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Highlights

Tensile deformation and failure of AlSi10Mg random cellular metamaterials

L. Salvi, B. Smaniotto, F. Hild, M.G. Tarantino

- The mechanisms of ductile fracture were revisited using random porous metamaterials
- Structural disorder was found to delay long-wavelength localization
- Void-sheeting was observed with increasing pore aspect ratio
- McClintock's model underestimated void growth data within the failure band

Tensile deformation and failure of AlSi10Mg random cellular metamaterials

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Abstract

Cellular metamaterials are an emerging class of porous materials, in which the topology of the solid is *precisely-engineered* to achieve attractive combinations of properties. This work specifically examines the tensile response of a novel type of cellular metamaterials, in which pores of arbitrary elliptical shape are randomly-dispersed into an aluminum alloy matrix. Their porous mesostructure is generated numerically via a random sequential absorption algorithm, and is fabricated by laser powder bed fusion from AlSi10Mg powders. The results – obtained by means of digital image correlation combined with X-Ray tomography - highlight the advantages offered by a random cellular topology. These are notably a low sensitivity to geometric imperfections (which result inevitably from manufacturing), coupled with the ability to delay long-wavelength strain localization (which is in turn responsible for failure). Structural disorder also leads to highly heterogeneous deformation patterns, which result from the interaction between geometric pores and promote void growth and coalescence during plastic straining. The presence of manufacturing defects, moreover, exacerbates void interaction and in turn promotes early plastic flow localization. Interestingly, experiments reveal that this class of porous solids effectively display features of the ductile fracture of metals, albeit at the mesoscale. For example, *void-sheeting* is observed with increasing pore aspect ratio and is accompanied by large geometric distortions of the voids upon coalescence. Measured data for the pore strains are, moreover, highly scattered and show a departure from McClintock's model predictions for the voids within the fracture band. Collectively, this study highlights the potential offered by random porous metamaterials, which can be harnessed to revisit the

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complex mechanisms of ductile fracture at the mesoscale.

Keywords:

Ductile fracture, strain localization, void growth and coalescence, cellular metamaterials, digital image correlation (DIC), laser-powder bed fusion (LPBF).

1. Introduction

2 Metallic cellular materials are a class of lightweight materials consisting of a solid metal
3 matrix and air. They contain pores of arbitrary geometry, whose content can be varied to
4 achieve solid relative densities typically between 80% and 0.1%. They can be classified into
5 *open-* and *closed-cell* solids depending on the connectivity of the pores [1]. The former ones are
6 called *metal sponges* and contain pores that are interconnected; the latter ones are designated
7 as *metallic foams* and consist of non-percolating voids, which are separated from each other
8 by portions of solid matrix. In general, open-cell metal sponges are fabricated by investment
9 casting, by replication processing and by sintering; closed-cell metallic foams are instead
10 produced by foaming techniques [1]. In recent years, the rapid progress in manufacturing
11 technologies has enabled the fabrication of metallic cellular structures with intricate, yet
12 *precisely-controlled*, pore features. Architected materials made of metals, commonly referred
13 to as metallic *metamaterials*, have been the subject of intensive research as they exhibit
14 properties that compare favorably with those of cellular metals produced by conventional
15 manufacturing processes [2, 3]. *Metallic lattices* fabricated by additive manufacturing are a
16 notable example [3, 4].

17 Metallic metamaterials offer attractive combinations of properties, as they combine opti-
18 mized architectures with unique features of metals. The latter ones comprise a large range
19 of physical and chemical properties (e.g., strength, ductility, electrical/thermal conductivity)
20 coupled with additional attributes that include recyclability and environmental durability.
21 Currently, metallic metamaterials are among the most attractive lightweight materials for
22 applications in energy and electronics [5], in transport and defense [6] as well as in biomedical
23 engineering and medicine [7]. Examples of their use include energy-absorbing structures [8],
24 biomedical implants and scaffolds [7], heat-exchangers [9] and electrodes [10] among oth-
25 ers [3, 11]. To reveal their potential in many engineering applications, a fine understanding
26 of their mechanical behavior is, often, instrumental.

27 The mechanical properties of architected materials were examined by Fleck et al. [2],
28 and results relevant to metallic metamaterials are summarized in [11, 3]. The subject now
29 comprises a vast body of theoretical and experimental work that investigates their elastic
30 (e.g., [12, 13, 14]), elastoplastic (e.g. [12, 15, 16, 17, 14, 13, 18, 19]), fracture (e.g., [20,
31 21, 22, 23, 24, 25]) and fatigue properties (e.g. [26, 27]). Collectively, these studies show
32 that the mechanical performance of metallic metamaterials mainly depends upon the internal
33 mesostructure. Important topological features comprise the pore connectivity (e.g., open or
34 closed) [12], the cell geometry [16] and the void arrangement (e.g., periodic or disordered) [14].
35 To date, the mechanical properties of metallic metamaterials have been mostly studied under
36 compression; their tensile deformation and failure have been comparatively less examined.
37 The reason for this discrepancy is two-fold. First, they have a high potential as structures
38 that absorb mechanical energy, for which compressive properties are of primary importance.
39 Second, in tension, more than in compression, internal damage is exacerbated leading to
40 a lower macroscopic ductility. At present, experimental studies on the tensile properties
41 of metallic metamaterials are (very) few and remain mostly limited to unit-cell based lat-
42 tices [14, 28, 24, 22, 17]. Examples include structures fabricated by laser powder bed fusion
43 (LPBF) using steel [24, 17], aluminum- [22] and titanium- [14] alloy powders, as well as low-
44 carbon steel honeycombs fabricated by water-jetting [28]. Collectively, these investigations
45 show that metallic metamaterials have low tensile ductility. Data for this property are highly
46 scattered [24], and display magnitudes that are by far lower than their compressive counter-
47 parts [25, 14]. For instance, titanium-based gyroid lattices produced by LPBF typically break
48 at strains less than 10% in tension and greater than 40% in compression [14]. Similar trends
49 are also observed for the flow stress and for the ultimate strength [14, 25].

50 The knockdown in tensile properties results from internal damage accumulation. The
51 latter in turn has multiple origins. One is associated with the presence of defects (such as,
52 e.g., microporosity [29] and geometric imperfections [13]), which result inevitably from man-
53 ufacturing and cause severe strain concentrations that eventually initiate the damage process.
54 Another one is the cellular topology, which dictates the deformation mechanisms [16] and
55 consequently affects the rate of damage accumulation. To date, the role of manufacturing
56 defects and topology on the failure mechanisms of metallic metamaterials produced by ad-
57 ditive manufacturing has been largely examined under compression (e.g., in [18, 19]). Their

58 combined effect on tensile failure has been scarcely documented [18]. This is the goal of the
59 present study.

60 In the following, the tensile response of metallic metamaterials produced by LPBF is
61 studied experimentally. Specifically, the metamaterials of this study contain cylindrical voids
62 randomly-dispersed into an aluminum matrix. Voids have arbitrary elliptical geometry and
63 are algorithmically generated prior to fabrication. Using digital image correlation (DIC) com-
64 bined with X-Ray computed tomography, the tensile deformation mechanisms of this class of
65 cellular solids are quantified at the mesoscale, and are shown to combine features of the duc-
66 tile fracture of metals with the defect sensitivity of additively-manufactured cellular materials.
67 The paper is organized as follows. In Section 2, the experimental methods comprising the
68 random porous design, manufacturing and testing are described. The results are presented
69 in Section 3, and give the effective tensile properties alongside the mesoscale deformation
70 fields. The experimental findings are rationalized in Section 4, where the emphasis is put
71 on elucidating the mechanisms of void growth and interaction during plastic straining. Data
72 for the average pore growth rate are compared with the well-known McClintock’s microme-
73 chanical model [30], and highlight the role of the random porous topology, combined with
74 manufacturing defects, on strain localization.

75 **2. Experiments**

76 This section describes the experimental methods used throughout the study. It begins
77 with the description of the random porous design (Section 2.1) and LPBF manufacturing
78 (Section 2.2). The testing protocol is presented in Section 2.3, while the basics principles of
79 Finite-Element (FE) based DIC (used to measure the kinematic fields during tensile defor-
80 mation) are recalled in Section 2.4.

81 *2.1. Porous design*

82 Metallic cellular architectures with randomly-dispersed cylindrical pores were designed,
83 fabricated and tested. Their porous structure was generated numerically, prior to fabrication,
84 using a random sequential adsorption (RSA) algorithm [31, 32, 33]. This procedure works
85 by placing (randomly, sequentially and irreversibly) non-overlapping particles in a cell of
86 arbitrary geometry and dimensions (e.g., 2D or 3D); it stops when the target particle content

107 is reached. Particles can have arbitrary ellipsoidal shape, size and orientation and can be made
108 from any constituent phase (e.g., solid or air), thereby enabling for the design of countless
109 composite architectures [31, 32, 33, 34, 35]. In the present work, the RSA algorithm was used
110 to generate 2D porous geometries with voided particles. More specifically, two types of porous
111 architectures were designed for this study, namely, one consisting of elliptical pores (Figure
112 1(a)), the other one of circular voids (Figure 1(b)). Both architectures had 20% nominal
113 porosity and contained 200 equisized voids. In particular, the elliptical pores had identical
114 orientation and constant aspect ratio $w = 0.3$ (this parameter designates the ratio of the pore
115 major to the minor axes).

116 Geometric models of the tested specimens were then created using the 2D RSA-generated
117 porous cells. To build the corresponding 3D models, the open-source software *Gmsh* was
118 used, and the meshes were then extruded out-of-plane prior to fabrication [33, 34]. Two
119 geometries were generated in this study, namely, cubic and dogbone. The first ones had
120 dimensions $12.5 \times 12.5 \times 12.5 \text{ mm}^3$ and were employed to examine the topological features
121 of the *as-manufactured* cellular architectures by metallographic analyses. The latter ones
122 were designed for mechanical testing. Specifically, dogbone samples had a nominal length of
123 110 mm and gauge section of $13.5 \times 4 \text{ mm}^2$ (Figure 1(c)). Their porous gauge area comprised
124 two identical porous cells (with dimensions $13.5 \times 12.5 \text{ mm}^2$), whereas the homogeneous heads
125 were 25 mm-wide. Each specimen therefore contained 400 pores (i.e. 200 per cell) that were
126 oriented, in the case of the elliptical pores (Figure 1(a)), with their major axis parallel to the
127 sample longitudinal axis.

128 The choice of the adopted porous design, comprising two nominally identical porous cells,
129 was made for a two-fold purpose. The first one was to assess the repeatability of the fabrication
130 process and, notably, of the manufacturing defects within the same specimen. The second one
131 was to probe the differences in the deformation behavior between the two porous cells. The
132 latter ones are likely to contain a different distribution/amount of manufacturing defects, as
133 shown by earlier work, e.g., for AlSi10Mg lattice structures produced by LPBF [19]. Last, in
134 the present design (Figure 1(c)), the pores between the two cells were separated by a small,
135 yet noticeable, matrix ligament. The latter extended through the whole specimen thickness
136 and resulted from the cell patterning. As shown in Sections 3.3 and 3.4, this region of the
137 specimens often contained internal defects (in the form of irregular cavities), which resulted

118 from the fabrication process and promoted strain concentrations.

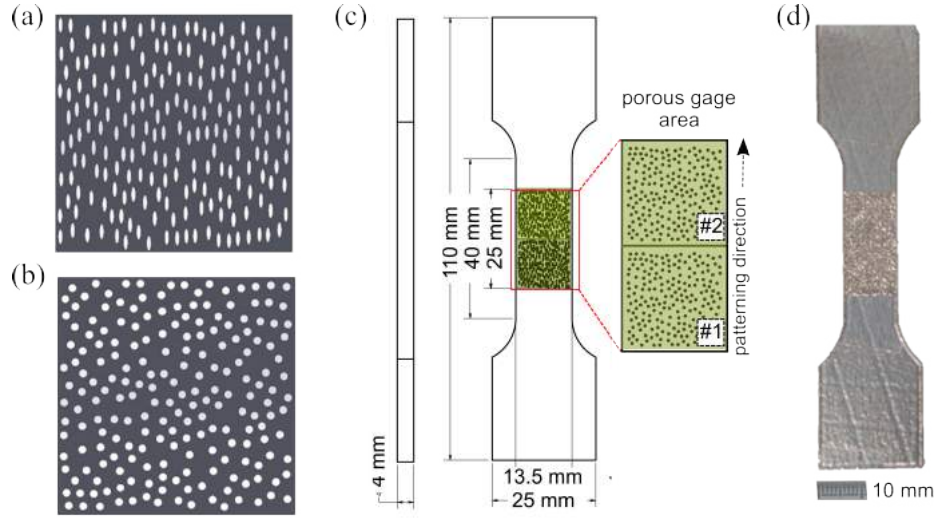


Figure 1: (a,b) Random porous architectures explored in this study. They are generated by means a numerical algorithm, and contain random distributions of equisized (a) elliptical and (b) circular pores at 20% volume fraction. The elliptical pores have identical orientation and constant aspect ratio $w = 0.3$. (c) Geometry of the tensile dogbone specimen comprising two identical porous cells. (d) Optical front-view of the *as-manufactured* tensile specimen.

119 A total of four specimen geometries were fabricated for this study using different porous
 120 realizations; their main characteristics are reported in Table 1. A short-hand designation is
 121 used in the form $M20-wY$ to distinguish each sample, where M denotes the specimen geometry
 122 (C for cubic and D for dogbone), 20 the nominal porosity and Y the pore aspect ratio w . In
 123 all specimens, pores had micrometric size with nominal (i.e., *as-designed*) average diameter of
 124 $500 \mu\text{m}$ (Table 1). This parameter corresponds to the average value between the pore major
 125 and minor axes, which have both identical values for specimens containing circular voids.

Table 1: Porous specimen specifications

Designation	No of pores	Average pore size	Pore aspect ratio
D20-w1	400	$450 \mu\text{m}$	1
D20-w03	400	$500 \mu\text{m}$	0.3
C20-w1	200	$450 \mu\text{m}$	1
C20-w03	200	$500 \mu\text{m}$	0.3

126 *2.2. Manufacturing*

127 The porous specimens were manufactured by LPBF (Figure 1(d)) using a SLM HL 125
 128 machine (SLM Solutions GmbH, Germany). It was equipped with a 500-W Yb:YAG laser,
 129 whose wavelength and spot diameter were 1070 nm and 70 μm , respectively. A commercial
 130 AlSi10Mg powder (TEKNA Advanced Materials Inc, Sherbrook, Canada) was employed for
 131 fabrication. It was produced by plasma spheroidization and contained, in addition to Al,
 132 ~ 10 wt% of Si and ~ 0.25 wt% of Mg. The specifications of its particle size distribution were
 133 provided by the manufacturer and are $D_{10} = 20$ μm and $D_{90} = 63$ μm . The fabrication was
 134 performed under argon atmosphere, and the oxygen content was maintained below 1000 ppm
 135 during the entire process. Printing was carried out using an incremental scanning strategy,
 136 with 67° rotation between two consecutive layers, and a constant powder layer thickness of
 137 30 μm . Since the specimens contained through-thickness cylindrical pores, their building
 138 direction was parallel to the longitudinal axis of the pores. Moreover, each sample was pro-
 139 duced using two contour passes followed by a hatch (i.e., volume). The selected manufacturing
 140 parameters are reported in Table 2.

Table 2: LPBF-process parameters used in this study

	Laser power	Scan speed	Hatch distance	Beam offset [†]
Volume	300 W	400 mm/s	170 μm	
Contour	300 W	1000 mm/s	100 μm	170 μm

[†] it designates the distance between the nominal contour of the part and the center of the laser beam

141 *2.3. Mechanical tests*

142 The porous dogbone specimens D20-w1 and D20-w03 (Table 1) were tested in tension.
 143 Prior to the experiments, they were subjected to a standard T6-like heat treatment in order
 144 to homogenize the microstructure [36]. It consisted in a solution annealing at 520°C for 1 h
 145 followed, in successive order, by water quenching at room temperature, and by artificial aging
 146 at 160°C for 6 h. Temperature and interval time recommendations provided in Ref. [36] were
 147 adopted. Several studies have shown that the T6-like heat treatment promotes microstructure
 148 recrystallization [36, 37], and in turn improves the alloy ductility and fatigue strength [38].

149 The experiments were conducted under a displacement control mode at a nominal strain
 150 rate of $\dot{\epsilon} \leq 10^{-4} \text{ s}^{-1}$ until sample failure. The experimental setup used in this study is shown
 151 in Figure 2. It comprised (i) a uniaxial testing machine equipped with a $100 \pm 0.01 \text{ kN}$ load cell;
 152 (ii) two optical cameras and (iii) an external unit for image acquisition. The two cameras were
 153 placed on either side of the specimen to track the deformation of the outer porous surfaces.
 154 In the following, they are designated as Cameras F and R (where F stands for front and R
 155 for rear, Figure 2). The first one employed a telecentric lens with a resolution $\simeq 25 \mu\text{m}/\text{px}$;
 156 the latter one was equipped with a standard lens with a resolution of $\simeq 50 \mu\text{m}/\text{px}$.

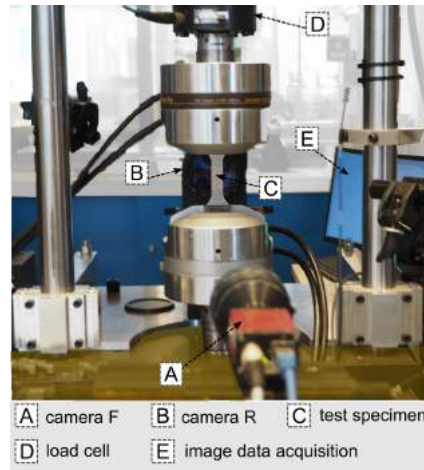


Figure 2: Tensile testing apparatus used in this study. It comprises a uniaxial testing machine equipped with a 100-KN load cell (D) and two optical cameras (A, B) placed on either side of the specimen surface (C). Images from both cameras were collected by means of an external data acquisition system (E) and were used to measure the local deformation fields of the specimen surfaces.

157 After tensile testing, specimens D20-w1 and D20-w03 were scanned using of a X-Ray
 158 tomograph (North Star Imaging X50+). A total of 900 radiographs per scan was collected
 159 while rotating the sample between 0 and 360° . An average of 10 images were taken at each
 160 rotation step. The X-ray tube was operated at an acceleration voltage of 150 kV using a
 161 tungsten transmission target with a $100 \mu\text{A}$ current. The resolution was $\simeq 18 \mu\text{m}/\text{voxel}$. A
 162 filtered-back projection algorithm was employed to reconstruct 3D images from the series of
 163 acquired radiographs. No filtering was applied.

164 *2.4. Strain measurements by Finite Element-based DIC*

165 Strain was quantified on two opposite surfaces of each dogbone specimen via DIC. Notably,
 166 the Green-Lagrange strain fields were computed from the displacement fields, obtained from
 167 each correlation process using a Finite Element discretization (Appendix 7). In what follows,
 168 the DIC method used throughout the study are briefly recalled.

169 *2.4.1. Digital Image Correlation*

170 The present section summarizes the basic principles of FE-based DIC. The correlation
 171 technique was used herein to measure displacement fields $\mathbf{u}(\mathbf{x})$ extracted from images ac-
 172 quired during the two tests. This operation is possible thanks to the application of gray level
 173 conservation

$$f(\mathbf{x}) = g(\mathbf{x} + \mathbf{u}(\mathbf{x})) \quad (1)$$

174 between two images g and f , corresponding respectively to the deformed and reference con-
 175 figurations. The correlation procedure aims at minimizing the gray level residuals $\varphi(\mathbf{x})$ over
 176 the chosen Region Of Interest (ROI) with respect to the nodal displacements gathered in the
 177 column vector $\{\mathbf{u}\}$

$$\varphi(\mathbf{x}) = f(\mathbf{x}) - g(\mathbf{x} + \mathbf{u}(\mathbf{x}, \{\mathbf{u}\})). \quad (2)$$

178 Thus, the aim of the DIC code is to iteratively minimize the sum of squared differences

$$\Phi_c^2 = \sum_{ROI} \varphi_c^2(\mathbf{x}) \quad (3)$$

179 with respect to the unknown degrees of freedom $\{\mathbf{u}\}$ of the parameterized displacement fields.
 180 Since a FE approach was adopted, the displacement field is written as

$$\mathbf{u}(\mathbf{x}) = \sum_n u_n \psi_n(\mathbf{x}), \quad (4)$$

181 where u_n denote the unknown nodal displacements, and ψ_n the corresponding shape functions.
 182 It is worth noting that the actual pore positions were recovered through backtracking (see
 183 Appendix 7).

184 In global approaches, the continuity of displacements is assumed *a priori*, which corre-
 185 sponds to a first regularization. To speed-up convergence to the solution, the displacement
 186 field can be further regularized by means of a penalty term that acts as a low-pass filter [39].

187 Regularization may apply on total or incremental displacements. The method adopted in the
 188 current work considered incremental regularization [40], from one step of the correlation anal-
 189 yses to the next, which constitutes a milder approach adapted for plasticity. The penalization
 190 is based on the equilibrium gap [41], which states that for force-free nodes the corresponding
 191 nodal force should cancel out

$$\Phi_m^2(\{\partial\mathbf{u}\}) = \{\partial\mathbf{u}\}^\top [\mathbf{K}]^\top [\mathbf{K}] \{\partial\mathbf{u}\}, \quad (5)$$

192 where $\{\partial\mathbf{u}\}$ is the column vector gathering all nodal displacement increments between image
 193 n and $n + 1$, $[\mathbf{K}]$ the rectangular stiffness matrix associated with the force-free nodes. The
 194 total cost function then becomes

$$\Phi_t^2 = \Phi_c^2 + \omega\Phi_m^2, \quad (6)$$

195 where the weight ω is selected by the user, and is proportional to a (regularization) length
 196 ℓ_{reg} raised to the power 4 [39]. In the present work, the DIC technique was implemented
 197 within Matlab and C++ kernels developed at our laboratory [42].

198 2.4.2. Regularization length and uncertainty

199 When starting any DIC analysis, the ROI is to be defined, meshed, and a regularization
 200 length must be selected in order to reduce uncertainties and help the convergence of the
 201 correlation algorithm. A value that does not filter out too many useful details is to be chosen.
 202 When performing the analyses, the uncertainty estimate was used to evaluate the accuracy
 203 of the measurements. In order to evaluate the measurement uncertainties, prior to starting
 204 the experiments *per se*, about 100 pictures were acquired for the unloaded specimen. In the
 205 present case, only the standard strain uncertainties are reported. They were assessed for each
 206 camera of each tested specimen. Table 3 shows that for the selected regularization lengths,
 207 the standard strain uncertainties were all less than $3 \cdot 10^{-4}$.

Table 3: Uncertainty quantification of longitudinal strain (E_{yy}) fields. In addition to the standard uncertainties, the regularization lengths (ℓ_{reg}) and mean element sizes (ℓ) are reported for each considered case

	ℓ_{reg}	Std(E_{yy})	ℓ	ℓ_{reg}	Std(E_{yy})	ℓ
	Camera F	Camera F	Camera F	Camera R	Camera R	Camera R
D20-w1	30 px	$3.1 \cdot 10^{-4}$	3.6 px	40 px	$1.9 \cdot 10^{-4}$	3.2 px
D20-w03	30 px	$1.2 \cdot 10^{-4}$	3.7 px	40 px	$2.9 \cdot 10^{-4}$	4.0 px

208 A regularization length of 30 px was adopted for Camera F acquisitions, which was char-
 209 acterized by a lower acquisition noise level; whereas a value of 40 px was adopted for Camera
 210 R. Beyond random factors such as the applied speckle pattern and stroke fluctuations (the
 211 hydraulic testing machine was on and the feedback loop was active), the uncertainty was also
 212 impacted by the camera, element size and regularization length. In this particular case, higher
 213 values of uncertainty around the porous region were due to smaller element sizes needed to
 214 describe the pore geometry. Despite these various sources, the standard uncertainty levels
 215 remained rather small in comparison to the strain levels that developed in the four analyses.

216 As reported in Table 3, the element sizes ℓ adopted for the porous specimens studied via
 217 DIC was of the order of 3-4 px. This very fine discretization was needed in order to capture
 218 local strain distributions around pores and required the use of mechanical regularization.

219 3. Results

220 This section reports the major experimental results of this study and includes both mi-
 221 crostructural observations (Section 3.1) and data measured during mechanical tests (Sec-
 222 tions 3.2 to 3.4). The macroscopic tensile response of the AlSi10Mg random cellular meta-
 223 materials is presented in Section 3.2, while the fracture surfaces are examined in Section 3.3.
 224 The latter also reports a thorough investigation, carried out by means of X-ray computed to-
 225 mography, of the matrix internal defects resulting from LPBF manufacturing. The mesoscale
 226 strain patterns measured by FE-based DIC, using meshes consistent with the pore topology,
 227 are displayed in Section 3.4.

228 *3.1. LPBF-manufactured cellular topologies*

229 Optical microscopy was used to examine the topological features of the as-fabricated cubic
230 porous samples C20-w1 and C20-w03 (Table 1). Micrographs are presented in Figure 3 and
231 are contrasted with the corresponding geometrical models.

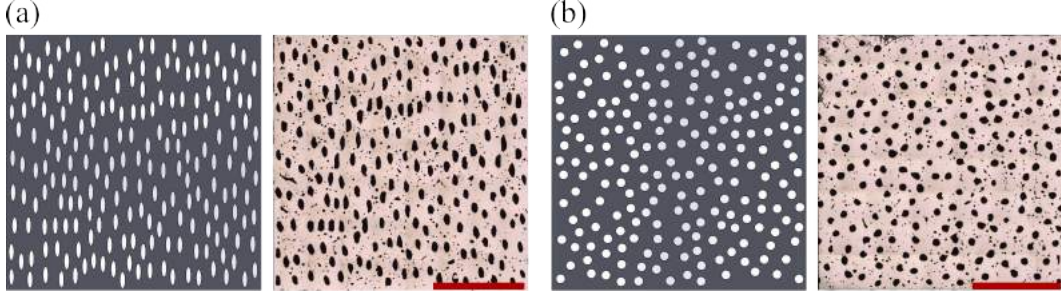


Figure 3: Optical micrographs of the AlSi10Mg cubic porous samples (a) C20-w03 and (b) C20-w1 containing respectively a random packing of elliptical and circular pores. The porous specimen specifications are given in Table 1. For both porous architectures, experimental micrographs are contrasted with the corresponding geometrical model. Scale bar: 3 mm.

232 For both porous architectures, the presence of manufacturing defects, such as matrix in-
233 ternal porosity and geometric imperfections, was observed. Internal matrix porosity occurred
234 in the form of small and irregular voids within the matrix. Topological defects were instead
235 distinguished by morphological mismatches between the as-designed and the as-manufactured
236 pore geometry. Notably, the printed pores were found to have non-uniform ellipticity, lower
237 roundness and more irregular size than their nominal (*defect-free*) counterparts. For example,
238 Figure 3(a,b) reveals that geometric voids were smaller near the cell edges and had a more
239 polyhedral shape, when they are closely-spaced.

240 *3.2. Stress-strain response*

241 The tensile response of AlSi10Mg dogbone samples D20-w03 and D20-w1 is reported in
242 Figure 4. For both pore geometries, data for the macroscopic engineering stress are plotted
243 against the longitudinal strain $\langle E_{yy} \rangle$. This last quantity is computed by means of FE-based
244 DIC (Section 2.4) and corresponds, unless specified otherwise, to the average axial strain
245 measured over the entire porous gauge area (Figure 1(c)). In the interest of comparison,
246 measurements from both cameras are reported using different line styles. Collectively, data in

247 Figure 4 show similar trends. In particular, three distinct stages of deformation corresponding
 248 to (i) initial elasticity, (ii) strain hardening up to the peak stress and (iii) final strain softening,
 249 are discerned from the stress-strain curves. For both porous architectures, the macroscopic
 250 tensile curve exhibits a departure from linearity at a strain level of about 0.1% (this value is
 251 denoted by the dotted line a in Figures 4(a,b)). Upon yielding, the response strain-hardens
 252 and the tensile flow stress increases continuously up to the peak load. The measured flow stress
 253 is notably higher for specimens D20-w03 (containing elliptical voids), and reaches its maximum
 254 value at a strain between 0.6 and 0.7% for both specimen geometries (Figures 4(a,b)). Beyond
 255 the peak stress, the tensile response of both samples is characterized by the occurrence of
 256 repeated stress drops; strain softening continues until final fracture of the sample. Overall,
 257 the results are also fairly reproducible. For each specimen geometry, the difference between
 258 the average strain measured by the two cameras remains very small until the attainment of
 259 the peak stress, and only increases slightly (past this point) as a result of damage softening.
 260 The highest strain difference is of the order of $\approx 0.3\%$, and is measured for specimen D20-w03
 261 (as seen by comparing, e.g., curves in Figures 4(a,b) past 1% strain denoted with the dotted
 262 line e).

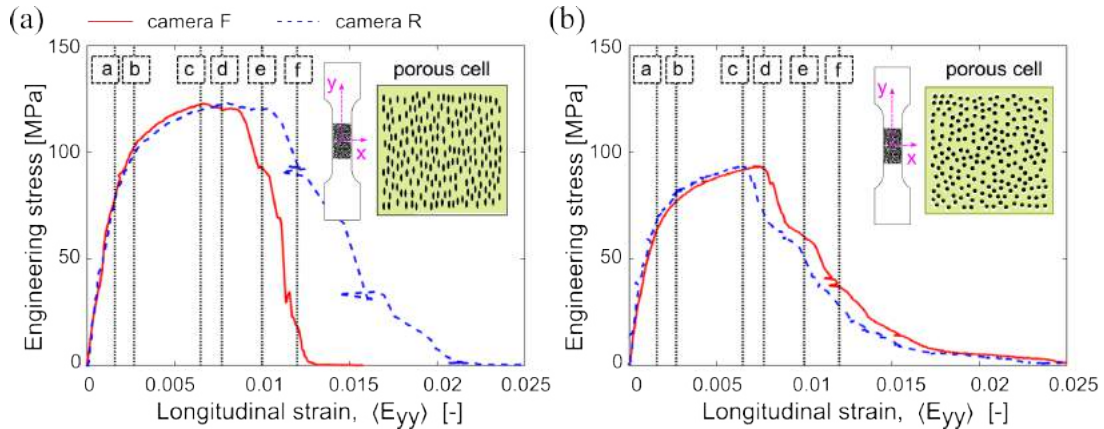


Figure 4: Experimental tensile stress-strain curves for the AlSi10Mg porous samples (a) D20-w03 and (b) D20-w1. The specimens specifications are given in Table 1, whereas their porous cell is shown in the inset of each figure. For both samples, the macroscopic stress is plotted against the average longitudinal strain $\langle E_{yy} \rangle$ computed over the entire porous gauge area by means of FE-based DIC. Strain measurements from images collected by the two cameras are denoted with different line colors. Note that Camera F of sample D20-w03 could capture only one porous cell.

263 *3.3. Post-mortem X-Ray Tomography*

264 Tomographic sections from reconstructed volumes of the dogbone samples, scanned after
265 mechanical testing, are presented in Figures 5(a,b). In the interest of comparison, images
266 corresponding to the outer and mid sections are reported for each porous geometry. As
267 seen, the fracture surfaces of the two mesoporous specimens display distinct characteristics.
268 In sample D20-w1 containing circular pores (Figure 5(a)), fracture occurred in the lower
269 porous cell and exhibited a band-like crack path. The latter was inclined with respect to
270 the loading axis and traversed a large number of voids across the specimen's width. By
271 contrast, the fracture path displayed by sample D20-w03 with unidirectional elliptical voids
272 (Figure 5(b)) was more irregular, and involved an extended region of the specimen comprised
273 between the two porous cells. Failure, moreover, encompassed only a few voids (Figure 5(b)).
274 Collectively, the observed fracture features are reproducible across the scanned volumes and
275 no significant differences can be discerned among the three tomographic sections, for each
276 specimen geometry.

277 Tomography also reveals the presence of matrix internal defects, such as small pores and
278 irregular cavities. To examine their morphological characteristics, a representative portion
279 of the matrix volume within sample D20-w03 was analyzed using the *Avizo* software. This
280 is highlighted in Figure 5(b) and consisted of $794 \times 214 \times 236$ voxels³ for a total volume of
281 $\sim 164.25\text{mm}^3$. Results of the tomographic reconstructions are presented in Figures 5(c,d),
282 whereby the Al-alloy matrix is shown with full transparency. To improve the presentation,
283 geometrical pores are depicted with blue color. Tomographic images show two populations of
284 matrix defects, whose total volume fraction was $\approx 2.4\%$. This last quantity was computed with
285 respect to the total volume of the region highlighted in Figure 5(b), using the *Avizo* software.
286 Matrix defects can be discriminated on account of their morphology, and are represented
287 with red and yellow colors (Figures 5(c,d)). The first ones have smaller sizes and display a
288 bubble-like shape. The second ones have larger and more irregular shape (see, e.g., defects
289 denoted as #1 and #2).

290 The findings of Figures 5(c,d) corroborate the metallographic observations reported in
291 Figures 3(a,b) (for different realizations of the same random topologies), and also indicate
292 that defects shown here result from LPBF. Notably, their distribution and morphology depend
293 on the adopted laser strategy. Figure 5(c) illustrates this point. It shows that bubble-like

294 defects formed in the matrix ligament surrounding the geometric pores, and lied along the
 295 laser *contour* trajectories.

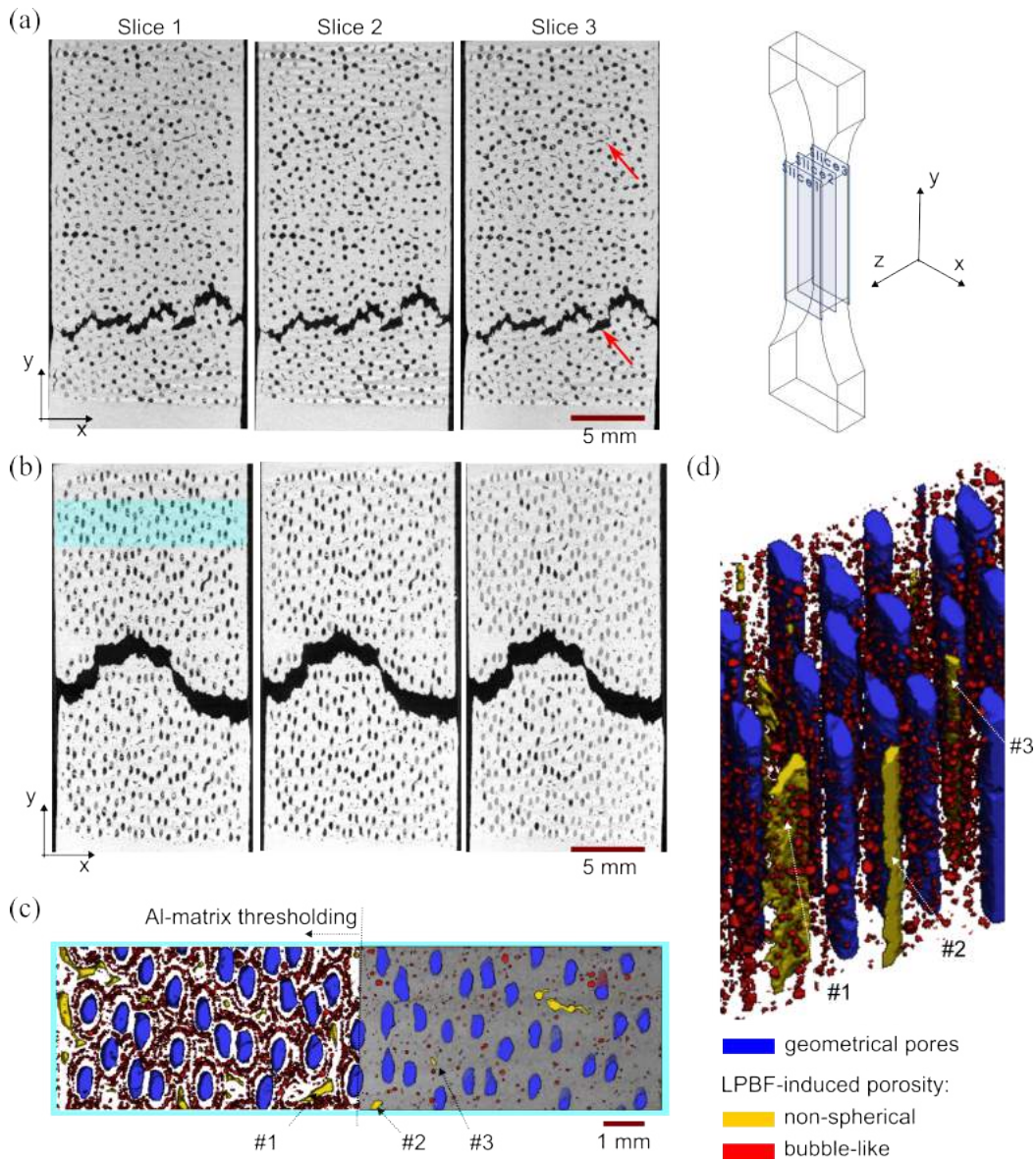


Figure 5: Tomographic sections of the reconstructed internal volume of specimens (a) D20-w1 and (b) D20-w03. They were scanned after mechanical testing. Representative examples of large matrix defects occurring at identical locations within the specimen are indicated with red arrows in sub-figure (a). (c) Front and (d) isometric view of a selected portion of the reconstructed volume for specimen D20-w03, highlighted with light-blue color in sub-figure (b) and showing the presence of process-induced porosity within the matrix. To improve the presentation, the Al-alloy matrix is shown with full transparency using the *Avizo* software.

296 Figure 5(d), instead, reveals that irregular defects (depicted with yellow color) extended
 297 consistently throughout the sample thickness, which in turn corresponds to the part build-
 298 ing direction. This observation is consistent with the tomographic section reconstructions
 299 reported in Figures 5(a,b). The latter ones also show that defects of this type occurred at
 300 about the same location within the specimen, irrespective of its cellular topology. An exam-
 301 ple is shown in Figure 5(a)-Slice 3 and is highlighted with red arrows. Figures 5(c,d) also
 302 indicate that bubble-like defects are present in the largest amount within the matrix of the
 303 analyzed sub-volume. To quantify their morphological characteristics, measurements of the
 304 volume and shape factor (sphericity) were obtained, for each reconstructed pore defect, using
 305 built-in *Avizo* functions. Their statistical distribution is reported in Figure 6. Figure 6(a) dis-
 306 plays the volume distribution computed from data obtained via the *Volume3D* function, and
 307 shows that most of the bubble-like defects did not exceed 10^{-5} mm³ in volume. Pores whose
 308 volume is greater than 10^{-3} mm³ are rare (see the inset in Figure 6(a)). The distribution in
 309 Figure 6(b) also indicates that most of the bubble-like pores are characterized by a sphericity
 310 factor close to unity. This last quantity was obtained via the *ShapeVA3D* function (where
 311 $\text{ShapeVA3D} = A^3/(36\pi V^2)$ was computed, for each pore, from its volume V and area A), and
 312 has a value equal to unity when the pore is a perfect sphere. Representative examples of the
 313 3D pore shapes associated with an increasing value of the sphericity factor, in parts built by
 314 LPBF, are given in Figure 9 in Ref. [43]. Collectively, the data of Figure 6(b) corroborate the
 315 observations in Figures 5(c,d) showing that bubble-like defects were mainly of spherical shape.
 316 To quantify their size, the pore equivalent diameter D_{eq} was computed for all defects with
 317 sphericity varying between 0.8 and 1.2 from the reconstructed volume (where the latter was
 318 assumed to be spherical). Measurements of mean equivalent diameter, at distinct sphericity
 319 intervals, are reported together with their standard deviations in the inset of Figure 6(b), and
 320 display values ranging from 40 μm to 80 μm .

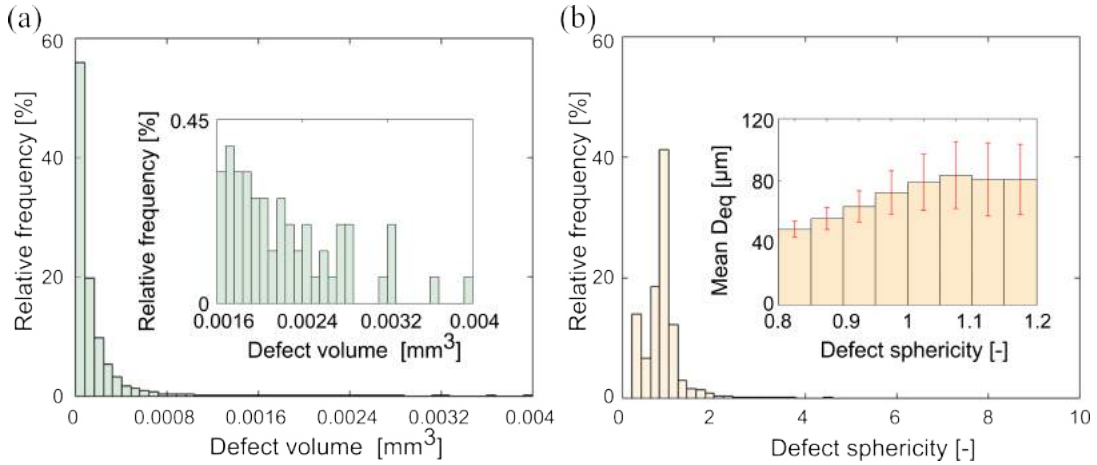


Figure 6: (a) Volume and (b) sphericity distributions of the bubble-like defects within the sub-volume highlighted with light blue color in Figure 5(b). The geometrical characteristics were extracted from the tomographic volume reconstructions using *Avizo* software. To improve the presentation, the relative frequency of the pore defects with volume larger than $1.6 \cdot 10^{-3}$ is shown in the inset of sub-figure (a) using a magnified scale. Likewise, data for the mean equivalent diameter D_{eq} are reported, together with their standard deviations, in the inset of (b) for all defects with a sphericity varying between 0.8 and 1.2. For each pore, D_{eq} was computed from its reconstructed spherical volume.

3.4. Mesoscale strain maps

The local strain fields could be computed by means of FE-based DIC (Section 2.4). For both porous specimens D20-w03 and D20-w1, the axial strain fields are presented in Figures 7 and 8, respectively. The mesoscale strain maps were computed from images collected by the two cameras and are reported for discrete instants of the deformation history. They were selected from the measured stress-strain curves, and correspond to the macroscopic strain levels $\langle E_{yy} \rangle$ denoted with dotted lines in Figures 4(a,b). Results show that:

- upon departure from elasticity, strains have inhomogeneous distribution (Figures 7(a) and 8(a)). When $\langle E_{yy} \rangle \approx 0.15\%$, areas of increased strain levels are observed across the porous surface of both tested specimens, and are uniformly distributed within the gauge area;
- upon strain hardening, distinct strain concentration bands start appearing in both porous specimens (Figures 7(b) and 8(b)). They have short lengths, encompassing groups of few closely-spaced pores, and widths larger than one pore-diameter. Inter-

335 estingly, the short bands occur at different locations in both specimens and display a
336 symmetric pattern in sample D20-w1 containing circular voids (Figure 8(b));

337 • near the peak stress, the existing short bands continue to accumulate strains and new
338 ones form with further plastic flow (Figures 7(c,d) and 8(c,d). The interactions between
339 pores become evident and occur, in most cases, within short bands inclined with respect
340 to the loading axis in both tested specimens;

341 • upon strain softening, a change in the deformation patterns is observed for both porous
342 specimens. Some of the short bands become inactive and the whole deformation con-
343 centrates around the fractured zone (Figures 7(e,f) and 8(e,f). Consistent agreement
344 is found between the strain patterns of Figures 7(f) and 8(f)) and the fracture paths
345 displayed in Figures 5(a,b).

346 The results from the two cameras are also similar, thereby indicating that the deformation
347 was pretty uniform along the sample thickness. This was expected given the specimen geom-
348 etry, which has extruded porous sections along the building direction. The slight difference
349 between the strain maps, measured with the two cameras, may be rationalized by considera-
350 tion of their different optical resolution. The latter was higher for camera F equipped with a
351 telecentric lens (Section 2), and allowed the deformation features to be more finely captured.
352 Examples are given in Figures 8(a) and 7(b), whereby zones of strain concentration around
353 large matrix defects could be measured with camera F images and benchmarked against to-
354 mographic scans for Slice 1 in Figures 5(a) and 5(b), respectively. Interestingly, the strain
355 maps also show that the strains concentrate in regions of the matrix ligament between the
356 two porous cells (Figures 8(c) and 7(c) obtained with camera F), where irregular pore defects
357 resulted from the LPBF contour strategy (Figure 5(a)). This finding is consistent with the
358 results of earlier work on AlSi10Mg lattice structures produced by LPBF [19], where large
359 irregular pores were observed in the matrix region connecting the elemental struts. Figure 8
360 also proves that, even though the two porous cells were nominally identical, the corresponding
361 strain patterns became different very early on with increasing macroscopic strain. It is be-
362 lieved that such differences were due to the manufacturing defects present within the matrix
363 of the two cells, Figure 5(a).

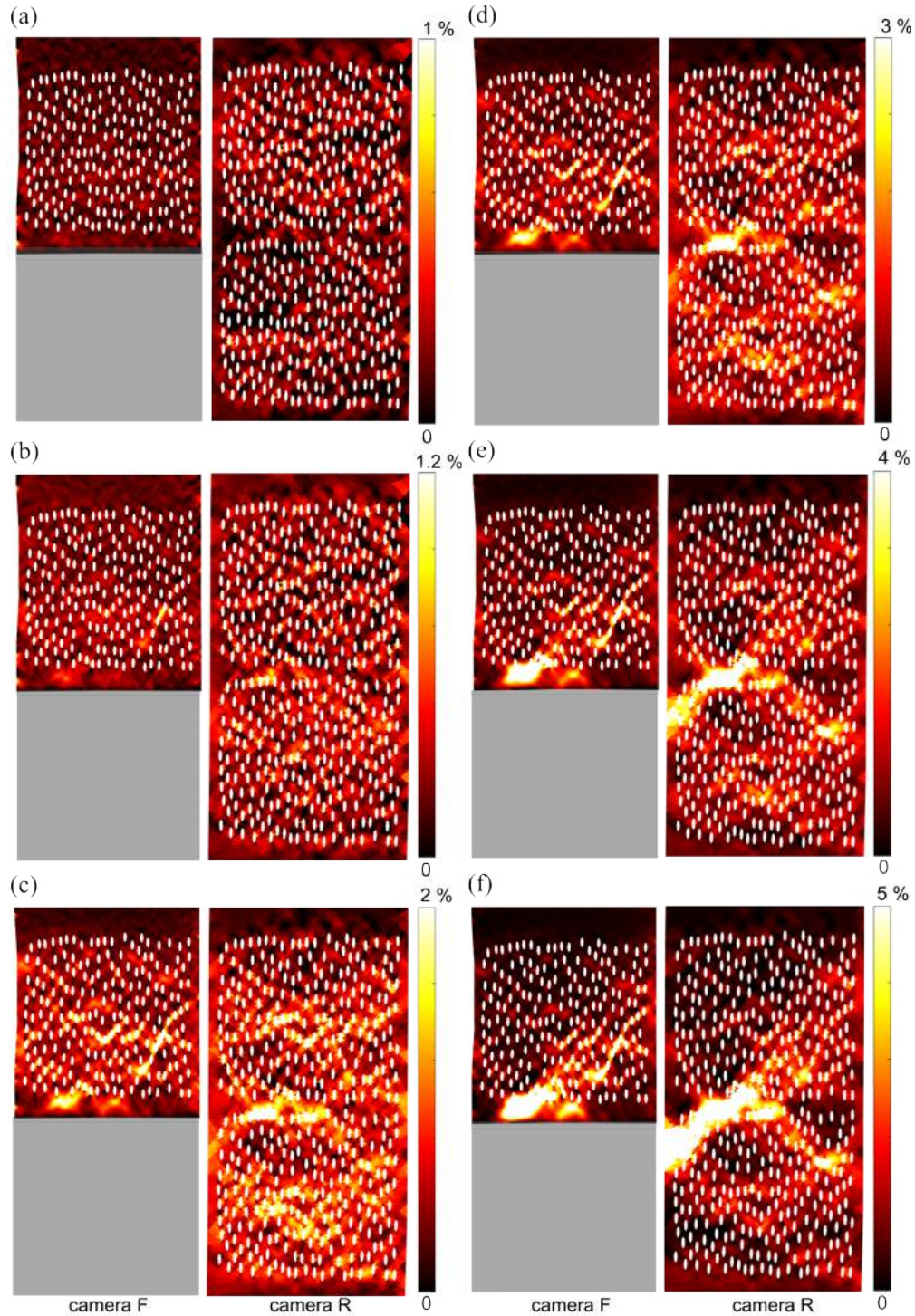


Figure 7: Local axial strain maps for specimen D20-w03 at discrete points of the deformation history. The macroscopic longitudinal strain $\langle E_{yy} \rangle$ is equal to (a) 0.16%, (b) 0.27%, (c) 0.65%, (d) 0.77%, (e) 1% and (f) 1.2%. These levels are depicted with dotted lines in Figure 4(a). The strain fields were computed by means of FE-based DIC using images collected simultaneously by both cameras. Note that since Camera F of sample D20-w03 could capture only one porous cell, the corresponding strain map is reported for the upper cell.

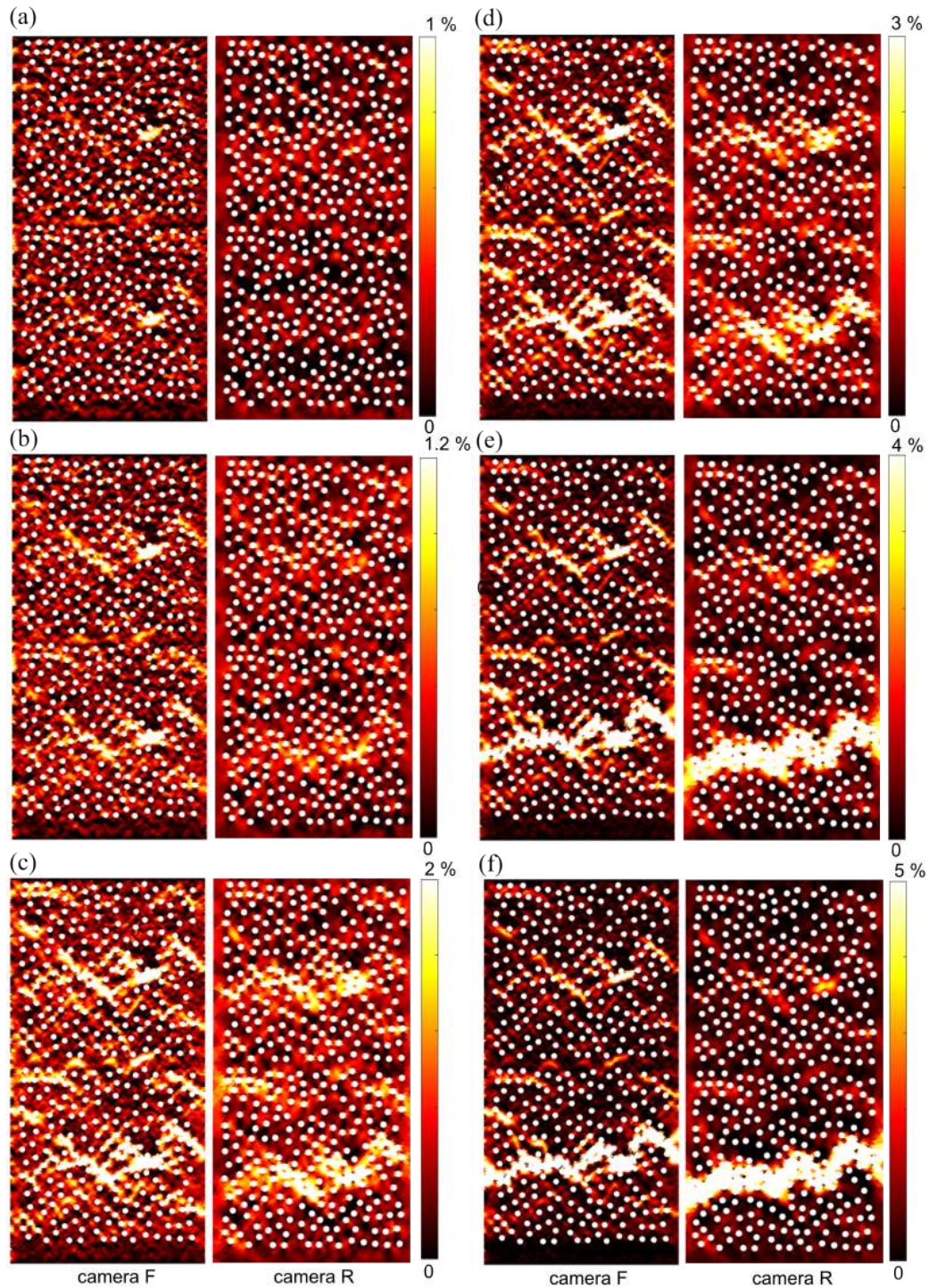


Figure 8: Local axial strain maps for specimen D20-w1 at discrete points of the deformation history. The macroscopic longitudinal strain $\langle E_{yy} \rangle$ is equal to (a) 0.16%, (b) 0.27%, (c) 0.65%, (d) 0.77%, (e) 1% and (f) 1.2%. These levels are depicted with dotted lines in Figure 4(b). The strain fields were computed by means of FE-based DIC using images collected simultaneously by both cameras.

364 4. Discussion

365 The main findings of this work are discussed hereafter. First, the effective tensile proper-
366 ties of the AlSi10Mg random cellular metamaterials are summarized in Section 4.1, and are
367 rationalized on account of the cellular topology combined with the presence of manufacturing
368 defects (Section 4.2). The mechanisms of internal damage accumulation are then examined in
369 Section 4.3, and are quantified using DIC data for the pore true strains. Measurements of the
370 average growth rate are also compared with predictions of the McClintock’s micromechanical
371 model (Section 4.4), and the main implications for models of ductile fracture are discussed.

372 4.1. Tensile properties

373 The random cellular architectures of this work have low elongations to failure despite the
374 inherent ductility of the alloy they are made from (Figure 4). Their tensile properties could
375 be computed from data in Figure 4 and are summarized in Table 4. In data presented herein,
376 the Young’s modulus along the longitudinal direction, i.e. E , was measured through linear-
377 regression of the elastic segment of the tensile stress–strain curve, and is reported together
378 with the R^2 coefficient. The yield stress $\sigma_{0.1\%}$ was computed at 0.1% offset strain, whereas
379 the ultimate tensile strength σ_{UTS} corresponds to the peak stress in the tensile response.
380 The elongation at the maximum load is denoted as ϵ_{UTS} . For each specimen tested, data
381 for E , $\sigma_{0.1\%}$ and ϵ_{UTS} were computed using strain measurements from each camera, and are
382 reported together with their average value. Measurements of σ_{UTS} are, instead, independent
383 of the camera sensor. In the interest of comparison, the tensile properties of the AlSi10Mg
384 matrix are also given in Table 4. They were measured in a separate set of experiments on
385 heat-treated bulk samples (produced and tested as described in Section 2), and are consistent
386 with data from literature [44, 36, 38].

387 Collectively, the results in Table 4 show that the two cellular architectures have similar
388 elastic stiffness but distinct plastic properties. For both porous specimens, the values of the
389 Young’s modulus E differ by less than 8% and are slightly higher for sample D20-w03 consist-
390 ing of elliptical voids. The yield stress $\sigma_{0.1\%}$ and ultimate tensile strength σ_{UTS} , instead, have
391 magnitudes that are up to 40% lower for specimen D20-w1 containing circular pores. Since
392 both samples have equal relative density and contain likely the same amount of internal ma-
393 trix porosity, the measurements in Table 4 indicate that the pore shape exerts no significant

Table 4: Tensile properties of the AlSi10Mg random cellular architectures of this study and corresponding values measured for the bulk alloy matrix in the same metallurgical state.

		E [GPa]	$\sigma_{y0.1\%}$ [MPa]	σ_{UTS} [MPa]	ϵ_{UTS} [%]
D20-w1	Camera F	38 ($R^2 = 0.988$)	78.6		0.73
	Camera R	39.7 ($R^2 = 0.92$)	82.8		0.65
	average	38.8 ± 0.8	80.7 ± 2.1	93.3	0.69 ± 0.06
D20-w03	Camera F	40.2 ($R^2 = 0.97$)	112.9		0.61
	Camera R	43 ($R^2 = 0.964$)	112		0.72
	average	41.6 ± 1.4	112.4 ± 0.4	122.8	0.67 ± 0.07
AlSi10Mg matrix	Camera F	72.6 ($R^2 = 0.991$)	122.6		14.8
	Camera R	65.9 ($R^2 = 0.883$)	128.6		15.2
	average	69.2 ± 4.7	125.6 ± 10	204.6	15 ± 0.5

394 influence during elastic deformation. This agrees well with what reported for other random
395 media, containing either solid [45] or voided particles [46]. Conversely, the effect of the void
396 shape becomes more important upon plastic straining in agreement with prior work [47]. In
397 particular, the observed tendency for the tensile strength to decrease as the aspect ratio w
398 increases (compare values of σ_{UTS} for samples D20-w1 and D20-w03 in Table 4) is consistent
399 with trends observed in porous metals produced by unidirectional solidification [48]. These
400 materials have similar cellular topologies and display values of σ_{UTS} that scale as ρ^{1+2w} , where
401 ρ is the relative density (see Equation (14) in Ref. [48]).

402 Equally important, data in Table 4 indicate that the measured ϵ_{UTS} are of the order
403 of $\sim 0.7\%$ and thus have levels which are one order of magnitude lower than for the bulk
404 alloy. These observations are consistent with what reported for most conventional Al-based
405 foams loaded in tension [49, 50, 51, 52], which typically fractured at strains below a few
406 percents. Prior work also shows that data for the tensile ductility of metallic random foams
407 are highly scattered (see, e.g., Figure 1 of Ref. [53]) and reflect a maximized influence of
408 internal damage accumulation on tensile deformation. For example, San Marchi et al. [51]
409 investigated the tensile response of replicated pure Al and Al-12Si foams at equal relative
410 density, and found that failure strains never exceeded a few percents even for foams made
411 from a very ductile metal (i.e., pure Al). Moreover, measured elongations-to-failure were up to

412 seven times lower for the Al-12Si foam containing brittle Si phases. Collectively, the findings
413 of this study highlighted the role of internal damage on the low tensile ductility exhibited
414 by metallic cellular materials and were also corroborated by several numerical studies by
415 Mangipudi et al. [54, 55].

416 The cellular materials produced for this study were made of peak-aged AlSi10Mg alloy
417 and thus contained nearly 10% of Si, typically in the form of globular particles with sizes
418 of a few μm [36, 44, 37]. In bulk form, the alloy exhibits a highly ductile behavior that
419 results from the nucleation and coalescence of microscopic voids (see Refs. [37, 36] and data
420 in Table 4), and is distinguished by the presence of equiaxed dimples observed across the
421 fractured surface. Prior work reported that the average size of the dimples depended on
422 the solution treatment temperature, and reached 5 μm when the alloy was peak-aged above
423 500°C [37]. Similar features were observed across the fracture surfaces of the broken sample
424 D20-w03 (Figures 9(a,b)), which showed the presence of equiaxed dimples within the matrix
425 (Figure 9(c)). The SEM fractography also revealed that the dimples had a size of a few
426 micrometers (in agreement with earlier work [37]), and were one order of magnitude smaller
427 than the spherical pore defects produced during LPBF fabrication (Figure 6(b)). The latter
428 ones (highlighted with dashed yellow circles in Figure 9(b)) exhibited morphological features
429 that are in quantitative agreement with the tomographic observations reported in Section 3.3.
430 Figure 9(a,b) also shows the presence of unmolten AlSi10Mg powder within the geometric
431 pores. This last observation is common in parts produced by LPBF, and in turn explains why
432 identical pores have different X-Ray attenuation across the thickness of the reconstructed
433 specimen volume (Figures 5(a,b)).

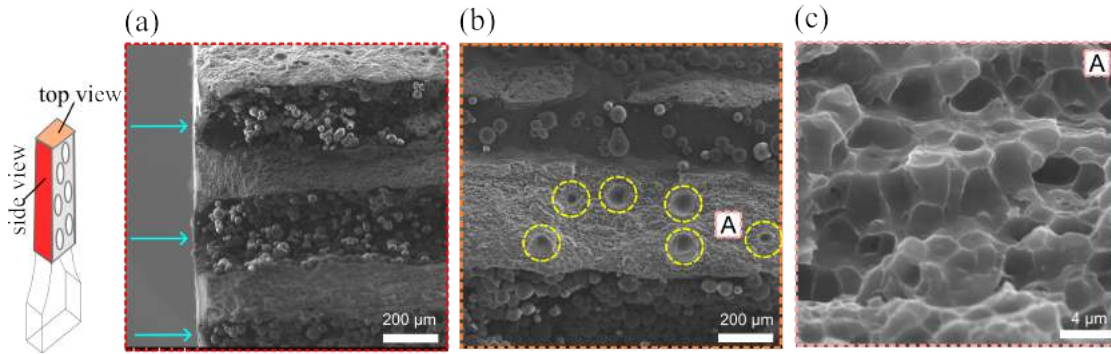


Figure 9: SEM micrographs of the fractured surfaces of sample D20-w03 containing elliptical pores. (a) Side and (b) top views, whereby the through-thickness geometric pores and the matrix internal defects are indicated with cyan arrows and dashed yellow circles in sub-figures (a) and (b), respectively. (c) Close-up view of the matrix region, denoted by [A] in sub-figure (b), revealing the presence of equiaxed dimples with size of a few micrometers.

434 The presence of dimples in the matrix ligament between geometric pores (Figure 9(c))
 435 also confirms that brittle fracture of the Si phase was not the dominant cause of internal
 436 damage accumulation in the present cellular metamaterials. Another reason for their reduced
 437 ductility must exist. This observation points to a geometric cause such as the cellular topology
 438 combined with the presence of manufacturing defects. The latter ones resulted from the LPBF
 439 process (Figures 3 and 5), and are known to exacerbate the mechanisms by which damage
 440 accumulates during plastic deformation [56, 57, 19, 18, 29].

441 4.2. LPBF-induced defects

442 The random cellular architectures of this work contained a non-negligible number of man-
 443 ufacturing defects, which occurred either as geometric imperfections or as matrix internal
 444 porosity (Figures 3,5). Their classification is given in Section 3.1 together with the descrip-
 445 tion of their morphological characteristics. Both types of defects are known to result from
 446 the laser scanning strategy and may have a detrimental effect on the mechanical proper-
 447 ties [56, 57, 19, 18, 29, 34]. Topological defects were not quantified in this study but were
 448 examined in Ref. [34], for similar porous structures fabricated using a large set of processing
 449 parameters and different metallic powders. Internal matrix porosity was instead analyzed
 450 using X-Ray tomography and was found to be of two types (i.e. pore- or crack-like, see
 451 Figure 5). Pores and cracks are commonly observed within the matrix of parts fabricated

452 by LPBF [29, 44], and factors governing their formation are well understood also for struc-
453 tures made of AlSi10Mg powders [44, 58]. Specifically, near-spherical pores result either from
454 trapped gas bubbles, during solidification of the base metal, or from the keyhole phenomenon
455 (which is in turn caused by the instability of the melting pool). Crack-like defects, also called
456 *lack-of-fusion* defects, are due either to a low energy density or to a poor overlap of the melt
457 pools. In the present study, the incomplete overlap between adjacent contour and hatch tracks
458 is likely the origin of the lack-of-fusion defects. This hypothesis is corroborated by several
459 observations. First, defects of this type (depicted with yellow color in Figure 5(c)) typically
460 appear within regions of the matrix where pores are largely spaced, which are constructed
461 with both contour and hatch tracks. Second, they occur at about the same location across
462 identical porous cells (Figures 5(a,b)), which were built using the same scanning strategy. Un-
463 like lack-of-fusion defects that are often through-thickness, rounded pores are much smaller
464 in size (Figure 6) and have a specific spatial distribution that corresponds to the laser con-
465 tour trajectory (Figure 5(c)). These observations are consistent with what was reported for
466 Ti6Al4V lattices produced by electron beam melting [57], and can be rationalized on account
467 of the viscous forces trapping gas bubbles within the laser contour trajectory.

468 The influence of manufacturing defects on the elastic properties was not investigated in
469 this study. It was examined in Ref. [34] for similar AlSi10Mg architectures, using image-based
470 FE simulations combined with experiments. This work rather elucidated the role of manu-
471 facturing defects on the damage process. Notably, it was found that strain first concentrated
472 around large matrix pores (see e.g. Figure 8(a)-Camera F for sample D20-w1), in agreement
473 with earlier studies on both periodic lattices [29, 19, 59, 57] and random foams [60] produced
474 by LPBF. By contrast, the pore geometric imperfections were found to exert no noticeable
475 influence on damage initiation. For example, similar deformation features were measured
476 across the two porous cells of sample D20-w1 (Figures 8(a)-(c)), despite the variations in
477 pore ellipticity and/or size between both cells (with identical nominal geometry). This find-
478 ing is consistent with what reported for other random porous media [33], and is of great
479 practical importance. Notably, it indicates that structural disorder effectively mitigates the
480 sensitivity to geometric imperfections. Reference [61] illustrates that point, highlighting the
481 advantage offered by *disordered metamaterials* over periodic lattices. In unit-cell based struc-
482 tures, the imperfection sensitivity is instead exacerbated and causes for the sudden formation

483 of shear/collapse bands that eventually lead to failure [62, 33, 63]. This deformation mecha-
484 nism is well documented for a large number of periodic lattices, including metallic structures
485 fabricated by LPBF [57, 13, 19, 59, 18, 56, 64].

486 4.3. Mechanisms of tensile failure

487 The materials produced for this study contained two populations of micrometric voids,
488 i.e., geometric and process-induced pores (Figure 5(d)). The first ones have larger cylindrical
489 shapes and were designed, with the RSA algorithm, to achieve a 20% porosity. The second
490 ones have smaller irregular geometry and do not exceed a few percent in volume (Section 3.3).
491 Both types of pores are embedded within the aluminum matrix and are likely to promote the
492 development of strained bands during tensile loading. It is well-known that in porous plas-
493 tic solids, strains concentrate between large primary voids, and that small secondary voids
494 (in the matrix material) accelerate the localization process [65]. This result was shown by
495 the pioneering analyses of Needleman [66] and Tvergaard [67] using an idealized material
496 model containing a square array of circular cylindrical voids, and agrees with the results
497 presented in this study. The latter ones reveal that regions of intense deformation occurred
498 between geometric pores and were heterogeneously distributed across the porous surface (Fig-
499 ures 7(a-d) and 8(a-d)). Only in the final stage of the fracture process, strains predominantly
500 concentrated around the failure zone (Figures 7(e) and 8(e)). This deformation behavior is
501 commonly observed in random porous media [54, 49, 33, 68] and is rationalized on the account
502 of the geometric disorder, which caused a diffuse straining activity and prevented the sudden
503 formation of a failure band.

504 To examine how the deformation accumulated in the present materials, the incremen-
505 tal strain maps for the two tested specimens are displayed in Figure 10. They measure
506 the changes of local longitudinal strains in discrete macro-strain intervals $\Delta\langle E_{yy} \rangle$, and were
507 computed for each cellular topology from the DIC-based displacement measurements corre-
508 sponding to Figures 7 and 8. Figure 10 shows that the mechanisms of strain accumulation
509 in both specimens were similar until departure from elasticity (Figure 10(a)), but differed re-
510 markably with increasing plastic straining (Figures 10(b-f)). Two main differences are noted
511 and are rationalized as follows. First, there is a tendency for more diffuse strain activity with
512 decreasing pore aspect ratio. This trend is particularly evident at the beginning of the plastic

513 stage (Figures 10(b-d)), and is explained by the pore morphology. Prior FE studies showed
514 that elliptical pores, with major axes aligned with the loading direction, lead to lower strain
515 concentrations than their circular counterparts, and in turn delay the damage process [69, 70].
516 The data in Figure 4 corroborate these findings, reporting a higher rate of strain hardening in
517 sample D20-w03 (with elliptical voids). The second difference is observed after the attainment
518 of the peak stress. With increasing pore aspect ratio, the strains concentrated in a single band
519 traversing groups of closely-spaced pores across the specimen's width (Figure 10(e,f)). This
520 observation is consistent with prior analyses for similar random porous geometries [71, 72, 73],
521 and is likely the consequence of the *cooperative* interaction between pores (both geometric and
522 process-induced). Prior works show that if voids are spatially close or favorably oriented (with
523 respect to the loading axis), plastic flow localizes into a band. This mechanism was first re-
524 ported by Ohno and Hutchinson [74] and by Needleman and Tveergard [75], who examined
525 respectively the effect of a non-uniform void distribution and of pore orientation on strain
526 localization. The subject now comprises a large body of work, which is relevant to ductile
527 fracture of metals [65].

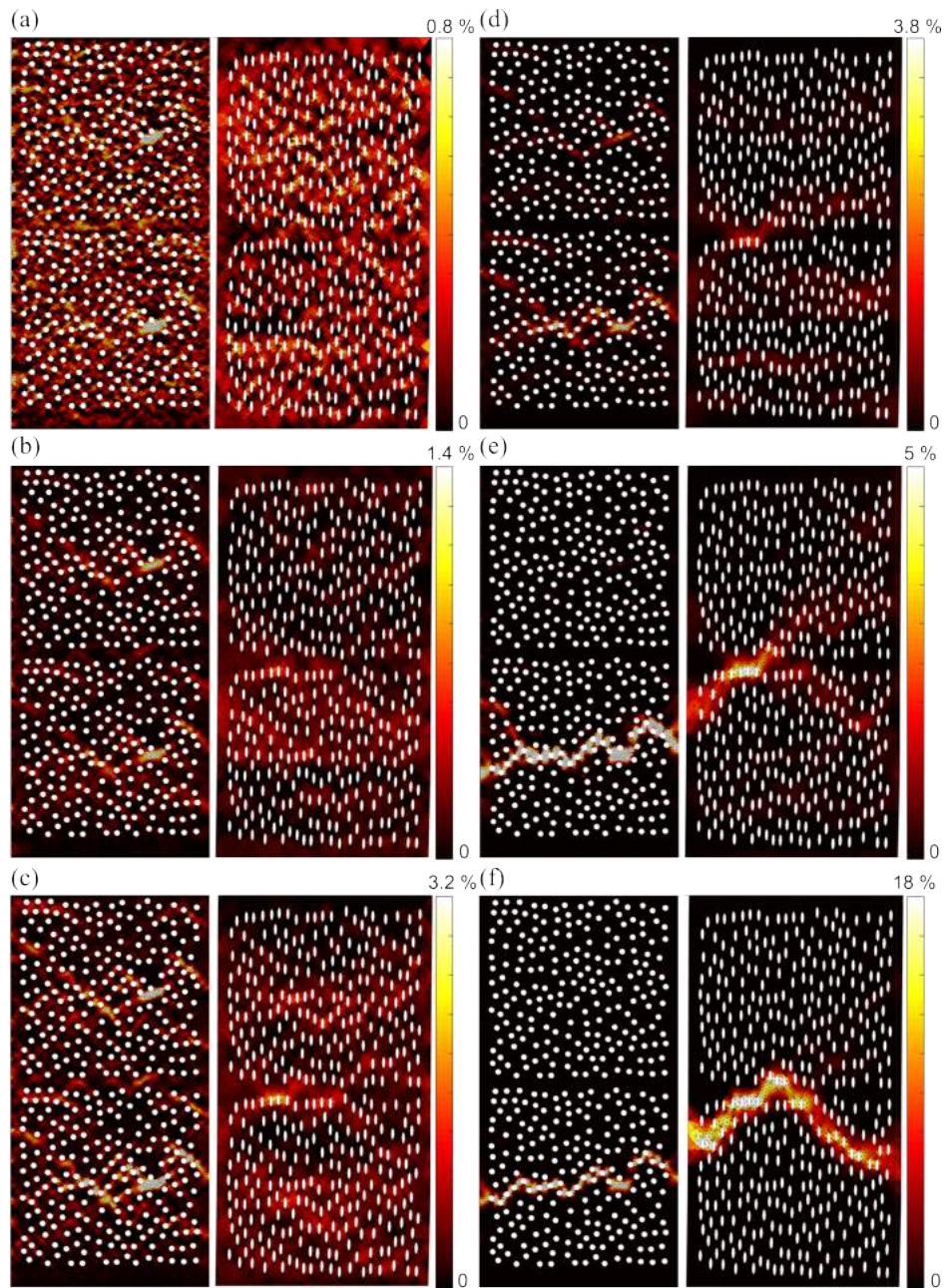


Figure 10: Incremental strain maps for specimens D20-w1 (left) and D20-w03 (right). They were computed from the displacement increments calculated via DIC from images collected by Cameras R and F, for specimen D20-w03 and D20-w1 respectively. Each reported map corresponds to a distinct macroscopic axial strain interval $\Delta\langle E_{yy} \rangle$, i.e., (a) 0% – 0.16%, (b) 0.16% – 0.27%, (c) 0.27% – 0.65%, (d) 0.65% – 0.77%, (e) 0.77% – 1%, (f) 1% – 1.2%. These strain levels are depicted with dotted lines in Figure 4 for each porous specimen.

529 cess, as damage will likely initiate in regions of highly concentrated strains. Two distinct
530 failure mechanisms are inferred from Figure 10 depending on the pore geometry. They are
531 summarized as follows, and show several common features with the mechanisms of ductile
532 fracture [65]. For sample D20-w03 with elliptical voids, failure was triggered by internal dam-
533 age accumulation and resulted from the progressive propagation of a microcrack forming upon
534 the first void coalescence. A distinct failure band only appeared at the very end of the fracture
535 process (Figure 10(f) for sample D20-w03). Void coalescence occurred by internal necking of
536 the matrix ligament between the pores, as suggested by the presence of dimples across the
537 fractured surface of sample D20-w03 (Figure 9(c)). The presence of scattered strained sites
538 (Figures 10(b-e)) also indicates that damage was diffuse over the porous surface, and likely
539 competed with strain hardening. Figure 4(a-Camera R) corroborates these findings, showing a
540 region of stress fluctuations around the peak load. Similar results were reported by Mangipudi
541 and Onck for random foams under tension [54]. In the case of sample D20-w1 with circular
542 voids, failure was instead triggered by plastic instability. The latter was distinguished by the
543 formation of a localized strain band (Figures 10(d-e) for sample D20-w1), and was accompa-
544 nied by elastic unloading in the outer material regions [76]. The DIC measurements provided
545 this evidence, but are not shown here for the sake of brevity. Fracture was likely produced by
546 void coalescence inside the band. Post-mortem tomographic images (Figure 5(a)) support this
547 observation showing a *jagged* crack path linking groups of pores oriented at $\approx 45^\circ$ with respect
548 to the macroscopic principal stress axis. Coalescence was induced by a shearing process. This
549 type of void coalescence mechanism, known as *void sheeting*, has been observed experimen-
550 tally first by Cox and Low [77] and by several other authors since then [78, 79, 80, 81]. It is
551 known to be favored by the presence of secondary voids and is accompanied with large pore
552 shape changes [65]. Tomographic observations (Figure 5) corroborate these findings, showing
553 evidence of large pore distortions upon coalescence (Figure 5(a)) as well as the presence of
554 smaller process-induced voids (Figure 5(c,d)).

555 4.4. Comparison with McClintock's micromechanical model

556 Plastic deformation is known to affect the rate at which pores grow in tension [65]. In this
557 study, the measured strain fields were highly heterogeneous (Figures 7 and 8), and exhibited
558 distinct accumulation mechanisms with varying pore geometry (Figure 10). It is therefore

559 reasonable to expect that the rate of void growth varied within the same specimen, and also
 560 between the two cellular topologies. Figure 11, which confirms this statement, reports for
 561 all geometric voids within each specimen, the pore true strain at discrete instants of the
 562 deformation history. This quantity is defined as the variation of the pore area over time (i.e.,
 563 $\ln(A/A_0)$, where A and A_0 denote respectively the current and initial pore areas), and was
 564 computed, for each pore, from its major and minor axes. The actual pore geometry was in turn
 565 quantified after interpolation (via a conic function) of the nodal pore coordinates, measured via
 566 FE-based DIC using meshes consistent with the porous architecture (Section 2.4). Figure 11
 567 highlights two major trends. First, it shows that void deformation was highly heterogeneous
 568 in both specimens and was higher for pores within areas of intense strain (Figures 11(a-d)).
 569 This last observation could be made by comparing pairwise Figures 11(a-d) with Figures 10(a-
 570 d) for specimen D20-w1 with circular voids, and was in turn expected as pores grow aided
 571 by plasticity [66, 82, 65]. Second, Figure 11 reveals that, upon macroscopic instability, void
 572 deformation was entirely localized inside the shear band (see Figures 11(e-f) for sample D20-
 573 w1 with circular pores). Pores in this band were no longer, even approximately, circular
 574 (Figures 11(e-f)) as a result of *void sheeting* [78, 79, 80, 81].

575 To quantify the rate of void growth during tensile straining, the change of the pore
 576 mean radius, $\ln(R_m/R_{m0})$, with the applied macroscopic deformation, $\langle E_{yy} \rangle$, is shown in
 577 Figures 12(a,b) for all geometric pores in each tensile sample. This geometric parameter was
 578 chosen for convenience as its initial value (i.e., R_{m0}) was similar for both cellular topologies
 579 (Table 1). For the sake of comparison, measurements corresponding to different geometric
 580 pores across each specimen are reported using different line colors. Collectively, the data in
 581 Figures 12(a,b) reflect the trends observed in Figure 11, indicating highly heterogeneous void
 582 growth across each specimen. Moreover, they show that the rate of void growth increased
 583 significantly upon the attainment of the maximum stress, for the pores in the fracture path.
 584 The latter ones are indicated in the inset figure for each cellular topology, and were found to
 585 grow significantly faster than all other pores in the specimen (Figures 12(a,b)). In particular,
 586 a higher void growth rate was measured in specimen D20-w1 with circular pores. Similar
 587 results were reported by Dubensky and Koss [71] and were rationalized on account of the
 588 cooperative interaction of the pores within the shear band, which eventually coalesce by a
 589 *void-sheeting* mechanism.

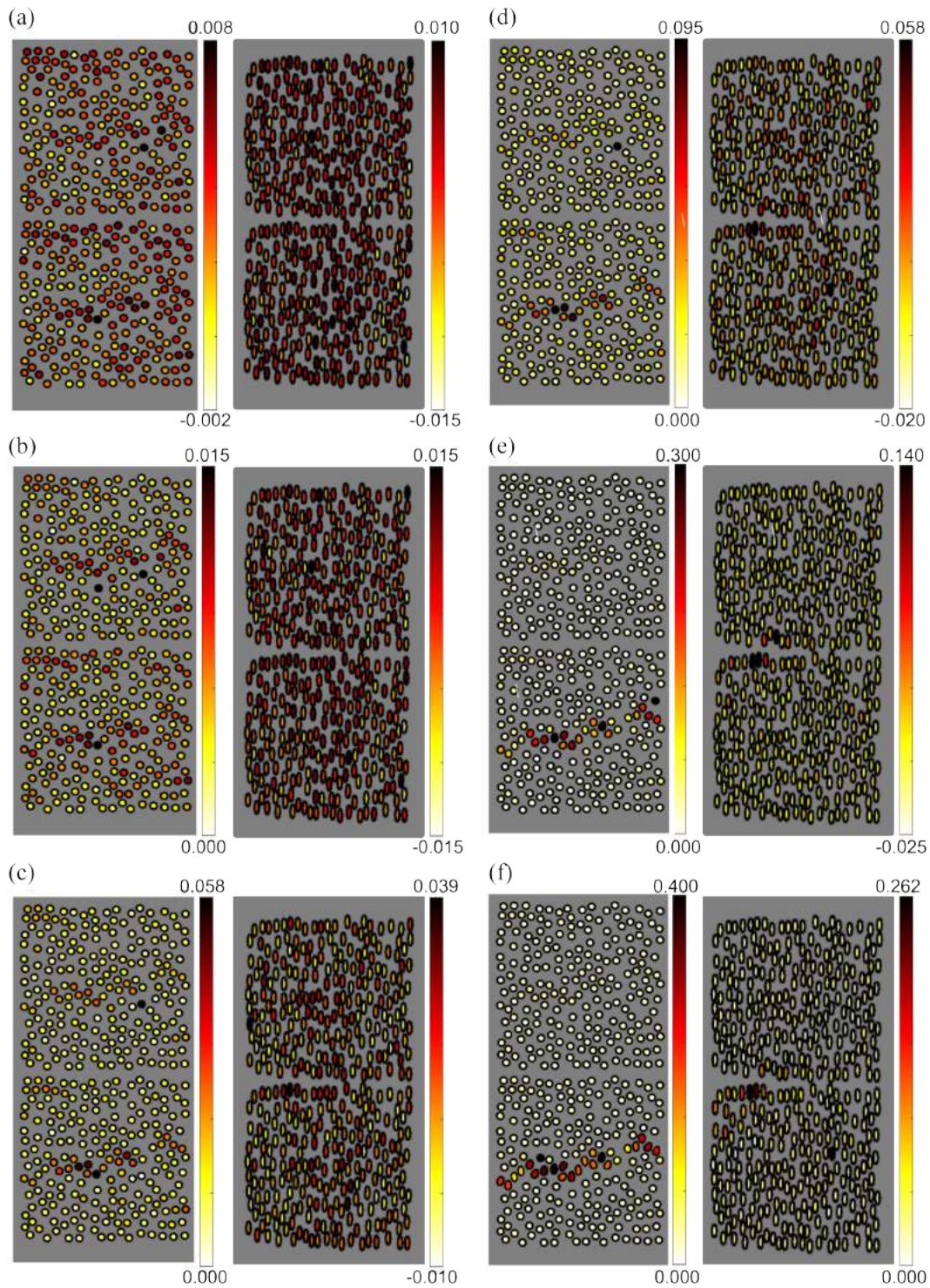


Figure 11: True pore strain maps (as determined from the pore area, i.e. $\ln(A/A_0)$) for specimens (left) D20-w1 and (right) D20-w03 at discrete points of the deformation history. Data for specimens D20-w1 and D20-w03 were computed from Camera F and Camera R, respectively. The macroscopic longitudinal strain $\langle E_{yy} \rangle$ was equal to: (a) 0.16%, (b) 0.27%, (c) 0.65%, (d) 0.77%, (e) 1% and (f) 1.2%. These levels are depicted with dotted lines in Figure 4 for each porous specimen. 31

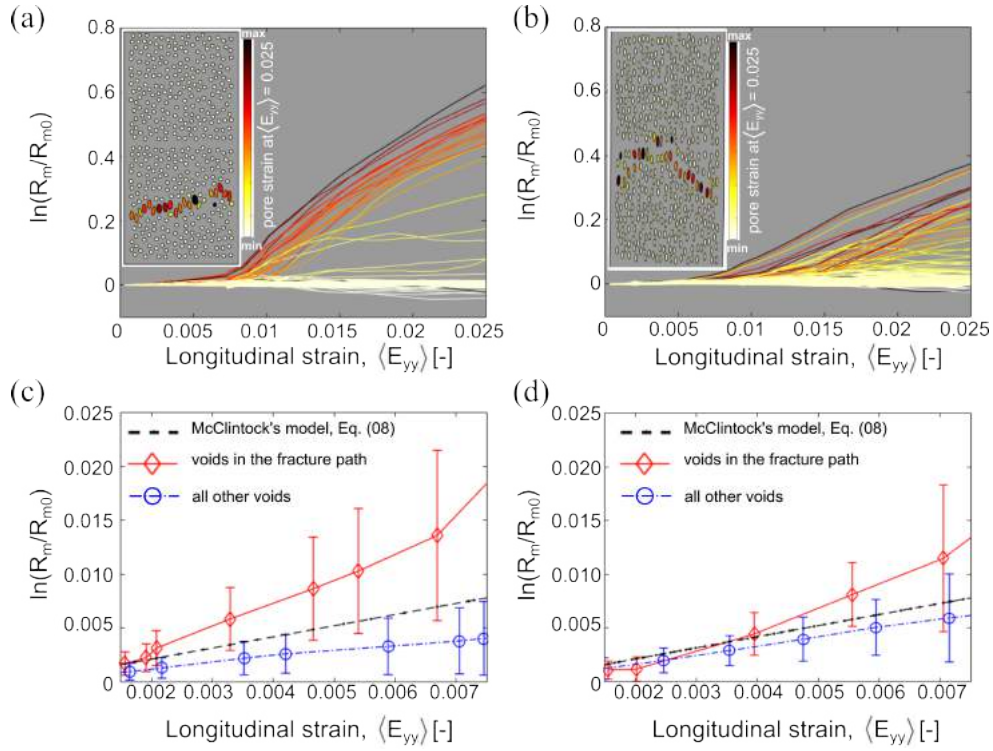


Figure 12: (a)-(b) Growth of the mean pore radius $\ln(R_m/R_{m0})$ as a function of the macroscopic strain $\langle E_{yy} \rangle$ in specimens (a) D20-w1 and (b) D20-w03. Data for specimens D20-w1 and D20-w03 were computed from Camera F and Camera R, respectively. Different line colors indicate different pores across each specimen. The latter ones are shown in their deformed state in the inset of each figure at $\langle E_{yy} \rangle = 2.5\%$. (c)-(d) Comparison between the McClintock's model predictions (Equation 8) and the experimental data for specimens (c) D20-w1 and (d) D20-w03. For each cellular topology, the average value of the mean pore radius growth is reported for two sets of pores (i.e., for the pores inside and outside the fracture band). Error bars represent the measurements standard deviation.

590 The effect of void interaction is not taken into account by the well-known McClintock
 591 void growth model [30]. The latter was formulated by solving the generalized plane-strain
 592 deformation problem of an isolated cylindrical hole in an infinite plastic medium. McClintock
 593 postulates that the pore mean radius (i.e., $\ln(R_m/R_{m0})$) increases with increasing equivalent
 594 applied strain (i.e., $\bar{\epsilon}$) and is given by (see Equation (23) in Ref. [30])

$$\ln \left(\frac{R_m}{R_{m0}} \right) = \frac{\sqrt{3}\bar{\epsilon}}{2(1-n)} \sinh \left(\frac{\sqrt{3}(1-n)}{2} \frac{\sigma_a + \sigma_b}{\bar{\sigma}} \right) + \frac{\epsilon_a + \epsilon_b}{2}, \quad (7)$$

595 where n and $\bar{\sigma}$ denote the strain-hardening coefficient and the equivalent macroscopic stress,
 596 respectively. The applied stresses σ_a and σ_b are those in the major and minor pore axes

597 (principal) directions, and ϵ_a and ϵ_b the corresponding strains. For an elliptical hole deformed
 598 in tension under plane strain conditions (i.e., $\epsilon_a = -\epsilon_b$ and the out-of-plane normal stress is
 599 $\sigma_z = (\sigma_a + \sigma_b)/2$ [30]), Equation (7) becomes

$$\ln \left(\frac{R_m}{R_{m0}} \right) = \frac{\langle E_{yy} \rangle}{(1-n)} \sinh(1-n). \quad (8)$$

600 The comparison between the predictions of McClintock’s model (Equation (8) with $n =$
 601 0.5, where the latter is given in Ref. [49] for an Al-matrix) and the experimental data is
 602 displayed in Figures 12(c,d) for applied macroscopic deformations up to $\langle E_{yy} \rangle = 0.75\%$ (i.e.
 603 c.a. up to $\approx \epsilon_{UTS}$, see Table 3.2). To make the comparison easier, the average value of the
 604 mean radius growth $\ln(R_m/R_{m0})$ is reported for each cellular topology. This quantity was
 605 computed from the measurements reported in Figures 12(a,b) for two sets of pores (i.e., for
 606 the voids inside and outside the fracture path, see insets of Figures 12(a,b)). For each set
 607 of experimental data, the standard deviations at discrete values of the applied axial strain
 608 are also reported. Figure 12(c,d) first shows that the growth of the mean pore radius was
 609 proportional to the applied macroscopic strain, and was lower for the holes outside the fracture
 610 band (compare the blue and red lines across Figure 12(c,d)). These findings are consistent
 611 with the results of prior studies [83, 84, 71], and agree well with McClintock’s analysis [30]. For
 612 this set of pores, the comparison between the predictions of Equation (8) and the experimental
 613 data is particularly satisfactory for specimen D20-w03 with elliptical holes (compare black and
 614 blue lines in Figure 12(d)), thereby confirming that void interactions were small outside the
 615 failure band. Conversely, McClintock’s model underestimated the measured data for the void
 616 growth inside the fracture path (see black and red lines in Figure 12(c,d)). This discrepancy
 617 is due to the model assumptions (neglecting void interaction [77]) and is more pronounced
 618 for specimen D20-w1 with circular voids. For the latter, void interactions occurred since the
 619 very early stages of deformation (Figure 8) and concentrated predominantly within the shear
 620 band (Figure 11).

621 The presence of a large irregular defect within the failure band of specimen D20-w1 (high-
 622 lighted with a red line in Figure 5(a)) likely exacerbated the effects of void interaction, pro-
 623 moting early plastic flow localization. This observation is consistent with prior studies on
 624 the effect of void cluster [74, 82, 85], and is likely the cause of the higher discrepancy with
 625 McClintock’s model predictions (Figure 12(c)). Tomographic reconstructions in Figure 5(a)

626 also corroborate this statement, providing experimental evidence of the interaction between
627 geometric voids and matrix defects. Notably, they showed that the crack-tip of the defect,
628 highlighted with a red line in Figure 5(a)-bottom cell, coalesced with its closest geometric
629 pores along the direction of the tensile axis.

630 Closing this section, it is also worth emphasizing some of the main limitations of microme-
631 chanical models based on periodic arrays of voids. Since the seminal works of Needleman [66]
632 and Tveergard [67], unit cell models have been successful in rationalizing salient features
633 of ductile fracture processes [86]). While models of this type will likely remain very use-
634 ful tools for investigation purposes, they fail to capture most of the findings produced in this
635 study. The latter ones notably included the effects of non-uniform pore distributions on strain
636 heterogeneities and localization (Figures 7, 8 and 10). This point was raised by several au-
637 thors [87, 76] and, in turn, opens up novel perspectives for this class of random metamaterials,
638 which may prove attractive *porous systems* for the study of ductile fracture at the mesoscale.

639 5. Conclusion

640 This work investigated experimentally the tensile response of metallic cellular metamate-
641 rials that contained randomly-dispersed cylindrical pores at $\approx 20\%$ in vol. fraction. Their
642 porous architecture was generated numerically via a random sequential absorption algorithm,
643 and was fabricated by LPBF from AlSi10Mg powders.

644 Using FE-based DIC (where the pore meshes were consistent with the cellular topology),
645 the mesoscale deformation mechanisms were quantified for this class of disordered porous
646 materials. Notably, it was shown that they are distinguished by non-uniform strain distribu-
647 tions and are triggered by the interaction between geometric pores. The latter was, in turn,
648 promoted by the presence of smaller pores and lack-of-fusion defects within the AlSi10Mg ma-
649 trix, as revealed by X-Ray tomography. Defects of this type resulted from the manufacturing
650 process and are typically observed in parts produced by LPBF.

651 Equally important, the results of this study showed that the mechanisms of strain con-
652 centration varied with the pore geometry and exhibited several common features with the
653 ductile failure mechanisms of metals, albeit at the mesoscale. For example, with increasing
654 pore aspect ratio, fracture occurred by *void-sheet* coalescence and resulted from macroscopic

655 plastic instability. Conversely, more elongated elliptical voids caused diffuse strained areas
656 and delayed the fracture process. Moreover, strain localization affected the void growth rate.
657 It was shown that the pores in the failure band grew much faster than all other pores and
658 that McClintock’s model could not predict the experimental data for the average void growth
659 rate in the fracture band.

660 The results of this study have important implications for micromechanical models used to
661 study void growth and coalescence, and in turn open up novel perspectives for this class of
662 random porous metamaterials as convenient model systems for the study of ductile fracture
663 of metallic alloys.

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674 **7. Appendix**

675 The backtracking procedure was instrumental to ensure that the DIC meshes were consis-
676 tent with the actual pore distribution. In this section, the backtracking technique, allowing
677 the pores to be correctly positioned on the adapted mesh for the two porous specimens is de-
678 tailed. The technique uses a correlation algorithm that instead of correlating two consecutive
679 images, as in classical DIC, registers a real and an ideal image [40]. A coarse auxiliary mesh
680 was used to measure the displacement field enabling for the best possible match. To evaluate
681 the quality of the converged solution, the gray level residuals (Section 2.4.1) were analyzed.
682 Grayish zones correspond to good correlation, whereas black and white zones highlight areas

683 with manufacturing defects. For instance, the white zones are associated with lack-of-fusion
684 defects that are present in the real volume section but not in the nominal image.

685 *From mask to mesh*

686 This first approach was applied to images acquired by Camera F. In that case, the ideal
687 image consisted in an indexed mask numerically reconstructed starting from the nominal
688 parameters of the porous matrix (i.e., pore positions, orientation and dimensions). A fine
689 triangular mesh was then created on the material indexed zone of the image via the open-
690 source software Gmsh [88]. In Figure 13(a), an example is shown for the mesh of specimen
691 D20-w03 for Camera F. The numerical mask was correlated to the real image (first frame
692 captured during testing) through a correlation algorithm thanks to a coarse auxiliary mesh
693 (Figure 13(b)). The coarse mesh assesses the displacement field the ideal mesh should undergo
694 to adapt to the real image. Thanks to an interpolation between the coarse and refined meshes,
695 it is possible to move back each node of the model-based mesh. In Figure 13(b), the coarse
696 mesh is depicted in orange. The correlation residuals after backtracking prove that the pores
697 were properly repositioned (Figure 13(c)).

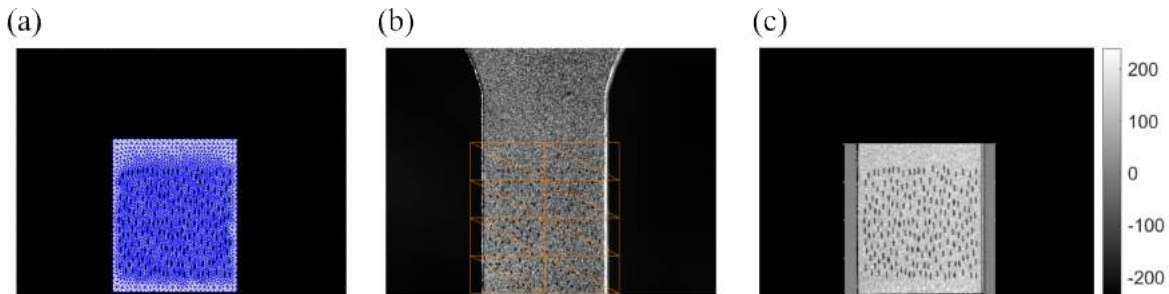


Figure 13: (a) Gmsh generated mesh to be backtracked and superimposed onto the nominal mask. (b) Reference image of Camera F for specimen D20-w03 with the coarse backtracking mesh depicted in orange. (c) Gray level residual map to backtrack the mesh of specimen D20-w03 and Camera F showing that the backtracking procedure was successful.

698 *From mesh to mask*

699 This technique was applied to the images acquired by Camera R. Its lower resolution
700 did not allow the numerically generated pores to be correlated to those present in the real
701 image. The correlation was made possible by the curved edges, still visible in Camera R

702 images, thanks to the lower zoom factor. In the absence of a numerical file containing all
703 positions of pores and edge lines, the numerical mask was built starting from the mesh of
704 the dogbone sample. A 2D .stp file of the full dogbone specimen (not only the porous part)
705 was available. A triangular fine mesh, displayed in Figure 14(a), was directly obtained on
706 the solid part by using the Abaqus interface [89]. The meshed specimen was then manually
707 prepositioned on the reference image and then backtracked. In that case, the indexed mask
708 was built starting from the Abaqus mesh, then imported in the DIC code, and the node to
709 pixel projector was used to determine which pixels belonged to the mesh. Once the mask in
710 Figure 14(b) was obtained, it was correlated with the reference image thanks to the coarse
711 auxiliary mesh represented in Figure 14(c). The displacement field obtained on the coarse
712 mesh was interpolated on the deformed mesh, thus obtaining the backtracked mesh shown in
713 Figure 15(b).

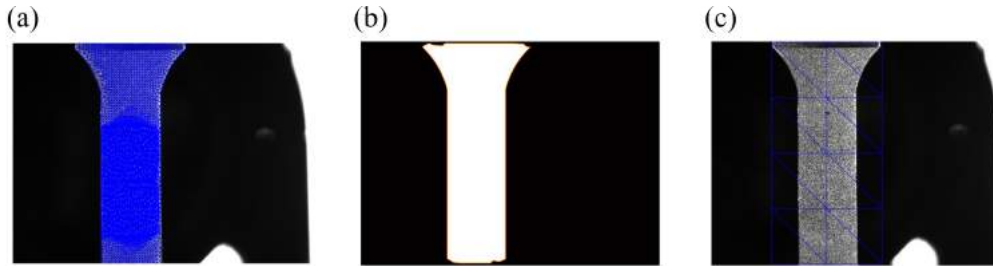


Figure 14: (a) Prepositioned mesh generated on the dogbone CAD model laid over the reference image.
(b) Reconstructed mask, and (c) auxiliary (blue) mesh in the reference image

714 The backtracked meshes for the two porous specimens and both cameras are shown in
715 Figure 15. They were utilized to report DIC analyses at the mesoscale level.

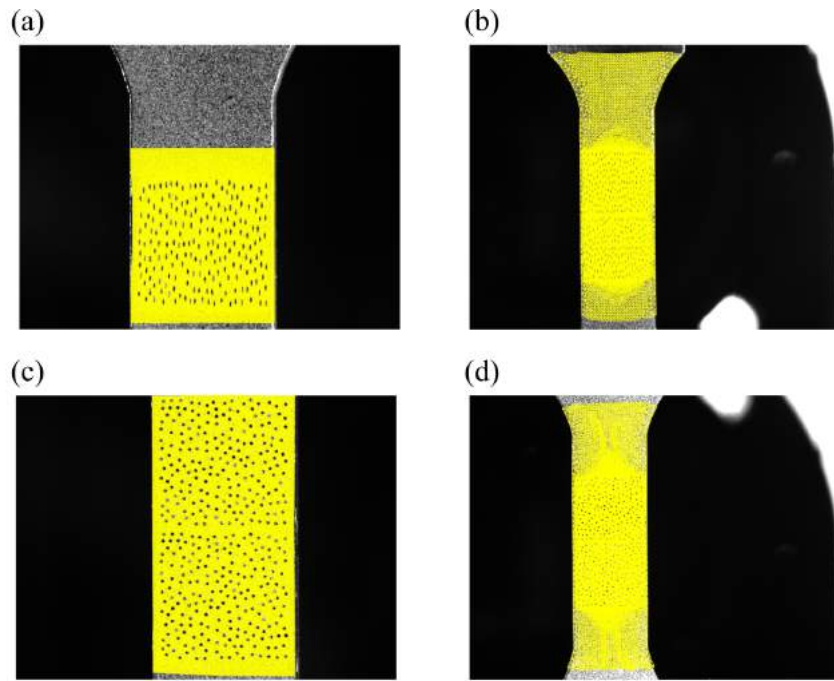


Figure 15: Backtracked meshes of sample D20-w03 for Cameras (a) F and (b) R acquisitions, and specimen D20-w1 for Cameras (c) F and (d) R acquisitions, respectively

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