



LPBF-fabricated superelastic Ti–Zr–Nb open cell lattice structures for orthopedic implants: Process optimization, geometric conformity, microstructure, static and fatigue mechanical behavior

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ABSTRACT

Additively manufactured orthopedic implants are required to simultaneously offer a high geometric accuracy, a low elastic modulus, and an excellent fatigue resistance. In this study, we optimized laser powder bed fusion (LPBF) process parameters to fabricate three open cell lattice structures from a superelastic Ti–Zr–Nb alloy: a rhombic dodecahedron, a sheet gyroid, and a Voronoi polyhedral. In these structures, the strut thickness ranged between 200 and 400 μm and the functional porosity, between 65 and 80%. Geometric conformity and defect porosity were quantified using microcomputed tomography, revealing that the optimized processing limited the maximum deviations in functional porosity to 9%, with the gyroid architecture demonstrating the highest dimensional accuracy. At the same time, the defect porosity in all the structures did not exceed 0.2%. After the heat treatment, microstructural analysis and mechanical testing confirmed the presence in these structures of metastable β -phase. Under quasi-static compression, the lattice specimens manifested a yield plateau associated with stress-induced martensitic transformation. Fatigue testing of Ti–Zr–Nb structures demonstrated a stable cyclic performance and competitive fatigue resistance up to 5×10^6 cycles as compared to their benchmark Ti–6Al–4V counterparts. The results obtained highlight the potential of LPBF-fabricated superelastic Ti–Zr–Nb lattice specimens for load-bearing implant applications, owing to their reduced stiffness and enhanced long-term mechanical durability.

1. Introduction

The difference in Young's modulus between bone tissue (typically 0.01–22 GPa depending on the bone type) and an implant material (≥ 100 GPa for conventional titanium alloys) generates stress shielding, reduces osteoblast activity and potentially leads to bone resorption and implant loosening [1]. Porous structures address this key problem of biomechanical compatibility effectively by significantly reducing the implants' stiffness to match that of bones [2,3]. Additionally, the developed internal surface of porous structures is generally considered favorable for osseointegration and bone tissue ingrowth, which may enhance the mechanical bond between implant and bone and enable a more uniform transfer of physiological loads. Among all the available manufacturing processes, additive manufacturing (AM) technologies,

such as laser powder bed fusion (LPBF) and electron beam melting (EBM), are the best suited to manufacture metallic load-bearing porous orthopedic implants [4–8]. LPBF offers higher manufacturing precision than EBM, making it preferable when producing porous structure-containing components [9,10].

To date, Ti–6Al–4V, the most widely used metallic material in medical practice, has most effectively realized the potential of 3D-printed orthopedic implants with porous structures [4,6–8,11,12]. By integrating open cell lattices, metallic biomaterials with a unique combination of low stiffness, high strength, and favorable osseointegration conditions have been developed. A wide range of porous architectures has been proposed for such applications, including regular strut-based, triply periodic minimal surface (TPMS), and stochastic structures [13]. These architectures differ in their stiffness, strength, and deformation

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behavior, thereby offering different design opportunities for load-bearing applications [14]. Auxetic metamaterials, which exhibit a negative Poisson's ratio, have also emerged as promising candidates because they may provide a favorable balance between strength and stiffness [15,16].

Oseointegration effectiveness depends directly on the implant porosity attributes [8,17–19]. At the same time, increasing functional porosity generally improves conditions for bone ingrowth and vascularization but may reduce the load-bearing capacity of the scaffold, which creates a fundamental design trade-off [20]. Analysis of the lattice structure characteristics presented in Ref. [8] shows that β -titanium alloys offer a significantly wider range of acceptable design parameters (functional porosity, cell size, and mechanical properties) than conventional $\alpha+\beta$ Ti–6Al–4V alloys. This advantage is mainly due to the substantially lower elastic modulus of β -titanium alloys, which broadens the room for design optimization and enables a better balance between biological integration and mechanical reliability. Consequently, many researchers have focused on the development of β -titanium-alloy-based implants using AM [21–27].

Among the most promising material candidates are metastable beta Ti–Zr–Nb alloys which are made exclusively from highly biocompatible components [28,29]. Depending on their composition and processing, Ti–Zr–Nb alloys demonstrate a Young's modulus as low as 40–60 GPa and a room temperature superelasticity with a recovery strain as high as 2.0–3.5% [30–32]. It can be reasonably suggested that this superelastic behavior would create better stress sharing conditions in porous structures, which are otherwise characterized by strongly non-uniform stress and strain distributions under load, thus improving their functional fatigue resistance. Prior works on LPBF-fabricated porous Ti–Ni shape memory alloys have demonstrated that open-cell architectures can retain superelastic functionality while substantially reducing the apparent stiffness [33,34].

In our previous studies on Ti–Zr–Nb alloys [35–37], the LPBF process and post-process conditions were optimized to produce fully dense (solid) parts with a low defect porosity (<0.4%) and an enhanced combination of mechanical and functional properties. In particular, these studies showed the need to adjust the chemical composition of a powder feedstock to compensate for titanium evaporation during processing, and therefore, to improve the extent of room temperature superelasticity [31]. Experimental studies have demonstrated the effectiveness of these adjustments, which yielded excellent bulk mechanical and functional properties, suitable for orthopedic implants, i.e., an ultimate tensile strength of $UTS = 725$ MPa, a Young's modulus of $E = 53$ GPa, an elongation to failure of $\delta = 16.5\%$, and a maximum recovery strain of $\epsilon_{max}^{se} = 2.6\%$ [31].

It can therefore be expected that transferring this combination of low elastic modulus and high mechanical resistance from bulk to porous material would provide a promising route toward orthopedic implants with reduced stress shielding and sufficient space for bone ingrowth and vascularization. However, manufacturing porous structures by LPBF presents specific challenges. LPBF process parameters for open cell lattice structures differ significantly from those for bulk components because a much higher precision is required to produce small features [38]. In addition, minimizing internal defects, especially pores within struts, is critical for improving the functional and mechanical properties of LPBF-fabricated porous structures. Such defects may form under both insufficient and excessive energy input, which alters powder melting and solidification conditions [39]. Because the reduced metal volume in porous structures changes the heat dissipation conditions, the melting process may become architecture-dependent, potentially causing excessive remelting, distortion, and increased defect sensitivity [6]. For structures with strut thicknesses as small as 0.2–0.3 mm, geometric accuracy and defect porosity become particularly critical, requiring structure-specific optimization of LPBF parameters [40]. One solution involves correcting the laser trajectory by reducing the distance between passes, which requires adjusting other LPBF parameters to maintain an

optimal level of energy input. However, changes in the heat transfer conditions during printing of porous Ti–Zr–Nb materials inevitably affect the microstructure and crystallographic texture formation, thus influencing the alloy's mechanical behavior, which is another area that needs to be explored.

The mechanical response of porous materials with defined internal geometry under cyclic loading is both scientifically interesting and practically important. Load-bearing implants experience repeated physiological loading in the human body and therefore require high fatigue strength. Since porous structures are generally weaker and more defect-sensitive than bulk materials, selecting an appropriate constitutive material, lattice design, and processing route is essential for maximizing fatigue performance [41,42]. Although a number of studies have evaluated fatigue behavior of additively manufactured lattice structures using standardized test approaches [42,43], systematic long-life fatigue data for LPBF-fabricated Ti–Zr–Nb open cell lattice structures remain limited. In addition, direct benchmarking against Ti–6Al–4V porous counterparts tested under identical conditions is still lacking.

In this work, we develop and optimize the LPBF process parameters to fabricate open cell lattice structures from superelastic Ti–Zr–Nb alloys using three representative architectures, i.e., a rhombic dodecahedron, a sheet gyroid, and a Voronoi polyhedra with a strut thickness ranging from 0.20 to 0.40 mm. We quantify the geometric fidelity relative to the CAD models of printed structures and analyze how the LPBF parameters and lattice geometries affect the melt-pool formation, the microstructure/texture, and the defect-related porosity of a final product. Finally, we compare the mechanical behavior of Ti–Zr–Nb printed structures with their Ti–6Al–4V equivalents under monotonic and cyclic loading to up to 5×10^6 cycles, demonstrating the potential of LPBF Ti–Zr–Nb open cell lattice structures for orthopedic applications.

2. Methodology

2.1. Material processing

Two Ti–Zr–Nb powder lots (#1 and #2) were produced using an identical technological workflow. Firstly, two 160 mm diameter – 200 mm long Ti–Zr–Nb ingots were fabricated by vacuum arc remelting (VAR). The ingots underwent multi-axis hot forging in the 900–1050 °C temperature range, reducing the diameter to 80–85 mm. The forged semi-finished products were then processed by electrode induction melting gas atomization (EIGA), and the resulting powders were sieved to obtain particle size distributions of $d_{10} = 15.8$, $d_{50} = 29.2$, $d_{90} = 50.3$ for Powder #1 (μm), and $d_{10} = 15.7$, $d_{50} = 32.9$, $d_{90} = 55.6$ for Powder #2 (μm). Finally, Powder #1 was employed for optimizing the LPBF process parameters, while Powder #2 was utilized for fabricating lattice specimens for static and fatigue mechanical testing.

Fig. 1 presents the SEM images of sieved powders, powder size distributions, and corresponding energy-dispersive X-ray spectroscopy (EDXS) maps of the particles. From Fig. 1a–d, it can be seen that both powder lots contain predominantly spherical particles with nearly Gaussian particle size distributions (Fig. 1b–e). The EDXS analysis of the particles reveals a uniform distribution of the major elements (Ti, Zr, Nb) (Fig. 1c–f). The apparent density and flow rate of both powder lots were determined according to the ISO 3923-1:2008 and ISO 4490:2018 standards, yielding 2.87 g/cm³ and 27 s/50 g for Powder #1, and 2.90 g/cm³ and 28 s/50 g for Powder #2, respectively (see chemical compositions in Table 1).

Note that the chemical composition of Powder 2 corresponds more closely to a target Ti–Zr–Nb alloy composition of ~67 at% Ti; ~18 at% Zr; ~15 at% Nb, potential exhibiting room temperature superelasticity in printed parts (see previous work on LPBF of a fully dense Ti–Zr–Nb [31]). Note also that differences in the powder particle size distributions and also in the ingot, powder, and printed parts chemical compositions are resulted from the VAR, EIGA and LPBF processes-related variabilities. While transforming ingot to powder and then powder to printed

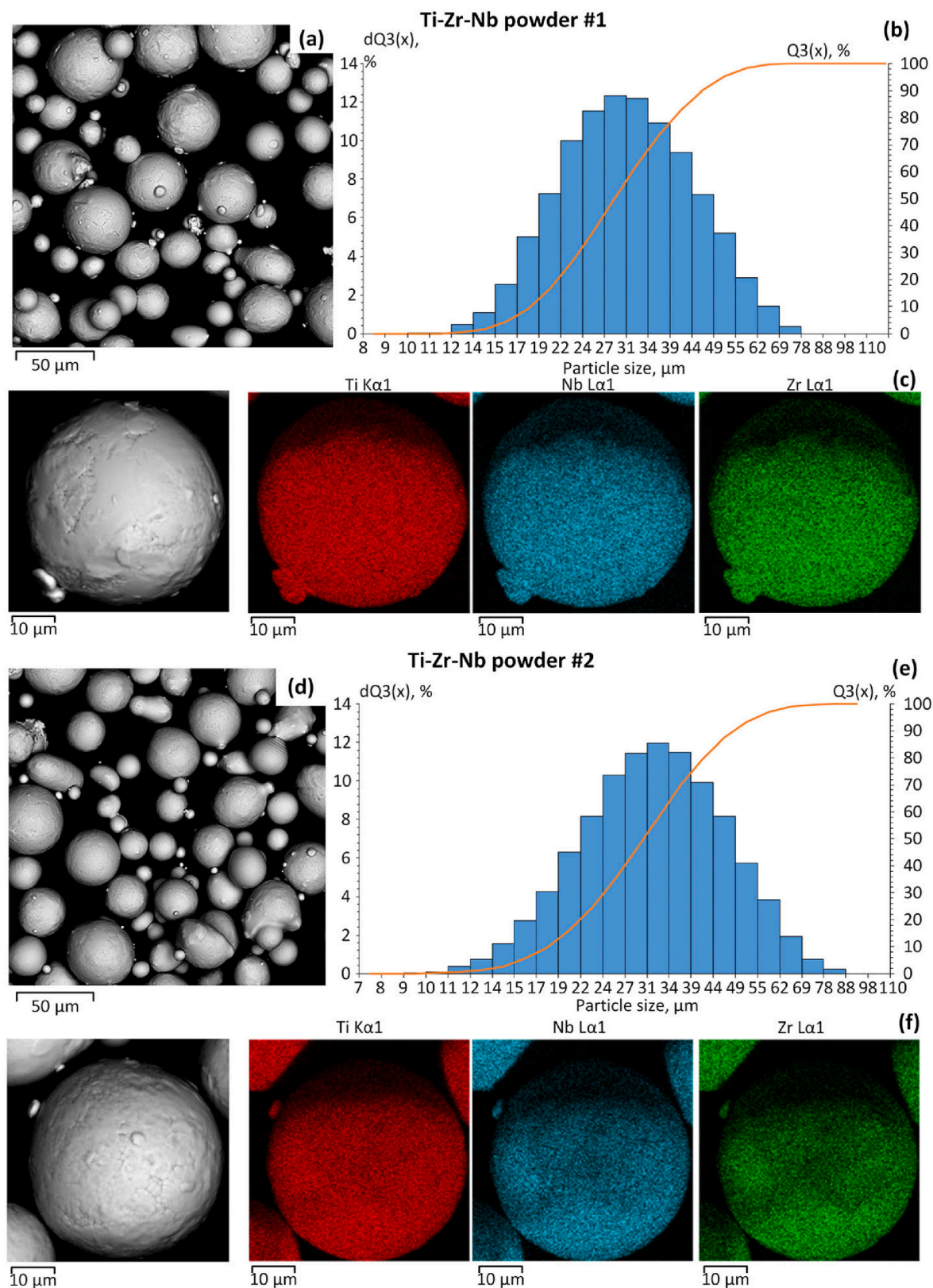


Fig. 1. (a, d) SEM images, (b, e) particle size distributions, and (c, f) EDXS maps of Ti-Zr-Nb powders (a-c) #1 and (d-f) #2.

part, the titanium content decreased from 66.3 to 64.3 for Powder #1 and from 68.7 to 67.0 for Powder #2 (at. %), while the oxygen content increased from ~ 0.075 to ~ 0.2 wt % for both materials, which confirm the previously observed changes in the chemical compositions of Ti-Zr-Nb alloys during the powder atomization and LPBF processes [29,

32].

2.2. LPBF parameter sets for bulk and lattice structures

A TruPrint 1000 (TRUMPF GmbH, Ditzingen, Germany) laser pow-

Table 1

Chemical composition of the Ti–Zr–Nb ingots, powder lots and LPBF-processed parts.

State		Main Elements (at. %)			Interstitial Elements (wt. %)			
		Ti	Zr	Nb	O	C	H	N
#1	Ingots	66.3 ± 0.5	18.2 ± 0.5	15.5 ± 0.5	0.090 ± 0.003	0.014 ± 0.003	0.004 ± 0.001	0.001
	Powder	65.2 ± 0.4	18.5 ± 0.4	16.3 ± 0.4	0.163 ± 0.001	0.014 ± 0.002	0.005 ± 0.001	0.001
	LPBF	64.3 ± 0.5	19.0 ± 0.5	16.7 ± 0.5	0.210 ± 0.005	0.016 ± 0.003	0.005 ± 0.001	0.001
#2 [31]	Ingots	68.7 ± 0.5	17.5 ± 0.5	13.8 ± 0.5	0.056 ± 0.003	0.018 ± 0.003	0.005 ± 0.001	0.001
	Powder	67.7 ± 0.4	17.7 ± 0.4	14.6 ± 0.3	0.092 ± 0.001	0.016 ± 0.002	0.005 ± 0.001	0.001
	LPBF	67.0 ± 0.5	18.2 ± 0.5	14.8 ± 0.5	0.167 ± 0.006	0.017 ± 0.002	0.005 ± 0.001	0.002

der bed fusion (LPBF) system equipped with a 200 W ytterbium fiber laser (30 μm spot diameter) was used. Based on the combined modeling and experimental validations reported in Ref. [36], the process window for printing high-density Ti–Zr–Nb parts was defined in the volume energy density (VED) (1) – build rate (BR) (2) coordinates in Fig. 2a:

$$VED(J/mm^3) = \frac{P}{v \times h \times t} \quad (1)$$

$$BR(cm^3/h) = v \times h \times t \quad (2)$$

and in the linear energy density (LED) (3) – scanning speed (v) coordinates in Fig. 2b:

$$LED(J/mm) = \frac{P}{v} \quad (3)$$

In addition to the VED and LED ranges identified in Fig. 2a and b, these figures contain one parameter set that was used to manufacture high-density ($\geq 99.5\%$) bulk reference specimens, i.e., $VED = 72 J/mm^3$ and $BR = 7 cm^3/h$, $LED = 0.129 J/mm$ and $v = 1080 mm/s$, the hatching distance $h = 0.06 mm$, and the layer thickness $t = 0.03 mm$. This parameter set was subsequently used as a baseline for designing the LPBF process for lattice structures.

Next, the scanning strategy was adjusted to allow the manufacture of thin struts in lattice structures. Specifically, while keeping a layer thickness of 0.03 mm, the hatching distance was reduced from 0.06 to 0.04 mm, which increased the number of single laser tracks within a 0.2 mm-wide region from 3 to 5 and decreased the h/t ratio from 2.0 to 1.5, thus increasing the degree of overlaps and therefore, the fraction of remelted material. This strategy was already used to improve the printability of thin elements in Refs. [27,44].

Furthermore, for further improving the dimensional accuracy of printed struts by reducing the size of a melt pool, it was decided to decrease simultaneously the laser power and the scanning speed values within the recommended VED-BR and LED-v spaces. These modifications yielded 16 LPBF parameter sets (Table 2) evenly distributed within the recommended VED-BR (LED-v) space and covering the 0.04–0.11 J/mm LED (Fig. 2b) and 35–90 J/mm³ VED values (Fig. 2a).

Table 2LPBF process parameters (layer thickness $t = 30 \mu m$; $h = 60 \mu m$ (baseline) and $40 \mu m$ (1–16 printing parameter sets)).

Regime #	P, W	v, mm/s	LED, J/mm	VED, J/mm ³	BR, cm ³ /h
Baseline	140	1080	0.126	72.00	7.00
1	140	1600	0.088	72.92	6.91
2	115	1600	0.072	59.90	6.91
3	140	1300	0.108	89.74	5.62
4	115	1300	0.088	73.72	5.62
5	90	1300	0.069	57.69	5.62
6	75	1300	0.058	48.08	5.62
7	60	1300	0.046	38.46	5.62
8	90	1000	0.090	75.00	4.32
9	75	1000	0.075	62.50	4.32
10	60	1000	0.060	50.00	4.32
11	45	1000	0.045	37.50	4.32
12	60	700	0.086	71.43	3.02
13	45	700	0.064	53.57	3.02
14	30	700	0.043	35.71	3.02
15	45	500	0.090	75.00	2.16
16	30	500	0.060	50.00	2.16

2.3. Lattice structures design and fabrication

To study the influence of LPBF regimes on the melt pool microstructure and defect porosity in bulk and arbitrary lattice specimens, and ultimately select the most favorable LPBF regime, both the bulk ($2 \times 3 \times 3 mm$ rectangular parallelepipeds) and porous rhombic dodecahedron unit cell specimens were printed. The lattice specimens were fabricated with two configurations: S0.25 (Fig. 3a) and S0.34 (Fig. 3b). The S0.25 specimens had a strut thickness of 0.25 mm, a unit cell size of 1.45 mm, overall dimensions of $5 \times 5 \times 5 (mm)$ and a design porosity of 72.6%. The S0.34 specimens had a strut thickness of 0.34 mm, a unit cell size of 2.0 mm, overall dimensions of $6 \times 6 \times 6 (mm)$ and a design porosity of 74.8%. To study the melt pool structure during their processing, additional S0.25 and S0.34 specimens were fabricated with pre-cut vertical planes (Fig. 3a and b).

Next, to study the static and fatigue mechanical behavior of porous materials, three different lattice structures were designed: rhombic dodecahedron (D), sheet gyroid (G), and Voronoi polyhedra (V). The

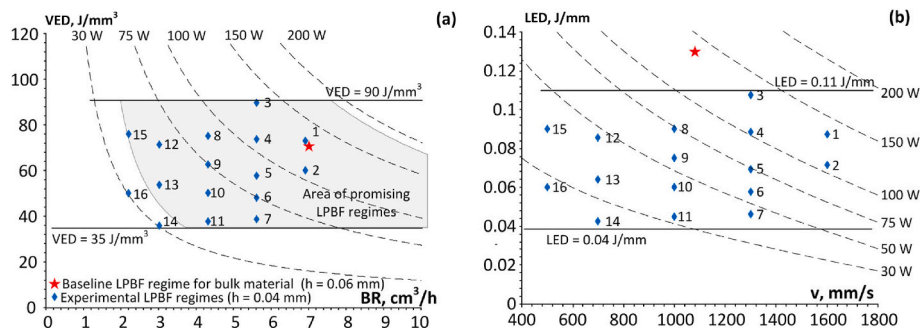


Fig. 2. LPBF process maps of developed LPBF regimes (red star for the baseline ($h = 0.06 mm$) and blue rhombus for experimental regimes ($h = 0.04 mm$) in the (a) VED-BR and (b) LED-v coordinates.

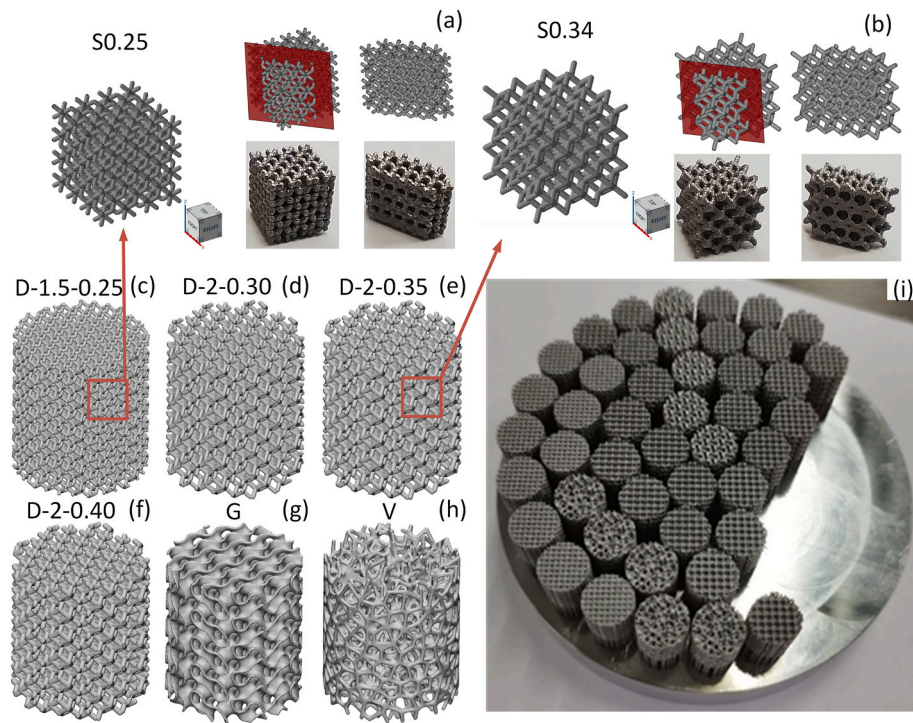


Fig. 3. CAD models and photographs of the experimental specimens of (a) S0.25 and (b) S0.34; CAD models of (c) D-1.5-0.25, (d) D-2-0.30, (e) D-2-0.35, (f) D-2-0.40, (g) G, and (h) V-type lattice specimens; (i) LPBF-ed specimens on the platform.

geometric characteristics of these structures were chosen considering both literature-based design criteria relevant to osseointegration and the processing capabilities [8]. For optimal osseointegration [45], improved vascularization and bone tissue formation [8,17], the pore size ranged from 0.1 to 1 mm. To keep the strength of lattice structures higher than that of bone tissue, while providing enough room for bone ingrowth, the maximum porosity level was limited to 75–80% [20]. Finally, to allow these structures to be printed using conventional LPBF equipment, the minimum strut and wall thicknesses were limited to 0.20 mm [8,46]. Considering the above requirements, 10 mm diameter – 11.3 mm height D-, G-, and V-type specimens with an equivalent porosity of 70–80% were designed using Materialize 3-matic software (Fig. 3c–h).

In more detail, the D-type lattice structures were selected to investigate how the lattice geometry and the functional porosity influence mechanical properties. Four D type specimens were designed to this end: D-1.5-0.25 with a $1.5 \times 1.5 \times 1.5$ mm unit cell size, a 0.25 mm strut size, and a 76% design porosity (Fig. 3c) and also D-2-0.3, D-2-0.35 and D-2-0.4 with a $2 \times 2 \times 2$ mm unit cell size and three strut thicknesses of 0.3, 0.35 and 0.40 mm, and corresponding design porosities of ~81, 76 and 70% (Fig. 3d,e,f). Note that the geometric attributes (cell size and strut thickness) of specimens D-1.5-0.25 and D-2-0.35 closely match those of specimens S0.25 and S0.34, respectively (Fig. 3). The next designed lattice structure was gyroid (G) with a unit cell size of $2.35 \times 2.35 \times 2.35$ mm, a wall thickness of 0.19 mm, and a design porosity of ~76% (Fig. 3g). The final designed structure was the Voronoi (V) lattice with a target pore radius of 0.65 mm, a strut thickness of 0.28 mm, an organic factor of 3.2 (which forms a chamfer at strut intersection nodes), and a design porosity of ~76% (Fig. 3h). Table 3 contains main geometric characteristics of all the lattice structures studied.

After fabrication, the build plate was depowdered, and each specimen was cleaned with compressed gas (Fig. 3i). The specimens were then separated from the build plate by electrical discharge machining (EDM), sealed in quartz tubes under vacuum, and subjected to post-fusion annealing (PFA) at 700 °C for 30 min, followed by water quenching. For comparison, two series of rhombic dodecahedron lattice structures (TAV-D-2-0.25 and TAV-D-2-0.35) with a $2 \times 2 \times 2$ mm unit

Table 3

Selected lattice structure geometric parameters.

Specimens	Nomenclature	Cell size, mm	Strut size, mm	Design porosity, %
D-type	D-1.5-0.25	$1.5 \times 1.5 \times 1.5$	0.25	76.0
	D-2-0.30	$2 \times 2 \times 2$	0.30	80.7
	D-2-0.35	$2 \times 2 \times 2$	0.35	75.8
	D-2-0.40	$2 \times 2 \times 2$	0.40	70.0
G-type	G	$2.35 \times 2.35 \times 2.35$	0.19	76.4
V-type	V	0.65 (target pore radius)	0.28	75.8

cell size and 0.25 and 0.35 mm strut thicknesses and ~85% and ~75% design porosities were manufactured from Ti-6Al-4V powder (AP&C, GE Additive, Canada). The following printing parameters were used on the same printing system: $P = 75$ W, $v = 1830$ mm/s, $h = 60$ μ m and $t = 20$ μ m. The powder PSD (ASTM B822) was $d_{10} = 23$, $d_{50} = 35$, and $d_{90} = 47$ (μ m); the flowability (ASTM B213), 25 s/50 g, and the bulk density, 2.55 g/cm³. After printing, the build plate was depowdered and specimens were cleaned with compressed gas. Before being cut from the build plate, the specimens were vacuum-annealed at 1010 °C for 45 min, cooled in vacuum to 200 °C, and finally air-cooled to room temperature.

2.4. Specimen characterization

Specimens obtained by LPBF were embedded in a two-component casting resin using cylindrical molds. The specimen surfaces were prepared for structural studies by multi-stage grinding and polishing. Mechanical grinding was performed first using abrasive paper with P320–P4000 grit sizes. Polishing followed on a piece of cloth for 30 min under a 25 N load to achieve a mirror-like surface. A SiO₂-based suspension with a 0.05 μ m particle size was applied. Solutions of liquid soap, H₂O₂ (3%), and ammonia were added to the suspension during polishing. To remove residual abrasive particles and suspension, the

specimens were finally placed in an ultrasonic bath with isopropyl alcohol for 5 min.

For the defect porosity analysis, surface images were captured using a Versamet-2 optical microscope (Union, Tokyo, Japan) equipped with a Nikon D90 camera (Nikon Corp., Tokyo, Japan). The original images (Fig. 4a) were binarized and then separated into the background (gray) and the specimen categories (white with black pores) (Fig. 4b). The analyses were performed using ImageJ software (Wayne Rasband, NIH, USA). The defect porosity was calculated using equation (4):

$$P_d = \frac{S_{\text{pore}}}{S_{\text{total}}} \quad (4)$$

where P_d is the defect porosity, S_{pore} is the pore area, and S_{total} is the total analyzed surface area. A minimum of 15 images were analyzed and statistically treated. To reveal the melt pool boundaries, specimens were surface-etched in a 1HF:3HNO₃:6H₂O solution for 3–30 s. The melt-pool structures were analyzed using a Versamet-2 optical microscope (Union, Tokyo, Japan) equipped with a Nikon D90 camera (Nikon Corp., Tokyo, Japan). The melt pool depth measurement schematic is shown in Fig. 4c.

The structural (functional) and defect porosities of printed parts were studied using a μ -CT NanoVoxel-1000 precision computed microtomography system (Sanying, China). Operating parameters included an X-ray tube voltage of 130 kV, a current of 40 μ A, and an image resolution of 10 μ m. The thickness of structural elements and also the functional (P_s) and defect ($P_{d(\mu\text{-CT})}$) porosities were obtained using VGStudio MAX software (Hexagon AB, Sweden). A sphere-fitting algorithm was applied to the reconstructed tomographic images to obtain distributions of the structural elements' dimensions and defect pores' equivalent diameters. To verify that the fabricated samples matched the original digital models, the same analyses were carried out on the reference CAD models, and deviations between the as-printed geometries and their CAD originals were analyzed. Throughout this manuscript, the term 'design porosity' (P_{CAD}) refers to the nominal open pore volume fraction of the CAD model; the term 'structural (functional) porosity' (P_s) refers to the effective open pore volume fraction; and the term 'defect porosity' (P_d) refers to unintended internal voids resulting from the LPBF process.

2.5. Static compression testing of lattice structures

The static compressive behavior of additively manufactured lattice specimens was characterized by uniaxial compression testing using an

LFV-L-25 universal testing system (Walter + Bai, Switzerland). The specimen geometries, design parameters, porosity levels, and fabrication conditions are described in Section 2.3 and illustrated schematically in Fig. 3. All tests were performed at room temperature under quasi-static conditions in accordance with ISO 13314 under displacement control with a constant crosshead speed of 5 mm/min. For Ti-6Al-4V lattice specimens, engineering stress–strain curves were used to extract the apparent Young's modulus (E), the conventional 0.2% offset yield strength ($\sigma_{0.2}$), and the maximum compressive stress (σ_{max}) as shown in Fig. 5a. For Ti-Zr-Nb lattice specimens, which exhibit a characteristic double-yielding behavior associated with stress-induced martensitic transformation, two distinct yield points were identified, specifically, the transformation yield strength (σ_{tr}) and the dislocation yield strength (σ_{dis}), and the maximum compressive stress (σ_{max}) as shown in Fig. 5b. For each lattice configuration, three specimens were tested under quasi-static compression conditions, and the reported values are presented as means $\pm$ standard deviations.

2.6. Fatigue compression testing of lattice structures

The fatigue specimens were identical in geometry, porosity level, and manufacturing conditions to those employed for static compression tests. Cyclic compression fatigue tests were performed under load control at a constant frequency of 15 Hz ($R = 0.01$). The applied cyclic load ranged from a minimum force of 0.025 kN to a predefined maximum load corresponding to 20, 30, 40, 50 and 80% of the characteristic yield strength of the material. For Ti-6Al-4V lattice specimens, the characteristic yield strength corresponded to the conventional 0.2% offset yield strength ($\sigma_{0.2}$), whereas for Ti-Zr-Nb lattice specimens, it corresponded to the dislocation yield strength (σ_{dis}). Each specimen was subjected to cyclic compression up to 5×10^6 cycles, which was defined as the run-out criterion. A large structural crack in the lattice specimens was considered a failure. The applied load was adjusted in increments of 0.05–0.1 times the yield strength until sufficient data (7–10 specimens) were obtained over the range of 10^5 to 5×10^6 cycles. The results of the fatigue tests were used to construct the Wöhler (S–N) curves, providing a comparative assessment of the fatigue performance of the LPBF-fabricated Ti-Zr-Nb and Ti-6Al-4V lattice specimens under cyclic compressive loading.

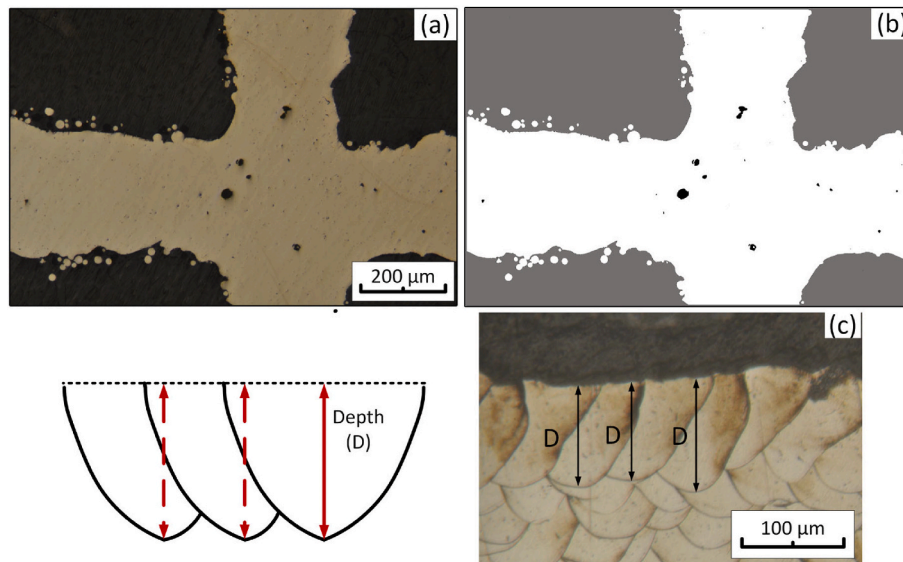


Fig. 4. (a) Metallographic and (b) binarized images of a S0.34 Ti-Zr-Nb lattice structure; (c) melt pool depth measurement schematic in bulk specimens; (#2 LPBF regime in all cases).

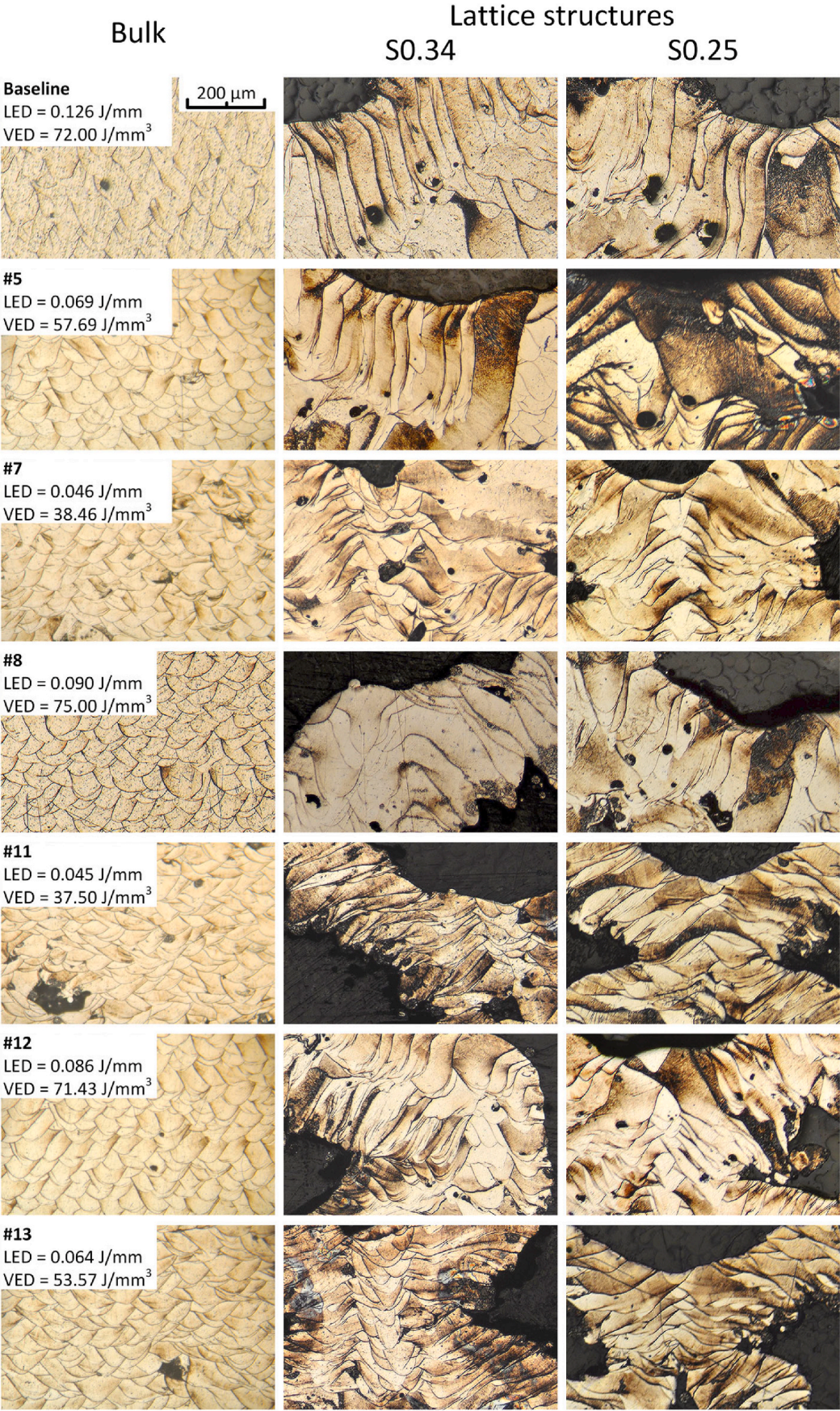


Fig. 6. Microstructure images of the bulk and lattice specimens produced using different LPBF regimes.

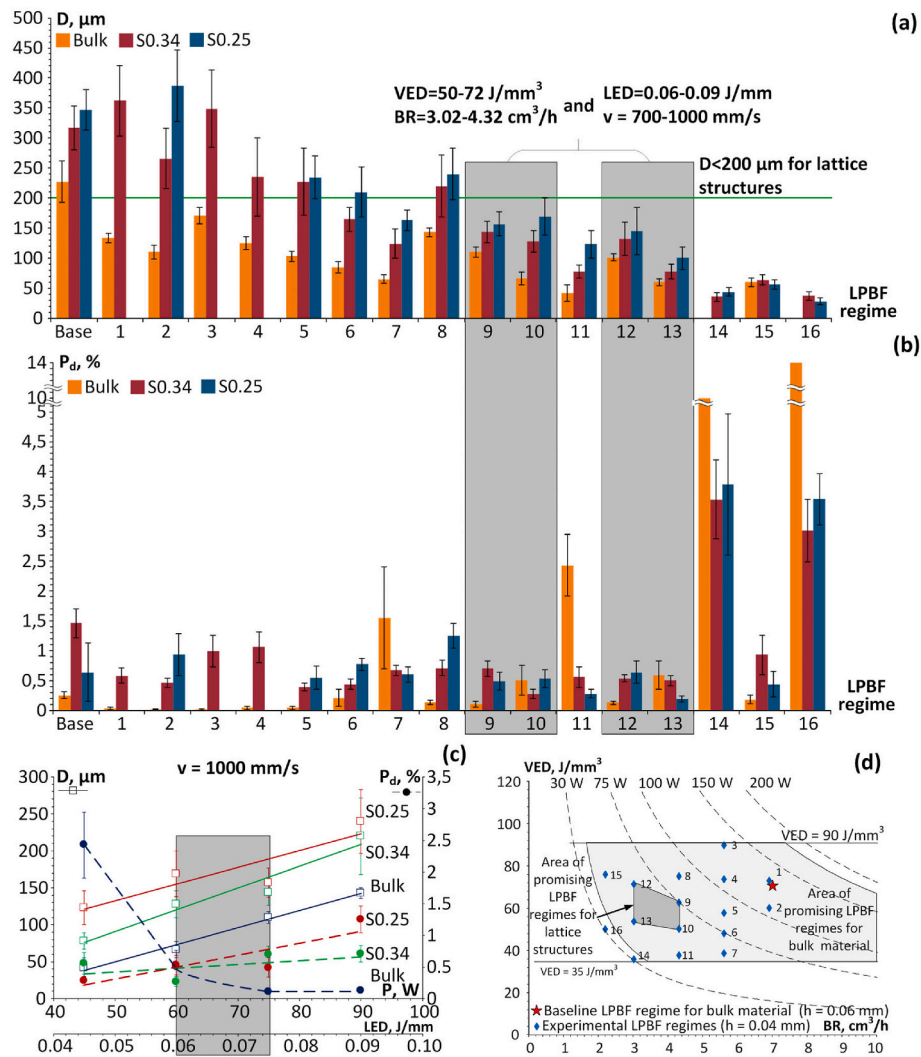


Fig. 7. (a) Melt pool depth (D) and (b) defect porosity (P_d) in the bulk and lattice specimens obtained using different LPBF modes; (c) variation of the D and (P_d) values as functions of the laser power/LED for a constant scanning speed of 1000 mm/s; (d) promising LPBF regimes. Melt pool depth values are unavailable for bulk specimens #14 and #16 because the excessively high defect porosity prevented their measurements.

remelting and adhesion of partially fused powder as compared to mode #12. The maximum falls within 0.24–0.25 mm, which is consistent with the nominal strut thickness of 0.25 mm. A similar trend is observed for specimen S0.34 (Fig. 8g and h). However, mode #12 yields larger deviations from the target strut size: the main peak is located at 0.27–0.28 mm, i.e., $\sim 0.07 \text{ mm}$ below the nominal value (Fig. 8g). The peaks associated with node sizes are less pronounced than in S0.25 and are centered at 0.32–0.33 mm and 0.43–0.44 mm.

The pore size distribution in specimens S0.25 manufactured using mode #12 (Fig. 8i) is bimodal. The first low-intensity peak corresponds to the inter-pore channels and is located at 0.36–0.39 mm, i.e., $\sim 0.50 \text{ mm}$ less than the calculated diameter. The second peak, associated with the primary pore population, is sharp and lies at 0.75–0.78 mm, corresponding to a deviation of $\sim 0.18 \text{ mm}$ from the calculated value. These discrepancies are primarily caused by material remelting and powder adhesion and are expected for lattice specimens with a rhombic dodecahedron (D) architecture.

Based on the above observations, mode #12 was identified as optimal among those considered, as it produces lattice structures with the fewest number of internal and external defects. To enable further application of mode #12, it is important to get rid of significant geometric deviations with respect to the nominal values obtained with this mode (0.04–0.07 mm). To reach this objective, the initial beam offset of

0.1 mm was reduced to 0.06 mm.

3.2. Microstructure and crystallographic texture of Ti–Zr–Nb lattice structures

The vertical cross-sections (XY) of bulk and lattice (S0.25, S0.34) specimens were analyzed using EBSD (see areas indicated by a red outline in the SEM images shown in the central column of Fig. 9). The EBSD analyses revealed differences in the as-built microstructure and crystallographic texture in the bulk and porous materials (Fig. 9a–c, e). In the bulk material (Fig. 9a), numerous low-angle boundaries (subgrain boundaries) with a misorientation angle of less than 15° are observed within the highly elongated grains formed in the build direction. These lattice defects are caused by elevated microstresses generated during the rapid solidification accompanying the LPBF process.

In lattice structures, the grain morphology differs significantly (Fig. 9c–e). The lattice design affects the heat dissipation path, promoting local directional crystallization. This leads to grain formation that partially follows the material's architecture and cooling pathway. Grains elongated in different directions along the struts of the rhombic dodecahedron structure are observed. This effect is more pronounced in specimens S0.25 (Fig. 9e), which have a smaller cell size and more compact strut junctions. Equiaxed grains are locally observed in the

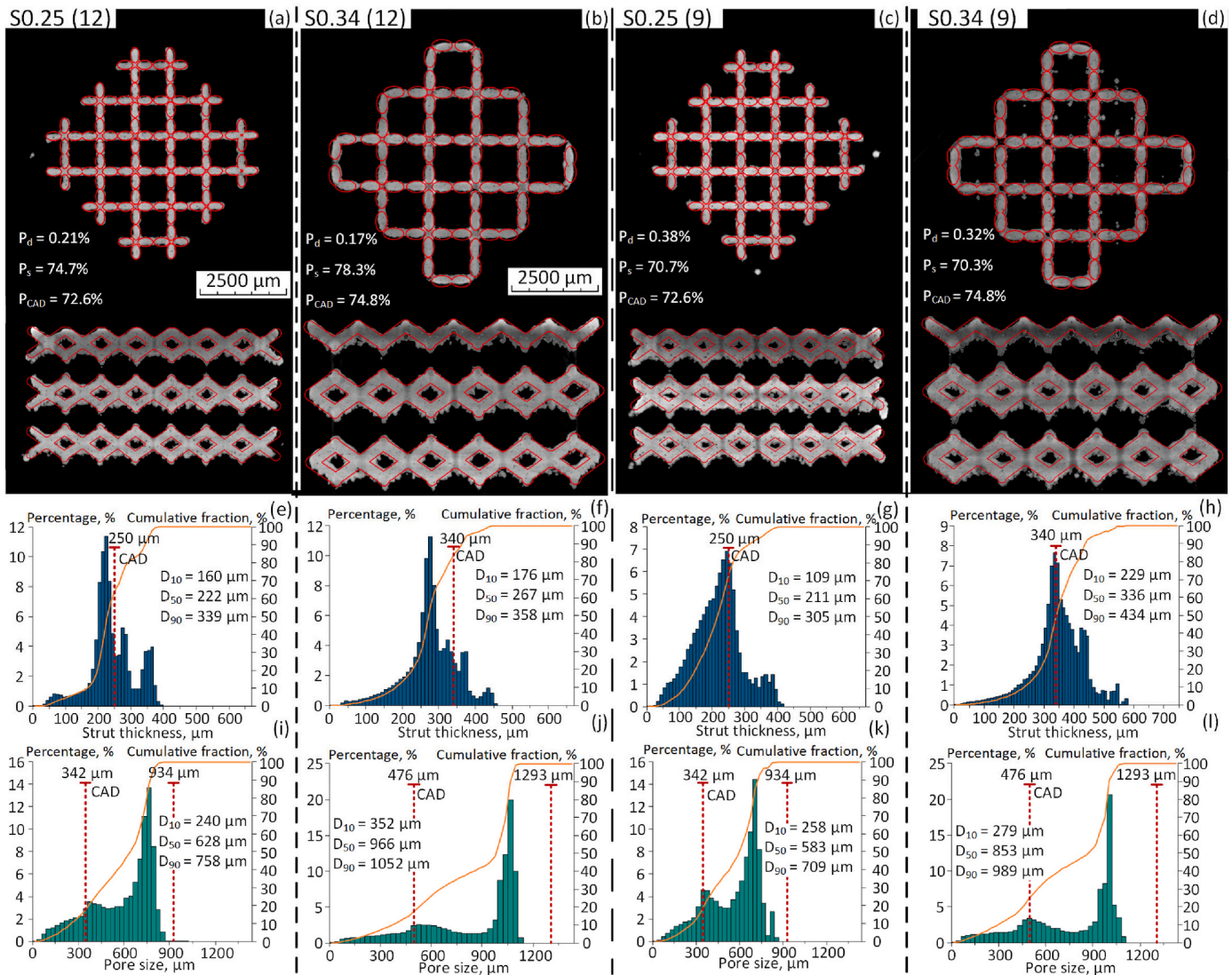


Fig. 8. – Micro-CT results of specimens S0.25 (a, c, e, g, i, k) and S0.34 (b, d, f, h, j, l) manufactured using LPBF modes #9 ($\text{VED} = 62.5 \text{ J/mm}^3$) (a, b, e, f, i, j) and #12 ($\text{VED} = 71.4 \text{ J/mm}^3$) (c, d, g, k, h, l): (a–d) horizontal and frontal projections of the reconstructed specimens compared with their CAD originals, (e–h) strut thickness distributions, (i–l) pore size distributions.

porous material structure, increasing in number as the strut thickness decreases. The number of low-angle boundaries in the porous material is significantly reduced as compared to the bulk material. Misorientations of less than 5° are rarely observed. The lower number of crystal structure defects, particularly subgrain boundaries, is attributed to the reduced thermal gradients and modified cooling conditions in the thin structural elements of lattice structures as compared to their bulk counterparts.

The crystallographic texture orientation shows a very strong maximum in the $\langle 100 \rangle_\beta$ direction (Fig. 9a), corresponding to the typical texture previously described for bulk Ti–Zr–Nb alloy [31]. In the structural elements of porous materials, this maximum is significantly weakened, and the crystallographic texture becomes random (Fig. 9c–e). This is explained by a wider distributed heat dissipation pathway determined by the lattice structure's geometry.

Heat treatment at 700°C does not significantly change the microstructure and crystallographic texture of the alloy in either bulk or porous states (Fig. 9b–d, f). The key grain morphology features formed during LPBF are preserved. In the bulk material, annealing reduces the number of low-angle boundaries (Fig. 9b). In the porous material, this effect is not observed due to a lower number of the as-built crystal structure defects (Fig. 9d–f).

3.3. Cell size, wall/strut thickness, structural and defect porosity in Ti–Zr–Nb lattice structures

The as-built geometry deviations of the D-type specimens manufactured using the baseline (D-2-0.35) and optimized #12 (D-1.5-0.25, D-2-0.35 and D-2-0.40) LPBF modes relative to their CAD geometries were assessed by micro-CT (Fig. 10a–d). Specimens produced using the baseline process mode exhibit pronounced layer remelting and powder adhesion on down-facing surfaces—typical defects associated with an excessive energy input (Fig. 10a). In the D-type structures, these defects are particularly critical because they partially block structural pores, thereby degrading the mechanical performance and compromising the suitability for osseointegration of the lattice structures.

Quantitative measurements corroborate these observations, revealing substantial deviations from the theoretical geometry (Fig. 10e–i). For the D-2-0.35 specimens manufactured using the baseline LPBF mode, the strut size distribution is trimodal: a strut peak at $0.45\text{--}0.46 \text{ mm}$ ($\approx 0.10 \text{ mm}$ above the nominal value) and two node-related peaks at $0.58\text{--}0.59 \text{ mm}$ (four-strut junctions) and $0.70\text{--}0.71 \text{ mm}$ (eight-strut junctions), indicating extensive remelting and powder attachment. The pore-size distribution (Fig. 10f) is also trimodal, with

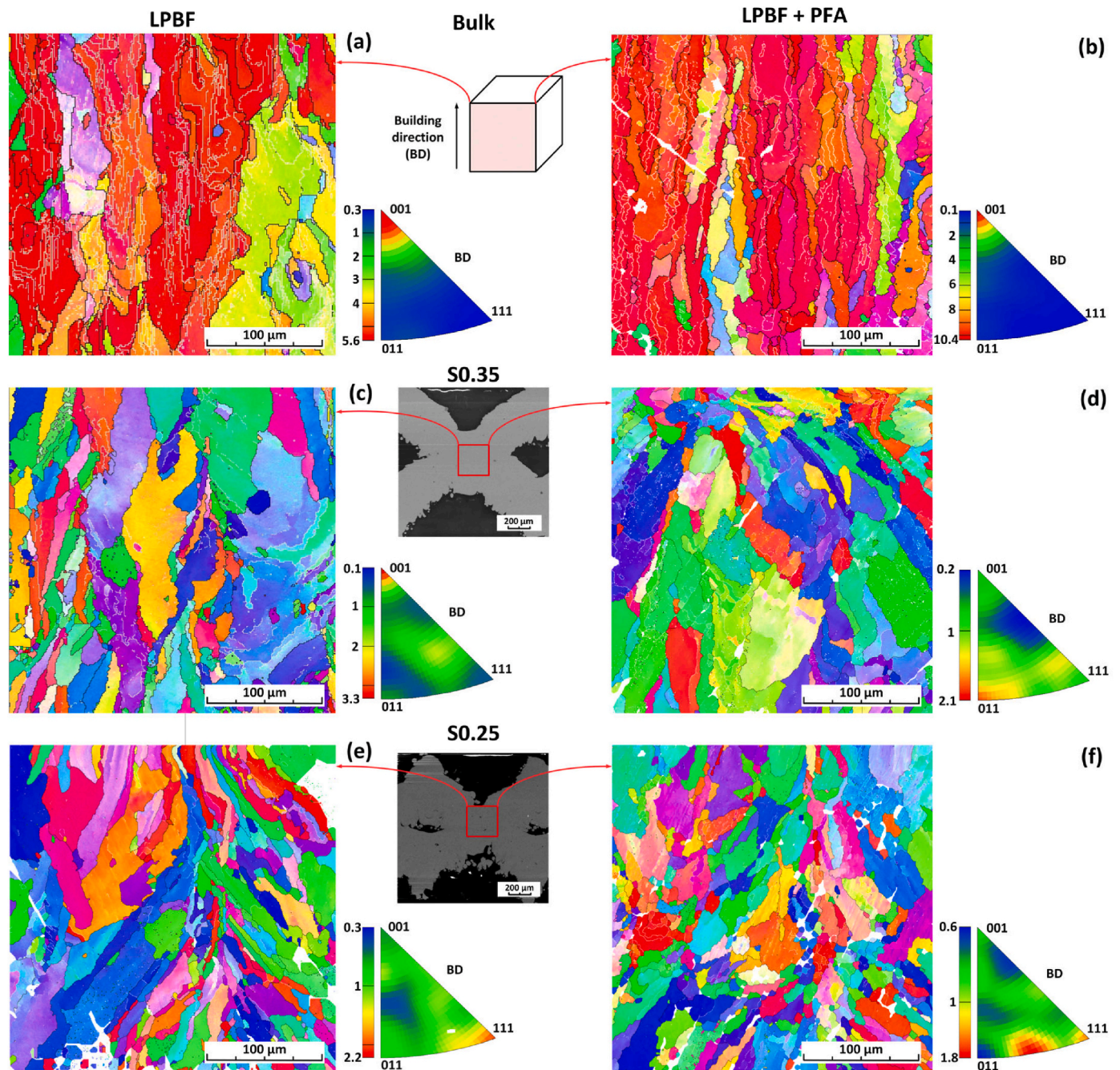


Fig. 9. – Microstructure and inverse pole figures of the (a, b) bulk and lattice (c, d) S0.34, (e, f) S0.25 specimens after (a, c, e) LPBF and (b, d, f) PFA.

peaks at 0.27–0.30 mm, 0.66–0.69 mm, and 1.05–1.08 mm. This reflects a departure from a classical rhombic dodecahedron towards the morphology with struts effectively elongated along the build direction.

Conversely to the LPBF-baseline mode (optimized for bulk material), the LPBF mode #12 optimized for both bulk and lattice materials yields strut-size distributions that closely match those of the CAD model (Fig. 10c). The deviation from the specified strut size is only 0.01–0.02 mm, i.e., within the measurement uncertainty. Two additional peaks are observed due to the localized defects with two node types (four- and eight-strut intersections). For D-2-0.35 (Fig. 10g), these peaks occur at 0.44–0.45 mm and 0.52–0.53 mm strut thickness range. The corresponding channel size peak for D-2-0.35 (Fig. 10k) lies in the 0.48–0.51 mm strut thickness range. The largest pores show comparatively small deviations from the nominal 1.283 mm value. Overall, specimens D-2-0.35 and D-2-0.40 reproduce the CAD contours well, with only limited

remelting and particle adhesion phenomena (Fig. 9c and d). More pronounced defects are observed in the D-1.5-0.25 specimens due to the smaller size of their unit cells (Fig. 10b).

D-1.5-0.25 (Fig. 10f) differs from the other D-type specimens by exhibiting a bimodal strut size distribution with an additional peak at 0.34–0.35 mm. This behavior is likely associated with defect formation in architectures with denser node arrangements arising from the reduced unit cell size. For the D-type specimens, pore-size histograms show one dominant peak and a tail of smaller pores generated by remelting during the LPBF process (Fig. 10f–h). Within this tail, a weak peak corresponds to the channel size. For D-2-0.35 and D-2-0.40, the channel sizes agree with the CAD specification, whereas the main pore sizes are reduced by 0.22–0.25 mm relative to the nominal value (Fig. 10f–f). Despite the larger overall geometric deviation in the D-1.5-0.25 specimen (Fig. 10j), the channel size (0.36–0.39 mm) agrees with

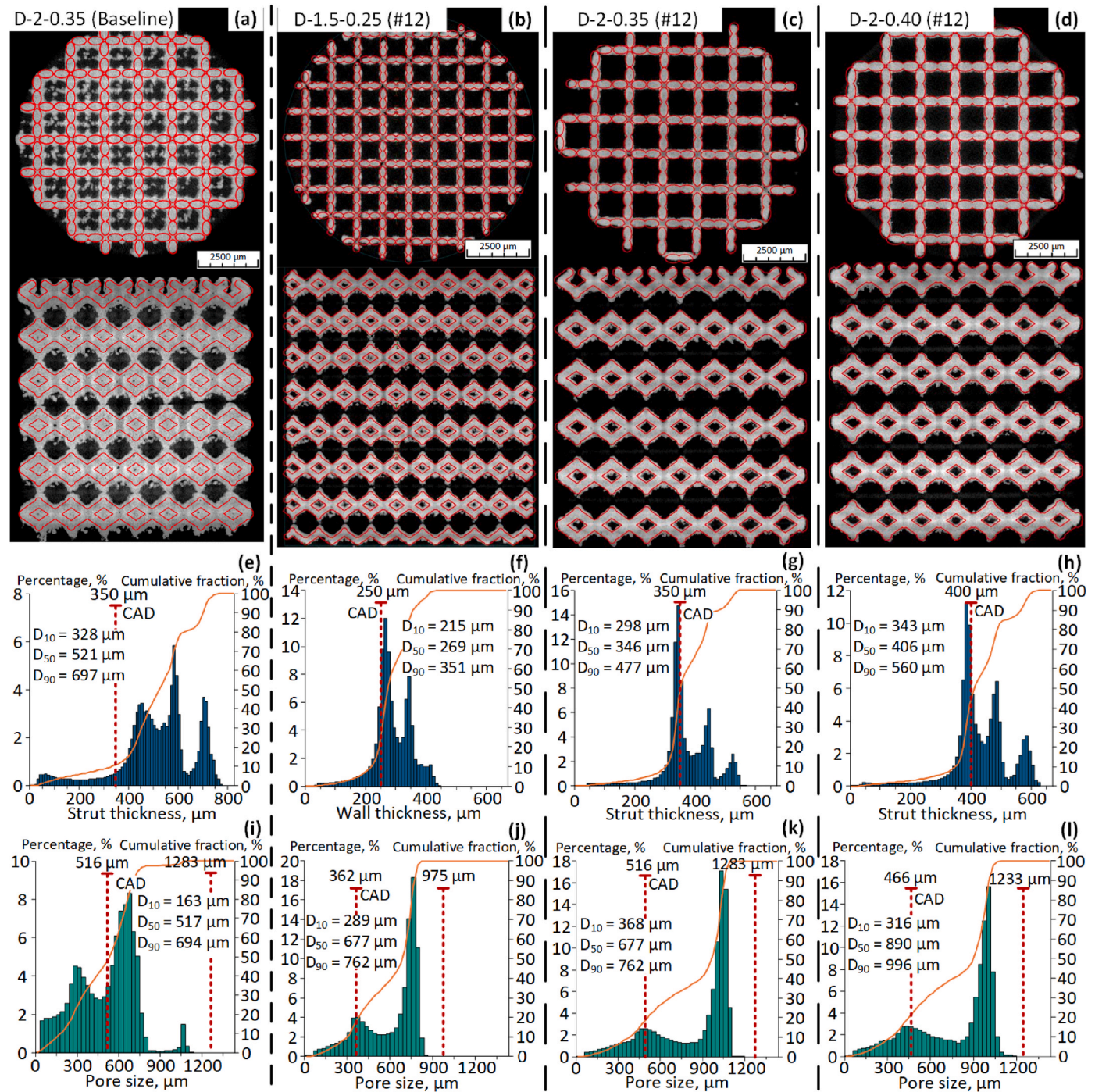


Fig. 10. – Micro-CT results of specimens (a, e, i, c, g, k) D-2-0.35, (b, f, j) D-1.5-0.25, and (d, h, l) D-2-0.40 manufactured using the (a, e, i) baseline mode and (b-d, f, h, j-l) #12 mode: (a–d) horizontal and frontal projections of reconstructed specimens as compared to the original CAD model, (e–h) strut thickness distributions, (i–l) pore size distributions (red dotted lines correspond to the CAD-related data).

the specified 0.36 mm value, while the main pore size (0.75–0.78 mm) is smaller by ~0.23 mm than the nominal 0.98 mm value.

Fig. 11 compares the macrostructures of V- and G-type specimens fabricated using the baseline and #12 mode. The geometric deviations produced with mode #12 are markedly smaller than those obtained using the bulk material baseline mode, as evidenced by comparing the projections of the reconstructed #12 specimens (Fig. 11b–d) with their baseline equivalents (Fig. 11a–c). Remelting zones and adhered powder particles are still visible on the down-facing surfaces of these structures. However, the number of these defects is substantially reduced.

For the V structures produced using the baseline LPBF mode

(Fig. 11e), the strut size distribution is bimodal, with peaks at 0.39–0.41 mm (distorted struts) and 0.53–0.55 mm (nodes and locally thickened elements), with the largest increases occurring at the nodes. The pore-size distribution (Fig. 11i) shows a dominant peak at 0.84–0.87 mm and a tail of small pores associated with remelting. The G structures produced using the LPBF baseline mode (Fig. 11g) exhibit a unimodal strut size distribution with a peak at 0.36–0.37 mm (vs. 0.19 mm nominal) and a unimodal pore size distribution with a peak at 0.60–0.61 mm (Fig. 11k).

It can be asserted that, when manufactured using the optimized mode #12, the V- and G-type geometries correspond more closely to

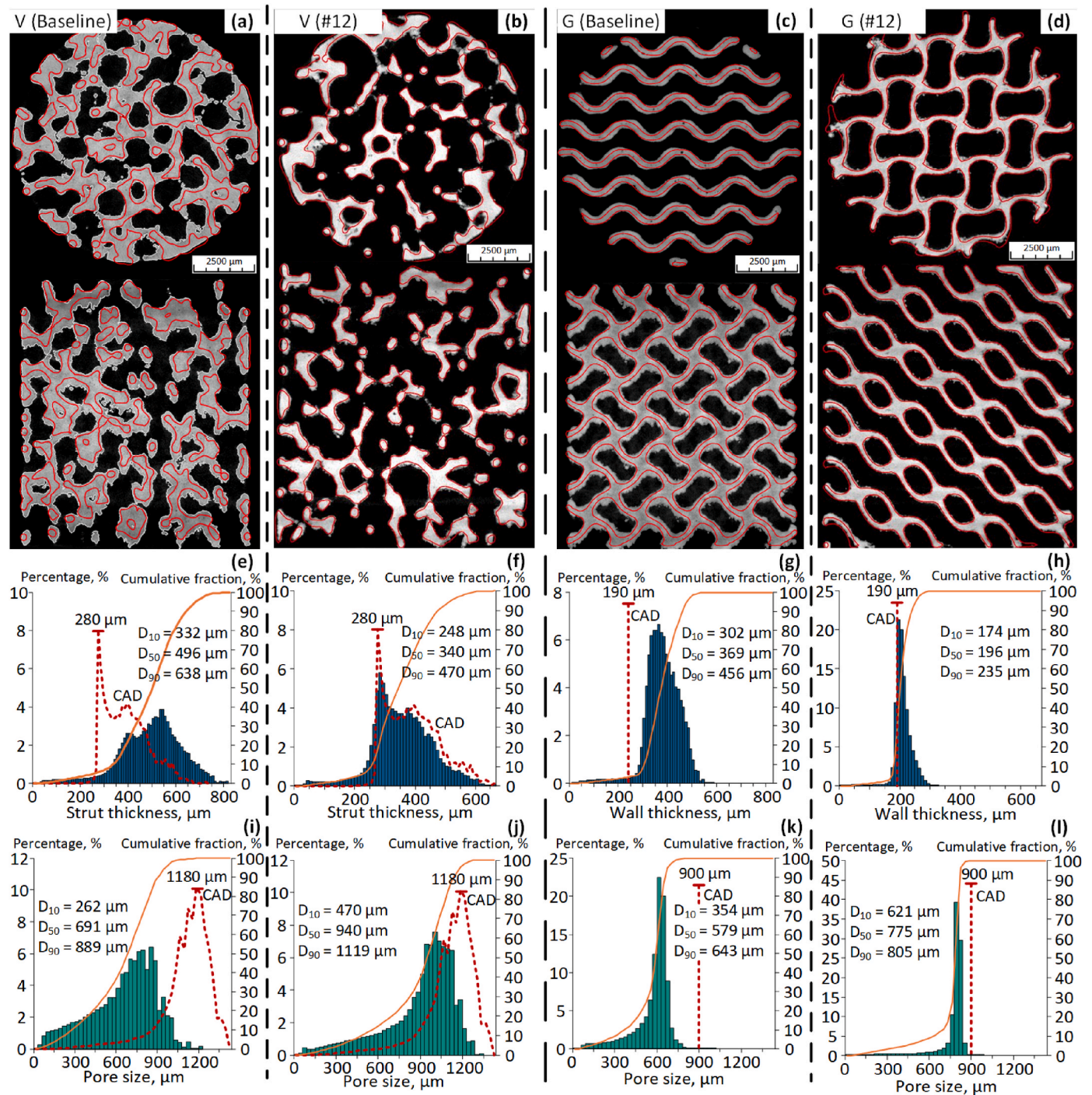


Fig. 11. – Micro-CT results of specimens (a, b, e, f, i, j) V and (c, d, g, h, k, l) G manufactured using (a, e, i, c, g, k) baseline and (b, f, j, d, h, i) #12 modes: (a–d) horizontal and frontal projections of reconstructed specimens compared with the original CAD models, (e–h) strut thickness distributions, (i–l) pore size distributions (red dotted lines correspond to the CAD-related data).

their CAD models than the D-type structures. For the V structures (Fig. 11f), the strut size distribution closely matches the CAD contour, with a peak at 0.28–0.29 mm, consistent with the target strut size of 280 μm . The pore-size distribution (Fig. 11j) shows a larger shift, with a peak at 0.99–1.02 mm relative to the CAD-based value. For the G structures (Fig. 11h), a single pronounced strut size peak at 0.19–0.20 mm matches the CAD specification (0.19 mm). The pore size distribution (Fig. 11l) also shows a distinct peak at 0.78–0.81 mm, representing the smallest deviation from the nominal pore size (0.90 mm) among all the investigated architectures.

The micro-CT data were further used to quantify the defect porosity (P_d) and the structural porosity (P_s) in the D-, V and G-type structures (Fig. 12). Under the baseline LPBF mode, the D-type lattices exhibit the largest reduction in structural porosity ($P_s - P_s(\text{CAD}) \approx 31\%$) and the highest defect porosity ($P_d \approx 1.49\%$). In contrast, the same lattices produced using the optimized mode #12 show a markedly improved quality, with a much smaller P_s discrepancy ($\approx 5\%$) and a low defect porosity ($P_d \approx 0.16\%$).

The G- and V-type lattices show smaller reductions in P_s in the baseline mode ($\approx 18\text{--}21\%$), indicating a better manufacturability of

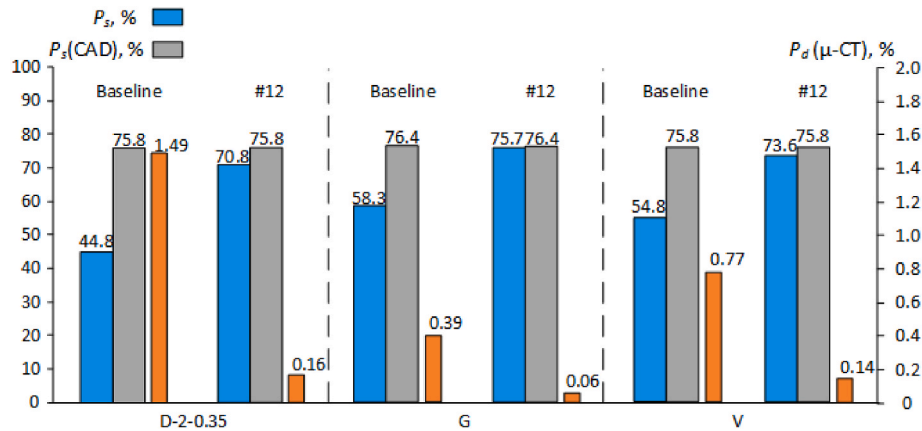


Fig. 12. Design, structural and defect porosity of specimens obtained using two printing parameter sets: baseline mode (reference) and #12 mode (optimized) for D-2-0.35, -V and -G.

these geometries under the reference regime; their defect porosity is also lower ($P_d \approx 0.39\%$ for G and $\approx 0.77\%$ for V). Despite these satisfactory results with the baseline mode, applying mode #12 substantially improves the geometric fidelity of these structures, reducing deviations from the target architecture to $<2\%$ for the V and G specimens. The G-type lattices exhibit the smallest deviations from the design porosity, consistent with their smooth, continuously varying topology without sharp angles or discrete nodes, in contrast to the rhombic dodecahedron (D) and Voronoi-type architectures.

For the D-type specimens produced using mode #12, the largest remaining deviations is observed for D-1.5-0.25 ($\approx 9\%$). D-type lattices with a larger unit cell size (D-2) show smaller deviations ($\approx 4\text{--}6\%$), likely due to a reduced number of nodes where pronounced remelting typically occurs. Importantly, the defect porosity of all the specimens manufactured with mode #12 does not exceed 0.2%, which is an excellent level for LPBF-fabricated porous components.

Overall, the micro-CT analysis demonstrates that the selected optimal LPBF mode #12 produces macrostructures that are sufficiently close to the design intent for multiple pore configurations. The largest deviations are observed for D-type structures, whereas the smallest deviations occur for G-type structures, consistent with prior reports on the LPBF process optimization targeted the printing of porous architectures [47].

3.4. Mechanical properties and fatigue behavior of Ti–Zr–Nb lattice structures

Fig. 13 summarizes the engineering stress–strain responses of the LPBF-fabricated lattice structures under quasi-static uniaxial compression. For comparison, the compressive response of a fully dense (bulk) TZN specimen is also included.

The bulk TZN alloy exhibits a well-defined double-yielding behavior (highlighted in green in Fig. 13a) corresponding to the stress-induced martensitic transformation (SIMT). The associated transformation yield strength is approximately 520 MPa. A magnified view of the low-stress region (Fig. 13b) highlights the characteristic mechanical response of architected lattice structures. The enlarged low-stress region (Fig. 13b) reveals that the architected porous TZN materials retain a distinct transformation-related yield strength as well, demonstrating that SIMT continues to be in effect even in the presence of substantial structural porosity, node/strut junctions, and local stress concentrators inherent to the lattice geometries. A similar double-yielding response under compression has previously been reported for porous superelastic NiTi alloys, confirming that SIMT can persist at the macroscale within architecturally porous superelastic alloys [33].

Among the investigated TZN architectures, the highest compressive

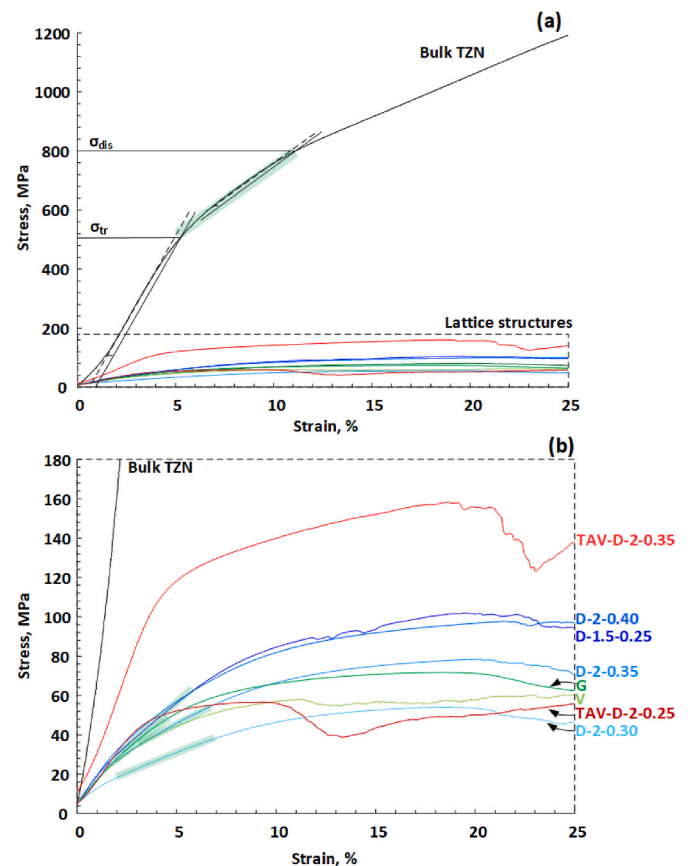


Fig. 13. (a) Engineering compressive stress–strain curves of the LPBF-fabricated lattice specimens; the bulk TZN specimen is included for reference. (b) Enlarged view of the initial deformation region; transformation-induced yield strength (green).

strength was obtained for lattices with the smallest unit cell sizes (D-1.5-0.25) and for those with the largest strut thicknesses (D-2-0.4). Conversely, decreasing the strut thickness leads to a systematic reduction in yield strength, consistent with the bending-dominated deformation mechanism of open-cell lattices. In such architectures, strength scales with relative density according to the established cellular-bulk scaling relations, and a reduction in the strut thickness (at a fixed topology) decreases both the load-bearing cross-section and the effective moment of inertia, promoting an earlier onset of the bending-driven

yielding and progressive collapse phenomena.

For identical unit cell parameters and strut thicknesses, the TAV lattice specimens exhibit higher compressive strengths than do their TZN counterparts. However, the post-yield response of the TAV lattices is characterized by more pronounced stress oscillations, indicative of the successive macroscopic collapse events of load-bearing struts and a reduced deformation stability. In contrast, the TZN lattice specimens show a smoother stress–strain response during post-yield deformation. This difference is consistent with the transformation-assisted deformation in the TZN structures, which assists in redistributing stresses, accommodating local strain concentrations, and thereby stabilizing progressive collapse at the structural scale.

Fig. 14 summarizes the compressive properties of the studied structures as a function of their structural porosity and effective stiffness. For the dodecahedron-based lattices, the yield strength decreases approximately linearly with increasing porosity for both alloy systems (see also Fig. 14a). In the TZN dodecahedron series, increasing the porosity from ~66 to ~76% reduces the yield strength from ~75 to ~35 MPa, in agreement with the expected relative density–strength dependence for bending-dominated cellular architectures [14]. The Voronoi-type TZN lattice specimens follow the same trend, showing a ~75% structural porosity and a ~50 MPa yield strength. At comparable porosities, the TAV lattice specimens consistently exhibit higher yield strengths than their TZN equivalents, thus reflecting significantly different yield strengths of parent materials, i.e., $\sigma_{dis} \sim 800$ MPa for TZN (Fig. 13a) versus $\sigma_{0.2} \sim 1000$ –1200 MPa for TAV.

Notably, the TZN gyroid achieves the best strength–porosity performance among the TZN designs, reaching ~60 MPa at a ~76% structural porosity. This value is comparable to, or slightly higher than that of the TAV dodecahedron lattice structures with a similar structural porosity, suggesting a more efficient load transfer enabled by the continuous minimal-surface topology, reduced stress concentrations at nodes, and a larger fraction of membrane-like load paths as compared to strut-based lattices.

The Young's modulus decreases monotonically with increasing structural porosity (Fig. 14b). At the same porosity level, the Young's modulus of the TZN lattice specimens is approximately twice as low as that of their TAV equivalents (e.g., at ~67% porosity: ~1.5 GPa for TZN versus ~3 GPa for TAV). This reduced stiffness is potentially beneficial for mitigating stress shielding in load-bearing implants. In addition, the TZN gyroid and Voronoi architectures show slightly higher moduli than predicted by the linear trend of the TZN dodecahedron series at comparable porosities, indicating a topology-driven stiffness enhancement attributable to improved load transfer pathways and reduced contributions of the bending-dominated deformation.

The strength–stiffness map in Fig. 14c (including reference ranges for trabecular and cortical bone [12]) highlights that for the same porosity, the TZN lattice specimens provide substantially lower stiffnesses than their TAV counterparts, while maintaining comparable mechanical strengths. This combination might translate into a more favorable strength–stiffness balance for orthopedic applications, where a low modulus is desired to reduce stress shielding without compromising mechanical integrity.

Fig. 15 presents the compression–compression fatigue behavior in the form of Wöhler (S–N) curves. Specimens surviving 5×10^6 cycles were treated as run-outs.

In the absolute stress–life coordinates (Fig. 15a), the TZN gyroids (TZN-G) exhibit the highest fatigue resistance. The fatigue data is separated into two main groups governed primarily by structural porosity (~66–71% versus ~75–76%), largely independent of the alloy system. This clustering confirms that the relative density is the dominant controlling parameter for fatigue performance in these lattices, consistent with the bending-dominated fatigue mode, where local stress amplitudes are strongly amplified by the reduced strut cross-section and geometric imperfections.

Within the dodecahedron series, the smallest-cell lattice specimens

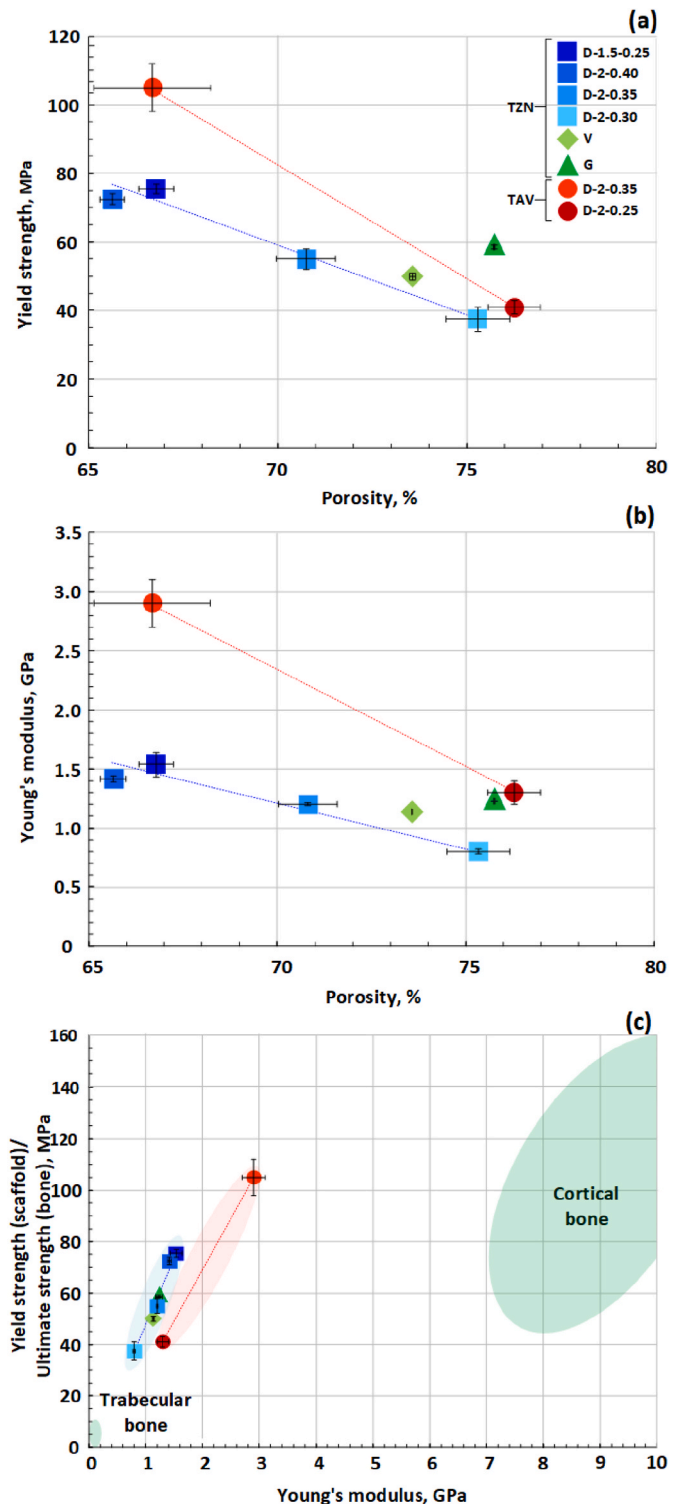


Fig. 14. Mechanical properties of LPBF-fabricated lattice structures: (a) yield strength (σ_{dis} for TZN and $\sigma_{0.2}$ for TAV) and (b) Young's modulus as a function of structural porosity; (c) yield strength as a function of Young's modulus, including reference ranges of ultimate strength and Young's modulus for trabecular and cortical bone [12,48,49].

(D-1.5-0.25) show superior fatigue life as compared to the larger-cell variants and even outperform the D-2-0.35 structures. For the TZN dodecahedron lattice specimens, reducing the strut thickness from 0.4 to 0.3 mm causes a marked decrease in fatigue strength (from ~25 to ~8 MPa), consistent with increased bending stresses, a greater sensitivity to

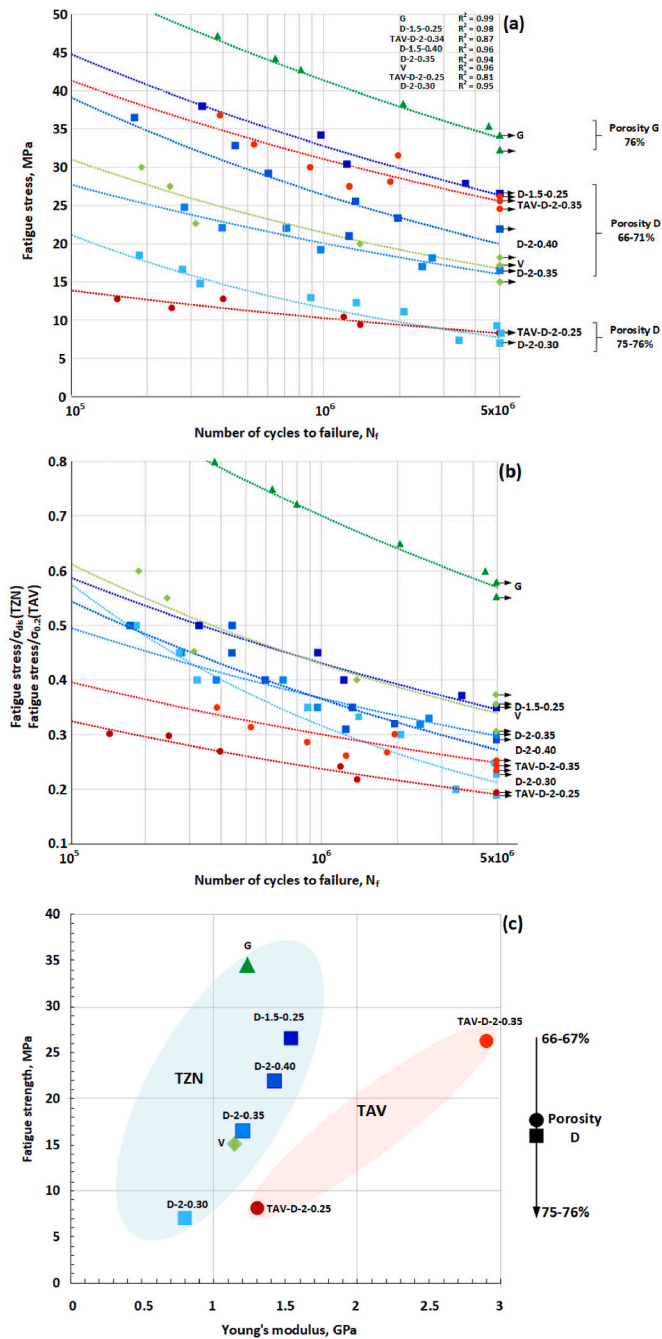


Fig. 15. Compression-compression fatigue behavior of the LPBF-fabricated lattice structures: (a) maximum fatigue stress versus cycles to failure (S-N curves); (b) normalized fatigue response (fatigue stress/ σ_{dis} for TZN and fatigue stress/ $\sigma_{0.2}$ for TAV) versus cycles to failure; (c) fatigue strength – Young's modulus diagram.

the surface roughness/notches typical of LPBF struts, and an earlier formation of critical damage in the thinnest load-bearing members. Interestingly, although D-2-0.3 structures exhibit the lowest high-cycle fatigue strength, they demonstrate an improved low-cycle fatigue life at higher stress amplitudes as compared with the TAV lattice specimens of similar porosities, suggesting an enhanced damage tolerance of the stress induced martensitic transformation (SMT)-capable TZN alloy under large cyclic strains.

Fig. 15b shows fatigue data normalized by the dislocation-related yield strength (σ_{dis} for TZN and $\sigma_{0.2}$ for TAV). The use for both structures (TZN and TAV) their respective dislocation yield strengths as

normalization parameters is justified by the fact that if these values are exceeded, both structures will accumulate irreversible strain under load and therefore, functionally fail. The gyroids remain distinctly separated, with a fatigue resistance-to-yield strength ratio of ~ 0.6 , whereas other lattice specimens fall within the ~ 0.2 – 0.35 range. For TZN, the normalization requires special consideration because the macroscopic yielding involves double yielding: the SMT-driven yielding followed by the dislocation-mediated plasticity. The TZN lattice specimens indicate a clearly improved normalized fatigue performance as compared to their TAV equivalents, supporting the hypothesis that transformation-assisted deformation promotes stress redistribution within the structure and delays crack initiation/propagation at the strut scale. Finally, correlating fatigue strength with Young's modulus (Fig. 15c) emphasizes that, at a comparable fatigue strength and structural porosity, the TZN lattice specimens deliver approximately a twofold lower Young's modulus than their TAV counterparts, which may represent an advantageous strength–stiffness balance for long-term load-bearing orthopedic use. A limitation of this comparison is that the TAV and TZN specimens were fabricated with different layer thicknesses (20 and 30 μm , respectively). Because a smaller layer thickness can improve surface quality, part of the difference in absolute fatigue life may be attributable to this processing effect, in addition to variations in material properties and lattice architecture.

Fig. 16 presents SEM micrographs of specimens collapsed after fatigue testing, used to analyze fracture features in lattice structures. At the macro scale, the rhombic dodecahedron specimens (TAV-D-2-0.35 and D-2-0.30) exhibit a similar fracture pattern (Fig. 16a and b). Cracks propagate along the specimen diagonal at an angle of about 45° . This fracture mode is typical of strut-based structures, particularly the rhombic dodecahedron [50–52]. Microcracks in these specimens (Fig. 16a and b, left) are localized at strut intersections (nodes), indicating that these locations act as stress concentrators, as shown previously [50,53]. Fatigue cracks initiate on strut surfaces, predominantly in sites where powder particles adhere and act as stress raisers [51,54]. The crack then propagates perpendicular to the struts, which is consistent with the classical failure mechanism of lattice structures under cyclic loading. The enlarged regions in Fig. 16a and b (right) show fracture surfaces whose morphology is generally consistent with that previously reported for similar materials [50,53,55]. At the same time, the presented fracture surfaces do not exhibit the characteristic dimpled relief typically associated with ductile overload fracture.

In the Voronoi-type specimen, the crack propagates diagonally through the structure along a tortuous path because of the random distribution of structural elements (Fig. 16h) [56]. In the sheet gyroid, a single main crack does not form; instead, numerous local cracks are observed throughout the structure volume (Fig. 16d). This may indicate a more uniform load distribution and the absence of stress concentrators like those present in the Voronoi and rhombic dodecahedron structures [52]. Microcracks show characteristic localization at sheet bending sites (Fig. 16d). The fracture surface characteristics are similar to those described above.

Overall, the results of the present work indicate that the mechanical response and fatigue life of LPBF lattices are governed primarily by their relative density and architecture, whereas the alloy selection mainly sets the “strength–stiffness” trade-off and the post-yield deformation stability. At a comparable porosity, TZN lattices exhibit markedly lower Young's moduli while maintaining similar levels of fatigue strength as their TAV counterparts, which indicates their potential for load-bearing orthopedic applications with a reduced risk of stress shielding. The most favorable property combination is obtained with the triply periodic minimal-surface (TPMS) architecture (gyroid), plausibly due to a more efficient load transfer pathway and reduced stress concentrations. A superior fatigue performance of the G-type architecture is likely due to three factors: (i) intrinsic topological advantages of the TPMS geometry, which provides continuous, node-free load paths with smooth curvature and reduced stress concentrations; and (ii) a higher geometric fidelity of

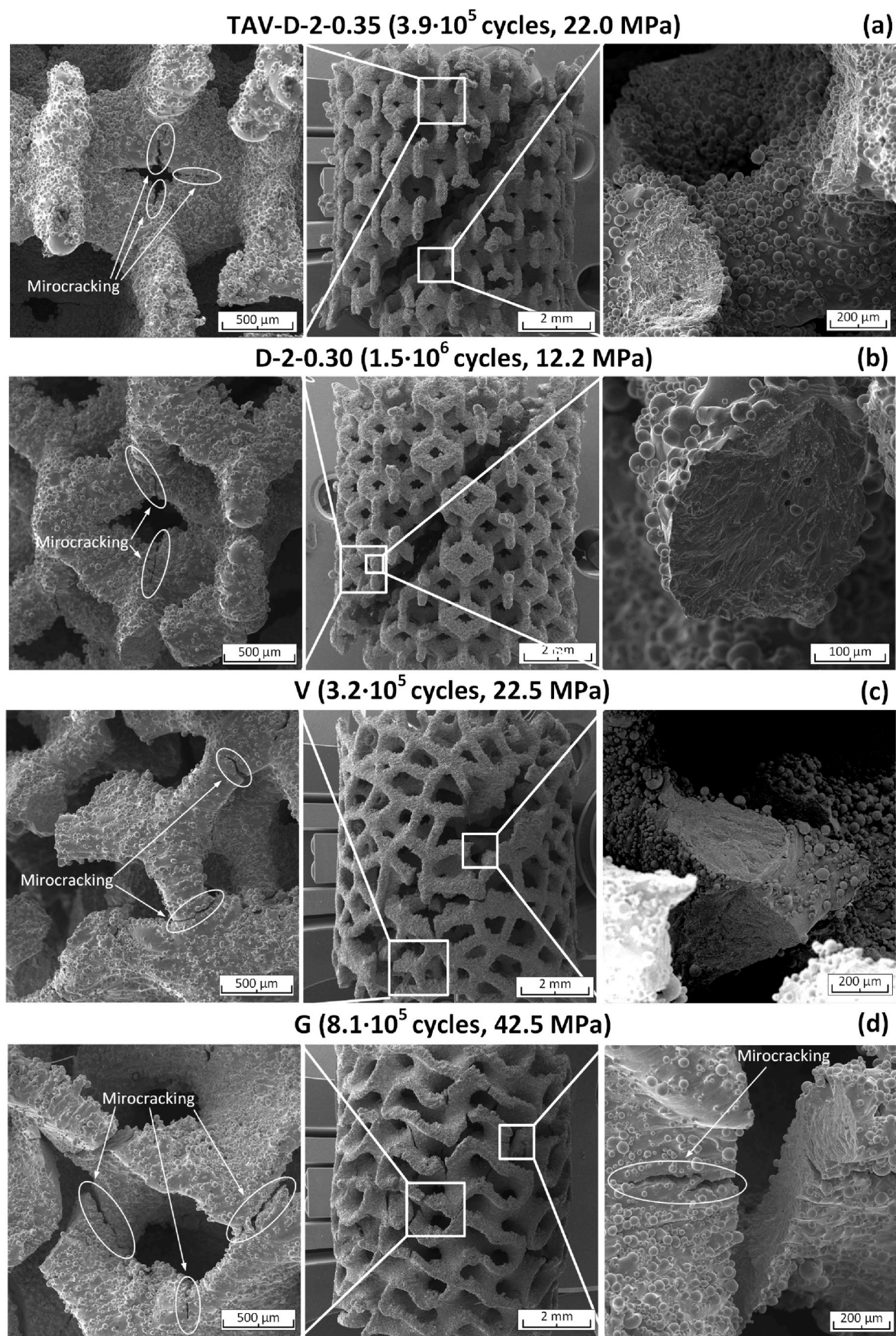


Fig. 16. SEM micrographs (general view in the center and enlarged fragments on the left and right) of collapsed specimens after fatigue tests: (a) TAV-D-2-0.35, (b) D-2-0.30, (c) V, and (d) G.

G-type structures under the applied LPBF processing conditions and (iii) the lowest defect porosity among all the investigated architectures [57]. Prior fatigue studies on Ti–6Al–4V TPMS LPBF lattices also reported a strong dependence of the fatigue resistance of these lattices on their architecture. They also consistently included gyroids among the best-performing TPMS topologies under cyclic loading [58,59]. In contrast, in dodecahedron (strut-based) lattices, strut thickness remains a critical parameter because it governs the bending stress levels and the sensitivity to LPBF-induced defects. This aligns with recent findings showing that AM-induced geometric deviations significantly alter local stress distributions and failure initiation sites [57]. An additional advantage of TZN is its stress/strain-induced martensitic transformation, which likely redistributes cyclic stresses and helps preventing catastrophic collapse under loading.

4. Conclusions

1. LPBF processing parameters were optimized for fabricating superelastic Ti–Zr–Nb (TZN) open cell lattice specimens with rhombic dodecahedron, gyroid, and Voronoi architectures. The results show that porous geometries exhibit deeper melt pools than bulk material under identical processing conditions, with the effect becoming more pronounced as the strut thickness decreases. This architecture-dependent thermal response strongly affects the defect formation and dimensional accuracy. A practical processing window ensuring limited melt-pool penetration and low defect porosity was identified at VED = 50–72 J/mm³, LED = 0.06–0.09 J/mm, P = 45–75 W and v = 700–1000 mm/s.
2. The mechanical response of LPBF TZN lattices is governed primarily by relative density and topology. Under monotonic compression, the yield strength of the TZN dodecahedron series decreased from ~75 to ~35 MPa as structural porosity increased from ~66 to ~76%, while the Voronoi lattice specimens exhibited ~50 MPa at ~75% porosity. The gyroid architecture provided the best strength–porosity combination, reaching ~60 MPa at ~76% porosity. At comparable structural porosity, TZN lattice specimens exhibited an apparent Young's modulus approximately two times lower than that of their Ti–6Al–4V (TAV) counterparts (e.g., ~1.5 vs ~3 GPa, at ~66% porosity).
3. Under compression–compression fatigue loading up to 5×10^6 cycles, the fatigue performance was controlled mainly by structural porosity, with the data clustering into two density-defined groups (~66–71% and ~75–76%) largely independent of the alloy type. The gyroid architecture exhibited the highest fatigue resistance and the highest fatigue-strength-to-yield-strength ratio (~0.6), whereas the other architectures remained within the ~0.2–0.35 ratio value. In the TZN dodecahedron series, decreasing the strut thickness from 0.4 to 0.3 mm reduced fatigue strength from ~25 MPa to ~8 MPa, highlighting the critical role of strut-scale bending stresses and LPBF-induced defects.
4. Overall, TZN lattices exhibit fatigue strength comparable to that of their TAV equivalents while offering substantially lower stiffness. Combined with TZN's transformation-assisted deformation capability, which likely promotes stress redistribution and stabilizes damage evolution under cyclic loading, these results suggest that LPBF-fabricated TZN lattice structures have promising mechanical properties for potential orthopedic applications, pending further validation under physiologically representative loading conditions.

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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