

Microstructure and loading direction dependent hardening and damage behavior of laser powder bed fusion AlSi10Mg

Lubin Song^a, Lv Zhao^{a,d,*}, Lipeng Ding^b, Yaxin Zhu^{a,d}, Minsheng Huang^{a,d}, Aude Simar^c, Zhenhuan Li^{a,d}

^a Department of Mechanics, School of Aerospace Engineering, Huazhong University of Science and Technology, Wuhan, China

^b Key Laboratory for Light-weight Materials, Nanjing Tech University, Nanjing, China

^c Institute of Mechanics, Materials and Civil Engineering, UCLouvain, Louvain-la-Neuve, Belgium

^d Hubei Key Laboratory of Engineering Structure Analysis and Safety Assessment, 1037 Luoyu Road, Wuhan, China

ARTICLE INFO

Keywords:

Laser powder bed fusion
AlSi10Mg
Strain hardening
Damage
Strain localization

ABSTRACT

The correlation between mechanical behavior and heterogeneous microstructure is still unclear for laser powder bed fusion AlSi10Mg. The present work investigates strain hardening and damage mechanisms with different build platform temperatures (35 °C and 200 °C) and loading directions (horizontal and vertical, building direction being vertical) to probe the effects of microstructure size and melt pool border orientation. The 35 °C microstructure involves load direction independent properties, with ductility determined by damage initiation occurring extremely close to final failure. In contrast, the 200 °C microstructure exhibits anisotropic mechanical behavior, as revealed by higher strain hardening capacity, earlier damage initiation and more dispersed damage distribution for the vertical loading direction. Although the vertical samples generally involve strain localization and crack propagation along the melt pool border, their ductility is not compromised compared to the horizontal counterparts. We attempt to explain these hardening and damage behaviors with stress partition and trade-off between phase stress and strain gradient.

1. Introduction

Metal additive manufacturing (AM) is an advanced technology that can realize complex three-dimensional (3D) components using layer-by-layer building strategy [1]. Laser powder bed fusion (LPBF), being a popular metal AM method, presents very promising application prospects in aerospace, automotive, biomedical and other fields [2–5].

Aluminum alloys have been widely used in engineering structures, owing to their high strength-to-weight ratio, low price and good resistance to corrosion [6,7]. Given these advantages, the development of AM Al alloys initiated almost at the same time as the emergence of LPBF process, as this process-material couple could promote a leap forward in lightweight structure design. Among LPBF Al alloys, AlSi10Mg stands out for its good printability and the possibility of forming Mg₂Si precipitates in post heat treatments [8–11]. Although LPBF AlSi10Mg is relatively mature, being subjected to many studies in the AM community, the correlation between mechanical behavior and microstructure is still unclear. Further understanding on how microstructure controls mechanical properties is not only beneficial to optimize AlSi10Mg, but

also heuristic to develop other AM Al alloys.

LPBF AlSi10Mg presents a multi-scale, hierarchical and heterogeneous structure, from nanoscale supersaturated solute atom clusters [12] to microscale Si-rich eutectic networks [13], and to mesoscale melt pool (MP) structure [14]. These structure characteristics are closely related to inherent features of LPBF process, i.e., high cooling rate, strong thermal gradient and layer-wise scanning strategy. Indeed, during the solidification of laser melted metal, the cooling rates were estimated to be in the order of 10³–10⁶ K/s [15], leading to supersaturated solid solution and fine Si-rich eutectic phase. Lefebvre et al. [12] and Van Cauwenbergh et al. [16] revealed a Si solid solution of roughly 3 at.% in α -Al, which is much higher than the equilibrium Si solubility (1.6 at.% [17]). Thijs et al. [18] comprehensively characterized the microstructure of LPBF AlSi10Mg, including the morphology of Si-rich eutectic phase, α -Al grains and melt pools. Due to specific thermal gradient and cooling rate, the Si eutectic phase was found to exhibit cellular shape enclosing the α -Al [19,20], the latter presenting columnar grain features and slight <001> fiber texture [13,18,21]. The Si-rich network generally presents a size in the range of 300–500 nm [13,22],

* Corresponding author. Department of Mechanics, School of Aerospace Engineering, Huazhong University of Science and Technology, Wuhan, China.
E-mail address: lvzhao@hust.edu.cn (L. Zhao).

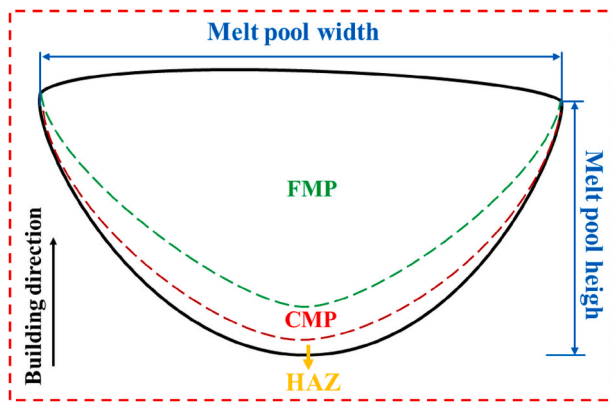


Fig. 1. Schematic drawing of spatial arrangement of the three melt pool zones (FMP, CMP and HAZ).

while the mean equivalent diameter of α -Al grains is normally of the order of 10 μm [18,23].

The layer-wise LPBF process not only forms mesoscale melt pool structures, but also modifies the Si-rich phase. When scanning the current powder layer, previously solidified material will be partially remelted, leading to three distinct zones at the vicinity of the melt pool border [18,20]: (1) fine structure presenting thin Si-rich eutectic network, also referred to as fine melt pool (FMP) zone, which occupies the majority of the whole melt pool; (2) coarse structure, also referred to as coarse melt pool (CMP) zone, in which the Si-rich eutectic phase still retains interconnected but becomes larger and thicker than that in the FMP; (3) heat affected zone (HAZ), where the interconnection of the Si network is significantly degraded. The spatial arrangement of the three zones on melt pool cross-section is schematically illustrated in Fig. 1. The CMP and HAZ are generally regarded as the melt pool border. Recently, Santos Macías et al. [13] characterized the 3D morphology of the Si-rich phase using focused ion beam/scanning electron microscopy (FIB/SEM) tomography, they revealed that the Si phase connectivity was very high in the FMP (>0.9) and CMP (>0.8), but dropped to a very low level in the HAZ (0.1–0.2).

The fine microstructure of LPBF AlSi10Mg imparts a higher mechanical strength compared to cast counterparts [13,24,25]. Wu et al. [26] attributed the high strength to the Si-rich networks impeding dislocation motion. Chen et al. [22] suggested that Orowan looping around discrete Si phase was the dominant strengthening mechanism. Takata et al. [27] pointed out that the presence of nanosized particles and solute atoms should promote a significant contribution to strength. Recently, Zhang et al. [28] demonstrated an important load bearing effect of the Si-rich phase during plastic deformation. Li et al. [29] revealed the presence of geometrically necessary dislocations (GND) at Al–Si interface and proposed these GNDs contribute to strain hardening.

LPBF AlSi10Mg generally presents anisotropic properties, its yield strength and elongation to fracture along the building direction (i.e., vertical samples) are mostly lower than those in the perpendicular direction (i.e., horizontal samples) [3,19,30], while strain hardening capacity being higher [19,30,31]. The underlying mechanism for the load direction dependent hardening has attracted little attention from the additive manufacturing community. The anisotropy in ductility is widely believed to be induced by the heterogeneous feature of the microstructure, as the melt pool border (i.e., CMP and HAZ) presents lower strength than the melt pool interior (i.e., FMP) according to nano-indentation tests [32]. In theory, vertical samples would experience pronounced strain localization at the CMP or HAZ, leading to early failure of material. This viewpoint is supported by the presence of melt pool traces on fracture surface of vertical samples [3,33,34]. Nevertheless, Zhao et al. [31] and Yan et al. [24] recently showed that sound LPBF AlSi10Mg presented very similar ductility in different loading

Table 1

Chemical composition (wt.%) determined by ICP OES.

Sample	Al	Si	Mg	Fe
LPBF 35 °C	Bal.	9.68	0.43	0.13
LPBF 200 °C	Bal.	9.60	0.43	0.14

directions. As a matter of fact, failure of metallic materials is generally controlled by damage initiation, growth and coalescence [35]. Different microstructures exhibit different strain levels for damage initiation [36], and damage evolution behaviors may also significantly vary as a function of damage size, spacing and location [37].

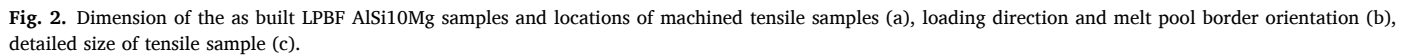
Few efforts have been devoted to systematic investigation of damage behavior of LPBF AlSi10Mg, compared to strengthening mechanisms. Zhao et al. [38] demonstrated that damage initiation was induced by breakage of Si-rich eutectic network. They further revealed a preference of damage initiation and crack path along CMP band for vertical samples [31]. Similar observations were reported by Maconachie et al. [39] where damage was found to be confined at CMP zone. Based on similar ductility between two distinct microstructure sizes, Santos Macías et al. [13] concluded that coarser Si-rich eutectic phase was more prone to damage, driving early damage initiation. However, the larger damage spacing also delayed coalescence events, leading to similar fracture strain as the finer microstructure. It should be noted that analyses and discussion of damage mechanism were mainly based on broken samples in the aforementioned studies. Although basic damage feature has been disclosed, some other important issues still remain to be tackled, for instance deformation level at damage initiation, correlation between damage distribution and local strength, and link between strain localization and macroscopic ductility.

This work investigates the strain hardening and damage behavior of LPBF AlSi10Mg with different build platform temperatures (35 °C and 200 °C) and loading directions (horizontal and vertical samples), with a focus on damage initiation, strain localization at melt pool border and their correlation to local property evolution. The 35 °C and 200 °C were chosen to obtain typical fine and coarse microstructures, respectively. This allows to probe whether microstructure size alters the dependence of mechanical behavior on loading direction. Note that 35 °C build platform grants higher mechanical strength, while 200 °C build platform mitigates residual stresses [40], both presenting practical interests. Tensile tests were carried out, with and without interruption/unloading to study damage before and upon macroscopic failure. Nano-indentation and electron backscatter diffraction (EBSD) were performed to probe strength change and deformation incompatibility at different MP zones prior to fracture. Based on these measurements and observations, hardening behavior, damage mechanism and their correlation to microstructure are thoroughly discussed.

2. Material and experiments

2.1. LPBF processing

The investigated LPBF AlSi10Mg samples were fabricated by an EOS M290 machine, equipped with a $252 \times 252 \times 25$ mm EN AW 5083 aluminum alloy build platform. Two sets of plates of $5 \times 35 \times 150$ mm were manufactured with the machine manufacturer suggested parameters (laser power = 390 W, layer thickness = 30 μm , hatch spacing = 0.19 mm, scanning speed = 1300 mm/s, scan rotation = 67° [13]) but with different build platform temperatures, i.e., 35 °C and 200 °C. The printed plates remained in as built state for the microstructure characterization and mechanical testing in the present work. The chemical compositions were measured by inductively coupled plasma optical emission spectroscopy (ICP OES) and are given in Table 1.



The microstructure of the built samples was characterized with scanning electron microscopy (SEM) and electron backscatter diffraction (EBSD). Metallurgical polishing and chemical etching (using 0.5% HF solution) were applied to obtain appropriate phase contrast. The SEM observations were carried out with the JSM-IT500A equipment. For EBSD analysis, no etching was performed, the sample polishing was finished with oxide polishing suspension (OPS) for 2 h to achieve good surface quality. EBSD characterization was carried out by a field emission SEM (TESCAN S800) equipped with a TSL OIM EBSD system named EDAX S800 GMH. A step size of 0.3 μm was employed to ensure a high resolution. The post-processing of EBSD data was conducted with the OIM software.

Tensile tests were performed in two directions to study the influence of melt pool border orientation on the mechanical behavior, see Fig. 2a. The samples with the loading direction perpendicular and parallel to the building direction were named as horizontal (H) and vertical (V) samples, respectively (see Fig. 2b). To easily recognize the combination of build platform temperature and loading direction, the samples were further designated as 35H, 35V, 200H and 200V, i.e. given first the built platform temperature (35 or 200 °C) and then the sample orientation. Fig. 2c presents the detailed sample size. It should be noted that a R10 rounding of the reduced section was carefully designed by finite element simulation to ensure a quasi-uniform stress state while constraining damage and fracture at the minimum section. This design would facilitate damage characterization before the onset of fracture.

end speed of 0.05 mm/min. To investigate damage onset, only a part of tensile tests were loaded until final failure, while the rest were interrupted before fracture. Five tests (3 broken and 2 interrupted) were performed for each sample type to ensure reproducibility. The tensile deformation of the interrupted tests generally exceeded 90% of the corresponding elongation to fracture. Fracture surface morphology was observed with SEM. All broken and interrupted samples were polished to mid-thickness plane to reveal their damage distribution. Moreover, EBSD analysis was performed on interrupted samples to probe the deformation induced microstructure change.

The 35 °C and 200 °C materials exhibit similar microstructural features, as shown in Fig. 2. The Si-rich eutectic phase manifests as

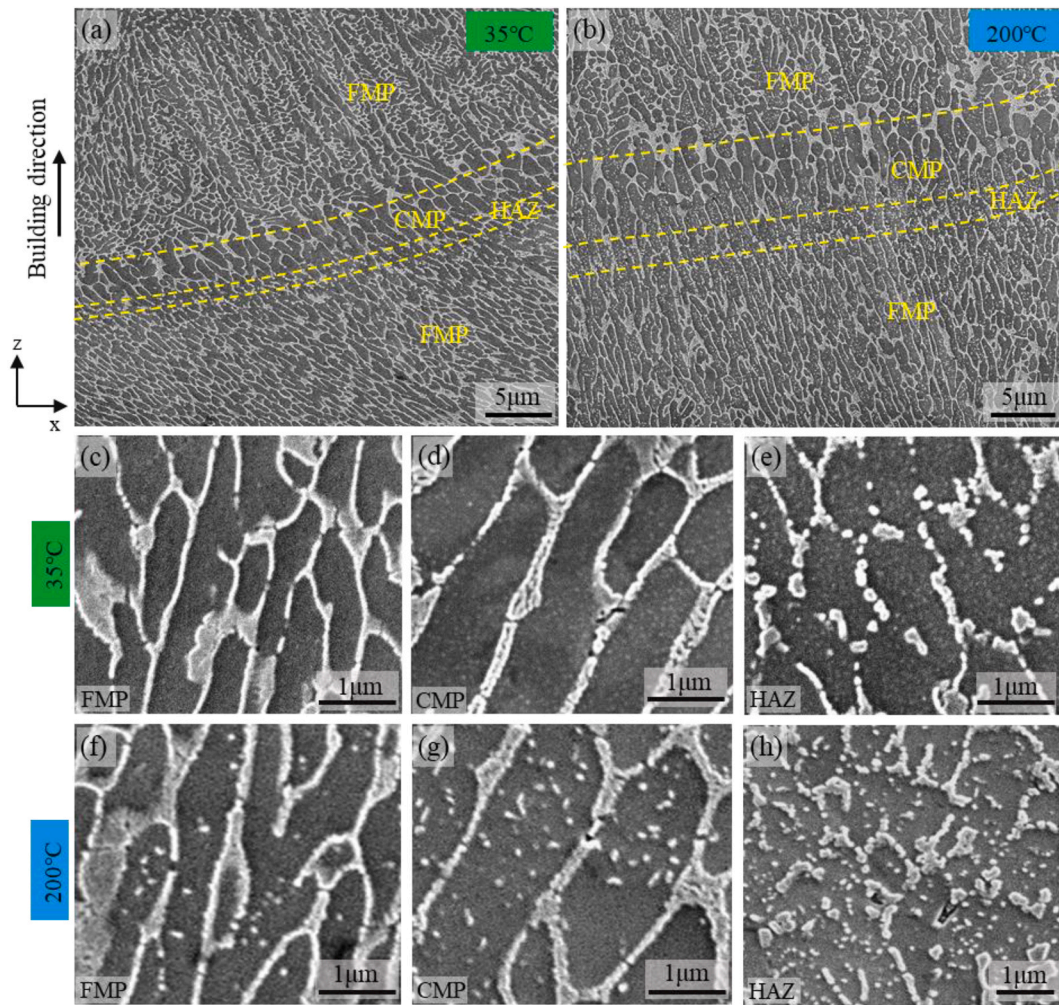


Fig. 3. Microstructure of as built LPBF AlSi10Mg. Typical Si-rich phase heterogeneity at the melt pool border for the 35 °C (a) and 200 °C (b) materials, morphology of FMP, CMP and HAZ for the 35 °C (c–e) and 200 °C (f–h) materials.

elongated (along the building direction) and interconnected network at the FMP and CMP (Fig. 3c, d and f, g). The connectivity significantly drops at the HAZ (Fig. 3e, h). Due to the relatively high build platform temperature reducing cooling rate and thermal gradient, the 200 °C microstructure is coarser at the level of three different microstructural features. First, both the CMP and HAZ regions are larger than those of the 35 °C material, as can be clearly seen in Fig. 3a and b. Second, the Si-rich eutectic network is coarser in terms of equivalent diameter and wall thickness for the 200 °C material (Fig. 3c, d and f, g). Third, the Si-rich

particles inside the Al cells are larger than those of the 35 °C material (see Fig. 3e and h). According to a previous study on the same materials manufactured with the same machine and parameters, the HAZ measures 1.5 μm vs. 3 μm , the CMP measures 5.5 μm vs. 9 μm (dimension along the building direction), and the Si-rich particles measure 12 nm vs. 60 nm, the larger values corresponding to the 200 °C material [13].

It should be noted that initial porosity of the samples studied in the present work is relatively low (i.e., 0.039% and 0.016%, for the 35 °C and 200 °C built platform temperatures respectively), as shown in our

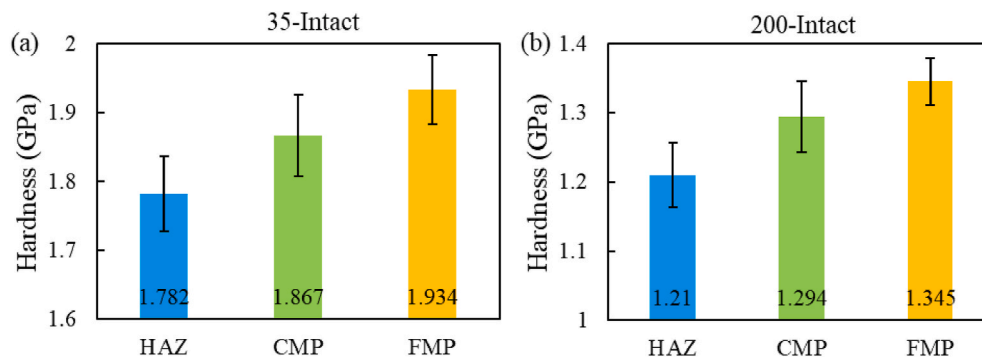


Fig. 4. Nano-hardness of three melt pool zones (FMP, CMP and HAZ) of intact 35 °C (a) and 200 °C (b) materials. The statistics are based on at least 20 data points for each zone.

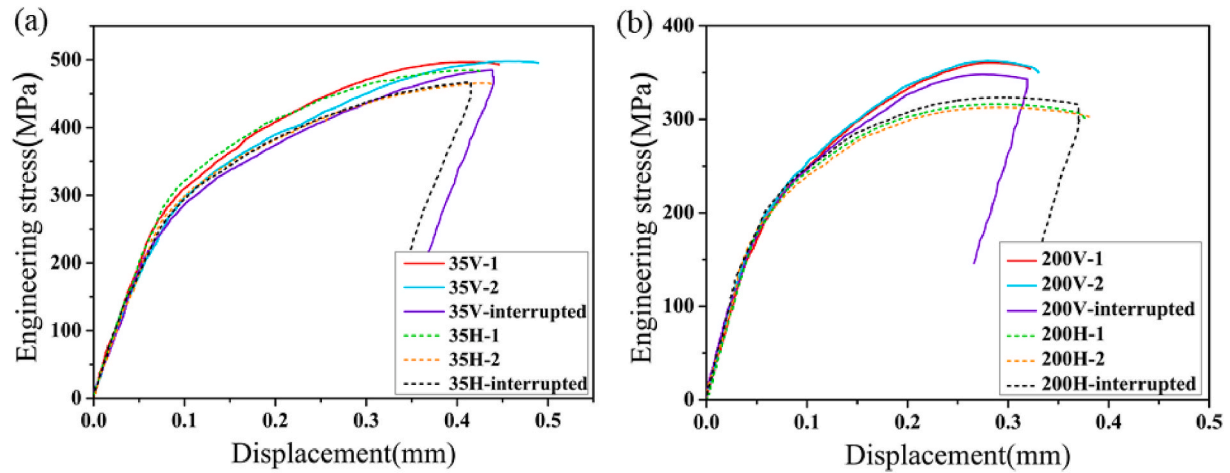


Fig. 5. Tensile curves for the 35 °C (a) and 200 °C (b) materials incorporating both broken and interrupted cases.

Table 2

Mean tensile properties of 35 °C and 200 °C materials. UTS denotes ultimate tensile strength. The values are recollected from 3 broken samples and 2 interrupted samples for each condition.

Designation	UTS (MPa)	Fracture strain	Loading displacement (mm)
35H	476 ± 6	0.105 ± 0.011	0.450 ± 0.03
35H-interrupted	/	/	0.415 and 0.44
35V	487 ± 10	0.107 ± 0.013	0.466 ± 0.03
35V-interrupted	/	/	0.44 and 0.416
200H	319 ± 5	0.103 ± 0.001	0.383 ± 0.005
200H-interrupted	/	/	0.369 and 0.373
200V	357 ± 8	0.096 ± 0.003	0.332 ± 0.01
200V-interrupted	/	/	0.313 and 0.286

previous work [40]. Such low porosity is expected to exert little impact on strength and ductility, as explained by Zhao et al. [38]. Therefore, the present work will only focus on the correlation between mechanical properties and the fine Si-rich phase size and shape.

3.2. Mechanical properties

The heterogeneous microstructure leads to local mechanical property variations which are revealed by nano-indentation tests. Fig. 4 presents the measured nano-hardness values for the three distinct zones of intact samples (i.e. prior to tensile loading), referred to as 35-Intact and 200-Intact. For both the 35 °C and 200 °C materials, the FMP and HAZ exhibit the highest and the lowest hardness, the relative difference being 7.9% and 10%, respectively. This is in line with the measurements by Delahaye et al. [32]. Due to coarser microstructure, the 200 °C

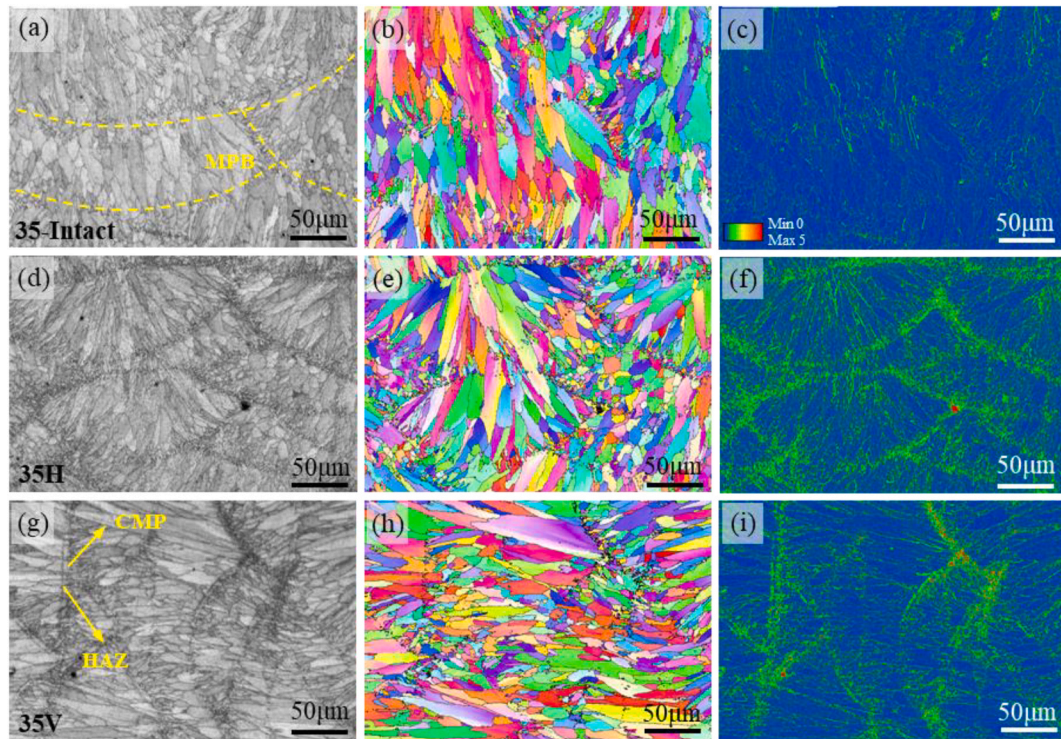


Fig. 6. EBSD image quality (IQ), inverse pole figures (IPF) coloring and kernel average misorientation (KAM) for intact (a)–(c), interrupted horizontal (d)–(f) and vertical (g)–(i) 35 °C samples.

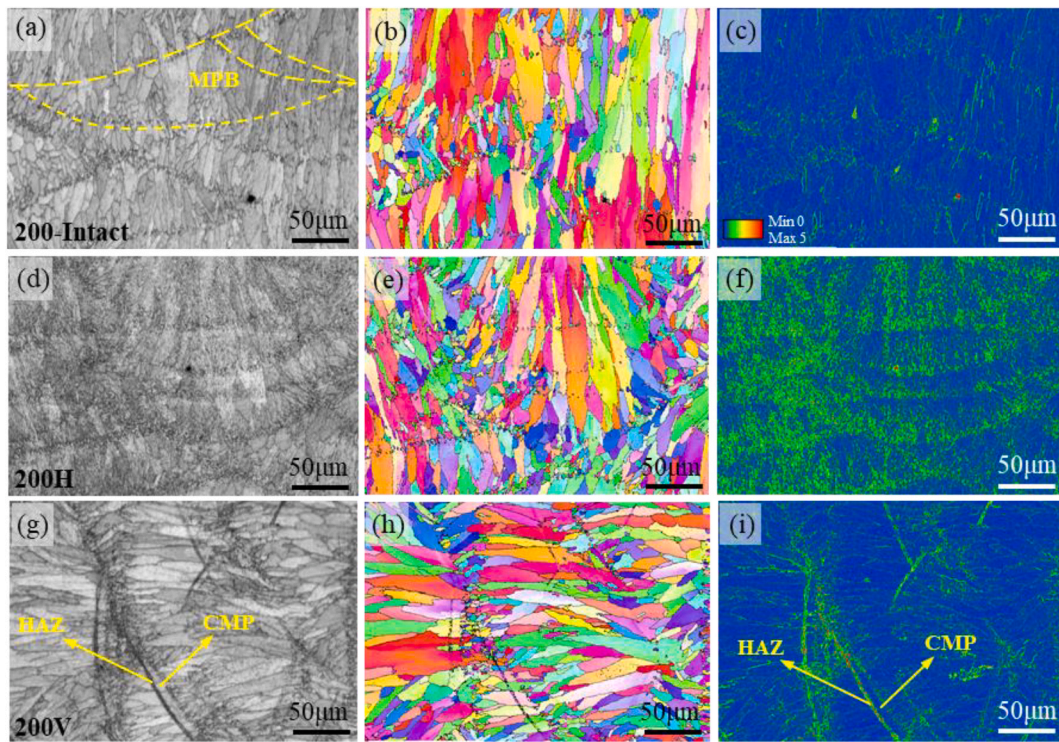


Fig. 7. EBSD image quality (IQ), inverse pole figures (IPF) coloring and kernel average misorientation (KAM) for intact (a)–(c), interrupted horizontal (d)–(f) and vertical (g)–(i) 200 °C samples.

material presents significantly lower hardness than the 35 °C counterpart in all three regions. The weaker melt pool border is expected to promote strain localization during tensile deformation.

The tensile behaviors are presented in Fig. 5, only two curves of broken samples and one curve of interrupted samples are shown for each condition to alleviate the figure illustration. The corresponding properties calculated from all tests are also given in Table 2. The tests present a good reproducibility. Note that as no extensometer was used, only the engineering stress vs. displacement relations are reported. Nevertheless, the true fracture strain was calculated with the section areas before and after the tests, using Eq. (1), see Table 2. These acquired experimental data are sufficient for the analysis of the present work seeking the link between strain hardening, damage and microstructure.

$$\epsilon_f = \ln(A_0/A_1) \quad (1)$$

Similar to the indentation test results, the tensile curves also reveal a higher strength for the 35 °C material (i.e., finer microstructure). The two build platform temperatures show different loading direction dependences. The 35 °C material exhibits very similar tensile behavior in the two loading directions (Fig. 5a), the hardening capacity being only slightly higher for the vertical direction. The true strain at fracture is very close between the two loading directions (see Table 2). Regarding the 200 °C material, a more remarkable anisotropy can be noticed, as reflected by the higher strain hardening capacity and lower elongation to failure of the vertical samples (Fig. 5b). However, as the vertical samples neck more significantly, their fracture strain is finally similar to that of the horizontal samples, see Table 2. Interestingly, all samples fail at a true strain of roughly 0.1, regardless of microstructure size, mechanical strength and loading direction. For the interrupted tests, unloading is performed just before fracture, as shown in Table 2.

3.3. Heterogeneous microstructure induced strain gradient

As the strength varies in the three melt pool zones (Fig. 4), microstructure is expected to experience heterogeneous deformation, leading

Table 3

Statistics of KAM values estimated at melt pool (MP) interior or border for intact and deformed 35 °C microstructures. Distinguished KAM values are provided for the CMP and HAZ in the 35V interrupted sample.

Location	35-Intact (°)	35H-interrupted (°)	35V-interrupted (°)
MP interior	0.41 ± 0.02	0.62 ± 0.03	0.46 ± 0.03
MP border	0.52 ± 0.04	0.83 ± 0.02	0.94 ± 0.12 (CMP) 1.04 ± 0.12 (HAZ)

to strain gradient. To reveal the dependence of strain gradient on microstructure size and loading direction, EBSD characterization has been conducted on the intact and interrupted samples. The image quality (IQ), inverse pole figure (IPF) coloring and kernel average misorientation (KAM) map are provided to thoroughly characterize the deformed microstructure. From the IQ and IPF graphs, it can be seen that the grains present very similar columnar morphologies between the 35 °C and 200 °C materials, with the long axes generally pointing toward the melt pool centers and several epitaxial grains crossing the melt pool borders (Figs. 6 and 7).

Fig. 6 and Table 3 present the results for the 35 °C material. Note that the KAM is a widely accepted indicator for density of geometrically necessary dislocations (GND) induced by plastic strain gradient [29,41]. It can be seen from Fig. 6c and Table 3 that the KAM values of the intact sample are non-zero, some high values are observed both in the MP interior and at the MP border. This is probably due to thermally induced GNDs formed at the Al/Si phase boundary during fast cooling [12]. As for the deformed samples, a strong strain gradient can be observed for both loading directions (see Fig. 6f and i). The vertical sample exhibits a more pronounced KAM concentration at the melt pool border than in the melt pool interior, compared to the horizontal sample (see Table 3). This indicates that the strain localization degree is higher for the vertical loading direction, as the weaker melt pool border is perpendicular to the load.

Table 4

Statistics of KAM values estimated at melt pool (MP) interior or border for intact and deformed 200 °C microstructures. Distinguished KAM values are provided for the CMP and HAZ in the 200V interrupted sample.

Location	200-Intact (°)	200H-interrupted (°)	200V-interrupted (°)
MP interior	0.39 ± 0.01	0.67 ± 0.03	0.42 ± 0.06
MP border	0.53 ± 0.04	0.84 ± 0.03	0.8 ± 0.10 (CMP) 1.3 ± 0.13 (HAZ)

Interestingly, the KAM concentration band has a similar thickness (roughly 8 μm) as the melt pool border (roughly 7 μm), suggesting simultaneous strain localization at the CMP and HAZ. This is further confirmed by similar KAM values in these two regions (see Table 3), both significantly higher than that of the FMP. Moreover, an increase of KAM value in the MP interior, in particular for the horizontal sample, can be clearly noticed (Table 3). This indicates that the GND density at the Al/Si boundary is enhanced during deformation, as also previously reported by Li et al. [29].

The EBSD results for the 200 °C material are shown in Fig. 7 and Table 4. The intact sample also exhibits non-zero GND density in the MP interior and at the MP border (Fig. 7a–c, Table 4). After deformation, the GND distributions are significantly different for the two loading directions. For the interrupted horizontal sample, the KAM values increase throughout the whole melt pool zones, no significant concentration at the MP border is found (Fig. 7d–f and Table 4). Regarding the interrupted vertical sample, a strong KAM concentration is observed, which appears to be confined in a band much thinner (roughly 4 μm) than the melt pool border (roughly 12 μm) (see Fig. 7g–i). From this geometrical

observation, it can be deduced that the strain gradient is more developed inside the HAZ for the vertical 200 °C sample, in contrast to the 35 °C counterpart where the CMP and HAZ present similar KAM values (compare Tables 3 and 4 for the vertical interrupted samples).

3.4. Damage and fracture behavior

Analysis of damage behavior is performed on both the broken and unloaded samples. Fig. 8 shows the typical fracture surface morphologies of the samples tested until fracture. A striking difference between the two loading directions can be noticed. The fracture surface of the vertical samples incorporates many melt pool traces (also called scan tracks), while the horizontal ones exhibit relatively flat fracture surfaces. This indicates that crack tends to frequently propagate along the melt pool border for the vertical samples. Such observations have been widely reported in previous works [3,19,21,33,34]. The Si-rich phase can be observed in high magnification SEM micrographs, as shown in Fig. S1 of the supplementary material.

A common feature of the fracture surfaces is that they all contain only few initial pores. No large lack of fusion defects are observed. This ensures that fracture is controlled by microstructure damage for the studied materials. To understand the difference in damage mechanism between the horizontal and vertical samples, it is necessary to figure out when damage initiates and how it is distributed in the microstructure of different samples.

To this end, the broken and interrupted samples are polished to mid-thickness plane, and the deformed microstructure is observed with SEM. All interrupted samples were investigated. As the observations are basically the same for each condition (see Table 2), only the interrupted tests with displacements reaching 92%, 94%, 96% and 94% of the

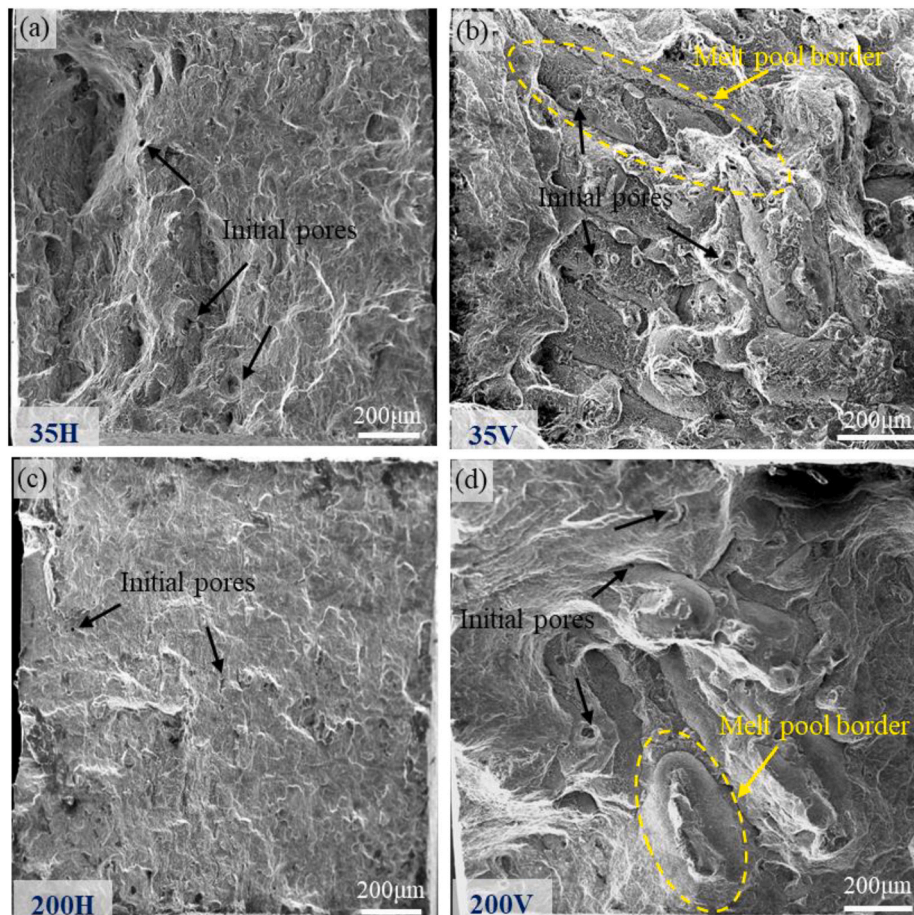


Fig. 8. Fracture surface morphology of 35H (a), 35V (b), 200H (c) and 200V (d) broken samples.

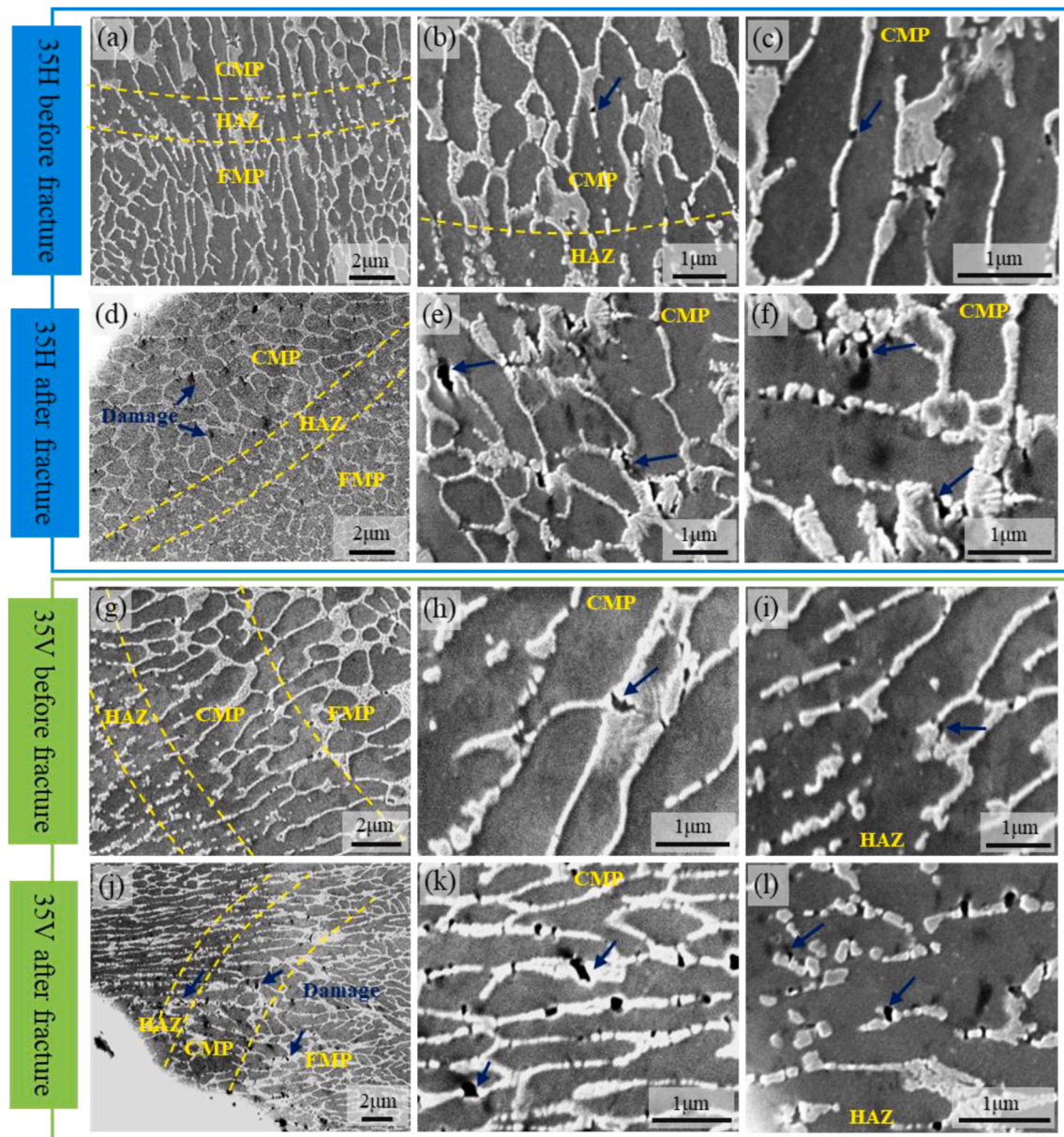


Fig. 9. Damage distribution at different MP zones of the 35 °C tensile samples. Interrupted horizontal sample (a)–(c), broken horizontal sample (d)–(f), interrupted vertical sample (g)–(i), broken vertical sample (j)–(l). The dark blue arrows point out the damage. (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

displacements at fracture of respectively the 35H, 35V, 200H and 200V samples are illustrated.

For each interrupted sample, the whole area of the reduced section is observed to probe the damage state, while only the area near the fracture surface is observed for the broken samples. The results for the 35 °C material are presented in Fig. 9. Surprisingly, the microstructure remains nearly intact for both the interrupted horizontal (Fig. 9a–c) and vertical (Fig. 9g–i) samples, though the deformation level is already very close to final failure. Regarding the broken samples, damage is mainly distributed at the Si-rich networks for the horizontal direction (Fig. 9d–f), while it can also be seen at the HAZ for the vertical direction (Fig. 9j–l), in agreement with previous literature [39].

From the comparison between the unloaded and broken samples, it can be deduced that damage initiates very late for the 35 °C microstructure, regardless of the loading direction. This poses a major challenge for in-situ mechanical testing aiming to capture the damage initiation and evolution, as mentioned by Zhao et al. [38].

As for the 200 °C material, damage evolution exhibits a strong dependence on the loading direction. Few damage sites can be observed for the interrupted 200H sample (96% of failure elongation), which are mainly located at the CMP as presented in Fig. 10a–c. However, substantial damages are already formed in the microstructure of the interrupted 200V sample (94% of failure elongation), see Fig. 10g–i. These damages are more or less homogeneously dispersed in the reduced section instead of being confined in a specific region. This is also confirmed by the damage distribution in the broken sample. Fig. 10j–l shows a high density of damage near the fracture surface of the vertical sample, with damage sites located at both interconnected (CMP or FMP, see Fig. 10k) and degraded (HAZ, see Fig. 10l) Si-rich networks. In contrast, damage is significantly less dense in the broken horizontal sample, and the HAZ exhibits very few damages, as can be seen in Fig. 10d–f.

To verify the early onset of damage for the 200V sample, the second interrupted test was unloaded at a lower deformation level (i.e., 86% of

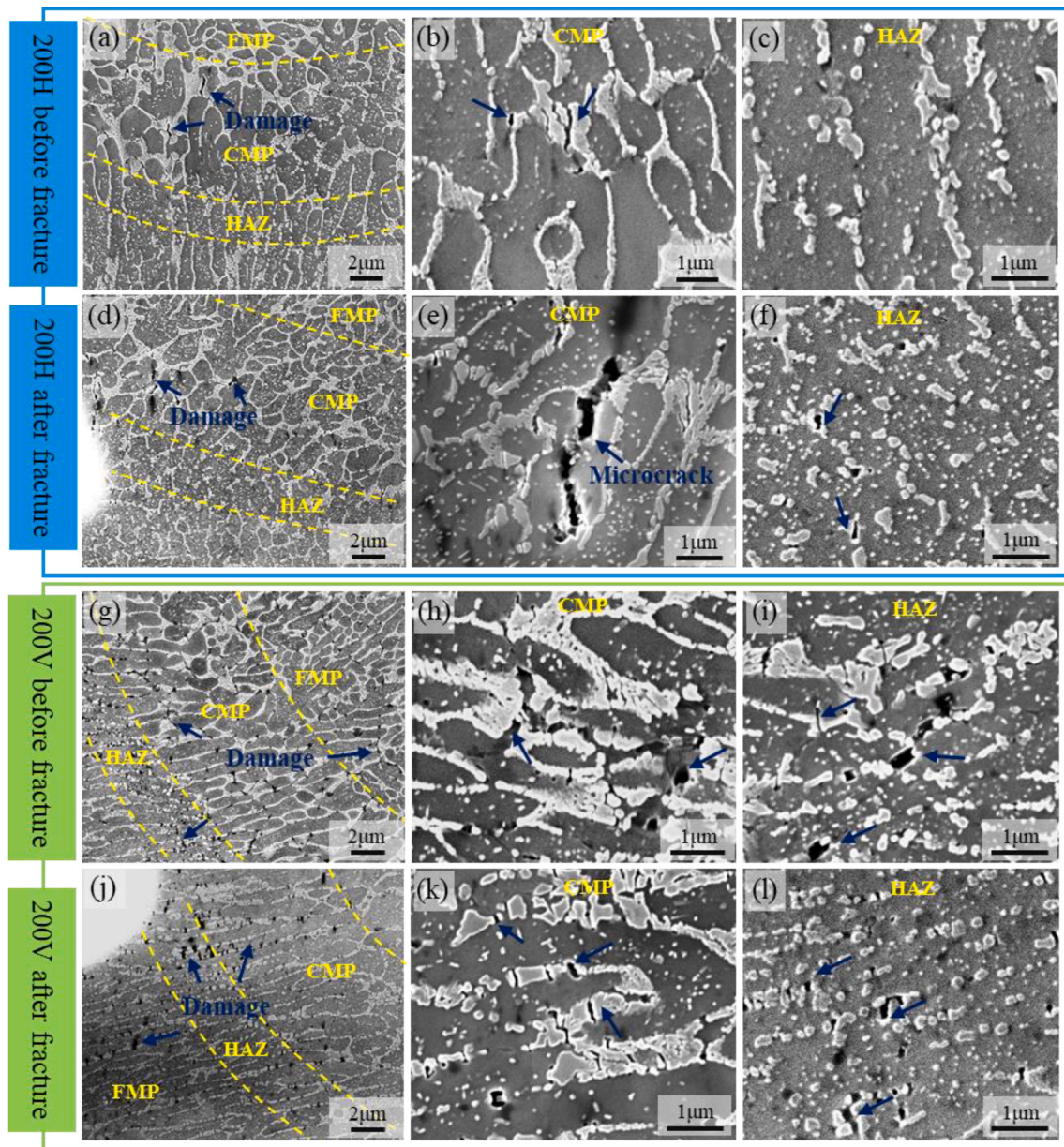


Fig. 10. Damage distribution at different MP zones of the 200 °C tensile samples. Interrupted horizontal sample (a)–(c), broken horizontal sample (d)–(f), interrupted vertical sample (g)–(i), broken vertical sample (j)–(l). The dark blue arrows point out the damage. (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

the elongation to fracture). The corresponding damage observations are presented in the supplementary material (see Fig. S2). It is found that damage already initiates. Nevertheless, both the damage size and density are much lower than that for the 94% counterpart. Moreover, damage is only randomly located at specific regions (mainly in CMP and HAZ), in contrast to the 94% case where a more homogeneous distribution is observed.

In order to further confirm the presence and distribution of damage, nano-indentation tests have been performed on the same interrupted samples as for the damage observations (Figs. 9 and 10). Note that local property degradation can be regarded as a damage indicator [42,43]. Examples of indent distribution across the melt pool border are shown in Supplementary Fig. S3, and the number of indentation points for each sample is given in Supplementary Table S1.

The hardness values of the interrupted samples are presented in Fig. 11. The results for the intact microstructures are re-plotted to clearly show the property changes at the FMP, CMP and HAZ. For the 35 °C

material (Fig. 11a), both the horizontal and vertical samples exhibit enhanced hardness in all three zones, due to strain hardening and lack of damage. In regard to the 200 °C material (Fig. 11b), increased hardness has also been noticed for the horizontal sample, similarly to the 35 °C counterpart. However, the vertical sample exhibits decreased hardness in all three zones, though the microstructure has been highly deformed (94% of elongation to fracture). This property degradation is consistent with the damage state [42], as Fig. 10g–i shows substantial voids in the interrupted 200V sample in all three zones. It is noteworthy that the simultaneous hardness decrease at the FMP, CMP and HAZ confirms the non-concentrated damage distribution in the 200V sample, in line with the SEM observations (Fig. 10g).

3.5. Summary of the most significant observations

The present work has investigated the mechanical behavior of fine (35 °C) and coarse (200 °C) LPBF AlSi10Mg microstructures, with a

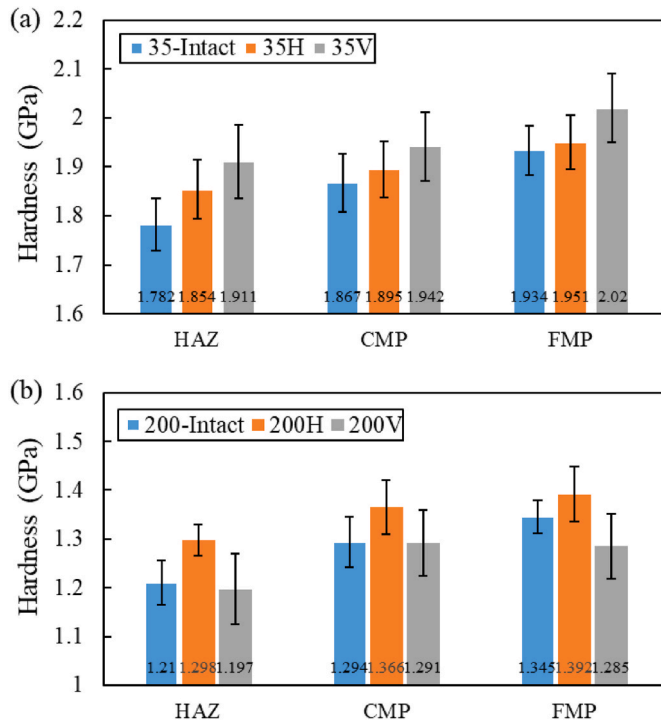


Fig. 11. Hardness values in the three melt pool zones of intact and loaded then interrupted tensile samples. 35 °C material (a), 200 °C material (b). For the HAZ, CMP and FMP, the 35H exhibits an increase by 3.9%, 1.5% and 0.9%, 35V by 6.8%, 4% and 4.4%, 200H by 7.3%, 5.6% and 3.5%, while 200V exhibits a decreases by 1%, 0.2% and 4.4%, compared to the intact state.

Table 5

Important observations and corresponding supports in the present work. The results for damage initiation, strain gradient and hardness evolution are obtained with the interrupted samples.

Important observations	Evidences
Microstructure: 35 °C-finer; 200 °C-coarser	Fig. 3
Flow behavior: 35 °C-nearly isotropic; 200 °C-anisotropic	Fig. 5
Strain gradient: 35V-similar in CMP and HAZ; 200V-stronger in HAZ	Figs. 6 and 7, Tables 3–4
Damage initiation: 35 °C-late for H and V; 200 °C-late for H, early for V	Figs. 9 and 10
Nanohardness evolution: 35 °C-rise for H and V; 200 °C-rise for H, drop for V	Fig. 11

focus on the strain hardening and damage evolution in the horizontal and vertical loading directions. The main findings are succinctly summarized in Table 5, with corresponding figures and tables recalled as evidences. The following discussion section tends to disclose the correlations between the microstructure, strain hardening and damage mechanism.

4. Discussion

4.1. Microstructure heterogeneity

LPBF AlSi10Mg exhibits three distinct melt pool zones, i.e., FMP, CMP and HAZ. The Si-rich eutectic phase mainly manifests as elongated and interconnected network, with coarsened structure at the CMP and degraded connectivity at the HAZ. This microstructure feature leads to local strength variation (Fig. 4) and load direction dependent hardening behavior (Fig. 5), as will be further discussed in the following sections.

The microstructure and its heterogeneity can be altered by the build platform temperature. As shown in the present work, the 200 °C

preheating has coarsened the microstructure in terms of nano-sized particles and Si-rich network (Fig. 3). Nevertheless, the relative hardness decreases for the CMP and HAZ with respect to the FMP are similar to those for the 35 °C microstructure (Fig. 4). The microstructure size and heterogeneity not only determine the mechanical strength and strain distribution, but also impact the damage evolution, as coarser secondary phase is more prone to damage [44]. The combined effect of strain localization and damage sensibility should be considered to understand the damage mechanism in such complex microstructure, which further shows a loading direction dependence (Figs. 8–10).

4.2. Strain hardening

The 35 °C material has been found to show nearly isotropic flow behavior, while the 200 °C counterpart exhibits a remarkably higher strain hardening in the vertical direction (see Fig. 5). As the microstructure contains two phases, the flow stress can therefore be divided into two parts: the stress of Al cells controlled by dislocation motion and the stress of Si-rich network determined by its load bearing capacity.

When analyzing the Schmid factor maps and distributions, as presented in Fig. 12, both microstructures show similar results in the two loading directions, with very slightly higher mean values for the horizontal direction. In this sense, the dislocation motion determined hardening behavior (i.e., stress within the Al cells) is expected to be independent of loading direction, in case where strengthening precipitates (i.e., nano-sized Si rich particles) are homogeneously distributed within the Al cells. Note that the homogeneous Si particle distribution has been demonstrated in our previous work [13].

As the Si phase presents much higher Young's modulus and strength than the Al phase, its load bearing can significantly contribute to macroscopic strain hardening of Al–Si alloy. According to the stress partition analysis performed by Zhang et al. [28] via X-ray diffraction, the Si-rich network exhibits a very high load bearing effect in LPBF AlSi10Mg, as reflected by high (35.4%) volume based load partition ratio (i.e., $f_{Si}\sigma_{Si}/\sigma_{Tot} = 35.4\%$, f_{Si} being only 9.6%) reached during flow stage. The load bearing effect of secondary phase depends on the strength of the matrix material and the aspect ratio (i.e., shape of the secondary phase). If the secondary phase presents identical strength to the matrix, its volume based load partition ratio will be exactly the second phase volume fraction. In contrast, if the secondary phase is much stronger, as for LPBF AlSi10Mg, its load partition will become more pronounced, as proved by the large gap between 35.4% and 9.6% mentioned above.

The difference in strain hardening behavior of the 35 °C and 200 °C materials can be rationalized with the strength of Al cells and the aspect ratio of Si-rich network:

- (1) On the one hand, the 200 °C microstructure involves much weaker Al cells than the 35 °C counterpart (see Fig. 4), mainly due to coarser Si-rich particles, as also demonstrated in Ref. [13] through strengthening models. This implies that the contribution to flow strength (i.e., volume based stress partition ratio) from the Si-rich network should be higher (respectively lower) for the 200 °C (respectively 35 °C) material.
- (2) On the other hand, as the Si-rich network presents elongated morphology in the building direction (see Fig. 3), its load bearing effect therefore depends on the loading direction. In fact, the effective aspect ratio of the Si-rich network is higher for the vertical direction, leading to stronger load bearing effect [45].

From the aforementioned points, the nearly isotropic hardening behavior of the 35 °C material can be explained by the fact that the load bearing effect of the Si-rich network is relatively low (due to strong Al cells) and thus cannot promote significant difference of hardening in the two loading directions. In contrast, for the 200 °C microstructure, the Si-rich network represents an important hardening component (due to

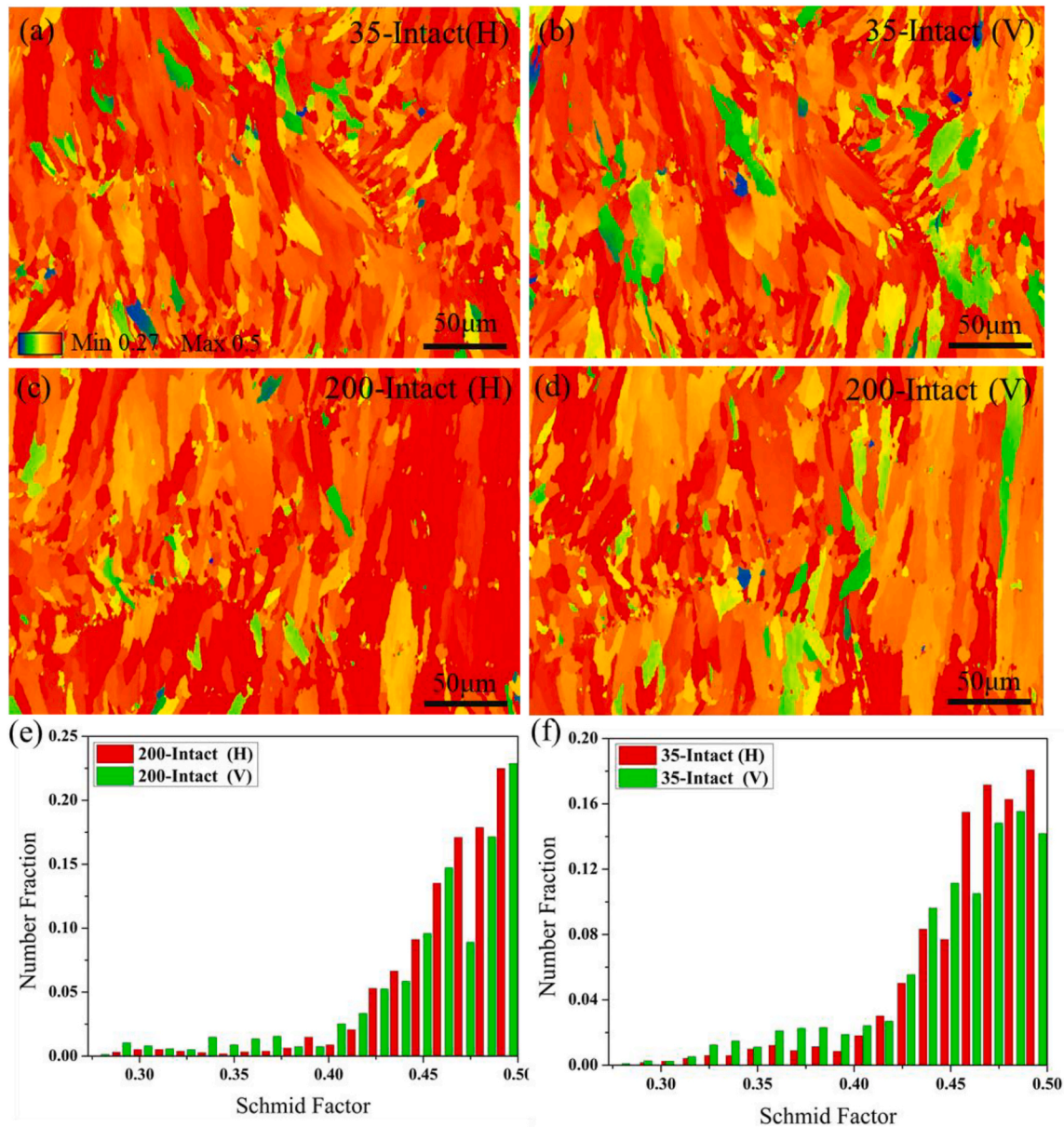


Fig. 12. Schmid factor map for the intact 35H (a), 35V (b), 200H (c), 200V (d) microstructures and corresponding Schmid factor distributions for the 35 °C (e) and 200 °C (f) materials. The mean Schmid factor values are 0.459, 0.449, 0.463 and 0.456 for the 35H, 35V, 200H and 200V, respectively.

weak Al cells), and its higher effective aspect ratio in the vertical direction therefore leads to stronger hardening behavior.

4.3. Strain localization at the melt pool borders

Due to the presence of weaker CMP and HAZ, strain localization will occur at melt pool border, leading to significant strain gradient. Moreover, the strain localization degree appears to depend on the loading direction as the weaker zone deforms more favorably when the melt pool border is perpendicular to the external load (i.e., vertical samples).

Nevertheless, strain localization might be mitigated if weaker region exhibits a strong hardening capacity. Indeed, as shown by the hardness values (see Fig. 11), the relative difference between the HAZ and FMP is reduced from 10% (intact state) to 5% (deformed state) for the 35 °C vertical samples, implying that the strain distribution tends to converge to homogeneous state. This phenomenon could be explained by the increased GND density across the melt pool border (see Figs. 6 and 7, Tables 3 and 4), thus enhancing the strain hardening of this initial

weaker region. The GNDs assisted hardening would alter the stress partition between the Al cells and the Si-rich network, since the Al cells become stronger due to the enhanced GND density. This in turn affects the damage behavior at the MP border, as will be discussed in the following section.

4.4. Damage evolution

A very interesting finding of the present work is that in the absence of significant porosity effect, the fine and coarse microstructures exhibit very similar ductility (roughly 0.1, see Table 2), regardless of the loading direction. However, the damage evolution behaviors leading to this fracture strain are different for the two microstructures and show a direction dependence.

For the fine microstructure (35 °C), damage initiates very late for both the horizontal and vertical samples (Table 5), as schematically represented in Fig. 13a and b. Our characterizations show that no obvious damage can be seen even at 94% of elongation to failure

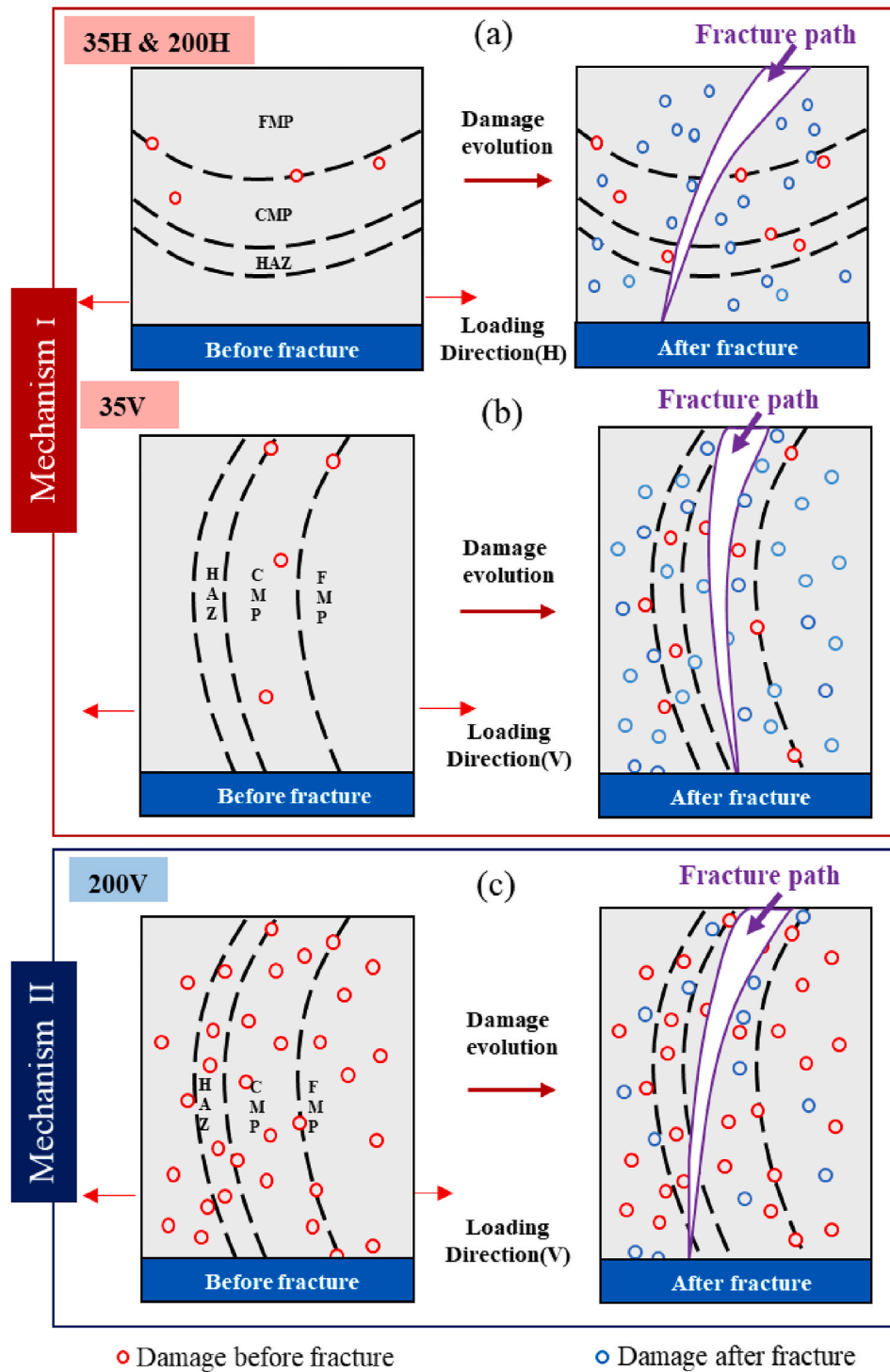


Fig. 13. Schematic drawing of damage evolution in 35H and 200H (a), 35V (b) and 200V (c) samples.

(Fig. 9). This further indicates a rapid damage coalescence following damage initiation. The similar ductility for the two loading directions can be rationalized with the stress partition between the Al cells and Si-rich network, which determines damage initiation. In fact, the fracture of brittle secondary phase (i.e., damage initiation) is controlled by a stress threshold [46,47]. As discussed in section 4.2, the load bearing of the Si-rich network would be insignificant for the strain hardening due to relatively high-strength Al cells of the 35 °C material, compared to the 200 °C counterpart. Indeed, the tensile curves of the 35 °C material have revealed very similar flow stage until final fracture for the two loading

directions (Fig. 5a). Considering the direction independent dislocation behavior deduced from close Schmid factor distribution (see Fig. 12), the similar flow behavior implies a close stress level of Si-rich phase as a function of macroscopic strain, rationalizing the similar strain at damage initiation for the two loading directions.

Regarding the coarse microstructure (200 °C), the damage behavior shows a significant load direction dependence, as revealed in Fig. 10. The damage initiation in the horizontal direction is also very late, similarly to the 35 °C material (Table 5), as schematically shown in Fig. 13a. In contrast, the vertical sample damages much earlier,

presenting quite dispersed damages sites, see the schematic drawing in Fig. 13c. This can also be related to the hardening behavior and explained by stress partitioning. As aforesaid (section 4.2) for the 200 °C material, the contribution of Si-rich network to hardening is on the one hand significant due to weak Al cells, and on the other hand more pronounced along the vertical direction due to higher effective aspect ratio of the Si-rich network. Consequently, the higher stress level in the Si-rich phase leads to earlier damage of vertical samples. Moreover, upon damage initiation in the vertical samples, the stress level in the Al cells would still be low, resulting in slow damage growth and coalescence, and allowing for additional damage initiation before final failure. In contrast, the horizontal samples damage relatively late, the damage therefore has little room for growth, and coalesces quickly due to high stress level in the Al cells, leading to more concentrated damage distribution.

For the 35V samples, the crack has been found to frequently propagate along the melt pool border (see Fig. 8b). The damage observations on broken samples further confirm that the crack path frequently coincides with the CMP, where the coarser Si-rich network is more prone to damage. Nevertheless, due to the presence of substantial GNDs at the melt pool border enhancing flow stress within the Al cells, the effective load bearing of Si-rich phase of the CMP and HAZ is expected to be lower than that of the FMP. Consequently, the CMP would not damage prematurely and compromise the material ductility.

In regard to the 200V cases, the KAM map of the interrupted sample (see Fig. 7i) shows a peculiar high strain gradient at the HAZ. Nevertheless, both the CMP and HAZ have been found to be preferred crack path. This can also be explained by the balance between phase stress and strain gradient: weaker HAZ tends to deform more due to lower strength; the increasing deformation, on the one hand raises stress in the Si-rich phase, and on the other hand builds up strain gradient and promotes GNDs which in turn alleviate the load bearing of the Si-rich phase, reducing the probability of its failure. This trade-off between phase stress and strain gradient might be responsible for the discord between strain localization with the damage distribution.

5. Conclusions

The mechanical behavior of LPBF AlSi10Mg with different platform temperatures (35 °C and 200 °C) and loading directions (horizontal and vertical) was investigated in the present work. Combining tensile test, nano-indentation, EBSD and SEM results, it was found that strain hardening and damage mechanisms present a dependence on microstructure and melt pool border orientation, with however nearly unchanged material ductility. The main conclusions are as follows:

- (1) The 35 °C and 200 °C build platform generates a fine and coarse microstructure, respectively. The fine microstructure exhibits nearly isotropic tensile properties, while the coarse one shows anisotropic hardening behavior, which can be explained by direction dependent load bearing effect of cellular Si-rich network.
- (2) Fracture of 35 °C material (i.e., fine microstructure) is controlled by damage initiation, as growth and coalescence stages are very short. Indeed, damage occurs very close to final failure and presents localized distribution.
- (3) Damage evolution shows an anisotropy for the 200 °C (i.e., coarse) microstructure. The vertical sample damages much earlier than the horizontal counterpart due to higher load bearing effect of the Si-rich network, resulting in decreased local hardness and probably promoting macroscopic necking.
- (4) The 200 °C vertical sample exhibits pronounced strain localization at the HAZ, while the CMP is even more damaged. This unexpected behavior may be explained by the trade-off between the Si phase stress and strain gradient (i.e., GND effect).

Data availability

The data that support the findings of this study are available from the corresponding author upon reasonable request.

CRediT authorship contribution statement

Lubin Song: Formal analysis, Investigation, Writing – original draft. **Lv Zhao:** Conceptualization, Formal analysis, Supervision, Writing – review & editing. **Lipeng Ding:** Formal analysis, Writing – review & editing. **Yaxin Zhu:** Investigation, Writing – review & editing. **Minsheng Huang:** Investigation, Writing – review & editing. **Aude Simar:** Investigation, Writing – review & editing. **Zhenhuan Li:** Investigation, Writing – review & editing.

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.

Acknowledgements

L.Z. acknowledges the support from the National Natural Science Foundation of China (No. 12002128) and the Fundamental Research Funds for the Central Universities, China (2021XXJS116). A.S. thanks the WALInnov LongLifeAM project (convention n°1810016) funded by the Service public de Wallonie Économie Emploi Recherche (SPW-EER) and the European Research Council (ERC) for the Starting Grant ALUFIX project (grant agreement n°716678). This work is also supported by the Jiangsu Key Laboratory of Lightweight Materials, Nanjing Tech University.

Appendix A. Supplementary data

Supplementary data to this article can be found online at <https://doi.org/10.1016/j.msea.2021.142484>.

References

- [1] P. Li, Y. Kim, A.C. Bobel, L.G. H Jr., A.K. Sachdev, S. Kumar, A.F. Bower, Microstructural origin of the anisotropic flow stress of laser powder bed fused AlSi10Mg, *Acta Mater.* 220 (2021), 117346.
- [2] S. Dai, Z.C. Deng, Y.J. Yu, K.Y. Zhu, Microstructure and constitutive model for flow behavior of AlSi10Mg by selective laser melting, *Mater. Sci. Eng.* 2021 (2021), 141157.
- [3] U. Patakham, A. Palasay, P. Wila, R. Tongsi, MPB characteristics and Si morphologies on mechanical properties and fracture behavior of SLM AlSi10Mg, *Mater. Sci. Eng.* 821 (2021), 141602.
- [4] E.O. Olakanmi, R.F. Cochrane, K.W. Dalgarno, A review on selective laser sintering/melting (SLS/SLM) of aluminium alloy powders: processing, microstructure, and properties, *Prog. Mater. Sci.* 74 (2015) 401–477.
- [5] Y.N. Hu, S.C. Wu, P.J. Withers, J. Zhang, H.Y.X. Bao, Y.N. Fu, G.Z. Kang, The effect of manufacturing defects on the fatigue life of selective laser melted Ti-6Al-4V structures, *Mater. Des.* 192 (2020), 108708.
- [6] N.T. Aboulkhair, M. Simonelli, L. Parry, I. Ashcroft, C. Tuck, R. Hague, 3D printing of Aluminium alloys: additive Manufacturing of Aluminium alloys using selective laser melting, *Prog. Mater. Sci.* 106 (2019), 100578.
- [7] E. Georgantzia, M. Gkantou, G.S. Kamaris, Aluminium alloys as structural material: a review of research, *Eng. Struct.* 227 (2021), 111372.
- [8] H.R. Kotadia, G. Gibbons, A. Das, P.D. Howes, A review of laser powder bed fusion additive manufacturing of aluminium alloys: microstructure and properties, *Additive Manufacturing* 46 (2021), 102155.
- [9] N.T. Aboulkhair, I. Maskery, C. Tuck, I. Ashcroft, N.M. Everitt, The microstructure and mechanical properties of selectively laser melted AlSi10Mg: the effect of a conventional T6-like heat treatment, *Mater. Sci. Eng.* 667 (2019) 139–146.
- [10] L. Girelli, M. Tocci, M. Gelfi, A. Pola, Study of heat treatment parameters for additively manufactured AlSi10Mg in comparison with corresponding cast alloy, *Mater. Sci. Eng.* 739 (2019) 317–328.
- [11] A.H. Maamoun, M. Elbestawi, G.K. Dosbaeva, S.C. Veldhuis, Thermal post-processing of AlSi10Mg parts produced by selective laser melting using recycled powder, *Additive Manufacturing* 21 (2018) 234–247.

- [12] W. Lefebvre, G. Rose, P. Delroisse, E. Baustert, F. Cuvilly, A. Simar, Nanoscale periodic gradients generated by laser powder bed fusion of an AlSi10Mg alloy, *Mater. Des.* 197 (2021), 109264.
- [13] J.G.S. Macías, T. Douillard, L. Zhao, E. Maire, G. Pyka, A. Simar, Influence on microstructure, strength and ductility of build platform temperature during laser powder bed fusion of AlSi10Mg, *Acta Mater.* 201 (2020) 231–243.
- [14] E. Brandl, U. Heckenberger, V. Holzinger, D. Buchbinder, Additive manufactured AlSi10Mg samples using selective laser melting (SLM): microstructure, high cycle fatigue, and fracture behavior, *Mater. Des.* 34 (2012) 159–169.
- [15] K.G. Prashanth, J. Eckert, Formation of metastable cellular microstructures in selective laser melted alloys, *J. Alloys Compd.* 707 (2017) 27–34.
- [16] P.V. Cauwenbergh, V. Samae, L. Thijs, J. Nejezchlebová, P. Sedlák, A. Iveković, D. Schryvers, B.V. Hooreweder, K. Vanmeensel, Unravelling the multi-scale structure–property relationship of laser powder bed fusion processed and heat-treated AlSi10Mg, *Sci. Rep.* 11 (2021) 6423.
- [17] X.P. Li, X.J. Wang, M. Saunders, A. Suvorova, L.C. Zhang, Y.J. Liu, M.H. Fang, Z. H. Huang, T.B. Sercombe, A selective laser melting and solution heat treatment refined Al–12Si alloy with a controllable ultrafine eutectic microstructure and 25% tensile ductility, *Acta Mater.* 95 (2015) 74–82.
- [18] L. Thijs, K. Kempen, J.P. Kruth, J.V. Humbeeck, Fine-structured aluminium products with controllable texture by selective laser melting of pre-alloyed AlSi10Mg powder, *Acta Mater.* 61 (2013) 1809–1819.
- [19] N. Takata, H. Kodaira, K. Sekizawa, A. Suzuki, M. Kobashi, Change in microstructure of selectively laser melted AlSi10Mg alloy with heat treatments, *Mater. Sci. Eng.* 704 (2017) 218–228.
- [20] X. Liu, C. Zhao, X. Zhou, Z. Shen, W. Liu, Microstructure of selective laser melted AlSi10Mg alloy, *Mater. Des.* 168 (2019), 107667.
- [21] M.J. Paul, Q. Liu, J.P. Best, X. Li, J.J. Kruzic, U. Ramamurthy, B. Gludovatz, Fracture resistance of AlSi10Mg fabricated by laser powder bed fusion, *Acta Mater.* 211 (2021), 116869.
- [22] B. Chen, S.K. Moon, X. Yao, G. Bi, J. Shen, J. Umeda, K. Kondoh, Strength and strain hardening of a selective laser melted AlSi10Mg alloy, *Scripta Mater.* 141 (2019) 45–49.
- [23] Q. Liu, H. Wu, M.J. Paul, P. He, Z. Peng, B. Gludovatz, J.J. Kruzic, C.H. Wang, X. Liu, Machine-learning assisted laser powder bed fusion process optimization for AlSi10Mg: new microstructure description indices and fracture mechanisms, *Acta Mater.* 201 (2020) 316–328.
- [24] Q. Yan, B. Song, Y. Shi, Comparative study of performance comparison of AlSi10Mg alloy prepared by selective laser melting and casting, *J. Mater. Sci. Technol.* 41 (2020) 199–208.
- [25] H. Wu, Y. Ren, J. Ren, L. Liang, R. Li, Q. Fang, A. Cai, Q. Shan, Y. Tian, I. Baker, Selective laser melted AlSi10Mg alloy under melting mode transition: microstructure evolution, nanomechanical behaviors and tensile properties, *J. Alloys Compd.* 873 (2021), 159823.
- [26] J. Wu, X.Q. Wang, W. Wang, M.M. Attallah, M.H. Loretto, Microstructure and strength of selectively laser melted AlSi10Mg, *Acta Mater.* 117 (2016) 311–320.
- [27] N. Takata, M. Liu, H. Kodaira, A. Suzuki, M. Kobashi, Anomalous strengthening by supersaturated solid solutions of selectively laser melted Al–Si-based alloys, *Additive Manufacturing* 33 (2020), 101152.
- [28] X.X. Zhang, A. Lutz, H. Andrä, M. Lahres, W.M. Gan, E. Maawad, C. Emmelmann, Evolution of microscopic strains, stresses, and dislocation density during in-situ tensile loading of additively manufactured AlSi10Mg alloy, *Int. J. Plast.* 139 (2021), 102946.
- [29] Z. Li, Z. Li, Z. Tan, D. Xiong, Q. Guo, Stress relaxation and the cellular structure-dependence of plastic deformation in additively manufactured AlSi10Mg alloys, *Int. J. Plast.* 127 (2020), 102640.
- [30] A. Hadadzadeh, C. Baxter, B.S. Amirkhiz, M. Mohammadi, Strengthening mechanisms in direct metal laser sintered AlSi10Mg: comparison between virgin and recycled powders, *Additive Manufacturing* 23 (2018) 108–120.
- [31] L. Zhao, J.G.S. Macías, T. Douillard, Z.-H. Li, A. Simar, Unveiling damage sites and fracture path in laser powder bed fusion AlSi10Mg: comparison between horizontal and vertical loading directions, *Mater. Sci. Eng.* 807 (2021), 140845.
- [32] J. Delahaye, J.T. Tchuindjang, J. Lecomte-Beckers, O. Rigo, A.M. Habraken, A. Mertens, Influence of Si precipitates on fracture mechanisms of AlSi10Mg parts processed by Selective Laser Melting, *Acta Mater.* 175 (2019) 160–170.
- [33] Z.H. Xiong, S.L. Liu, S.F. Li, Y. Shi, Y.F. Yang, R.D.K. Misra, Role of melt pool boundary condition in determining the mechanical properties of selective laser melting AlSi10Mg alloy, *Mater. Sci. Eng.* 740–741 (2019) 148–156.
- [34] X. Li, D. Yi, X. Wu, J. Zhang, X. Yang, Z. Zhao, Y. Feng, J. Wang, P. Bai, B. Liu, Y. Liu, Effect of construction angles on microstructure and mechanical properties of AlSi10Mg alloy fabricated by selective laser melting, *J. Alloys Compd.* 881 (2021), 160459.
- [35] A. Pineau, A.A. Benzerga, T. Pardoen, Failure of metals I: brittle and ductile fracture, *Acta Mater.* 107 (2016) 424–483.
- [36] L. Babout, E. Maire, R. Fougères, Damage initiation in model metallic materials: X-ray tomography and modelling, *Acta Mater.* 52 (2004) 2475–2487.
- [37] T. Pardoen, J.W. Hutchinson, Micromechanics-based model for trends in toughness of ductile metals, *Acta Mater.* 51 (2003) 133–148.
- [38] L. Zhao, J.G.S. Macías, L. Ding, H. Idrissi, A. Simar, Damage mechanisms in selective laser melted AlSi10Mg under as built and different post-treatment conditions, *Mater. Sci. Eng.* 764 (2019), 138210.
- [39] T. Maconachie, M. Leary, J. Zhang, A. Medvedev, A. Sarker, D. Ruan, G. Lu, O. Faruque, M. Brandt, Effect of build orientation on the quasi-static and dynamic response of SLM AlSi10Mg, *Mater. Sci. Eng.* 788 (2020), 139445.
- [40] L. Zhao, J.G.S. Macías, A. Dolimont, A. Simar, E. Rivière-Lorhèvre, Comparison of residual stresses obtained by the crack compliance method for parts produced by different metal additive manufacturing techniques and after friction stir processing, *Additive Manufacturing* 36 (2020), 101499.
- [41] D. Ben, Y. Ma, H. Yang, L. Meng, X. Shao, H. Liu, S. Wang, Q. Duan, Z. Zhang, Heterogeneous microstructure and voids dependence of tensile deformation in a selective laser melted AlSi10Mg alloy, *Mater. Sci. Eng.* 798 (2020), 140109.
- [42] C.C. Tasan, J.P.M. Hoefnagels, M.G.D. Geers, A critical assessment of indentation-based ductile damage quantification, *Acta Mater.* 57 (2009) 4957–4966.
- [43] C.C. Tasan, J.P.M. Hoefnagels, M.G.D. Geers, Indentation-based damage quantification revisited, *Scripta Mater.* 63 (2010) 316–319.
- [44] F. Hannard, S. Castin, E. Maire, R. Mokso, T. Pardoen, A. Simar, Ductilization of aluminium alloy 6056 by friction stir processing, *Acta Mater.* 130 (2017) 121–136.
- [45] X.X. Zhang, H. Andrä, Crystal plasticity simulation of the macroscale and microscale stress-strain relations of additively manufactured AlSi10Mg alloy, *Comput. Mater. Sci.* 200 (2021), 110832.
- [46] F. Beremin, Cavity formation from inclusions in ductile fracture of A508 steel, *Metall. Mater. Trans.* 12 (1981) 723–731.
- [47] F. Hannard, T. Pardoen, E. Maire, C.L. Bourlot, R. Mokso, A. Simar, Characterization and micromechanical modelling of microstructural heterogeneity effects on ductile fracture of 6xxx aluminium alloys, *Acta Mater.* 103 (2016) 558–572.