

Advancing multi-material laser powder bed fusion through topology optimized design

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ABSTRACT

Recent advancements in Multi-Material Laser Powder Bed Fusion (MM-LPBF) offer the unique ability to additively manufacture highly complex structures by selectively depositing multiple material systems within a single layer in the conventional laser powder bed process. MM-LPBF combines the fine resolution of typical LPBF processes with the ability for high resolution spatial control of multiple material placements. The alternating active phase (AAP) algorithm is applied to design a multi-material Messerschmitt-Bölkow-Blohm (MBB) beam, minimizing compliance while also considering manufacturability. Topologically optimized solutions were fabricated through MM-LPBF (316 L stainless steel and CuCrZr) and evaluated through flexural testing with digital image correlation (DIC) and mechanical performance was compared to finite element analysis (FEA). Microstructure characterization of the bi-material interface revealed localized MM-LPBF process-specific defect formation. Mechanical testing revealed progressive stages of failure initiating in the bulk CuCrZr regions and propagating to interfacial regions. DIC analysis indicated that stiffness of the multi-metal MBB structure was within 5.3% of the predicted stiffness from FEA. Findings from this study demonstrate that highly complex multi-material topologically optimized (MM-TO) designs, which were otherwise not plausible to manufacture through either traditional or AM methods, are now feasible at high resolution. The applied method can be extended to other applications which would benefit from multi-objective criterion such as multi-material heat exchangers, biomedical devices, and energy storage devices. Finally, this research highlights the need for further MM-LPBF process development to reduce bulk porosity and interfacial defects, as its processing physics differ significantly from single material LPBF and other metal AM processes.

1. Introduction

1.1. Multi-material advanced manufacturing

Multi-material laser powder bed fusion (MM-LPBF) processing capability through selective powder deposition (SPD) provides a relatively new approach to additively manufacture highly complex structures with finer resolution that cannot be achieved through functionally graded materials via directed energy deposition [1]. MM-LPBF allows for the fabrication of components with spatially varying material properties both within (intra-) and across (inter-) layers, which can improve local performance and function according to regional design requirements. Additive manufacturing (AM) technology provides an

unprecedented degree of geometrical design freedoms not otherwise available through traditional manufacturing. Fig. 1 shows the selective powder deposition technique which replaces the traditional recoating system in a standard single material LPBF machine. Although this study leverages two different material systems, the presented MM-LPBF method can process three materials, and in the future, could be extended to more than three materials. The three material drums rotate over the build plate and selectively deposit material based on voxelated information as demonstrated in Fig. 1. These material regions are then sequentially melted with material-specific laser parameters to form a monolithic multi-material component [1].

Incorporating multiple materials into a single component can produce design advantages for both assembly part consolidation and

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unnecessary material reduction [2,3]. This study first conducts multi-material topology optimization (MM-TO) using the Messerschmitt-Bölkow-Blohm (MBB) boundary conditions [4], incorporating process-induced manufacturability consideration, by implementing an Alternating Active Phase (AAP) algorithm [5]. Second, it demonstrates that SPD in MM-LPBF enables the fabrication of such computationally optimized structures with both high geometric complexity and compositional variation.

MM-LPBF combines near-net shape geometric and surface roughness resolution with a typical resolution of 150–300 μm in the XY-plane for the spatial tailoring of dissimilar materials, 316 L and CuCrZr in this study, based on functional design requirements (mechanical performance in this study). For instance, advanced heat exchanger designs can now include copper fins for efficient heat dissipation combined with a stainless-steel base for structural strength [6]. In most engineering applications (e.g., aerospace, automotive, space), weight is considered a critical design factor. CuCrZr alloys are attractive for high-performance applications due to their excellent thermal and electrical conductivity combined with stable mechanical properties at elevated temperatures, making them suitable for aerospace, electronic cooling, and fusion systems [7,8]. In many of these systems, CuCrZr must be integrated with structural alloys possessing superior mechanical and corrosion resistance, motivating its pairing with stainless steels, such as 316 L. The material system of 316 L stainless steel and CuCrZr was selected to represent a technologically relevant and highly dissimilar material pairing. CuCrZr provides high thermal conductivity, while 316 L offers superior structural stiffness, ductility, and corrosive resistance. However, combining CuCrZr with 316 L stainless steel through MM-LPBF remains a technical challenge due to disparities in melting temperature, thermal expansion coefficient, and diffusivity, as well as the severe

immiscibility of Fe-based and Cu-based alloys [9,10]. These factors promote the formation of heterogeneous microstructures, copper-rich clusters, porosity, and interfacial cracks, which may degrade mechanical performance [1,11,12]. Prior studies have demonstrated that processing strategies such as laser beam shaping [13], transition layers [14], surface treatments [15], and hot isostatic pressing (HIP) [16] could mitigate interfacial defects by improving energy distribution and stress accommodation. Nonetheless, residual stress build-up and material compatibility remain as prominent defect formation mechanisms [9], [17].

Failure modes in multi-material metal components fabricated by MM-LPBF are governed by differences in constituent material properties, interface integrity (presence or lack of process-induced defects), and part geometry. In complex multi-material structures, sequential failure has been observed, where the more ductile or weaker alloy deforms or fractures first, followed by a sudden failure of the stronger alloy. This sequential failure has been reported both in 316L–17–4PH bi-metallic lattices [18] and in CuSn–19Ni200 bi-metallic porous structures [19], where interfacial defects were not prominent enough to induce premature failure. At the microscale, interfacial defects such as dendritic cracking, and porosity degrade load-bearing capacity, as demonstrated in a study on 904LSS–CuSn10 gyroid flexural specimen [1]. For the CuCrZr–316LSS pairing, these challenges are expected to be particularly severe due to the extreme conductivity difference, combined with the aforementioned immiscibility of Fe and Cu [11,12]. Defect formation mechanisms in complex builds can be exacerbated by material mixing behavior at the material interface. Prior studies by Li *et al.* have demonstrated that the orientation of the interface relative to the recoating blade influences interfacial mixing during the fabrication of vertical interfaces using the SPD system [20]. Due to the complex

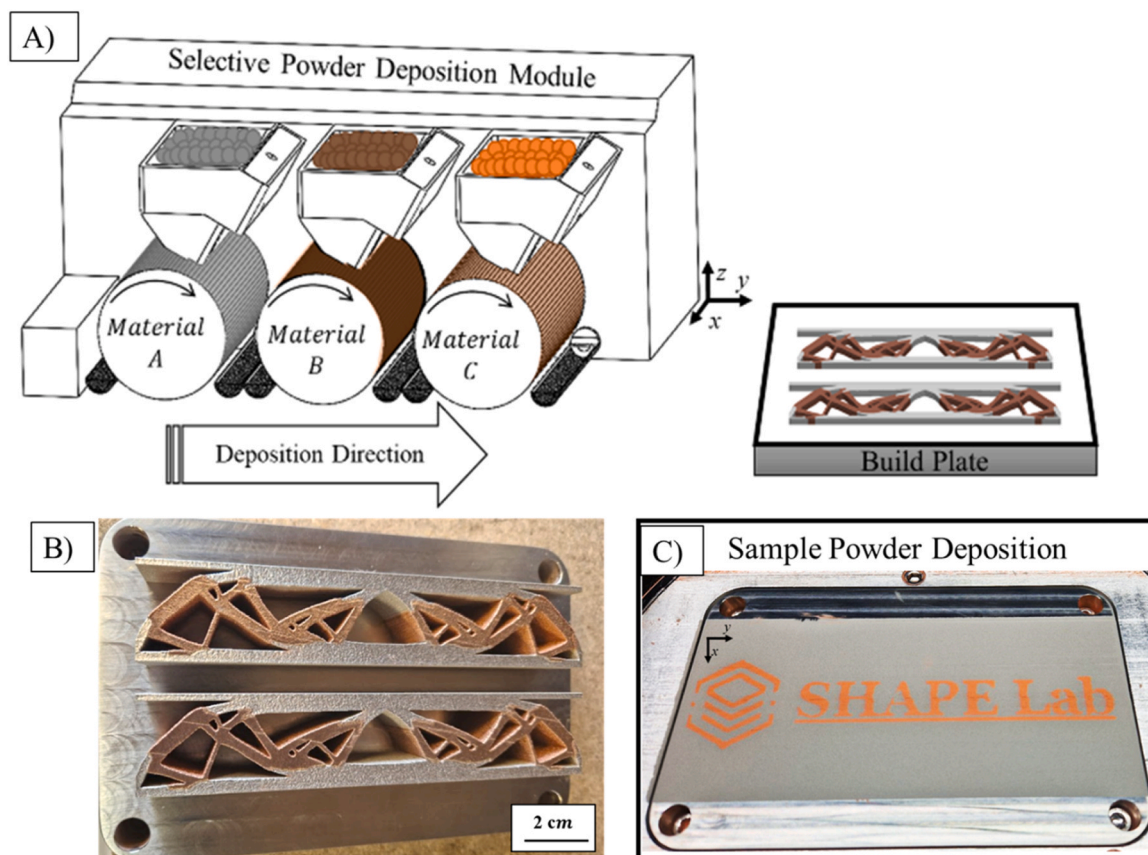


Fig. 1. A) Selective powder deposition (SPD) Aerosint module showing three material hoppers and drums to deposit up to three materials within a single layer. B) Image of MBB beams on build plate in as-build orientation C) Sample powder deposition shows 316 L stainless steel and CuCrZr unmelted powder selectively deposited across a single layer.

geometry of the MM-TO beam presented in this study, no single build orientation can provide an optimal configuration that minimizes interfacial mixing across all interface locations. All interfaces investigated in this study were fabricated at a build orientation perpendicular to the build plate and predominantly parallel to the recoating direction to minimize this variability in interfacial mixing. Prior work has demonstrated that build orientation can produce multiple distinct interface types (e.g., material A deposited on B versus B on A), each with significantly different mixing behavior [1]. The orientation selected here was therefore chosen to avoid the formation of such heterogeneous interface conditions within a single component. A schematic of the build volume is provided in Fig. 1. In complex multi-material geometries or graded structures, interfacial defects can drive premature collapse or delamination of the interfacial bonding region, followed by catastrophic failure of the copper-rich regions when thermal and mechanical stresses concentrate at the interface [1,3,21]. Collectively, these findings demonstrate that MM-LPBF offers substantial promise for the processing of highly complex, high-performance, multi-material architectures generated through computational design solutions.

1.2. Topology optimization design strategies for multi-material manufacturing

Topology optimization (TO) is a numerical method that optimizes the material distribution within a given design space, for a given set of loads (e.g., thermal, structural), boundary conditions, and user-defined constraints [22,23]. TO is a well-adapted design method for AM across processes such as L-PBF, directed energy deposition [24], binder-jetting [25], material extrusion [26], photopolymerization [27], and material jetting [28]; hybrid AM methods such as AM-augmented castings [29,30]. Well-recognized application domains for topologically optimized design methods range from biomedical devices [31,32], heat exchangers [33], and mechanical structures [34]. One classic benchmark problem in TO is the MBB beam under flexural bending, making it a relevant model for understanding structural performance under nominal load conditions [4]. A flexural loading condition was selected to produce a spatially varying stress distribution, as opposed to a uniaxial loading case, which would result in a largely uniform stress state and therefore limited design variability when using multiple materials. Applying this benchmark within the context of MM-LPBF establishes a framework for integrating TO principles into the computational design of additively manufactured multi-material structures.

The ability to achieve MM-LPBF has often been limited to the material science and process engineering aspects of multi-material fabrication to improve the quality of the discrete dissimilar material interface. Significant research efforts are involved in the thermo-mechanical modeling [35–38], laser parameter optimization [39–42], and materials characterization of a bi-metallic interface [43–45]. However, limited reported research has addressed the geometrical and material freedoms provided by multi-material fabrication from the design perspective, despite its potential to unlock superior structural and functional performance.

Numerical MMTO methods have since been developed to systematically explore these design freedoms. Density-based approaches such as SIMP (solid isotropic material with penalization) and RAMP (rational approximation of material properties) have been extended to multiphase interpolation [46], allowing allocation of multiple materials within a discretized domain. However, they are limited in graded or intermediate density representation, which are more applicable for advanced manufacturing of functionally graded materials. Evolutionary methods like MM-BESO (bi-directional evolutionary structural optimization) [47,48] enforce discrete 0–1 design variables, eliminating grey regions at an increased computational expense. Level-set methods [49–52] provide sharp boundaries between phases and are suitable for coated or multi-layered structures. Likewise, phase-field approaches [53–56] track the interface dynamically as a continuous transition region governed by

thermodynamic models. The AAP algorithm, initially proposed by Tavakoli et al. in 2014 [5], offers a complementary perspective by splitting a multiphase problem into a series of binary-phase sub-problems solved iteratively with existing single-material solvers [57,58]. This approach provides the freedom to apply any single material TO algorithm within the two activated phases [2].

However, while these frameworks provide mathematically designed solutions, most studies remain purely computational and do not fabricate or experimentally test their design solutions. This gap arises because multi-material additive manufacturing (MMAM), particularly for metals, is still at an early stage of maturity. Challenges such as powder handling, interfacial bonding between dissimilar alloys, residual stresses, and minimum feature resolution often prevent direct translation of optimization results into manufacturable parts. As a result, many studies validate performance only through computational simulations, with physical demonstrations largely limited to polymer-based jetting or extrusion systems, where material compatibility and processing constraints are less severe. For example, Nguyen et al. demonstrated multi-color polymer printing of a physical topologically optimized prototype via material jetting using 2.5D geometry [59]. Similarly, Zhang et al. produced high-resolution (18 μm) multi-color polymeric parts via material jetting with changing features and materials in all three dimensions [57]. With continued progress in MMAM, particularly powder bed fusion systems capable of selective powder deposition, this study exists to bridge the current divide between computationally optimized designs and experimentally verified multi-material structures.

To the authors' knowledge, no reported study has successfully fabricated and experimentally evaluated a multi-metal topologically optimized design. This work applies an AAP algorithm for simultaneous multi-material additive design while incorporating MM-LPBF-specific manufacturability considerations, enabling the fabrication of highly complex multi-material architectures that were previously unattainable with conventional or single material AM. It further evaluates the performance of the manufactured component, both in terms of material compatibility and mechanical behavior, including an incremental failure analysis which combines experimental DIC with theoretical FEA to assess the effects of process induced defects, such as interface cracking, porosity, and stress localization. These contributions establish a direct link between advanced design techniques, processing, and performance to demonstrate the capability of MM-LPBF to produce defect tolerant high performance multi-material components for applications which benefit from multi-objective design criteria, including heat exchangers, biomedical devices, and energy storage systems.

2. Methodology

2.1. Alternating-active phase (AAP) for multi-material topology optimization

A multi-material topology optimization method, known as the AAP algorithm [5], is adopted in this study to explore optimal multi-material distributions. An implementation of the AAP algorithm in Matlab is adopted herein [5]. The goal is to determine the layout of n distinct material phases (including void) within the design domain Ω , such that the objective function is minimized under the specified boundary conditions. The material distribution is governed by local volume fraction fields ρ_i ($i = 1, \dots, n$), under the constraint that $\rho_i \in V(\Omega)$, where V is the admissible design space. Since the design domain is distinct and non-overlapping, volume fractions of all phases must sum to unity. Overall, the optimization problem is formulated as:

$$\begin{aligned} \text{minimize:} & \quad C(\rho, U) = U^T K U \\ \text{subject to:} & \quad K(\rho) U = F \\ & \quad \sum_{i=1}^n \rho_i = 1 \\ & \quad \int_{\Omega} \rho_i d\Omega \leq V_i \\ & \quad 0 < \rho_{i,\min} \leq \rho_i \leq 1 \end{aligned} \quad (1)$$

where C is chosen as the mechanical compliance. K and F are the global stiffness matrix and force vector, respectively. U is the global displacement vector to be solved from the equilibrium equation. $\rho_{i,\min}$ is the lower bound of design variable to avoid numerical stability.

The above formulation shares great similarity with the well-known SIMP approach. Effective material properties are related to elemental relative densities using a power law relationship to penalize intermediate densities. Indeed, AAP adopts an outer iteration strategy where binary-phase sub-problems are iteratively solved using SIMP as an internal solver (Fig. 2A). The number of sub-problems is determined by the number of distinct two-material combinations that can be formed from the total set of n materials, which is given by

$$nC_2 = \frac{n(n-1)}{2} \quad (2)$$

These sub-problems are partially solved using a binary-phase TO solver. Specifically, only two material phases, a and b , will be activated within the current iteration, while the remaining $(n-2)$ phases are treated as fixed regions (Fig. 2B).

$$\rho^a + \rho^b = 1 - \sum_{i=1, i \neq a, b}^n \rho_i \quad (3)$$

where ρ^a and ρ^b denote the volume fraction field of active phases a and b , respectively.

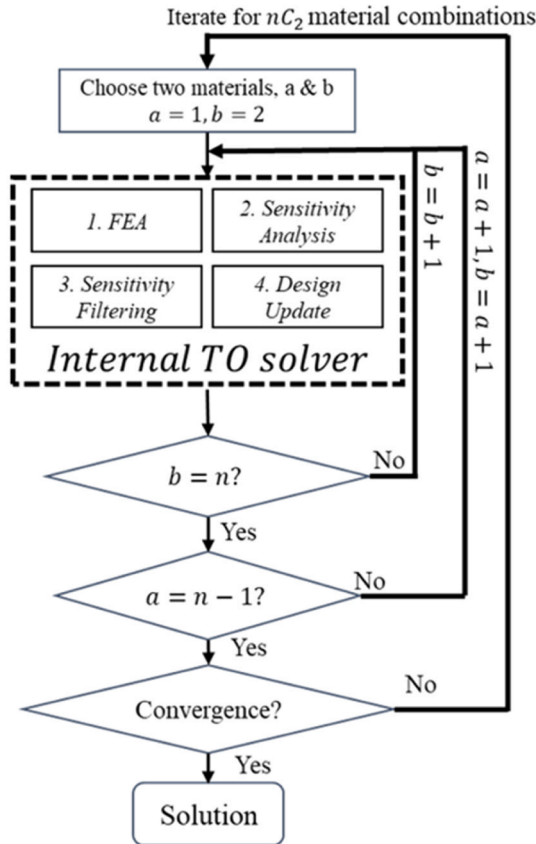
2.1.1. Design parameters specification

As shown in the experimental setup of flexural testing shown in Fig. 3, the MBB beam is subjected to symmetric loading conditions, which allows design optimization using only half of the structure to save computational cost. The physical dimension of a full beam is designed to be 150 mm by 25 mm. Quadrilateral elements are chosen to discretize the half beam design domain into 450 by 150 pixels. The geometric resolution of the selective powder deposition in MM-LPBF was taken to be 500 μm [60] (a conservative estimation in comparison to the theoretical resolution of 150 μm), and a mesh independency filter was imposed to incorporate the minimum feature size requirement ($fr = 5$). Different values of filter radii have been tested to show the effect of filter radius on the minimum feature size within the optimal structures. The effects of the filter radius can be observed in the Supplementary information in Appendix A.

The structural optimization of the MBB beam involves two solid material candidates in addition to the void phase, with relevant properties outlined in Table I.

The volume fraction of each material phase is a user-defined constraint. For this study, the void phase volume fraction (V_{void}) is set to 50% of total volume, where the two metal phases share equal volume fraction ($V_{\text{SS}} = V_{\text{Cu}} = 25\%$). Within the adopted AAP code [5], the following settings were used: number of distinct materials $p = 3$, penalty parameter $q = 3$, and a stopping criterion defined by a design field change smaller than a tolerance of 0.05. Fig. 4 shows the resultant design adopting the above-mentioned set-up, which converges after 200 iterations. After applying symmetry to obtain the full domain result, a 2 mm thick 316 L stainless steel plate is added to the top and bottom to

A) Alternating-Active Phase Flow Diagram



B) 3-Material Alternating Active Phase Sequence

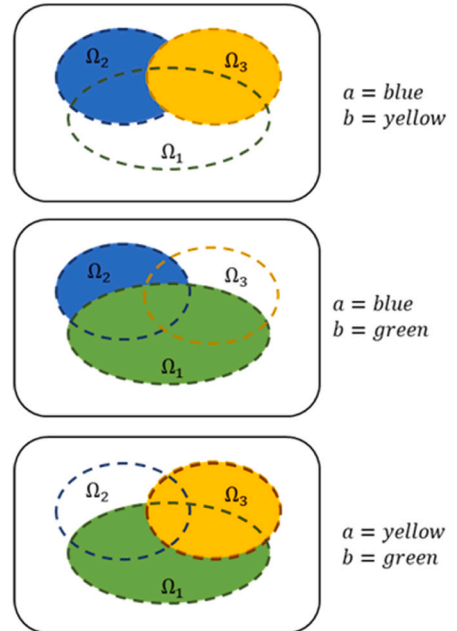


Fig. 2. A) Alternating-active phase optimizer flow chart showing the outer loop (iterating alternate active phases across binary phases for nC_2 material combinations found by Eq. 2) and inner loop (traditional binary-phase TO solver). B) The sequence of phase activation in a three-phase optimization example, where the two active phases are denoted by solid-colored regions and their summed relative density is defined by Eq. 3.

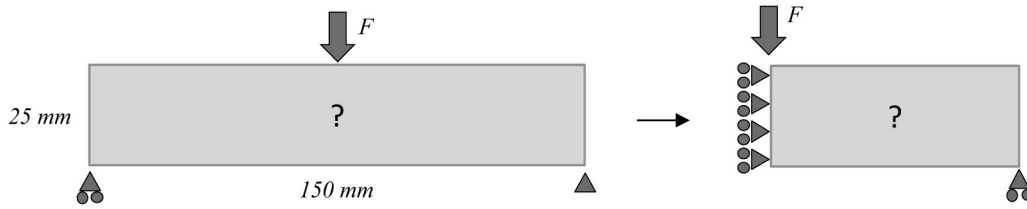


Fig. 3. Loading condition of the MBB beam with physical dimension labeled.

Table I
Material attributes of three material phases.

Phase ID	Material	Density [g/cm ³]	Elastic Modulus [GPa]	Thermal Conductivity [W/mK]	Color
1	316 L Stainless Steel	8.00	193	16.3	Grey
2	CuCrZr	8.94	136	330	Brown
3	Void	1×10^{-9}	1×10^{-9}	0.05	White

ensure compatibility with the three-point bending test setup. Planar extrusion was then performed to translate the 2D design into a 3D manufacturable structure.

2.2. Materials and fabrication

Specimens were manufactured in an Aconity MIDI+ (Aconity GmbH, Germany), equipped with an Aerosint SPD v1 (Aerosint SA, Belgium), with a single mode fiber laser at a continuous wavelength of 1070 nm, layer thickness of 40 μm , and an effective spot size of 80 μm . Table II summarizes the material-dependent laser parameters used for each material section. The interlayer rotation is 67° for all specimens. The chamber was maintained at a maximum O₂ PPM of 800 using argon gas. Tables III and IV show the chemical composition of both materials, as provided by the supplier. Material powder size distribution provided by the powder supplier was found to be D10 18.3 μm , D50 27.6 μm , and D90 41.2 μm for 316 L stainless steel and D10 7 μm , D50 18 μm , and D90 30 μm for CuCrZr. A material-wise sequential material deposition and laser exposure strategy was used for each layer. CuCrZr powder was deposited first, followed by 316 L stainless steel powder, after which laser exposure was applied sequentially to the CuCrZr and 316 L stainless steel regions.

The specimens were printed lying flat, such that all 316LSS-CuCrZr material interface planes are perpendicular to the build plate. They were printed 2 cm thick, then sliced into 5 mm thick beams using electron discharge machining (EDM) to provide a reasonable fracture load for a 20 kN load cell. Three random beams were selected for mechanical testing with digital image correlation (DIC).

2.3. Microstructural and hardness investigation

A microscopy coupon (1 cm³) was manufactured in the same build



Fig. 4. Resultant multi-material (316 L stainless steel & CuCrZr) MBB beam designed through the AAP algorithm, with the objective of minimizing mechanical compliance.

conditions as the MBB beams, with the material interfacial plane standing perpendicular to the build plate, similar to all interfaces seen in the MBB test specimens. The sample was mounted and polished for microstructural microscopy using first optical microscopy for a wide-field investigation of internal defects. Scanning electron microscopy (SEM) and energy dispersive x-ray spectroscopy (EDS) for an indication of the compositional gradient across the interface.

Nanoindentation was identified as a method to derive local interfacial material properties. Hardness was measured using a Hysitron Ti950 Nanoindenter (Minneapolis, MN) fit with a Berkovich tip. Indentation was conducted using a maximum cutoff load of 6 mN and a dwell time of 3 s. The TriboScan (10.2.0.1) software was used during indentation to record the loading/unloading curve and calculate the reduced Young's modulus (E_r). Young's modulus was calculated using the Oliver and Pharr method, described in detail in Appendix A of Griffiths et al. [61]. Nanoindentation was conducted within a well-mixed interfacial region to quantify interfacial hardness, as well as within bulk 316 L stainless steel and CuCrZr regions to determine corresponding bulk material properties of each constituent. The results of each region are provided in Fig. 8 below. The Young's modulus values found through indentation from the constituent material regions were used for FEA analysis.

2.4. Finite element analysis

Numerical finite element analysis (FEA) was performed using Ansys Static Structural (2023 R2) simulation. Each material section was treated as a single part file and was meshed independently using linear, 4-node tetrahedron mesh elements with a maximum size constraint of 0.15 mm. Adjacent material sections were contact bonded at their interfaces to represent an idealized defect-free beam to establish a baseline mechanical response, enabling the detrimental effects of the processing-induced defects observed experimentally to be assessed by comparison. Material properties necessary for a linear-elastic FEA simulation are restricted to Young's modulus and Poisson's ratio. Young's modulus for both materials was calculated through the Oliver and Pharr method from nanoindentation across the interface. For the Oliver and Pharr method, as well as the FEA simulation, Poisson's ratio must be assumed. In this case, the Poisson's ratio of CuCrZr was taken as 0.33 [62] and the Poisson's

Table II
Material-dependent laser parameters.

	Power [W]	Speed [$\frac{\text{mm}}{\text{s}}$]	Hatching Distance [μm]
316 L Stainless Steel	150	600	80
CuCrZr	500	600	120

Table III

316 L stainless steel powder composition sourced from Thyssenkrupp Material Trading (GmbH, Germany) [78].

	Fe	C	Si	Mn	P	S	Cr	Ni	N	Mo
316 L stainless steel	Bal.	0.03	1.00	2.00	0.045	0.015	16.50 – 18.50	10.00 – 13.00	0.10	2.00 – 2.50

Table IV

CuCrZr powder composition sourced from Eckart TLS (GmbH, Germany).

	Cu	Cr	Zr	Fe	Si
CuCrZr	Bal.	0.50 – 1.20	0.03 – 0.3	≤ 0.08	≤ 0.10

ratio of 316 L stainless steel was taken to be 0.28 [63]. The loading and support cylinders were modeled using linear hexahedral elements with a maximum size constraint of 0.5 mm. The supporting cylinders were fixed, with a surface-to-surface contact with the MBB beam. The loading cylinder was given a tie constraint to the upper surface of the beam. Force control was applied to the loading cylinder within the experimental elastic region from 0 N - 4500 N in four solution steps. Maximum displacement was recorded at each force, and stiffness was calculated as the relative change in force to cross head displacement. Strain probes were used at two locations along the beam to compare simulated to experimental strain found through DIC, directly under the loading cylinder, and at the first point of failure.

2.5. Flexural testing for mechanical performance

Three beams were selected for mechanical testing with DIC. All three specimens were tested under three-point flexural loading conditions on MTS load frames equipped with 50 kN load cell. While ASTM D7250/D7250M testing standard for sandwich beams assumes homogenous material distribution across the beam, relevant resting conditions, such as the constant quasi-static displacement rate of 0.008 mm/s, were applied, as there is no current standard for flexural testing of multi-material structures [64]. Testing was terminated after a 90% drop in maximum stress or a complete delamination of adjacent material sections, or after excessive densification was observed.

The results of the DIC provide an experimental strain map of the beam. Due to the complex geometry and irregular material assignment, there is no current method to experimentally calculate a regional or global stress distribution across the beam. As a result, the FEA model will serve as a preliminary baseline to validate the results provided by the experimental DIC strain maps. Additionally, local displacement-strain curves were generated within the elastic region to determine regional strains prior to failure, in comparison to the predictive FE model.

2.6. As-built porosity and internal defects

Micro-computed tomography (micro-CT) was performed on the 1 cm microscopy cube and one random MBB test specimen to evaluate the internal porous defects of the as-built parts. Micro-CT was performed (phoenix v|tome|xL, Waygate Technologies, NY, USA) with a voxel size of 10 μ m and 25 μ m for the microscopy cube and MBB test specimen, respectively. Micro-CT volume data was imported into the Avizo (version 2022.1, Amira-Avizo, FEI, Thermo Fisher Scientific, USA) software and morphological analysis included relative pore density for each material region, and part inspection for process-specific surface and internal defects. The as-built density calculated from the micro-CT volume data of the microscopy cube for both material sections was found to be 99.979%, and 99.930% for the 316 L stainless steel and CuCrZr sections, respectively. It is important to note that this density only reflects voids caused by lack of fusion, keyhole, or gaseous porosity, while internal void space due to cracking is unaccounted for in this porosity calculation.

3. Results

The integration of MM-LPBF with computationally topology optimization enables strategic multi-metal design for the fabrication of highly complex parts with spatially varying properties tailored for local design requirements. A TO-designed part for MM-LPBF was successfully designed using an alternating-active phase algorithm, and manufactured using selective powder deposition. The optimized MBB beam, designed under the classic MBB loading conditions and aimed at minimizing compliance, illustrates how Design for AM principles specific to MM-LPBF can be applied to computationally design a complex multi-material 3D mechanical structure.

3.1. Interfacial and internal morphology

Optical images across the interfaces (Fig. 5), and micro-CT (Fig. 6) of the beams demonstrate patterned defects at both the interface, and in the bulk material regions. CuCrZr showed porosity within the bulk material regions, providing a defect porosity of 99.930%. It is well known that internal impurities and pores negatively contribute to material conductivity [65] and mechanical strength. Process-induced defects in CuCrZr are consistent with other literature [66,67], which have demonstrated a difficulty in processing this alloy due to its high reflectivity, melting point and thermal conductivity.

Micro-CT analysis (Fig. 6) of the beam reveals porosity in the CuCrZr regions of the beam, and in close proximity to the material interface.

EDS map analyses (Fig. 7) of the microscopy coupon reveal a well-mixed interfacial region between the CuCrZr and steel components, extending approximately 450 μ m. Within the steel, the onset of cracking is characterized by numerous micron-scale fractures rather than a singular primary crack. Material diffusion appears to favor CuCrZr diffusing into the stainless steel regions, likely influenced by the sequence of material melting during laser processing.

Reduced Young's modulus (E_r) and indentation hardness (H) were determined by the TriboScan testing software, according to the Oliver and Pharr method. Nanoindentation was performed across the CuCrZr/316 L stainless steel interface to evaluate the transition in mechanical properties and to inform static structural finite element simulation. The results show a gradual increase in hardness from CuCrZr to 316 L stainless steel, with interfacial values falling between the two materials. The hardness of CuCrZr measured 1.62 ± 0.043 GPa, increasing to 3.25

± 0.065 GPa in 316 L stainless steel. Similarly, the elastic modulus transitioned from 123.97 ± 5.10 GPa in CuCrZr to 195.121 ± 2.44 GPa in 316 L stainless steel.

3.2. Mechanical failure behavior under flexural loading

Flexural analysis was performed on the multi-material, topologically optimized beam, focusing on both global and local responses under bending loads. Due to the complex geometric and material composition of the beam, DIC was used during flexural testing to quantify regions of high strain during the onset of failure. To evaluate the performance of the beam, a finite element analysis (FEA) was used to provide a baseline for failure. The results of the flexural test and finite element analysis are summarized in Fig. 9 and Fig. 10 below.

Up to 10% (~ 1 mm) of crosshead displacement, the 316 L stainless steel and CuCrZr topologically optimized beam exhibits linear elastic behavior with uniform strain distribution and no localized cracking. Above 4000 N, CuCrZr struts begin to buckle (Location B, Fig. 9C),

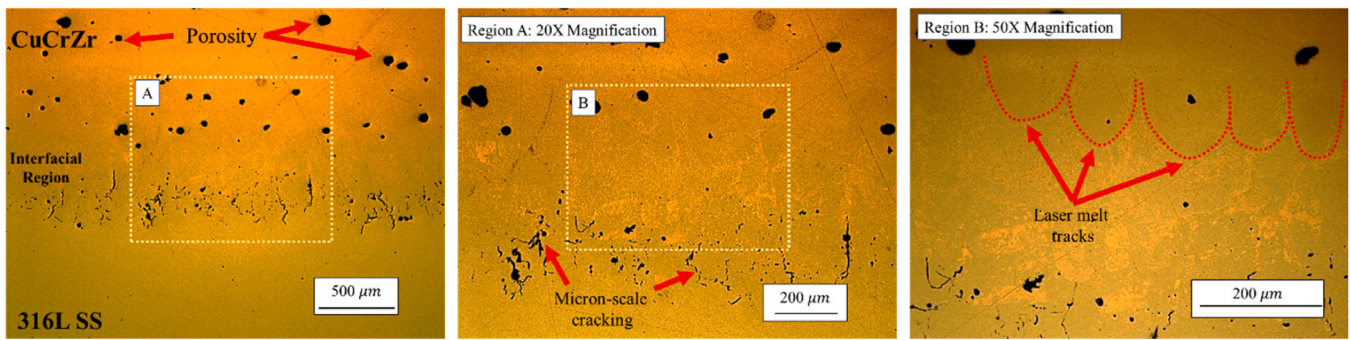


Fig. 5. Optical Microscopy of the interfacial region at 10X, 20X and 50X magnification for characterization of superficial defects near the interface.

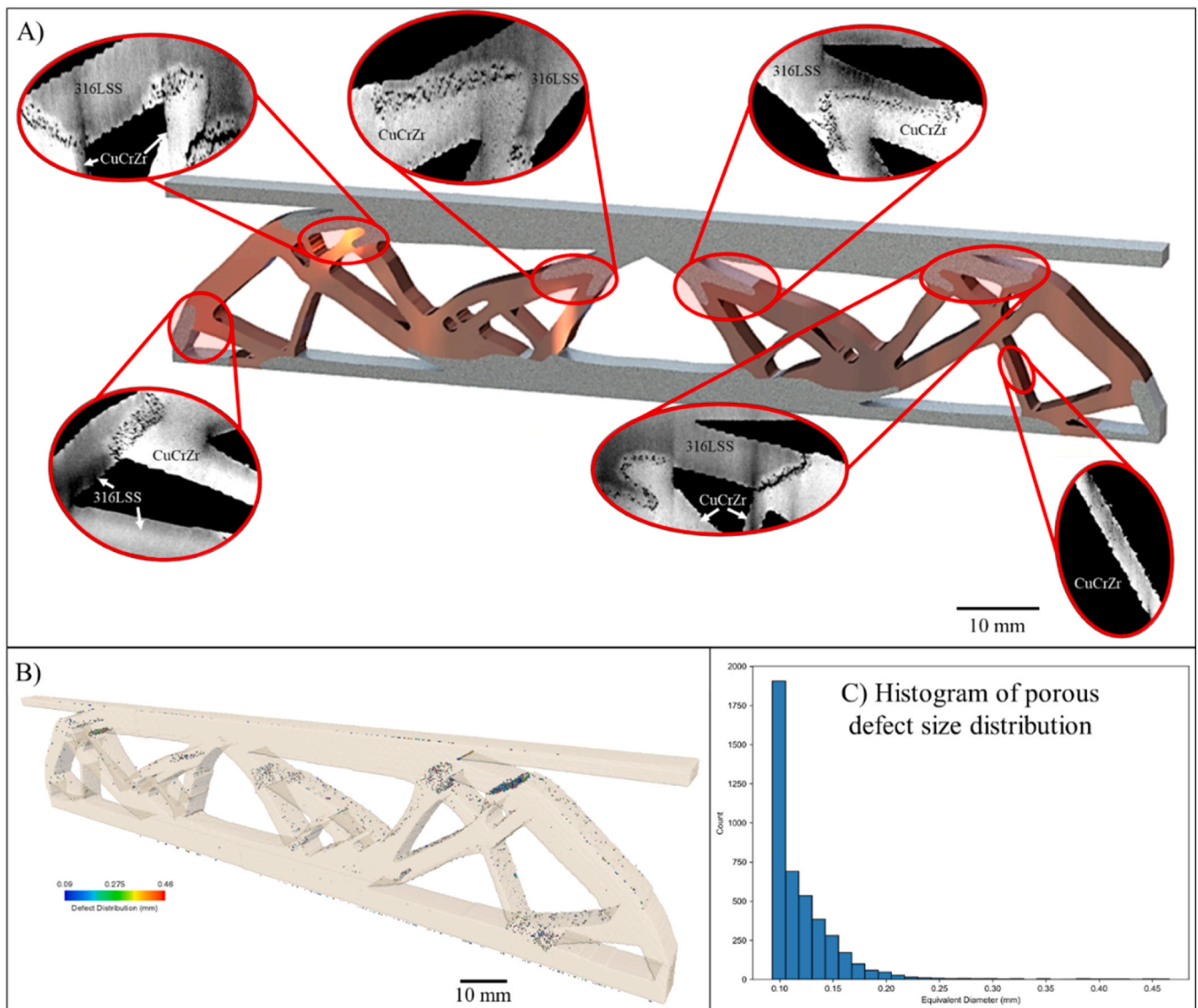


Fig. 6. Micro-CT analysis of MBB beam highlighting internal defect distributions within CuCrZr (brown) and 316 L stainless steel (grey) regions. A) CAD rendering of the beam with sectioned callouts at selected locations along the beam length. Each callout presents the corresponding micro-CT slice with material regions explicitly labelled. B) 3D reconstruction of the complete beam with the spatial distribution of internal defects overlaid. C) Histogram showing the pore size distribution measured across the full beam.

causing stepwise load drops but no immediate global collapse. Incremental strut failures continue until $\sim 15\%$ displacement, when high local tensile stresses emerge at the 316 L stainless steel–CuCrZr interface

above the supporting test cylinder.

Between 2.5 mm and 4 mm (Locations C and D, Fig. 9D, E), the same stages of failure are mirrored on the opposing side of the beam. Ongoing

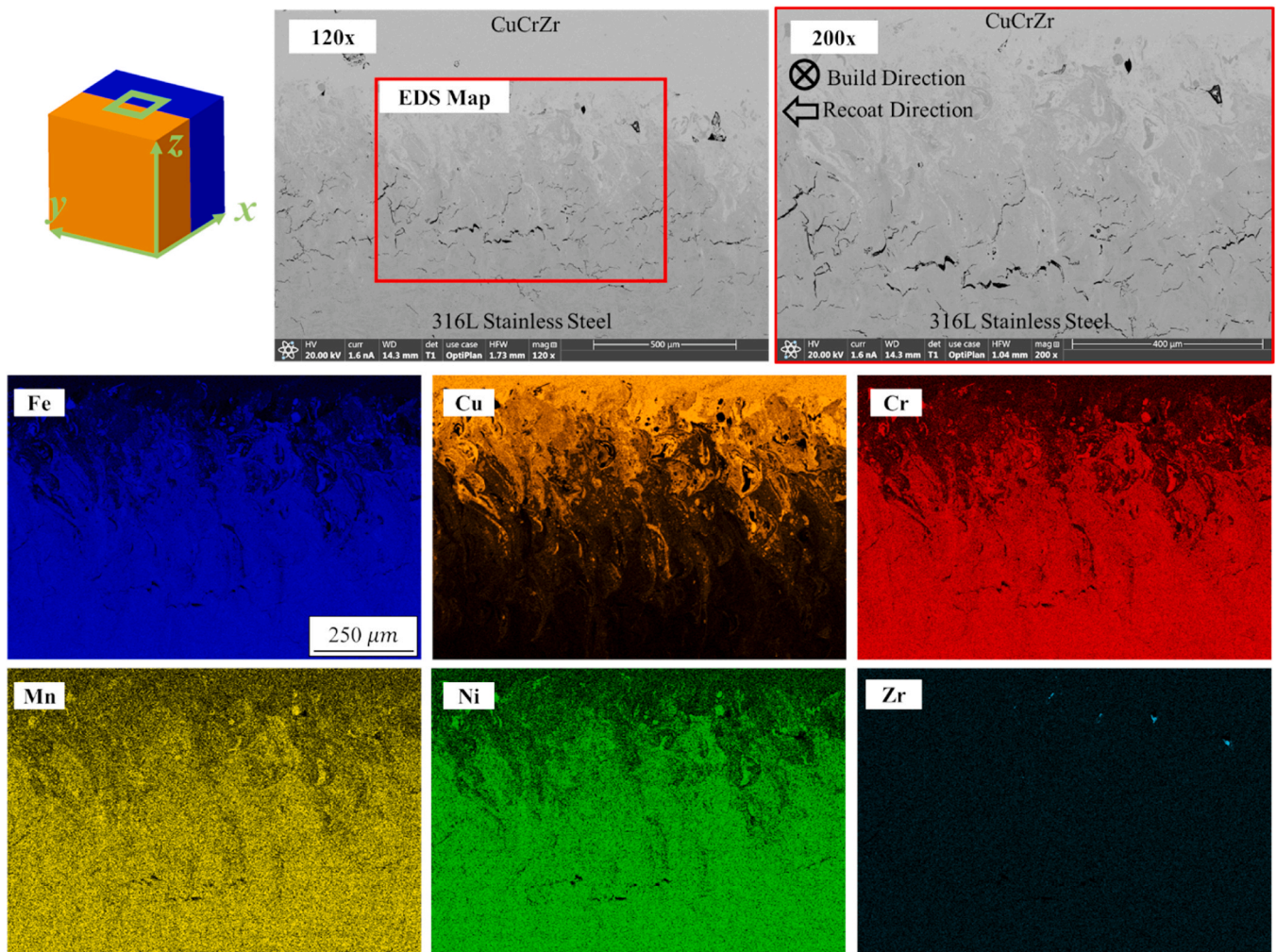


Fig. 7. A) Electron discharge spectroscopy (EDS) of the interfacial region at 200X magnification for interfacial characterization of superficial defects near the interface.

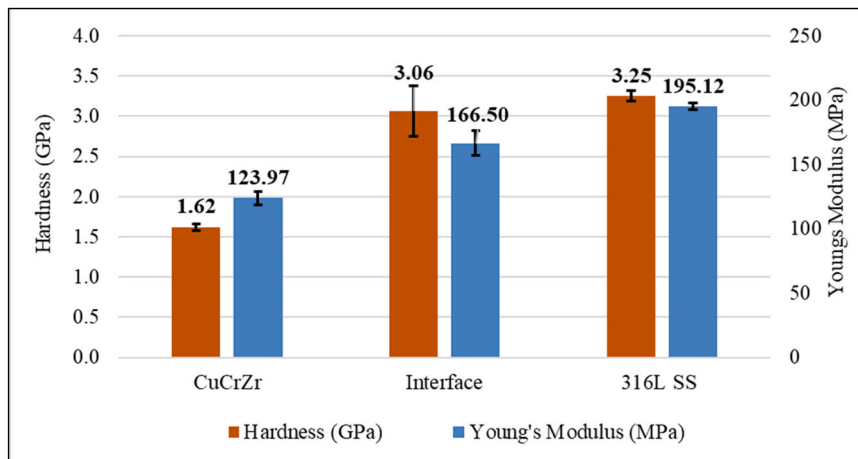
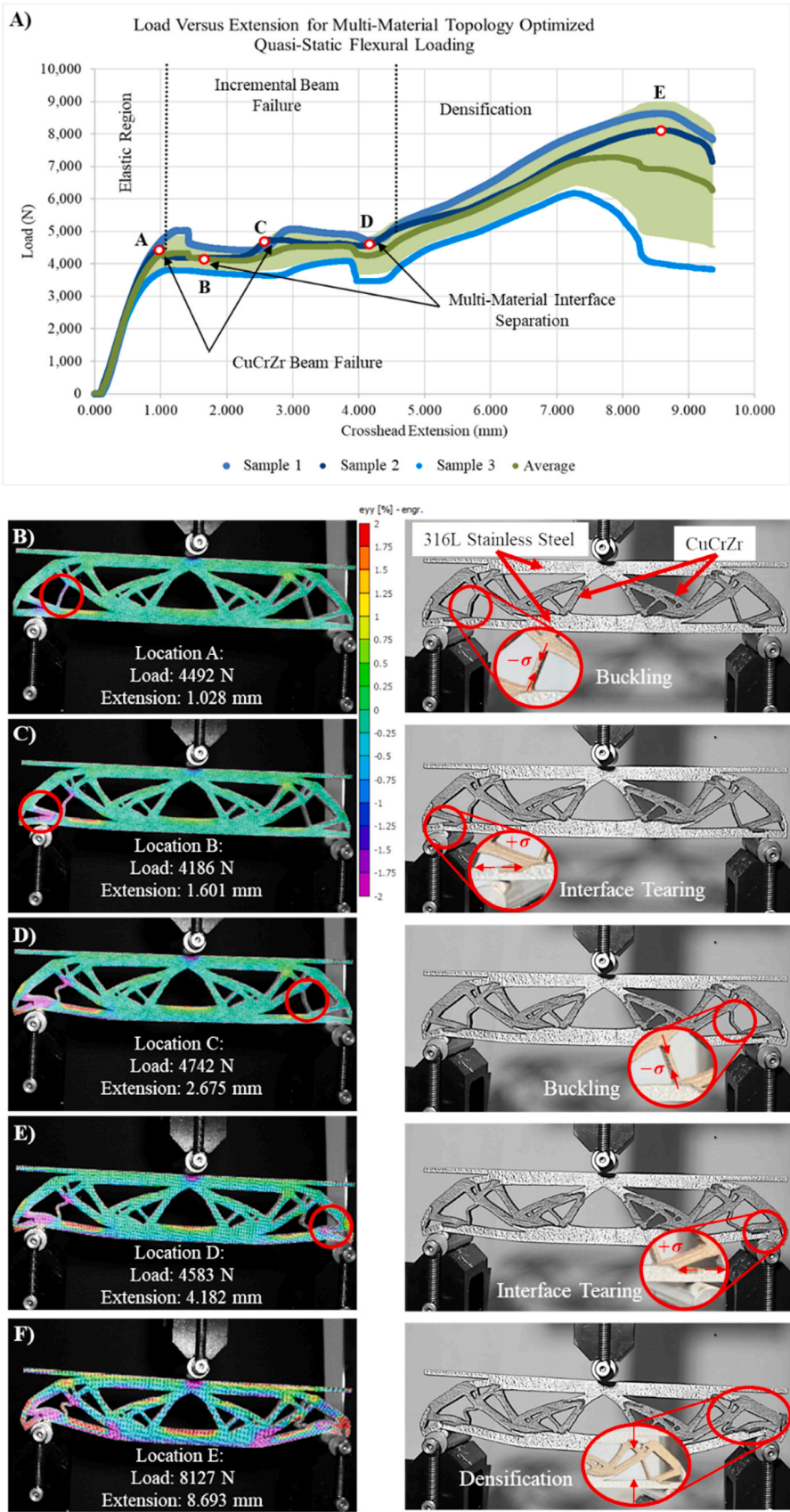


Fig. 8. Computed hardness and Young's Modulus across the CuCrZr and 316 L stainless steel interface manufactured via MM-LPBF taken from the microscopy witness coupon printed with the material interface perpendicular to the build plate, as seen in the interfaces presented in the MBB test specimen.

fractures and delamination initiate a “densification” phase from 4.5 mm to 9.5 mm, characterized by pronounced deformation and folding. Beyond ~9.5 mm, widespread strut fracture and extensive interfacial delamination result in final structural collapse, often including complete

failure of the largest outer struts. Mixed-mode fracture (buckling and tensile/interfacial delamination) dominates in the terminal failure stage.

FEA was performed on the multi-material MBB beams to evaluate the



(caption on next page)

Fig. 9. A) Experimental results from DIC analysis showing load-extension curves for all three flexural beams and average with standard deviation. The response is segmented into key deformation and failure stages: the initial elastic regime, progressive beam failure associated with strut buckling or interfacial tearing at locations B, C, and D, followed by beam densification, which is characteristic of strut-based structures. B-F) Representative stages (A-E) corresponding to the labelled points in the load-extension curve, showing DIC strain maps (left) and corresponding black and white images of the mechanical testing (right). The DIC maps illustrate strain distribution across the beam, while red circles indicate primary failure locations at each stage. The black-and-white images are provided to show the multi-material distribution during testing.

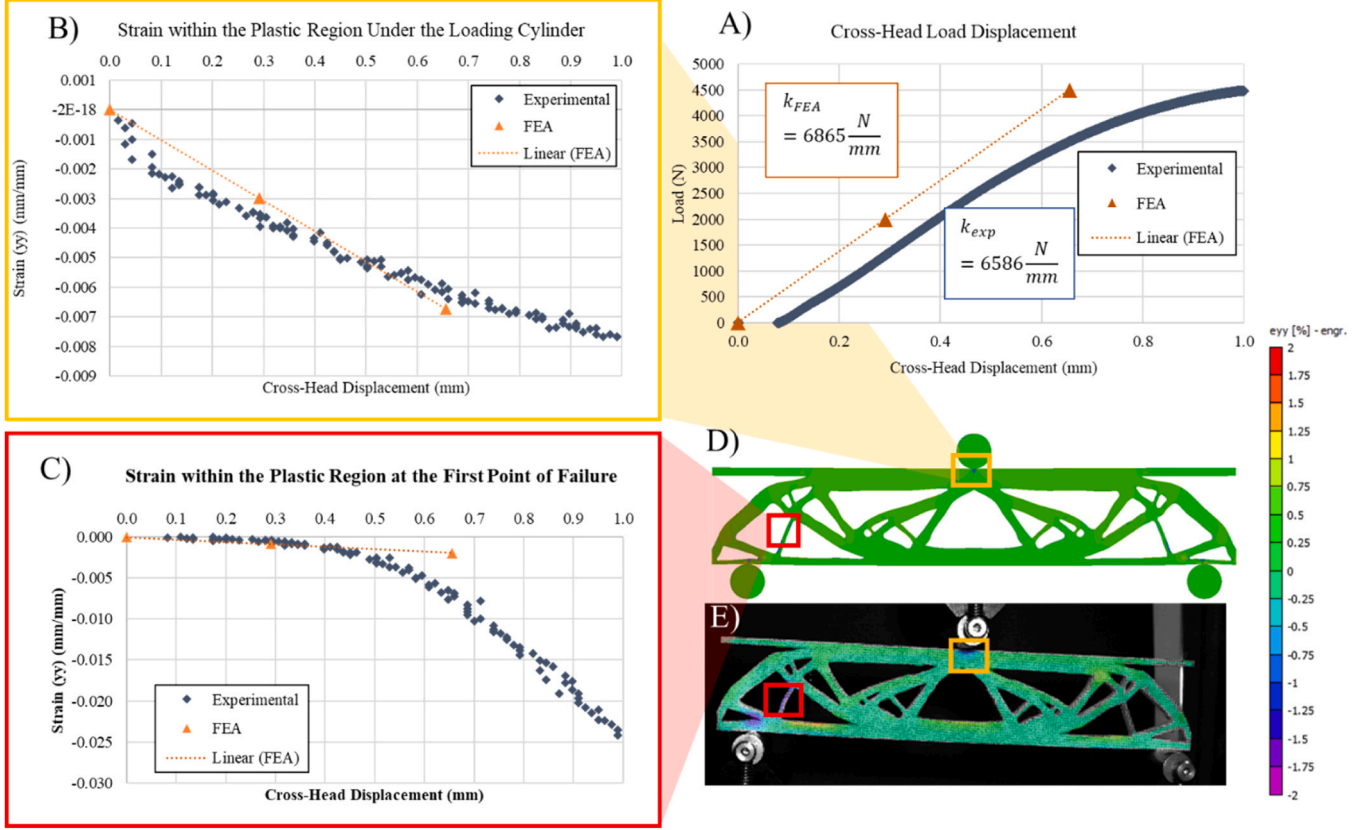


Fig. 10. A) Global cross-head load to displacement comparison within the elastic region, comparing experimental measurements and FEA predictions of stiffness. The experimentally measured stiffness and FEA-predicted stiffness are $k_{exp} = 6586 \frac{N}{mm}$ and $k_{FEA} = 6865 \frac{N}{mm}$, respectively, corresponding to a 5.3% difference. B) Point-probe comparison of strain as a function of crosshead displacement directly under the loading cylinder, obtained from DIC measurement and FEA results. C) Point-probe comparison of strain to cross-head displacement at the first point of failure. D) FEA-predicted strain field (y-normal strain, mm/mm) flexural loading and E) Experimental DIC strain both captured at 4500 kN, or location A presented in Fig. 9.

experimental mechanical response under flexural loading under ideal surface and density conditions. The FEA model incorporated the local material properties of as-built CuCrZr and 316 L stainless steel found using nanoindentation (Fig. 8) and assumed ideal bonding at bi-material interfaces. The simulations predict a stiffness of $k_{FEA} = 6865 \frac{N}{mm}$, which is 5.3% higher than the experimentally observed stiffness, $k_{exp} = 6586 \frac{N}{mm}$. This minor prediction error is attributed to model assumptions which do not fully capture physical imperfections, such as minor porosity in the bulk CuCrZr section, interfacial defect cracking, or residual stresses introduced during metal AM. These factors likely reduced the stiffness of the experimental beams relative to the idealized model.

4. Discussion

4.1. Computational design

To further analyze the effect of volume fraction on objective performance, a design sweep was conducted by varying the volume fraction of 316 L stainless steel, $V_{ss} = \{0, 0.15, 0.25, 0.35, 0.528\}$. The total mass remains constant by setting the volume fraction of CuCrZr following the equation:

$$V_{Cu} = \frac{(0.25 \cdot \rho_{ss} + 0.25 \cdot \rho_{Cu}) - V_{ss} \cdot \rho_{ss}}{\rho_{Cu}} \quad (4)$$

Optimization results with varying volume fraction constraints are reported in the Supplemental document Appendix B, and Fig. 11. As the volume fraction of 316 L stainless steel increases, the structural deformation decreases. This trend can be attributed to the higher stiffness of 316 L stainless steel compared to CuCrZr, which enhances the overall mechanical resistance of the structure.

The thermal performance of the structures was also evaluated by replacing the point load in Fig. 3 with a thermal load. The thermal load is applied as a heat source at the center of the top surface of the beam, while the beam is treated as a heat sink, as illustrated in Fig. 11. This additional analysis was conducted to understand how integrating a high-conductivity material, such as CuCrZr, with 316 L stainless steel may influence the heat dissipation towards achieving functionally optimized components that balance mechanical strength with effective thermal management. The measurement of thermal compliance (C_{th}) is an indication of thermal potential energy associated with the structure, equivalent to the elastic potential energy in the mechanical case, and a lower C_{th} corresponds to more efficient heat dissipation under the

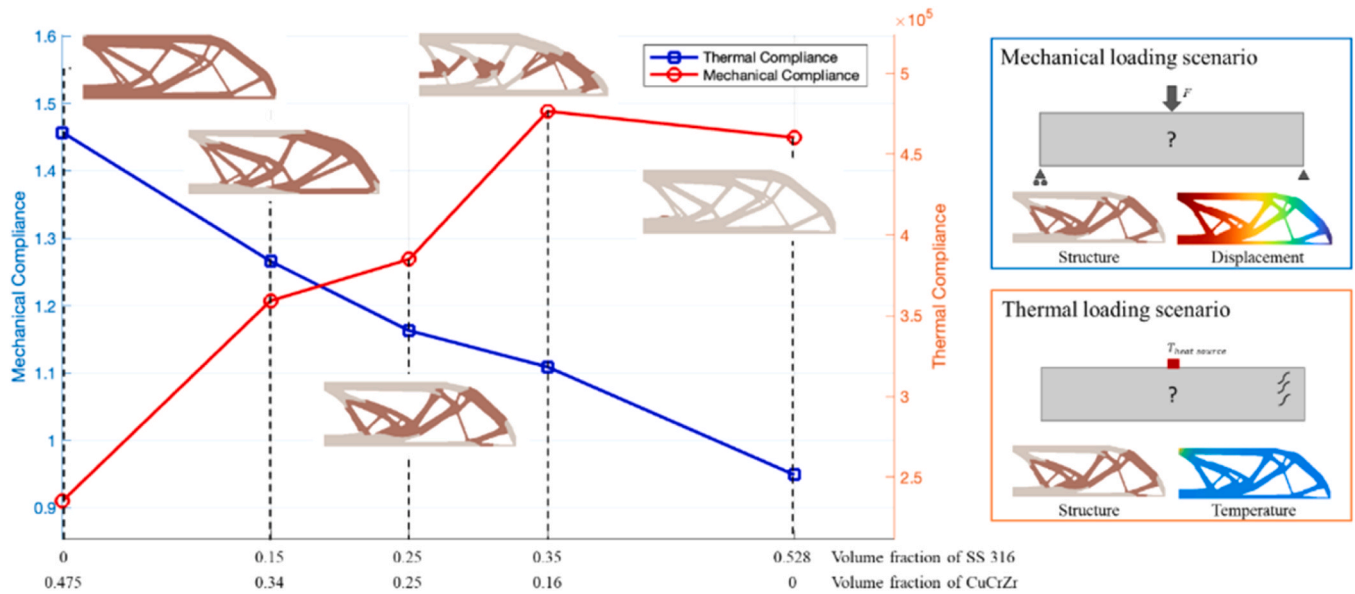


Fig. 11. Optimal structures obtained from AAP with varying volume fraction constraints while maintaining constant mass. The loading conditions for the mechanical and thermal analyses are provided, together with the resulting displacement and temperature fields.

prescribed thermal boundary conditions. Please note that the mechanical and thermal analyses were conducted separately on the structures. The CuCrZr's higher thermal conductivity leads to a general decrease in thermal compliance as its volume fraction increases. However, fluctuations occur when CuCrZr content falls below a certain threshold, likely due to an incomplete thermal pathway hindering effective heat dissipation. Future work will focus on multi-objective optimization, considering both mechanical and thermal performance at the design stage.

4.2. Interfacial and internal morphology

The printability of the beam highlights critical challenges inherent to MM-LPBF. A number of processing challenges local to the interface contribute to the presence of cracking or porosity defects (Fig. 7A). A mismatch of thermal conductivity and expansion between these two alloys presents unique thermal boundary conditions to the melt pool, and during solidification that are not apparent in single-material LPBF. Additional challenges pertaining to the solidification behavior of a mixed copper-based and austenetic steel melt pool are known to present material compatibility issues due to insolubility at high concentrations of copper, and an extreme difference in melting and solidification temperatures [1,68].

These processing challenges have shown to be dependent on the orientation of the interface, with respect to the build direction [1]. However, due to the inherent design complexity of this beam, there was no single orientation that avoided defect-prone orientations. As a result, the beam was oriented with all interfaces perpendicular to the build plate, which has shown to produce cracking defects at the interface for similar material combinations. SEM images (Fig. 7) at the interface show a pattern of micron-scale cracking adjacent to the interface, favoring the steel material sections. This patterned cracking perpendicular to the interface has been shown to impact mechanical performance [1,21], and is suggestive of material incompatibility. Additionally, an uptake in internal porosity was observed to localize near the interfacial regions, as evident in the micro-CT in Fig. 6. This defect behavior is likely thermally driven, as the stainless steel regions tend to retain heat due to a relatively low thermal conductivity, in comparison to the CuCrZr. This effect becomes more significant in areas adjacent to the interface, where a high volumetric energy density (VED) melt pool interacts with both materials. The resulting conductivity mismatch can induce localized melt

pool instability, promoting pore formation and incomplete bonding at the interface. Recent advancements increasingly treat the mixed-material interfacial region as a distinct third processing zone, with tailored laser parameters and scan strategies designed to mitigate localized thermal or metallurgical instabilities [69,70]. This defect formation mode can be further investigated through melt pool dynamics simulations using computational fluid dynamics (CFD) to capture interfacial flow behavior and transient heat transfer effects.

Local nanoindentation across the bi-metal interface of CuCrZr and 316 L stainless steel reveals a gradual transition in mechanical properties. In this study, hardness at the interface was observed to be intermediate between that of the bulk CuCrZr and 316 L stainless steel. Bulk CuCrZr manufactured via single-material LPBF has shown reported hardness ranging from 1.28 GPa (130 HV) to 1.88 GPa (192 HV) [66, 71]. The hardness reported adjacent to the interface, at CuCrZr indent location is consistent with this range. Printed 316 L stainless steel has been reported to have a hardness ranging from 1.92 GPa [72] to 3.68 GPa [73]. The hardness reported in this study at the 316 L stainless steel indent is again consistent within this range. At the interface, nanoindentation measurements suggest effective interfacial bonding and a heat-affected zone with modified microstructural characteristics local to the interface.

4.3. Incremental mixed-mode mechanical failure

During the initial phase of loading (up to approximately 10% (1 mm) of crosshead displacement), the beam, composed of 316 L stainless steel and CuCrZr segments, deforms within its elastic limits. Both constituent materials exhibit linear elastic behavior, as supported by the linear slope of the global load-extension curve in this region. DIC strain fields indicate a uniform distribution of strain, suggesting that neither localized yielding nor cracking has initiated in either alloy. Consequently, the beam behaves as a single, elastically deforming member without discernible signs of damage or plastic deformation.

The flexural test results of the multi-material MBB beam using DIC reveal a progressive and incremental failure pattern across all tested beams. Based on prior research on the mechanical testing of MM-LPBF components [1], failure was expected to originate from the interface due to the likelihood of defects (observed in both the micro-CT in Fig. 6 and SEM/EDS imaging in Fig. 7) and stress concentration due to material discontinuity at the boundary. Despite the localized porous and cracking

defects, the bending stress condition of this flexural test places most defect-prone regions under compressive stress, which will effectively compress and close these pores, and limit their contribution to the onset of failure. As the applied load surpasses approximately 4 kN, initial buckling is observed in a single-material strut, representing the first tangible sign of localized failure (Location A, Fig. 9B). This beam buckling does not cause an immediate, global collapse. Instead, the struts continued to fail incrementally, leading to characteristic “step-like” load drops in the load-displacement curve. Each fracture event briefly reduces the overall carrying capacity, followed by partial load recovery until the occurrence of the next localized failure.

Beyond approximately 15% (1.5 mm) of deflection, high local tensile stresses concentrate at the interface between the 316 L and CuCrZr layers, particularly in regions directly above the supporting test cylinder (Location B). In multi-material systems, any bonding imperfections or mismatches in material properties can predispose the interface to delamination. Internal porosity can arise during LPBF processes due to insufficient or excess laser energy input while melting powders [74]. Internal pores serve as stress concentrators, increasing localized stresses and accelerating crack initiation under load [74,75]. These defects also reduce the effective load-bearing area, leading to a larger magnitude of stress distributed across a smaller cross section. The combination of tensile forces combined with the onset of micro-cracks frequently initiates an interfacial fracture. SEM and optical microscopy imaging (Fig. 5 and Fig. 7) of the microscopy coupons confirm the presence of micron-scale cracking and porosity along or near these interfaces.

After one side of the beam experiences CuCrZr buckling and interfacial delamination, a mirrored sequence of failures occurs on the opposite side at higher displacements from 25% to 40% (~2.5 mm to 4 mm) at Locations C and D, respectively. These discrete fracture steps provide insight into the progressive failure mechanism. As further incremental fractures occur and interfacial delamination propagates, increasing portions of the beam lose load-carrying capacity. The progression culminates in what can be described as “densification,” wherein broken or partially fractured segments collapse onto one another, producing a more compacted or folded cross-section [76]. This stage was observed from ~50% - 100% (4.5 mm and 9.5 mm) of crosshead displacement, and is marked by substantial permanent deformation.

Ultimately, beyond approximately 100% (9.5 mm) extension, the combined effects of numerous fractured struts and widespread interfacial delamination render the beam unable to sustain additional load. The load-extension curve then exhibits a final drop, indicating overall structural collapse. In many tests, the largest outer struts in the beam subsequently fail completely. At this stage, the failure mechanism is often characterized by a mixed-mode combination of buckling and tensile/interfacial delamination, reflecting the complexity inherent in multi-material systems under bending loads.

Incremental stages of failure for complex structures fabricated via LPBF have been shown first in single material lattice and scaffolding structures [77]. More recently, few papers have explored the failure mechanisms of multi-material lattice structures. Another study investigated bi-metallic lattices under compressive loading of 316 L stainless steel and 17-4PH steel [18]. They found that when the materials were loaded in series, they experienced sequential failure in parallel. Here, the more ductile material failed first, which then propagated the load unto the stronger 17-4PH steel which failed subsequently at a higher load. Likewise, another study also investigated the progressive failure of multi-material lattice structure loaded in parallel [19]. They found an initial collapse of the weaker material section, followed by the second catastrophic collapse of the stronger material section, and a final densification of the struts. These studies highlight the interaction of localized material properties in multi-material complex structures, where the mechanical response is governed by non-uniform stress distribution across the test specimens. This phenomenon also emphasizes the importance of understanding and predicting the load redistribution mechanisms of complex organic geometries produced from

computational design tools.

5. Conclusion

This study represents a significant step towards expanding the application domain of multi-metal additive manufacturing through evolving computational design tools for an optimal regional distribution and allocation of materials. By adapting an AAP algorithm to incorporate the manufacturing considerations of MM-LPBF, this work addresses critical gaps in current manufacturing and design research to demonstrates the manufacturing feasibility of optimized multi-metal parts with specific performance objectives. This work provides a comprehensive exam of the mechanical behavior and failure mechanisms of a multi-metal design produced through topology optimization and manufactured via MM-LPBF using the classical MBB boundary conditions. The findings reveal progressive failure of the beam is determined by as-built pores in bulk CuCrZr material sections and across the bi-material interface. Bulk CuCrZr defects are likely a result of both its well-known difficulty in processing this material due to its high reflectivity and thermal conductivity, as well as evolving complex thermal boundary conditions in regions adjacent to the stainless steel. Structural FEA computed a numerical beam stiffness 5.3% higher than that of the experimental beam. The findings of this investigation open new possibilities for advanced engineering applications of multi-material components, enabling the production of high-resolution, multifunctional metallic components with enhanced regional performance and reliability.

Immediate future efforts will extend this MMTO framework to support multiple objective functions, enabling material placement driven by competing structural and thermal boundary conditions, and to three-dimensional design spaces to fully exploit the geometric freedoms of inter-layer MM-LPBF. In parallel, a process-aware MMTO formulation will be developed by incorporating manufacturing constraints beyond machine resolution, such as orientation effects, and representative defect characteristics (e.g., lack-of-fusion, keyhole porosity, and interface cracking) through probabilistic or defect-tolerant approaches. However, first-principles models developed for single-material LPBF are not directly transferable to MM-LPBF, and defect prediction in multi-material systems remains an open challenge. Physics-based models capable of reliably predicting defect type, size, and spatial distribution as functions of process parameters and material combinations are not yet mature for integration into a TO loop. Accordingly, future work will focus on improved characterization of interface-driven failure mechanisms and the incorporation of MMAM constraints and material variability into the MMTO framework, including elastoplastic models derived from experimental post-yield behavior, to enable more accurate prediction of deformation and failure beyond the elastic regime.

CRedit authorship contribution statement

Jacklyn Griffiths: Writing – review & editing, Writing – original draft, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **Jier Wang:** Writing – review & editing, Writing – original draft, Data curation, Conceptualization. **Kaitlyn Hartman:** Writing – review & editing, Formal analysis, Data curation. **Avez Shaikh:** Writing – review & editing, Formal analysis. **Xiaochen Yu:** Writing – review & editing, Writing – original draft, Data curation, Conceptualization. **Jeremy Schreiber:** Writing – review & editing, Supervision, Methodology, Funding acquisition, Formal analysis. **Ajit Panesar:** Writing – review & editing, Supervision, Conceptualization. **Guha Manogharan:** Writing – review & editing, Supervision, Methodology, Funding acquisition, Conceptualization.

Declaration of Competing Interest

The authors declare that they have no known competing financial

interests or personal relationships that could have appeared to influence the work reported in this paper.

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Appendix A. Supporting information

Supplementary data associated with this article can be found in the online version at [doi:10.1016/j.addma.2026.105247](https://doi.org/10.1016/j.addma.2026.105247).

Data availability

Data will be made available on request.

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