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High Strain-Rate Compression Behavior of Polymeric Rod and Plate Kelvin Lattice Structures

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Abstract

The compressive high strain-rate behavior of polymeric Kelvin lattice structures with rod-based or plate-based unit cells was investigated through experimental techniques and finite element simulations. Polymeric lattice structures with 5x5x5 unit cell geometries were manufactured on the millimeter scale using vat polymerization additive manufacturing and tested at low (0.001/s) and high (1000/s) strain-rates. High strain-rate experiments were performed and validated for a viscoelastic split-Hopkinson (Kolsky) pressure bar system (SHPB) coupled with high-speed imaging and digital image correlation (DIC). Experimental results at both low and high strain-rates show the formation of a localized deformation band which was more prevalent in low relative density specimens and low strain-rate experiments. Strain-rate effects of lattice specimens strongly correlate with effects of the base polymer material; both bulk polymer and lattice specimen demonstrated strain-rate hardening, strain-rate stiffening, and decreased fracture strain under dynamic loading. Results show mechanical failure properties and energy absorption depended strongly on the relative density of the lattice specimen and exhibited distinct scaling between relative density and geometry type (rod, plate) and loading rate. High relative density plate-lattices demonstrated inferior mechanical properties to rod-lattices; however, there exists a critical relative density for a given mechanical property (17%-28%) below which plate-lattices outperform rod-lattices of similar mass. High strain-rate explicit finite element simulations were performed and showed good agreement with the mechanical failure trends and deformation modes observed in the experiments.

Keywords: Lattice materials, Additive manufacturing, Plate/Rod-lattices, Strength, Failure

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1. INTRODUCTION

Cellular materials demonstrate excellent mechanical properties for energy absorption and impact mitigation due to their ability to undergo large deformations at low internal forces (Gibson & Ashby, 1997). Historically, stochastic foams and 2D honeycomb structures have been widely investigated and optimized for these applications (Sun & Li, 2018). Recent improvements in additive manufacturing techniques have allowed for the expansion of cellular material design space through precise control of microstructure in polymeric and metallic materials. As a result, it is now feasible to easily manufacture *lattice structures*, a class of periodic cellular materials composed of a connected network of rods, plates, or shells, on the nano-, micro-, and millimeter length scales (Tofail et al., 2018; Zhang et al., 2020).

Lattice structures are commonly characterized through geometry, unit cell deformation mechanism, and relative density (ρ^*/ρ) which is defined as the ratio of lattice mass density (ρ^*) to base material mass density (ρ) and is representative of the volume fraction occupied by solid material (Ashby, 2006). Lattice structures such as rod-lattices, plate-lattices, or shell-lattices may demonstrate *stretching-dominated* or *bending-dominated* behavior in unit cell members (Deshpande et al., 2001; Al-Ketan et al., 2018). Stretching-dominated structures have been shown to exhibit about ten times the stiffness and three times the strength of bending-dominated structures such as stochastic foams at $\rho^*/\rho = 0.1$ (Deshpande et al., 2001). Plate-lattices have been shown through simulation and experiment to outperform rod-lattices of equal mass (Tancogne-Dejean et al., 2018) and geometries such as the octet-cubic plate-lattice have been shown computationally to attain the theoretical Hashin-Shtrikman (H-S) (Hashin & Shtrikman, 1963) upper bounds on isotropic elastic stiffness (Berger et al., 2017). This result has been supported by experiments with pyrolytic carbon octet-cubic nanolattices which have demonstrated quasi-static stiffnesses and strengths close to the H-S and Suquet upper bounds (Crook et al., 2020). Smooth-shell lattices have also been shown to outperform particular rod-lattices

of equal mass (Bonatti & Mohr, 2019) and experiments on milliscale metallic lattices of varying topology show superior mechanical performance in sheet-triply periodic minimal surface (TPMS)-based geometries than rod- or skeletal-TPMS based geometries (Al-Ketan et al., 2018). Rod-, plate-, and shell-lattices also appear in nature through biological cellular materials. Cancellous bone, toucan beak, and woodpecker cranial skull all exhibit a combination of rod, plate and shell structures (Mosekilde, 1990; Seki et al., 2005; Wang et al., 2011).

Lattice structure mechanics inherently depend upon lattice topology and base material behavior. Many studies have taken advantage of material size effects on the nanoscale to attain high specific-strengths (Schaedler et al., 2011; Bauer et al., 2014) and flaw insensitivity (Montemayor et al., 2016; Zhang et al., 2019). Geometric effects from node geometry (Portela et al., 2018) and manufacturing imperfections such as strut waviness or thickness variation (Liu et al., 2017) also affect the mechanical properties. Large geometric defects such as missing struts severely weaken the mechanical response of lattice structures and demonstrate different strength and stiffness scaling properties than ideal geometries (Meza et al., 2017).

Dynamic experiments on lattice structures have shown lattice strain-rate behavior is material dependent. Both strain-rate-strengthening and weakening effects have been observed in polymeric octet-truss lattice specimens depending on the base material used (Ling et al., 2019). Experimental work by Tancogne-Dejean et al. suggests that the dynamic-strengthening effect present in stainless steel 316L octet-truss lattice specimen is attributed to the strain-rate effects of the base material (Tancogne-Dejean et al., 2016). Hazeli et al. report dynamic experiments on metallic heat-treated lattice structures that exhibit strain-rate-hardening and similar deformation trends at low and high strain-rates for octet-truss, rhombic dodecahedron, and dode-medium unit cells (Hazeli et al., 2015). Experiments on polymeric plate-Kelvin lattices have shown the position of deformation bands can be dependent on strain-rate, with high strain-rate deformation occurring at lattice edges and low strain-rate deformation occurring in the middle of the specimen (Duan et al., 2019). The dynamic response of cellular

materials spans two regimes: a transitional dynamic regime and a shock regime. The transitional dynamic regime encompasses dynamic behavior effects prior to shock-formation at some critical impact velocity (Sun & Li, 2018). This paper will focus on the transitional dynamic regime of lattice structures, however the shock regime has previously been studied and remains an area of active research (Hawreliak et al., 2016; Messner et al., 2017; Lind et al., 2019; Portela et al., 2021).

The behavior of various rod- and plate-lattices has been individually explored, but not actively compared for a single unit cell geometry under dynamic loading. Low-velocity impact experiments have been carried out on polymeric plate-lattices (Andrew et al., 2021) and high strain-rate experiments have been carried out on a single unit cell of a metallic plate-lattice (Tancogne-Dejean et al., 2019), but high strain-rate experimental characterization of full plate-lattices is lacking. In our work, we perform high strain-rate and low strain-rate experiments and simulations on polymeric rod- and plate-lattices with a 5x5x5 Kelvin unit cell geometry. Lattice specimens are manufactured using digital light processing (DLP) vat polymerization and loaded at low strain-rates using a servo-hydraulic machine and at high strain-rates using a polycarbonate split-Hopkinson (Kolsky) pressure bar. The mechanical strength and failure properties are extracted and the dynamic effects on lattice specimen are studied. Finally, finite element simulations performed with Abaqus/Explicit are used to complement experiments, which show good qualitative agreement with high strain-rate results.

2. MATERIALS AND METHODS

2.1. Specimen Design and Manufacturing

Lattice specimens were designed using rod and plate structures with a Kelvin unit cell (Sullivan et al., 2008). Rod (R) geometries were designed by placing rod structures along the edges of a unit cell and plate (P) geometries were designed by placing plate structures along the faces of a unit cell. Computer Aided

Design (CAD) models were constructed in SolidWorks (Dassault Systems) using a 5x5x5 Kelvin unit cell geometry composed of rod or plate structures with relative densities of 15%, 20%, 25%, and 30%. Relative density was calculated in SolidWorks by taking the volume fraction of each unit cell specimen within a space-filling bounding box. Unit cell lengths were kept constant and the thickness of the structural elements was varied to obtain different relative densities. Small holes were placed in the plate specimen to avoid manufacturing constraints of closed-cell structures. Baseplates were attached on two opposite faces to improve manufacturability and ensure planar loading during testing. Table 1 shows the CAD dimensions for each of the lattice unit cell specimen and the calculated relative densities.

Specimen		L (mm)	t (mm)	ρ^*/ρ (-)	m_{CAD} (g)	m_{exp} (g)
Rod (R)		1.10	0.51	0.15	1.48	1.49 ± 0.06
		1.10	0.63	0.20	1.80	1.80 ± 0.06
		1.10	0.75	0.25	2.16	2.16 ± 0.02
		1.10	0.89	0.30	2.58	2.57 ± 0.13
Plate (P)		1.25	0.18	0.15	1.64	1.64 ± 0.09
		1.25	0.25	0.20	1.93	1.92 ± 0.15
		1.25	0.34	0.25	2.20	2.22 ± 0.22
		1.25	0.43	0.30	2.45	2.50 ± 0.15

Table 1: Dimensions of Kelvin cell characteristic length L , structure thickness t , relative density ρ^*/ρ , as-designed mass m_{CAD} , and experimental mass m_{exp} for each lattice specimen. Small holes in plate specimen have diameters of 0.225 mm on square faces and 0.300mm on hexagonal faces.

Specimens (Figure 1) were manufactured using vat polymerization additive manufacturing (3D printing) with Digital Light Processing (DLP) technology. DLP 3D printers utilize a layer-by-layer manufacturing technique to cure liquid photoresin into solid photopolymer using UV light (Tofail et al., 2018). An Autodesk Ember DLP 3D printer was used with Colorado Photopolymer Solutions PR57-W photoresin. The printer uses a 405 nm wavelength light for 2.8 s exposure time per 25 μm layer. Figure 1b shows an example of the cross-sectional illumination during printing of a rod specimen. Manufacturing supports were attached to the baseplates to improve print success and later removed. The print build area allowed for two lattice specimens to be printed from one manu-

facturing run. After fabrication, specimens were rinsed in an isopropyl alcohol bath to remove any excess photoresin. The small holes in plate specimen (square faces - 0.225 mm diameter, hexagonal faces – 0.300 mm diameter) allow uncured photoresin to be removed from the interior of the unit cells. Lattice dimensions showed around 5% shrinkage from as-designed CAD values in the print direction and $< 5\%$ in lateral directions.

The density of the photopolymer was calculated as 1200kg/m^3 and used to approximate the as-designed masses of each specimen in SolidWorks. The as-designed and average experimental masses with standard deviation for each specimen are listed in Table 1. The small difference between masses and low shrinkage demonstrate that the resolution of the printer can resolve the as-designed dimensions and indicate the experimental relative densities are consistent with the as-designed values.

Figure 1: Design and manufacture of rod and plate lattice specimen. (a) CAD model of R15 specimen and (b) cross-sectional illumination of DLP manufacturing technique for layer highlighted in (a). Manufactured photopolymer lattice specimen with (c) R15, (d) R30, (e) P15, and (f) P30 geometries.

In addition to the lattice structures, bulk photopolymer specimens were printed for mechanical characterization of the base material. Cylindrical specimens with 10 mm diameter and 5 mm width were manufactured using the same print parameters as the lattice specimens.

2.2. Low Strain-Rate Experiments

Low strain-rate experiments were performed using a servo-hydraulic actuator machine (MTS Model 358.10) with a 13.3 kN axial load capacity. Lattice specimens were compressed at a rate of 1 mm/min and bulk photopolymer specimen were compressed at a rate of 0.2 mm/min until the load capacity of the machine was reached. These loading rates correspond to a nominal strain rate of $\dot{\epsilon} \approx 0.001/s$. A Fastec IL5 High-Speed Camera with a 100 mm Tokina AT-X Pro lens was used to take experimental images at 24 frames per second with a continuous light source (Technique FOI-150-UL). Images were resampled and analyzed at 1 fps to limit the amount of data collected from the long timescale of the experiment. 2D digital image correlation (DIC) software (Vic2D, Correlated Solutions, Columbia, SC) was used to determine the displacement of the hydraulic crosshead. A random speckle was placed on the crosshead and a subset size of 55 pixels with a step size of 5 pixels was used in the DIC analysis.

2.3. High Strain-Rate Experiments

High strain-rate experiments were performed using the split-Hopkinson (Kolsky) pressure bar (SHPB) technique (Ramesh, 2008). Figure 2 shows a general diagram of the SHPB set-up composed of an impactor, input, and output bar with strain gauge locations. A pressurized gas gun is used to propel the impactor into the input bar which generates a stress wave that compresses the sample between the input and output bars. An 7075 aluminum SHPB system was used for testing of bulk photopolymer specimen and a polycarbonate (PC) system was used for testing of lattice specimen. A Vishay 2310B signal conditioning amplifier and 2.5 GHz Tektronix DPO 3014 digital oscilloscope were used for both SHPB systems to record raw strain gauge voltage data.

A Shimadzu HPV-X2 camera with a 100 mm Tokina AT-X Pro lens was used to take high-speed images of the lattice during dynamic deformation. A Cavitar Cavilux Smart laser unit was used as a light source to send 20 ns incoherent laser pulses. A field of view of 55 mm x 32 mm (7 pixels/mm) was used to capture 128 images at 100,000 fps. 2D DIC code (Vic2D) was used to extract

the displacements and velocities of the input- and output-bar interfaces. A random speckle was applied to the ends of the input and output bars (Figure 2) and low-pass imaging with an average subset size of 35 pixels with a step size of 3 pixels was used in the DIC analysis.

Figure 2: Split-Hopkinson (Kolsky) pressure bar experimental set-up for high strain-rate experiments. Strains experienced by gauges are conditioned and recorded via oscilloscope and used to trigger high-speed camera image capture. An example of incident, reflected, and transmitted strain signals is shown.

2.3.1. 7075 Aluminum SHPB System

The Al 7075 SHPB system consisted of impactor (0.45 m), input (1.83 m), and output (1.83 m) bars of diameter 19.05 mm with Omega SGD-2D/350-LY11 strain gauges (1/4 bridge circuit) located at the midpoints of the bars. Bending effects are eliminated by taking the average of two strain gauges on diametrically opposing sides of the bar. Conventional one-wave SHPB analysis (Ramesh, 2008) was used to obtain the macroscopic stress-strain response of the test specimen. It is assumed excessive dispersion or attenuation does not occur and no dispersion correction is applied.

2.3.2. Polycarbonate SHPB System

The polycarbonate system consisted of impactor (0.46 m), input (1.83 m), and output (1.83 m) bars of diameter 24.50 mm with Vishay EA-06-031DE-350/LE strain gauges (1/4 bridge circuit) located at the midpoints of the bars, in a similar arrangement to the Al 7075 system. Due to poor heat conduction in the polycarbonate, the strain gauges were operated at a low input voltage (1.4 V) to avoid thermal drift in measurements. In a viscoelastic medium such as polycarbonate there is considerable dispersion and attenuation of stress waves.

Corrections of strain measurements are necessary to extract accurate specimen stress-strain response in a viscoelastic SHPB. A non-conventional SHPB analysis technique is thus required and is briefly described here.

The dispersion and attenuation effects inherent in a viscoelastic medium are corrected by applying a phase shift in the frequency domain to the strain measurements. Following the work of Bacon (Bacon, 1998): the velocities (v) and forces (F) (for some position x) in a viscoelastic SHPB can be written in the frequency domain in terms of angular frequency, $\omega = 2\pi f$, where f is frequency in Hz:

$$\hat{v}(x, \omega) = \frac{-i\omega}{\gamma} [\hat{P}(\omega)e^{-\gamma(\omega)x} + \hat{N}(\omega)e^{\gamma(\omega)x}] \quad (1)$$

$$\hat{F}(x, \omega) = \frac{-\rho A \omega^2}{\gamma^2} [\hat{P}(\omega)e^{-\gamma(\omega)x} + \hat{N}(\omega)e^{\gamma(\omega)x}] \quad (2)$$

$$\gamma(\omega) = \alpha(\omega) + i \frac{\omega}{c(\omega)} \quad (3)$$

$\hat{P}(\omega)$ and $\hat{N}(\omega)$ represent the Fourier transforms of the strains at $x = 0$ (corresponding to the measurement location) due to the waves traveling in the directions of increasing and decreasing x , respectively. $\gamma(\omega)$ is the wave propagation coefficient defined by attenuation coefficient $\alpha(\omega)$ and phase velocity $c(\omega)$. Velocity and force time histories in the bars are obtained using an Inverse Fourier Transform on Eqs. (1) and (2). The propagation coefficient is found using a one-point measurement experimental technique (Bacon, 1998) and the average propagation coefficient computed over 10 experiments is used in the analysis (see Appendix A).

We validate the polycarbonate SHPB by comparing bar-interface velocities computed using SHPB (i.e. Eq. (1)) and DIC techniques. Error in measurement is defined in time as the percent difference of the velocity measurements at the bar-interfaces and is propagated onto force measurements. Figure 3a compares the input bar and output bar interface velocities for a R20 specimen. There is good agreement of peak velocities within 5 – 8% on the input bar and

within 3 – 4% on the output bar over the course of loading. Considering force equilibrium, Figure 3b shows the incident and transmitted (left and right, respectively) force-time histories of rod specimen. We observe good agreement between the right-interface and left-interface forces for all specimen geometries. This indicates dynamic equilibrium is attained during high strain-rate loading experiments. An improvement in dynamic equilibrium is observed for higher relative density specimen. Additionally, plate specimen demonstrated similar force-time histories to rod specimen.

Figure 3: (a) SHPB-computed and DIC-computed interface velocities of a R20 specimen showing good agreement and (b) force time histories of rod specimens of various relative densities (15%, 20%, 25%, 30%) during loading. Error in measurement is defined as the percent difference between SHPB-computed and DIC-computed bar-interface velocities. This error is propagated onto the force measurements.

2.4. Numerical Simulations

Numerical simulations were carried out using Abaqus/Explicit (Dassault Systems, Providence, RI) finite element analysis. The simulation framework included a 5x5x5 lattice specimen geometry compressed between two rigid plates (Figure 4a). The bottom plate remained fixed while a constant velocity was applied to the top plate. The velocity was implemented as a smoothed amplitude step during the initial 10% of loading to decrease numerical oscillations. The “Brittle Cracking” Abaqus/Explicit material definition was used; this model assumes linear elastic behavior with damage from tensile cracking. This constitutive relation and the material parameters used in the numerical model were

chosen to match the qualitative macroscopic stress-strain response and deformation behavior in dynamic experiments. A critical stress σ_C is defined using values for stiffness, $E = 1000$ MPa (Young's modulus) and $\epsilon_C = 0.2$ (strain). A linear loss of strength is assumed from ϵ_C up to some $\epsilon_D = 0.3$ (strain) at which the element is removed from the simulation (Figure 4b). Mass density was experimentally determined and defined as 1200 kg/m^3 and Poisson's ratio was taken as $\nu = 0.35$, which is a typical value for polymers. A general contact algorithm was defined for self-contact of the specimen and surface-to-surface contact properties with no normal separation were defined for the rigid plate-lattice interactions. Tetrahedral meshing with a local mesh size of 0.20 mm and quadratic C3D10M elements with deletion was used. Mass scaling was used to increase the computational timestep which corresponded to $< 0.1\%$ percent change in mass for all specimens.

Figure 4: a) Simulation framework includes two rigid plates (defined by a reference point RP) and a 5x5x5 lattice specimen imported from SolidWorks. (b) Parameters used in “Brittle Cracking” material model and material response.

3. RESULTS AND DISCUSSION

3.1. Base Material Characterization

The mechanical behavior of lattice structures is known to depend on both the structural unit cell geometry and base material behavior (Schwaiger et al., 2019; Zhang et al., 2020). Therefore, strain-rate characterization of the base photopolymer material is necessary to understand the mechanical response of the lattice specimen. Low strain-rate experiments on bulk photopolymer specimen were performed with a servo-hydraulic MTS testing machine and DIC-computed displacements. High strain-rate experiments were performed using an Al 7075 SHPB system with one-wave analysis (Ramesh, 2008). Force equilibrium was observed during high strain-rate experiments and it is assumed the specimens were in dynamic equilibrium.

Figure 5 shows the compressive **true** stress-strain curves for the photopolymer at various average strain-rates. Strain-rate was calculated for each specimen using one-wave analysis and taken as the average value from the start of loading until unloading of the stress wave. Specimens experienced approximately uniform strain-rate for $\epsilon > 0.03$. All specimens demonstrated elastic, yielding, and strain-hardening material responses. Strain-rate-stiffening in modulus and strain-rate-hardening effects are observed. The dynamic enhancement factor D (defined as the ratio of the high strain-rate value over the corresponding low strain-rate value) for material strength for $\dot{\epsilon} = 0.001 - 1000/s$ is $D \approx 6$ at strain $\epsilon = 0.05$ and $D \approx 3.5$ at $\epsilon = 0.10$. Brodnik et al. have performed crack propagation experiments and simulations on a dyed form of the base photopolymer (PR-57 Black) and demonstrated brittle material behavior with a fracture toughness of $0.2MPa\sqrt{m}$ (Brodnik et al., 2020). High-speed experimental images revealed fracture of specimen during dynamic testing. Specimens that were observed to fracture in a typical brittle failure mode of axial splitting during compressive loading are marked in Figure 5. Fracture was observed during high strain-rate experiments with $\epsilon > 0.25$ and no fracture was observed during low strain-rate experiments.

Figure 5: Compressive **true** stress-strain response of the bulk photopolymer at various strain-rates.

3.2. Low Strain-Rate Experiments on Lattice Specimens

Low strain-rate experiments were performed on rod and plate specimen with relative densities of 15%, 20%, 25%, and 30% at a nominal strain rate of $\dot{\epsilon} \approx 0.001/s$. Each geometry was tested with at least two specimens. The engineering stress of the specimen was computed using force measurements from the load cell divided by the full area of the lattice and the engineering strain was computed using crosshead displacements obtained using DIC divided by the length of the specimen. The use of DIC-measured displacements allowed for correction of undesired crosshead motion.

Figure 6a shows the low strain-rate stress-strain response of lattice specimens. Good repeatability of the results is demonstrated by the modest differences across stress-strain response for each specimen geometry. All lattice specimens demonstrated a general mechanical response with five distinct regions: I) an initial linear response at low strains around 0.05; II) a monotonically increasing non-linear response until a peak stress is reached; III) a large drop in stress following the peak stress; IV) a long approximately constant plateau “collapse

stress”; and V) followed by steep stiffening. These regions represent the I) nominally elastic response, II) yielding and strain-hardening of the specimen until failure, III) specimen collapse into a localized band, IV) crushing of the localized band, and V) densification. Approximate regions for R30 specimen are shown in Figure 6a. The local stress fluctuations of the crushing region represent the microstructural response and individual failure of each layer normal to the loading. It is observed that densification strain decreases with relative density and stiffening occurs sooner; this is expected due to the increase in material volume of the higher density specimens.

Experimental images as seen in Figure 6b show that the specimens initially deformed uniformly, and slight misalignments of baseplates were corrected during low strains less than 0.1. See Supplementary Video 1 for visualization of the full deformation. As loading continued, individual strut failures started to occur—initiation was largely sensitive to and would localize at any geometric defects present from manufacturing. As the lattice localized at initial failures, the loss of strength developed over the formation of a deformation band. The lattice then progressively crushed from the localized band until densification, which led to the sharp steepening of the stress-strain curve (Figure 6a). Lower relative density (15%, 20%) specimens were observed to have sharper localizations and narrower collapse bands than higher relative density (25%, 30%) specimens. Mostly transverse (normal) deformation bands were present at lower relative densities and a combination of normal and shear deformation was observed at higher relative densities.

Figure 6: (a) Low strain-rate (0.001/s) stress-strain response of lattice specimens of various relative densities with approximate regions of the mechanical response (I, II, III, IV, and V) for R30 shown; and (b) deformation images for rod (R) and plate (P) specimens with $\rho^*/\rho = 15\%, 20\%, 25\%, 30\%$ at $\epsilon = 0, 0.2, 0.4$.

3.3. High Strain-Rate Experiments of Lattice Specimens

High strain-rate experiments using a polycarbonate SHPB system were carried out on rod and plate specimens with relative densities of 15%, 20%, 25%, and 30% at an average nominal strain rate of $\dot{\epsilon} \approx 1000/s$. At least two specimens were tested for each geometry to ensure repeatability and consistency of material behavior. The engineering stress of the specimen was computed using the transmitted SHPB force obtained following dispersion correction (Section 2.3.2, Eq. 2) divided by the full area of the lattice. The engineering strain was computed using net displacements of the SHPB interfaces measured using DIC

divided by the initial axial length of the specimen, L_s . The instantaneous strain-rate experienced by specimens was calculated as the net difference in velocity of the SHPB interfaces measured using DIC divided by the length of the specimen: $\dot{\epsilon}(t) = (v_L(t) - v_R(t))/L_s$. During a given experiment, the instantaneous strain-rate reached a mostly uniform value for $\epsilon > 0.15 - 0.20$. The average nominal strain-rate was then calculated as the average value of measurements during initial loading (for $\epsilon > 0.01$) until unloading of the stress wave. This value varied from 930/s to 1180/s for all specimen.

Figure 7a shows the compressive high strain-rate stress-strain response. Again, we observe repeatability across specimen geometries. Similar to low strain-rate experiments, all specimens demonstrated a four-region mechanical response with I) an initial linear region, II) yielding and strain-hardening, III) strut fracture and failure, and IV) localization collapse. Densification was not fully observed due to the finite loading pulse of the SHPB. Approximate regions for R30 specimen are shown in Figure 7a. In contrast to the low strain-rate response, the drop in stress experienced as the localization band forms is less steep. The lack of local stress peaks shows a progressive crushing (rather than individual layer failure) for the duration of loading. High strain-rate experiments produced high velocity debris which appeared to increase in volume with relative density. The high velocity debris produced an interesting result in the P30 specimen which experienced no stress in the strain range of 0.2-0.25 due to complete ejection of the material.

Figure 7b shows the high strain-rate experimental images at $\epsilon = 0, 0.2$, and 0.4 for all specimens. Images show large localizations proceeded by progressive crushing at these strains. See Supplementary Video 2 for visualization of the full deformation. Deformation bands for R15 and P15 specimen appeared nearly planar and normal to the loading direction. An interesting observation is the progressive ‘folding’ of the P15 specimen unit cells onto each other during crushing compared to the strut fracture of the R15 specimen. Some non-axial shearing failure was observed in higher density specimens, leading to an ‘X’-shaped deformation band in the P30 specimen.

Figure 7: (a) High strain-rate (1000/s) stress-strain response of lattice specimens with approximate regions of the mechanical response (I, II, III, and IV) for R30 shown; and (b) deformation images for rod (R) and plate (P) specimens with $\rho^*/\rho = 15\%$, 20%, 25%, 30% at $\epsilon = 0$, 0.2, and 0.4.

3.4. Mechanical Properties and Energy Absorption

The mechanical properties of lattice specimen were evaluated from the stress-strain response. The following mechanical properties were identified: (1) failure stress, σ_{fail} , (2) failure strain, ϵ_{fail} , (3) stiffness, S , and (4) specific energy absorption, E_{abs} . σ_{fail} is defined as the maximum value of stress the specimen sustains and ϵ_{fail} the corresponding strain. Stiffness, S , is defined as the secant modulus (σ/ϵ) of the stress-strain curve at $\epsilon = 0.03$. The specific energy absorption is defined as: $E_{abs} = \frac{1}{\rho^*} \int_0^{0.5} \sigma d\epsilon$, where ρ^* represents the mass density of the lattice specimen. We note that all specimens demonstrate dynamic equilibrium (as seen in the force-time histories in Figure 3b) by the time the specimen experiences failure. However, by this definition, force equilibrium is not attained

at the time the stiffness S is calculated. We assume because the transmitted strain is used in the analysis this can serve as a comparative measure across experiments. We can further assume this is an appropriate approximation based upon the initial linear shape of the curves up to $\epsilon = 0.03$.

Cellular materials and lattice structures are widely characterized by scaling laws with relative density (Gibson & Ashby, 1997; Ashby, 2006). Figure 8 shows the mechanical failure properties plotted against relative density on a log-log scale. For some property, $X \propto (\rho^*/\rho)^k$ where the scaling exponent, k , can be extracted from the slopes of the lines in Figure 8 and the results for the defined mechanical failure properties are shown in Table 2.

Figure 8: Mechanical properties of (a) σ_{fail} , (b) ϵ_{fail} , (c) S , and (d) E_{abs} at high and low strain-rates for all lattice specimens as a function of relative density. The scaling exponents (k) in the relation $(\rho^*/\rho)^k$ of each quantity can be found in Table 2.

Figure 8 shows the mechanical properties are clearly affected by strain-rate. σ_{fail} , S , and E_{abs} increase and ϵ_{fail} decreases during high strain-rate loading for all specimens. These trends draw parallels to the mechanical behavior of

the polymeric materials, which exhibit stiffening, strengthening, and decreased fracture strains under dynamic loading (Siviour & Jordan, 2016).

Scaling exponents shown in Table 2 show a strong correlation between mechanical properties and the relative density of lattice specimens. A higher value of the exponent represents a larger effect of relative density. At both low and high strain-rates, σ_{fail} , S , and E_{abs} show strong positive scaling (dependence) with relative density while ϵ_{fail} appears to be independent of relative density. Rod specimens also exhibited higher values of σ_{fail} , S , and E_{abs} at higher relative densities while plate specimens exhibited higher values at lower relative densities. The transition density at which rod or plate specimens outperform the other is represented by the intersection of the scaling fit-lines and varies from about 0.17 to 0.28 depending on the property. The trends in properties show plate specimens perform well at low relative densities, but are less effective than rod specimens for energy absorption at higher relative densities. High strain-rate plate specimens show a slightly lower scaling for E_{abs} than low strain-rate plate specimen (0.89 vs 1.48). This may be explained by poor mechanical performance due to complete ejection of the material for $\epsilon = 0.2 - 0.25$. Stress concentrations on plate unit cells may also be more prone to failure by fracture. Failure strain ϵ_{fail} was observed to be lower in plate specimens than rod specimens at both loading rates and all relative densities.

$Geometry^{Rate}$	σ_{fail}	ϵ_{fail}	S	E_{abs}
Rod^{High}	3.30	0.08	3.70	2.04
Rod^{Low}	3.54	0.13	3.55	2.05
$Plate^{High}$	2.86	0.19	2.64	0.89
$Plate^{Low}$	2.96	0.66	2.79	1.48

Table 2: Scaling of experimental mechanical failure properties with the relative density, ρ^*/ρ of the lattices, $(\rho^*/\rho)^k$. Scaling exponents (k) are found by fitting a linear correlation for the log-log plots in Figure 8.

Scaling laws have been largely developed through cellular models derived by Gibson and Ashby (Gibson & Ashby, 1997) and have been applied to metallic foams (Ashby et al., 2000). A "bending-dominated" open-cell foam scales with relative density, (ρ^*/ρ) with stiffness, $S_{open} \propto (\rho^*/\rho)^2$ and compressive strength,

$\sigma_{c,open} \propto (\rho^*/\rho)^{3/2}$. Closed-cell foams scale with relative density with $S_{closed} \propto (\rho^*/\rho)^2 + (\rho^*/\rho)$ and compressive strength, $\sigma_{c,closed} \propto (\rho^*/\rho)^{2/3} + (\rho^*/\rho)$. Adaptation of these scalings laws to brittle foams produce mostly similar results but with $\sigma_{c,closed} \propto (\rho^*/\rho)^{3/2} + (\rho^*/\rho)$. Experiments on open- and closed-cell foams show a range of scaling exponents from $k = 1 - 3$ for stiffness, and $k = 1 - 2$ for compressive strength. The results from the present investigation demonstrate higher scaling values for stiffness and compressive strength in the lattice specimens than metallic foams. These values indicate the lattice specimens have a higher dependence on relative density than metallic foams and are not well-modelled within this density regime using classical approaches.

3.4.1. Dynamic Rate Effects

The strain-rate effects on the mechanical failure properties can be quantified using the dynamic enhancement factor, D . The factor D for each property is defined as the ratio of the average high strain-rate value to the corresponding average low strain-rate value. These ratios are plotted in Figure 9 for each relative density and specimen type. Error in the factor D was found using error propagation from measurement uncertainty (determined from the percent difference in bar-interface velocities as shown in Figure 3).

For all specimens, $D(\sigma_{fail}) \approx 3.5 - 4$. A higher $D(\sigma_{fail})$ was generally observed in rod specimens, but cannot be confirmed due to large uncertainty. For ϵ_{fail} , $D(\epsilon_{fail})$ is approximately 1 for the P15 specimen, but ≈ 0.8 for all others. A value below 1 represents a lower fracture strain during dynamic loading. ϵ_{fail} appears to be the same across strain-rates for the P15 specimen while it decreases at high strain-rates for all other specimens. $D(S)$ of the stiffness, S , follows similar trends to σ_{fail} and $D(S) \approx 4 - 4.5$ for all specimens. $D(E_{abs})$ for the specific energy absorption was observed to generally be lower for plate specimens than rod specimens and appeared to decrease with ρ^*/ρ for plate specimens. This decrease may be associated with poor mechanical performance due to the ejection of material previously discussed. The values of $D(\sigma_{fail}) \approx 3.5 - 4$ correspond well with the D values for strength in the bulk

polymer material of $D \approx 6$ at $\epsilon = 0.05$ and $D \approx 3.5$ at $\epsilon = 0.1$ (Figure 5 in Section 3.1).

Figure 9: Enhancement factor, D for a) σ_{fail} , b) ϵ_{fail} , c) stiffness, S , and d) specific energy absorption, E_{abs} , for rod (R) and plate (P) geometries of $\rho^*/\rho = 15\%, 20\%, 25\%, 30\%$. A D of 4 (σ_{fail} , S , E_{abs}) and 1 (ϵ_{fail}) is marked to estimate the D extracted from base material characterization.

The Kelvin unit cell is known to display “bending-dominated” behavior (Ashby, 2006) and can be expected to have some dependence on both tensile and compressive properties of the base material. The value for D of the photopolymer is measured in compression, but may serve as a baseline approximation for the high strain-rate effect. A value of $D \approx 4$ is estimated from base material characterization and is marked in Figure 9. Slight deviations in loading rates, geometric defects, and tensile weakening effects may explain the deviation of lattice specimen D values from that of the photopolymer. Overall, the differences in lattice specimen D values are small from $D = 4$ and suggest that the dynamic strengthening of the lattice specimen is mostly due to the rate hardening of the base polymer material.

3.5. Numerical Simulations

High strain-rate finite element simulations were performed on all lattice specimens using Abaqus/Explicit (Dassault Systems). The velocity of the top rigid plate was set to correspond to an engineering strain-rate of $\dot{\epsilon} = 1000/s$. The engineering stress-strain response of the specimen was computed using the nodal forces and displacements of the rigid end plates. Engineering stress was calculated using the rigid-plate force over the full area of the lattice and averaged across the two plates to account for dynamic equilibrium. Engineering strain was calculated using the rigid-plate displacement and the axial length of each specimen. The “Brittle Cracking” material model includes element removal and thus introduces an inherent mesh sensitivity. A mesh sensitivity study was carried out and required a mesh size of 0.20 mm to attain physically relevant and converging simulations. The explicit nature of the simulation required consideration of dynamic equilibrium of the specimen. The initial bumps in the stress-strain response at $\epsilon < 0.02$ reflect that force equilibrium is not attained. Beyond strain $\epsilon > 0.02$, stress equilibrium is reached indicated as evidence by the stress balance of the end plates, $\sigma_{top} \approx \sigma_{bot}$. A constant strain-rate-stiffening in modulus is assumed through the definition of E (Young’s modulus) in the material model. Low strain-rate experiments on bulk photopolymer (Figure 5) allow extraction of the quasi-static modulus $E \approx 400$ MPa. Thus, assigning $E = 1000$ MPa in numerical simulations introduces an approximate stiffening-effect which was defined to qualitatively match experimental stress-strain response and deformation behavior. This effect agrees with experimentally observed strain-rate effects of the photopolymer material (Section 3.1).

Figure 10a shows the engineering stress-strain response for rod-lattice specimens and Figure 10c shows the response for plate-lattice specimens. Despite the simplistic model, the simulation results capture the general mechanical behavior and formation of the deformation band for each specimen. Numerical stress-strain curves exhibit the same four-region mechanical response observed in all experiments (I: elastic response, II: yielding and strain-hardening, III: strut fracture and failure, and IV: localization collapse). This suggests the me-

chanical behavior of these lattice specimen is largely dominated by brittle failure and that a brittle failure model is an effective model for the base photopolymer material.

Figure 10: (a) The computed engineering stress-strain response of rod specimen compared to average experimental values; and (b) experimental and simulation deformation images showing nominal strain at $\epsilon = 0.20$ for R15, R20, R25, and R30. Similar plots for plate specimen of (c) engineering stress-strain response and (d) deformation images for P15, P20, P25, and P30.

Figure 10b and 10d compare the experimental and simulation deformation images for rod and plate specimen, respectively, at $\epsilon = 0.20$. Good qualitative agreement and trend is found between experimental and simulation deformation modes. Lower density (15%, 20%) specimens showed a more localized deformation band while higher density (25%, 30%) specimens show more uniform deformation. A "X-shaped" deformation pattern can be observed in simulation of plate specimens which was also present in the high strain-rate experiments on P25 and P30 specimen. This deformation mode shape has also been observed in literature in plate-Kelvin cell lattice structures (Duan et al., 2019) and the in-plane response of hexagonal honeycombs (Ruan et al., 2003). Sim-

ulation deformation images also reveal higher stress concentrations at unit cell boundaries in plate specimens than in rod specimens. This supports the role of stress concentrations in the weakening of plate specimen due to failure by tensile fracture.

The numerical stress-strain response also provides good validation of the mechanical failure property trends. Figure 11 shows the mechanical failure properties plotted against relative density on a log-log scale for high strain-rate experiments and simulations. The experimentally observed trends of mechanical performance are also observed in the numerical results: rod specimens exhibit higher values of σ_{fail} , S , and E_{abs} at higher relative densities, while plate specimens exhibit higher values at lower relative densities.

Figure 11: Comparison of scaling of mechanical properties with relative density for high strain-rate experiments and simulations. Trends are shown for (a) σ_{fail} , (b) ϵ_{fail} , (c) S , and (d) E_{abs} . In both simulation and experiment, plate-lattices perform better at lower relative densities and rod-lattices perform better at higher relative densities for σ_{fail} , S , and E_{abs} . No distinct trends are observed for ϵ_{fail} .

Table 3 shows the scaling exponent values k for both high strain-rate experiment and simulation. In both experiment and simulation, rod specimens exhibit higher scaling exponent values for σ_{fail} , S , and E_{abs} and no clear trend is observed for ϵ_{fail} . Differences in values may stem from experimental geometric defects, differences in measurement resolution, or simplistic material modeling. Simulation results show ϵ_{fail} decreases slightly in plate specimen and increases slightly in rod specimen on the order of 1%. This can be assumed to be outside of the experimental resolution and thus consistent with the weak experimental scaling of ϵ_{fail} with ρ^*/ρ . The match between qualitative scaling in both experiment and simulation trends emphasizes there is a transition in mechanical performance of the lattice specimen for the defined mechanical properties.

<i>Geometry</i>	σ_{fail}	ϵ_{fail}	S	E_{abs}
<i>Rod^{Exp}</i>	3.30	0.08	3.70	2.04
<i>Rod^{Sim}</i>	2.82	0.13	2.74	1.75
<i>Plate^{Exp}</i>	2.86	0.19	2.64	0.89
<i>Plate^{Sim}</i>	1.74	0.15	1.71	0.78

Table 3: Scaling of high strain-rate experimental and simulation mechanical failure properties with the relative density, ρ^*/ρ of the lattices, $(\rho^*/\rho)^k$. Scaling exponents (k) are found by fitting a linear correlation with the log-log plots in Figure 11.

4. SUMMARY AND CONCLUSIONS

The high strain-rate compressive behavior of rod and plate polymeric Kelvin lattice structures was experimentally and numerically explored. Mechanical properties such as failure strength, stiffness, and specific energy absorption of lattice specimens show dependence on strain-rate and relative density. Strain-rate experiments on lattice specimen and the base photopolymer material suggest the observed dynamic effects of strain-rate-hardening, strain-rate-stiffening, and decreased fracture strain may be mostly attributed to the parent polymeric material. A brittle failure material model with strain-rate dependent stiffness (Young's modulus) is shown to be sufficient to simulate lattice structure high strain-rate mechanical response. At both low and high strain-rates, experimental and simulation results show there exists a transition relative density under

which plate-lattices outperform rod-lattices of similar mass (relative density) for each mechanical property. Experimental and simulation deformation images suggest no change in dynamic collapse mechanisms, but reveal the formation of a localized failure band more prominent in low strain-rate experiments and low relative density specimens.

The mechanical effects of applying rod or plate geometries to unit cells must be analyzed per geometry, loading rate, and relative density for optimal design. Despite poor performance at higher relative densities, polymeric plate-Kelvin-lattice structures offer improved mechanical properties for lightweight, energy absorbing materials on the millimeter length scale and may be readily manufactured using current technologies for engineering applications. Generalization of this analysis to other materials and higher strain-rates remains an area for future research.

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Appendix A. Viscoelastic Parameters for PC SHPB Correction

The wave propagation coefficient $\gamma(\omega)$ was used to correct dispersion and attenuation effects of viscoelastic strain measurements in the polycarbonate SHPB system. $\gamma(\omega)$ can be decomposed into the attenuation coefficient $\alpha(\omega)$ and phase velocity $c(\omega)$ as in Eq. (3) and may be experimentally determined using a one-point measurement technique (Bacon, 1998). Figure A.12 shows the experimental results and average values of the attenuation coefficient and phase velocity for the polycarbonate SHPB system used in this study. Strains and striker velocities of wave propagation coefficient experiments were comparable to that of high strain-rate experiments on lattice specimens.

Figure A.12: Wave propagation coefficient for the polycarbonate SHPB system realized through (a) attenuation coefficient, $\alpha(\omega)$, and (b) phase velocity, $c(\omega)$, as a function of frequency, ω . The dark solid lines represent the averaged value of the measurements.

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Highlights

- Plate-lattices mechanically outperform rod-lattices at low relative densities
- Deformation bands localize sharply at low strain-rates and low relative densities
- Lattice strain-rate behavior is consistent with that of the base polymer material
- Explicit dynamic finite element simulations are able to capture lattice behavior at high strain-rates

Declaration of interests

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

☐ The authors declare the following financial interests/personal relationships which may be considered as potential competing interests: