

Article

In Situ Ageing Assessment During the Manufacturing of a Large-Scale A357-AlSi7Mg0.6 Component by Laser Powder Bed Fusion

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Abstract

Additive manufacturing (AM) is increasingly being considered to produce large aluminum components for aeronautical applications. Larger parts lead to longer build durations, numerous lasers and may result in prolonged exposure to high temperatures if heat dissipation is insufficient, causing in situ ageing during manufacturing and a consequent reduction in the mechanical properties of as-built aluminum alloys. This study investigates the relationship between process parameters, heat dissipation, and resulting mechanical properties in large-scale A357-AlSi7Mg0.6 aluminum alloy manufactured by laser powder bed fusion (LPBF). In the context of developing geometrically complex casings or components using LPBF equipment with multiple lasers (up to 12), mechanical testing of first prototypes revealed up to a 30% reduction in mechanical properties in certain regions, attributed to insufficient heat dissipation during the build. Thermal modelling and in situ experimental measurements have shown that the high thermal mass and extended build times of large components lead to non-uniform temperature distributions, promoting undesired microstructural evolution and advanced over-ageing, effects that are generally less pronounced in smaller specimens reported in the literature. Hardness mapping on samples subjected to monitored ageing ranges, as well as on additively manufactured rods with deliberate geometries for heat accumulation, confirmed that areas with insufficiently optimized support show significant decreases in hardness due to inefficient heat dissipation. A numerical model was developed to link the hardness evolutions to both ageing time and temperature, enabling improved predictions of microstructure and properties as a function of build and heat-removal strategies. These findings emphasize the critical importance of support structure optimization for both mechanical support and thermal management during LPBF, as well as the need for tailored process and post-process heat treatments to achieve consistent and reliable mechanical properties in large, additively manufactured aluminum alloy components.

Keywords: additive manufacturing; laser power bed fusion; selective laser melting; A357; AlSi7Mg0.6; Al-Si-Mg ternary alloys; precipitation; in situ ageing; artificial ageing; heat dissipation; thermal simulation



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1. Introduction

Additive manufacturing (AM) encompasses various manufacturing processes that are increasingly used in the automotive, aerospace, and biomedical industries [1]. Selective laser melting (SLM), also known as laser powder bed fusion (LPBF), is the most common direct fabrication process that is used to produce metallic components [2]. An important weight reduction can be obtained using this process compared to conventional casting or forging processes [1]. Producing parts layer by layer using LPBF requires the repetition of these three following steps: deposition of a powder layer (20–100 μm thick) onto the build platform using a scraper; selective melting of the powder by a laser spot; and lowering of the build platform. At least 50 parameters must be finely chosen to obtain the suited thermomechanical properties of the built parts [3], also depending on the size of the parts that must be built.

Aluminum alloys are essential structural materials in aeronautic and automotive applications, owing to their low density, superior corrosion resistance, and balanced mechanical properties [1]. For traditional casting, extrusion, and cold-work operations, alloy selection is extensive; however, options for laser powder bed fusion (LPBF/SLM) remain relatively limited, though rapidly evolving [4]. Among the few widely adopted grades, AlSi10Mg and AlSi7Mg0.6 stand out, with their LPBF-compatible powder forms introduced primarily for the automotive industry since 2011 and for aerospace since 2015 [1]. Silicon enhances castability and compensates for solidification shrinkage, while magnesium increases strength through solid solution and precipitation hardening mechanisms [5].

Rapid cooling rates inherent to LPBF ($>10^6$ K/s) induce supersaturated solid solution in α -Al matrices, with silicon concentrations markedly higher than equilibrium values (equilibrium content of 1.65 wt. % for AlSi7Mg0.6 alloy), sometimes approaching 5 wt. % [6]. The precipitation sequence after quenching and ageing in hypo-eutectic Al-Si-Mg alloys follows the transformation from supersaturated solid solution to Guinier–Preston zones, then to β'' (Mg_5Si_6 or $\text{Mg}_5\text{Al}_2\text{Si}_4$) coherent and β' semi-coherent precipitates, finally leading to incoherent β - Mg_2Si phases under overaged conditions [7–9]. Some researchers distinguish β' precipitates from type A (U1), type B (U2) and type C (B') phases [9]. Every phase belongs to a different space group and has a specific chemical composition. β'' nano-scale precipitates are responsible for the highest strength of materials by acting as an obstacle to dislocations [10]. As-built LPBF microstructure comprises fine α -Al cells bordered by a silicon-rich eutectic network (from around 200 nm to 800 nm [6]), which coarsens substantially upon heat treatment, affecting mechanical properties [6].

Several thermal post-process strategies have been developed to optimize the mechanical performance of Al-Si LPBF alloys [6,7,11–13]. Solution heat treatment (505–540 $^{\circ}\text{C}$), followed by water quenching and artificial ageing at 150–225 $^{\circ}\text{C}$, facilitates the precipitation sequence and relieves residual stresses, but excessive solutionizing can degrade strength due to eutectic coarsening and loss of fine β'' precipitates [11]. The coarsening of these cell walls is observed with spherical eutectic silicon reaching almost 1 μm in diameter [6]. In addition, aluminum has a strong affinity for hydrogen, which leads to the formation of microporosity and blistering on the surface of the parts during the high-temperature solution treatment stages (approx. 500 $^{\circ}\text{C}$). Direct ageing (DA) at 150–225 $^{\circ}\text{C}$ without prior solution treatment leverages the as-built fine microstructure, increasing hardness by up to 15% [6,10], although it only partially relieves residual stresses. Intermediate stress relief treatments (200–300 $^{\circ}\text{C}$) dissipate residual stress effectively but risk further microstructural degradation and diminished hardness [12]. The complex relationship between alloy chemistry, process-induced microstructure, and heat treatment protocols underscores the need for integrated optimization to achieve targeted mechanical properties and reliability in LPBF Al-Si-Mg components [7,11,12,14,15]. Hardness mapping performed

on samples subjected to controlled ageing, as well as on additively manufactured samples designed with geometries to promote heat accumulation, confirmed that regions with not thermally optimized support structures experience significant reductions in hardness due to inefficient heat dissipation [16,17].

Many researchers have proposed simulation and experimental methods [18–20] for predicting the temperature distribution and residual stress during the LPBF process. Li and Gu [21] conducted a parametric analysis of the thermal behaviour of aluminum alloy powder during LPBF, demonstrating the pivotal role of laser power, scanning speed, and layer thickness in influencing the thermal field, melt pool morphology, and final part quality. Building on this, Li et al. [22] correlated numerical simulations with experimental investigations to reveal the significant impact of thermal management on the formation of defects such as porosity and cracking, emphasizing the need for precise process control. Yang et al. [23] further contributed with an analytical model capable of predicting the temperature evolution in AlSi10Mg powder beds under LPBF conditions, supporting both process optimization and microstructural prediction. Complementarily, Du et al. [24] developed a temperature field prediction model that incorporates various boundary conditions and the material's thermophysical properties, offering deeper insights into heat transfer mechanisms and their effect on solidification dynamics. Finally, Ansari et al. [25] employed both modelling and experimental approaches to assess temperature distribution and melt pool size, confirming that thermal profile control is essential for ensuring high-quality production of LPBF-fabricated aluminum alloy components. Collectively, these studies advance the understanding of the thermal phenomena governing LPBF of aluminum alloys and lay the groundwork for the development of more robust process monitoring and control strategies aimed at enhancing material performance, especially from an industrial point of view with a macro evaluation of thermal evolution during the build associated with the link with hardness and mechanical properties correlations. In this paper, the effect of part geometry on heat dissipation, in situ ageing during manufacturing, and the resulting residual mechanical properties of as-built A357-AlSi7Mg0.6 parts produced by AM is investigated. A comprehensive database of ageing conditions enabled the prediction of the effects of in situ ageing temperature and duration on the mechanical properties of AM parts. By coupling this approach with thermal modelling of the heterogeneous thermal history experienced during manufacturing, a model was developed to predict the spatial distribution of mechanical properties in as-built parts.

The genesis of this approach stems from the need to address the challenges associated with controlling the properties of aluminum alloys during additive manufacturing, particularly for large-scale components undergoing complex thermal cycles. This requirement became sharply apparent when transitioning from the XLINE2000R system (two lasers, 11-day build) to the NXG12 SLM600 (12 lasers, 3-day build) on the same part. While the reference case on the legacy machine did not reveal significant inhomogeneities, the drastic increase in productivity—via multiplication of laser sources and reduction of overall build time—on the SLM600 highlighted and exacerbated localized over-ageing phenomena and thermal gradients within the part. A similar observation can be made with the laser power implemented on the NXG12 SLM600 around 650 W compared to those usually implemented for more conventional machines less than or equal to 400 W. This observation directly underscored the necessity for a predictive multi-scale methodology, combining advanced thermal simulations with bespoke metallurgical diagrams, to reliably anticipate and control the distribution of properties in safety-critical components produced by additive manufacturing.

2. Materials and Methods

2.1. Materials and Additive Manufacturing Process

All samples have been built in an argon flow atmosphere using the Nikon NXG12 SLM600 equipped with 12 lasers of up to 1000 W (used at 650 W) power each at SAMC (Safran Additive Manufacturing Campus) at Le Haillan (France) on a build platform pre-heated to 150 °C. The build platform temperature of 150 °C was selected to minimize residual stress and prevent cracking during the solidification process. The AlSi7Mg0.6 gas-atomized powder with an average particle size from 27 to 46 µm has been received from the manufacturer (Tekna Advanced Materials, Sherbrooke, QC, Canada). Manufacturing parameters have been developed to get the optimum mechanical properties with reduced pore size and distribution. The chemical composition of the powder is given in Table 1.

Table 1. Chemical composition of AlSi7Mg0.6 powder used in this study, in wt.%.

Al	Si	Mg	Fe	Ti	Mn	Cu	Zn
Base	6.50	0.45	-	-	-	-	-
	7.50	0.70	0.19	0.20	0.10	0.05	0.07

Three kinds of samples have been prepared for this study: common cylindrical samples for tensile tests after ageing conditions, rods for thermal gradient ageing and riser/feeder type samples for heat accumulation generation.

2.2. Samples and Heat Treatments

A thermal gradient inside a tubular furnace (Carbolite-Gero TG3 12/60/600, Ne-hausen, Germany) [9,26–28] has been maintained on four rods of length 325 mm and section 3.5 × 7 mm. One extremity has been maintained at 170 °C and the other end at 240 °C during periods of 6 h, 50 h, 100 h or 200 h (Figure 1), one rod per condition. Cylindrical samples from 4 to 14 mm in diameter and from 10 to 60 mm in length have been exposed to ageing treatments at a temperature between 150 °C and 300 °C during 2 h to 5000 h (see Table 2) in a Memmert UN110 oven (Mettmert GmbH + Co. KG, Schwabach, Germany) (at least three samples per condition). The riser/feeder type samples (funnel-shaped test tubes) have been produced as presented in Figure 2a, but the building has stopped prematurely due to extreme overheating at middle height (blistering or bubbling of the surface and cohesion with the scraper stopping the build) as seen in Figure 2b. Dimensions of the funnel-shaped test tubes are presented in Table 3 according to schematics in Figure 2c, one sample per condition.

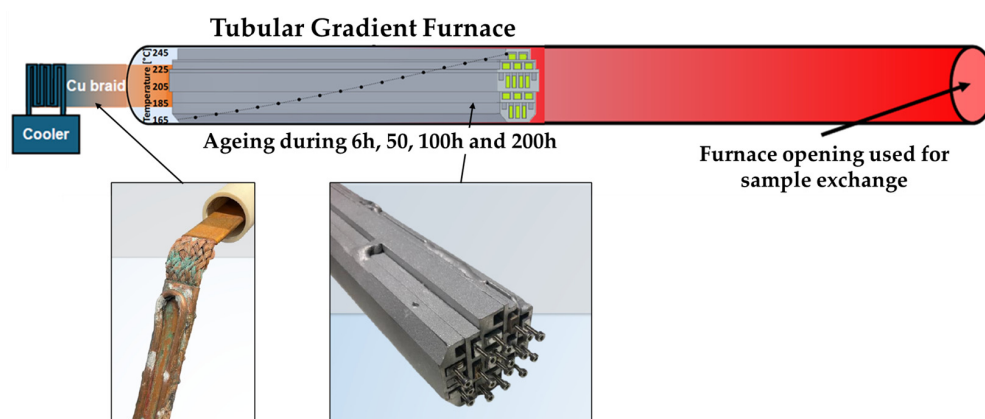


Figure 1. Schematic representation of the position of the samples (in green) within the tubular furnace for thermal gradient sampling preparation [9,26–28]. Reprinted from Ref. [26].

Table 2. Ageing conditions of the cylindrical samples studied.

Temperature (°C)	150 °C	200 °C	225 °C	275 °C	300 °C
Ageing duration (h)	1500 h	2.5 h	22 h	41 h	120 h
	5000 h		71 h	125 h	378 h
			233 h		500 h

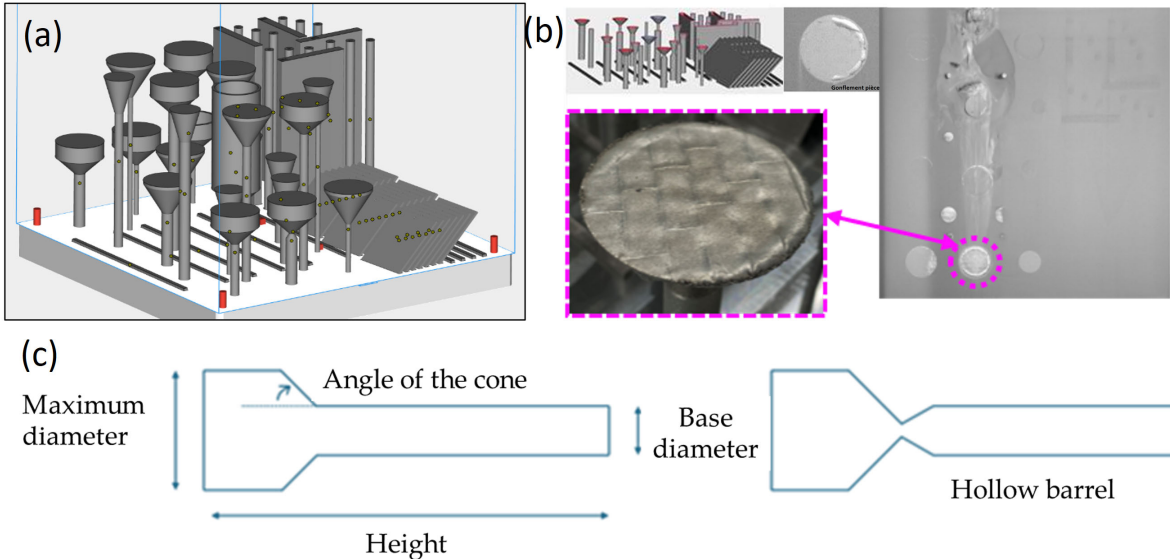


Figure 2. (a) 3D view of the initial build plate; (b) the build was prematurely interrupted at approximately mid-height because severe overheating caused surface swelling and material distortion, leading to interference with the recoater (120 mm only); (c) schematic dimensions of the samples.

Table 3. Dimensions of the funnel-shaped test tubes of initial height 160 mm but produced at only 120 mm.

Reference *	Base Diameter (mm)	Angle of the Cone (°)	Maximum Diameter (mm)	Comment
S-SA-SB	8.7	15	82	/
S-SA-MB	15.1	15	82	/
S-SA-LB	19.5	15	82	/
S-MA-SB	8.7	30	82	/
S-MA-MB	15.1	30	82	/
S-MA-LB	19.5	30	82	/
S-HA-SB	8.7	45	82	/
S-HA-MB	15.1	45	82	/
S-HA-LB	19.5	45	82	/
H-HA-SB	8.7	45	82	The cone is hollow
Hollow barrel	20	45	82	Hollow barrel, and there is a pinch at the base of the cone, thickness 1 mm.

* S: solid, H: hollow, HA: high angle, MA: medium angle, SA: small angle, LB: large base, MB: medium base, SB: small base.

2.3. Microstructural Analysis

For microstructure observations and microhardness, samples have been cut, polished and prepared with Struers Secotom-20, Accutom-100 or Discotom-100, Struers Tegramin-30 and Struers Litopress-30 (Ballerup, Denmark). Samples have been chemically etched with Keller’s reagent (2.5 mL HNO₃ + 1.5 mL HCl + 1 mL HF + 95 mL distilled water) for 25 s. Observations of the samples have been made with Axio Imager.Z2m (Zeiss, Rueil-Malmaison, France) and ZEISS EVO25 SEM (Zeiss, Rueil-Malmaison, France). The

Transmission Electron Microscope (TEM) JEOL 2100F from the SIMAP laboratory (Grenoble, France) has been used. The preparation method used for the TEM observation was the one proposed by Bello et al. [29]. The technology ASTAR has been used to help identify the phases present, unfortunately without success due to the small size of the β'' -Mg₅Si₆ needles. This package is an automatic crystallographic orientation and indexing tool developed for TEM [30].

2.4. Mechanical Testing

Brinell microhardness tests have been performed using the equipment Qness 250E (QATM, Golling, Austria). An indenter of 1 mm and a load of 10 kgf have been set. Thus, HBW 1/10 tests have been performed. Vickers microhardness tests have been performed using the Wilson VH3300 (Buehler, Leinfelden-Echterdingen, Germany) equipment and a force of 0.2 kgf. In accordance with the ASTM E8 standard [31], cylindrical tensile samples, with a diameter of (4.0 ± 0.1) mm and a gauge length of 20 mm, were machined from the cylinders to evaluate the mechanical properties of the heat-treated samples.

3. Results

3.1. Microstructure in Heat-Treated Conditions

The observed microstructure (Figure 3a,b), according to building direction orientations, is defined by a fine network of silicon constituent eutectic (highlighted with yellow arrows in Figure 3c), which envelops α -aluminum cells ($<1 \mu\text{m}$ diameters). Within these cells, hardening via precipitation processes following the GP zone to β'' -Mg₅Si₆ transformation is active during direct ageing. This cellular configuration is consistently found throughout the melt pool (MP) zone, from its core outward to its periphery. Notably, at the edges of the molten pool, localized heat flows generated during the powder melting stage promote the growth and coarsening of the silicon eutectic (shown in magnified microstructure from Figure 3a), ultimately leading to the breakdown of the silicon network structure (Figure 3c). Such a coarsening will have a direct impact on ductility and fracture toughness.

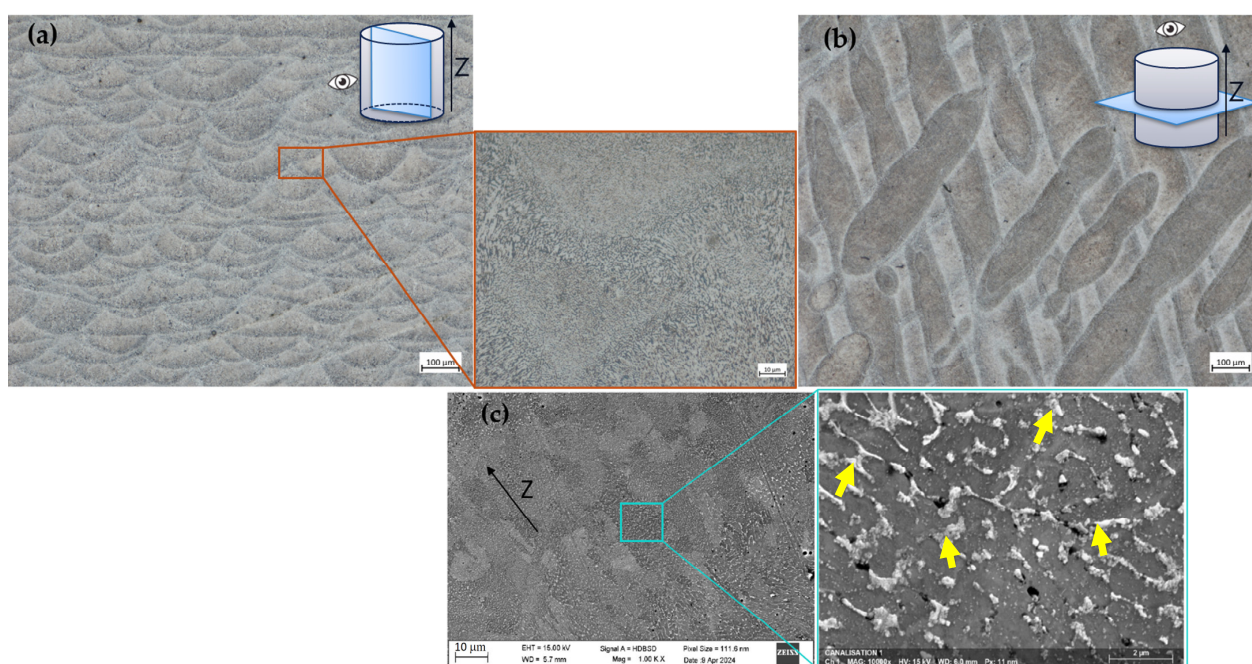


Figure 3. OM micrograph of the AlSi7Mg0.6 sample DA at 170 °C during 6 h in (a) vertical view; (b) horizontal view; (c) SEM micrograph of an AlSi7Mg0.6 sample. The yellow arrows indicate the Si-eutectic constituent particles that form the Si-network around α -Al cells.

At the bottom of the molten pool, the microstructure is coarser compared to the core region. This can be attributed to differences in the solidification rates across these zones [32,33]. In additive manufacturing processes or during rapid solidification, the central area of the molten pool typically experiences higher cooling and solidification rates, which promote the formation of finer cellular structures (Figure 3c). In contrast, at the bottom of the pool, the heat is dissipated less efficiently due to the proximity to the previously solidified layers and the substrate, resulting in a slower solidification rate. This reduced cooling rate at the bottom provides more time for solute redistribution, leading to the formation of larger grains and a coarser microstructure. Therefore, the difference in solidification speeds between the pool's core (fast) and bottom (slow) directly influences the scale and morphology of the microstructural features observed as well as the distribution of residual stress.

3.2. Ageing Profiles and Hardness Evolution

In order to generate hardness charts as a function of the time–temperature pair for AlSi7Mg0.6 parts produced by LPBF, hardness measurements (Figure 4) were conducted along the rod samples aged within a thermal gradient as described in Section 2.2. As these samples are relatively small and were built with the large base of the laths in contact with the build plate, no significant in situ thermal ageing is expected. Four rod samples were aged under a thermal gradient ranging from 170 °C to 240 °C for durations between 6 h and 200 h. The resulting hardness mappings along the thermal gradient and as a function of ageing time are presented in Figure 5.

A time–temperature equivalence approach was implemented as described in [9,26–28]. This model assumes that the evolution of mechanical properties during ageing is governed by a single thermally activated mechanism, typically the diffusion of the solute species involved in precipitation hardening. Under this assumption, the evolution of mechanical properties depends on the diffusion length, given by:

$$L = \sqrt{D_0 \exp\left(-\frac{E_a}{RT}\right) t} \quad (1)$$

With D_0 the diffusivity coefficient and E_a the activation energy, T the heat treatment temperature and t the duration.

As hardness decreases monotonically during ageing, two ageing conditions yielding the same hardness are assumed to correspond to the same diffusion length:

$$L_1 = L_2 \Rightarrow \exp\left(-\frac{E_a}{RT_1}\right) t_1 = \exp\left(-\frac{E_a}{RT_2}\right) t_2 \quad (2)$$

The activation energy can then be determined by plotting all experimental hardness data as a function of an equivalent ageing time at a reference temperature and adjusting the activation energy to collapse the data onto a single master curve.

$$t_{eq} = t_2 \exp\left(\frac{E_a}{R} \left(\frac{1}{T_{ref}} - \frac{1}{T_2}\right)\right) \quad (3)$$

Once the activation energy is identified, the mechanical properties resulting from any heat treatment can potentially be predicted by calculating the corresponding diffusion length, or equivalently the equivalent ageing time at a given temperature, and comparing it with the experimental database. It is subsequently possible to verify the physical relevance of this activation energy. In this context, it is expected that the activation energy should be relatively close to the activation energy for the diffusion of silicon and magnesium in

the aluminum matrix, which are 118 kJ/mol and 121 kJ/mol, respectively (calculated with ThermoCalc TCAL9).

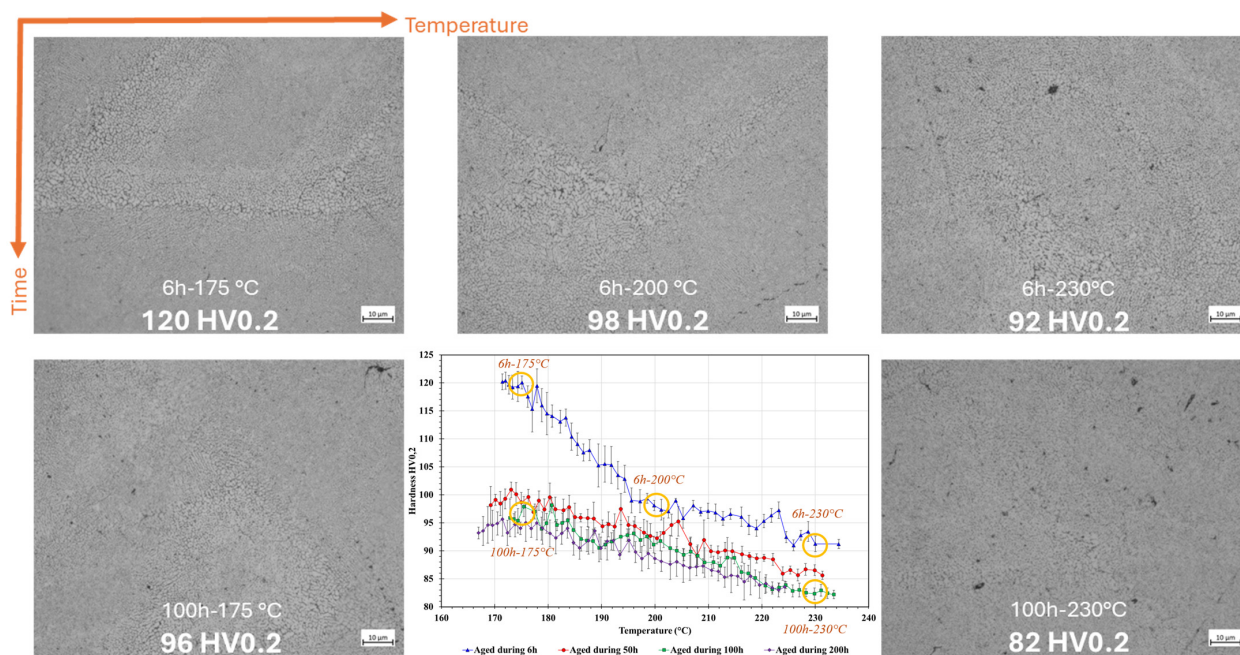


Figure 4. Vickers microhardness as a function of temperature for different ageing times associated with optical micrographs (OM) observations of the areas of the thermally gradient rods exhibiting the greatest differences in hardness (10 μm scale bar).

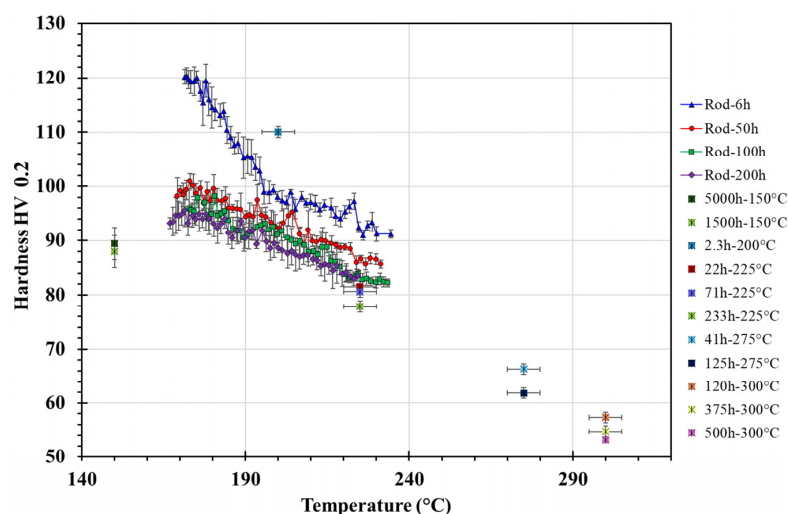


Figure 5. Vickers microhardness as a function of temperature for different ageing times.

Based on complementary ageing conditions from Table 2, all hardness results are presented in Figure 5. Since the rods had not undergone any prior heat treatment (as-built state), it is logical to observe a peak in hardness after ageing at 170 $^{\circ}\text{C}$ for 6 h, which corresponds to the industrial standard for achieving the best mechanical properties in LPBF-produced AlSi7Mg0.6 alloy. Beyond this peak, in the overaging regime, the coalescence and dissolution of precipitates and Si lead to a progressive decrease in hardness.

The build containing the funnel-shaped specimens could not be fully manufactured due to swelling observed during the fabrication of one of the samples. Excessive temperatures led to the formation of a molten material bubble (see Figure 2b), preventing further powder spreading and halting the process at 120 mm height. Only the small feeders

(160 mm) were partially fabricated and studied. Hardness measurements were taken over the entire height of 11 samples, with results shown in Figure 6. Hardness generally decreases along the sample length, with maximum values (~120 HB) at the build platform and minimum values (~45 HB) near the top of the S-HA-SB sample. Funnel-shaped samples with larger shaft diameters exhibit less pronounced hardness reduction, while a finer cone angle also mitigates hardness decrease. The hollow cone sample (H-HA-SB) demonstrates higher hardness than its solid counterpart with identical shaft diameter and cone angle (S-SA-SB). The sample featuring a hollow shaft and pinched cone has unusually low overall hardness, despite its large shaft diameter (20 mm) and low cone angle (15°), parameters typically associated with high hardness. Notably, its section at the pillar matches the slender samples (8.7 mm diameter). At the pinched region, hardness rises compared to adjacent zones. Optical microscopy was used to investigate the microstructure at different hardness levels, as shown in Figure 6. No discernible differences were observed between areas with 111 and 85 HB. Silicon coalescence becomes apparent from 72 HB, but is pronounced only at 45 HB, where the α -Al matrix is largely free of precipitates, and silicon appears fully spheroidized.

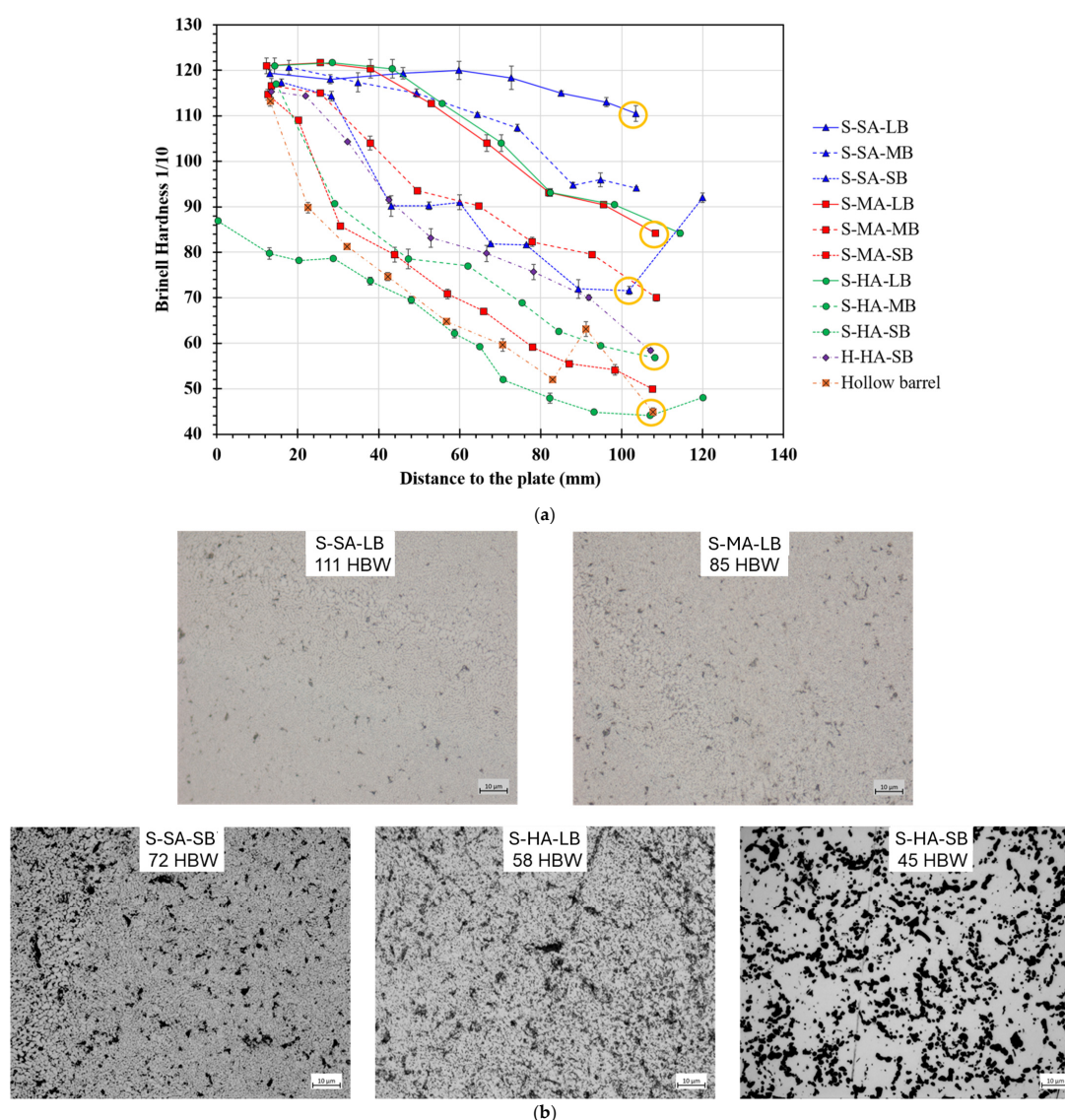


Figure 6. (a) Hardness evolution as a function of the distance from the plate for different funnel-shaped test samples based on dimensions of Table 3 with (b) associated OM for the top of some built conditions (orange circles), (10 µm scale bar).

Two samples were prepared by electropolishing thin discs for TEM observations. These were the samples obtained from thermal gradient rods with the highest hardness (170 °C-6 h-120 HV0.2) and one with the lowest hardness (240 °C-100 h-82 HV0.2). The samples were first observed by SEM after electropolishing (see Figure 7a,b). The coalescence and spheroidization of the silicon network are already visible under SEM after ageing at 240 °C-100 h. Comparing Figure 7c,d, the cells of the 240 °C-100 h sample do not appear to contain the fine precipitates (1 to 5 nanometers diameter) present in the [170 °C-6 h] sample, which are presumably β'' -Mg₅Si₆.

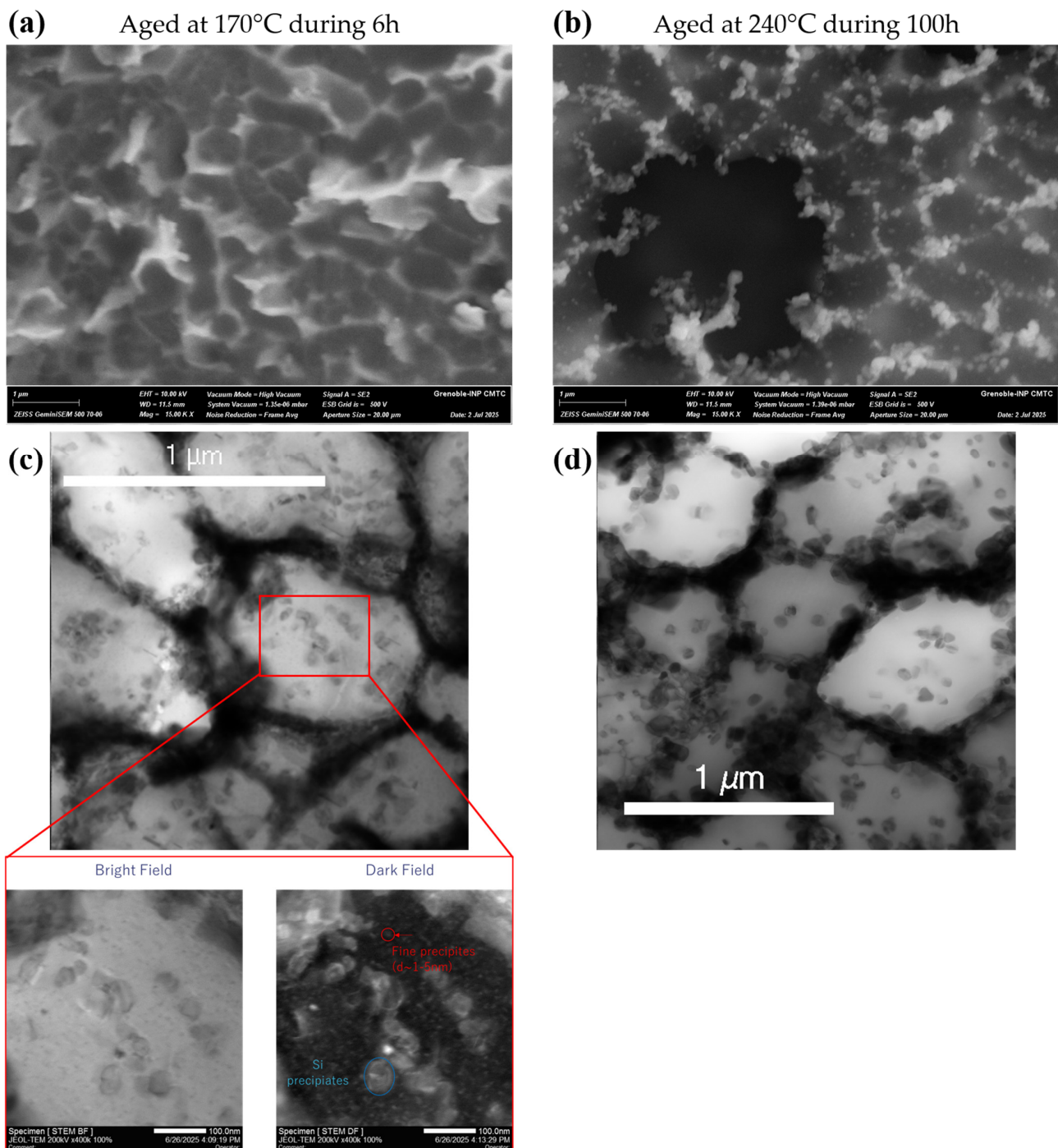


Figure 7. (a) SEM observation of samples (170 °C-6 h) and (b) (240 °C-100 h) from thermal gradient rods, after electropolishing; (c) bright and dark-field zooms of the interior of an α -Al cell from the sample [170 °C-6 h] observed by TEM; (d) TEM observation of the sample [240 °C-100 h].

Hardness evolution as a function of equivalent time was investigated for the four thermal gradient rods (Figure 4) and complementary samples according to Table 3, as described in Section 2. The fitted activation energy is 125 kJ/mol. Figure 8 illustrates the hardness variation at 200 °C for all four rods and complementary samples as a function of equivalent time. For instance, achieving a hardness of 80 HV0.2 at this temperature requires an ageing duration slightly exceeding 1000 h. These curves can be generated for a wide range of selected tempering temperatures, allowing direct comparison of hardness evolutions under different heat treatment conditions. The application range is nevertheless restricted in temperature, with a maximum temperature around 300 °C beyond which the equilibrium volume fraction of the β'' -Mg₅Si₆ phase drops. Parameters for the fit of this experimental curve have been identified with a double-Gompertz law as detailed in the Appendix (Tables A1–A3). The Gompertz model is an empirical sigmoidal equation commonly used to describe processes with an initial lag phase, a rapid growth stage, and a final deceleration, which is quite correlated to metallurgical kinetics occurring in this context.

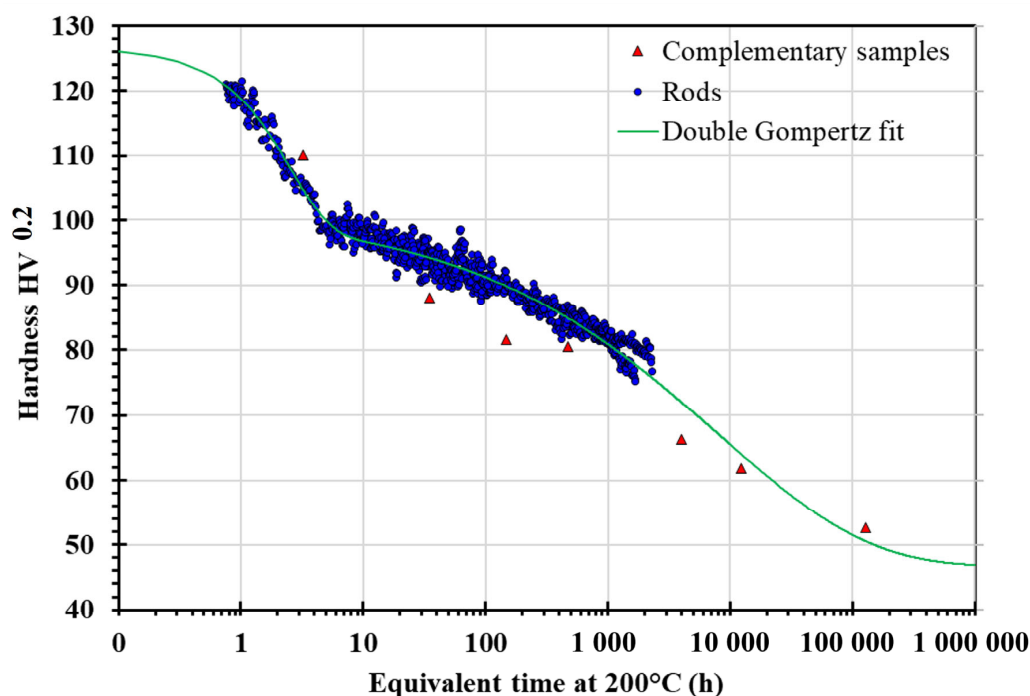


Figure 8. Hardness as a function of equivalent time for the 4 thermal gradient rods and complementary samples. $T_{\text{Ref}} = 200\text{ °C}$, $E_a = 125\text{ kJ/mol}$.

3.3. Mechanical Properties Relationships

Complementary ageing treatments were performed on additional rods to complete the curve in Figure 4, see Figure 5, and to complete Figure 9. The ageing conditions were selected using the Arrhenius-based model as described in Section 3.2. Targeted hardness values were chosen to supplement mechanical strength (UTS and YS) curves as a function of Brinell hardness (HB) (as shown in Figure 9). Conversions between Vickers hardness (HV) and Brinell hardness (HB) were carried out in accordance with ISO 12865:2013(F) [34] for aluminum alloys.

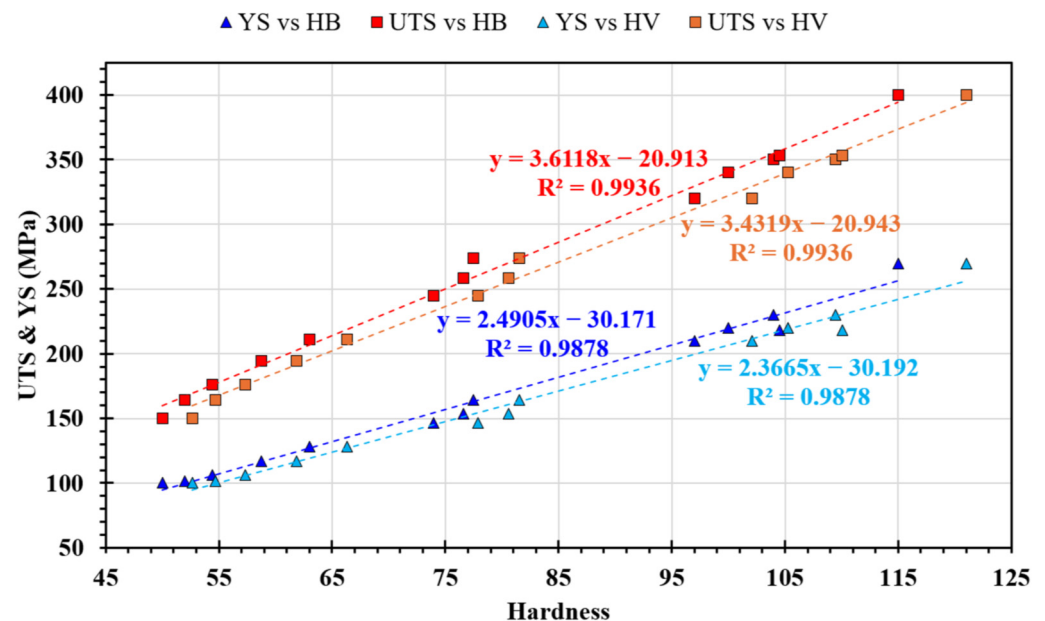


Figure 9. Ultimate tensile stress UTS and yield stress as a function of hardness (HB or HV) for AlSi7Mg0.6 samples manufactured by LPBF.

4. Predicting the Hardness of AlSi7Mg0.6 LPBF-Built Parts

A numerical approach has been developed to predict local hardness in laser-based additive manufacturing of alloys sensitive to thermal ageing. This methodology integrates a calibrated thermal simulation—based on the Arrhenius-type law from Section 3.2—correlating the local thermal history during the layer-by-layer process with final microstructural and hardness evolution. The process involves:

- Simulating the temporal evolution of the temperature field during LPBF and final cooling (to reach ambient temperature) to estimate local exposure times and peak temperatures.
- Establishing a one-to-one correspondence rule between equivalent thermal exposure and hardness, validated experimentally.
- Optimizing process parameters (laser power, scanning speed, platform temperature, support configuration) to minimize discrepancies between predicted and measured hardness.
- Iteratively refining the support design and simulation model until all estimated hardness values exceed a predetermined threshold.

This predictive tool enables optimization of support geometries and build parameters, minimizing trial-and-error and reducing development costs.

The comprehensive predictive approach permits optimization of support configuration and process parameters while ensuring homogeneous hardness. The methodology integrates calibration of a finite element thermal model using experimental hardness measurements on support elements, followed by numerical simulation of temperature fields and local thermal histories throughout the build. By correlating equivalent exposure times at reference temperatures—calculated via an Arrhenius-type relationship—with post-ageing hardness data, the model establishes a direct correspondence between local temperature profiles and hardness. Iterative adjustment of key parameters (e.g., laser power, scan strategy, platform/gas temperature, powder compaction) minimizes discrepancies between simulated and measured properties and hardness. The protocol could also account for heat transfer through residual powder, and it enables real-time thermal monitoring for improved calibration. A schematic flowchart of the calculation process for post-processing thermal simulation results during the manufacturing of LPBF components

to locally predict the final hardnesses obtained on the parts is presented in Figure 10. This process can be summarized in six steps:

1. Selection of the study area at the node of the mesh resulting from the finite element calculation.
2. Determining the temperature change experienced by the part during the build.
3. Discretization of the previous curve.
4. Calculation of the equivalent time to obtain the same hardness at another (fixed and arbitrarily) chosen temperature based on Equation (3) with $E_a = 125$ kJ/mol.
5. Calculation of the total equivalent time at the chosen temperature by summing all the previous time periods.
6. Calculation of the final local hardness from the total equivalent ageing time using the fitted double-Gompertz master curve shown in Figure 8 and detailed in Appendix A (Figure A1).

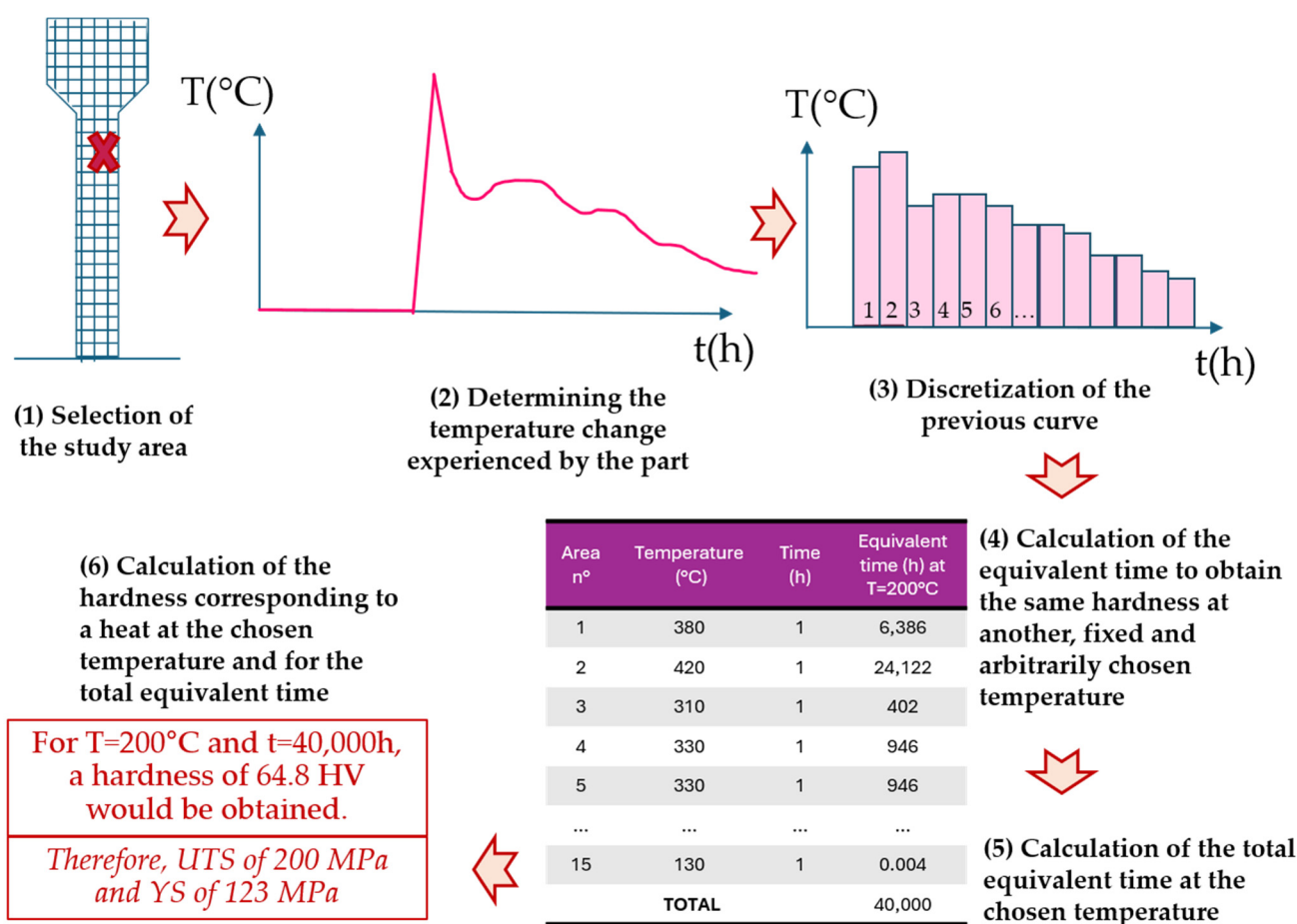


Figure 10. Flowchart of the calculation process for post-processing thermal simulation results during the manufacturing of LPBF components to locally predict the final hardnesses obtained on the parts.

Figure 11 is an extract of some thermal simulations with post-treatment for hardness predictions on parts for funnel-shaped specimen geometries according to different boundary conditions in the heat balance layer-by-layer process. Accurate hardness prediction requires accounting not only for the build duration but also for the cooling period after the end of laser processing, because significant ageing may continue until the part returns to ambient temperature. The post-build cooling stage contributed between 15% and 35% of the total equivalent ageing time, depending on geometry and build duration, emphasizing its significance in predicting final hardness.

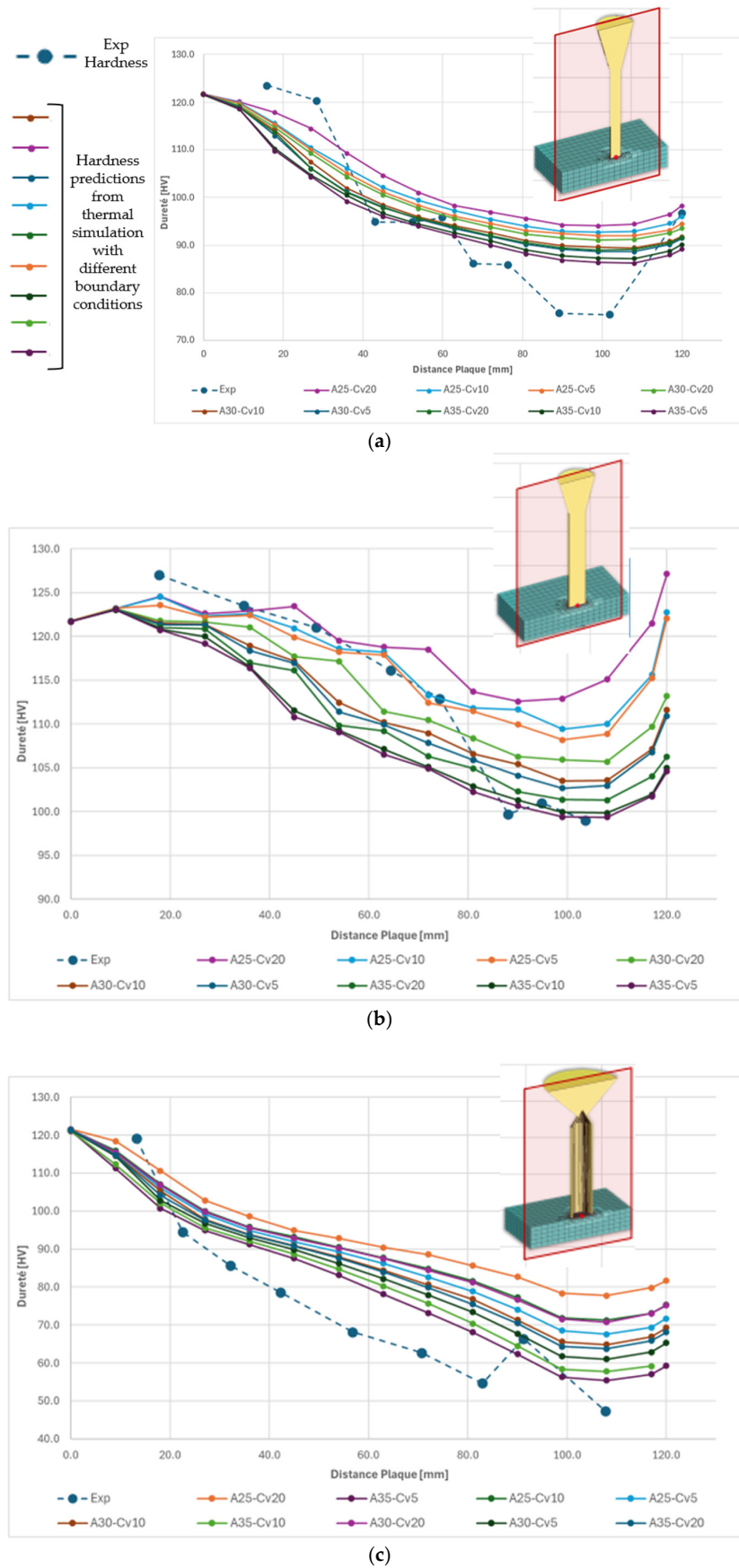


Figure 11. Comparison between experimental hardness and thermal simulation with post-processing in predicting final hardness obtained on three types: (a) S-SA-SB, (b) S-SA-LB and (c) Hollow barrel parts for funnel-shaped specimen geometries.

5. Discussion

5.1. Inhomogeneity in Hardness and Properties Resulting from LPBF Manufacturing

LPBF parts frequently display vertical gradients in hardness and microstructure, with the lower regions—proximal to the heated build platform—being susceptible to over-ageing and coarsening of β'' -Mg₅Si₆ precipitates and the eutectic silicon phase [6–8,12]. Indeed, right after the fabrication of this alloy, a gradient of mechanical properties along the height of samples is often observed. From hardness profiles, microstructural observations and DSC analyses, it has been deduced that pieces built by LPBF are exposed to in situ hardening, which is inhomogeneous. Indeed, a Differential Thermal Analysis (DTA) has been performed by Chambrin et al. [7]. An exothermic peak at around 230 °C has been observed only in the top region of the samples. Thanks to the DSC analysis of Fiocchi et al. [14], this peak has been identified as being linked to the formation of β'' precipitates. As this peak does not appear in the middle or at the bottom of the samples, it has been deduced that these regions are exposed to in situ ageing that maximizes the hardness by the formation of β'' precipitates. The bottom region, close to the build platform, often displays lower hardness than the middle of the sample [6–8,12]. It has been determined that this region has been exposed to over-ageing conditions: β'' precipitates are present, and the eutectic network is coarser. This second element is responsible for lower hardness. The in situ ageing conditions applied at the bottom of the samples simply correspond to the fabrication time of the samples and the temperature of the built platform [11]. These trends are greatly dependent on the manufacturing parameters, particularly the temperature of the built platform.

Residual tensile stress is observed at the surface of samples built by LPBF [32]. This is due to the very fast cooling of the sample after the melting of the powder. At the macroscale, the heat is quickly evacuated from the core of the sample, but it stays stuck at the boundary of it because of the bad heat dissipation in the powder. This gradient of temperature in samples built by LPBF is the opposite of that the one observed in cast and quenched aluminum parts, the surface cools faster than the core whereas in LPBF parts, the heat evacuates faster by conduction from the core to the surface of the part than from the surface to the surrounding (by convection with air, by conduction with the powder or by conduction with the built platform). Residual tensile stress at the surface can cause the opening of small surface cracks, thus the premature break of samples. It is then important to release this residual stress. It can be completely released by a SHT as shown in [6]. Residual compression stress is even observed at the surface of these samples with the associated quench [35].

The realization of a SHT allows solving these two issues, but the mechanical properties obtained after this heat treatment are poor. The hardness, ultimate tensile strength and yield strength can significantly drop after an SHT of 505 °C during 4 h according to [7]. This value is approximate since it depends on the temperature of the built platform and the localization of the considered zone (bottom or top of the sample). The decrease in the mechanical properties is attributed to the coarsening of eutectic silicon and the dissolution of the hardening β'' -Mg₅Si₆ precipitates.

5.2. Ageing Effects on Mechanical Properties

The Arrhenius-based model developed in this study has proven to be highly effective in predicting the evolution of hardness in AlSi7Mg0.6 parts produced by selective laser melting (SLM) and subsequently subjected to various post-process ageing treatments. The calculated activation energy (125 kJ/mol) is in strong agreement with known diffusion coefficients for silicon and magnesium in aluminum, substantiating a diffusion-driven mechanism for the observed hardness variations. A notable finding is the steeper slope

(Figure 8) of the time–temperature–hardness relationship for specimens experiencing a 6 h anneal at temperatures below 195 °C. This regime is classically dominated by precipitation hardening due to β'' and β' phase formation. However, the possible concomitant dissolution of metastable precipitates may necessitate a more nuanced distinction in modelling, as their simultaneous occurrence complicates the interpretation of thermally activated processes.

The model demonstrates robustness even under more severe thermal exposures (275–300 °C), yielding hardness values consistent with experimental observations. Minor discrepancies in the model gradient may be partially attributed to a 5 °C uncertainty in oven control during sample preparation, demonstrating that a ± 5 °C uncertainty can yield up to ± 2 HV deviation in hardness. Remarkably, the high-throughput experimental approach underpins the model's reliability across a broad range of process conditions, as the superposition of multiple thermally activated phenomena follows the same diffusion energy barriers.

Contrary to expectations from the classical literature [7,11], where distinct hardness plateaus and multiple slope transitions are often reported in TTT curves, the current results exhibit fewer clear inflexion points, as in Figure 5. In the as-built condition, unaged LPBF AlSi7Mg0.6 specimens exhibit a baseline hardness near 100–110 HV, contingent on process parameters; thus, any region exhibiting <100 HV is indicative of over-ageing, as confirmed across virtually all investigated funnel-shaped specimens. Direct ageing at 170 °C for 6 h is expected to increase the as-built hardness by approximately +20 HV.

Of particular significance is the influence of funnel-shaped specimen geometry. Risers with a large shank diameter and low draft angle consistently maintain hardness above 110 HV through much of their height, reflecting favourable thermal history with exposure to near-plateau temperatures (~ 150 °C). Uniform hardness at the build platform interface across most risers suggests highly consistent temperature–time exposure at the base, except for geometries with highly constricted cross-sections, where excessive local heating (>230 °C) produces significant hardness reductions (<90 HV).

A gradient in hardness with increasing height was systematically observed, consistent with progressively harsher thermal histories, and corroborated by optical microscopy that shows the onset of eutectic silicon spheroidization as hardness drops below 75 HB, becoming extensive at 45 HB. For intermediate hardness ranges (85–110 HB), no microstructural distinctions are observable optically, yet TEM data confirms coarsening and dissolution of hardening phases.

Sectional geometry emerges as the principal factor in thermal and therefore hardness evolution, surpassing the effects of built time or platform temperature alone. Specimens with a relatively constant section maintain higher hardness, highlighting the necessity to control heat extraction pathways. The observed effect in heat dissipation—faster at the core than near the edge—is evidenced by radial hardness profiling and is consistent with the presence of residual tensile stresses at SLM component surfaces.

Anomalous phenomena, such as increased hardness at the necking of hollow risers, are explained by localized heat trapping, complicating thermal management and ultimately halting part build due to defect formation. The overall hardness reduction correlates directly with geometric constrictions, further validating the thermal bottleneck hypothesis. Direct ageing heat treatments (150–180 °C, several hours) optimize mechanical performance only for parts without significant in situ ageing and fail to completely relieve surface tensile residual stresses. In contrast, a full solution treatment (505 °C, tens of minutes) followed by artificial ageing (150–175 °C, ~ 10 h) enables both stress relief and mechanical property recovery. Adjusting the build-plate temperature poses a trade-off between minimizing residual stress (higher plate temperatures) and maximizing core property retention (lower

plate temperatures). The current study helps to attain conformal jobs with an 80 °C build-plate temperature.

Numerical simulations based on the Arrhenius law successfully forecast final hardness as a function of thermal exposure, confirming the exponential dependency on ageing time at a fixed temperature. Ultimately, effective heat dissipation during LPBF is paramount to achieving homogeneous and optimal mechanical performance in AlSi7Mg0.6. Laser-induced heat is predominantly evacuated through the build plate; geometric features that impede or localize heat—specifically abrupt section changes—are detrimental to hardness and fatigue properties. While supported structures can enhance heat flow, they insufficiently counteract the negative impacts of increased layer thickness (to reduce building time and heat accumulation), which tends to reduce both hardness and fatigue performance. Porosity increases with thicker layers have a negligible effect on hardness, but potential adverse impacts on fatigue cannot be ruled out.

6. Conclusions

- Direct ageing treatments at 150–180 °C for a few hours maximize mechanical properties in as-built alloys, though surface residual tensile stresses persist. Two thermally activated mechanisms are confirmed during both in situ and ex situ ageing: (1) coalescence and evolution of the eutectic silicon network, and (2) precipitation hardening due to nanometric precipitates.
- Eutectic silicon forms a continuous network at peak mechanical properties; prolonged exposure above 170 °C leads to larger cells and network discontinuity, with coarse (>1 µm) Si precipitates appearing below 70 HV of hardness.
- Nanometric hardening precipitates are observable by TEM in as-built samples, but become undetectable after substantial ageing (e.g., 230 °C for 200 h).
- Arrhenius-based numerical modelling reliably predicts the time–temperature dependence of alloy hardness.
- Efficient heat evacuation during LPBF build of AlSi7Mg0.6 is crucial for achieving homogeneous mechanical properties; large and continuous support sections, providing stable heat-conduction paths to the build platform, improve heat extraction and reduce local over-ageing.
- Spatially resolved hardness predictions allow identification and avoidance of regions below critical hardness thresholds, supporting automated quality validation. This process significantly reduces costly prototyping cycles, enabling reliable and efficient manufacture of components meeting strict mechanical property requirements.

Nevertheless, it should be noted that the present study is limited by the scope of the experimental geometries and process parameters investigated, as well as by the absence of comprehensive ductility, fatigue, and in-depth defect characterization. Future research should focus on the statistical validation of the proposed models across a broader range of industrial component shapes, the integration of non-destructive evaluation techniques, and the direct correlation of microstructural evolution with ductility, fatigue, and in-service performance.

7. Patents

P. H., L. F. and R. B. are co-inventors on a utility patent application that includes the numerical methodology, the alloy compositions, processing conditions, and the microstructural features described in this paper.

Author Contributions: P.N.: Writing—review and editing, writing—original draft, visualization, methodology, investigation, conceptualization. A.D.: Writing—review and editing, visualization, supervision, investigation, conceptualization. P.H.: Writing—review and editing, supervision, resources, project administration. R.B. (Romain Bergeron), P.M., T.P., E.L., L.F. and R.B. (René Billardon): Writing—review and editing, visualization, supervision, investigation, conceptualization. All authors have read and agreed to the published version of the manuscript.

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Abbreviations

The following abbreviations are used in this manuscript:

AA	Artificial Ageing
AM	Additive Manufacturing
DA	Direct Ageing
LPBF	Laser Powder Bed Fusion
SHT	Solution Heat Treatment
SLM	Selective Laser Melting
WQ	Water Quench

Appendix A

Identification of a Double-Gompertz Law for Age Hardening of AlSi7Mg0.6 (L-PBF)

Problem Statement and Objective:

To identify the parameters of a double-Gompertz law representing Vickers microhardness (HV0.2) as a function of $x = \ln(t_{eq})$, based on experimental data (file: new7.txt). The fitting procedure was performed under constraints (asymptotic values and imposed points with tolerances) and employed a specific weighting for certain ranges of (x) .

Data

All hardness points have been implemented into a master file:

- Number of points: 1204
- Domain: (x) in $([-0.267; 7.749])$, (H) in $([75.2; 121.4])$ HV0.2
- Weights applied to data points:
 - $(w = 3)$ for $(x < 1.8)$
 - $(w = 1.5)$ for $(7.2 < x < 8.0)$
 - $(w = 1)$ otherwise

Model and Constraints

Imposed model (double-Gompertz):

$$H(x) = H_0 + dH_1 * \exp(-\exp(n_1 * (x - x_1))) + dH_2 * \exp(-\exp(n_2 * (x - x_2)))$$

Constraints (with tolerances)

- Asymptote at $(-\infty)$: $(H(-\infty) = 128 \pm 2)$ HV, i.e., $(H_0 + dH_1 + dH_2 \in [126, 130])$
- Asymptote at $(+\infty)$: $(H(+\infty) = 46.5 \pm 2)$ HV, i.e., $(H_0 \in [44.5, 48.5])$
- Fixed Point 1: $(H(11.75) = 52.6 \pm 2)$ HV, i.e., $(H(11.75) \in [50.6, 54.6])$
- Fixed Point 2: $(H(8.3) = 70 \pm 2)$ HV, i.e., $(H(8.3) \in [68, 72])$

Identification Procedure

1. Data reading, cleaning, and sorting by (x) .
2. Assignment of weights $(w(x))$ according to the specified regions.
3. Definition of the objective function: weighted least squares (SSE).
4. Non-linear constrained optimization (bounds and inequalities) carried out using SLSQP, with multiple initializations to avoid local minima.
5. Evaluation: error metrics and diagnostic plots (fit curve, residuals, histogram, QQ-plot).

Results**Optimal Parameters:****Table A1.** Identified parameters for the fit.

Parameter	Value
(H_0)	46.661
(dH_1)	25.872
(dH_2)	54.820
(n_1)	1.3436
(x_1)	0.9346
(n_2)	0.3557
(x_2)	9.0285

Verification of Constraints:**Table A2.** Values for constraints into the fit.

Quantity	Value
$H(-\infty) = H_0 + dH_1 + dH_2$ [HV]	127.35 HV
$H(+\infty) = H_0$ [HV]	46.661 HV
$H(8.3)$ [HV]	72.000 HV
$H(11.75)$ [HV]	50.600 HV

Fitting Quality (on experimental data):**Table A3.** Confirmation of the fit.

Indicator	Value
RMSE [HV]	1.6487
MAE [HV]	1.2968
R^2	0.9659
Weighted RMSE [HV]	1.6522
Weighted MAE [HV]	1.3016
Weighted R^2	0.9788
Max error $ e $ [HV]	8.3149

Diagnostic Figures

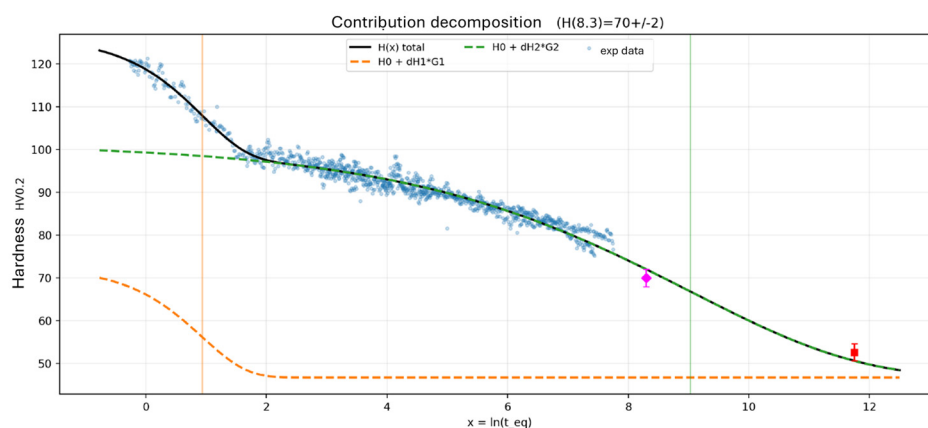


Figure A1. Contribution decomposition for the curve fitting.

Conclusions

The double-Gompertz law was successfully fitted under the imposed constraints (asymptotes and two fixed points) and the specified weighting strategy. Diagnostic plots (residuals and QQ-plot) indicate the quality of the fit. To further constrain the second transition (parameters (n_2 , x_2 , dH_2)), it is recommended to acquire additional experimental measurements for ($x > 7.75$), particularly between 8 and 12.

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