforced epoxy composites. *Journal de Physique IV* 10:275–280
- Nakai K, Yokoyama T (2014) Through-thickness tensile strength of carbon/epoxy laminated composites under impact loading. In: 16th international conference on experimental mechanics, Cambridge, UK, 7–11 July
- Nakai K, Yokoyama T (2016) Dynamic stress strain properties of carbon/epoxy laminated composites in through-thickness direction: tension and compression. In: The 9th international symposium on impact engineering, Tainan, Taiwan, 5–9 September
- Naik NK, Yernamma P, Thoram NM, Gadipatri R, Kavala VR (2010) High strain rate tensile behavior of woven fabric E-glass/epoxy composite. *Polymer Test* 29:14–22
- Gilat A, Goldberg RK, Roberts GD (2007) Strain rate sensitivity of epoxy resin in tensile and shear loading. *J Aerosp Eng* 20(2):75–89
- Li Z, Lambros J (1999) Determination of the dynamic response of brittle composites by the use of the split Hopkinson pressure bar. *Compos Sci Technol* 59:1097–1107
- Gama BA, Lopatnikov SL, Gillespie JW (2004) Hopkinson bar experimental technique: a critical review. *Appl Mech Rev* 57(4):223
- Grey GT III (2000) Mechanical testing and evaluation (ASM International). *ASM Handb* 8:462–476
- Yokoyama T, Nakai K (2004) High strain-rate compressive characteristics of laminated composites in the through-thickness direction. In: SEM X international congress & exposition on experimental & applied mechanics, Costa Mesa, California, USA, 7–10 June
- Reu PL, Miller TJ (2008) The application of high-speed digital image correlation. *J Strain Anal Eng Des* 43(May):673–688
- Moulart R, Pierron F, Hallett SR, Wisnom MR (2011) Full-field strain measurement and identification of composites moduli at high strain rate with the virtual fields method. *Exp Mech* 51(4):509–536
- Pierron F, Forquin P (2012) Ultra-high-speed full-field deformation measurements on concrete spalling specimens and stiffness identification with the virtual fields method. *Strain* 48(5):388–405
- Pierron F, Zhu H, Siviour CR (2014) Beyond Hopkinson's bar. *Philos Trans R Soc A* 372:20130195
- Zhu H (2015) A novel methodology for high strain rate testing using full-field measurements and the virtual fields methods. Ph.D. thesis, University of Technology of Troyes, France
- Koohbor B, Kidane A, Lu W-Y, Sutton MA (2016) Investigation of the dynamic stress–strain response of compressible polymeric foam using a non-parametric analysis. *Int J Impact Eng* 91:170–182
- Fletcher L, Van Blitterswyk J, Pierron F (2018) A novel image-based inertial impact (IBII) test for the transverse properties of unidirectional composites at high strain rates. Submitted to the *Journal of Dynamic Behaviour of Materials*
- Yoon SH, Giannakopoulos I, Siviour CR (2015) Application of the Virtual Fields Method to the uniaxial behavior of rubbers at medium strain rates. *Int J Solids Struct* 69–70:553–568
- Yoon SH, Winters M, Siviour CR (2016) High strain-rate tensile characterization of EPDM rubber using non-equilibrium loading and the virtual fields method. *Exp Mech* 56(1):25–35
- Yoon SH, Siviour CR (2017) Application of the virtual fields method to rubbers under medium strain rate deformation using both acceleration and traction force data. *J Dyn Behav Mater* 3(1):12–22
- Lukić BB, Saletti D, Forquin P (2018) On the processing of spalling experiments. Part II: identification of concrete fracture energy in dynamic tension. *J Dyn Behav Mater* 4:56–73
- Forquin P, Lukić BB (2018) On the processing of spalling experiments. Part I: identification of the dynamic tensile strength of concrete. *J Dyn Behav Mater* 4:34–55
- Pierron F, Grédiac M (2012) The virtual fields method: extracting constitutive mechanical parameters from full-field deformation measurements. Springer, Berlin
- Avril S, Grédiac M, Pierron F (2004) Sensitivity of the virtual fields method to noisy data. *Comput Mech* 34:439–451
- Clarkson E, Valliappan T, Tan D (2016) CYTEC (Formerly Advanced Composites Group) MTM45-1/12K AS4 145gsm 32%RW Unidirectional Qualification Statistical Analysis Report FAA Special Project Number SP3505W1Q Test Panel Fabrication Facility. Technical report, National Institute for Aerospace Research (NIAR), Tulsa, OK
- Piro JL, Grédiac M (2004) Producing and transferring low-spatial-frequency grids for measuring displacement fields with moiré and grid methods. *Exp Tech* 28(4):23–26
- Pierron F (2018) The grid method (GM). http://photodyn.org/wp-content/uploads/2016/03/Report_UniversityPrintCentre_2.pdf
- Pierron F (2018) The grid method (GM). http://photodyn.org/wp-content/uploads/2016/03/Report_UniversityPrintCentre.pdf
- Grédiac M, Sur F, Blaysat B (2016) The grid method for in-plane displacement and strain measurement: a review and analysis. *Strain* 52(3):205–243
- Pierron F, Cheriguéne R, Forquin P, Moulart R, Rossi M, Sutton MA (2011) Performances and limitations of 3 ultra high-speed imaging cameras for full-field deformation measurements. *Appl Mech Mater* 70:81–86
- Herráez MA, Burton DR, Lalor MJ, Gdeisat MA (2002) Fast two-dimensional phase-unwrapping algorithm based on sorting by reliability following a noncontinuous path. *Appl Opt* 41(35):7437–7444
- Wang P, Pierron F, Thomsen OT (2013) Identification of material parameters of PVC foams using digital image correlation and the virtual fields method. *Proc Soc Exp Mech* 53(6):1001–1015
- Rossi M, Lava P, Pierron F, Debruyne D, Sasso M (2015) Effect of DIC spatial resolution, noise and interpolation error on identification results with the VFM. *Strain* 51(3):206–222
- Toussaint E, Greiac M, Pierron F (2006) The virtual fields method with piecewise virtual fields. *Int J Mech Sci* 48(3):256–264
- Rossi M, Pierron F (2012) On the use of simulated experiments in designing tests for material characterization from full-field measurements. *Int J Solids Struct* 49:420–435

39. Lukić B, Saletti D, Forquin P (2017) Use of simulated experiments for material characterization of brittle materials subjected to high strain rate dynamic tension. *Philos Trans R Soc A* 375:20160168
40. Hufenbach W, Hornig A, Zhou B, Langkamp A, Gude M (2011) Determination of strain rate dependent through-thickness tensile properties of textile reinforced thermoplastic composites using L-shaped beam specimens. *Compos Sci Technol* 71(8):1110–1116
41. Gerlach R, Siviour CR, Wiegand J, Petrinic N (2013) The strain rate dependent material behavior of S-GFRP extracted from GLARE. *Mech Adv Mater Struct* 20(7):505–514
42. Gerlach R, Siviour CR, Wiegand J, Petrinic N (2012) In-plane and through-thickness properties, failure modes, damage and delamination in 3D woven carbon fibre composites subjected to impact loading. *Compos Sci Technol* 72:397–411
43. Govender RA, Louca LA, Pullen A, Fallah AS, Nurick GN (2012) Determining the through-thickness properties of thick glass fiber reinforced polymers at high strain rates. *J Compos Mater* 46(10):1219–1228
44. Brodnik NR, Hsueh CJ, Johnson MT (2018) A nonconstraining templated powder grid for measurement of strain. *Strain* (in press)



A Novel Image-Based Inertial Impact Test (IBII) for the Transverse Properties of Composites at High Strain Rates

L. Fletcher¹ · J. Van-Blitterswyk¹ · F. Pierron¹

Received: 16 June 2018 / Accepted: 23 January 2019 / Published online: 6 February 2019
© The Author(s) 2019

Abstract

Current methods for testing the high strain rate properties of composites require multiple assumptions that limit achievable strain rates. Therefore, this study presents a new method for testing the transverse properties of composites at high strain rates using ultra-high speed imaging. The image-based inertial impact test developed here uses the reflection of a compressive stress wave to generate tensile stress in the specimen. Throughout the test, full-field displacement measurements are taken. The acceleration and strain fields are then derived from the displacement fields. The acceleration is then used to calculate the average stress in the specimen. This paper describes the optimisation of the experimental configuration using simulations and the experimental validation of the technique. The elastic modulus and tensile strength were identified at strain rates of $\sim 2000 \text{ s}^{-1}$. The results showed an increase of 8% in elastic modulus and an increase of 57% in strength compared to quasi-static values.

Keywords Composites · High strain rate · Full-field measurement · Tensile strength · Virtual fields method

Introduction

In many applications, composite materials undergo high strain rate dynamic loading, for example: crash, impact and blast loading. For these cases, it is essential that the strain rate dependent material properties are accounted for in the design of composite structures. Computational power has exponentially increased in recent decades, allowing for extensive simulations of damage formation in composite structures. However, the usefulness of these simulations is dependent on the quality of the input constitutive models. The computational prediction of damage formed from impact loading in composites requires a full understanding of the complete failure envelope at high strain rates. This

understanding can only be provided by detailed and careful experimental studies that are used to calibrate material models. Unfortunately, practical difficulties related to high strain rate testing have limited the availability of data for the transverse properties of composites at high strain rates.

The strain rate dependent behaviour for the transverse direction in composites has been investigated in several previous studies [1–5]. However, it is difficult to directly compare data from these studies due to differences in the composite materials being tested (e.g. different fibre/matrix materials or different fibre volume fractions). The general trends for unidirectional (UD) composites show that the transverse properties exhibit strain rate sensitive behaviour. Specifically, the tensile elastic modulus shows a small increase or no change at high strain rates whereas the transverse tensile strength significantly increases with strain rate [1–4]. This behaviour is expected as the transverse properties are dominated by that of the matrix material. In general, the matrix is a polymeric material (e.g. epoxy) that shows relatively strong strain rate dependence [6, 7].

Current strength data for unidirectional composites at strain rates greater than 100 s^{-1} has been produced using the Split Hopkinson Pressure Bar (SHPB) technique [1–3, 5]. The SHPB technique is nominally capable of reaching strain rates on the order of 1000 s^{-1} provided that a number of assumptions are met. Two key assumptions of the SHPB

Dr. Fletcher and Prof. Pierron are members of the Society of Experimental Mechanics.

✉ L. Fletcher
l.c.fletcher@soton.ac.uk
J. Van-Blitterswyk
j.van-blitterswyk@soton.ac.uk
F. Pierron
f.pierron@soton.ac.uk

¹ Engineering Sciences, University of Southampton,
Southampton SO17 1BJ, Hampshire, UK

analysis are: 1) quasi-static stress equilibrium and 2) one dimensional wave propagation (i.e. no dispersion effects). Violation of either of these assumptions leads to unreliable test results. The quasi-static equilibrium assumption is equivalent to neglecting inertial acceleration effects in the specimen. Once the quasi-static stress equilibrium condition is satisfied, one dimensional wave theory can be used to infer the material properties from strain gauge measurements on the bars of the SHPB apparatus. In order for this assumption to be satisfied, there are strict constraints on specimen size and geometry. Small specimens (relative to the overall bar length) are generally used such that the stress wave can reverberate multiple times and damp out quickly. Furthermore, the reverberation of the stress wave occurs during the initial (i.e. elastic) part of the test. Thus, it is generally accepted that in most cases it is not possible to accurately measure the elastic modulus using standard SHPB techniques and only an apparent modulus can be obtained [8, 9]. Recently, it has been shown that for specific test configurations at moderate strain rates (400 s^{-1}) inertial effects are probably small enough that the elastic modulus can be obtained with the standard SHPB analysis [10]. However, this finding is not generalisable to higher strain rates with lower wave speed materials such as the transverse composite samples considered here.

Achieving quasi-static stress equilibrium is particularly challenging when testing composite materials in transverse tension. The main reason for this is the brittle failure mode which is characterised by small strains to failure [1, 2]. As the specimen fails at small strains, it can fail before inertial effects have damped out and the quasi-static stress equilibrium assumption will not be reasonably satisfied. The small strain to failure also means that the signal to noise ratio is low for measurement of the specimen strain using the gauges in the bars. The constraints of the quasi-static stress equilibrium assumption effectively limit the strain rate that can be obtained using the standard SHPB technique to under 1000 s^{-1} for the transverse tensile properties of a unidirectional composite [1, 2]. The limitations of the standard SHPB technique for transverse properties of composites are shown clearly in a study by Melin and Asp [1]. In this study, Melin and Asp used a moiré fringe technique to analyse the strain field and compared this to the strain predicted by the standard SHPB technique. For this case, it was found that the SHPB theory underestimated the true strain field in the tested specimens. Additionally, in the study by Melin and Asp, it was only possible to measure the transverse properties up to an effective strain rate of 800 s^{-1} . This limitation is due to the quasi-static stress equilibrium assumption of the SHPB technique as mentioned previously.

The second key requirement of the SHPB testing is the need for the stress wave imparted on the specimen to be one-dimensional. Violation of this assumption is termed dispersion. The effects of dispersion generally manifest as oscillatory

behaviour on the stress-strain response. This oscillatory behaviour is shown by the transverse stress-strain curve of figure 8 in ref. [2] at a nominal strain rate of 400 s^{-1} . Pulse shaping techniques may be used to reduce the effects of dispersion by removing the high frequency content from the input pulse. However, this reduces the achievable strain rate. In any case, the effects of dispersion are never fully removed by pulse shaping [8]. To address the significant limitations for testing the transverse properties of composites at high strain rates, new test methods are required that do not rely on the assumptions of the standard SHPB technique.

Recent studies have shown that image-based full-field measurement techniques can provide a viable alternative for developing new high strain rate methods [11–13]. Full-field measurements provide a rich data set and allow for a multitude of measurement points across a test specimen. Techniques such as digital image correlation (DIC) and the grid method have been used extensively for quasi-static testing. When these techniques are combined with an inverse identification technique (e.g. the virtual fields method (VFM)) they can provide data to calibrate entire material models from a single test [14, 15]. These techniques are now being applied to high strain rate testing due to advances in high speed imaging technology [11–13, 16, 17]. The recent advent of ultra-high speed cameras (as defined in ref. [18]) with framing rates greater than 1 MHz has allowed for the imaging of stress wave propagation in materials. The inertial acceleration fields during stress wave propagation can be used to reconstruct stress information. In this way, the acceleration fields are used as an embedded load-cell in the specimen. This concept has been successfully demonstrated experimentally in several studies [11–13, 19]. The use of ultra-high speed full-field measurements combined with the use of acceleration fields as an embedded load cell provides a promising new avenue for designing high strain rate tests for composite materials. Therefore, the overall aim of this study is to design a high strain rate test method for the transverse properties of composites using ultra-high speed full-field measurements. The following section of this paper describes the theory and concept design of the image-based inertial impact (IBII) test configuration. The next section of the paper then describes the finite element modelling used to validate the proposed test and optimise the test configuration. Following the finite element test design, the experimental campaign is discussed. The final sections of this paper outline the limitations and future applications of the new method along with a summary of the main findings.

Concept Design and Theory

Imaged-Based Inertial Impact (IBII) Test Concept

The aim of this work is to design a dynamic test to determine both the transverse stiffness and strength of a unidirectional

composite. For this purpose, an image-based inertial impact (IBII) test has been designed similar to the test presented in Pierron and Forquin [13] for measuring the tensile properties of concrete. The basic principle behind this test is briefly described here. Consider the unidirectional composite specimen shown in Fig. 1, impacted by the force $F(t)$. A short compressive pulse is induced in the specimen. Then, the compressive pulse travels to the free edge of the specimen and reflects becoming tensile. Finally, the tensile wave travels through the specimen causing failure. Throughout the test, full-field displacement measurements are taken over the specimens' surface. From the full-field displacement data, the strain and acceleration fields can be calculated. The stiffness and strength of the material can then be determined from the kinematic fields using the methods described in the following sections. The indirect loading configuration is advantageous for testing brittle materials in tension as it takes away the need to grip the specimen and the associated difficulties with misalignment of the grips. Additionally, this configuration only requires that one edge of the sample is loaded with the other edges free with known boundary conditions. This simplifies the derivation of the relationships between the kinematic fields and the desired material properties as described in following sections.

In order to use the IBII test configuration, the compressive strength of the material must be higher than its tensile strength to avoid causing compressive failure with the initial pulse. The IBII test is well suited for testing the transverse properties of unidirectional composites as the compressive strength is much higher than the tensile strength. A unidirectional carbon fibre has a transverse compressive strength typically on the order of 200–300 MPa whereas the transverse tensile strength is in the range 40–80 MPa, depending on the matrix material. At high strain rates, both of these strength values are typically thought to increase [1, 2, 4, 5]. Based on these properties, it is possible to select the compressive input

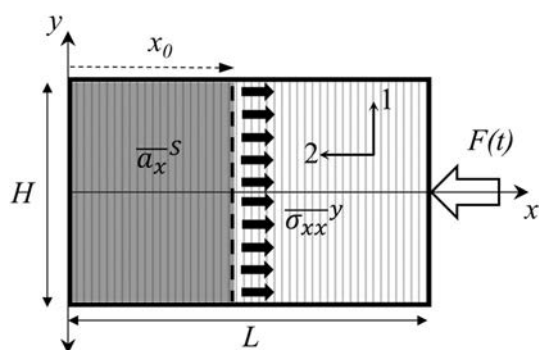


Fig. 1 Schematic of a unidirectional composite specimen impacted transverse to the fibre direction. Note that this is a free-body diagram of the test at an arbitrary time point t . The variables listed here illustrate those used in eq. 16 with the greyed out area showing the area over which the acceleration surface average is calculated

pulse to cause tensile failure while remaining well below the compressive strength. It should be noted here that the transverse tensile strength at higher strain rates is thought to increase on the order of 30% [1, 2]. This projected increase in the tensile strength needs to be accounted for in the test design to ensure that the reflected tensile pulse causes failure in the specimen.

The Virtual Fields Method

Consider the schematic of a dynamically impacted composite specimen shown in Fig. 1. The underlying problem to address here is to relate the measured kinematic fields to the material constitutive parameters. In order to do this, the principle of virtual work is used. This approach is referred to as the virtual fields method (VFM). The interested reader is referred to ref. [15] for a detailed discussion of the VFM formulation. It is noted here that the principle of virtual work is simply the integral form of the stress equilibrium equation. The principal of virtual work, in absence of volume forces, is given by:

$$-\int_V \boldsymbol{\sigma} : \boldsymbol{\epsilon}^* dV + \int_{\delta V} \mathbf{T} \cdot \mathbf{u}^* dS = \int_V \rho \mathbf{a} \cdot \mathbf{u}^* dV \quad (1)$$

where V is the volume of the specimen and δV is the boundary surface of the volume. $\mathbf{T}$ is the Cauchy stress vector acting on the boundary surface δV , and $\boldsymbol{\sigma}$ is the Cauchy stress tensor. The virtual displacement field is given by $\mathbf{u}^*$ and the virtual strain tensor (derived from $\mathbf{u}^*$) is given by $\boldsymbol{\epsilon}^*$. The acceleration field is denoted as $\mathbf{a}$ and ρ is the material density. The matrix and vector dot products are denoted as $:$ and $\cdot$ respectively. The first term of eq. 1 is referred to as the internal virtual work, the second term is the external virtual work and the third term is the acceleration virtual work. It should be noted that all mechanical fields in this equation are functions of time and space. In order to simplify the equations here and in the following section, the time and space variables as function notation have been omitted.

In the VFM, the principle of virtual work is used to generate integral equilibrium equations that relate stress components to external forces and accelerations. The only conditions are that the virtual displacement vectors are continuous and piecewise differentiable. It is worth noting that the standard textbook statement that $\mathbf{u}^*$ has to be kinematically admissible is not a mathematical requirement for the equation but a need for unknown reaction forces at imposed displacement boundaries not to appear in the equation when it is used to solve the direct problem (as in the finite element method). Here, $\mathbf{u}^*$ has to be ‘VFM-admissible’ so that no unwanted unknowns appear in the equation (like a distribution of pressure in the grips of a test machine when only a

resultant from a load-cell is measured). It is noted here that the selected virtual fields must also be C^0 continuous.

Additionally, a number of simplifying assumptions are made: (1) the density, thickness and stiffness of the specimen do not vary in space; (2) the kinematic fields are uniform through the thickness; and (3) the specimen is in a state of plane stress. As can be seen on Fig. 1, the stress field will be predominantly longitudinal, so the VFM analysis can concentrate on the σ_{xx} term. For this purpose, generic virtual fields that cancel out the contributions of σ_{yy} and σ_{xy} are defined as follows:

$$\begin{cases} u_x^* = f(x) \\ u_y^* = 0 \end{cases} \begin{cases} \epsilon_{xx}^* = f'(x) \\ \epsilon_{yy}^* = 0 \\ \epsilon_{xy}^* = 0 \end{cases} \quad (2)$$

where function $f(x)$ is continuous and piecewise differentiable. Therefore, the VFM equation becomes:

$$-\int_S \sigma_{xx} \epsilon_{xx}^* dS + \int_l T_x f(x) dl = \rho \int_S a_x f(x) dS \quad (3)$$

where S is the surface of the specimen and l is the boundary of the specimen.

Constitutive Model and Assumptions

At this stage, it is necessary to substitute the stress in eq. 3 using an appropriate constitutive model. The material being analysed here is a unidirectional carbon fibre composite. For this material, a linear elastic orthotropic constitutive law is appropriate. Only the shear response can exhibit some non-linearity but this test will not produce significant shear strain in the material axes, and eq. 3 has been constructed to remove the shear contribution.

The stress–strain relationship for the σ_{xx} component can be written as:

$$\sigma_{xx} = Q_{22} \epsilon_{xx} + Q_{21} \epsilon_{yy} \quad (4)$$

where Q_{22} is the plane-stress stiffness transverse to the fibre direction and Q_{21} is the plane-stress stiffness associated with Poisson's effect. They are provided by:

$$Q_{22} = \frac{E_{22}}{1 - \nu_{12}\nu_{21}}; Q_{21} = \nu_{12}Q_{22} \quad (5)$$

where E_{22} is the Young's modulus in the transverse direction and ν_{12} and ν_{21} are the major and minor Poisson's ratios respectively. Additionally, the Young's modulus in the fibre direction E_{11} is related to the transverse modulus E_{22} as follows: $\nu_{21}E_{11} = \nu_{12}E_{22}$. Typical values of these parameters for a unidirectional carbon fibre composite are: $E_{11} = 135$ GPa, $E_{22} = 10$ GPa and $\nu_{12} = 0.3$, so $\nu_{21} = 0.022$. Factorizing Q_{22} in eq. 4, the following equation is obtained:

$$\sigma_{xx} = Q_{22}(\epsilon_{xx} + \nu_{12}\epsilon_{yy}) \quad (6)$$

Most of the ϵ_{yy} in the test will arise from Poisson's effect because of the σ_{xx} stress. Separating ϵ_{yy} into the part arising from Poisson's effect denoted $\epsilon_{yy}^{(P)}$ and the part caused by some possible presence of σ_{yy} stress, denoted $\epsilon_{yy}^{(\sigma_{yy})}$, eq. 4 becomes:

$$\sigma_{xx} = Q_{22}(\epsilon_{xx} + \nu_{12}(\epsilon_{yy}^{(P)} + \epsilon_{yy}^{(\sigma_{yy})})) \quad (7)$$

Now, we can replace $\epsilon_{yy}^{(P)}$ by $-\nu_{21}\epsilon_{xx}$ and factorize ϵ_{xx} . The term $1 - \nu_{12}\nu_{21}$ simplifies with the denominator of Q_{22} in eq. 5 and 7 becomes:

$$\sigma_{xx} = E_{22}\epsilon_{xx} + Q_{21}\epsilon_{yy}^{(\sigma_{yy})} \quad (8)$$

Because of the high E_{11} stiffness, $\epsilon_{yy}^{(\sigma_{yy})}$ will be very small and the second term in eq. 8 is negligible with respect to the first one. This will be checked in the finite element validation step later on this article, "Numerical Design and Validation" section. Therefore, in the rest of the article, the constitutive model that will be used is simply $\sigma_{xx} = E_{22}\epsilon_{xx}$ and the VFM equation becomes (assuming that the material is homogeneous at the scale of the experiment):

$$-E_{22} \int_S \epsilon_{xx} \epsilon_{xx}^* dS + \int_l T_x f(x) dl = \rho \int_S a_x f(x) dS \quad (9)$$

Virtual Fields For Stiffness Identification

In the present work, several virtual fields have been used for different purposes. In the following section, a set of virtual fields are described that have been used for identifying the transverse elastic modulus E_{22} .

Virtual Field 1

Consider the simplest case, where $f(x) = 1$ and $f'(x) = 0$. This virtual field describes a virtual rigid translation along the x axis. Substituting this virtual field into eq. 9 removes the contribution of all virtual strain components. This leaves the external virtual work and acceleration virtual work terms, as follows:

$$\int_l T_x dl = \rho \int_S a_x dS \quad (10)$$

Equation 10 can be evaluated at any axial section along the specimen length, denoted x_0 . This axial section is shown as a dashed line in Fig. 1. As the traction vector $\mathbf{T} = \boldsymbol{\sigma} \times \mathbf{n}$, T_x is given by the stress component σ_{xx} . The resulting equation now becomes:

$$\int_{-H/2}^{H/2} \sigma_{xx} dy = \rho \int_{-H/2}^{H/2} \int_0^{x_0} a_x dx dy \quad (11)$$

In practice, the integral terms in eq. 11 are calculated using full-field displacement measurements obtained using the grid method or DIC. If these measurements have sufficient spatial resolution, the integrals can be approximated as discrete sums over a portion of the field of view. For example, the acceleration virtual work term can be approximated as follows:

$$\int_{-H/2}^{H/2} \int_0^{x_0} a_x dx dy \approx \sum_{i=1}^n a_x^{(i)} s^{(i)} \quad (12)$$

$$\frac{S}{n} \sum_{i=1}^n a_x^{(i)} = x_0 H \overline{a_x^S} \quad (13)$$

where n is the total number of measurement points on the specimen surface bounded by the area S [$0 \leq x \leq x_0$, $-H/2 \leq y \leq H/2$], $a_x^{(i)}$ is the acceleration measured at point i and $s^{(i)}$ is the small surface area associated with data point i . The small surface area $s^{(i)}$ is the same for all data points as the measurement grid is regular. $\overline{a_x^S}$ is the surface average of the acceleration over the field of view to the point x_0 . The overline notation is used in subsequent equations to denote spatial averaging and a superscript S denotes the spatial average over a portion of the field of view. The integral in the external virtual work term can be expressed in a similar manner:

$$\int_{-H/2}^{H/2} \sigma_{xx} dy \approx \sum_{j=1}^m \sigma_{xx}^{(j)} s_l^{(j)} \quad (14)$$

$$\frac{H}{m} \sum_{j=1}^m \sigma_{xx}^{(j)} = H \overline{\sigma_{xx}^y} \quad (15)$$

where m is the total number of points along the axial slice at $[x = x_0, -H/2 \leq y \leq H/2]$, $\sigma_{xx}^{(j)}$ is the axial stress at point j and $s_l^{(j)}$ is the small section of the line associated with point j . The superscript y coupled with the overline notation indicates the average over the line at x_0 . The expressions in eqs. 13, 15 can be substituted into eq. 11 to give:

$$\overline{\sigma_{xx}^y} = \rho x_0 \overline{a_x^S} \quad (16)$$

Equation 16 is hereafter referred to as the stress-gauge equation. The variables in eq. 16 are illustrated on the free body diagram in Fig. 1. The stress-gauge equation relates the mean axial stress to the mean of the acceleration field with the density acting as a calibration factor. Introducing the constitutive equation as defined previously, this equation becomes:

$$E_{22} \overline{\epsilon_{xx}^y} = \rho x_0 \overline{a_x^S} \quad (17)$$

By plotting $\rho x_0 \overline{a_x^S}$ as a function of $\overline{\epsilon_{xx}^y}$, a linear response should be obtained, the slope of which provides E_{22} .

Virtual Field 2

Additional virtual fields can also be used for stiffness identification. For instance, $f(x) = x - L$ and $f'(x) = 1$. This function has been selected such that $f(L) = 0$. However, since $f(x)$ does not depend on y , any virtual field as defined in eq. 2 would only involve the resultant stress at the impact edge, which can be obtained from eq. 16 when $x_0 = L$, so for instance, $f(x) = x$ would lead to the linear combination of eqs. 16 and 18 below:

$$\overline{\epsilon_{xx}^S} E_{22} = -\rho \overline{a_x(x-L)^S} \quad (18)$$

for $\overline{\epsilon_{xx}^S} \neq 0$, where $\overline{a_x(x-L)^S}$ is the average of the axial acceleration over the field of view weighted by the function $x - L$. Similarly, $\overline{\epsilon_{xx}^S}$ is the average of the axial strain over the field of view. Using eq. 18 the transverse modulus can be determined at any frame of the dynamic test. The formula in eq. 18 is hereafter referred to as the manual VFM as it involves the manual definition of the function $f(x)$. Obviously, this raises the question regarding the choice of virtual field and how this affects the identified stiffness. In particular, using a virtual field that does not vary in time may lead to both terms in the fraction of eq. 18 to be zero and thus, to times when E_{22} is unidentifiable. Moreover, for an ideal data set, with sufficiently rich acceleration and strain fields, the identified stiffness is insensitive to the choice of virtual field. However, in practice, the experimental acceleration and strain fields contain noise. Thus, the appropriate selection of virtual fields is dependent on minimising the effects of noise on the identification procedure. This problem has been previously investigated and methods have been developed to automatically select virtual fields that minimise the effects of noise [15, 20, 21]. This type of virtual field formulation is termed ‘special optimised’. In the following section, the formalism of the special optimised virtual fields is developed for the case of identifying the transverse modulus of a unidirectional composite.

Virtual Field 3: Special Optimised

This follows the procedure presented in ref. [20]. Since the constitutive model is simplified here compared to the full orthotropic case in ref. [20], the main equations will be briefly recalled. Here, a set of piecewise functions have been used to expand the virtual fields.

To comply with the formulation in eq. 2, the virtual mesh is formulated with only u_x^* degree of freedom and nodes with the same x coordinate must have the same u^* value. If the number of nodes along the x and y axes are denoted by ‘ n ’ and ‘ m ’,

respectively, then this gives a total of mn conditions. The contribution of the impact load is also cancelled out by setting all $u^* = 0$ for $x = L$ (impact edge), though this is not compulsory (see previous remark for virtual field 2). This results in an additional m conditions. A typical virtual mesh is presented in Fig. 2.

The speciality condition is given by $\int_S \epsilon_{xx} \epsilon_{xx}^* dS = 1$. This forces the virtual fields to ‘follow’ the deformation waves to produce non-zero internal and acceleration virtual works to ensure identification. This provides another constraining equation for the virtual fields. With this condition, eq. 9 becomes:

$$E_{22}^{app} = -\rho \int_S a_x u^*(x) dS \quad (19)$$

Here, E_{22}^{app} is introduced to reflect the fact that the strain data is corrupted by noise. For exact data, $E_{22}^{app} = E_{22}$. If strain noise is present and is represented by a Gaussian distribution Π_x of standard deviation γ , then eq. 9 becomes:

$$E_{22} = \gamma E_{22} \int_S \Pi_x \epsilon_{xx}^* dS - \int_S \rho a_x u_x^* dS \quad (20)$$

It is further assumed that the noise amplitude is much smaller than the magnitude of the measured strains. Therefore, one can substitute the actual stiffness parameters with their approximate counterparts. This is to keep the process analytical. Therefore, eq. 20 can be rewritten as follows.

$$E_{22} \simeq \gamma E_{22}^{app} \int_S \Pi_x \epsilon_{xx}^{*(1)} dS + E_{22}^{app} \quad (21)$$

Following through the same noise minimisation procedure as detailed in the work by Avril et al. [20] and approximating the integral quantities by discrete sums, the variance associated with E_{22} is approximated by eq. 22.

$$\begin{aligned} V(E_{22}) &\simeq \gamma^2 \left(\frac{S}{N} \right)^2 \left[(E_{22}^{app})^2 \sum_{i=1}^N (\epsilon_{xx}^*(M_i))^2 \right] \\ &= \left(\frac{S}{N} \right)^2 E_{22}^{app} \cdot \mathbf{G} \cdot \mathbf{Q}_{22}^{app} \end{aligned} \quad (22)$$

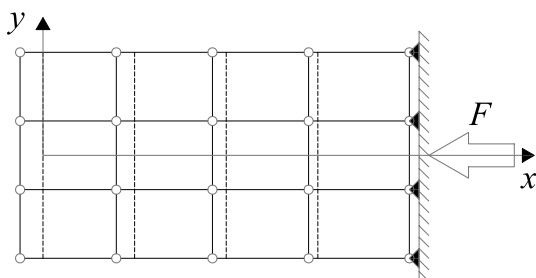


Fig. 2 Virtual mesh for the identification of E_{22} . Undeformed mesh shown in dashed lines. Only longitudinal virtual DOFs are allowed

where S is the surface area of the sample which has N discrete measurement points. M_i denotes the i th measurement point and $\epsilon_{xx}^*(M_i)$ is the virtual strain evaluated at this point. Since only one parameter is identified, the matrix $\mathbf{G}$ in ref. [20] is reduced to eq. 23.

$$\mathbf{G} = \sum_{i=1}^N (\epsilon_{xx}^{*(i)}(M_i))^2 \quad (23)$$

Solving the minimisation problem using the Lagrangian multipliers method [20] produces a single set of optimal virtual field coefficients, which are used to obtain E_{22} .

$$E_{22} = -\rho \frac{LH}{N} \left[\sum_{i=1}^N a_x \mathbf{Y}^* \cdot \tilde{\mathbf{u}}_x^* \right] \quad (24)$$

In eq. 24 the vector denoted $\mathbf{Y}^{(i)*}$ contains the unknown coefficients of the virtual fields, or $\tilde{\mathbf{u}}_x^*$.

Virtual Fields for Strength Identification

It is also possible to use virtual fields to derive simple stress relationships from acceleration that can be used to approximate the stress state in a test. This can then be used to infer the failure stress in simple cases. Consider the stress-gauge formula given by eq. 16, this represents the mean axial stress in the specimen at a particular axial location. In this case, the stress is fully resolved in the x direction (i.e. it provides the mean stress at each cross section). However, the distribution of the stress in the y direction is only represented by an average value. Therefore, this will only be a reasonable approximation of the actual stress if the stress is uniform through the width. Only then will this equation be useful to identify strength.

Additional rigid body virtual fields can be used to derive measures of mean stress for strength measurement. This includes: (1) a virtual rigid body translation in the y direction giving the mean shear stress $\overline{\sigma_{xy}}^y$ and (2) a virtual rigid body rotation giving the average of the first moment of the axial stress $\overline{\sigma_{xx}y}^y$. These are derived in the following section. Consider the following virtual field that describes a rigid body translation in the y direction:

$$\begin{cases} u_x^* = 0 \\ u_y^* = 1 \end{cases} \begin{cases} \epsilon_{xx}^* = 0 \\ \epsilon_{yy}^* = 0 \\ \epsilon_{xy}^* = 0 \end{cases} \quad (25)$$

Applying the same assumptions and reasoning as for the first rigid body virtual field, the principle of virtual work becomes:

$$\int_{-H/2}^{H/2} \sigma_{xy} dy = \rho \int_{-H/2}^{H/2} \int_0^{x_0} a_y dx dy \quad (26)$$

$$\overline{\sigma_{xy}}^y = \rho x_0 \overline{a_y}^S \quad (27)$$

where $\overline{\sigma_{xy}}^y$ is the average shear stress in the cross section at location x_0 and $\overline{a_y}^S$ is the surface average of the acceleration from the free edge to x_0 . Next, consider the virtual field that describes a rigid body rotation about the origin:

$$\begin{cases} u_x^* = y \\ u_y^* = -x \end{cases} \begin{cases} \varepsilon_{xx}^* = 0 \\ \varepsilon_{yy}^* = 0 \\ \varepsilon_{xy}^* = 0 \end{cases} \quad (28)$$

With the same assumptions as before, the principle of virtual work becomes:

$$\begin{aligned} \int_{-H/2}^{H/2} \sigma_{xx} y dy - x_0 \int_{-H/2}^{H/2} \sigma_{xy} dy \\ = \rho \int_{-H/2}^{H/2} \int_0^{x_0} (a_x y - a_y x) dx dy \end{aligned} \quad (29)$$

$$\overline{\sigma_{xx}}^y - x_0 \overline{\sigma_{xy}}^y = \rho x_0 (\overline{a_x}^S - \overline{a_y}^S x) \quad (30)$$

Substituting eq. 27 into eq. 30 gives the following equation for the average of the first moment of axial stress in terms of weighted averages of the acceleration field:

$$\overline{\sigma_{xx}}^y = \rho x_0 (x_0 \overline{a_y}^S + \overline{a_x}^S - \overline{a_y}^S x) \quad (31)$$

where $\overline{\sigma_{xx}}^y$ is the average of the first moment of the axial stress over the cross section at x_0 . The acceleration terms in eq. 31 are spatial averages over the specimen surface up to the transverse slice x_0 . Equations 16, 27 and 31 can now be used to provide an x -resolved linear approximation of the σ_{xx} stress along y . Denoting $\sigma_{xx}(LSG)$ the linear approximation of the σ_{xx} stress along y (LSG for ‘Linear Stress-Gauge’), $\sigma_{xx}(LSG)$ can be expressed as follows:

$$\sigma_{xx}(LSG) = \sigma_{xx}^a + \frac{y}{H} \sigma_{xx}^b \quad (32)$$

where σ_{xx}^a and σ_{xx}^b are the coefficients of the linear function. Consider the first rigid body virtual field describing a translation in the x direction. Equation 32 can be substituted into eq. 11 to obtain:

$$\int_{-H/2}^{H/2} \left(\sigma_{xx}^a + \frac{y}{H} \sigma_{xx}^b \right) dy = \rho \int_{-H/2}^{H/2} \int_0^{x_0} a_x dx dy \quad (33)$$

Evaluating the integrals and simplifying with the overline notation yields the coefficient σ_{xx}^a :

$$\sigma_{xx}^a = \rho x_0 \overline{a_x}^S \quad (34)$$

which is, not surprisingly, eq. 16. Next, consider the virtual field describing a rigid body rotation. Substituting eqs. 32 into 29 gives:

$$\begin{aligned} \int_{-H/2}^{H/2} \left(\sigma_{xx}^a + \frac{y}{H} \sigma_{xx}^b \right) y dy - x_0 \int_{-H/2}^{H/2} \sigma_{xy} dy \\ = \rho \int_{-H/2}^{H/2} \int_0^{x_0} (a_x y - a_y x) dx dy \end{aligned} \quad (35)$$

Simplifying the integrals in eq. 35 yields:

$$\sigma_{xx}^b \frac{H}{12} - x_0 \overline{\sigma_{xy}}^y = \rho x_0 (\overline{a_x}^S - \overline{a_y}^S x) \quad (36)$$

The term $\overline{\sigma_{xy}}^y$ can be expressed in terms of acceleration averages using eq. 27. If the relation in eq. 27 is substituted into eq. 36, the coefficient σ_{xx}^b can be expressed as:

$$\sigma_{xx}^b = \frac{12 \rho x_0}{H} (\overline{a_x}^S - \overline{a_y}^S x + x_0 \overline{a_y}^S) \quad (37)$$

Finally, a linear approximation of the stress distribution is obtained by combining the results for each of the linear coefficients with eq. 32:

$$\sigma_{xx}(LSG) = \rho x_0 \overline{a_x}^S + \frac{12 \rho x_0 y}{H^2} (\overline{a_x}^S - \overline{a_y}^S x + x_0 \overline{a_y}^S) \quad (38)$$

The eq. 38 will be hereafter referred to as the linear stress-gauge formula. This equation is useful as it allows for an improved approximation of the stress field in the y direction compared to the standard stress-gauge formula in eq. 16. A valid question that arises in the application of the linear stress-gauge formula is: ‘‘How accurate is the linear approximation compared to the true stress field?’’. This can be addressed by using the constitutive law identified with the previously described methods. The stress field in the specimen can be reconstructed from the strain using the identified constitutive law. This can then be compared to the stress predicted by the linear stress-gauge approach to assess the accuracy of the approximation. Obviously, once the assumptions of the constitutive law are violated then this comparison no longer holds. For instance, if the identified constitutive law is linear elastic then once the material exhibits non-linear behaviour (plasticity, damage, failure etc.), these two measures of stress will diverge. In fact, this divergence can be useful in identifying the point at which a linear elastic material begins to damage and fail. This approach will be used in this study to identify the transverse tensile strength.

The VFM equations using rigid body virtual fields can all be used to approximate the stress state in a material at failure and therefore provide an estimate of the materials strength. Compared to other virtual fields, these equations provide a stress information which is resolved in the x direction (i.e. averages over a cross-section) while other virtual

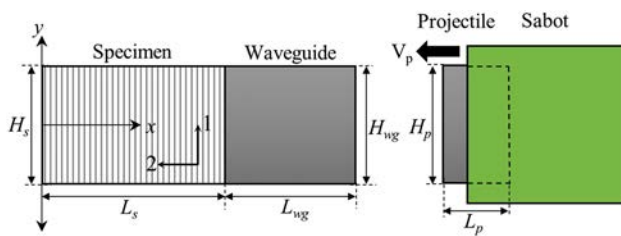


Fig. 3 Schematic of test configuration showing the main test components and relevant geometric variables. For the experiments and the finite element model the specimen is a thin rectangular plate and the remaining components are cylindrical. For the finite element modelling the sabot is neglected

fields provide some spatial average over some portion or all of the (x, y) plane. Virtual fields that provide stress averages over a large planar area cannot be used for strength identification due to inadequate spatial resolution. It is worth noting however that the stress averages are not resolved in the y direction. Only a first moment can be derived (see eq. 38) that can provide a linear approximation of the stress through the width. Whether the stress components could be reconstructed with better spatial resolution from acceleration is an open problem. Fundamentally, this is difficult to address as three stress components are required from only two acceleration components.

Numerical Design and Validation

The first step in designing a new test technique is to validate the proposed methodology using numerical simulation. Here, numerical modelling has two purposes: (1) select test parameters to maximise the reflected tensile stress while minimising the input compressive stress; and (2) validate the stiffness identification procedures described previously in "Concept Design and Theory" section.

Model Configuration and Parametric Design Sweep

In order to experimentally implement the impact test in Fig. 1, a number of components are required; these are shown in Fig. 3. To impart the initial compressive pulse into the rectangular specimen, a short cylindrical projectile is used. The projectile is encased in a nylon sabot to reduce friction with the gas gun barrel. In order to minimise effects of small misalignments between the projectile and specimen, a cylindrical waveguide is used between the specimen and the projectile. The waveguide acts like an end tab on typical quasi-static test specimen. Instead of mitigating the effects of the grips in a mechanical test machine the waveguide allows for three dimensional effects in the input stress pulse to be mitigated by the Saint Venant effect before

Table 1 Summary of parametric sweep variables for test design simulation

Parameter	Min	Max	Increment
Projectile length, L_p (mm)	10	50	5
Projectile velocity, V_p (m/s)	5	50	5
Waveguide length, L_{wg} (mm)	40	60	10

transmitting the loading pulse into the specimen [22]. The choice of geometry for the various components in Fig. 3 influences the ratio of input compressive stress to reflected tensile stress. In order to achieve the desired tensile stress and optimise the ratio of input compressive stress to tensile stress, a parametric design sweep is required. This design sweep was conducted using an explicit dynamics finite element model described in the following section. Before describing the finite element model, the variables in Fig. 3 will be considered with respect to experimental constraints.

Based on the available experimental apparatus, the geometry of the specimen and test components are constrained. This removes several variables from the parametric design sweep. The bore of the gas gun to be used for the impact experiments is limited to projectiles of 45 mm diameter (with a 50 mm diameter sabot). For simplicity, the projectile diameter, waveguide diameter and specimen height have been fixed at 45 mm ($H_s = H_{wg} = H_p = 45\text{mm}$). Further constraints on specimen geometry arise from the aspect ratio of the camera used for the full-field measurements. Specifically, to maximise the spatial resolution of the full-field displacement measurements, the aspect ratio of the specimen should match the aspect ratio of the camera. For the experiments performed in this study, a Shimadzu HPV-X camera was used with an aspect ratio of 1.6:1. Additionally, an allowance is required for rigid body motion of the sample during the test such that the free edge of sample does not move out of the field of view. Therefore, the sample length L_s is fixed at 70 mm allowing for 2 mm of rigid body motion.

The remaining test parameters can be chosen based on a design sweep to optimise the test configuration. The remaining parameters to be selected include: projectile length L_p , projectile velocity V_p , and waveguide length L_{wg} . The ranges of these variables that were simulated are summarised in Table 1. Based on the composite material that will be used for the experimental studies, the transverse tensile strength was assumed to be $\sim 50\text{ MPa}$. This was later found to be appropriate based on the quasi-static test campaign and includes an allowance for an increase with strain rate (see "Composite Material and Quasi-Static Tests" section). Data was not available for the transverse compressive strength of this specific material. Therefore, the compressive strength was conservatively estimated to be at least 200 MPa based on the results in ref. [5, 23]. This gives a design envelope of

$\overline{\sigma_{xx}^y} > 50 \text{ MPa}$ during tensile loading and $\overline{\sigma_{xx}^y} > -200 \text{ MPa}$ during the compressive loading. While it is possible to optimise the design based on the local stress rather than the average the average stress has been selected here as it will give a conservative prediction for the maximum tensile stress that is reached.

Finite Element Implementation

The explicit dynamics simulations for this study were conducted using ANSYS APDL LS-DYNA (v 16.2). A schematic of the test configuration is shown in Fig. 3. The parametric design sweep was conducted using a 3D finite element model. The specimen was meshed with quadrilateral SOLID164 elements (8-node, linear, reduced integration) and the cylindrical waveguide and projectile were meshed with tetrahedral SOLID168 elements (10-node, quadratic). The sabot shown in Fig. 3 was neglected in the 3D simulation for computational efficiency. Contact was modelled between all components using the default, automatically generated, three dimensional contact algorithm in LS-DYNA.

A convergence study was conducted using a parametric sweep of mesh size, time step and beta damping (stiffness proportional damping). Note that beta damping was used to reduce the numerical errors that occur as high frequency oscillations in the acceleration fields. The simulation parameters were selected from this sweep in order to minimise the error between the average stress $\overline{\sigma_{xx}^y}$ calculated from the accelerations (eq. 16) and the average stress extracted from the finite element model. The selected simulation parameters are summarised in Table 2. As the specimen behaviour was of greatest interest the mesh size of the projectile and waveguide was set to be more coarse than the specimen with an element edge length of 4 mm (compared to 0.5 mm elements for the specimen).

The boundaries of all components were assumed to be free. The only condition applied to the simulation was the initial velocity of the projectile. The total simulation time was set such that the stress wave generated by the impact between the projectile and waveguide could travel through the waveguide and the specimen to the free edge, reflect and travel back through the length of the specimen again. All materials were modelled as homogeneous and linear elastic. The unidirectional carbon fibre composite specimen was modelled as orthotropic with material properties $\rho = 1600 \text{ kg/m}^3$, $E_{11} = 135 \text{ GPa}$, $E_{22} = E_{33} = 10 \text{ GPa}$, $\nu_{12} = \nu_{13} = 0.02$, $\nu_{23} = 0.3$, $G_{12} = G_{13} = 5 \text{ GPa}$ and $G_{23} = 3.85 \text{ GPa}$. The projectile and waveguide material were modelled as Aluminium 6061-T6 with isotropic material properties $\rho = 2700 \text{ kg/m}^3$, $E = 70 \text{ GPa}$ and $\nu = 0.33$. The initial distance between the projectile and the waveguide was set based on the projectile

Table 2 Explicit dynamics simulation parameters

Specimen mesh size	0.5 mm
Nominal time step	$0.5t_{crit} = 0.1\mu\text{s}$
Beta damping	0.1
Data output step	0.5 μs

velocity such that there were three output steps before the projectile contacted the waveguide.

Parametric Sweep Results

The parametric sweep results showed that the length of the waveguide (L_{wg}) had minimal effect on the input compressive and reflected tensile stresses. This was the case as long as the waveguide was at least twice the projectile length. When the projectile is the shortest component, the projectile length controls the pulse duration (i.e. the pulse duration is twice the time it takes to traverse the length of the projectile). Therefore, as long as the waveguide is at least double the projectile length, it will not truncate the overall pulse duration. Based on practical considerations, the length of the waveguide was selected to be 50 mm. This choice was made to minimise the mismatch in mass between waveguide and projectile to reduce the potential for the projectile to reflect back towards the barrel and cause potential damage to it.

The results of the parametric sweep for the remaining parameters are shown in Fig. 4. It is desirable to maximise the proportion of input compressive stress that is reflected as tensile stress. Therefore, the ratio of reflected tensile stress to input compressive stress is shown in Fig. 5. It is interesting to note that in Fig. 5, the curves are effectively insensitive to the projectile velocity and collapse onto each other. The combination of Figs. 4 and 5 suggests that the projectile length determines the ratio of reflected tensile to compressive stress while the projectile velocity determines the magnitude of the input compressive stress.

From the data shown in Figs. 4 and 5, it is possible to select a wide range of parameters that will achieve the desired design constraints. The design envelope that satisfies the requirements is given by: projectile velocity [$20 \leq V_p \leq 40 \text{ m/s}$] and projectile length [$10 \leq L_p \leq 50 \text{ mm}$]. The configuration selected for the experiments is as follows: waveguide length of $L_{wg} = 50 \text{ mm}$, projectile length of $L_p = 25 \text{ mm}$ and projectile velocity of at least $V_p = 30 \text{ m/s}$. This configuration leads to a reflected average tensile stress of 105 MPa and an input average compressive stress of -188 MPa . It should be noted that the selected configuration is not the most efficient. The reason for this selection was based on economical reasoning; that is, the projectile design was suitable for testing several materials (e.g. soda-lime

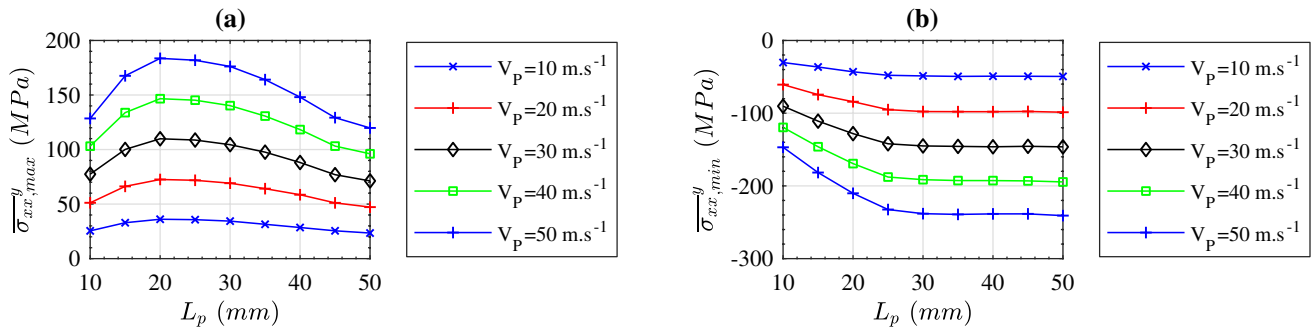


Fig. 4 Parametric sweep results for the 3D finite element model **a** the maximum average tensile stress $\overline{\sigma_{xx,max}}$ and **b** the maximum input compressive stress $\overline{\sigma_{xx,min}}$ plotted against the projectile length L_p

and projectile velocity V_p . The waveguide length $L_{WG} = 50$ mm. Note that every second velocity value has been plotted for clarity

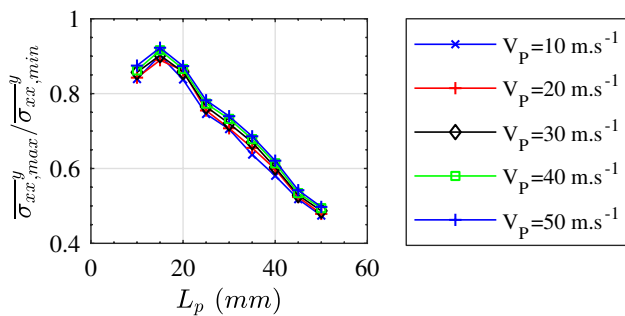


Fig. 5 Ratio of reflected average tensile stress to input compressive stress from the 3D finite element model $\overline{\sigma_{xx,max}} / \overline{\sigma_{xx,min}}$. Note that every second velocity value has been plotted for clarity

glass) using an IBII test and allowed for a large batch of projectiles to be made, reducing costs.

For the selected test configuration the finite element model predicts that the maximum average tensile stress occurs at $x = 11.25$ mm and $t = 42.5$ μ s. The σ_{xx} stress field at $t = 42.5$ μ s is shown in Fig. 6a. The stress field in Fig. 6a is interesting as the peak stress occurs in the centre of the specimen. Therefore, it is possible that the specimen will fail in the centre rather than at the edges. The predicted maximum average tensile stress at different axial locations is shown in Fig. 6b. Analysis of Fig. 6b shows that just after the first tensile peak another occurs near the centre of the specimen length. If the first tensile peak does not cause failure it is possible that a failure will occur near the centre of the specimen length. If it is assumed that the failure strength of the tested material is at least 50 MPa then the model predicts that failure will occur between $x = 5$ to 40 mm and $t = 41$ to 52 μ s.

As the purpose of this parametric design sweep is to design a high strain rate test, it is also interesting to analyse the predicted strain rates that are achievable. In Fig. 7, the strain rate fields are shown for a time before the wave has reflected (a), and following the reflection of the wave

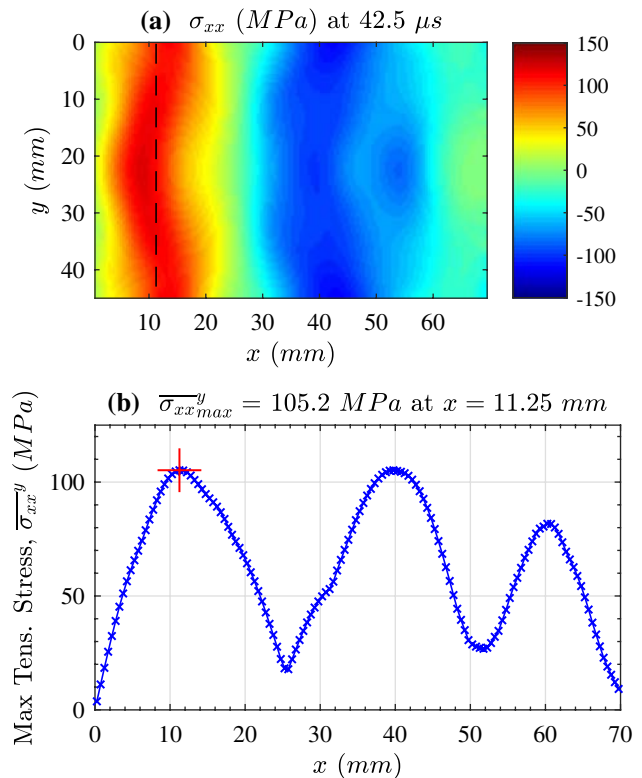


Fig. 6 **a** Finite element model (3D case) stress field σ_{xx} for the selected test configuration indicating the time and location of maximum average tensile stress. **b** Maximum average tensile stress as a function of axial location. The predicted maximum average tensile stress is highlighted with a red cross. (Color figure online)

(b). These figures show that for the IBII test the strain rate field is heterogeneous over the specimen surface. Further measures of the axial strain rate can be obtained by averaging over the whole field of view as in Fig. 7c or averaging over the width at a given axial position as in Fig. 7d. For the selected test configuration, the finite element model predicts peak width averaged strain rates on the order of 2000 s^{-1} .

Stiffness Identification Validation

Here a finite element model will be used to validate the stiffness identification methods proposed in "Virtual Fields For Stiffness Identification" section. In order to have sufficient mesh density for the acceleration and stress field to be in equilibrium it was necessary to run a 2D plane stress model that only considered the sample. The validity of using a 2D plane stress assumption for the specimen was assessed using the results of the 3D model as detailed in the Appendix. An additional convergence study was conducted on the finite element model for the stiffness identification procedures. The results of this study showed that the previously selected damping and time step parameters were suitable. However, a finer mesh of 0.25 mm was required. The sample was meshed using PLANE162 elements (4-node, reduced integration, plane stress) and

the loading was applied using a pressure pulse applied to the impacted edge of sample. The magnitude and temporal evolution of the loading pulse was set to replicate the loading conditions observed in the 3D model. All other edges of the sample were considered to be free.

Full-field strain and acceleration maps were extracted from the finite model at each output time step. The acceleration and strain fields are shown in Fig. 8 at a time point when the compressive wave has fully entered the specimen. The acceleration fields from the finite element model were used to calculate the average stress $\overline{\sigma_{xx}^y}$ according to eq. 16. Stress-strain curves were then plotted at all axial sections along the specimen length using the average axial strain $\overline{\epsilon_{xx}^y}$. The stress-strain curves were fitted with a linear function to obtain the elastic modulus E_{22} . A stress-strain curve reconstructed near the specimen centre ($x = 35$ mm) is shown in Fig. 9a. The identified

Fig. 7 Predicted strain rate from the 3D finite element model. Strain rate maps over the specimen surface at $t = 24.5 \mu\text{s}$ (a) and at $t = 50 \mu\text{s}$ (b). c Average strain rate over the specimen surface $\overline{\epsilon_{xx}^y}$ for the test duration and d axial averaged strain rate $\overline{\epsilon_{xx}^y}$ at $x = 35$ mm. (Color figure online)

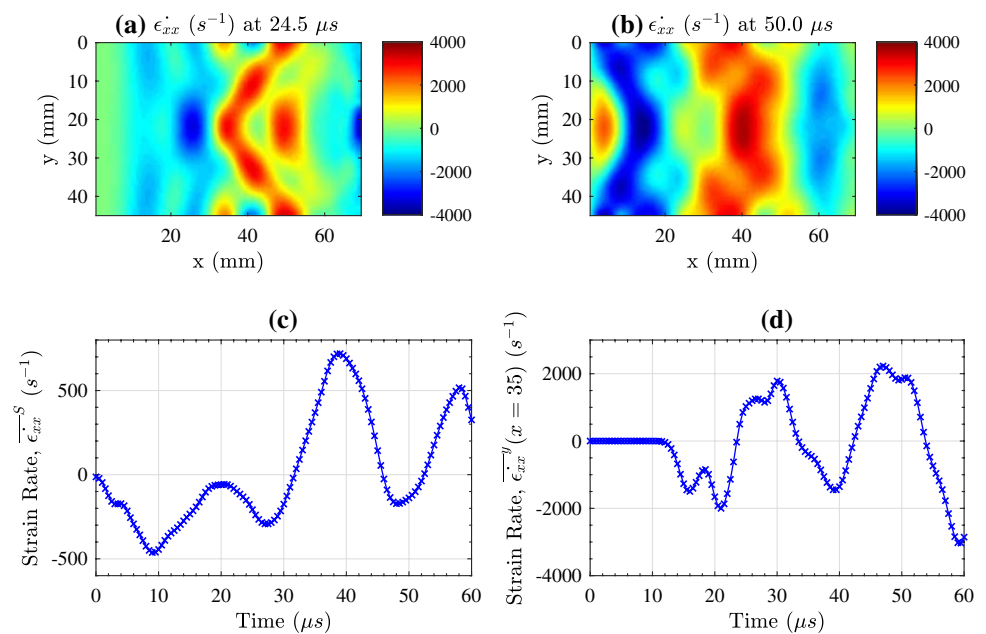
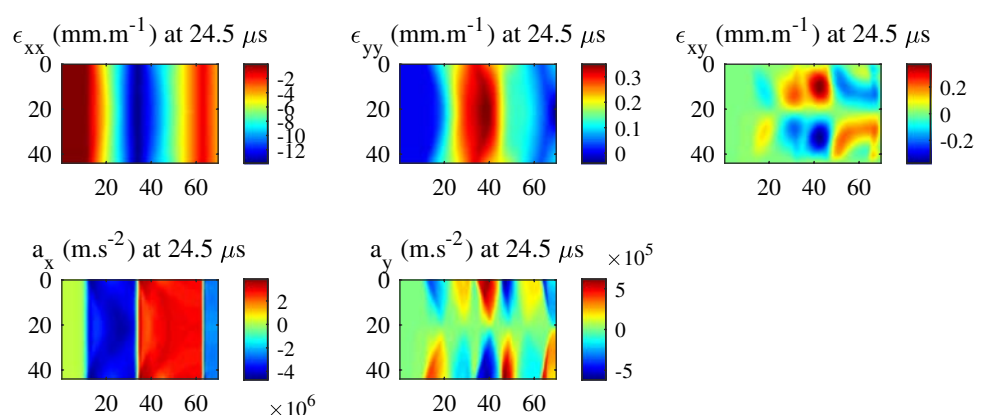


Fig. 8 Finite element model (2D case) strain and acceleration fields for the configuration: wave guide length $L_{wg} = 50$ mm, projectile length $L_p = 25$ mm and projectile velocity $V_p = 30$ m/s at $t = 24.5 \mu\text{s}$. (Color figure online)



elastic modulus from linearly fitting all stress-strain curves along the specimen length is shown in Fig. 9b. Analysis of Fig. 9 shows that the identification from the finite element data is stable and within 2% of the input value over the specimen length.

The stiffness identification procedure was also validated for the manual and special optimised virtual fields using the finite element data. The identification using the manual virtual field procedure is shown in Fig. 10a and the identification using the special optimised virtual field is shown in Fig. 10b. For the special optimised virtual fields, a virtual mesh of 5 by 1 elements was used. A virtual mesh refinement study was conducted and showed that the identification converged with a virtual mesh of 3 by 1 elements. Here, 5 by 1 elements are used to be consistent with the convergence study performed on experimental data. For both types of virtual fields, the identification is of poor quality at the start of the test when the wave is entering the

specimen. The reason for this is that there is inadequate spatial information to identify the material properties (i.e. strain and acceleration). The manual virtual fields show additional points of instability which correspond to a situation where both terms in eq. 18 come close to zero. This instability is not evident in the special optimised virtual fields. The reason for this is that the special optimised virtual fields follow the input spatial information leading to an improved identification. This is shown in Fig. 11 where the u_x^* special optimised virtual field highlights the acceleration peak.

When the identification is stable the variation is low and well under 1% of the input target value. All stiffness identification methods show excellent agreement with the input stiffness value when applied to finite element data. This validates the approach and the next section is dedicated to the experimental implementation of the method.

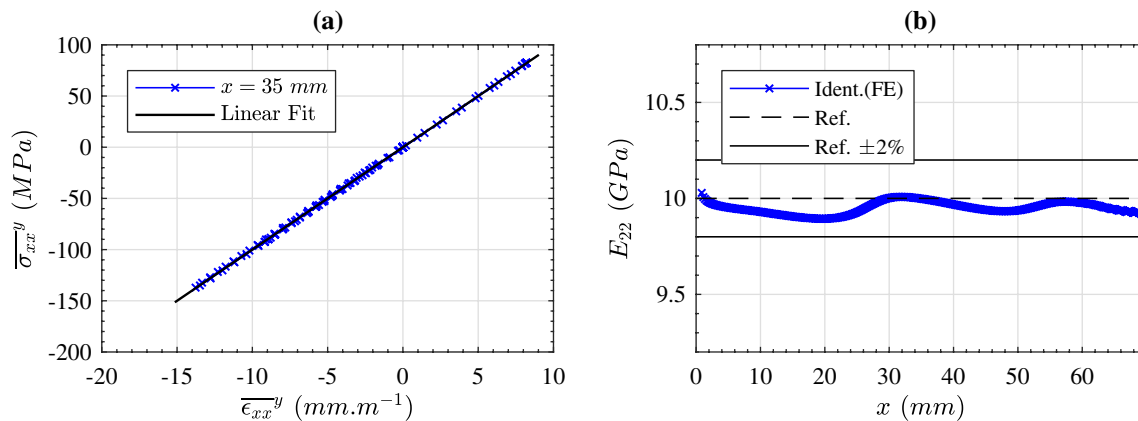


Fig. 9 **a** Stress-strain curve reconstructed from the 2D finite element model acceleration and strain fields near the specimen centre $x = 35$ mm. **b** Identified elastic modulus from fitting stress-strain curves over specimen length for the finite model

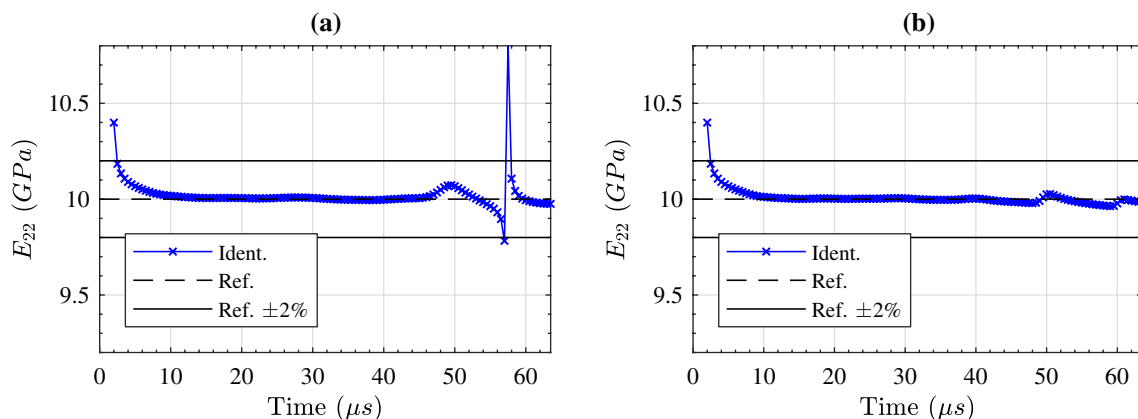


Fig. 10 **a** Identification over time for the manually defined virtual field applied to the 2D finite element model. **b** Identification over time for the special optimised virtual fields applied to the 2D finite element model

Experimental Methodology

Composite Material and Quasi-Static Tests

The composite material used in this study was a unidirectional pre-preg tape with a low temperature cure epoxy matrix, Gurit SE70. The layup of the material was unidirectional with 15 layers ($[90^\circ]_{15}$). The layers of composite pre-preg tape were mounted on a ground aluminium plate that acted as a mould. The composite was then cured in a vacuum bag under 1 bar pressure in an autoclave oven at 110°C for 50 minutes. The final cured plate had a thickness of 4.75 mm with a nominal fibre volume fraction of 55%. Six specimens were machined from the plate using an irrigated diamond saw to the nominal in-plane dimensions of 70×45 mm. The density of the material was measured using a micro balance and water immersion to be $1530 \pm 25 \text{ kg/m}^3$.

Quasi-static testing of the material was also undertaken using the same pre-preg material. For the quasi-static testing, six specimens were cut from a unidirectional plate ($[90^\circ]_8$) cured using the same method described above. General testing conditions and specimen configuration conformed to the standard ISO 527-5:2009 [24]. All quasi-static testing was conducted on an Instron electromechanical load frame with a 50 kN load cell. Strain measurements were performed using back-to-back unidirectional strain gauges mounted in the centre of the specimen and aligned with the long axis of the specimen. The strain rate for the quasi-static tests was $1 \times 10^{-4} \text{ s}^{-1}$. It was not attempted to measure the minor Poisson's ratio ν_{21} here as it is very small and hard to measure accurately with this configuration. The results of the quasi-static testing are summarised in Table 3 and are provided as a comparison with the high strain rate data

presented later in the "Experimental Results and Discussion" section. The average quasi-static strength is low compared to manufacturer data for this pre-preg (nominally $\sigma_F = 42 \text{ MPa}$). Subsequent microscopic observations showed that the cured pre-preg plate contained porosities, explaining the low quasi-static strength. The scope of this study is to develop the framework for the IBII test, not to fully characterise the properties of a particular composite system. Following this first validation of the IBII test method subsequent studies will need to be performed on well characterised composite systems.

Impact Apparatus

The IBII tests were carried out using a custom built gas gun facility at the University of Southampton. The waveguide and projectile were cylinders machined from an Aluminium

Table 3 Summary of experimental quasi-static testing results for transverse properties of the Gurit SE70 pre-preg conducted at a strain rate of $1 \times 10^{-4} \text{ s}^{-1}$

Specimen	E_{22} (GPa)	σ_{22}^F (MPa)
1	7.43	34.3
2	7.27	39.2
3	7.31	39.6
4	7.32	31.3
5	7.20	37.9
6	7.21	38.5
Mean	7.29	36.8
SD	0.085	3.32
COV (%)	1.17	9.01

Note that SD represents the standard deviation and COV represents the coefficient of variation

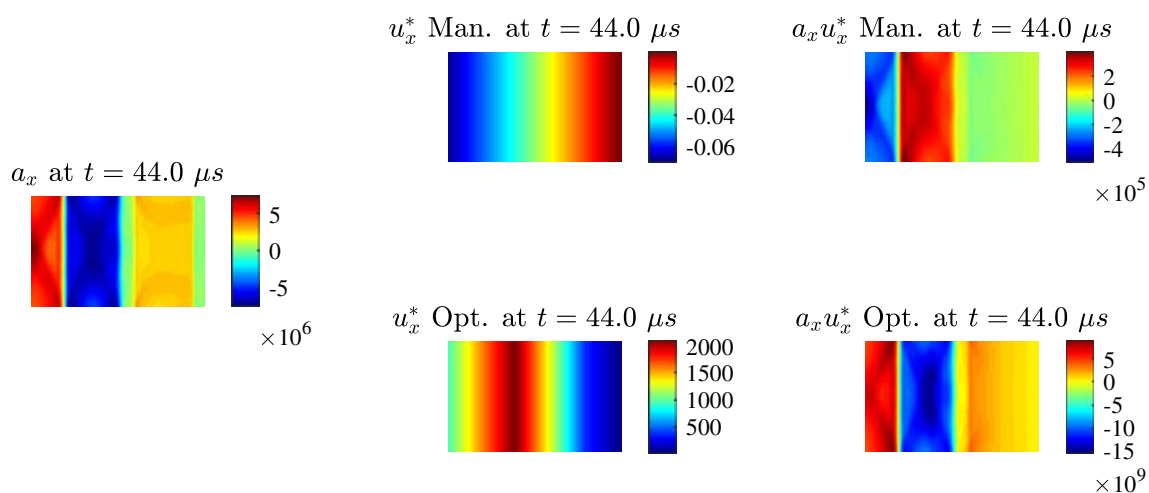


Fig. 11 Comparison of manual (Man.) and special optimised (Opt.) virtual fields at $t = 44 \mu\text{s}$ applied to the 2D finite element simulation. The product of the virtual and kinematic fields shows that the optimised virtual fields follow the areas of high input signal. (Color figure online)

6061-T6 rod with dimensions conforming to those selected in the parametric design sweep. The specimen was bonded to the cylindrical wave guide using cyanoacrylate glue. The waveguide was placed on a wedge-shaped foam stand 200 mm in front of the barrel exit. The foam stand does not directly attach to the waveguide, the waveguide is held in the centre of 'v' shaped stand by gravity. A custom mounting block was used to align the foam stand and wave guide with the gas gun barrel. A photograph of the impact chamber and a mounted specimen is shown in Fig. 12.

Two infra-red light gates were mounted in front of the exit of the barrel, these were connected to an Arduino Uno micro controller. The light gates were used to calculate the projectiles speed and to automatically trigger the photographic flash accounting for the flash rise time, 100 μ s (Bowens Gemini Pro 1000). A copper contact trigger was used on the impacted face of the waveguide to trigger the camera. A delay of 11 μ s was used between the trigger event and image recording to account for the time taken for the input pulse to traverse the length of the waveguide before entering the sample. The gas gun reservoir pressure was set for a nominal impact speed of 35 m/s. The actual impact speed for all samples was 35.3 ± 5.9 m/s (mean $\pm$ standard deviation). It should be noted that accurate control of the projectile speed is not required here as the response of the specimen is fully characterised by the measured full-field kinematics.

Imaging and Full-Field Measurement Setup

All experimental imaging was performed using a Shimadzu HPV-X ultra-high speed camera and a Sigma 105 mm macro lens. Additional imaging parameters used in this study are summarised in Table 4. The grid method was used for all

full-field displacement measurements as it offers improved compromise between spatial and measurand resolution compared to DIC [25]. This is especially important here given the low number of pixels on the Shimadzu HPV-X (400 $\times$ 250 pixels). The main drawback of the grid method is that a regular grid needs to be deposited on the surface of the specimen. However, the regularity of the grid can be relaxed if an iterative displacement calculation procedure is used as described in ref. [26]. We note here that there is a similar requirement for optimal use of DIC as extremely large speckles require the use of large subsets, compromising spatial resolution and extremely small speckles cause aliasing. Thus, for optimal use of DIC strict control of speckle size is required and producing a high quality speckle pattern presents similar challenges to producing a regular grid. For this study, the grid method was used along with the iterative displacement calculation described in ref. [26]. Grids with a pitch of 0.9 mm were bonded to the specimens using the process outlined in ref. [27]. The grid pitch and sampling were selected such that the sample covered most of the camera field of view, maximising the use of available camera pixels.

There are practical issues associated with the technology used in the Shimadzu HPV-X camera, most notably, the low pixel fill factor. If not corrected for by slightly defocusing (i.e. blurring) the grid image, this can lead to a systematic error in the strain which manifests as an artificial fringe pattern in the strain fields [28, 29]. This raises the question of 'how much should the image be blurred to remove the fill factor problem?'. Therefore, prior to each impact experiment, the image was blurred and an out-of-plane movement test was conducted. The out-of-plane movement test was performed by taking a series of static images, then moving the camera back 5 mm using a translation stage attached to the tripod. A second series of static images were then taken

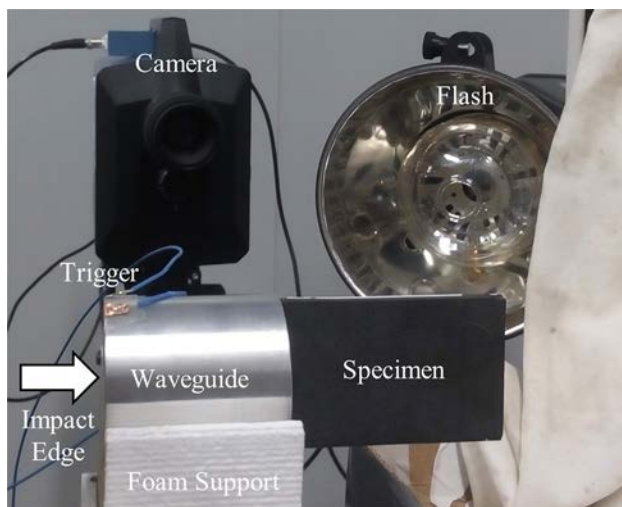


Fig. 12 Photograph of a transverse carbon fibre test specimen in-situ prior to testing

Table 4 Summary of imaging, grid and smoothing parameters used for the experiments

Camera	Shimadzu HPV-X
Pixel array size	400 $\times$ 250
Total frames	128
Inter frame time	0.5 μ s
Integration (shutter) time	0.2 μ s
Grid pitch	0.9 mm
Grid sampling	5 pixels/period
Field of view	72 $\times$ 45 mm
Raw displacement resolution	0.3 μ m
Spatial smoothing	Gaussian 25 $\times$ 25 pixels
Temporal smoothing	Savitsky-Golay, 3rd order over 17 frames
Strain resolution	80 μ m/m
Acceleration resolution	6×10^5 m/s ²

and, correlated to the first using the grid method. The out-of-plane movement leads to an artificial in-plane hydrostatic strain because of the change in magnification. If the blurring is not sufficient to mitigate the fill factor problem, then fringes appear in the strain fields which should nominally be uniform spatially. This process is repeated until the fringes almost completely disappear from the strain fields. The camera is then moved back to its original position and another series of static images are taken prior to testing. For further information on the full-field measurement performance of ultra-high speed cameras with specific detail on the associated problems with the fill factor, the reader is referred to [28, 29].

Experimental Data Processing

The raw grey-level images were processed using an open source grid method processing code [26]. More detailed information on the grid method theory, practical implementation of the method and data processing can be found in a detailed review by Grédiac et al. [26]. In brief, the grid method uses a priori knowledge that the pattern on the specimen is regular and nominally sinusoidal. Signal processing algorithms can then be used to determine the phase at each camera pixel sampling of the grid. The phase is related directly to the displacement field of the specimen. Phase maps taken from the specimen surface can exhibit sharp discontinuities when the in-plane deformation is large. To account for this, spatial unwrapping is required. The algorithm used here was developed in ref. [30] and comes pre-packaged with the grid processing code of Grédiac et al. [26]. Discontinuous jumps in the phase can also occur between frames (i.e. in time) if the rigid body motion of the test sample is larger than a grid pitch. For the impact test considered here, the rigid body motion is small between frames (much less than the grid pitch) and it increases monotonically. Therefore, the temporal discontinuities can be accounted for by adding integer multiples of the 2π to the phase maps to ensure that the average displacement increases monotonically over the test duration. This is referred to as temporal unwrapping. The temporal unwrapping function used here was developed in-house. Subsequent processing of the full-field displacement maps was performed using custom-built Matlab programs (v R2015a).

Spatial and temporal smoothing were used to reduce the effects of camera noise and reach the required strain and acceleration resolutions. Smoothing filters were applied directly to the displacement fields before numerically differentiating in space (for strain) or in time (for acceleration). The numerical differentiation method was a centred finite difference with a forward or backwards difference at start or end of the data series (as implemented by the Matlab function *gradient*). For the strain calculation, only a spatial filter

was used (i.e. no temporal smoothing) on the displacement field before numerical differentiation to obtain the strains. To obtain the acceleration fields only a temporal filter was used on the displacement fields (i.e. no spatial smoothing). The smoothed displacement fields were then numerically differentiated in time to obtain the velocity fields and then differentiated again with the same technique to obtain the acceleration. Relevant smoothing parameters are summarised in Table 4. For the Gaussian smoothing filter, problems arise on the borders of the data. These were dealt with by using the in-built symmetric padding option in Matlab.

While it is beyond the scope of the present work, it should be noted that the optimal method for determining filter parameters (and the resulting random and systematic errors) is to use image deformation simulations [31, 32]. The underlying principle of this technique is described in detail in refs. [31–34]. Briefly, the idea behind this technique is to use displacement fields taken from a known solution to the underlying kinematics that encodes the desired constitutive parameters. If an analytical solution is available this can be used but for the dynamic case considered here the displacement fields can be taken from a finite element model. These displacement fields are then used to synthetically deform a set of images to replicate the spatial/temporal resolution and noise of the camera. These images can then be processed using exactly the same procedure as the experimental data (i.e. same number of pixels, frame rate, grid pitch etc.). The identified parameters can then be compared to the known constitutive parameters of the finite element model to determine the error. By sweeping a number of different processing parameters (i.e. different smoothing kernel sizes) with different copies of noise applied to the images the random and systematic error for each filter combination can be determined. Obviously, the largest smoothing kernels will result in the smallest random error but heavy smoothing will truncate the peaks in the signal leading to a large systematic error. The opposite is true for no smoothing. However, there is an optimum compromise between systematic and random error that can be determined using the image deformation procedure. Work is currently under way by the authors to develop an image deformation software pipeline to allow for the assessment of measurement error in identifying the stiffness and strength using the IBII technique described in this paper, but this is beyond the scope of this first paper.

Apart from smoothing to reduce random noise effects, there are further practical considerations associated with all full-field techniques, specifically the handling of missing data at the specimen border. For the grid method, one grid pitch worth of data at the specimen border is lost. The missing data on the border has been shown to have an effect on material models identified using the VFM combined with DIC [35]. However, in the study by Rossi et al. [35], it was shown that padding or extrapolating the data can lead to a

significant improvement in the identified parameters. For the impact test considered here, the missing data on the specimen border will affect the identification procedure. In order to address this problem, the strain and acceleration data were linearly extrapolated over the missing data region using the extrapolation option of the Matlab function *interp*. The extrapolation was performed after all smoothing and numerical differentiation procedures.

The raw images, data processing program and associated output are provided for all tested specimens in the data repository detailed at the end of the manuscript.

Experimental Results and Discussion

Full-Field Measurement Results

For all subsequent images and analysis presented, specimen 4 was selected as a typical representation of all specimens tested. Note that additional images and videos are provided for all tested specimens in the data repository mentioned at the end of the manuscript. The full-field displacement maps for specimen 4 are shown in Fig. 13 for the approximate time that the compressive wave is halfway through the sample ($t = 24.5 \mu\text{s}$), and at a time after the wave has reflected becoming tensile ($t = 52 \mu\text{s}$). The corresponding acceleration and strain maps are shown in Figs. 14 and 15 respectively.

The acceleration map shown in Fig. 14a clearly shows the front and back edge of the input stress pulse. The

acceleration maps for a_x show peaks on the order of $10^6 g$. The experimental strain maps also confirm that the magnitude of the ϵ_{yy} strains are significantly lower (approximately one order of magnitude) than the ϵ_{xx} strains. For the case of the ϵ_{yy} strains shown in Fig. 15c, d, it is clear that the signal is much noisier than for the ϵ_{xx} strains. This provides experimental evidence to validate the assumption that the contribution of the ϵ_{yy} strain is small when formulating the constitutive law in the "Constitutive Model and Assumptions" section. The maps of the $\dot{\epsilon}_{xx}$ strain rate are shown in Fig. 16. These strain rate maps are heterogeneous and show that different regions of the material experience different strain rate amplitudes. It is difficult to gain a thorough understanding of a dynamic experiment from static images alone therefore, videos of the kinematic fields are provided for each specimen in the data repository associated with this paper.

From the surface average of the acceleration field the average loading force on the impact edge can be reconstructed: $\overline{F}_x^y = m_s \overline{a}_x^S$ where m_s is the mass of the sample and $\overline{F}_x^y$ is the average force applied to the sample at the impact edge. The loading pulse for specimen 4 is shown in Fig. 17 showing peak loads on the order of 15 kN in compression. The processed full-field acceleration and strain data can now be used to identify the stiffness and strength using the methods described in the "Concept Design and Theory" section. The stiffness identification is presented in the following section "Stiffness Identification" while the strength identification is presented in the "Strength Identification" section.

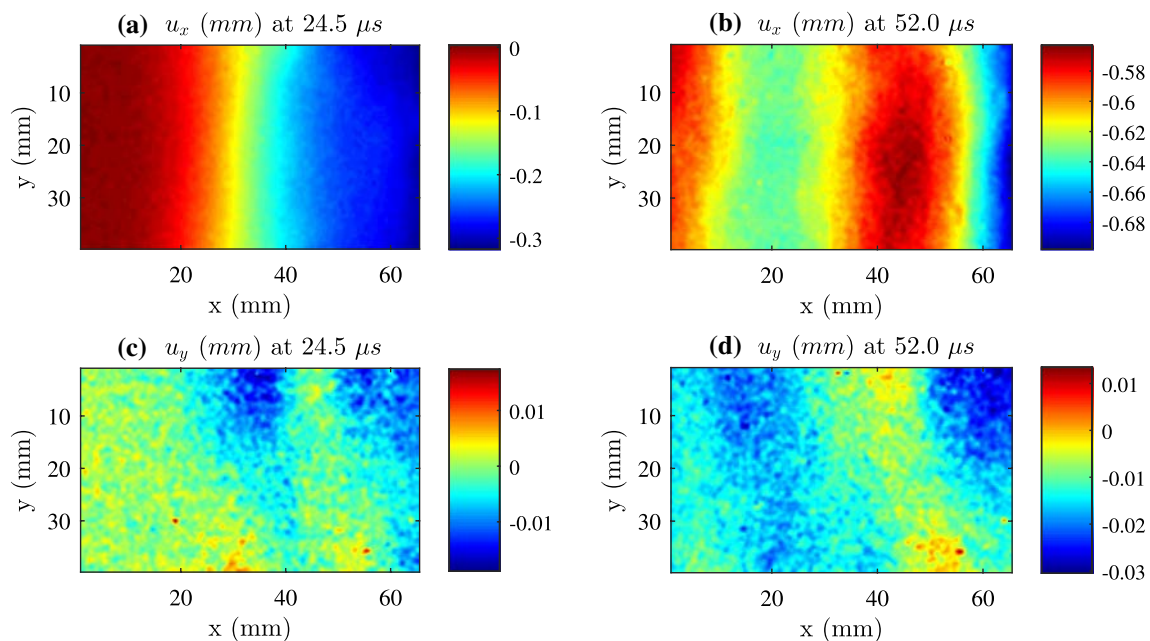


Fig. 13 Experimental displacement fields for specimen 4. **a, b** u_x and **c, d** u_y at times $24.5 \mu\text{s}$ and $52 \mu\text{s}$. (Color figure online)

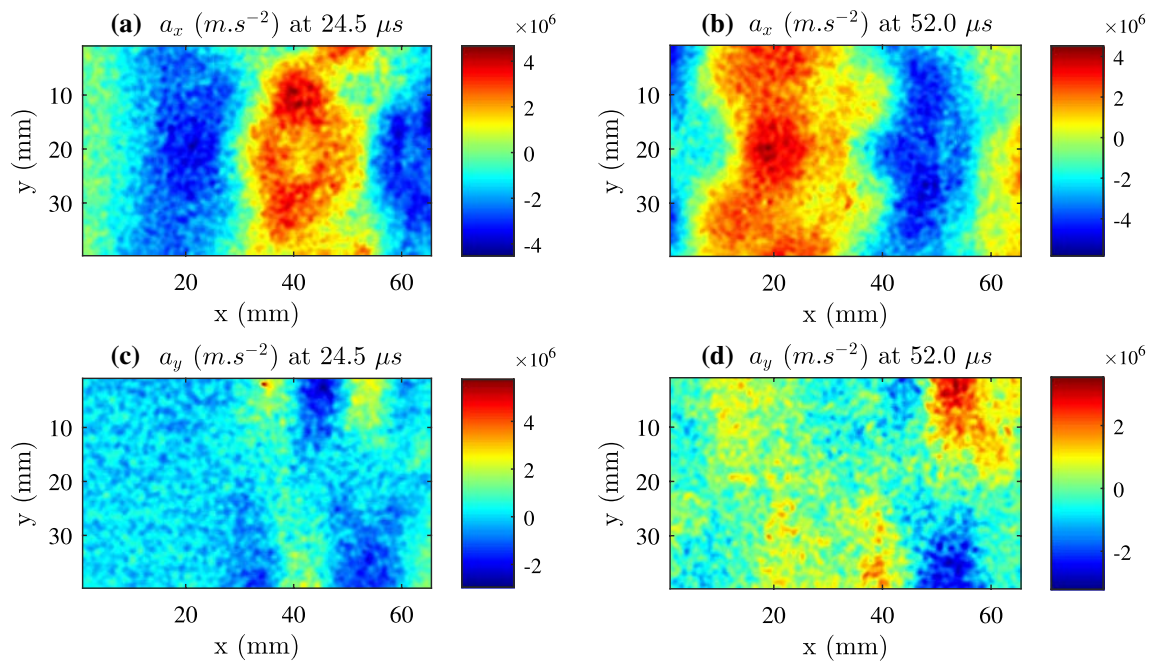


Fig. 14 Experimental acceleration fields for specimen 4. **a, b** a_x and **c, d** a_y at times 24.5 μs and 52 μs . (Color figure online)

Stiffness Identification

The stress-gauge equation was used to calculate the average stress $\overline{\sigma_{xx}^y}$ from the acceleration fields for all six specimens (eq. 16). The peak average compressive stress for specimen 4 was 76 MPa. The experimental compressive stress is well below the given design criterion of an input pulse below 200 MPa. Therefore, the compressive pulse is highly unlikely to cause any compressive damage and corrupt the stiffness identification.

The mean axial stress was used to plot stress-strain curves for different sections along the specimen length. The transverse stiffness E_{22} was then identified by linearly fitting the stress-strain curve up to the point of maximum compressive stress in each section. This avoided using any tensile response as some micro-damage may appear early and compromise the measurement of the initial stiffness of the material. Stress-strain curves for several sections along the length of specimen 4 are shown in Fig. 18a. The identified elastic modulus as a function of position for specimen 4 is also shown in Fig. 18b. Similarly, the identified modulus as a function of length is shown for all six specimens in Fig. 19. Analysis of Figs. 18 and 19 shows that the data near the specimen edges is of poor quality. This is due to several factors including: 1) the data extrapolation on the borders; 2) edge effects in the strain coming from spatial smoothing; and 3) physical defects in the grids at the edges leading to outliers in the kinematic data. If the modulus data at both specimen ends is ignored, a representative modulus

can be obtained by taking the average over the middle 50% of the specimen. For specimen 4, shown in Fig. 18, this results in an identified transverse modulus of 7.9 GPa. The method described previously was used to identify the transverse modulus for all tested specimens. The results are summarised in Table 5. The average modulus identified over all specimens was 7.9 GPa with a coefficient of variation (COV) of 3.2%. The coefficient of variation for the stiffness measurement using this approach is low considering the low pixel count of the camera used. The COV also compares well with that obtained in the quasi-static testing campaign results in Table 3.

The transverse elastic modulus was also identified using the manual and special optimised virtual fields. The transverse stiffness is shown as a function of time for both the manual and optimised fields in Fig. 20 for specimen 4. A mesh convergence study was conducted for the special optimised virtual fields. The stiffness identification was found to stabilise with a virtual mesh of 5×1 elements. Therefore, a 5×1 virtual mesh was used for all specimens. The identification over time for all specimens is shown in Fig. 21. Note that in Fig. 21, the identification has been stopped after the appearance of a macro crack on the strain maps. The identification for all specimens is stable over a relatively large portion of the test, apart from several time frames at which the acceleration signal is low. When the wave first enters the specimen, there is minimal acceleration signal leading to an unstable identification. Once the wave has fully entered the specimen, the identification becomes stable. A

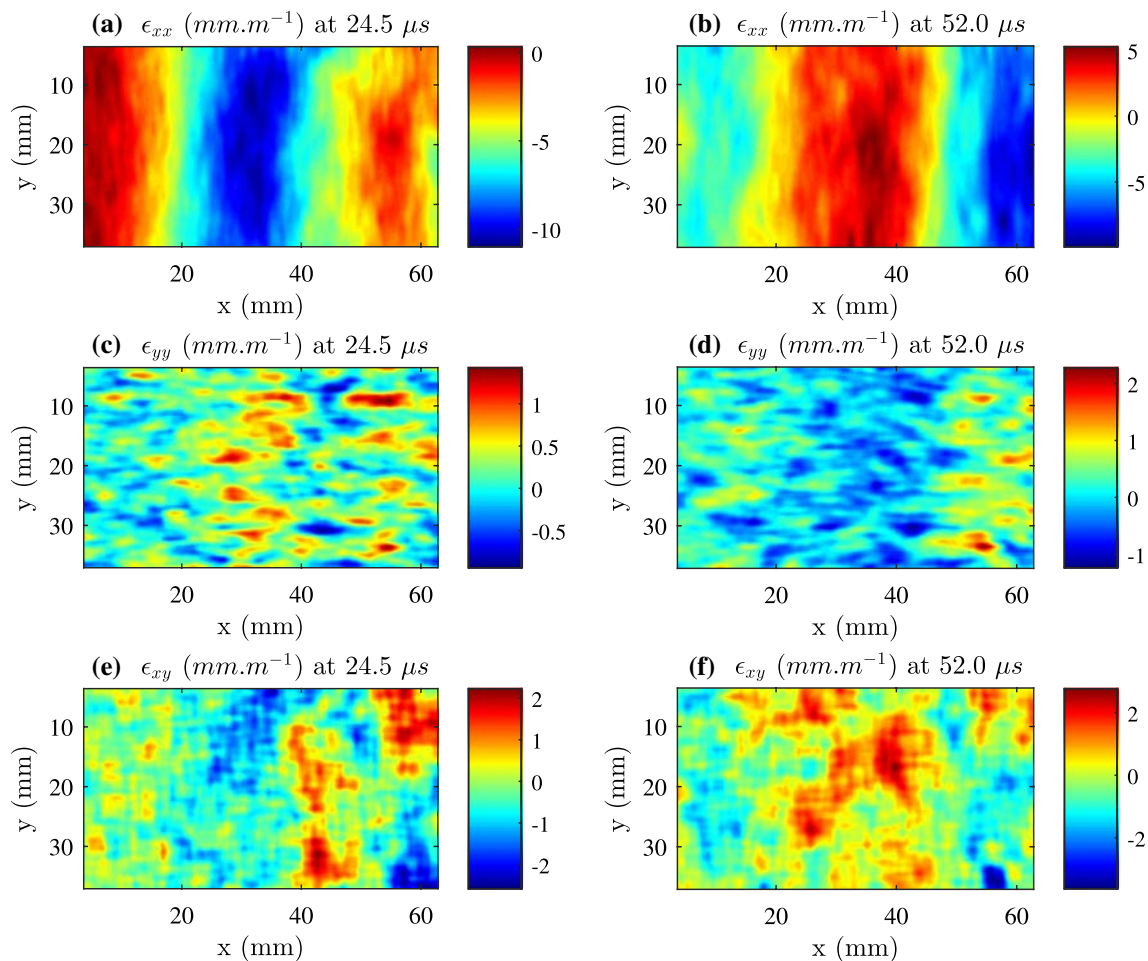


Fig. 15 Experimental strain fields for specimen 4. **a, b** ϵ_{xx} ; **c, d** ϵ_{yy} ; and **e, f** ϵ_{xy} at times 24.5 μs and 52 μs . (Color figure online)

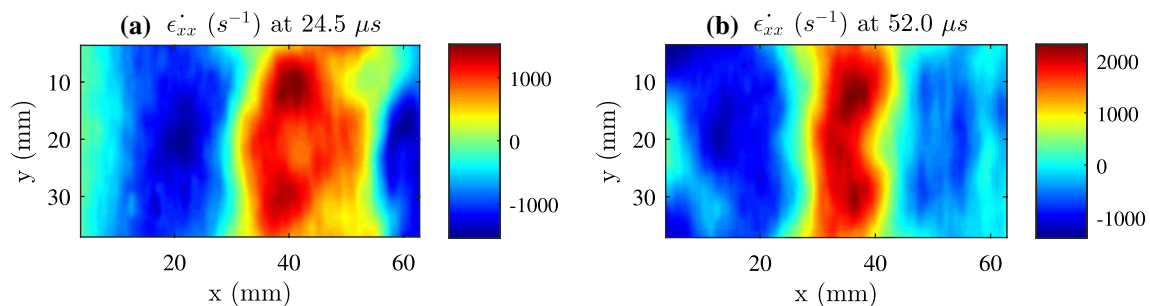


Fig. 16 Experimental strain rate fields for specimen 4. $\dot{\epsilon}_{xx}$ at times **a** 24.5 μs and **b** 52 μs . (Color figure online)

similar instability occurs again once the wave reflects off the free edge leading to low acceleration signal. It is also shown in Figs. 20, 21 that the identified stiffness drops at the end of the test. This is due to the specimen cracking and failing, rendering the virtual fields analysis non-physical (i.e. non-physical strain information due to cracks). In order to make the modulus identified with the the VFM comparable to the

stress-gauge equation, the time frames for the initial compressive loading is used. For this purpose, a representative modulus was identified by averaging between 15 and 30 μs . The average transverse modulus identified for specimen 4 is 7.8 GPa using both the manual and optimised virtual fields. The stiffness identified for each specimen using the VFM is summarised in Table 5. The manual and optimised VFM

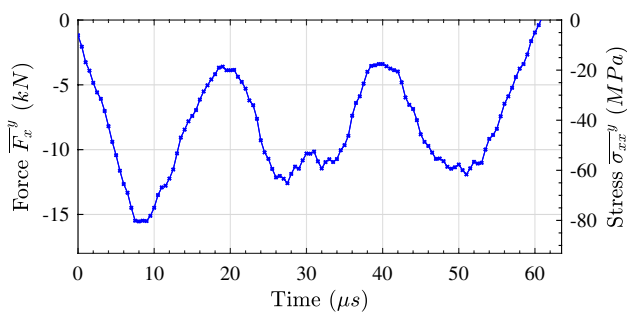


Fig. 17 Experimental loading pulse applied to specimen 4 reconstructed using the average acceleration over the field of view. Note the multiple loading pulses coming from the wave speed mismatch between the projectile and the specimen

Table 5 Identified transverse elastic modulus from experimental data using the stress gauge approach (SG), the manual (VF Man.) and optimised (VF Opt.) virtual fields method: including the mean, standard deviation (SD) and coefficient of variation (COV)

Specimen	E_{22} SG (GPa)	E_{22} VF Man. (GPa)	E_{22} VF Opt. (GPa)
1	8.3	9.9	8.2
2	7.8	8.3	7.9
3	8.2	8.4	8.2
4	7.8	7.8	7.8
5	7.6	8.0	7.6
6	7.9	9.0	7.1
Mean	7.9	8.6	7.8
SD	0.3	0.8	0.4
COV (%)	3.2	8.8	5.5
% Diff. to QS	+ 8.8	+ 18	+ 7.0

Percentage difference with respect to the quasi-static (QS) value is also provided

values show distinct differences when compared across all specimens (see Fig. 21a, b). The reason for this is that the optimised virtual fields method follows the areas of the kinematic fields that contain the largest signal whereas the manual fields act like a static spatial filter regardless of the input signal. Therefore, it is expected that the optimised virtual fields results will be more robust across all time frames, which is what is seen here.

The identification of the transverse elastic modulus is quite consistent between specimens and identification methods except for the manually defined virtual fields. Comparing the results to the quasi-static reference modulus (7.29 GPa) shows an increase of 8.8% when compared to the modulus obtained with the stress-gauge equation, and an increase of 7.0% when compared to the modulus obtained using the optimised virtual fields. This increase in elastic

modulus is expected for the transverse properties of a uni-directional composite as the transverse properties are dominated by the matrix material. For the tested material, the matrix is epoxy. Polymers such as epoxy show some strain rate dependence and are generally thought to have increasing stiffness with increasing strain rate [6, 7]. This increase in stiffness with increasing strain rate is due to a reduction in polymer chain mobility with increasing strain rate [6, 7].

It is difficult to compare the results of this study with previous studies using the SHPB technique at this strain rate. As previously mentioned, the influence of inertial effects only allow an apparent modulus to be obtained with the SHPB [8, 9]. Further complications arise in comparing the transverse properties of composites as they are highly dependent on the matrix material. Differences in the chemical composition of the epoxy resin will lead to differences in the strain rate dependence. For example, more or less cross-linking will change the polymer chain mobility and hence, the strain rate dependence of the material. Here, a tentative comparison can be made with the results of a study by Kwon et al. [36]. Kwon et al. used a high speed hydraulic test machine coupled with DIC to analyse the strain rate dependence of CFRP at strain rates between 0.001 and 100 s⁻¹. The results of this study show an increase of 25% in the transverse modulus from the lowest to highest strain rate tested. This is significantly higher than the percentage increase found in this study. A possible reason for this could be the difference in the matrix material of the composites tested, or experimental difficulties in measuring the applied force at higher strain rates.

When identifying the high strain rate properties of a material, it is important to consider the strain rate at which the material is tested. While this task is trivial for monotonic quasi-static testing, it is more complicated for high rate dynamic testing. The strain rate for high strain rate dynamic tests will inevitably be heterogeneous and transient due to inertial effects (i.e. stress wave propagation). This is evident when considering the strain rate maps in Fig. 16. This raises the question: what strain rate corresponds to the identified modulus values? While there is no definitive answer to this question at this stage, it is possible to extract the peak width averaged strain rate for each specimen to give an indication of the strain rates achieved. A summary of the peak tensile and compressive strain rates for each specimen is given in Table 6. Note that these peak strain rate values are the maximum strain rates before the first macro crack appears on the raw strain maps. The width averaged strain and strain rate history for several axial sections of specimen 4 is also provided in Fig. 22. Note that these axial sections correspond to the stress-strain curves shown in Fig. 18a. Analysis of Fig. 22 shows that for most of the test, when the strain magnitude is moderate the strain rate tends to be high. Therefore it may be possible to define an “effective” strain rate using

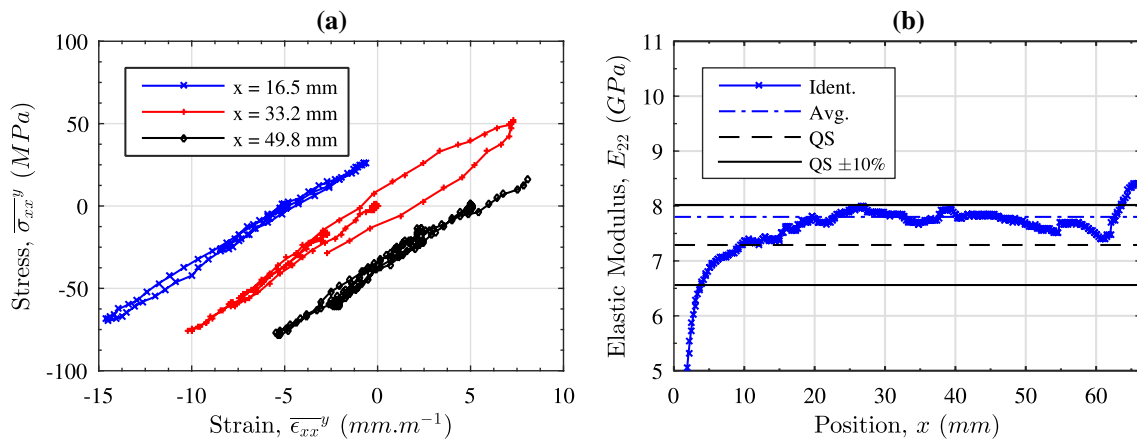


Fig. 18 **a** Stress–strain curves at 25%, 50% and 75% of the specimen length from the free edge. Note that all stress–strain curves start near the origin but have been offset by 5 mm/m for clarity. **b** Transverse

elastic modulus identified using the stress-gauge equation as a function of specimen length including comparison to the quasi-static value (QS). Note that this is experimental data taken from specimen 4

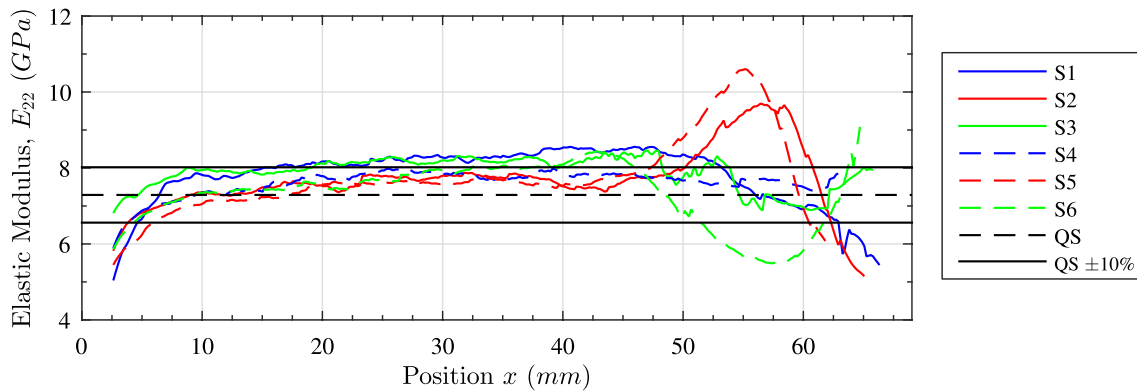


Fig. 19 Transverse modulus identified using the stress-gauge equation for all six specimens as a function of the specimen length. The quasi-static (QS) value is provided for comparison purposes. Note

that the extrapolated data on the specimen ends has been removed. Data markers have been removed for clarity. (Color figure online)

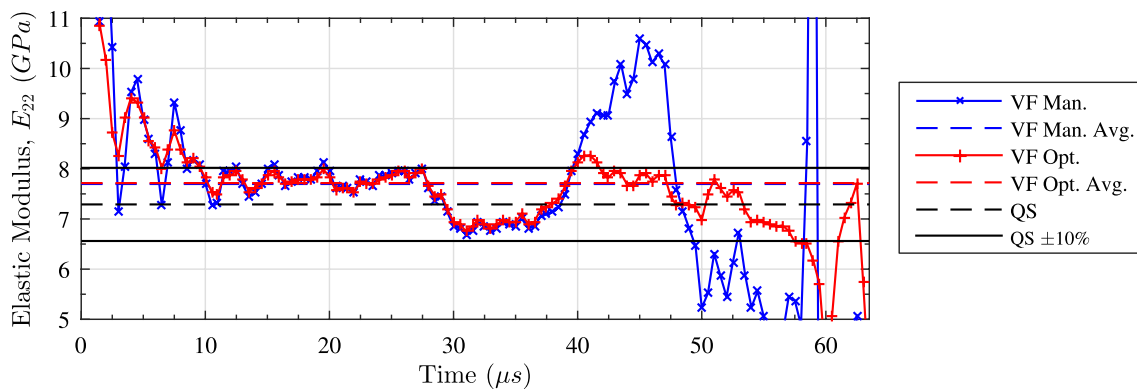


Fig. 20 Transverse elastic modulus identified from experimental data using the manual (VF Man.) and optimised virtual fields (VF Opt.) method as a function of time for specimen 4. The quasi-static (QS) value is provided for comparison purposes. (Color figure online)

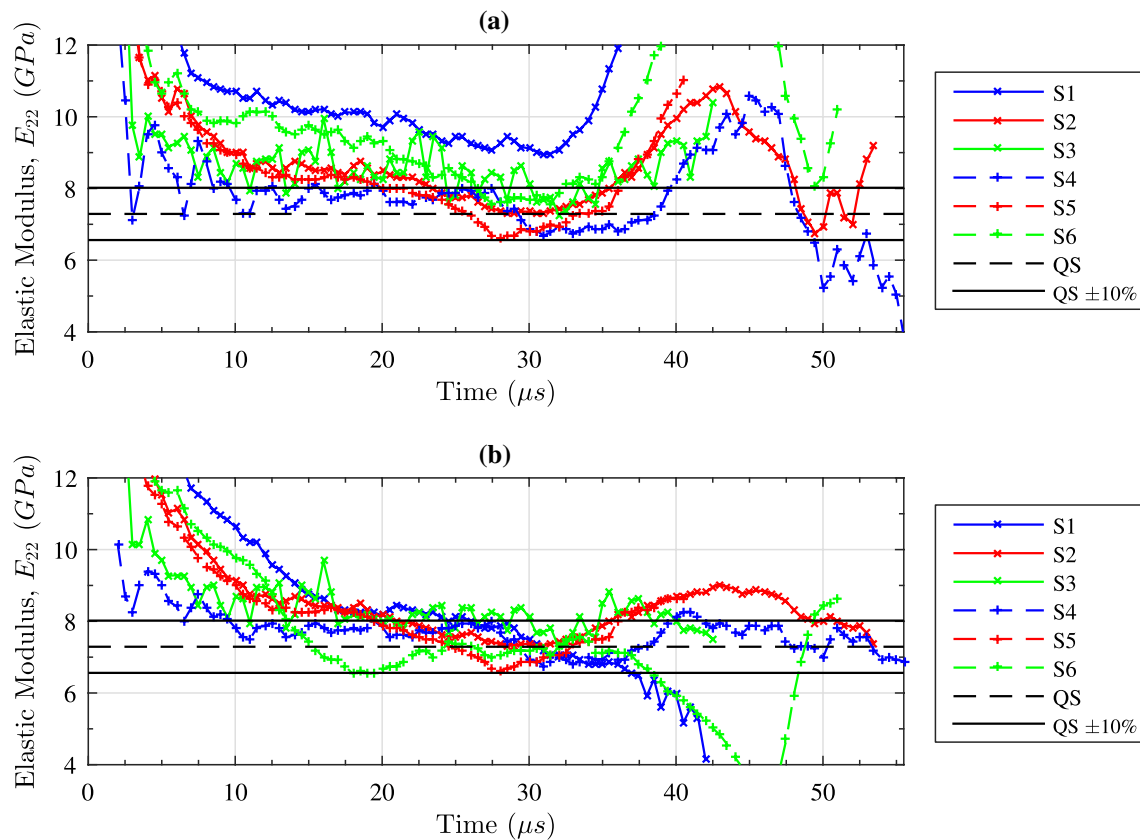


Fig. 21 Transverse elastic modulus as a function of time for all six experimental specimens using **a** the manual virtual field and **b** the optimised virtual fields method. Note that the first five time steps have been removed as there is inadequate spatial information for reli-

able identification. Additionally, the identification has been stopped after the first macro crack appears on the strain maps. (Color figure online)

Table 6 Peak tensile (tens.) and compressive (comp.) width averaged strain rate $\overline{\epsilon_{xx}^y}$ from the experimental data

Specimen	Peak tens. $\overline{\epsilon_{xx}^y}$ (s^{-1})	Peak comp. $\overline{\epsilon_{xx}^y}$ (s^{-1})
1	2200	– 1800
2	2700	– 2200
3	3100	– 2800
4	2200	– 1800
5	2900	– 2400
6	3100	– 2500

this data. However, this is beyond the scope of the present work. For all specimens, the peak width averaged strain rate is on the order of $2000\ s^{-1}$. E_{22} is not very rate sensitive here, with only an 8% increase between quasi-static and a few $1000\ s^{-1}$, so this is not really critical here. If a material exhibited very strong strain rate sensitivity, then a strain rate sensitive material model could be included in the VFM

formulation to be identified. This may have to be implemented in the future for certain materials as required.

Strength Identification

In order to identify the strength, there are three separate issues to address: (1) determining the location of failure; (2) determining the time of failure; and (3) deriving the failure stress at the time of fracture. The first problem is addressed by analysis of the raw (i.e. unsmoothed) strain maps throughout the test. Fig. 23 shows the raw strain map for specimen 4 once a macro crack has formed. Here the crack is clearly shown as an area of (non-physical) strain concentration.

The second and third issues present some interesting challenges when using an ultra-high speed camera with a low pixel resolution. It is possible that the material fails and a crack starts to open before it is directly observable on the strain maps. Further, it is also possible that the material will damage or plastically deform before a macro crack forms. Given that plasticity in polymers tends to be suppressed with

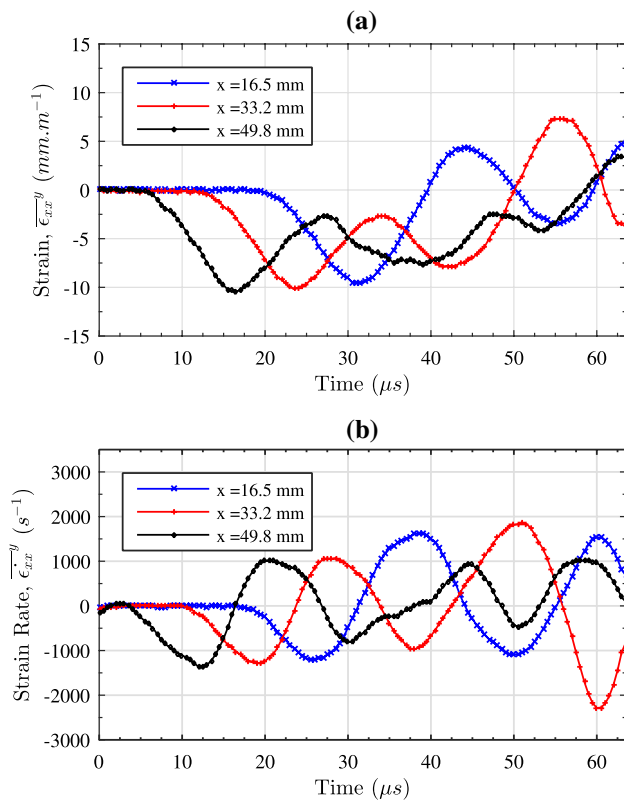


Fig. 22 Width averaged experimental data for specimen 4. **a** Strain $\overline{\epsilon_{xx}^y}$ and **b** strain rate $\overline{\dot{\epsilon}_{xx}^y}$ history corresponding to the stress-strain curves in Fig. 18a

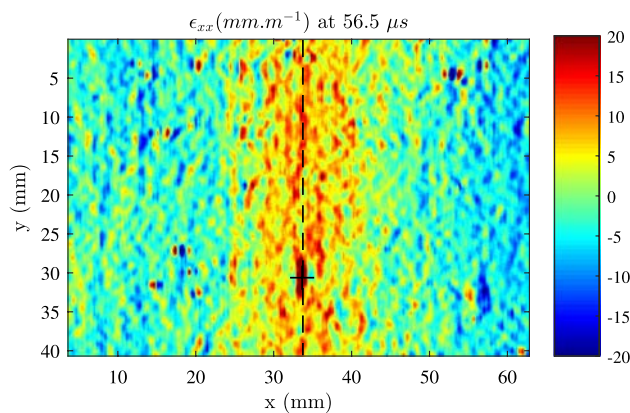


Fig. 23 Raw unsmoothed strain map (ϵ_{xx}) from the experimental data of specimen 4 showing the fracture location. (Color figure online)

increasing strain rate it is more likely that a damage process is occurring [6]. Therefore, accurately determining the failure time becomes challenging. It is possible to partially resolve the problem of identifying the time of failure by using two different types of stress information. Specifically, the mean stress calculated from the acceleration and the stress calculated from the identified material model can be

compared at the fracture location. When the material starts to form damage, the stress calculated using the strains will diverge from the stress calculated using the accelerations. The point at which these two measures diverge can be used to identify the the point of damage onset (or plasticity) in the material. After a critical amount of damage is formed, a macrocrack will nucleate and the material will fracture. Here, the fracture was assumed to occur at the maximum tensile stress measured using the acceleration after the two different methods of estimating the stress have diverged. The fracture time is then the frame at which the maximum stress occurs.

In order to compare both stress measures in the experimental data, the mean stress over a virtual gauge area of 11 (x) by 21 (y) pixels was used. The spatial average over the virtual gauge area will be denoted using a superscript A following the overline notation, e.g. $\overline{\sigma_{xx}^A}$ (LSG) denotes the average stress in the virtual gauge area using the linear stress-gauge equation (eq. 38). This gauge area is shown on the stress maps in Fig. 24a, b. For the purpose of calculating the stress from the strains, the average elastic modulus from the stress-gauge and special optimised virtual field was used. Note that the individually identified elastic modulus was used for each specimen. The divergence of the two stress measures is shown clearly in Fig. 24c, d. From Fig. 24c, the two stress measures begin to diverge just after 53.0 μ s and the maximum stress calculated from the acceleration occurs at 55.5 μ s. The divergence of these measures is shown as non-linear behaviour in the stress-strain curve over the virtual gauge region in Fig. 24d. This non-linear behaviour prior to failure suggests that there may be plastic deformation or a damage process in the composite prior to the formation of a macro crack. However, the stress being derived from acceleration, this apparent softening could also come from temporal ‘leakage’ from the crack as the acceleration is temporally smoothed. This will need to be investigated further in the future using image deformation simulations. Using the average stress in the virtual gauge area from the linear stress gauge equation, the failure strength was identified to be 52 MPa (this point is highlighted using a red cross in Fig. 24d).

It was possible to identify the transverse tensile strength for specimens 1–5. The identified strength for each of these specimens is summarised in Table 7. For comparison purposes, the strength derived using the standard stress-gauge equation is also provided in Table 7. It can be seen that both values are very consistent for specimens 3–5, but differ significantly for specimens 1 and 2. This shows that the LSG approach provides a more flexible stress measure to accommodate for some stress non-uniformity in the lateral direction. The strength values obtained through the LSG approach are very consistent and present a comparable COV to the quasi-static results. This is certainly the result

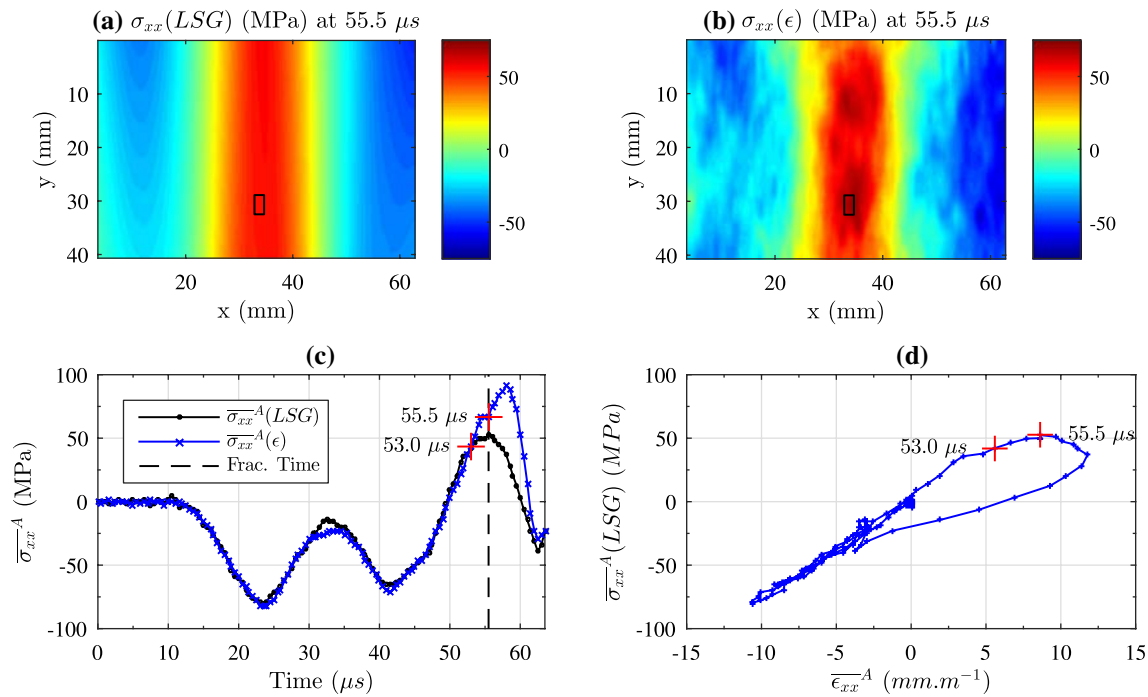


Fig. 24 Strength identification diagnostics for the experiment data from specimen 4. **a** Approximate stress field at the identified fracture time using the linear stress-gauge equation. **b** Approximate stress field at the identified fracture time using the strain and constitutive law. The virtual gauge area is shown as a black rectangle in **a**, **b**. **c**

Average stress in the virtual gauge area using the linear stress-gauge equation $\bar{\sigma}_{xx}^A(LSG)$ and stress calculated from the strains $\bar{\sigma}_{xx}^A(\epsilon)$. **d** Stress-strain curve over the virtual gauge area at the fracture location. (Color figure online)

of the very neat and consistent way of loading the material in tension through the waves while the quasi-static results will be more affected by possible specimen bending in the test machine. Additionally, the comparison of the average stress in the gauge area calculated using the acceleration $\bar{\sigma}_{xx}^A(LSG)$ and the strains $\bar{\sigma}_{xx}^A(\epsilon)$ is shown in Fig. 25. For specimens 1–5 both stress metrics are reasonably consistent up to the point of fracture.

Unfortunately, it was not possible to identify the transverse tensile strength for specimen 6. The reason for this was due to a significant inconsistency between the stress distribution predicted using the stress-gauge equation and the stress distribution reconstructed using the identified constitutive law. This inconsistency was only evident after the wave had reflected. Note that this does not affect the identified elastic modulus for this specimen as the modulus identification procedure only uses the portion of the test prior to the wave reflecting and becoming tensile (i.e. the portion of the test where both measures of stress are consistent). It is unclear why this divergence occurred. However, it is suspected that through-thickness dispersion following the wave reflection may be responsible. This could be due to slight inconsistencies in machined specimen geometry (e.g. non-square edges through-thickness). This suggests that specimen machining procedures are critical for this type of test. To confirm that

through-thickness dispersion is responsible for these inconsistencies, back-to-back imaging of the sample is required. Future experiments are planned to assess this phenomenon.

Table 7 Identified transverse tensile strength for all experimental specimens

Specimen	Strength $\bar{\sigma}_{xx}^F(SG)$ (MPa)	Strength $\bar{\sigma}_{xx}^F(LSG)$ (MPa)
1	27	55
2	46	69
3	58	58
4	53	52
5	51	54
6	–	–
Mean	47	58
SD	12	6
COV (%)	25	11
% Diff. to QS	+ 28	+57

The strength has been calculated using the stress-gauge equation ($\bar{\sigma}_{xx}^F(SG)$) and the linear stress-gauge equation ($\bar{\sigma}_{xx}^F(LSG)$). The percentage difference with respect to the quasi-static (QS) value is also provided for comparison

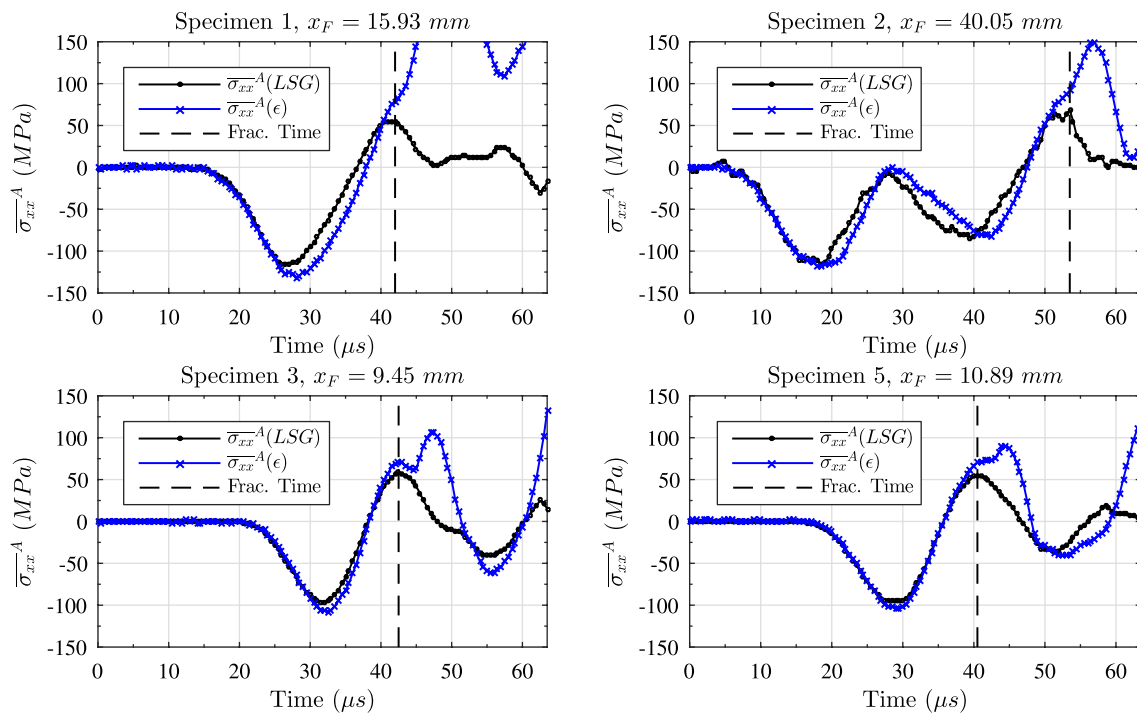


Fig. 25 Comparison of the average stress in the gauge area calculated using the acceleration $\sigma_{xx}^A(LSG)$ and the strains $\sigma_{xx}^A(\epsilon)$ for all remaining experimental specimens. The fracture time for each speci-

men is shown with a black dashed line and the axial location of fracture x_F is given in the header of each sub-plot. Note that specimen 4 is shown in Fig. 24c

Table 8 Peak mean tensile (tens.) and compressive (comp.) strain rate over the virtual gauge area ($\overline{\epsilon_{xx}^A}$) at the fracture location for each experimental specimen

Specimen	Peak tens. $\overline{\epsilon_{xx}^A}$ (s^{-1})	Peak comp. $\overline{\epsilon_{xx}^A}$ (s^{-1})
1	2700	– 1900
2	2900	– 1800
3	3300	– 1800
4	2000	– 1200
5	2900	– 1800
6	–	–

The limitation of through-thickness heterogeneity is discussed further in the "Future Work and Limitations" section.

In order to assess the strain rate associated with the measured strength of each specimen, the peak average strain rate over the virtual gauge area was extracted. The peak tensile and compressive average strain rate is provided for each specimen in Table 8. The strain rate history for the virtual gauge area on specimen 4 is provided in Fig. 26. The peak strain rates observed at each fracture location are similar in magnitude to the maximum peak strain rates in Table 6. All specimens experience a peak tensile strain rate on the order of $2000\ s^{-1}$.

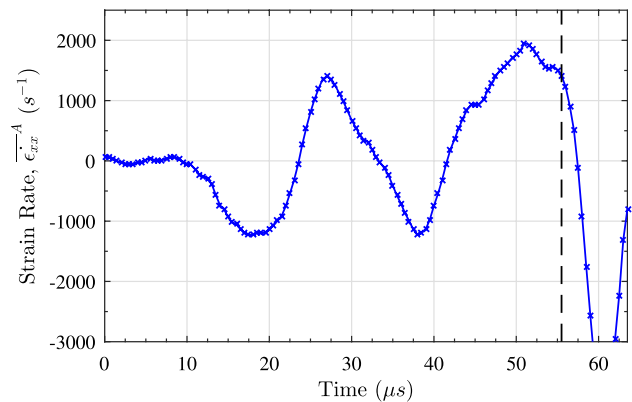


Fig. 26 Average strain rate history in the virtual gauge area (fracture location) for specimen 4

The results in Table 7 show that there is a significant increase in the strength when compared to the quasi-static value (57%). This increase in strength is expected as the transverse properties of a unidirectional composite are dominated by the matrix material [6, 7]. A previous study by Melin and Asp [1] showed a 37% increase in the transverse tensile strength at a strain rate of $800\ s^{-1}$. While a study by Gilat et al. [2] showed a 33% increase in the transverse tensile strength at a strain rate of $400\ s^{-1}$. These results contrast

with the results of Kwon et al. which predict a 65% increase in the transverse tensile strength at a strain rate of 100 s^{-1} . The results of the present study fall within the bounds of previous results. This variability in results with previous studies is likely to come from difference in polymer matrix chemistry, as previously mentioned.

The accuracy of the identified strength is dependent on a multitude of factors arising from the imaging system and the data processing method. The key parameters associated with the imaging system come from the spatial and temporal resolutions of the camera. A higher spatial resolution will allow for greater strain resolution and more precise identification of the fracture strain and location. A higher temporal resolution will allow for more accurate determination of the fracture time and fracture stress. For the present series of experiments, the frame rate was set to 2 MHz such that a full reflection of the wave could be obtained. For all specimens in this study, it was observed that both measures of stress diverged significantly within a single multiple of the inter frame time. Thus, in future studies, it would be desirable to scale down the specimen such that a full reflection could be imaged while recording at a frame rate of 5 MHz. This would allow for greater temporal resolution for identifying the fracture time and allow for further temporal information to be obtained regarding the fracture processes in the material.

Future Work and Limitations

The results of this study have successfully demonstrated the feasibility of an IBII test for measuring the transverse properties of composites at high strain rates. However, there are a number of limitations that need to be discussed in order to better understand the scope of the results and the potential applications of the technique described here. These limitations are listed below.

Through-Thickness Heterogeneity

For all identification procedures used in this study, it is assumed that the stress and acceleration fields are uniform through the thickness. Any deviations from through-thickness uniformity will generate errors in the identified properties. Through-thickness dispersion was the suspected cause for the inconsistency between stress and strain for specimen 6 after the wave had reflected. This led to an inability to reliably identify the strength for this specimen. Future experiments need to be conducted using two cameras in a back-to-back configuration to assess the extent of any errors generated from through-thickness dispersion. However, as five of the six specimens tested did not exhibit this

behaviour, it is unlikely that this is a significant limitation of this new test method.

Stress Field Approximation

This limitation refers to the use of the linear stress-gauge equation to approximate the stress at a particular location on the specimen for strength identification. Indeed, the linear stress gauge approximation is only appropriate if the actual stress distribution is linear across the specimen width at the considered cross-section. This was qualitatively verified for each specimen by comparing the stress calculated with the constitutive law with the stress calculated from the acceleration (see Fig. 25). Additionally, there are issues associated with temporal leakage effects in the acceleration. Any temporal leakage in the acceleration will be present in the derived stress information. The overall contribution of these temporal smearing effects is currently unknown but it is tractable using image deformation simulations.

With these limitations in mind, it is useful to consider the future development and potential applications of the IBII test presented in this work.

Measurement Error Characterisation

The use of image deformation simulations has been consistently mentioned as a method to assess the effects of camera noise, spatial and temporal resolutions. The accuracy of the identified properties are dependent on spatial and temporal resolutions of the camera, as well as the selected smoothing parameters. In order to fully characterise the errors arising from measurement resolution and noise, it is necessary to conduct image deformation simulations similar to refs. [31, 32]. Image deformation allows for the characterisation of random errors arising from noise as well as the systematic errors arising from the camera temporal and spatial resolutions. This will also allow for the effects of temporal leakage on the strength identification to be quantified. The authors are currently developing a software image deformation pipeline for this purpose. This image deformation pipeline will be coupled with explicit dynamics models that include cohesive surfaces to assess the strength identification procedure described previously.

Strain Rate Control

In this work, strain rate was not considered as part of the test design criteria. Obviously, it would be desirable to control the peak strain rate such that a wider range of strain rates could be tested. This will require further finite element simulations to understand which test configuration variables have most influence on the strain rate. It is expected that the speed and stiffness of the impactor are the two main drivers.

Application to Brittle Materials

This test method can be applied to any material that has a compressive strength significantly higher than its tensile strength. Brittle materials that fall into this category include concrete, glass, graphite and ceramics. These materials are notoriously difficult to test at high strain rates for reasons that are similar to transverse testing of unidirectional composites (i.e. small strains to failure leading to corrupting inertial effects). The IBII test presented in this work has already been successfully applied to test the high strain rate properties of concrete in ref. [13], and has significant potential to provide new understanding of the high strain rate properties of other brittle materials.

Shear/Tension and Through-Thickness Testing

The present study focused on the in-plane transverse properties of a uni-directional composite. The current methodology can be modified to analyse coupled shear/tension testing by using samples cut at various angles to the fibre direction. Another extension of the present method would be to analyse the through-thickness tensile strength of thick composite plates using a scaled-down test sample.

In summary, the design space for the IBII test proposed in this study is significant. The ultra-high speed camera technology that is the foundation of this work will only improve in the future. Increases in both spatial and temporal resolutions of ultra-high speed cameras will lead to reduced measurement errors and increased efficacy of the test method proposed herein. It is anticipated that this technique could become a new standard test in the future.

Conclusion

This work presents a new methodology for testing the transverse properties of fibre composites at high strain rates. The type of test proposed is an image-based inertial impact (IBII) test. This test uses a short input compressive pulse which reflects and becomes tensile causing failure of the test sample. Throughout the test, ultra-high speed full-field measurements are taken and the concept of acceleration as a load-cell is used to reconstruct stress information in the specimen. The IBII test was successfully used to identify the transverse elastic modulus and tensile strength of a carbon fibre composite at peak strain rates on the order of $2000\text{--}3000\text{ s}^{-1}$. The main findings of this study can be summarised as follows.

1. Explicit dynamics simulations are a reliable tool for designing this IBII test giving suitable design predictions for the required test configuration.
2. It is possible to cause transverse tensile failure in a composite material with a relatively low impact speed, on the order of $30\text{ to }35\text{ ms}^{-1}$.
3. Measurement quality with current ultra-high speed camera technology is sufficient to allow for both stiffness and strength identification in the same test.
4. The average transverse elastic modulus of the tested composite material was identified as 7.9 GPa using the stress-gauge equation and 7.8 GPa using the optimised virtual fields at a peak average strain rate on the order of 2000 s^{-1} . This represents an 8% increase compared to the quasi-static value.
5. The average transverse tensile strength was identified as 58 MPa at a strain rate on the order of $2000\text{--}3000\text{ s}^{-1}$. This represents a 57% increase compared to the quasi-static value.

Work is currently under way to extend the IBII method to combined shear-tension/compression testing as well as to the through-thickness properties of fibre composites. This new methodology can be applied to any brittle material that has a compressive strength greater than its tensile strength. Application of this new technique to brittle materials such as glass, concrete and ceramics will lead to experimental data that has previously not been available for these materials at high strain rates. Test methods based on full-field measurement have significant potential for high strain rate testing. Future advances in imaging technology will allow for the production of experimental data that was not previously attainable.

Acknowledgements This material is based on research sponsored by the Air Force Research Laboratory, under Agreement Number FA9550-17-1-0133. The authors are grateful to the grant programme manager, Dr David Garner from EOARD/AFOSR. The U.S. Government is authorized to reproduce and distribute reprints for Governmental purposes notwithstanding any copyright notation thereon. The views and conclusions contained herein are those of the authors and should not be interpreted as necessarily representing the official policies or endorsements, either expressed or implied, of the Air Force Research Laboratory or the U.S. Government. Dr Lloyd Fletcher and Prof. Fabrice Pierron acknowledge support from EPSRC through Grant EP/L026910/1.

Compliance with Ethical Standards

Data Provision All data supporting this study are openly available from the University of Southampton repository at: <https://doi.org/10.5258/SOTON/D0608>.

Open Access This article is distributed under the terms of the Creative Commons Attribution 4.0 International License (<http://creativecommons.org/licenses/by/4.0/>), which permits unrestricted use, distribution, and reproduction in any medium, provided you give appropriate credit to the original author(s) and the source, provide a link to the Creative Commons license, and indicate if changes were made.

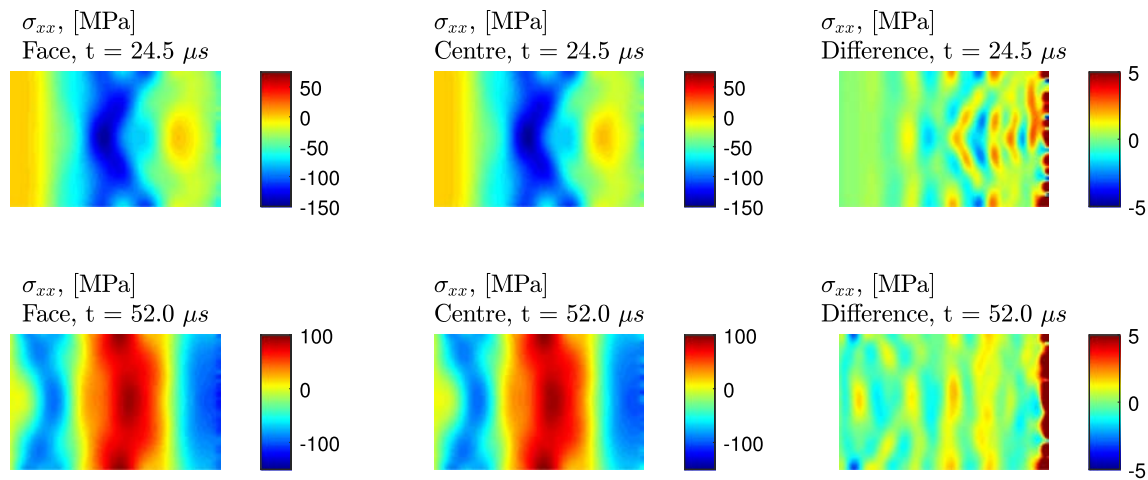


Fig. 27 Comparison of the in-plane stress component σ_{xx} at the outer face and centre plane of the specimen for the 3D finite element model. The difference between the fields shows that there is minimal

variation through-thickness for most of the surface of the specimen. (Color figure online)

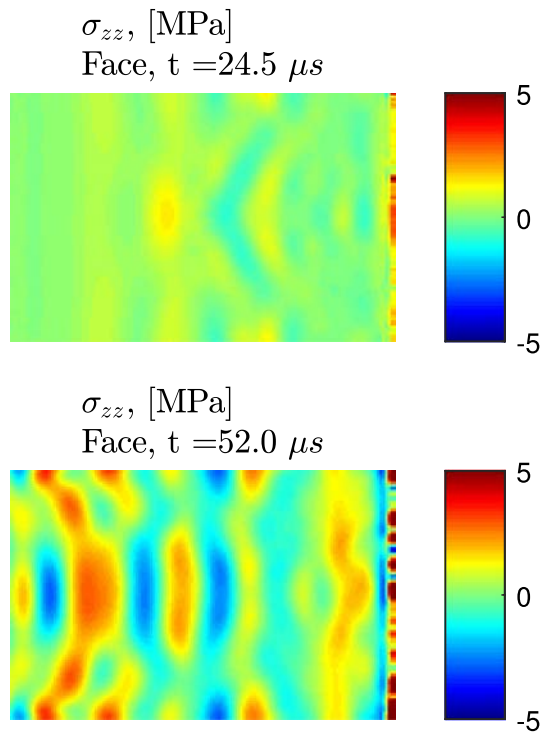


Fig. 28 Out of plane stress component σ_{zz} for the outer face of the 3D finite element model. As the out of plane component is small compared to the in-plane stresses (see Fig. 27) the plane stress assumption is reasonably satisfied. (Color figure online)

Appendix

The purpose of this section is to verify the plane stress assumption and to assess the uniformity of the kinematic fields through-thickness using the 3D finite element model

described in the "Finite Element Implementation" section. Here, the parameters of the finite element model were set to replicate the selected design case: waveguide length of $L_{wg} = 50$ mm, projectile length of $L_p = 25$ mm and projectile velocity of at least $V_p = 30$ m/s.

The in-plane stress component σ_{xx} was extracted for the outer face and central plane of specimen. A comparison of the stress fields at two different time points in the simulations is shown in Fig. 27. These time points were selected as the first corresponds to a point at which the loading pulse has mostly entered the specimen and the second corresponds to time point after the wave has reflected off the free edge. Analysing the difference of the two fields shows that it is small apart from near the impact edge. This difference likely arises from three dimensional effects where the sharp edge of the specimen is in contact with the flat surface of the waveguide. However, this difference quickly disappears within a few elements of the contact surface.

The out of plane stress component σ_{zz} was also extracted from the surface of the specimen. The out of plane stress for the surface of the specimen is shown in Fig. 28. Comparing Fig. 28 to Fig. 27 shows that the out of plane stress is generally much less than the in-plane stress. This demonstrates that the assumptions of plane stress and through-thickness uniformity are reasonable.

References

1. Melin LG, Asp LE (1999) Effects of strain rate on transverse tension properties of a carbon/epoxy composite: studied by moiré photography. *Compos A Appl Sci Manuf* 30(3):305–316

2. Gilat Amos, Goldberg Robert K, Roberts Gary D (2002) Experimental study of strain-rate-dependent behavior of carbon/epoxy composite. *Compos Sci Technol* 62(10–11):1469–1476
3. Taniguchi Norihiko, Nishiwaki Tsuyoshi, Kawada Hiroyuki (2007) Tensile strength of unidirectional CFRP laminate under high strain rate. *Adv Compo Mater* 16(2):167–180
4. Daniel IM, Werner BT, Fenner JS (2011) Strain-rate-dependent failure criteria for composites. *Compos Sci Technol* 71(3):357–364
5. Koerber H, Xavier J, Camanho PP (2010) High strain rate characterisation of unidirectional carbon-epoxy IM7-8552 in transverse compression and in-plane shear using digital image correlation. *Mech Mater* 42(11):1004–1019
6. Gerlach Robert, Siviour Clive R, Petrinic Nik, Wiegand Jens (2008) Experimental characterisation and constitutive modelling of RTM-6 resin under impact loading. *Polymer* 49(11):2728–2737
7. Jordan Jennifer L, Foley Jason R, Siviour Clive R (2008) Mechanical properties of Epon 826/DEA epoxy. *Mech Time-Depend Mater* 12(3):249–272
8. Gama Bazle A, Lopatnikov Sergey L, Gillespie Jr, John W (2004) Hopkinson bar experimental technique: a critical review. *Appl Mech Rev* 57(4):223–250
9. Lopatnikov SL, Gama BA, Krauthouser K, Gillespie G (2004) Applicability of the classical analysis of experiments with split Hopkins pressure bar. *Tech Phys Lett* 30(2):102–105
10. Sutton MA, Gilat A, Seidt J, Rajan S, Kidane A (2018) Full field deformation measurements in Tensile Kolsky Bar experiments: studies and detailed analysis of the early time history. *J Dyn Behav Mater* 4(1):95–113
11. Moulart R, Pierron F, Hallett SR, Wisnom MR (2010) Full-field strain measurement and identification of composites moduli at high strain rate with the virtual fields method. *Exp Mech* 51(4):509–536
12. Pierron F, Zhu H, Siviour C (2014) Beyond Hopkinson's bar. *Phil Trans R Soc A* 372(2023):20130195
13. Pierron F, Forquin P (2012) Ultra-high-speed full-field deformation measurements on concrete spalling specimens and stiffness identification with the virtual fields method. *Strain* 48(5):388–405
14. Avril S, Bonnet M, Bretelle A-S, Grédiac M, Hild F, Lenny P, Latourte F, Lemosse D, Pagano S, Pagnacco E, Pierron F (2008) Overview of identification methods of mechanical parameters based on full-field measurements. *Exp Mech* 48(4):381
15. Pierron Fabrice, Grédiac Michel (2012) The virtual fields method: extracting constitutive mechanical parameters from full-field deformation measurements. Springer, New York
16. Yoon S-H, Winters M, Siviour CR (2016) High strain-rate tensile characterization of EPDM rubber using non-equilibrium loading and the virtual fields method. *Exp Mech* 56(1):25–35
17. Koohbor Behrad, Kidane Addis, Wei-Yang Lu, Sutton Michael A (2016) Investigation of the dynamic stress-strain response of compressible polymeric foam using a non-parametric analysis. *Int J Impact Eng* 91:170–182
18. Reu PL, Miller TJ (2008) The application of high-speed digital image correlation. *J Strain Anal Eng Des* 43(8):673–688
19. Zhu Haibin (March 2015) *A novel methodology for high strain rate testing using full-field measurements and the virtual fields methods*. PhD, University of Troyes, France
20. Avril S, Grédiac M, Pierron F (2004) Sensitivity of the virtual fields method to noisy data. *Comput Mech* 34(6):439–452
21. Avril S, Pierron F (2007) General framework for the identification of constitutive parameters from full-field measurements in linear elasticity. *Int J Solids Struct* 44(14):4978–5002
22. Zhu H, Pierron F (2016) Exploration of Saint-Venant's principle in inertial high strain rate testing of materials. *Exp Mech* 56(1):3–23
23. Hsiao HM, Daniel IM, Cordes RD (1999) Strain rate effects on the transverse compressive and shear behavior of unidirectional composites. *J Compos Mater* 33(17):1620–1642
24. International Organization for Standardization (2009) *ISO 527-5:2009 - Plastics – Determination of tensile properties – Part 5: Test conditions for unidirectional fibre-reinforced plastic composites*. International Organization for Standardization, Geneva, Switzerland
25. Grédiac M, Blaysat B, Sur F (2017) A critical comparison of some metrological parameters characterizing local digital image correlation and grid method. *Exp Mech* 57(6):871–903
26. Grédiac M, Sur F, Blaysat B (2016) The grid method for in-plane displacement and strain measurement: a review and analysis. *Strain* 52(3):205–243
27. Piro J-L, Grédiac M (2004) Producing and transferring low-spatial-frequency grids for measuring displacement fields with moiré and grid methods. *Exp Tech* 28(4):23–26
28. Pierron Fabrice, Cheriguene Rachid, Forquin Pascal, Moulart Raphael, Rossi Marco, Sutton M A (2011) Performances and limitations of three ultra high-speed imaging cameras for full-field deformation measurements. *Appl Mech Mater* 70:81–86
29. Rossi Marco, Pierron Fabrice, Forquin Pascal (2014) Assessment of the metrological performance of an in situ storage image sensor ultra-high speed camera for full-field deformation measurements. *Meas Sci Technol* 25(2):025401
30. Herráez Miguel Arevallilo, Burton David R, Lalor Michael J, Gdeisat Munther A (2002) Fast two-dimensional phase-unwrapping algorithm based on sorting by reliability following a noncontinuous path. *Appl Opt* 41(35):7437–7444
31. Rossi Marco, Pierron Fabrice (2012) On the use of simulated experiments in designing tests for material characterization from full-field measurements. *Int J Solids Struct* 49(3):420–435
32. Lukić Bratislav, Saletti Dominique, Forquin Pascal (2017) Use of simulated experiments for material characterization of brittle materials subjected to high strain rate dynamic tension. *Phil Trans R Soc A* 375(2085):20160168
33. Lava P, Cooreman S, Coppieters S, De Strycker M, Debruyne D (2009) Assessment of measuring errors in DIC using deformation fields generated by plastic FEA. *Opt Lasers Eng* 47(7):747–753
34. Lava P, Cooreman S, Debruyne D (2010) Study of systematic errors in strain fields obtained via DIC using heterogeneous deformation generated by plastic FEA. *Opt Lasers Eng* 48(4):457–468
35. Rossi M, Lava P, Pierron F, Debruyne D, Sasso M (2015) Effect of DIC spatial resolution, noise and interpolation error on identification results with the VFM. *Strain* 51(3):206–222
36. Kwon Junbeom, Choi Jaeyoung, Huh Hoon, Lee Jungju (2017) Evaluation of the effect of the strain rate on the tensile properties of carbon-epoxy composite laminates. *J Compos Mater* 51(22):3197–3210