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Influence of Heat Treatment and Thermal Aging on the Fatigue Performance of LPBF Fabricated AlSi10Mg Alloy

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Abstract: This investigation explores fatigue performance of laser powder bed fusion (LPBF)-fabricated AlSi10Mg alloy subjected to various post-processing treatments: stress relief (SR), Hot Isostatic Pressing (HIP), and HIP + T6, all followed by prolonged thermal aging at 177 °C for up to 1000 h. Microstructural evolution, hardness, density, and uniaxial fatigue performance, including SEM fractography, were examined to recommend practical service life thresholds. Microstructural analysis confirmed that HIP and HIP + T6 treatments effectively eliminated process induced lack of fusion porosity, which was identified as the direct cause of premature elastic regime failure in SR-treated samples. While HIP + T6 achieved the highest initial hardness, both HIP-treated groups showed susceptibility to thermal degradation. The HIP-only samples, in particular, suffered a fatigue life reduction of up to 38% after just 10 h of aging. Fracture analysis showed that unaged SR samples failed in a comparatively more brittle manner, characterized by tear ridges, whereas HIP and HIP + T6 samples displayed an initially ductile fracture with prominent dimple features. Prolonged aging caused the HIP + T6 samples to replace ductile dimple features with a mix of dimple features along with uneven patterns, which may correlate with their decreased fatigue performance. The results of this paper conclusively demonstrate that while HIP post-processing provides robust initial fatigue strength, its benefits are severely compromised by thermal exposure. And for applications limited to the elastic regime, SR post-processing may be suitable, as long as printing parameters are optimized to prevent lack of fusion defects.

Keywords: AlSi10Mg alloy; laser powder bed fusion; stress relief; hot isostatic pressing; microstructure; fractography; tensile; hardness and fatigue testing

1. Introduction

Some of the first significant applications of aluminum alloys involved aircraft construction. Up to around 1920 aircraft were constructed from wood and fabric. However, in 1920 the German aircraft manufacturer Junkers built the first all metal airplane (Junkers F.13) utilizing Duralumin, an Al-4% Cu, 1% (Mn + Mg) developed in Germany in 1909 [1]. This alloy was the precursor to Al 2024 alloy with ~ 4.4% Cu and <1% (Mn + Mg) used in more modern aircraft manufacture in the US [1]. This alloy along with most other Al alloy compositions, allowed for variations in design strategies in the manufacture of aircraft and many other products because of heat treatments



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and aging schedules which promote various sizes, morphologies, distributions and compositions of precipitates which variously contribute to precipitation hardening and alloy strengthening [2–4].

Because of the light weight, good strength-to-weight ratio and corrosion resistance of aluminum and its alloys, commercial applications in aerospace, automotive, marine and other industrial areas flourished by the middle of the twentieth century. This included the proliferation of various casting processes to produce a wide range of industrial components and structures [5]. Al-Si alloys, especially AlSi10Mg emerged as a preferred, castable alloy for a wide range of component products including heat exchangers, gear box designs, complex fasteners, marine pumps and impeller designs, among many others [5].

Around the start of the current century, additive manufacturing (AM) involving electron and laser beam fusion of prealloyed powders rapidly emerged as a means to fabricate even more complex component and structure geometries beyond the capabilities of castings. Consequently, AlSi10Mg became a preferred candidate for AM, especially laser powder bed fusion (LPBF) processing because of its availability in powder form and its excellent printability [6].

The heat treatment of hyper-eutectic AlSi10Mg alloy primarily involved the production and control of Si precipitate particle sizes, morphology and distributions; including aging treatments which variously contributed to changes in hardness, tensile strengths and ductility quantified by residual elongations or strain-to-failure. Beder et al. [7] have shown that significant increases in the strength of cast AlSi10Mg are achieved by aging treatments. Correspondingly, considerable studies examining the mechanical properties of heat treated LPBF-fabricated AlSi10Mg components have demonstrated similar findings, although specific schedules for stress relief, HIP, and numerous solutionizing treatments at various temperatures and treatment times have shown a wide variation in tensile strengths, ductility, hardness, and fatigue behavior [8–19].

While the vast majority of heat-treatment studies of LPBF-fabricated AlSi10Mg components have involved aging temperatures < 200 °C, and treatment times < 10 h [8–19], Merino et al. [20] recently completed a comprehensive study of heat treatment and aging schedules for LPBF-fabricated AlSi10Mg up to 1000 h. The effects of these comprehensive heat treatment and aging schedules on the tensile properties and hardness were examined by comparing optical metallography with the corresponding tensile and hardness testing results.

The research results presented in the present study represent an extension of the work by Merino et al. [20] by examining the effects of stress relief annealing, HIP and HIP + T6 treatments and final aging at 10, 100, and 1000 h on the uniaxial fatigue and fatigue fracture behavior of LPBF-fabricated AlSi10Mg alloy. To our knowledge, this is the first study to correlate prolonged thermal aging of up to 1000 h across SR, HIP, and HIP + T6 treatments with fatigue life and fracture morphology in this alloy, providing design guidance for LPBF AlSi10Mg components used at elevated temperatures.

2. Materials and Methods

2.1. Material Feedstock and LPBF System

Gas-atomized AlSi10Mg powder (Valimet AM 103C grade) was used for specimen fabrication. This powder was analyzed by Dynamic Image Analysis using a Retech Camsizer X2, manufactured by Microtec MRB (Verder Scientific Group), Haan, Germany, to measure powder morphology and particle size. Its particle size distribution revealed D10, D50, and D90 values of 25 µm, 39 µm, and 59 µm, respectively. The powder exhibited a spherical morphology, which is known to be suitable for the LPBF process to facilitate flowability of powder, as shown in Figure 1.

Specimens were fabricated using an EOS M290 system equipped with a 400 W Yb-fiber laser, an F-theta lens enabling scan speeds up to 7 m/s, a spot diameter of 100 µm, and a build volume of 250 × 250 × 325 mm³. The optimal process parameters were selected as shown in Table 1.

Table 1. Build process parameters.

Laser Power	300 W
Build Plate Preheat Temperature	80 °C
Scanning Speed	1000 mm/s
Hatch Distance	130 µm
Layer Thickness	30 µm
Print Pattern	Striped (7 mm width, 20 µm overlap)

Cylindrical rods (14 mm D × 80 mm H) were printed for Z orientation and rectangular bars (14 mm L × 17 mm W × 76 mm H) for XY orientation, to ensure a balanced representation of directional effects. Rectangular bars were machined to produce six samples, and each rod makes a single sample. For both orientations, the as-built contour surface was removed when machining to the final sample geometry, consisting of a continuous radius

between ends and a gauge diameter of $D = 6.35$ mm. The gauge section was polished longitudinally in accordance with ASTM E466-15. A total of 500 samples were fabricated for the cross-study that included Merino's work [20], from which 144 samples were selected randomly across five different builds for fatigue testing.

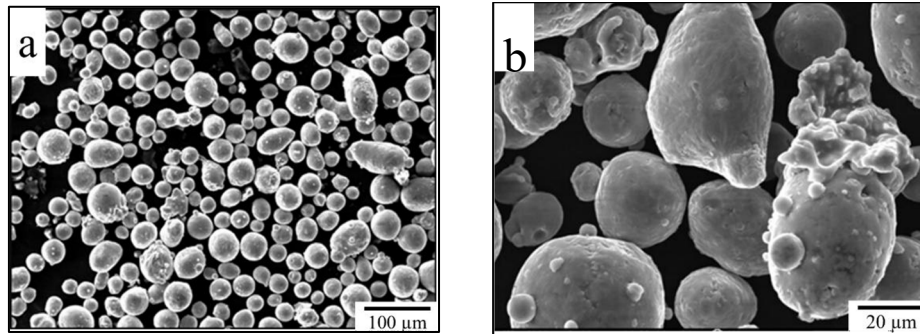


Figure 1. SEM image of powder particles at (a) 200× magnification, and (b) 900× magnification.

2.2. Heat Treatments and Thermal Aging

Equal amounts of specimens were subjected to three heat treatments: stress relief (SR), hot isostatic pressing (HIP), and hot isostatic pressing followed by T6 (HIP + T6). SR samples were heated at 285 °C for a 2 h hold, and air cooled. HIP was performed at 516 °C and 100 MPa for 3 h in an inert argon environment and cooled below 93 °C. In HIP + T6, specimens were HIP processed and followed by solution treatment of 530 °C for a 6 h hold, quenched in water, and aged at 160 °C for 360 min. As shown in Table 2, all specimens were divided into four aging groups (0, 10, 100, 1000 h) aged at 177 °C, a temperature chosen to simulate a common service environment [20]. Temperatures were monitored by external thermocouples and recorded every 30 min to ensure consistency throughout the entire aging duration.

Table 2. Heat treatments and aging conditions.

Variant	Heat Treatment	Aging Time (h)	Condition
1	SR	0	As-SR (unaged)
2	SR	10	SR + 10 h at 177 °C
3	SR	100	SR + 100 h at 177 °C
4	SR	1000	SR + 1000 h at 177 °C
5	HIP	0	As-HIP (unaged)
6	HIP	10	HIP + 10 h at 177 °C
7	HIP	100	HIP + 100 h at 177 °C
8	HIP	1000	HIP + 1000 h at 177 °C
9	HIP + T6	0	As-HIP + T6 (unaged)
10	HIP + T6	10	HIP + T6 + 10 h @ 177 °C
11	HIP + T6	100	HIP + T6 + 100 h @ 177 °C
12	HIP + T6	1000	HIP + T6 + 1000 h @ 177 °C

2.3. Metallographic Preparation and Characterization

Fatigue specimens were sectioned at the unstrained grip section along the Y-Z, X-Z, and X-Y planes and mounted into black epoxy and phenolic powder using a QATM Opal 460 Hot Mounting Press (Mammelzen, Germany) to produce round metallographic samples. The metallographic preparation followed Medrano's procedure [21], beginning with 320 and 800 grit SiC grinding papers at 300 rpm with 25 N force for 1 min with running water. Pre-polishing was performed on a hard synthetic cloth disc with 9 μm diamond suspension at 150 rpm with 25 N for 3:30 min, then intermediate polishing on a soft wool cloth disk with 3 μm diamond suspension at 150 rpm with 20 N for 5 min, and finally fine polishing on a soft synthetic cloth with 0.1 μm fumed silica at 150 rpm with 10 N for 10 min. Microstructure was revealed by immersion in Keller's reagent for 30 s. A Keyence VHK-700, manufactured by Keyence Corp. in Osaka, Japan, and a JEOL, JMS-IT500 EEM, manufactured by JEOL Ltd. in Tokyo, Japan, were used to analyze the microstructure and fracture surfaces.

2.4. Density Measurement & Hardness Testing

Density measurements and hardness values were taken from Merino's results on mechanical properties [20]. Using ImageJ 1.8.0 software, relative density was estimated from cross-sectional metallography by quantifying percentage of area occupied by pores in the analyzed area.

2.5. Fatigue Testing

Fatigue testing was conducted in accordance with ASTM E466-15. Two stress levels were selected for fatigue testing at each orientation and aging condition: (1) within the elastic regime and (2) in the plastic regime, beyond plastic limit. Table 3, contains yield strength and UTS values from Merino's study [20] for each heat treatment and aging condition. Under each stress level, three specimens with the same printing orientation and the same aging time were tested. Tests were performed at room temperature under a fatigue rate (sinusoidal waveform) at 50 Hz and $R = 0.1$ until failure or 5 million cycles (runout).

Table 3. Mechanical properties of heat treated AlSi10Mg [20].

	Variant	YS (MPa)	UTS (MPa)	ϵ (% Elong)
As Built	-	237 ± 4	463 ± 3	7.6 ± 1.0
SR	1	153 ± 8	249 ± 10	21.3 ± 1.7
SR	2	154 ± 8	246 ± 9	21.6 ± 1.8
SR	3	174 ± 3	271 ± 4	16.5 ± 1.2
SR	4	155 ± 2	245 ± 5	14.8 ± 2.0
HIP	5	88 ± 4	144 ± 1	32.2 ± 1.2
HIP	6	93 ± 2	144 ± 1	31.8 ± 1.0
HIP	7	93 ± 1	138 ± 1	30.5 ± 3.1
HIP	8	80 ± 1	127 ± 1	33.3 ± 1.3
HIP + T6	9	240 ± 8	305 ± 8	16.2 ± 1.4
HIP + T6	10	232 ± 7	283 ± 10	14.5 ± 1.5
HIP + T6	11	159 ± 3	210 ± 2	16.2 ± 0.6
HIP + T6	12	94 ± 4	144 ± 6	28.9 ± 2.7

3. Results and Discussion

3.1. Microstructure

SR samples (Figure 2) show typical eutectic Si network of ultra fine cell structures on the α -Al matrix, along with clear delineation of melt pool boundaries highlighted by the red stitched lines.

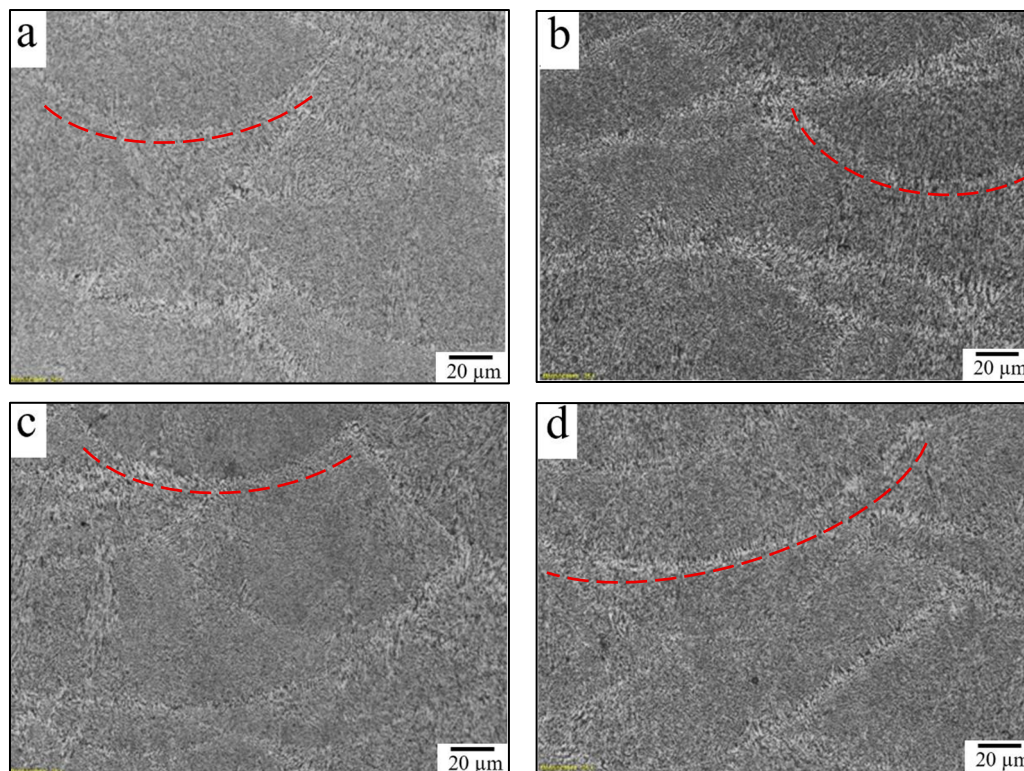


Figure 2. Microstructure of SR AlSi10Mg of (a) 0, (b) 10, (c) 100, and (d) 1000 h at 500 $\times$ magnification. Examples of melt pool boundaries are highlighted by the red stitched lines.

Prolonged thermal exposure does not seem to significantly affect SR microstructure at the micron scale, as no major differences were observed with subsequent aging. This stability can be attributed to the fact that decomposition of the Si network does not occur until temperatures close to 300 °C [22–25]. However, Zhao [22] reports that the decrease in strength and increase in ductility from the as built to the SR condition can be explained by the precipitate coarsening within the Al cells, which reduces the hardening capacity. This mechanism helps to explain the results of Merino [20] in Table 3.

Microstructure of HIP samples (Figure 3) is characterized by a uniform distribution of larger equiaxed silicon particles on the aluminum matrix. In contrast to SR microstructure, melt pool boundaries are no longer present. When comparing the HIP unaged condition (Figure 3a) to the 1000 h aged (Figure 3d), Si particles appear coarser and interparticle spacing is wider in the aged condition, while the unaged condition shows finer and more densely distributed particles.

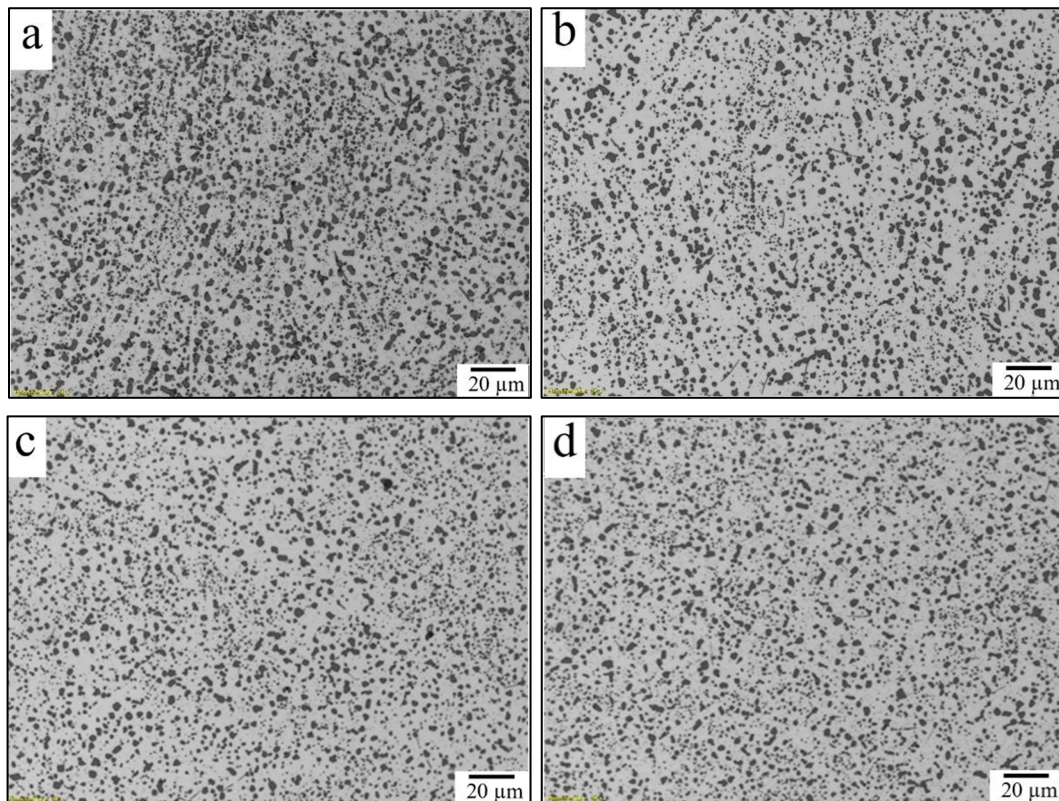


Figure 3. Microstructure of HIP AlSi10Mg with aging time of (a) 0, (b) 10, (c) 100, and (d) 1000 h at 500× magnification.

All HIP + T6 variations (Figure 4) presented slightly coarser silicon particles and larger interparticle spacings compared to HIP alone (Figure 3). However, aging had opposite effects on particle size for the two conditions: HIP samples showed coarser particles and wider interparticle spacing at 1000 h compared to the unaged condition, whereas HIP + T6 samples showed the opposite trend, with the 1000-h aged condition containing finer, more evenly distributed particles than the unaged condition. Normally, this refinement of particles would be expected to increase strength by precipitation, but Merino's UTS values in Table 3 show the opposite trend [20], with the 1000 h aged HIP and HIP + T6 conditions being weaker than the unaged state. For both HIP and HIP + T6, the decline in strength (Table 3) and hardness (Table 4) with aging may be attributed to coarsening of finer, nanoscale hardening precipitates within the α -Al matrix through an overaging mechanism. As these nanoscale precipitates are below the resolution of the microstructural imaging performed in this study, this mechanism is proposed based on aging behavior in Al-Si-Mg alloys, and further characterization (TEM) would be needed to confirm contributions of precipitate overaging to mechanical behavior.

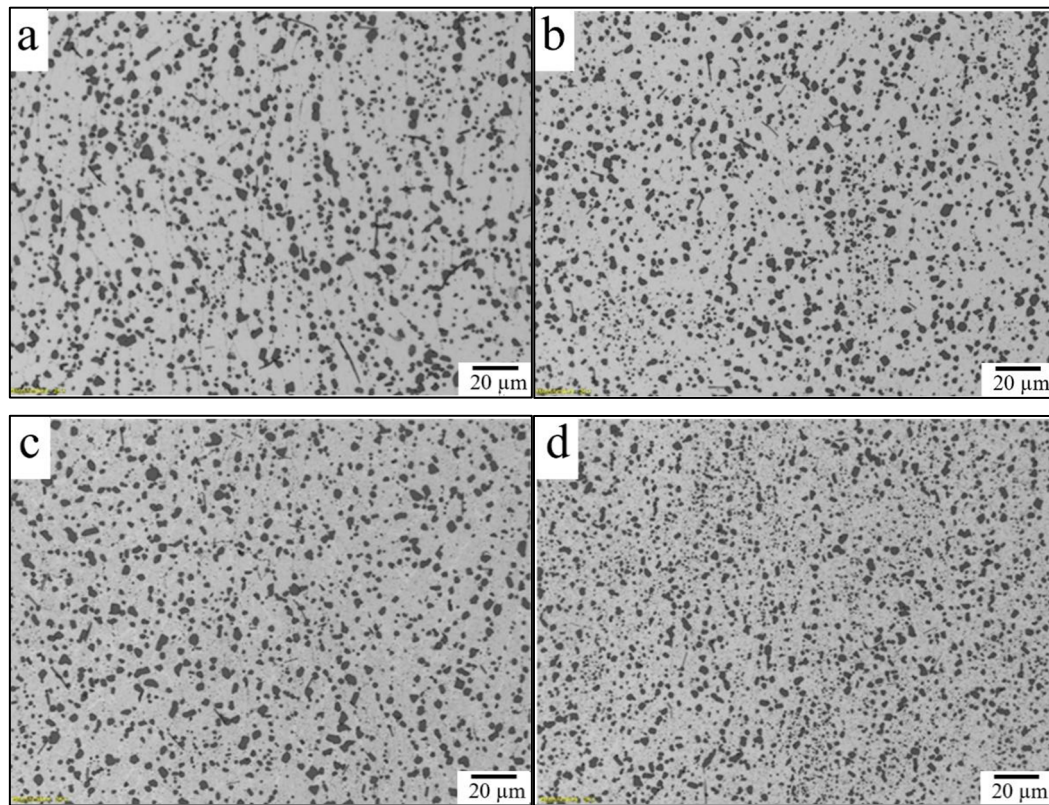


Figure 4. Microstructure of HIP + T6 AlSi10Mg of (a) 0, (b) 10, (c) 100, and (d) 1000 h at 500× magnification.

Table 4. Material properties of AlSi10Mg under heat treatment and aging time at 177 °C.

XY-Build Direction					Z-Build Direction				
Heat Treatment	Aging (h)	Hardness (HV)	Density (g/cm ³)	Relative Density (%)	Heat Treatment	Aging (h)	Hardness (HV)	Density (g/cm ³)	Relative Density (%)
SR	0	93	2.65	99.13	SR	0	89	2.65	99.19
	10	95	2.65	99.08		10	98	2.65	99.15
	100	94	2.65	99.06		100	105	2.65	99.19
	1000	86	2.65	99.07		1000	84	2.65	99.15
HIP	0	60	2.65	99.25	HIP	0	59	2.65	99.29
	10	59	2.65	99.28		10	60	2.66	99.5
	100	49	2.65	99.29		100	60	2.66	99.49
	1000	45	2.65	99.33		1000	47	2.65	99.25
HIP + T6	0	106	2.65	99.34	HIP + T6	0	121	2.66	99.44
	10	110	2.65	99.27		10	110	2.65	99.33
	100	73	2.65	99.18		100	81	2.65	99.31
	1000	59	2.65	99.19		1000	58	2.65	99.25

3.2. Density and Hardness

The pycnometer density measurements (Table 4) revealed a highly consistent density of 2.65–2.66 g/cm³ across all heat treated groups. However, a more notable difference emerged in relative density values, where the HIP and HIP + T6 groups exhibited higher values, proving to effectively reduce porosity compared to the SR samples. Hardness of HIP samples decreased with aging, starting from ~60 HV in unaged samples to ~45 HV after 1000 h of aging. HIP + T6 unaged samples in the Z direction presented the highest hardness out of all variables at 121 HV and followed the same trend of decrease in hardness as HIP samples, starting from 106 HV–121 HV and decreasing to ~59 HV at 1000 h. The overall reduction in hardness with longer aging duration in both HIP groups correlates with the observed decrease in YS and UTS, and is consistent with previous studies [17,21]. SR samples maintained an intermediate hardness, beginning from 89 HV–93 HV and similarly decreasing to ~84 HV after 1000 h of aging. However, unlike the HIP groups, the peak in hardness for SR is not in the unaged condition but rather in the 10–100 h aging range, with 105 HV at 100 h in the Z direction. Similarly, Zhou [26] reported that artificially aged samples exhibited peak hardness at around 10 h.

3.3. SR Fatigue Behavior

Table 5 presents the testing results of the SR-treated samples. At the elastic stress testing of 124 MPa, unaged samples reached the 5 million cycle limit for both XY and Z orientations. In 10 h aged samples, while XY samples reached 5 million cycles, Z samples showed more scatter, with two runoffs and one failing at 0.92 million cycles. At 100 h aging, XY samples still reached 5 million cycles, while Z samples showed two failures, averaging 2.3 million cycles. Such variability in Z builds can be attributed to the higher sensitivity of vertical orientations to defect alignment and melt pool morphology [27]. For 1000 h aged SR samples, both XY and Z samples reached runoff.

Table 5. Fatigue properties of SR treated AlSi10Mg measured by number of load cycles to failure (in millions).

Test Matrix of AlSi10Mg SR Samples		Thermally Aged at 177 °C							
		0 h		10 h		100 h		1000 h	
		XY	Z	XY	Z	XY	Z	XY	Z
Elastic Regime	124 MPa	5	5	5	5	5	0.59	**	5
		5	5	0.92	5	5	5	5	5
		5	5	5	5	5	1.21	5	5
	AVG	5	5	3.6	5	5	2.3	5	5
Plastic Regime	193 MPa	0.21	0.15	0.15	0.05	0.38	0.28	0.16	0.15
		0.14	0.14	0.21	0.04	0.3	0.11	0.13	0.09
		0.16	0.21	0.26	0.09	0.27	0.08	0.17	0.08
	AVG	0.17	0.17	0.21	0.06	0.32	0.16	0.15	0.11

** Specimen failed at grip section.

In the plastic stress testing of 193 MPa, unaged SR samples averaged 0.17 million cycles for both XY and Z orientations. Fatigue life showed some variation with aging, with 100 h aged samples averaging 0.32 million cycles for XY and 0.16 million cycles for Z, and 1000 h aged samples averaging 0.15 million cycles for XY and 0.11 million cycles for Z.

The sudden failure region in Figure 5b,c is very representative of the Z printing orientation compared to the XY orientation (Figure 5a), where multiple melt pool boundaries are visible on the fracture surface. These features, indicated by the red arrows in Figure 5b,c, form a distinct “zig-zag” or “weave” pattern. However, these features do not appear to significantly influence fatigue life, consistent with the results presented here and previous studies [14]. Multiple defects are visible on the fracture surfaces of the samples in Figures 5 and 6. For example, lack of fusion defects are enclosed by the red stitched lines in Figure 5b, while unmelted particles are indicated by yellow arrows in Figures 5a–c and 6. The effect of these defects is discussed below.

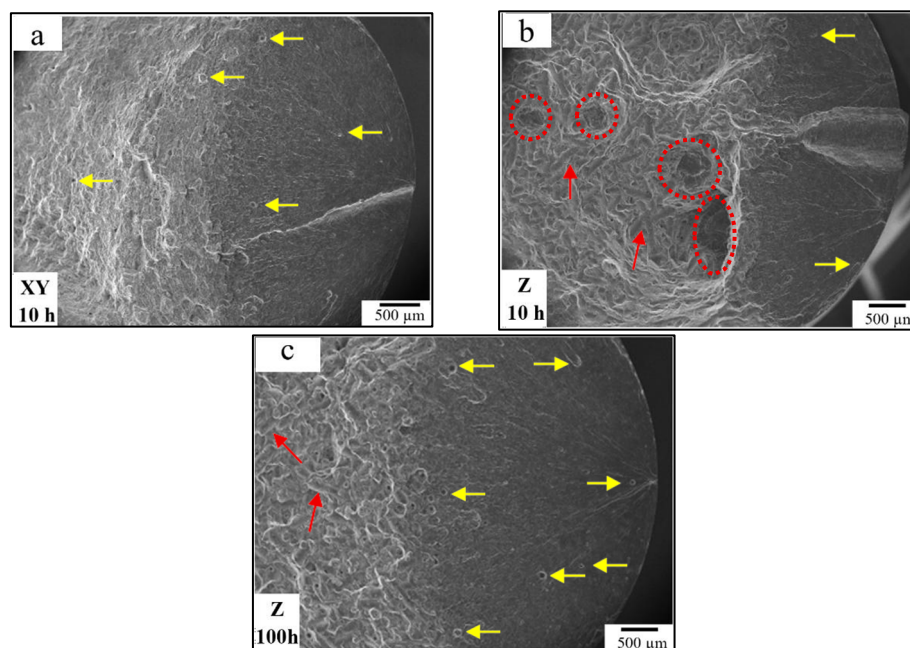


Figure 5. SEM fractographs of SR treated and aged AlSi10Mg samples after fatigue testing in the plastic regime: (a) XY sample aged for 10 h, (b) Z sample aged for 10 h, (c) Z sample aged for 100 h. Red arrows indicate melt pool boundaries, red stitched lines highlight lack-of-fusion pores, and yellow arrows mark unmelted particles.

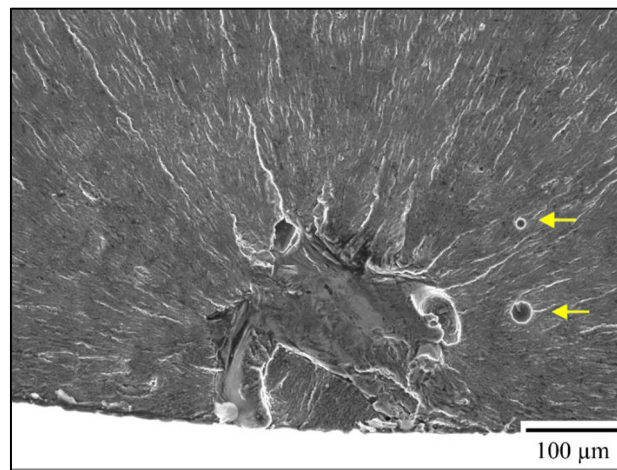


Figure 6. Defect evaluation at crack initiation site in an SR treated sample tested in the elastic regime, showing a lack of fusion pore. Unmelted particles marked by yellow arrows.

3.4. SR Defect Evaluation

It was determined that the premature failure of SR samples in the elastic regime could be directly attributed to process-induced porosity. SEM examination revealed these to be lack of fusion pores, ranging in size from 100–200 μm (Figure 6). These internal defects can act as critical stress concentrators under cyclic loading conditions and their presence provides preferential sites for crack initiation, even within the nominal elastic stress range. This explains the premature fatigue failures observed in some SR samples in the elastic regime despite other SR samples reaching the 5 million cycle limit. In addition to these stress concentrators, multiple unmelted particles were present on the fracture surfaces of the SR samples (Figures 5 and 6), marked by the yellow arrows. As noted by Cáceres [28], even low levels of porosity can alter deformation mechanisms by providing easy paths for fracture, significantly reducing both tensile ductility and tensile strength. Multiple comparable findings have been reported in previous fatigue studies of AlSi10Mg and other additive manufacturing alloys [29–31].

3.5. HIP Fatigue Behavior

Table 6 presents the testing results of the HIP-treated samples. At 62 MPa on the elastic regime testing, all samples, regardless of aging or orientation, reached the 5 million cycle limit without failure. This result demonstrates robust fatigue performance under elastic stress, consistent with literature reports [17].

Table 6. Fatigue properties of HIP treated AlSi10Mg measured by number of load cycles to failure (in millions).

Test Matrix of AlSi10Mg HIP Samples		Thermally Aged at 177 °C							
		0 h		10 h		100 h		1000 h	
		XY	Z	XY	Z	XY	Z	XY	Z
Elastic Regime	62 MPa	5	5	5	5	5	5	5	5
		5	5	5	5	5	5	5	5
		5	5	5	5	5	5	5	5
	AVG	5	5	5	5	5	5	5	5
Plastic Regime	124MPa	1.22	1.21	0.77	0.72	0.73	0.63	0.02	0.01
		1.19	1.25	0.9	0.82	0.53	0.59	0.02	0.01
		1.53	**	0.75	1.06	0.49	0.68	0.02	0.02
	AVG	1.31	1.23	0.81	0.87	0.58	0.63	0.02	0.01

** Specimen not considered.

At 124 MPa (plastic regime), aging significantly reduced fatigue life. Unaged XY specimens averaged 1.31 million cycles, while 1000 h aged samples failed around 22,000 cycles. Z orientation specimens showed a similar trend, with fatigue life decreasing from 1.23 million to just 12,000 cycles. These results show that as few as 10 h of aging at 177 °C can reduce fatigue life by up to 38%. It was also observed that no obvious difference in the number of cycles can be observed between printing orientations, for example, the specimens aged for 100 h, either Z or XY fractured close to the 600,000 cycles.

Figure 7 presents SEM fractographs of HIP treated Z and XY specimens. A notable reduction in fracture cross sections is evident with longer aging, observed in unaged samples (Figure 7a,b) to the 1000 h aged samples

(Figure 7g,h). Unaged, 10 h, and 100 h aged samples (Figure 7a–f) show two primary fracture textures: a smooth texture with concentric lines indicating crack propagation (inside red stitched arcs), and a rough surface from rapid rupture. These distinct textures diminish in specimens subjected to longer aging, where increased ductility allows for greater deformation and a more uniform stress distribution across the cross section, resulting in a more uniform fracture texture (Figure 7g,h). The smooth (crack propagation) and rough (sudden fracture) textures are present in the 0 to 100 h specimens (Figure 7e,f), but not in the 1000 h, combined with a decrease in cross sectional area, this is clear evidence of increased ductility. Increased ductility correlated with thermal exposure time and reduced fatigue life, suggesting a trade-off between toughness and fatigue resistance [21].

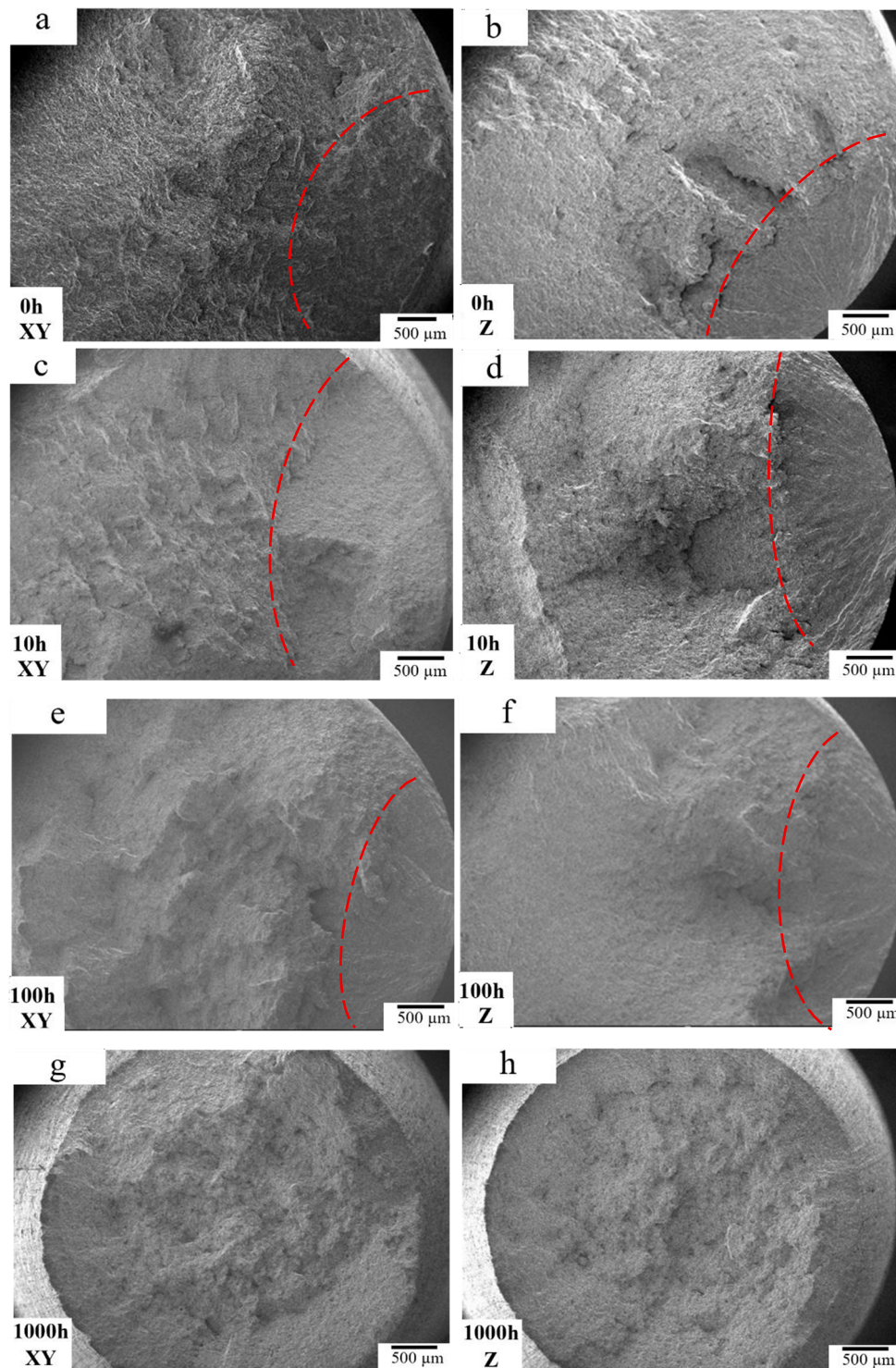


Figure 7. SEM fractographs of HIP treated and aged AlSi10Mg samples (XY and Z) after fatigue testing in the plastic regime: (a) XY built, 0 h aged; (b) Z built, 0 h aged; (c) XY built, 10 h aged; (d) Z built, 10 h aged; (e) XY built, 100 h aged; (f) Z built, 100 h aged; (g) XY built, 1000 h aged; (h) Z built, 1000 h aged. Red stitched arcs indicate the boundary of the crack propagation region.

3.6. HIP + T6 Fatigue Behavior

Table 7 presents the testing results of the HIP + T6-treated samples. At the elastic stress testing of 145 MPa, all samples from 0 to 100 h aging, and across both XY and Z orientations, successfully reached the 5 million cycle limit without failure. However, at 1000 h of aging, one XY sample reached 0.25 million cycles, and Z samples averaged a significantly lower life of 0.08 million cycles. In the plastic stress testing of 193 MPa, the unaged HIP + T6 samples had scattered fatigue lives, with XY averaging 0.76 million cycles and Z averaging 3.49 million cycles due to 2 samples being runoffs. With increased aging, the fatigue life significantly decreased following the same trend as HIP, with samples at 100 h aging showing an average of 0.19 million cycles for both XY and Z orientations.

Table 7. Fatigue properties of HIP + T6 treated AlSi10Mg measured by number of load cycles to failure (in millions).

Test Matrix of AlSi10Mg HIP + T6 Samples		Thermally Aged at 177 °C							
		0 h		10 h		100 h		1000 h	
		XY	Z	XY	Z	XY	Z	XY	Z
Elastic Regime	145 MPa	5	5	5	5	5	5	0.2	0.24
		5	5	5	5	5	5	0.25	0.002
		5	5	5	5	5	5	0.01	0.004
	AVG	5	5	5	5	5	5	0.15	0.08
Plastic Regime	193 MPa	1.8	5	3.2	0.4	0.25	0.09	*	*
		0.26	5	0.24	0.11	0.19	0.23	*	*
		0.21	0.47	0.92	5	0.13	0.26	*	*
	AVG	0.76	3.49	1.45	1.84	0.19	0.19	0	0

* Specimen fractured before reaching the maximum test load.

The 1000 h aged samples that failed in the elastic regime testing are presented in Figure 8a,b. Their fracture surfaces show no crack propagation and exhibit a high reduction in cross-sectional area, indicating an increase in ductility similar to the trend seen in HIP only samples. While HIP + T6 initially outperforms HIP in the plastic regime, consistent with reports of T6 fatigue benefits [14,17,30,32], this advantage diminishes with aging. By 100 h aging, HIP + T6 performance falls below HIP, and both conditions severely underperform after 1000 h of aging. This suggests that the precipitation hardening effect of T6 is outweighed by matrix softening at prolonged aging times.

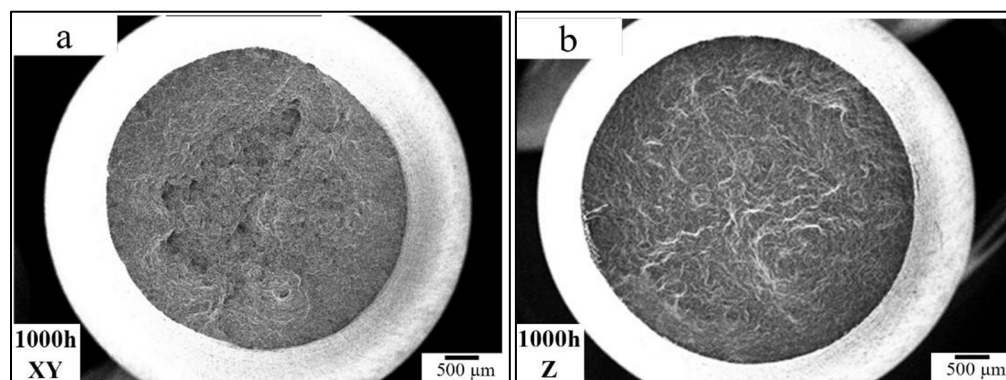


Figure 8. SEM fractographs of HIP + T6 treated and aged AlSi10Mg samples (XY and Z) after fatigue testing in the elastic regime. Under these stress conditions, failure occurred only in samples aged for 1000 h: (a) XY built, aged 1000 h, and (b) Z built, aged 1000 h.

These findings point to mechanisms that together, help explain the fatigue behavior observed across conditions and aging times (Figure 9). HIP treatment reduces the impact of lack-of-fusion porosity, removing dominant crack initiation sites responsible for premature failure of SR samples. This explains the improvement in fatigue life for HIP and HIP + T6 (up to 100 h) relative to SR in the elastic regime (Figure 9c,d), where fatigue life is sensitive to defects. At the plastic stress level (Figure 9a,b), the stresses applied exceed yield strength and fatigue resistance is more dependent on the matrix resistance to cyclic plastic strain. As discussed previously, the decline in strength and hardness with aging may be attributed to overaging of nanoscale hardening precipitates within the α -Al matrix. Softening of the matrix reduces the resistance to cyclic plastic deformation, providing an explanation for the fatigue life degradation observed in HIP and HIP + T6 samples at the plastic stress level.

While HIP + T6 samples show a difference in fatigue life between XY and Z orientations at 0 h of aging, this difference is not observed in HIP only samples, despite both undergoing homogenization that eliminates melt pool

boundaries. This suggests that build orientation difference in results from HIP + T6 is more likely attributed to data scatter, given the limited number of samples per group.

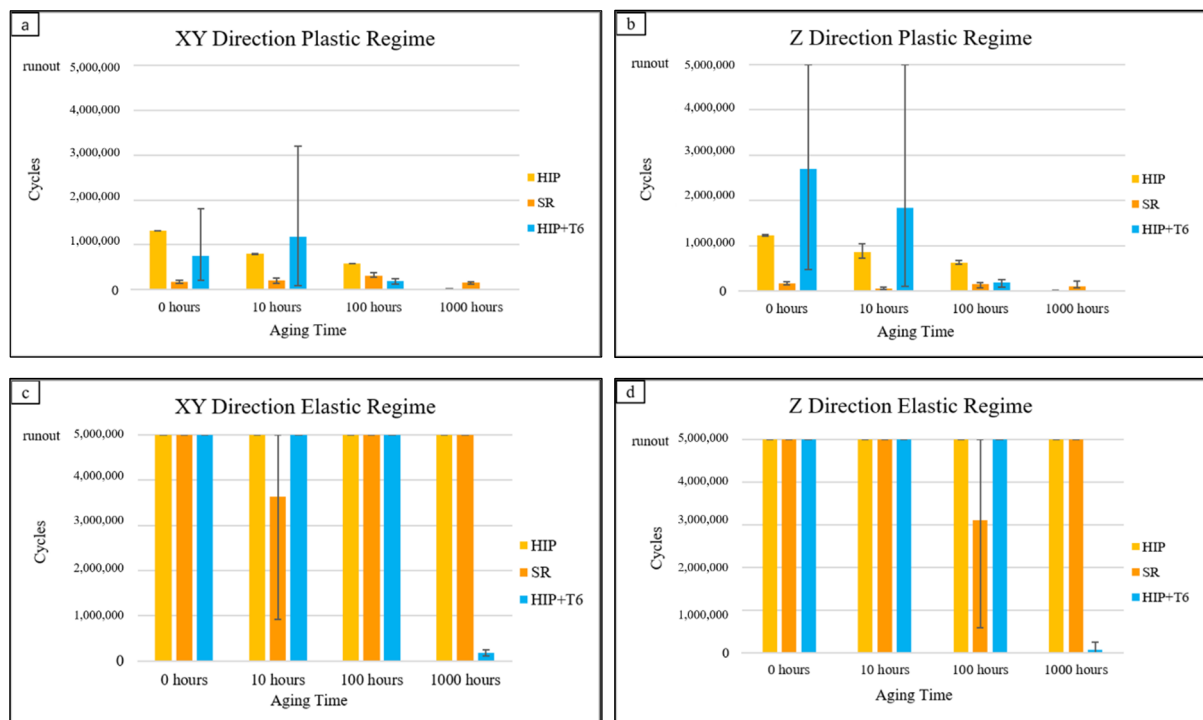


Figure 9. Fatigue life (cycles) as a function of aging time (0, 10, 100, and 1000 h) for SR, HIP, and HIP + T6 treated AlSi10Mg samples in (a) XY direction, plastic regime; (b) Z direction, plastic regime; (c) XY direction, elastic regime; and (d) Z direction, elastic regime. Bars represent the average of three specimens tested per condition, with error bars indicating the maximum and minimum values.

3.7. Fractography and Ductility Trends

The SEM imaging of the fracture surfaces in Figure 10 revealed differences in fracture morphology among the heat treatment conditions and aging times. All SR samples, presented a predominantly quasi-brittle behavior, characterized by flat facets and tear ridge patterns, with only minimal evidence of shallow dimples. In contrast, HIP samples in the unaged condition, showed a classical ductile fracture with deep and defined dimple features, indicating a highly ductile fracture mode. After 1000 h of aging, HIP samples continued to display a ductile fracture surface with even deeper dimple features and a more uniform fracture pattern compared to the 0 h samples, suggesting 1000 h aged samples sustained ductility during the prolonged thermal exposure. Similarly, HIP + T6 unaged samples also displayed well-defined dimple features comparable to the HIP only samples, confirming ductile fracture mode. However, after 1000 h of aging, HIP + T6 samples showed noticeably shallower dimples and regions with rough and uneven patterns. This change in fracture surface morphology may be correlated to the underperformance of HIP + T6 samples in the elastic regime fatigue testing after prolonged thermal exposure.

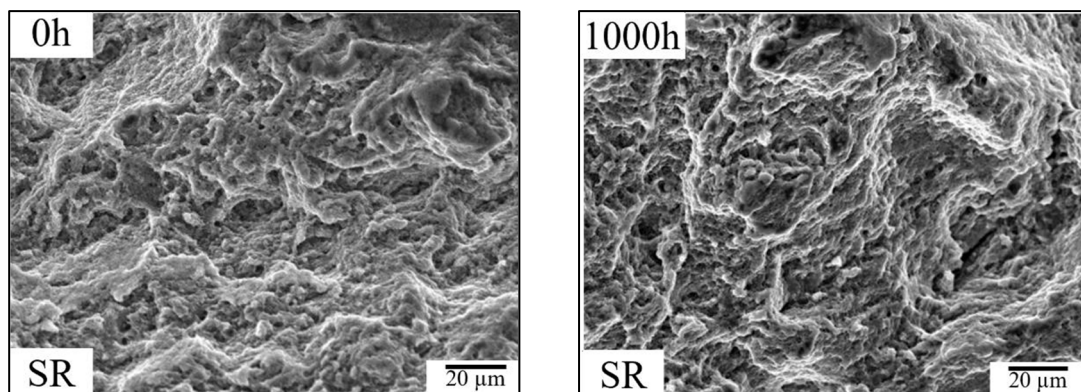


Figure 10. Cont.

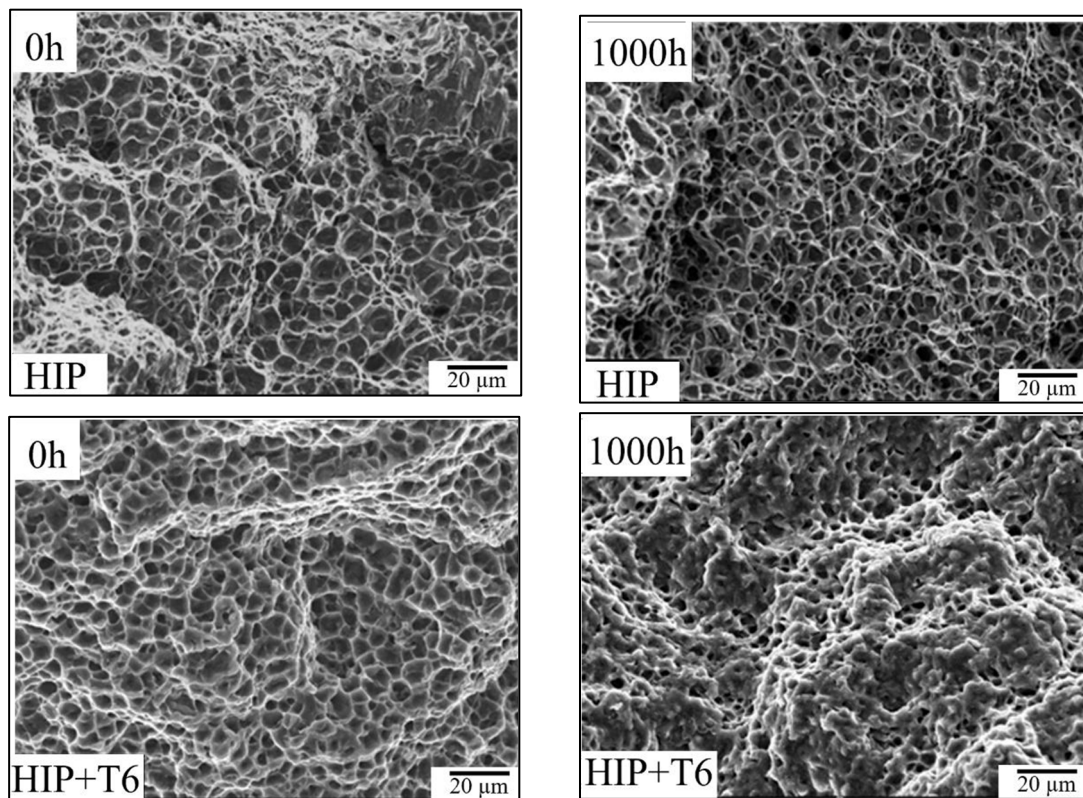


Figure 10. SEM fractographs showing the morphology and dimple presence within the sudden fracture region across heat treatments and aging times.

4. Conclusions

This investigation into the fatigue behavior of LPBF AlSi10Mg alloy under thermal aging revealed distinct performance characteristics for different post-processing selections.

The microstructural analysis showed that SR samples retained the fine, cellular eutectic Si network, while HIP and HIP + T6 samples exhibited a uniform distribution of larger, equiaxed Si particles and refinement of such particles with prolonged aging. Density measurements confirmed that HIP and HIP + T6 treatments were effective in reducing porosity, helping in fatigue performance. Hardness generally decreased with prolonged aging across all conditions, though SR samples showed a peak hardness within the 10–100 h aging range.

The premature failure of SR samples in the elastic regime was directly linked to process-induced porosity, specifically lack-of-fusion pores. Both HIP and HIP + T6 treatments effectively mitigated these defects, leading to improvement in fatigue performance. However, samples treated with HIP showed a fatigue life reduction of up to 38% after just 10 h of aging, indicating a high sensitivity to thermal exposure. We did not find a clear outperformer between printing orientations in terms of fatigue life.

Fracture analysis showed that unaged SR samples failed in a more brittle manner, characterized by tear ridges, while HIP and HIP + T6 samples exhibited a ductile fracture with prominent dimple features. Prolonged aging had different effects on the two groups: HIP only samples became more ductile over time, with deeper dimple features, while HIP + T6 samples transitioned to a mixed fracture mode after 1000 h, as ductile dimple features were replaced by rough, uneven patterns. This shift in fracture morphology directly correlates with the decreased fatigue performance of aged HIP + T6 samples.

In conclusion, the results presented in this paper demonstrate that HIP treatment provides robust fatigue performance in the elastic regime, but both HIP and HIP + T6 groups are highly susceptible to fatigue degradation at service temperatures. While the HIP + T6 treatment provides strong initial properties, its performance declines with prolonged aging, suggesting that under the specific conditions tested in this study, fatigue performance at 177 °C degrades significantly beyond 100 h of aging. This should be considered as a preliminary indication given the limited number of specimens tested per condition. Future work is proposed to include a 500 h aging condition to analyze degradation in the 100–1000 h range, and CT scanning prior to testing to characterize defects that could influence fatigue performance. SR may be suitable for elastic regime applications, provided that printing parameters are carefully optimized to avoid lack-of-fusion defects.

Author Contributions

L.A.M.: Original draft, conception, writing, review and editing; E.A.: Writing, editing; L.E.M.: Draft revision and writing, review, editing; D.A.A.: Writing, editing; M.B.: Provision of experimental samples, data, editing; J.T.: Editing, review; J.C.: Review, editing; F.M.: Supervision, conceptualization, review, editing. All authors have read and agreed to the published version of the manuscript.

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Conflicts of Interest

The authors declare no conflict of interest.

Use of AI and AI-Assisted Technologies

No AI tools were used for this paper.

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