



Build-Height Influence on Electrical Conductivity and Mechanical Properties of LPBF-Manufactured AlSi7Mg Alloy: Role of Networked and Composite-Like Microstructures

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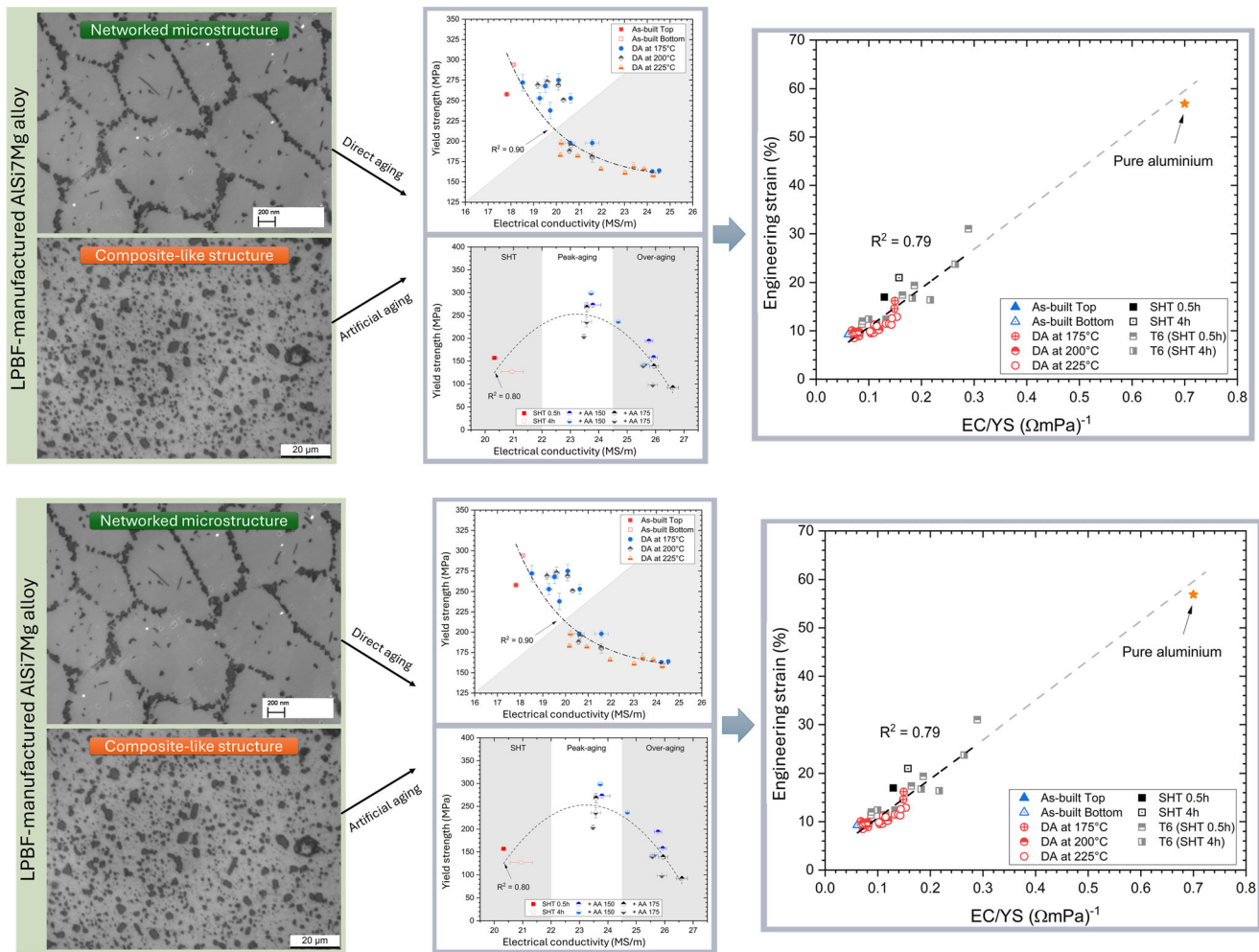
This study presents the variations in electrical conductivity in 300 mm-tall AlSi7Mg samples produced by laser powder bed fusion, a height greater than that typically examined in previous studies. Electrical conductivity was correlated with the microstructure and mechanical properties in both as-built and heat-treated specimens. Two-dimensional image analysis confirmed a density of ~ 99 pct in the as-built state with pore diameters ranging from 2 to 15 μm , which shifted slightly after exposure to 505 °C/4 h (resulting in a density of 97.7 pct). The limited porosity exhibited negligible impact on electrical conductivity. In the as-built state, an electrical conductivity gradient was developed along the build height due to the precipitation and sub-grain structures present in the bottom regions, compared to a supersaturated solid solution presents in the top regions. Direct aging at temperatures between 175 and 225 °C enhanced the electrical conductivity by 20 to 35 pct, primarily due to precipitation phenomena and coarsening of the Si-eutectic network (especially at temperatures above 200 °C). Following solution treatment, the Si-decorated α -Al cells transformed into a composite-like structure in which Si particles were embedded in the Al matrix, leading to 27 to 31 pct increase in electrical conductivity and an 8 to 15 pct increase in strength. A proportional relationship between the electrical conductivity-to-yield strength ratio and elongation was also identified, emphasizing the strong dependence between microstructure, electrical conductivity, and mechanical behavior.

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I. INTRODUCTION

LASER powder bed fusion (LPBF) is an additive manufacturing process that produces metal components through a layer-by-layer method. This technique enables the manufacturing of physical objects with complex geometries and shapes very close to the desired final dimensions.^[1] Additionally, the capability to build lattice structures within components results in weight reduction, which is further enhanced when using light-weight alloys such as AlSi7Mg. For these reasons, the AlSi7Mg alloy can be fabricated by the LPBF process to manufacture, *e.g.*, electrical vehicle battery housings with complex geometries.

AlSi7Mg is an age-hardenable Al alloy widely used to additively manufacture components due to its low shrinkage and short solidification temperature range.^[2] These features result in a short liquid lifetime, leading to a very fine microstructure and improved strength compared to cast Al alloys.^[3] To reduce the anisotropy in both the microstructure and the mechanical properties, or to improve strength, ductility, and electrical conductivity of built components, they are commonly subjected to various industrial heat treatments.^[4] A common practice involves stress relief around 325 °C to reduce the residual stresses developed during the LPBF process, as also suggested by the metal powder suppliers. However, exposure to these annealing temperatures may reduce the degree of anisotropy in the mechanical behavior at the expense of strength.^[5,6] To avoid

material softening and to preserve the networked microstructure of LPBF Al–Si–Mg samples, components can be manufactured on a build platform pre-heated to temperatures below 200 °C.^[7] This is *in situ* heat-treatment that increases strength due to the supersaturated solid solution (SSS) formed as a result of very rapid melt pool solidification. This SSS leads to the precipitation of β'' and semi-coherent β' precipitates,^[8,9] which strengthen the α -Al matrix *via* the cutting and Orowan mechanisms.^[10] These conditions preserve the LPBF-manufactured microstructure formed by α -Al cells decorated with fine Si-eutectic particles, which form an interconnected network.

Another approach to reducing residual stresses and improving the strength-ductility ratio is applying T6 heat treatment.^[6,11–15] The conventional T6 heat-treatment, which involves solution and aging heat treatments, generally increases ductility but decreases strength compared to the T5 condition when applied to the same LPBF Al–Si–Mg alloys.^[16] Several studies^[4,6,8] have reported that the T6 heat-treatment results in lower strength-to-ductility ratio values compared to the direct-aged condition. Specifically, Padovano *et al.*^[6] reported values of 28 to 36 for Al–Si–Mg alloys in the T6 condition, while Aversa *et al.*^[8] observed values of 83 to 100 for the T5 condition.

In industrial applications, LPBF Al–Si–Mg alloys are also employed in electrically and thermally conductive applications, such as coils for electrical motors, antenna design, or other electromagnetic devices.^[17–19] Consequently, electrical conductivity (EC) or resistivity is one of the main physical parameters required to be measured. EC is directly influenced by electron scattering at various features in the matrix, such as alloying elements in solid solution, dislocations, grain boundaries and precipitates, as described by Matthiessen’s model.^[20,21] For these reasons, EC varies with the applied heat treatments, which affect the as-built microstructure and consequently the mean free path of free electrons. As a result, EC measurements can also be considered to evaluate the degree of aging and microstructural evolution within the α -Al matrix, as widely supported in.^[22–24]

Cabrera-Correa *et al.*^[21] investigated EC variations between the as-built and stress-relieved LPBF-manufactured Scalmetalloy®, highlighting the effects induced by the microstructural variations in terms of dislocations, precipitates, grain boundaries, and solid solution. Guo *et al.*^[25] analyzed the influence of defects and residual stresses on the EC of LPBF-manufactured Inconel 738 LC. Silbernagel *et al.*^[17] evaluated the electrical resistivity of as-built, annealed, and T6-like heat-treated LPBF AlSi10Mg alloy using a four-wire Kelvin resistance method. Spurek *et al.*^[26] carried out *in situ* EC measurements during the LPBF process of AlSi10Mg alloys. Recently, Marola *et al.*^[27] reported EC variations under three different heat treatment conditions of LPBF-manufactured waveguide components fabricated in AlSi7Mg alloy.

To the best of the authors’ knowledge, the limited number of studies available in the literature that investigate EC have performed only a small set of EC measurements, and the alloys examined were, in only a

few instances, aluminum alloys produced by LPBF, such as Scalmetalloy or AlSi10Mg. Most previous works instead focused on entirely different alloy systems (*e.g.*, steels and Inconel). Notably, only one study to date has examined AlSi7Mg produced by LPBF.

Within this context, the present work addresses several novel aspects that have not been previously investigated:

- (i) A refined analysis of the top/bottom surface differences in EC;
- (ii) An assessment of the correlation between precipitation gradients and EC;
- (iii) A comprehensive evaluation of T5 (175 °C to 225 °C for 0 to 512 h) and T6 (505 °C for 0.5 h followed by artificial aging at 150 °C to 175 °C for 0 to 512 h) heat treatments, with a particular focus on short-term, peak-aged, and long-term conditions to elucidate the coupled evolution of mechanical properties and EC throughout aging;
- (iv) The introduction of a correlation between eutectic silicon morphology, dispersion, and EC;
- (v) The development of quantitative relationships linking mechanical properties, EC, and Vickers hardness (HV); and
- (vi) Most notably, the proposal of a new factor relating the EC/YS ratio to elongation, ϵ (pct), which reveals a linear trend validated using data from other LPBF-manufactured alloys and pure aluminum.

Finally, EC results are correlated with microstructural observations and mechanical properties to identify the optimal heat treatment conditions. The overarching goal is to enable the design of LPBF AlSi7Mg components with tailored microstructural features and property combinations suitable for advanced industrial applications.

II. MATERIALS AND METHODS

Argon gas-atomized AlSi7Mg powder particles, whose chemical composition is listed in Table I, were used to manufacture both bars and billets. Both geometries were manufactured to investigate variations in the microstructure, mechanical properties, and EC in as-built and heat-treated conditions. The powder particles were spherically shaped, with approximately 95 pct of them falling within the range of 20 to 60 μm range. The LPBF process was performed with a SLM@280 machine (Nikon SLM Solutions, Lübeck, Germany)

Table I. Nominal Chemical Composition (Wt Pct) of AlSi7Mg Alloy as Supplied by Manufacturer (Tekna®)

Alloy	Al	Si	Mg	Mn	Fe	Zn	Cu
AlSi7Mg	bal.	7.04	0.59	0.006	0.06	0.011	< 0.005

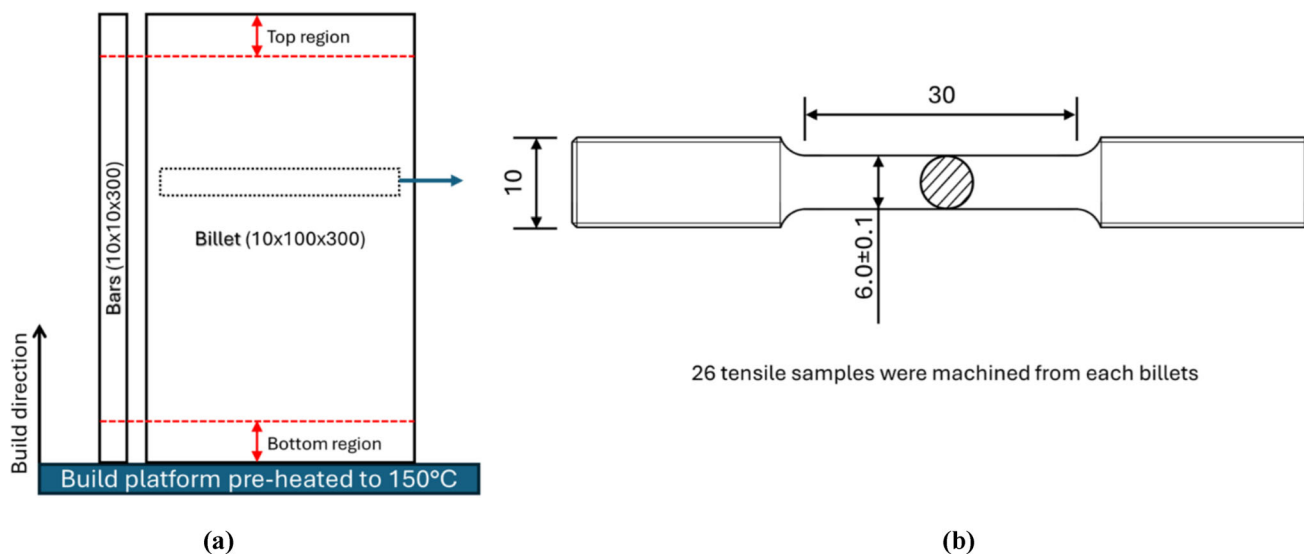


Fig 1—(a) Schematic representation of bars and billets LPBF-manufactured on the pre-heated build platform and (b) technical drawing of a cylindrical tensile sample machined from LPBF-manufactured billets. All dimensions are expressed in mm.

using 370 W of laser power, a scanning speed of 1400 mm/s, a layer thickness of 90 μm , a hatch spacing 70 μm , and a laser beam diameter of 110 μm . These process parameters had been previously optimized by the materials supplier. Samples were built on a platform pre-heated to 150 °C and maintained at this temperature throughout the LPBF process.

Bars measuring 10 × 10 × 300 mm³ were printed to evaluate the effects of the pre-heated build platform along the build direction [Figure 1(a)] on Vickers microhardness and EC. Another set of billets (10 × 100 × 300 mm³) was also printed to evaluate mechanical properties. For each billet, cylindrical tensile specimens, having dimensions dictated by the ASTM E8/E8M standard [Figure 1(b)], were machined horizontally with respect to the build platform [see Figure 1(b)] and labeled as “bottom” and “top” samples. “Bottom” and “top” refer to the regions closest and farthest from the build platform pre-heated to 150 °C, respectively.

In the as-built condition, EC and Vickers microhardness were measured along the entire height (300 mm) of the specimens. The same characterizations were performed on the bottom and top samples machined from the billets, after direct aging, solution heat treatment (SHT), and artificial aging (AA).

Aging temperatures of 150, 175, 200, and 225 °C were chosen to optimize the mechanical properties between the bottom and top regions in the as-built state. Based on the effects observed between 0.5 and 4 hours soaking time at 200 and 225 °C, only the exposure temperatures of 150 and 175 °C were evaluated for prolonged time (up to 512 hours). Both 150 and 175 °C also represent typical service temperatures for AlSi7Mg components. Aging treatments were performed both directly on as-built samples and following solution heat treatments at 505 °C for 0.5 and 4 hours (SHT + AA). Solution temperatures and times were chosen in accordance with

prior studies published in the literature.^[28,29] Heat treatments were conducted in a muffle furnace (Nabertherm GmbH, Lilienthal, Germany), with a K-type thermocouple placed in direct contact with the samples to control the temperature. After each heat treatment, specimens were water-quenched to room temperature.

The microstructure of both the as-built and heat-treated bars was investigated by light microscopy using a Leica DMI8 optical microscope (Wetzlar, Germany) equipped with LAS-X software. The very fine networked microstructure of the AlSi7Mg samples was observed using a scanning electron microscope (Auriga Compact FIB-SEM, Zeiss, Oberkochen, Germany) equipped with EDS analysis. Prior to microstructural analysis, specimens were mechanically ground with SiC papers (from P80 to P4000) and then polished with silica suspension.

Transmission electron microscopy (TEM) analyses were carried out using a PhilipsTM CM-200® operated at 200 keV and equipped with a liquid-nitrogen-cooled double tilt specimen holder. TEM thin discs were prepared mechanically by polishing, dual dimpling down to 25 to 35 μm , and then ion milling, using a GatanTM PIPS operated with Ar + at progressively abutting incident angles from 8° down to 4° to obtain a sufficiently wide electron transparency TEM disc volume.

The statistical analysis of the Si-eutectic network was performed using LAS-X software on a population of 500 particles taken from several micrographs at a magnification of ×1000 for T6 samples and at 80 k × for as-built samples.

Considering the influence of pores on the EC measurements, a statistical analysis of pore size and density was also performed taking a set of six images at ×100 and three at ×1000. Density values (ρ) were calculated as follows:

Table II. Percentage of Pores Measured on xz- and xy-Plane of the AlSi7Mg Bars in As-Built, Direct-Aged and Artificially Aged Conditions and Their Densities

Investigated Conditions	xz-Plane (Pct)	xy-Plane (Pct)	Density (Pct)
As-Built	1.2 ± 0.2	0.9 ± 0.1	98.9 ± 0.2
Direct Aging at 225 °C/4 h	1.2 ± 0.1	1.4 ± 0.2	98.8 ± 0.1
505 °C/4 h + 175 °C/512 h	2.1 ± 0.3	2.5 ± 0.4	97.6 ± 0.2

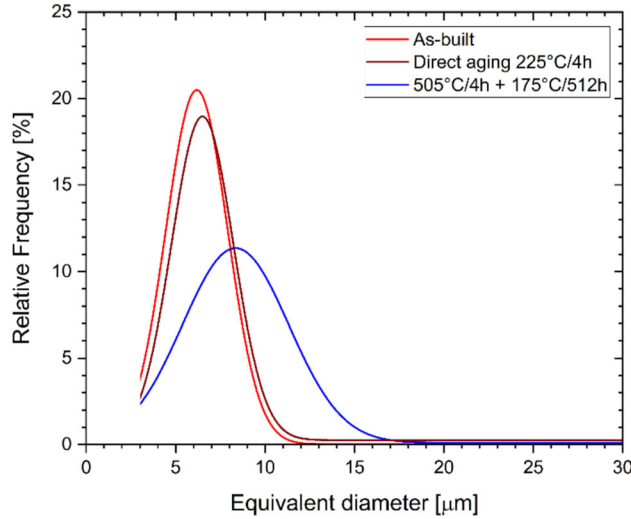


Fig. 2—Statistical distribution of pore equivalent diameters in as-built, direct aged and solution-treated AlSi7Mg samples.

$$\rho = 1 - \frac{\sum_{i=1}^n (A_p)_i}{\sum_{j=1}^6 \bar{A}_j} \quad [1]$$

where A_p represents the pore's area [μm^2] and $\bar{A}$ indicates the micrograph area [μm^2]. The associated errors were calculated according to the equations discussed in Reference 30.

EC testing was performed at room temperature using a calibrated SIGMATEST 2.070.01 device (Foerster, Reutlingen, Germany) equipped with an 8 mm titanium probe and operating at a frequency of 60 kHz. Each value is the average of four measurements.

To evaluate the microhardness variation along the 300 mm height of the as-built bar [Figure 1(a)], Vickers microhardness profiles were measured three times along the build direction (z-axis) at 10 mm intervals between two adjacent indentations and using a Leica VMHT tester (Leica, Wetzlar, Germany) with a load of 500 gf and a dwell time of 15 seconds. A load of 200 gf was used to evaluate any anisotropy between the xz- and xy-planes in as-built specimens.

Tensile tests were conducted at room temperature using a servo-hydraulic machine (Z100, Zwick/Roell, Ulm, Germany) at a constant strain rate of $1.6 \cdot 10^{-3} \text{ s}^{-1}$. For each as-built and heat-treated condition, three tensile tests were carried out on cylindrical tensile samples [Figure 1(b)], prepared according to ASTM E8/E8M.

III. RESULTS AND DISCUSSION

A. Density and Microstructural Investigations

The density values listed in Table II, obtained via 2D image analysis of gas and lack-of-fusion pores and calculated using Eq. [1], confirm the high quality of the LPBF-manufactured AlSi7Mg samples in both the as-built and heat-treated conditions.

Additionally, Figure 2 shows that the pores formed in the LPBF-manufactured AlSi7Mg samples were statistically small in size. In fact, approximately 95 pct of the pores fell within the range of 2 to 10 μm in equivalent diameter for both the as-built and direct-aged samples. Furthermore, only exposure to 505 °C/4 h shifted the peak pore frequencies in the as-built and direct-aged conditions from 6.10 to 8.50 μm (Figure 2) due to the diffusion-controlled gas entrapped within the gas pores. These conditions increased the number of pores from approximately 500 (as-built state) to 2500 to 3000 after SHT at 505 °C.

Figure 3 shows the microstructure of the as-built AlSi7Mg alloy, with Figures 3(a) and (b) acquired on the xz-plane (parallel to the build direction, Figure 1) and on the xy-plane (parallel to the build platform), respectively. Figures 3(c) and (d) are higher magnification micrographs of both the center and edge of the melt pool, respectively, as shown in Figure 3(a). Figure 3(a) exhibits a layered architecture in which the overlap of melt pools forms a fish-scale structure, typical of LPBF-manufactured materials. In contrast, Figure 3(b) depicts the cross-sectioned laser scan tracks and the elliptically shaped melt pools. Figures 3(a) and (b) also illustrate the presence of gas porosity [Figure 3(a)] and lack-of-fusion defects [Figure 3(b)], which account for the density values reported in Table II. The spherical gas pores originate from gases dissolved within the melt pool, which remained entrapped during the solidification process, as extensively discussed in References 31–33. Conversely, lack-of-fusion pores originate from melt pool instabilities.^[4] The density values of the as-built samples were determined by both gas and lack-of-fusion pores detectable during observations, namely by pores having sizes higher than the resolution limit of the optical microscope. This aspect is supported by both Figure 3(a) (red arrow) and Lysak *et al.*,^[34] who analyzed gas pores in the LPBF AA6061 alloy. The reduction in sample density after SHT at 505 °C is driven by an increase in the number of pores, which is attributable to diffusion-controlled movement of gas present in pores that were undetectable in the as-built

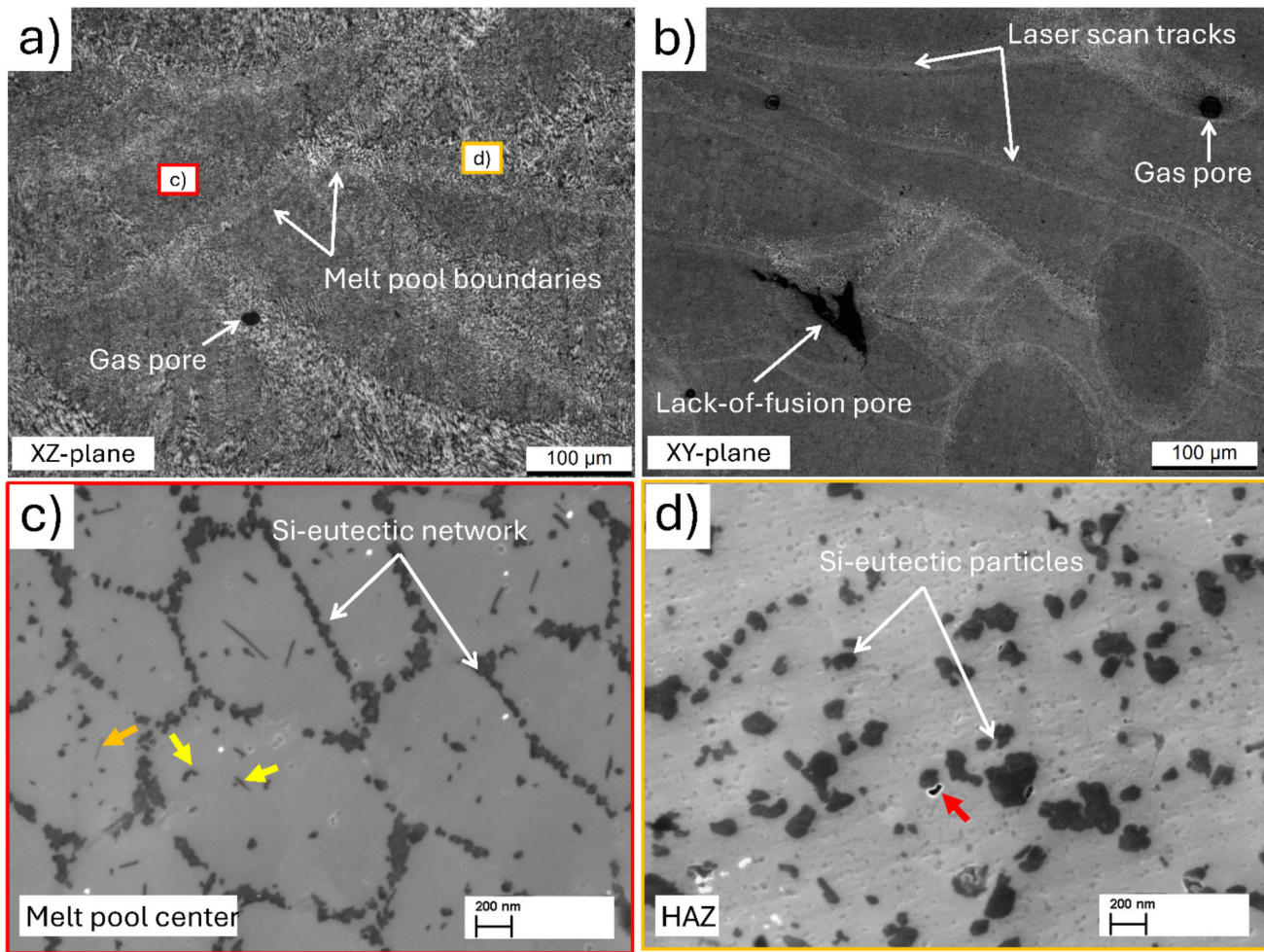


Fig. 3—Microstructure of as-built AlSi7Mg samples acquired on (a) xz-plane and (b) xy-plane. (c), (d) High-magnified SEM micrographs of the (a) melt pool center and (b) its boundary representing heat-affected zone (HAZ). Yellow arrows indicate rod- and circle-shaped Si particles, while the orange arrow indicates an Fe-rich intermetallic phase, and the red arrow points to a small pore.

Table III. Vickers Hardness and EC Variations Along xz- and xy-Plane of as-Built AlSi7Mg Samples

Investigated Plane	AlSi7Mg	
	HV0.2	EC [MS/m]
xy	128 ± 5	17.44 ± 0.07
xz	114 ± 3	18.02 ± 0.10

samples. Simultaneously, the high SHT temperature drives pore coarsening among adjacent pores.

Overall, Table III highlights that the microstructural anisotropy observed along the xz- [Figure 3(a)] and xy-plane [Figure 3(b)] leads to variations in both Vickers microhardness and EC.

The Vickers measurements decreased from the xy-plane to the xz-plane, likely due to the greater density of melt pool boundaries [compare Figures 3(a) and (b)], where the networked structure characterizing the melt pool center [Figure 3(c)] coarsened due to the

diffusion-controlled mechanisms of Si atoms. This phenomenon is triggered by the heat fluxes generated during melt pool solidification.^[35] The discontinuous Si-eutectic network [Figure 3(d)], formed by coarser Si-eutectic particles, made the material softer.^[36] These microstructural features slightly increased the EC measurements, as the coarse and discontinuous Si-network imposed less resistance to the movement of free electrons. In other words, the larger volume of extended and interconnected α -Al regions (the spacing between Si-particles) along the xz-plane, compared to the xy-plane, increases the mean free path of free electrons.

Figure 3(c) also shows α -Al cells decorated by fine Si-eutectic particles that contain rod- and circularly shaped Si nanoparticles, as indicated by the yellow arrows. Moreover, the orange arrow points to needle-like particles that are likely Fe-rich particles, forming during the melt pool solidification due to the low solubility of Fe in the Al lattice.^[37]

Figure 4 illustrates the microstructural changes that occurred during direct at 225 °C/4 h [Figure 4(a)], solution heat-treatment at 505 °C/0.5 h [Figure 4(b)], and artificial aging [Figure 4(c)] in the as-built

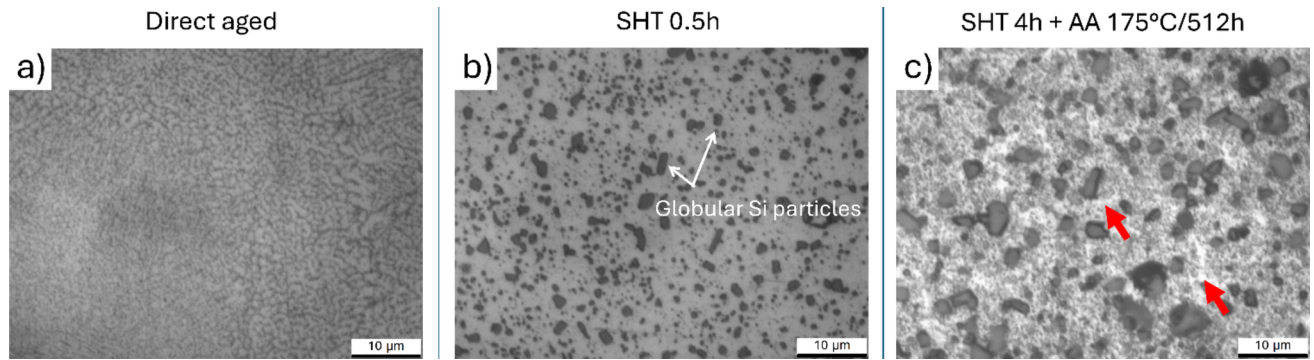


Fig. 4—Microstructures of the AlSi7Mg alloy after: (a) Direct aging at 225 °C/4 h, (b) SHT at 505 °C/0.5 h, (c) long-term T6 (505 °C/4 h + 175 °C/512 h). Red arrows indicate the unetched zones relating to precipitate free zones.

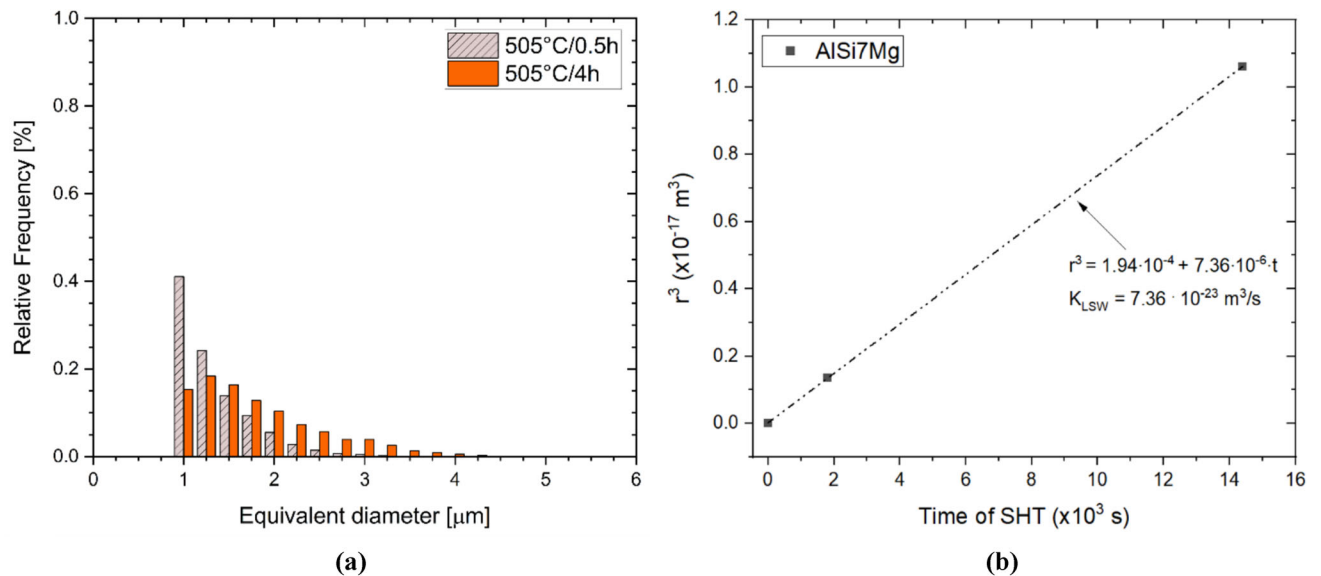


Fig. 5—(a) Variation of the equivalent diameters of Si-eutectic particles and (b) their constant coarsening rate during the exposure to 505 °C/0.5 h and 505 °C/4 h. The term t indicates the exposure time to solution temperature of 505 °C.

microstructure. Direct aging, performed at the highest temperature analyzed in this work (225 °C/4 h), appears not to alter the networked microstructure of the AlSi7Mg alloy [Figure 4(a)], compared to the as-built state.

However, exposure to 225 °C made the Si-eutectic network coarser and more discontinuous than that shown in Figure 3, as detailed in our previous work.^[38] In that study, we demonstrated that direct aging at 225 °C/4 h increased the thickness of the Si-eutectic network by approximately 28 pct. The solution treatment at 505 °C/0.5 h [Figure 4(b)] transformed the networked microstructure [Figures 3(c) and (d)] into a composite-like structure, where globular Si-eutectic particles are embedded within the α -Al matrix. When the exposure time was increased to 4 hours [Figure 4(c)], the Si-eutectic particles coarsened further, as confirmed by the equivalent diameter distributions in Figure 5(a). The peak of the particle size distribution shifted toward higher values, with a corresponding decrease in the peak frequency; this is due to the lower amount of Si-particles

counted in the sample solution-treated for 4 hours (~1450) compared to the sample solution-treated for 0.5 h (~1860). Following SHT at 505 °C/0.5 h, the average size of the as-built Si-eutectic particles increased exponentially from (70 ± 20) nm to (1.05 ± 0.45) μ m due to the diffusion-controlled flux of Si atoms and the Ostwald ripening mechanism. These mechanisms continued to promote coarsening during the 4 h soaking time at 505 °C. This trend reflects that samples solution-treated for 4 h [Figure 4(c)] contained fewer but larger Si particles compared to those solution-treated for 0.5 h [Figure 4(b)]. The diffusion-controlled coarsening of Si-eutectic particles can be described by the Lifshitz-Slyozov-Wagner (LSW) model, given by the following equation:

$$r^3 - r_0^3 = K_{LSW} \cdot t \quad [2]$$

where r^3 and r_0^3 represent the cubes of the final and initial average particle radii (m^3), respectively, K_{LSW} is the coarsening rate constant (m^3/s), and t is the exposure time (s). Figure 5(b) shows the linear relationship

between the cube of the average particle radius and exposure time at 505 °C. The slopes of the lines correspond to the K_{LSW} values described in Eq. [2], namely to a coarsening rate of $7.3610^{-23} m^3/s$. Abdelaziz *et al.*^[39] proposed a K_{LSW} value of $6.3810^{-23} m^3/s$ in a 356 alloy solution-treated at 550 °C, while Liang *et al.*^[40] reported similar values ($K_{LSW} = 2.7 - 9.310^{-23} m^3/s$) in a cast AlSi8Mg0.6 alloy solution-treated at 525 °C.

Exposure to solution temperatures also promoted the coarsening of the needle-like Fe-rich particles that formed during melt pool solidification. The EDS spectrum acquired from the acicular phase indicated that the intermetallic located between two Si-eutectic particles corresponds to Al_5FeSi . This was confirmed by the intensity ratio of the Al, Si, and Fe peaks in the

spectrum and its needle-like morphology (Figure 6). As previously noted regarding Figure 3, this intermetallic phase forms during the melt pool solidification and coarsens during the exposure to elevated temperatures (such as the SHT temperature of 505 °C).

B. Vickers Microhardness, Mechanical Properties, and EC

Figure 7(a) shows the Vickers microhardness and EC measured along the build direction of the as-built 300 mm-tall AlSi7Mg bars [Figure 1(a)]. Both profiles exhibit a decreasing trend from the bottom (the region closest to the pre-heated build platform) to the top (the zone farthest from the pre-heated build platform). The same decreasing trend is shown in Figure 7(b) by the stress-strain curves, with tensile strengths decreasing from the bottom samples (300 MPa YS and 428 MPa UTS) to the top samples (261 MPa YS and 407 MPa UTS). Conversely, ductility increases from the bottom to the top samples from approximately 9 to 12 pct.

The decrease in both Vickers microhardness and strength is related to the precipitation phenomena induced by the build platform being heated to 150 °C. High-magnified TEM images (Figure 8), performed on both the top [Figures 8(a), (b)] and bottom [Figures 8(c), (d)] regions, revealed that the bottom samples [Figure 8(b)] exhibited α -Al cells with a higher density of precipitates [20.9 pct, Figures 8(b), (d)] compared to the top regions [3.9 pct, Figures 8(a), (c)], as discussed in our previous work.^[41] Additionally, the TEM micrographs of the bottom samples depicted the presence of sub-grain structures likely due to the Bérnard-Marangoni instability, promoted by the build platform temperature. Additionally, no significant variations in the dimensions of either the Si-eutectic network or the α -Al cells were observed between the top and bottom samples. Therefore, exposure to the pre-heated build platform indeed induced the formation of a greater

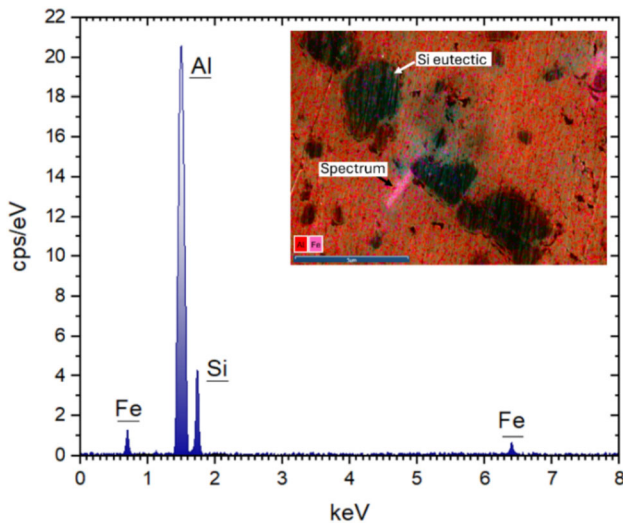
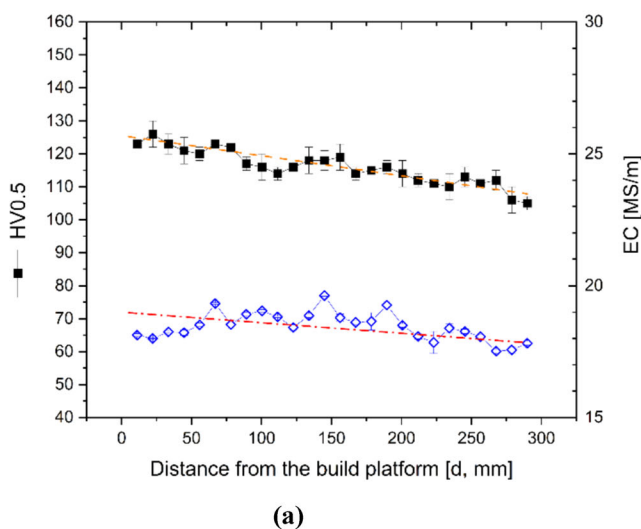
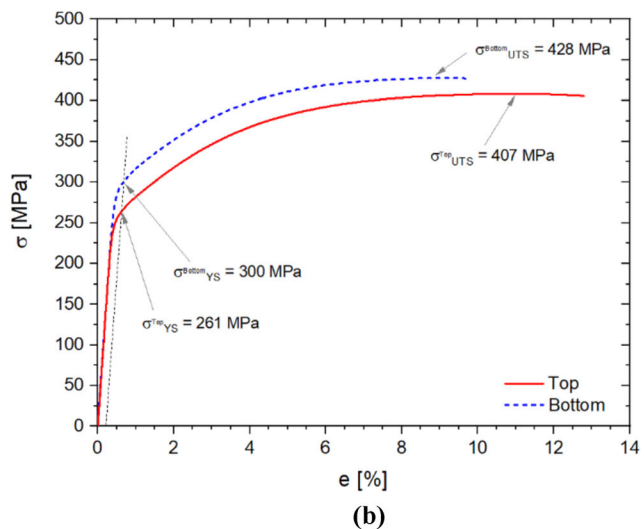


Fig. 6—Spectrum EDS acquired on acicular phase indicated by the black arrow within the EDS map.



(a)



(b)

Fig. 7—(a) Vickers microhardness and EC values measured along the build direction of as-built AlSi7Mg bars. (b) Representative stress-strain curves of both bottom and top samples. The term d indicates the distance from the pre-heated build platform.

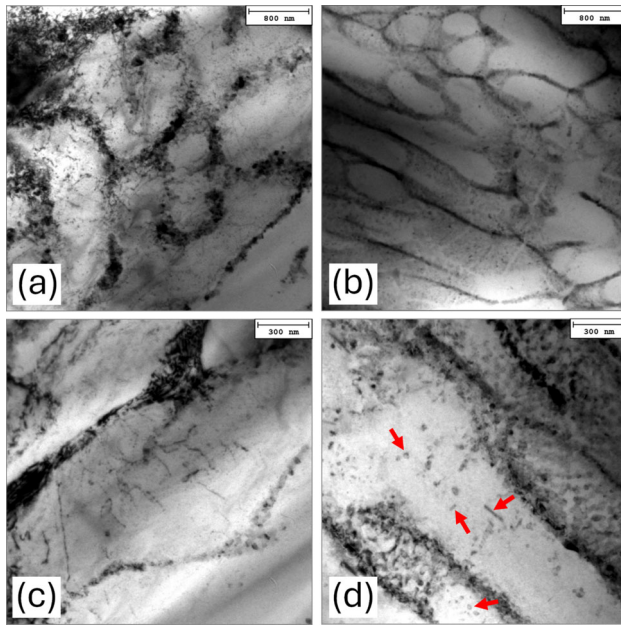


Fig. 8—High-magnified BF-TEM images of top (a), (c) and bottom (b), (d) samples in as-built conditions. Al-200 was the acquisition plane. The g vectors are in the same direction of those indicate in Figure 10. (Reprinted with permission from Ref. [41]).

amount of both semi-coherent β' and incoherent β -Mg₂Si precipitates [indicated by red arrows in Figure 8(d)] within the α -Al cells of the bottom [Figures 8(b), (d)] compared to the top [Figure 8(a), (c)] specimens, as well as the presence of sub-grain structures.

The presence of the as-built structure and the gradient in precipitate amount from the bottom to the top regions result in a slight variation in EC values [Figure 7(a)]. Therefore, the decreasing trend from the bottom to the top samples may be caused by the higher reduction of SSS during aging and the consequent precipitate formation, which relaxes the Al lattice and extends the mean free path of electrons, at the bottom rather than at the top.^[42,43] Finally, considering the density values in Table II, it can be concluded that the variation in EC values of the as-built samples is due to the microstructural changes rather than the presence of voids. This conclusion is also supported by,^[44–46] which investigated the influence of pore size, distribution, and morphology on EC. Hakamada *et al.*^[45] found that electrical resistivity increases with pore diameters greater than 700 μm . Similarly, Li *et al.*^[46] reported that porosity levels above 10 pct drastically reduce EC values. The same observations can be drawn for the solution-treated samples.

Figure 9 shows the variations in UTS, YS, and EC values during direct aging at 175 [Figure 9(a)], 200 [Figure 9(b)], and 225 $^{\circ}\text{C}$ [Figure 9(c)] for the bottom and top samples. Both the UTS and YS reached maximum values of (426 ± 10) MPa and (275 ± 8) MPa, respectively, after 2 hours at 175 $^{\circ}\text{C}$. Conversely, both UTS and YS of bottom samples decreased linearly with exposure time at 175 $^{\circ}\text{C}$ confirming that peak-aging conditions were achieved during the LPBF process.

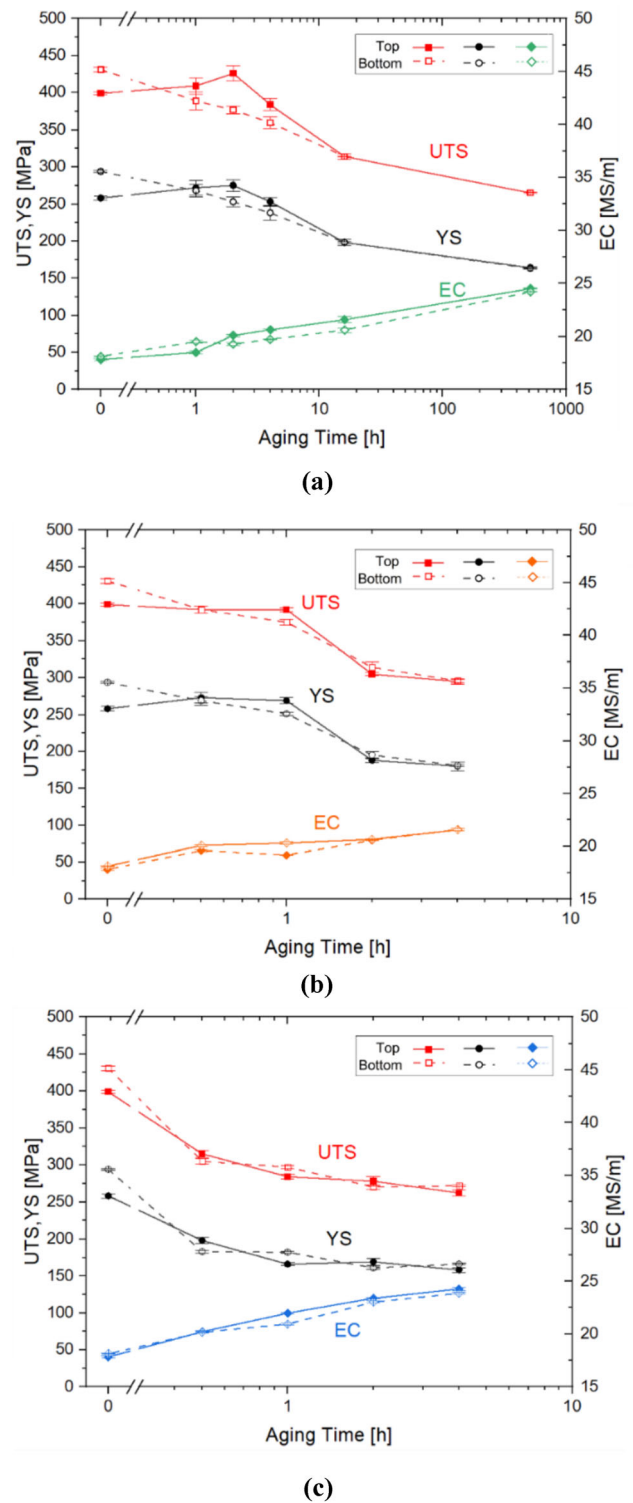


Fig. 9—Variation of mechanical properties (UTS, YS) and EC during exposure to (a) 175 $^{\circ}\text{C}$, (b) 200 $^{\circ}\text{C}$, and (c) 225 $^{\circ}\text{C}$ of both top and bottom AlSi7Mg samples.

As a matter of fact, the precipitates formed during the LPBF process (Figure 8) coarsened during the exposure to the aging temperature. On the other hand, the double-aging phenomena in the bottom samples [smallest precipitates in Figure 10(b), (c)] can only have

Peak-aged conditions

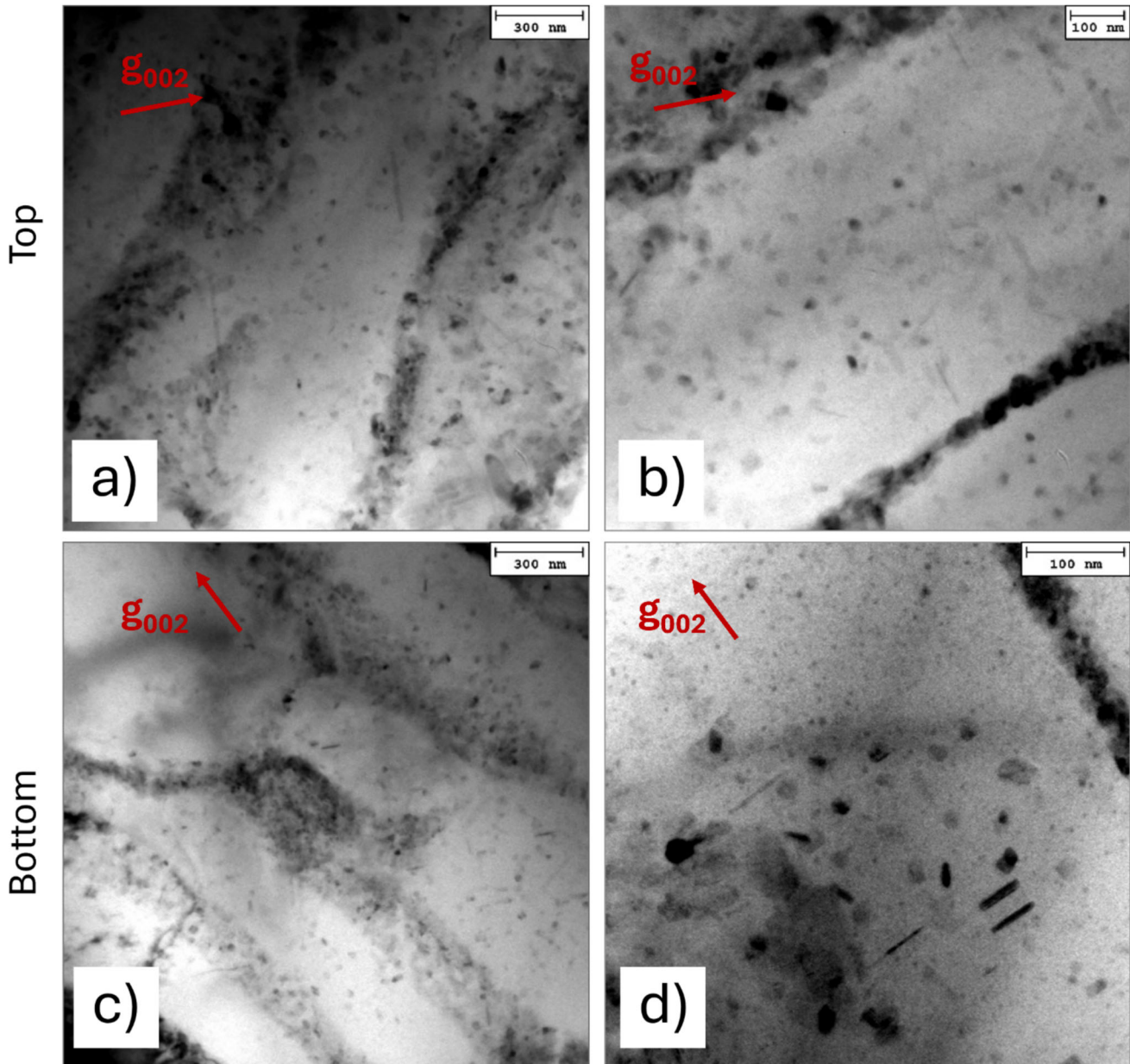


Fig. 10—High-magnified BF-TEM micrographs of the peak-aged (a), (b) top and (c), (d) bottom samples at 175 °C. The crystallographic plane is Al-(200) and the corresponding g vector is reported in each of the here reported micrographs.

mitigated the negative effects caused by coarsening phenomena of those already present in the as-built samples. After 16 hours of exposure, the mechanical properties of the bottom and top samples became homogenized. Ductility increased only after a 512 hours soaking time, as indicated by the stress-strain curve in Figure 9(a). Regarding EC, the measurements continuously increased with aging time due to the precipitation phenomena and the associated reduction in SSS. Figure 9(b) shows the aging curves at 200 °C where the top samples reached peak-aging conditions after 1 hours of exposure, although the first homogenization

in strength was reached at 0.5 hours soaking time. Specifically, the decreasing trends of both UTS (392 ± 5) MPa and YS (273 ± 7) MPa for the bottom samples intersected with the increasing trends for top specimens (UTS of 392 ± 6 MPa and YS of 269 ± 9 MPa). The strengths were again homogenized at direct aging times of 2 and 4 hours. The EC measurements followed the aging profiles, showing a complete overlap after 1 hours at 200 °C. The direct aging at 225 °C [Figure 9(c)] did not exhibit peak-aging conditions due to the coarsening phenomena of the precipitates and the softening caused by the Si-eutectic network, where Si particles became

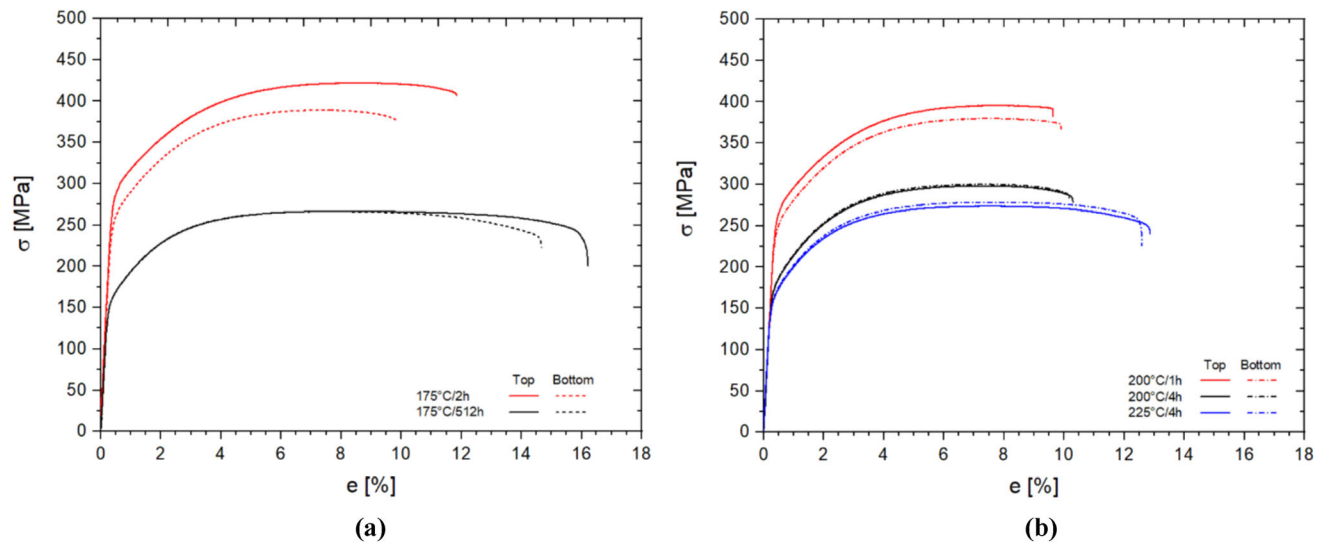


Fig. 11—Representative stress-strain curves of both bottom and top samples after direct aging at (a) 175 °C and (b) 200 °C and 225 °C.

larger and more widely spaced. Due to the microstructural variations occurred during direct aging at 225 °C, the EC measurements of both the bottom and the top samples rapidly increased because of free electrons encountered fewer obstacles along their mean free path. At 175 °C, the highest EC value of (25.52 ± 0.03) MS/m was observed after 512 hours soaking time, likely due to coarsening phenomena affecting precipitates and Si-eutectic network. The EC measurements of 200 °C direct aging samples fall between those measured on samples direct aging at 175 °C (black lines) and at 225 °C (blue lines). This is because the exposure temperature of 200 °C did not greatly influence the Si-eutectic network size in the way that direct aging at 225 °C does, but it promoted more coarsened precipitates than during aging at 175 °C.

Indeed, the peak-aging state shown in Figure 10 exhibited a high precipitate density compared with the matrix illustrated in Figures 8(a), (b). Unlike those conditions, the bottom samples [Figures 10(c), (d)] displayed both significant coarsening of the precipitates already present in the as-built α -Al cells [Figures 8(c), (d)] and double-aging phenomena (*i.e.*, the appearance of new and smaller precipitates). As discussed in Reference 41, we showed that statistical analysis of the precipitates revealed a bimodal size distribution, with the first peak around 5 to 10 nm (new precipitates) and the second peak around 20 to 22 nm. The temperature of 175 °C did not induce any coarsening of either the α -Al cells or the Si-eutectic network, confirming what was previously discussed for the aging temperatures of 200 and 225 °C. In the latter case, it can be stated that the higher aging temperatures promoted the precipitation of coarse phases.^[41,47]

The over-aging state at 175 °C of the bottom samples reduced the strength of the as-built state. Within this framework, exposure to 512 hours led to a significant decrease in strength and to a homogenization of mechanical properties between the top and bottom samples. On the other hand, direct aging temperatures

between 200 and 225 °C [Figure 11(b)] only provided a reduction in strength (ranging from 20 to 30 pct) due to the presence of coarsened precipitates and the Si-eutectic network. Because this Si network was not fully broken down, the ductility values did not show significant increase compared to the as-built state or solution-treated samples (as will be discussed later). Ductility increased from (9.5 ± 0.7) pct in the as-built state to (15.5 ± 1.2) pct or (12.9 ± 1.1) pct under the 175 °C/512 h and 225 °C/4 h conditions, respectively.

Figure 11 shows the effects of both precipitation and Si-eutectic coarsening on the strength of the as-built top and bottom samples during direct aging. The precipitation of Mg_2Si alone in the top as-built samples increased both the YS and UTS by 10 and 6 pct [Figure 11(a)^[42,43]] during exposure of the as-built samples to 175 °C/2 h (peak-aging). Focusing on elongation values, the presence of the networked microstructure [Figures 3, 4(a)] limits the enhancement of the as-built ductility of 9 to 12 pct [Figure 7(b)].

Figure 12 illustrates the tensile properties and EC after exposure at 505 °C/0.5 h [Figures 12(a), (b)] and 505 °C/4 h [Figure 12(c), (d)] followed by aging at 150 °C [Figures 12(a), (c)] and 175 °C [Figures 12(b), (d)]. After the solution heat treatments, both the UTS and YS decreased with the exposure time, from (281 ± 7) MPa and (157 ± 8) MPa to (247 ± 6) MPa and (135 ± 5) MPa, respectively, due to the coarsening phenomena that affected the Si-eutectic particles [Figures 4(b), (c)]. Another important factor for the strength reduction is the lower amount of SSS present in the Al lattice of samples solution-treated at 505 °C/4 h compared to those solution-treated at 505 °C/0.5 h. In fact, the longer SHT time reduced the high concentration of alloying elements in the Al lattice of the as-built samples, allowing it to approach the equilibrium value governed by the phase diagrams. These findings were supported by the reduction in EC values from (21.25 ± 0.48) MS/m in solution-treated samples at 505 °C/4 h to (20.32 ± 0.09) MS/m in solution-treated

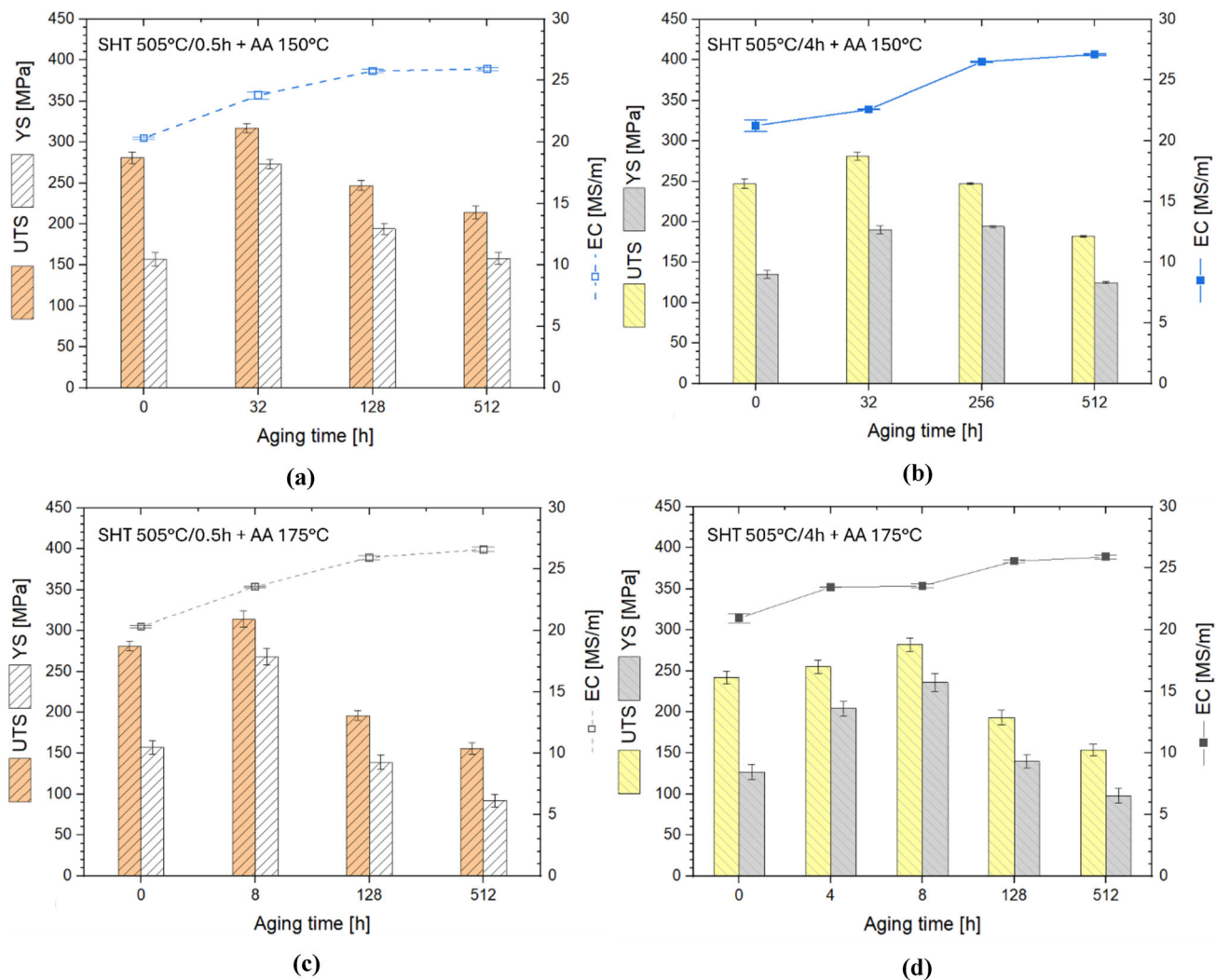


Fig. 12—Variation of mechanical properties (columns) and EC (continuous and dotted lines) of as-built AlSi7Mg samples after T6 heat-treatments: 505 °C/0.5 h (a), (c) + 150 °C (a) or + 175 °C (c), and 505 °C/4 h (b), (d) + 150 °C (b) or + 175 °C (d). The highest peaks of strength represent the peak-aging conditions obtained during artificial aging of samples solution-treated at 505 °C/0.5 h and 4 h.

samples at 505 °C/0.5 h. As a matter of fact, the lower amount of alloying elements in the Al lattice is the key factor in enhanced conductivity, as will be later discussed in Eq. [4].

Exposure to the aging temperature of 150 °C [Figures 13(a), (b)] recovered the UTS and YS of the solution-treated samples, allowing them to reach peak-aging conditions. However, coarsened precipitates formed in over-aged conditions drastically reduced these values. Ductility followed the opposite trend as briefly summarized in Figure 13. In fact, elongation values were reduced from the solution-treated condition to peak-aging state due to precipitation phenomena.

The chemically etched microstructure of the artificially aged samples revealed both dark pits within the α -Al matrix and lighter zones, as indicated by the red arrows. In the first case, the long-term exposure to 175 °C led to the formation of incoherent β -Mg₂Si precipitates uniformly distributed throughout the α -Al matrix. During the exposure to the aging temperature of 175 °C, the supersaturated solid solution promoted

formation of a large amount of Mg₂Si particles (Figure 14). The peak-aging conditions exhibit needle-like secondary phases [Figure 14(a)] that coarsened after 512 hours of exposure [Figure 14(b)]. In fact, the average precipitate size increased from (37 ± 20) nm under peak-aging conditions to (50 ± 20) nm under the over-aged state, as reported in.^[41] The presence of these precipitates promotes the formation of point-like chemical attack in the α -Al matrix, resulting in the formation of pits. In the second case, the lighter zones may correspond to precipitate-free zones, where the absence of precipitation maintained the matrix unetched. These observations are supported by.^[48]

Long exposure times at 175 °C led to a clear recovery in elongation compared with direct aging at 150 °C for 512 hours. This effect may be attributed to the reduced hindrance to dislocation motion caused by the smaller amount of precipitates formed during the 512 hours at 175 °C compared with those formed at 150 °C. Finally, the anisotropy in elongation between the samples solution-treated for 0.5 and 4 hours is related to the

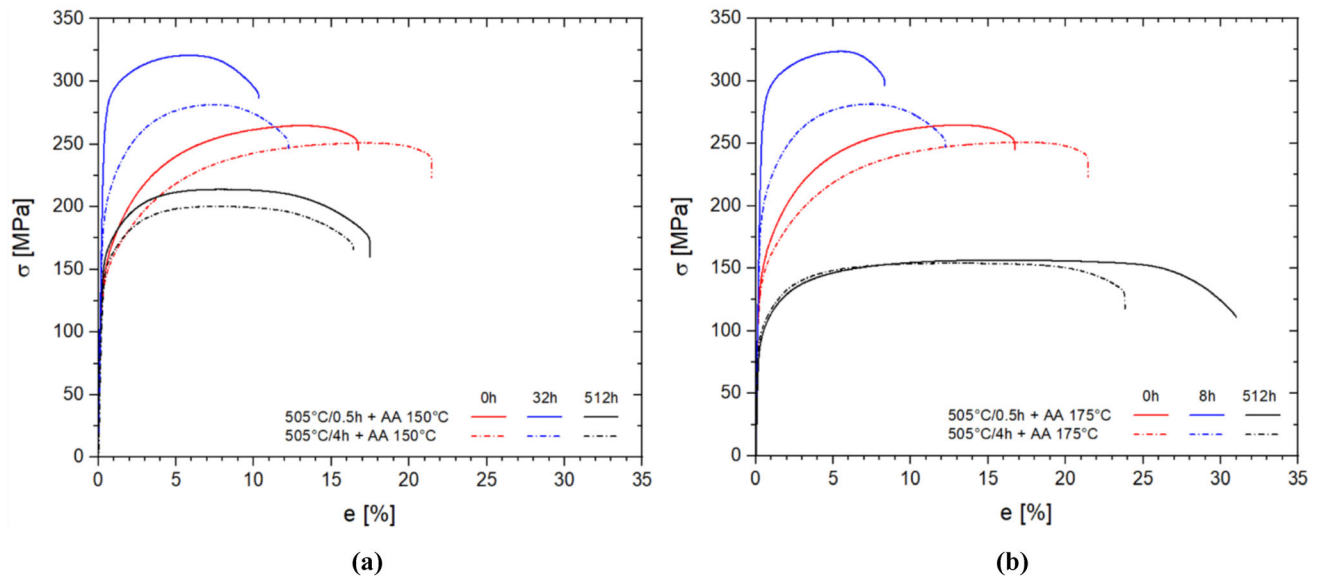


Fig. 13—Representative stress-strain curves of solution heat-treated samples and artificially aged at (a) 150 °C and (b) 175 °C.

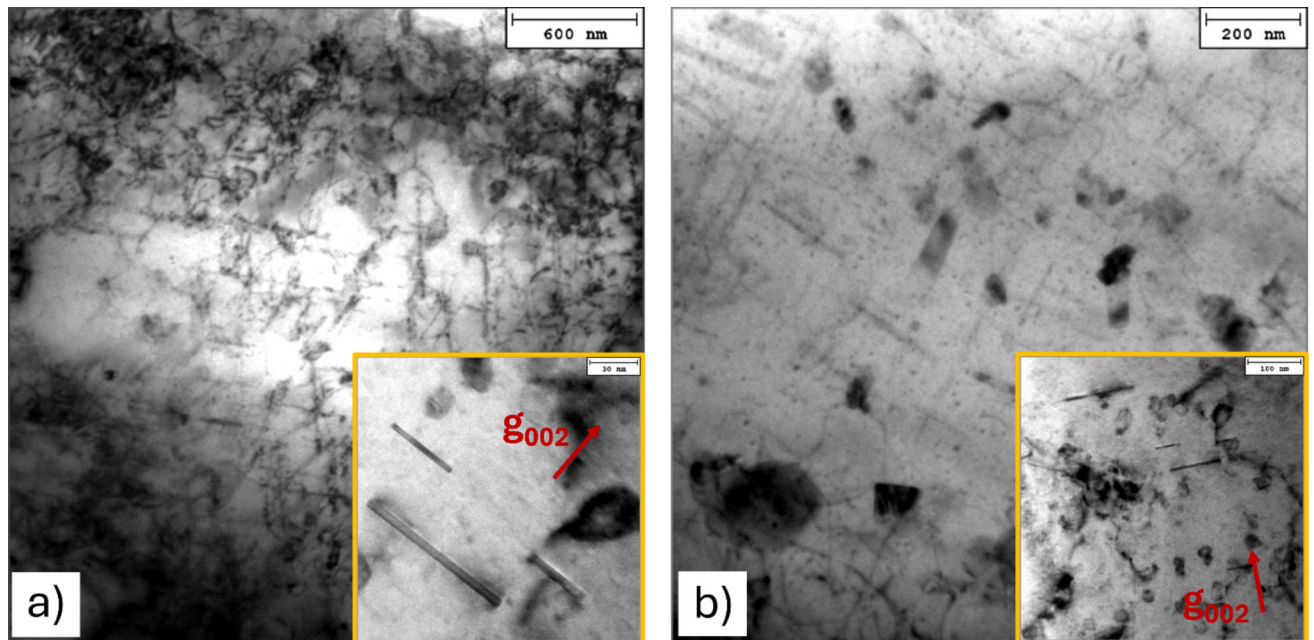
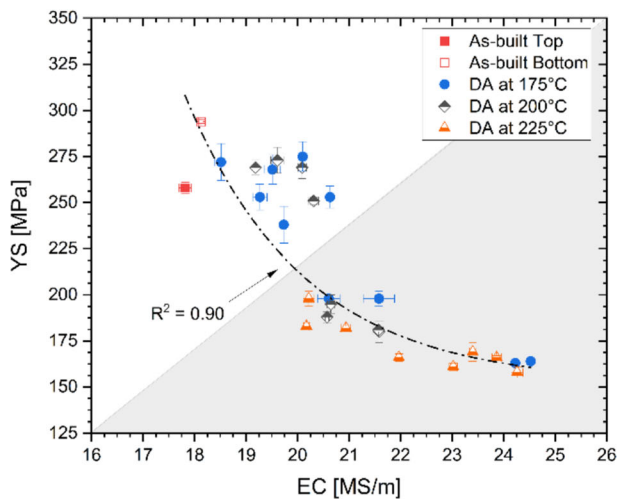


Fig. 14—High-magnified BF-TEM micrographs of solution-treated samples at 505 °C for 0.5 h and aged in (a) peak-aging and (b) over-aging conditions (512 h). The crystallographic plane is Al-(200) and the corresponding g vector is reported in each of the here reported micrographs.

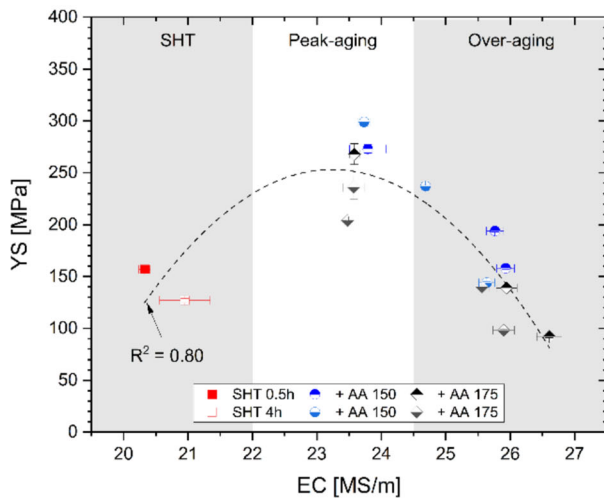
different distributions of the Si-eutectic (Figure 4). As a matter of fact, these Si particles govern the crack propagation during failure.^[49] The EC values in Figure 12 increased monotonically with aging time due to the reduction in SSS and the formation of precipitates. Similar trends in mechanical properties and EC were observed for samples aged at 175 °C [Figures 12(c), (d)]. Due to the diffusion-controlled growth of β -Mg₂Si and its dependence on temperature, aging at 175 °C led the samples to reach peak-aging more rapidly [8 hours in Figures 12(c), (d) compared to 32 hours in Figure 12(a), (b)]. For the same reasons, the UTS and

YS reached after long-term exposure times at 175 °C were lower than those measured at 150 °C. Additionally, the presence of over-aged conditions (Figure 13) caused further degradation of the UTS and YS compared to the solution-treated state. In this case as well, the EC followed a monotonic increase with aging time.

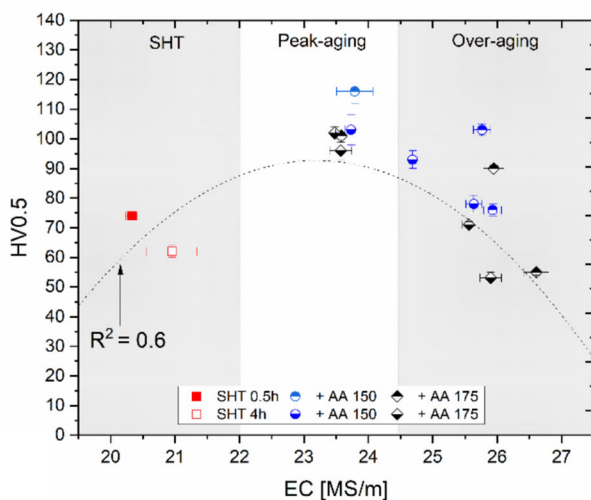
Figure 15 illustrates the correlations between the YS and EC [Figures 15(a), (b)] and between HV and EC [Figures 15(c), (d)] of the AlSi7Mg samples during direct aging [Figures 15(a)] and artificial aging that followed the solution treatment [Figure 15(b)]. Due to the microstructural changes that occur during direct aging,



(a)



(b)



(c)

Fig. 15—Correlations between (a), (b) YS-EC and (c) HV-EC in AlSi7Mg samples in: (a) as-built and direct aged conditions, (b), (c) in T6 conditions.

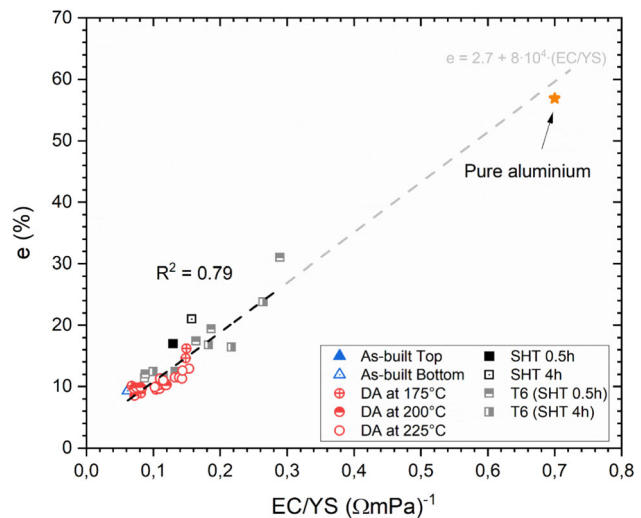


Fig. 16—Variation of elongation values of as-built and heat-treated AlSi7Mg in function of EC/YS ratio. The black dotted line represents the linear fit, while the gray one extends it.

a decrease in YS corresponds to a conductivity improvement [Figure 15(a)]. During short-time exposure ($t < 32$ h) at 175 and 200 °C, the precipitation reduced the SSS, thereby increasing the EC (data contained in the white zone). In contrast, both over-aging conditions and the coarsened Si-eutectic network drastically reduced YS in favor of enhanced EC [data contained in the gray zone of Figure 15(a)]. With regard to the solution-treated samples [Figure 15(b)], precipitation phenomena play a key role in the variation of EC-YS and EC-HV. From the solution-treated state to peak-aged one, both YS or HV and EC increased. This is because semi-coherent precipitates hinder dislocation movement while simultaneously depleting the Al lattice of Si and Mg alloying elements, thus improving EC. When samples reached over-aging conditions, incoherent precipitates offered less resistance to dislocation movement compared to semi-coherent phases, resulting in decreased YS or HV. Meanwhile, the greater reduction in SSS and smaller amount of precipitates caused by formation of incoherent phases resulted in enhanced conductivity. Lastly, it can be concluded that precipitates play a key role in conductivity changes during aging, especially when compared to samples with a networked microstructure where the Si network contributes more significantly to electron scattering.^[22,50,51]

Considering that EC is closely related to the YS (Figure 15), due to their direct dependence on precipitation phenomena, grain dimensions, presence of a composite-like microstructure rather than a networked one, the EC/YS ratio may serve as a valuable parameter in the design of materials for specific applications. The EC/YS ratio provides a measure of the trade-off between electrical efficiency and mechanical reliability in engineering design, effectively distinguishing materials that are highly conductive but relatively soft ($EC/YS \sim 1$) from high-strength materials exhibiting low EC values. Based on investigations made on pure aluminum in Refs. [52,53], its EC/YS ratio is equal to 0.7. This is a

significantly higher value than those depicted in Figure 16, where the EC/YS values of LPBF-manufactured AlSi7Mg samples ranged from 0.06 to 0.3. This lower EC/YS ratio results from the lower EC and significantly higher YS of the as-built and heat-treated AlSi7Mg samples compared to pure aluminum. Figure 16 also depicts that the EC/YS ratio is directly proportional to elongation. The strong linear fit model derived from as-built and heat-treated LPBF-manufactured samples accurately approximates the elongation of the pure aluminum samples and LPBF-manufactured AlSi7Mg alloys from the literature^[5,54] indicated in Figure 16. The following linear fit indeed predicts an elongation of approximately 58.9 pct:

$$e_{\text{pct}} = 2.7 + 8.05 \cdot 10^4 \frac{EC}{YS} \quad [3]$$

Generally, the linear correlation between ductility and the EC/YS ratio highlights that relatively soft materials ($EC/YS \sim 1$) are highly deformable. Dislocations can glide freely over long distances, matching the increase in the mean free path of the free electrons. Conversely, high density of semi-coherent and incoherent precipitates, lattice distortion, fine grains, and a networked microstructure pose severe obstacles to dislocation movement. This condition results in YS improvement, and a simultaneous reduction in both ductility and EC. According to the literature, indeed, the contribution related to the total resistivity can be directly correlated with those describing the total strength and, consequently, ductility. For electrical resistivity (inverse of EC), Matthiessen's rule establishes that the total resistivity (ρ_{total}) is given by^[21,55]:

$$\rho_{\text{total}} = \rho_{\text{Al}} + \rho_{\text{D}} + \rho_{\text{GB}} + \rho_{\text{SS}} + \rho_{\text{PREC}} \quad [4]$$

where ρ_{Al} is the resistivity of pure Al, ρ_{D} is the resistivity contribution from dislocations, ρ_{GB} is the resistivity contribution from grain boundaries, ρ_{SS} is that of the solid solution, and ρ_{PREC} is the contribution of the precipitates. These terms contribute to increasing the resistivity in comparison to pure Al (ρ_{Al}), due to the rise in electron scattering. In other words, they reduce the mean free path of the free electrons and, consequently, reduce the material's conductivity, as also stated in the Introduction section. Simultaneously, the same microstructural features expressed in Eq. [4] also govern the strengthening contributions, as follows^[56,57]:

$$\sigma = \sigma_{\text{Al}} + \sigma_{\text{D}} + \sigma_{\text{GB}} + \sigma_{\text{SS}} + \sigma_{\text{PREC}} \quad [5]$$

where σ_{Al} is the strength of the pure Al crystal lattice, σ_{D} is the strength related to dislocations, σ_{GB} is the contribution from grain boundaries which is governed by the Hall-Petch equation, σ_{SS} is that of the solid solution, and σ_{PREC} is the contribution governed by precipitates.

IV. CONCLUSIONS

The present work investigated the roles played by the microstructural variations on electrical conductivity of laser powder bed fused AlSi7Mg. From the discussed results, the following conclusion on electrical conductivity can be drawn:

Both as-built and heat-treated AlSi7Mg samples exhibited high-density values despite an increase in amount of pores during the solution heat treatment. Nevertheless, the small pore dimension and densities higher than 97.7 pct minimally impacted on electrical conductivity measurements.

- (1) The anisotropic microstructure between xy- and xz-plane, *i.e.*, the different amount of heat-affected zones around the melt pool boundaries, varies the electrical conductivity by around 4 pct. The Vickers microhardness consequently changes by 11 pct.
- (2) The electrical conductivity followed a parabolic curve, from bottom to top regions of the 300 mm height sample, reaching the maximum in the middle zone. The sub-grain structures and Si-nanoparticles present in α -Al cells of bottom samples and the amount of SS retained in top samples increased resistivity with respect to the middle zone.
- (3) The exposure to solution temperature coarsened the Si-eutectic particles with a constant growth rate of $7.36 \cdot 10^{-23} \text{ m}^3/\text{s}$ and increased the as-built conductivity of 13 and 16 pct after 0.5 and 4 hours soaking time, respectively. The composite-like microstructure of coarsened Si particles within the α -Al matrix makes precipitation hardening the key role in electrical conductivity improvement with respect to the as-built samples, the solutioned specimens exhibited a simultaneous increase in mechanical properties (12 to 17 pct of ultimate tensile strength and 40 to 80 pct of yield strength) and electrical conductivity (11 to 17 pct).
- (4) Exposure to direct aging at 175 °C/1 h represents the optimal condition for homogenizing properties between top and bottom regions, without significantly increasing the electrical conductivity (< 20 MS/m).
- (5) The EC/YS ratio is linearly correlated with the elongation through the math equation of

$$e_{\%} = 8.05 \cdot 10^4 \frac{EC}{YS} + 2.7.$$

Ultimately, it can be concluded that the optimal heat treatment to achieve a good balance between electrical conductivity and yield strength is the solution heat treatment followed by aging at 150 or 175 °C under peak-aging conditions. Conversely, to achieve higher electrical resistivity, a networked microstructure must be maintained. In this case, direct aging treatments performed at temperatures below 200 °C for short exposure times (< 2 h) are the best choice. This approach yields

components with high YS while reducing the differences in strength between the bottom and top regions of the components. However, the presence of the networked microstructure limits the enhancement of electrical conductivity. Finally, the 300 mm-tall AlSi7Mg samples lack stability in both mechanical properties and electrical conductivity during exposure at temperatures of 150 °C and 175 °C, which may represent the operating temperatures of these AlSi7Mg components.

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AUTHOR CONTRIBUTIONS

Conceptualization: E.G.; Methodology: E.G.; Validation: E.G., E.C.; Formal analysis: E.G., L.C.; Investigation: L.C., M.C.; Resources: E.G., E.C.; Data curation: E.G., L.C., M.C.; Writing—original draft: E.G.; Writing—review and editing: E.G., L.C., E.C., M.C.; Visualization: E.G., M.C.; Supervision: E.G., E.C.; Project administration: E.G.; Funding acquisition: E.G., E.C..

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DATA AVAILABILITY

Data will be made available on request.

COMPETING INTERESTS

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