

A study of AlSi5Cu3Mg alloy produced by laser powder bed fusion: porosity assessment, microstructure, mechanical properties

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Abstract. After having determined the LPBF additive manufacturing parameters for the AlSi5Cu3Mg alloy by means of a design of experiment method, three tempers are studied on the manufactured test pieces: as built, direct aging and T6. The study reviews the impact of these three tempers on porosity assessment, microstructure and mechanical properties. It appears that the microstructures in the as built and direct aging tempers are often comparable to those of the AlSi7Mg0.6 and AlSi10Mg alloys which are used as references. However, a significant difference appears with the T6 temper, which does not show any change in porosity for the AlSi5Cu3Mg alloy, unlike the two other alloys. Moreover, due to a high density of type θ'' and/or θ' fine precipitates, the T6 temper features a high yield strength but also an almost isotropic behaviour with good elongation. The analysis of the mechanical behaviour of the AlSi5Cu3Mg alloy in the three tempers is completed with an analysis of the strain hardening rate which is put into perspective with an EBSD analysis of the dislocation density, thus highlighting a close relationship between the microstructures (especially fine dendritic structures) and a high dislocation density. Lastly, a technical and ergonomic study is presented which compares the AlSi5Cu3Mg and AlSi7Mg0.6 alloys. Finally, we explain the interest of the T6 temper for the AlSi5Cu3Mg alloy after LPBF additive manufacturing.

Keywords: Laser powder bed fusion / selective laser melting / AlSi5Cu3Mg aluminium alloy / heat treatment / T6 / microstructure / mechanical properties

1 Introduction

Laser powder bed fusion (LPBF), or selective laser melting (SLM), is an additive manufacturing technique characterised by the fact that the parts are built layer by layer: a laser melts the metallic powder based on the 3D data entered into the computer. This process is now well-known and described in many documents [1].

Many metallic materials are suitable for LPBF additive manufacturing, and aluminium alloys are no exception. However, aluminium alloys which can be processed by LPBF need to have good flowability and weldability properties and must not be susceptible to solidification cracking (interdendritic or intergranular cracking) [2]. This is why the first aluminium alloys available on the LPBF market were casting alloys, such as the historical AlSi10Mg alloy. Today, AlSi10Mg is still the reference aluminium alloy for LPBF [3] along which AlSi7Mg0.6 can be now

found. With their silicon content, these two alloys are well suited for LPBF additive manufacturing; as a matter of fact, the silicon content prevents cracking phenomena [2,4]. During the past few years, more alloys have appeared on the LPBF market, such as the AlSi12 and AlSi9Cu3 alloys, which are also casting alloys and which supplement the standard offer. Later, taking into account the specific features of LPBF, modified alloys have been offered, such as the Al-Mg-Zr alloys (e.g. the Addalloy commercial alloys), Al-Mg-Sc alloys (e.g. Scalmalloy), or alloys with addition of ceramic particles (A20X, RAM alloys). The purpose of the widening of the available offer is to better answer the needs of the market, particularly in terms of mechanical properties, productivity, cost, etc.

The AlSi5Cu3Mg alloy, which is also a casting alloy, can potentially offer interesting properties in LPBF. However it has never formed the subject of studies in this field, unlike the AlSi10Mg, AlSi7Mg0.6, AlSi12 and AlSi9Cu3 alloys which are covered by substantial LPBF literature regarding subjects such as the optimisation of operating conditions [5–11], post-build heat treatments in

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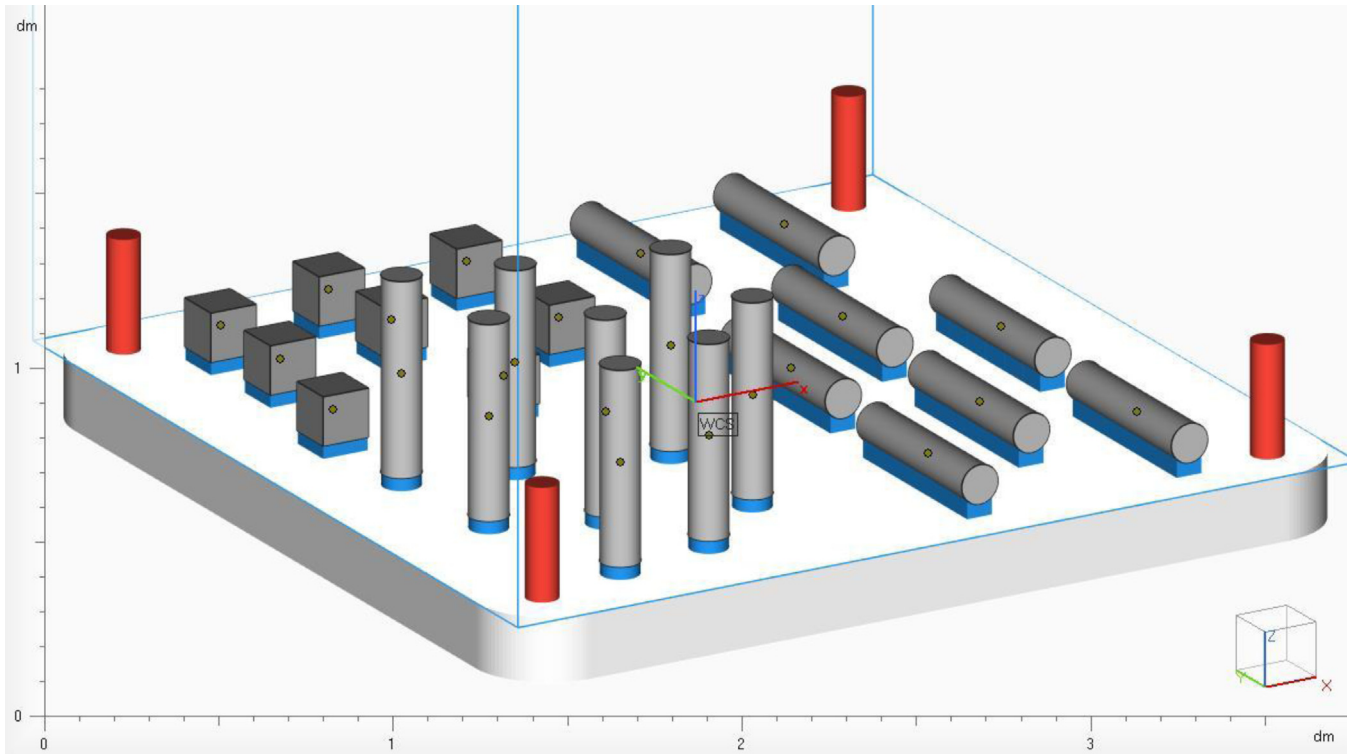


Fig. 1. Example of build platform based on the parameters defined in [Section 2.1](#).

relation to the microstructure and the mechanical properties [3,5,9,12–20], fatigue behaviour [21–25] and industrial applications [26].

The AlSi5Cu3Mg alloy is currently used for the casting of parts for the aerospace or automotive industry (turbocharger turbines) [27]. This alloy has good casting properties, the presence of copper makes machining easier and increases the resistance to high temperature, while magnesium allows the application of heat treatments when high mechanical properties are required [27].

The main objective of this study is thus to present and analyse the static mechanical properties and the metallurgical structures (in relation to the post-build heat treatments carried out) of the AlSi5Cu3Mg alloy after LPBF additive manufacturing. Some industrial data, such as productivity and cost, is also presented in comparison with the AlSi7Mg0.6 alloy.

2 Materials and methods

2.1 LPBF machine

The laser melting machine used for this study is an SLM 280 HL machine (manufactured by SLM solutions) equipped with a YAG laser with a power of 700 W. The build platforms are made of aluminium alloy. The build platform temperature is 150 °C. It should be noted that the temperature of the built platform can have an impact on the state of precipitation of the aluminium alloy at the end of manufacture. However, the temperature of 150 °C, which remains below the conventional temperatures for the artificial aging of foundry aluminium alloys (generally from

160 °C to 200 °C), limits the pre-precipitation effect and makes it possible to maintain an almost uniform hardness in the samples produced especially when the manufacturing height (related to the manufacturing time) is less than 160 mm [28]. This is the case of this study ([Fig. 1](#)). All the tests are carried out with shielding gas (argon with minimum purity of 99.99%). The layer thickness called e is set to 50 μm . The building strategy used is a stripe pattern with rotation of 67° from one layer to another [29–31].

The parameters used for the manufacturing of the AlSi5Cu3Mg alloy test pieces (refer to [Sect. 2.2](#)) are determined via a design of experiments (DOE) approach. The laser power is set to 650 W [32]. Thus, only the manufacturing speed v (in mm/s) and the hatch spacing h (in μm) are used as factors for the DOE. Excessively low or excessively high values of these factors may lead to a lower level of densification of the alloy and, therefore, to fabricated samples with poor mechanical properties [7,8]. We used a Box Behnken design of experiments [33] in order to meet our requirements. The output value or selection criterion of this DOE is the closed porosity measured by means of the Archimedes method. The parameters obtained and, therefore, used to make the AlSi5Cu3Mg test pieces are as follows: $P = 650$ W (fixed), $v = 1800$ mm/s, $h = 140$ μm and $e = 50$ μm (fixed). Thus, the volumetric energy density [32] E_v is equal to 51.6 J/mm³.

2.2 Powder and test pieces

Approximately 100 kg of AlSi5Cu3Mg alloy powder were procured from ECKA Granules, at a price of € 70/kg. The chemical composition of the powders is verified by a physical

Table 1. Technical data of the AlSi5Cu3Mg alloy powder.

Particle size	$D_{10} = 24.3 \mu\text{m}$ $D_{50} = 39.7 \mu\text{m}$ $D_{90} = 63.9 \mu\text{m}$ Dxx: diameter for which xx% of the particles, in volume, are smaller. Span $(D_{90} - D_{10})/D_{50} = 1.0$ Span: factor characterising the width of the particle size distribution [34].												
Morphology	Spherical type, with very few satellites												
		Si	Fe	Cu	Mn	Mg	Cr	Ni	Zn	Pb	Sn	Ti	O
Chemical composition (% by weight)	ECKA	4.7 ± 0.3	0.13 ± 0.01	3.0 ± 0.2	<0.005	0.23 ± 0.01	<0.005	<0.005	0.049	<0.005	<0.005	0.046	0.092
	EN 1706 AlSi5Cu3Mg	4.5–6.0	<0.60	2.6–3.6	<0.55	0.15–0.45	–	<0.10	<0.20	<0.10	<0.05	<0.25	–

method of chemical analysis called ICP-AES (Tab. 1). It complies with standard NF EN 1706.

The particle size distribution of the powder is measured with a Mastersizer 3000 E laser particle sizer from MALVERN in Mie diffusion mode. It can be noted that the measured particle size (Tab. 1) is rather conventional for powders intended for use in LPBF and displays a rather low and centred spread in terms of particle size distribution. A smaller span value indicates a narrower particle size distribution or a shift to coarser particle sizes within the same particle size distribution (when the differences between D_{90} and D_{10} stays constant and D_{50} increases) [34].

The test pieces used for this study are of three types:

- Cube (dimensions $15 \times 15 \times 15$ mm) (Fig. 1).
- Cylinder for tensile testing (cylinder with a diameter of 12 mm and a length of 60 mm), either flat (XY plane) or vertical (Z direction) (Fig. 1).

The manufacturing direction is along the Z axis as shown in Figure 1. The red cylinders (Fig. 1) represent the areas where the build platform is attached to the machine and the blue areas show the manufacturing supports.

2.3 Heat treatments

The equipment used for the heat treatments consists of a forced convection air furnace. This furnace is designed to apply heat treatments on aluminium alloys (temperatures $\leq 650^\circ\text{C}$) and has excellent temperature homogeneity ($\Delta T \leq 6$). Cold water (approximately 20°C) is used as the quenching fluid. The time required to transfer the load (test pieces) to the quenching medium is less than 7 seconds in all circumstances (standard SAE AMS2772E).

The heat treatments performed on the LPBF-processed AlSi5Cu3Mg alloy are as follows:

- As built: no additional heat treatment.
- Direct aging: from 5 h to 18 h at 170°C (to be more precise direct artificial aging on the as built temper).
- T6 temper (solution heat treatment + quenching + artificial aging): 2 h at 510°C + water quenching + 18 h at 160°C . Artificial aging for 18 h at 160°C is a common

alternative to aging for 10 h at 170°C to improve the mechanical properties of the AlSi5Cu3Mg alloy resulting from casting.

2.4 Characterisation tests

The porosity measurements were carried out using the Archimedes or “three weighings” method [3,35]. This is the method most often used for quickly determining the porosity of parts and test pieces made by LPBF. Porosity measurements by image analysis were also carried out on the cube test pieces. These measurements were performed using an Axio Imager M2m (Zeiss) optical microscope [35].

The Brinell hardness measurements were carried out at room temperature with an Emco Duramin 500 testing machine (in accordance with standard EN ISO 6506-1). At least three measurements were performed on each cubic test piece (only the average is given).

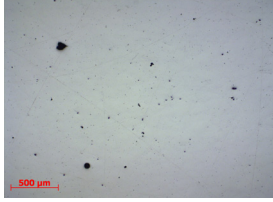
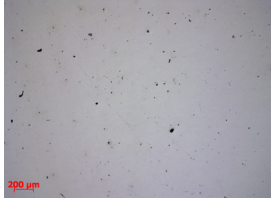
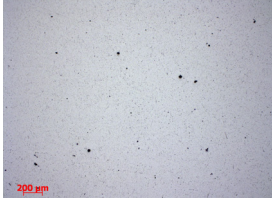
The tensile tests were carried out in accordance with standard ISO 6892-1 at room temperature, using a Zwick Z250 testing machine (250 kN). The testing speed is set to 5 mm/min or 0.003 s^{-1} .

The samples for micrographic examinations were prepared conventionally (cutting, coating, polishing, etc.) before examination using a Zeiss Axio Imager M2m optical microscope. For information, etching for the examinations is carried out using Keller’s reagent.

The samples for scanning electron microscope (SEM) examinations were prepared conventionally and examined using a field emission gun scanning electron microscope (FEG-SEM) JEOL JSM-IT800. This SEM is equipped with an Energy Dispersive X-ray Spectroscopy (EDS) sensor Oxford Ultim Max 100 and an Electron Back Scattering Diffraction (EBSD) camera Oxford CMOS Symmetry. Note that the EBSD examined samples were subjected to a final polishing operation on a vibrating table (Presi Vibrotech 300 model).

In order to measure the electrical conductivity of non-magnetic metals (which is the case of aluminium alloys), we used the eddy current technique. The instrument used is a Fisherscope MMS pc equipped with an ES40 probe, which

Table 2. Porosity rates by image analyses of the three tempers.

	As built	Direct aging	T6
Average (in %)	0.25	0.29	0.31
Standard deviation	0.04	0.13	0.14
Image			

is itself equipped with a thermocouple. Thus, the Fisherscope instrument instantly corrects the electrical conductivity according to the temperature. We used a frequency of 60 kHz in order to penetrate deep into the alloy.

3 Results and discussion

3.1 Porosity rate and chemical composition

The porosity rate is measured for the three tempers (Tab. 2): as built, direct aging and T6 temper. For direct aging, heat treatment for 10 hours at 170 °C is preferably used which highlights a maximum hardness for this configuration (Sect. 3.3)

With the AlSi5Cu3Mg alloy, we found that:

- The porosity rate in the as built temper is in line with the rates usually found in the literature [5,7,8–10].
- The porosity rate does not change with the temper, unlike in the AlSi7Mg0.6 alloy [36].

Several studies focus on the impact of heat treatments, in particular of the T6 temper, on AlSi10Mg alloys from LPBF and more specifically on the evolution of porosity after heat treatment. Currently, there is no consensus on this topic in the literature. Thus, for certain authors there is no evolution of the porosity after solution heat treatment [37–39]; and two works by Majeed [40,41] even show a slight improvement in density. Conversely, Girelli [42], Yang [43] and Bagherifard [44] observe a substantial increase in porosity. Similarly, if we take up the conclusions of Mauduit's study [36] on the AlSi7Mg0.6 alloy, it is stated that the porosity rate significantly increases when the alloy changes from the as-built temper to the T6 temper (from 0.43% to 1.01%). These pores, which increase in terms of quantity and size, are attributed to the gases (hydrogen) trapped in the molten pool during the solidification process. During the solution heat treatment, this hydrogen migrates and forms new (and larger) pores. This conclusion agrees with the analyses of Weingarten [45] on the AlSi10Mg alloy. He also states that this is due to the solution heat treatment temperature. As a matter of fact, hydrogen starts to migrate at approximately 525 °C. As the solution heat treatment temperature of the AlSi7Mg0.6 alloy is equal to 540 °C, this phenomenon appears. However, in the case of the AlSi5Cu3Mg alloy, the solution heat treatment temperature is only 510 °C;

therefore, the phenomenon does not appear. Thus, unlike the AlSi7Mg0.6 and AlSi10Mg alloys, the AlSi5Cu3Mg alloy is perfectly suited for the T6 temper.

The chemical composition of the AlSi5Cu3Mg alloy was checked on a cube type test piece: the analysis was compliant and is presented in Table 3. The chemical composition of the AlSi5Cu3Mg alloy did not change much in comparison with the powder delivered by ECKA Granules. We noticed a slight enrichment in Si (from 4.7% to 5.2% by weight) and a slight loss of Cu (from 3.0% to 2.9% by weight) and Mg (from 0.23% to 0.16% by weight). It is worth noting that the loss of chemical elements is rather common in LPBF, especially for Mg, which has a low boiling point (1091 °C) [2,36]. The slight loss of Cu and possibly the low Si enrichment can be explained by measurement uncertainties (Tabs. 1 and 2).

The AlSi5Cu3Mg alloy after LPBF additive manufacturing has a low Mg content and its chemical composition is close to that of the AlSi5Cu3 alloy (EN 1706) or the A319 grade.

3.2 Metallurgical structure

In the following, we will detail the structures of the three tempers presented in paragraph 2.3.

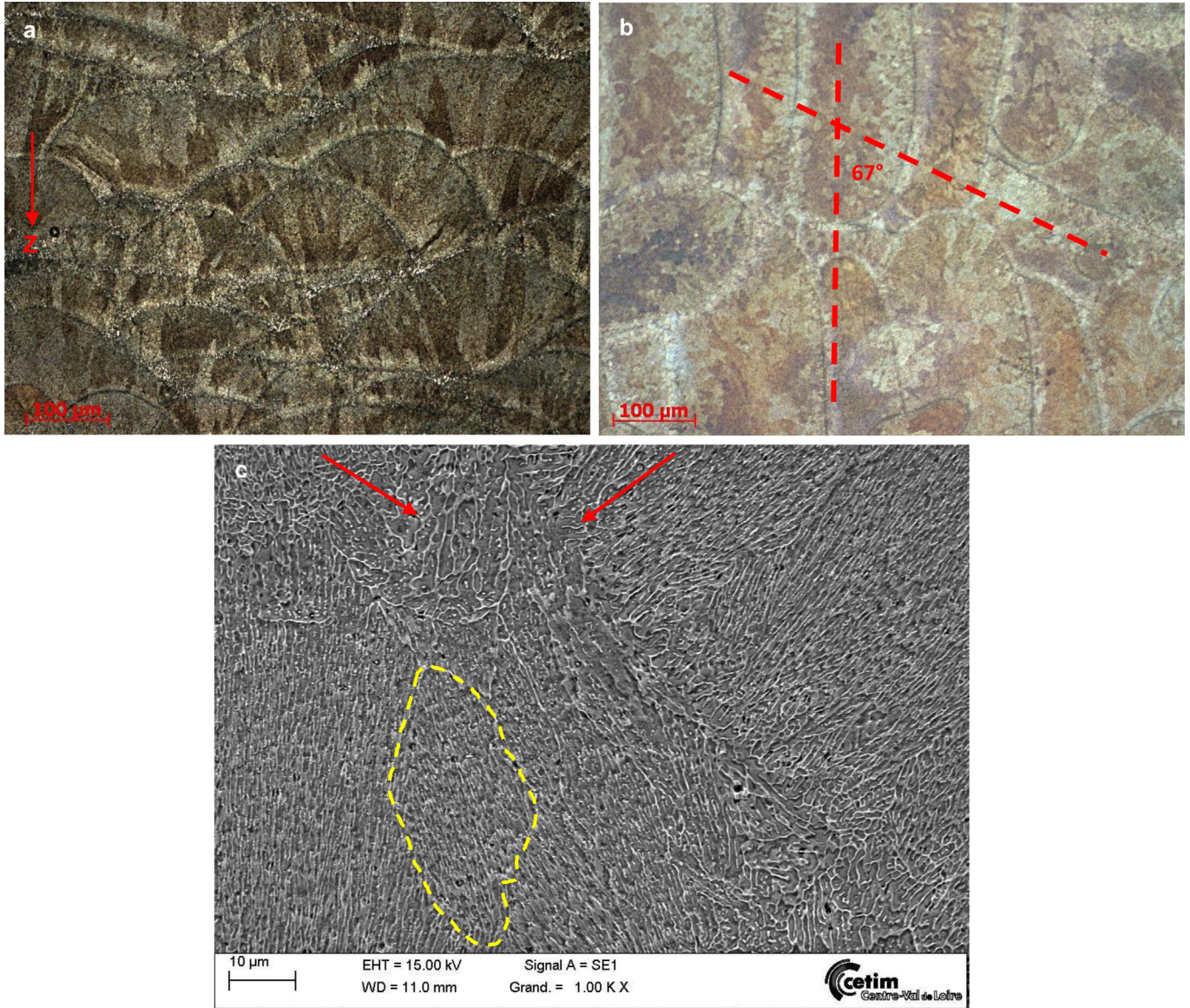
As an introduction, we present, the decomposition sequence of the solid solution of a ternary alloy of the Al-Si-Cu type. Indeed, in view of the very low Mg content of the alloy after LPBF, it can be considered as an alloy of the Al-Si-Cu type and not a quaternary Al-Si-Cu-Mg alloy. Thus, according to Mohamed [46], the majority decomposition sequence involved in hardening is:



The sequence begins with the decomposition of the solid solution and the clustering of Cu atoms; this clustering then leads to the formation of coherent, disk-shaped Guinier Preston (GP) zones. At room temperature aging conditions, GP zones arise homogeneously; these zones appear in the form of two-dimensional, copper-rich disks. After the formation of the GP zones, the θ'' phase corresponding to the formation of platelets parallel to the {100} planes of the matrix is observed and is coherent

Table 3. Chemical composition of the AlSi5Cu3Mg alloy after LPBF – As-built temper.

% by weight	Si	Fe	Cu	Mn	Mg	Cr	Ni	Zn	Pb	Sn	Ti
Test cube	5.2 ±0.3	0.13 ±0.01	2.9 ±0.2	<0.005	0.15 ±0.01	<0.005	<0.005	0.040	<0.005	<0.005	0.030
EN 1706 AlSi5Cu3Mg	4.5–6.0	<0.60	2.6–3.6	<0.55	0.15–0.45	–	<0.10	<0.20	<0.10	<0.05	<0.25

**Fig. 2.** As built temper – (a) Micrographic examination Z direction – (b) Micrographic examination XY plane – (c) FEG-SEM examinations Z direction

with the matrix. As aging proceeds, the θ'' phase starts to dissolve and the θ' phase begins to form. This θ' phase also has a platelet-like shape and is composed of Al and Cu atoms in an ordered tetragonal structure. However, as it grows, the θ' phase loses its coherence with the matrix. The θ phase is the Al_2Cu equilibrium compound which is fully incoherent with the matrix. At room

temperature, aging only leads to the GP zones. With artificial aging below 200°C , the decomposition of the solid solution stops at the θ'' phase or the θ' phase. The maximum level of hardening is obtained with these phases. Beyond this temperature, the θ phase, incoherent with the matrix, forms and results in lower mechanical properties [47].

In the presence of Fe, it is possible to encounter other phases in this alloy, such as $\text{Al}_7\text{Cu}_2\text{Fe}$ and Al_5FeSi [46]. As previously indicated, $\beta\text{-Mg}_2\text{Si}$ and $\text{Q-Al}_5\text{Mg}_8\text{Cu}_2\text{Si}_6$ are not (or very little) present due to the low Mg content [18,19].

3.2.1 As-built temper

Figure 2 shows the build tracks in the XY plane (Fig. 2b) and along the Z direction (Fig. 2a). The observations are comparable to what is documented in the literature on the AlSi7Mg0.6 and AlSi10Mg alloys [3,5,8,10,11,13,36]. That is:

- Along the Z direction (building direction): The build tracks appear in the form of waves – the edge of the track is highlighted.
- In the XY plane: The building strategy is highlighted: the tracks run crosswise to each other with a rotation of 67° between layers.

On Figure 2a, like for the AlSi7Mg0.6 alloy [36], the grains highlighted by the metallographic etching process develop along paths orthogonal to the isotherms, in keeping with the laws of solidification. Thus, the grains are oriented perpendicularly to the tangent of the track edges.

The FEG-SEM examination of the as-built temper (Fig. 2c) revealed fine dendrites or dendritic cells of solid solution aluminium together with a eutectic [3,5,8,10,11,13,36]. The structure is coarser at the track edge (example red arrow): this is an area which was thermally affected by the construction of the neighbouring tracks. We also found that the fine dendrites are oriented in the same way by zones and that these represent grains (example yellow dotted line).

Thus, the as-built metallurgical structure of the AlSi5Cu3Mg alloy is not very different from that of the AlSi7Mg0.6 alloy.

Figure 3 helps to better study the eutectic which comes along the solid solution aluminium. On the backscattered electron image (Fig. 3a), which allows chemical contrast, the morphology of the eutectic appears in two forms: dark (red arrows) and bright (blue arrows). Considering the observation, the bright zones are related to heavier chemical elements, which can be the case of Cu compared to Al, Si or Mg (present in the alloy). The EDS analysis (Figs. 3b–3f) confirms the presence of Cu in the eutectic, in particular the bright detected zones in Figure 3a. Si is also present in the eutectic. Thus, we can conclude that there are probably three eutectics: Al-Si, Al- Al_2Cu and a ternary eutectic of the Al-Si- Al_2Cu type.

3.2.2 Direct aging (10 h at 170°C)

The structure of the alloy has not significantly changed compared to the as-built temper. As before, the following observations can be made:

- Presence of build tracks in the form of waves along the Z direction, and grains perpendicular to the tangent of the track edges.
- Tracks running crosswise to each other at 67° and highlighting the building strategy (rotation of 67° between layers).

- Fine dendrites of solid solution aluminium along with a eutectic whose morphology has not changed.

3.2.3 T6 temper

After application of the T6 temper to the AlSi5Cu3Mg alloy, the metallurgical structure of the alloy significantly changes, even if the “traces” of the build tracks can still be seen (Fig. 4a). We noticed the appearance of silicon polyhedrons, already observed in the AlSi7Mg0.6 and AlSi10Mg alloys [3,36], with sizes ranging from approximately 0.5 to $4\text{ }\mu\text{m}$. It can be noted that these polyhedral compounds are mainly distributed at the grain boundaries (Fig. 4b). Rao et al. [48] indicate that the possible origin of the silicon polyhedrons is due to a supersaturation in Si of the solid solution aluminium in the as-built temper, probably as a result of the very fast cooling rate of 10^5 – 10^6°C/s [32]. Thus, the solubility of Si in the solid solution aluminium exceeds the value of 1.6% (by weight) and reaches values of approximately 5% (by weight). During the solution heat treatment, the supersaturated solid solution aluminium decomposes and the solubility of the silicon “naturally” drops back below 1.6% (by weight) by forming silicon polyhedrons visible in the T6 temper (Figs. 4b and 4c). In part, the Si polyhedrons also come from the destabilization of the Al-Si eutectic by the solution heat treatment.

The FEG-SEM examinations (Figs. 4c and 4d) highlight the following, in addition to the silicon polyhedrons:

- Very few Al-Cu-Fe type acicular intermetallic compounds (probably $\text{Al}_7\text{Cu}_2\text{Fe}$), approximately 3 – $4\text{ }\mu\text{m}$ in size.
- Fine precipitates aligned with the grain boundaries (see example in the red area, in Fig. 4c).

3.2.4 Comparison of as-built, direct aging and T6 temper at high magnification

Figure 5, which presents backscattered electron FEG-SEM images at higher magnification ($50,000\times$) of as-built, direct aging and T6 temper, provides additional elements to understand the evolution of microstructure and precipitation. In Figure 5a, does not give more information than previously, that is to say fine dendrites of solid solution aluminium along with three eutectics (Al-Si, Al- Al_2Cu and Al-Si- Al_2Cu). However, it will be interesting to compare this figure to other temper configurations (direct aging, and T6).

In direct aging, we noted a change in the structure with the appearance of a low density of fine precipitates (example of the red area in Fig. 5b) present in the form of fine platelets (length $< 50\text{ nm}$) crossing each other at 90° . These precipitates are probably of the θ'' type, i.e. a transition phase towards the stable and incoherent Al_2Cu phase [46]. This phase precipitated as a result of the direct aging treatment applied (10 hours at 170°C). We also noted that the three eutectics were still present and were not destabilised by the direct aging treatment. This last

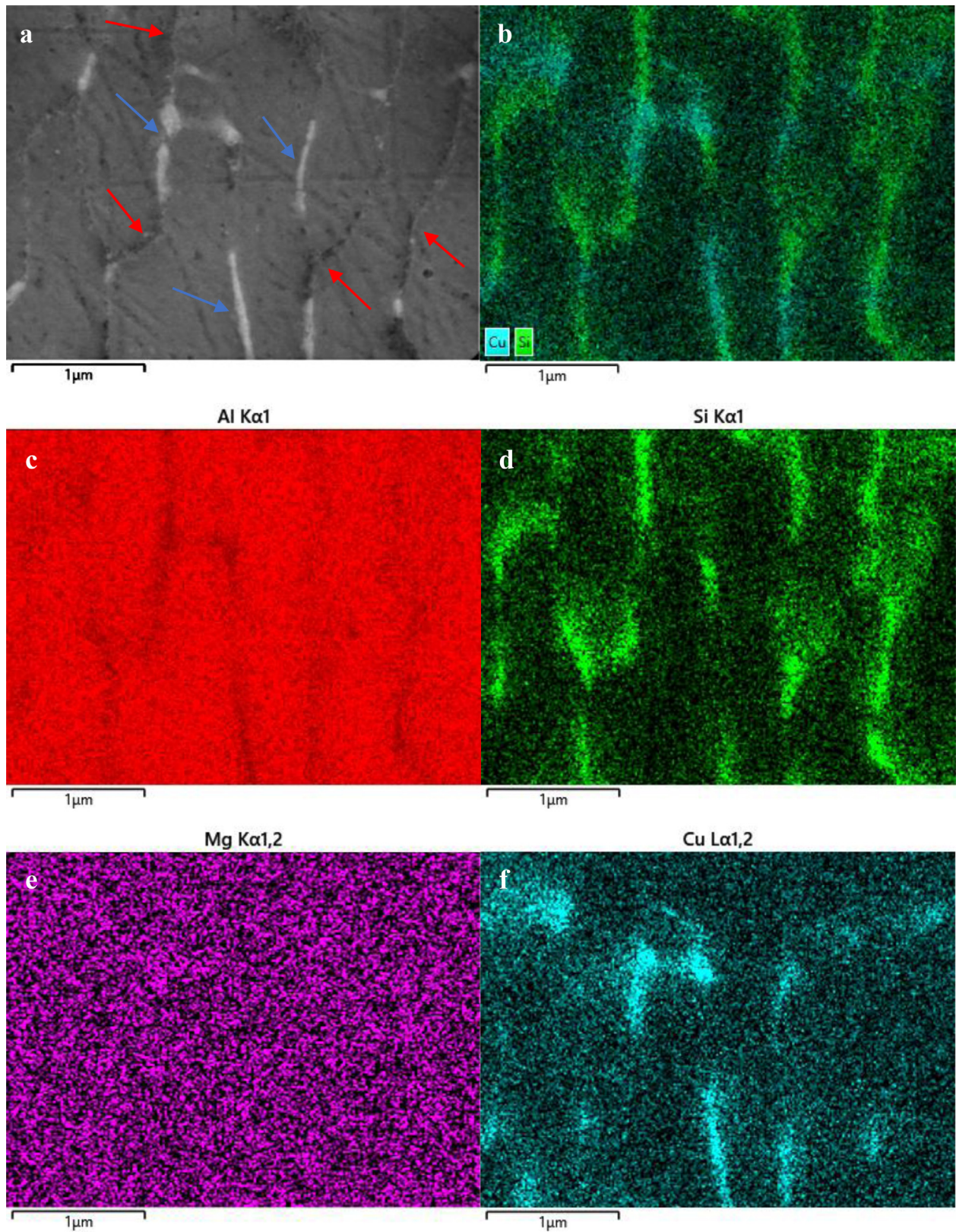


Fig. 3. AlSi5Cu3Mg alloy – Analysis of the eutectic – as built. (a) FEG-SEM examination: electronic image. (b) to (f): EDS analyses (mapping of chemical elements).

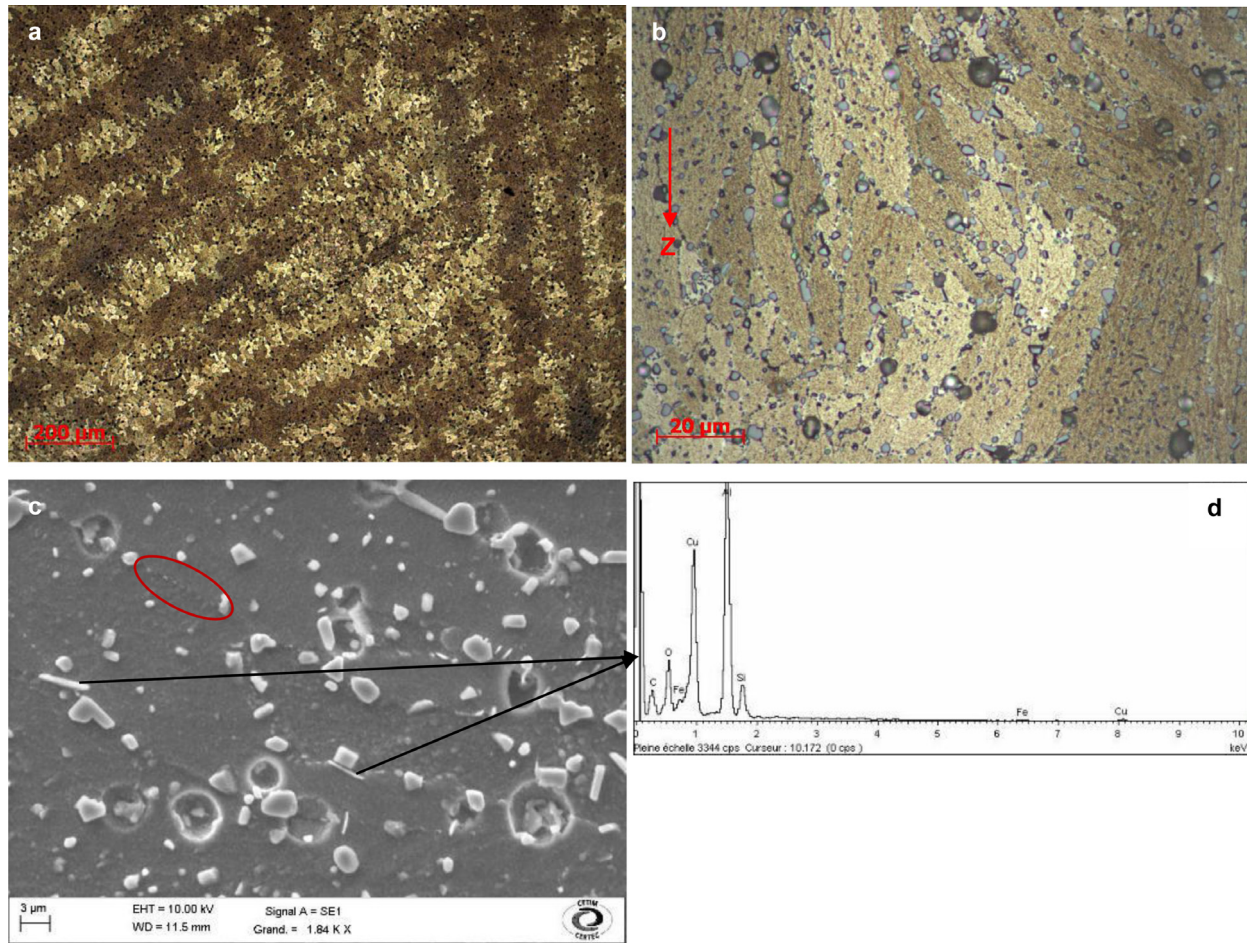


Fig. 4. AlSi5Cu3Mg alloy, T6 temper. (a) Micrographic examination in the XY plane after etching; (b) Micrographic examination along the Z direction after etching; (c) FEG-SEM examinations (secondary electron mode); (d) EDS analysis, chemical element identification spectrum.

point partly explains the low precipitation detected. Indeed, a large quantity of Cu remains trapped in the eutectic and cannot participate in the precipitation.

Based on the information of the precipitation sequence previously described [46], Figure 5c, which presents T6 temper, can be commented as follows:

- We can note the appearance, in very significant density, of fine platelet-shaped precipitates with a length less than 100 nm and crossing each other at 90°: this is probably the θ'' phase.
- Moreover, we can note the presence of a silicon polyhedron (left-hand part of the figure), approximately 1 μm in size (which confirms the previous observations – Fig. 4).
- In the centre of Figure 5c, the examination reveals the presence of a precipitate-free zone on both sides of a grain boundary [47] over a width of approximately 100 nm. Two mechanisms for the creation of these zones denuded of precipitates are proposed, either by depletion of vacancies at the grain boundaries or by depletion of solute [47].
- Now, eutectics are not present (destabilized by solid solution treatment).

Precipitation is “complete” since all the available Cu has been placed in solid solution and has therefore precipitated as a result of the artificial aging of the T6 temper.

3.3 Hardness and electrical conductivity

The T6 temper applied to the AlSi5Cu3Mg alloy features a very high hardness level, much higher than the other tempers previously studied, i.e. as built and direct aging (10 h at 170 °C) (Tab. 4). Thus, as already indicated in Section 3.1, the AlSi5Cu3Mg is suitable for the T6 temper but it also seems to respond well to heat treatments. It is therefore important to study the mechanical properties of this temper. Note that the direct aging temper shows an improved hardness level compared with the as-built temper, however without reaching the hardness level of the T6 temper.

Among the experimental methods for studying the structural precipitation of aluminium alloys, there are macroscopic physical methods that use the macroscopic variations of physical properties (electrical, thermal, dimensional, etc.) of the alloy depending on its state of

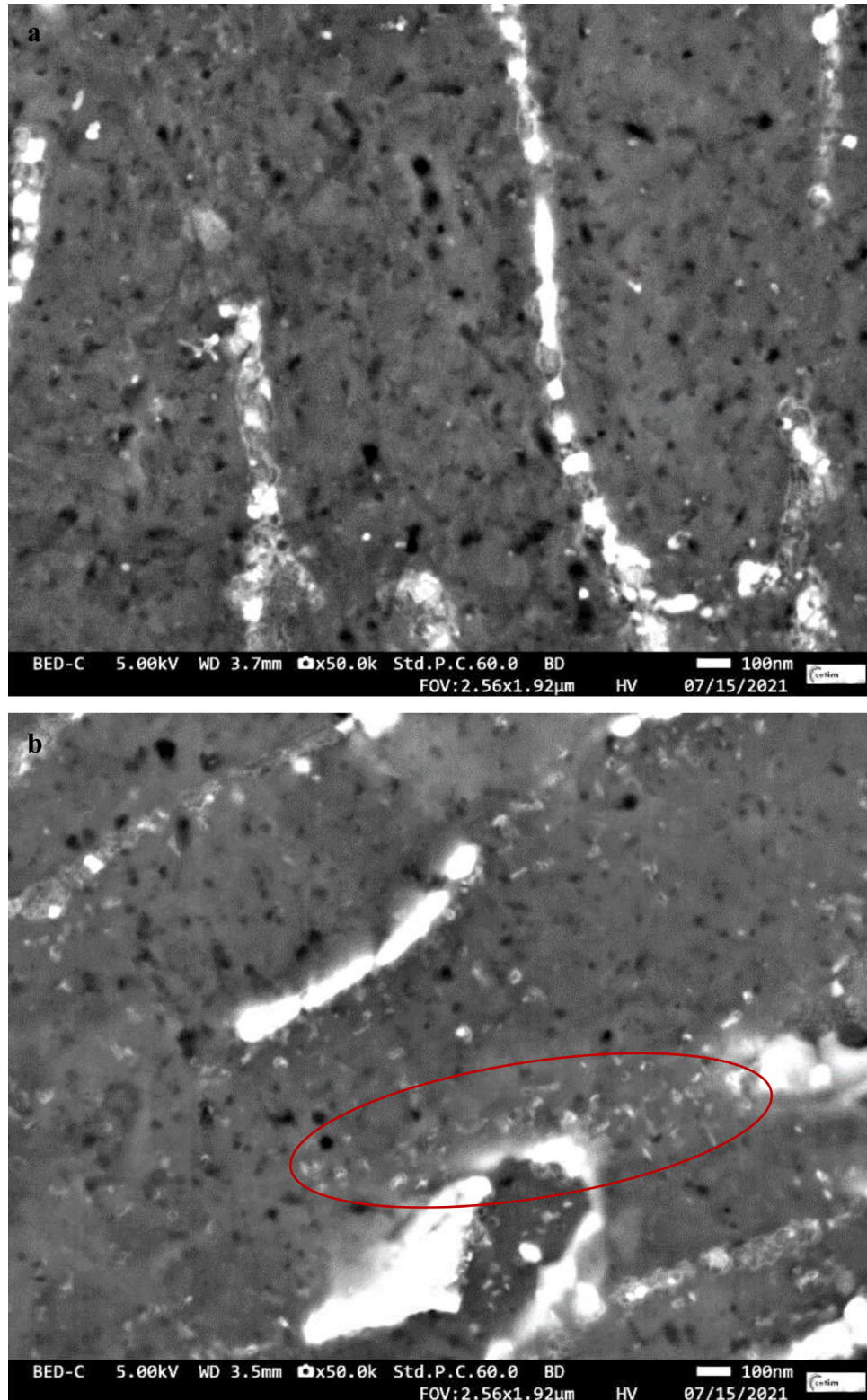


Fig. 5. Backscattered electron FEG-SEM examination (50,000x). (a) As-built AlSi5Cu3Mg alloy. (b) AlSi5Cu3Mg alloy in the direct aging temper (10 hours at 170 °C). (c) AlSi5Cu3Mg alloy in T6 temper.

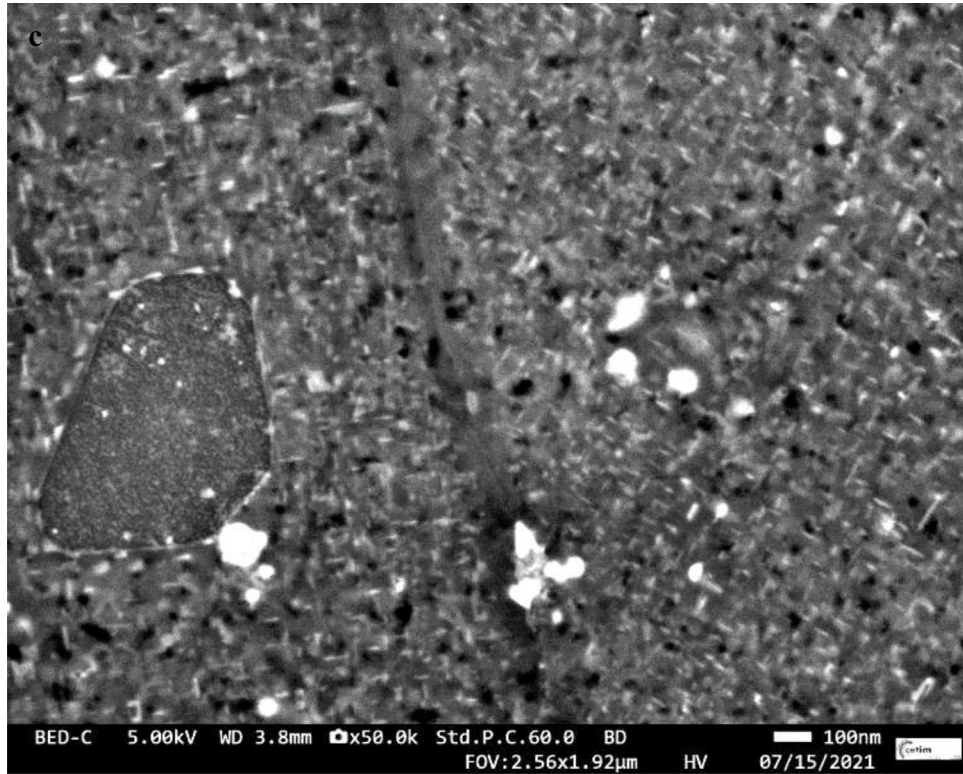


Fig. 5. (Continue).

Table 4. HBW hardness and electrical conductivity of various tempers of the AlSi5Cu3Mg alloy.

Tempers	HBW 5/250 hardness	Electrical conductivity (in MS/m)
As-built	95.8±1.9	19.65±0.02
Direct aging (10 hours at 170 °C)	109±2.3	21.99±0.02
As quenched W (5 minutes) (solution heat treatment + quenching)	74.3±1.4	19.78±0.06
T4 (solution heat treatment + quenching + natural aging)	108.2±1.8	19.20±0.02
T6 (solution heat treatment + quenching + artificial aging)	128.4±2.6	22.27±0.02

precipitation. The electrical conductivity and, conversely, the electrical resistivity are governed by the state of decomposition of the solid solution of the studied aluminium alloy [47].

It can be noted that the electrical conductivities in the as-quenched W temper and in the as-built temper (Tab. 4) are approximately identical. However, it is the as-quenched temper which features the maximum quantity of solid solution elements. We can assume that all the Cu (i.e. 3% by weight) and up to 1.6% of Si by weight are present in solid solution. Nevertheless, as shown by Figure 3, in the as-built temper, a significant quantity of Cu is present in the eutectic, therefore not in solid solution. Thus, with almost equivalent electrical conductivity values, this

implies that an amount of Si above 1.6% by weight is present in solid solution in the as-built temper. We also confirm the assertion of Rao et al. [48] about a silicon supersaturation of the solid solution (Sect. 3.2.3).

As indicated in Section 2.3, an isothermal heat treatment is carried out at 170 °C (up to 18 h) on the as-built temper in order to determine the optimum artificial aging. The hardness and the electrical conductivity are measured and plotted in Figure 6. Between the as-built temper and artificial aging for 5 h (18,000 s) at 170 °C, the hardness increases then stabilises (around 108 HBW), irrespective of the duration of the artificial aging treatment. This seems to indicate that the maximum mechanical properties are reached after 5 h of artificial

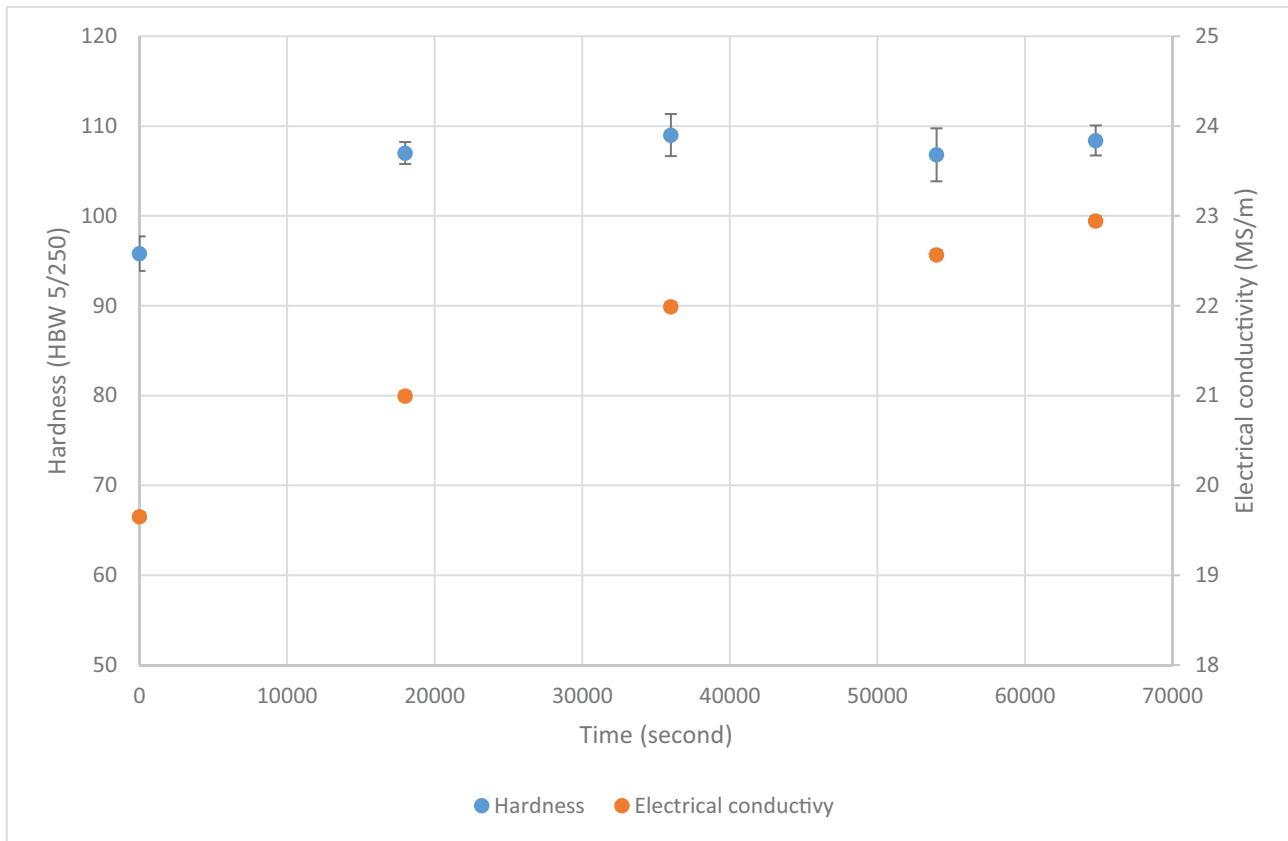


Fig. 6. AlSi5Cu3Mg alloy – Change in hardness and electrical conductivity as a function of the isothermal artificial aging at 170 °C.

aging at 170 °C. However, even if the hardness has significantly increased with respect to the as-built temper, it remains far from the 128.4 HBW measured in the case of the T6 temper. Nevertheless, the mechanical properties in the as-built temper and after direct aging (10 h at 170 °C) will also be studied along with those of the T6 temper.

Figure 6 shows a change in the state of precipitation as a result of the increased electrical conductivity. However, as previously indicated, this change only results in a moderate increase in hardness in a first step and eventually results in a stagnation of the hardness in a second step. We can note that the hardness of the T4 temper (natural aging) is equivalent to that obtained after direct aging (after 5 h of holding time), which implies that the precipitation in the direct aging treatment does not provide a high level of hardening. This is explained by the fact that a significant quantity of Cu remains trapped in the eutectic (which is not destabilized by the direct aging) and therefore cannot participate in the precipitation (Fig. 5b).

3.4 Static mechanical properties

Figure 7 presents the static mechanical properties obtained for the following tempers: as built, direct aging (10 h at 170 °C) and T6.

The as-built temper features interesting mechanical properties: the R_m value is close to 400 MPa. However, compared to the AlSi7Mg0.6 alloy and the AlSi10Mg alloy, the mechanical properties are lower [3,36]. These lower

properties are probably related to a lower Si content (only 5.2% by weight) and thus to a lower solid solution hardening effect. As a matter of fact, as demonstrated by Rao et al. [48], the solid solution aluminium is supersaturated with silicon in the as-built temper, due to the very high cooling rates of approximately 10^5 to 10^6 °C/s. As a result, the solubility of silicon in the solid solution aluminium exceeds 1.6% (by weight), although this value is the maximum limit of solubility in aluminium at the temperature of the eutectic.

Direct aging (10 h at 170 °C) increases the $R_{p0.2}$ value and also very lightly increases the R_m value. The elongation (A%) decreases. The improved $R_{p0.2}$ can be explained by the fine precipitation of the θ'' phase in low density, as shown in Figure 5b. But, in comparison with the AlSi7Mg0.6 alloy in an equivalent temper [36], the mechanical properties obtained are lower. This difference can be explained by a lower Mg content in the AlSi5Cu3Mg alloy than in the AlSi7Mg0.6 alloy. However, the addition of Cu could have been expected to allow a greater increase in the mechanical properties. But as we indicated in Section 3.2.5, the Cu remains trapped in the eutectic.

On the other hand, the T6 temper is completely relevant with the AlSi5Cu3Mg alloy, unlike the AlSi7Mg0.6 alloy. As a matter of fact, high and very interesting mechanical properties are obtained: the R_m value reaches 430 to 440 MPa, the $R_{p0.2}$ value reaches 360 MPa and the A% value remains at an interesting level, at approximately 12%. Moreover, the anisotropy observed

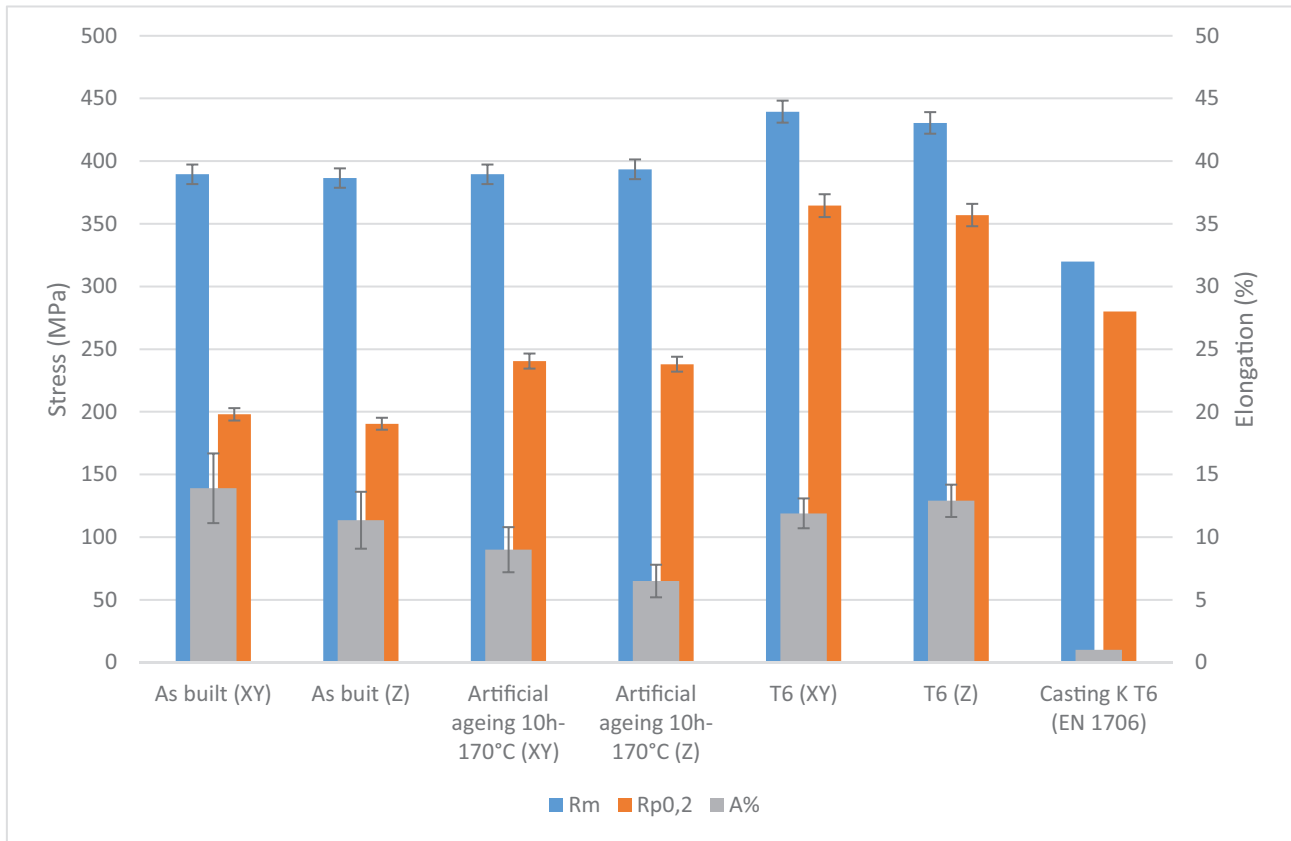


Fig. 7. Static mechanical properties of the AlSi5Cu3Mg alloy for different tempers.

in the as-built temper and in the direct aging temper does not seem to exist anymore; this can be explained by the “disappearance” of the build tracks as observed in Figure 4. As a matter of fact, due to the layer-by-layer construction, the stacked build tracks create some anisotropy along the Z direction. The mechanical properties found are also markedly higher than the minimum values of standard NF EN 1706 (casting K T6), and they are equivalent to the values reached with a 2024 T6 wrought alloy. These good mechanical properties are the result of a fine and dense precipitation of the θ'' phase, as observed on Figure 5c. As a matter of fact, it is an admitted fact that the yield strength of an alloy stems from the ability of obstacles, such as precipitates, to hinder the motion of moving dislocations [46,49]. The yield strength increases simultaneously with the size of the precipitates, as long as the precipitates are sheared by the dislocations.

Figure 8 highlights a significant anisotropy on the elongation in the as-built temper and after direct aging (10 h at 170°C) (same for the AlSi7Mg0.6 alloy [36]). The T6 temper can be considered as isotropic. So as to provide an explanation on this anisotropy, we examined the fracture surfaces of the tensile test pieces built along the Z direction and in the XY plane, for the three studied tempers: as built, direct aging and T6.

Figure 9 is specific to the as-built temper and gives rise to the following remarks:

- On the fracture surface of the tensile test pieces built along the Z direction, the build tracks are visible; the tracks cross each other at 67°, thereby revealing the building strategy (Fig. 9a). Moreover, the direction of the fracture of the tensile test pieces built along the Z direction shows a tendency to break perpendicularly to the axis of the tensile stress. At the microscopic scale, the presence of dimples is observed over large areas but also smoother areas corresponding to dendrites of solid solution (Fig. 9a – red dotted line).
- Conversely, on the fracture surface of the tensile test specimens built in the XY plane, the build tracks are not visible and the direction of fracture is oriented at 45° with respect to the stress (Fig. 9b).

The same remarks can be expressed in the case of the test pieces after direct aging (10 h at 170°C).

Figure 10 presents the T6 temper and gives rise to somewhat different remarks:

- It is more difficult to see build tracks on the fracture surfaces of the tensile test pieces built along the Z direction. The fracture appears in the form of two bevels at 45° (Fig. 10a). At the microscopic scale, the entire surface of the fracture facies is made up of dimples which are deep and wide compared to the as built temper where the dimples are numerous, shallow and not very wide (Fig. 10a).

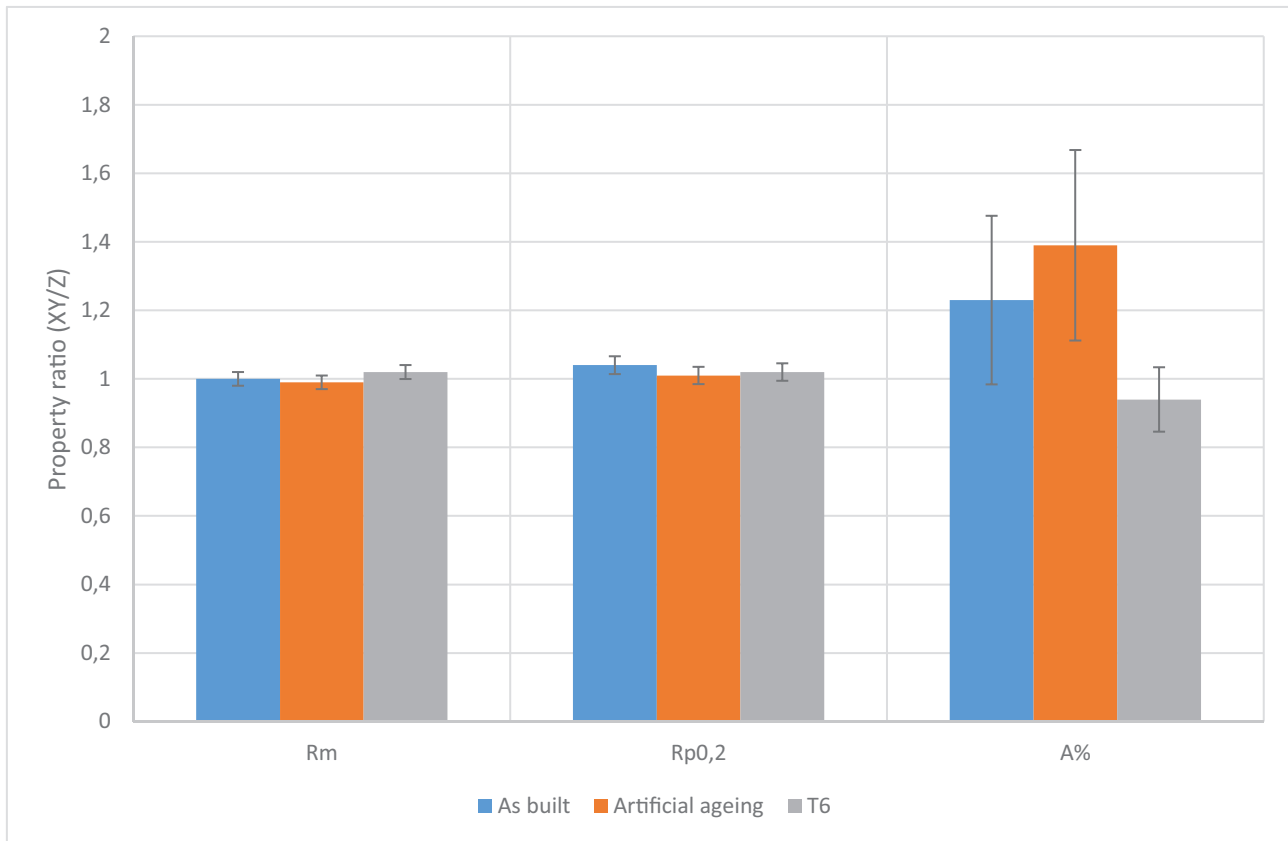


Fig. 8. Anisotropy of the static mechanical properties.

- For the test pieces built in the XY plane, there are no signs of the build tracks, and the direction of fracture is oriented at 45° with respect to the stress (Fig. 10b).

The fractures oriented at 45° with respect to the tensile testing direction are indicative of a ductile material which usually features good elongation properties. However, a fracture perpendicular to the tensile testing direction indicates a brittle material with rather low elongation properties [50]. In the case of the as built temper, this illustrates the fact that the Z direction has lower elongation properties compared to the XY plane. In addition, the visibility of the build tracks on the as built fracture facies in the Z direction can be explained by the fact that the edge of the build tracks is composed of coarser dendritic cells created by another thermal alteration of the neighbouring build tracks [3,12]. This coarser structure is more brittle as shown in Figure 9a where the area circled in red suggests some brittleness; as indeed the numerous, small and shallow dimples that are normally bonded to less ductile material (lower elongation) [50]. Moreover, the Al-Si-Al₂Cu eutectic is a brittle structure [51].

When a tensile test is carried out on a test piece built in the XY plane, the tensile forces can be schematically represented as illustrated in Figure 11a (a possible example is shown with blue arrows). In this configuration, one can see that a fracture can follow the edge of a

build track (blue dotted line) in a build plane; however this fracture will have to run in a completely different direction in the next or previous build plane in order to follow again the edge of a build track (red dotted line) [52]. Therefore, the fracture may sometimes follow the edge of the tracks but it may also run in the middle of a track.

When a tensile test is carried out on a test piece built along the Z direction, the tensile forces can be schematically represented as illustrated in Figure 11b (an example is shown by blue arrows). In this configuration, a fracture can initiate in any build plane and follow, in all cases, the edges of the build tracks without any difficulty (red dotted line) [52].

These comments help to explain the findings made on the fracture surfaces but also the very significant anisotropy of the elongation along the Z direction for the as-built and direct aging tempers. The T6 temper is more isotropic, due to the disappearance of the dendritic structure and the build tracks.

Figure 12 shows examples of engineering stress-strain curves for the three studied tempers. Note that the curves in the T6 temper display the usual morphology of that temper, that is to say with a perfectly elastic-plastic trend. Moreover, the curves after direct aging (10 h at 170 °C) have the same shape as the curves of the as-built temper. The only difference is a shift in the yield strength; this shift being present on the entire curve.

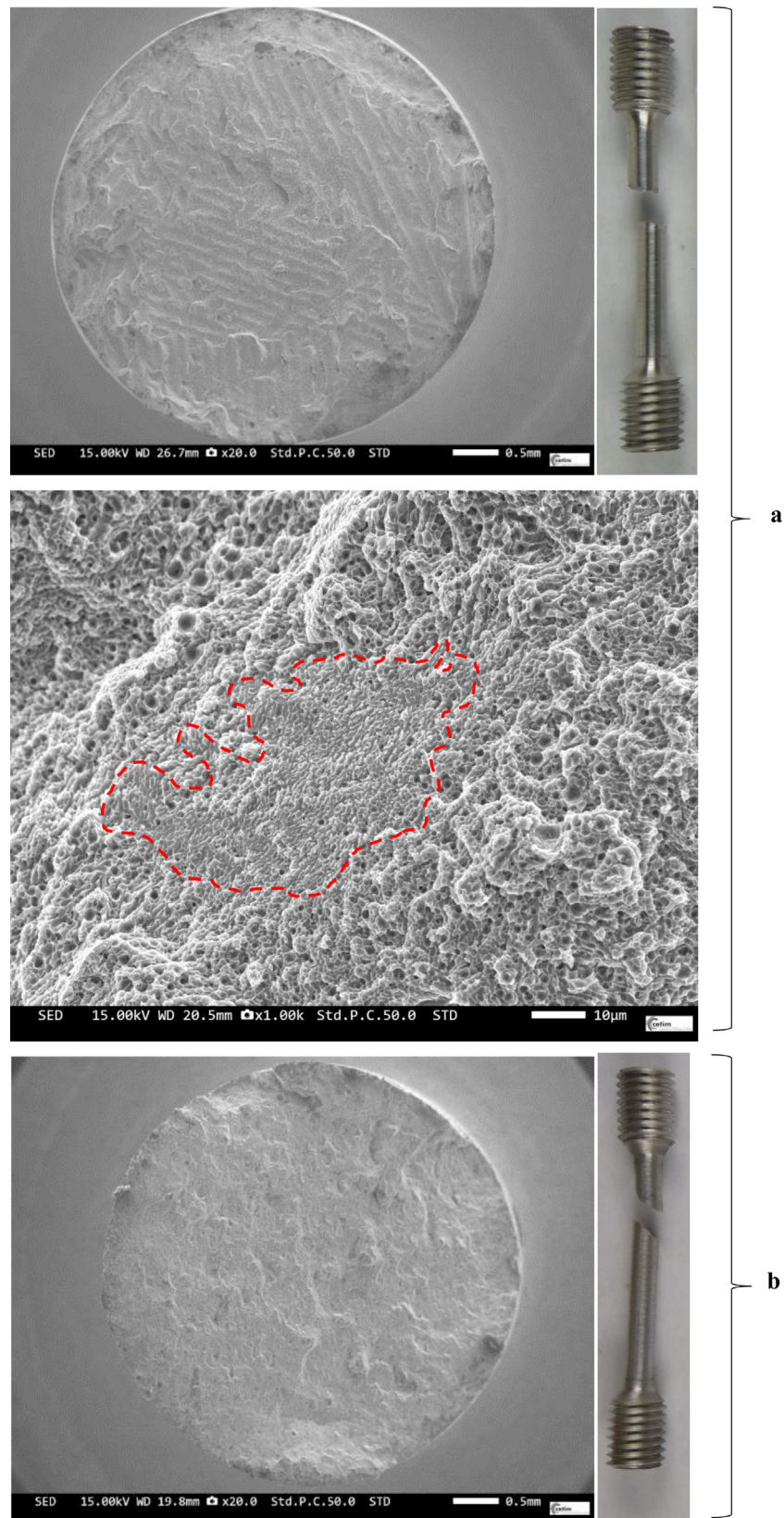


Fig. 9. Fractography – Tensile test pieces, as built. (a) Z direction – (b) XY direction.

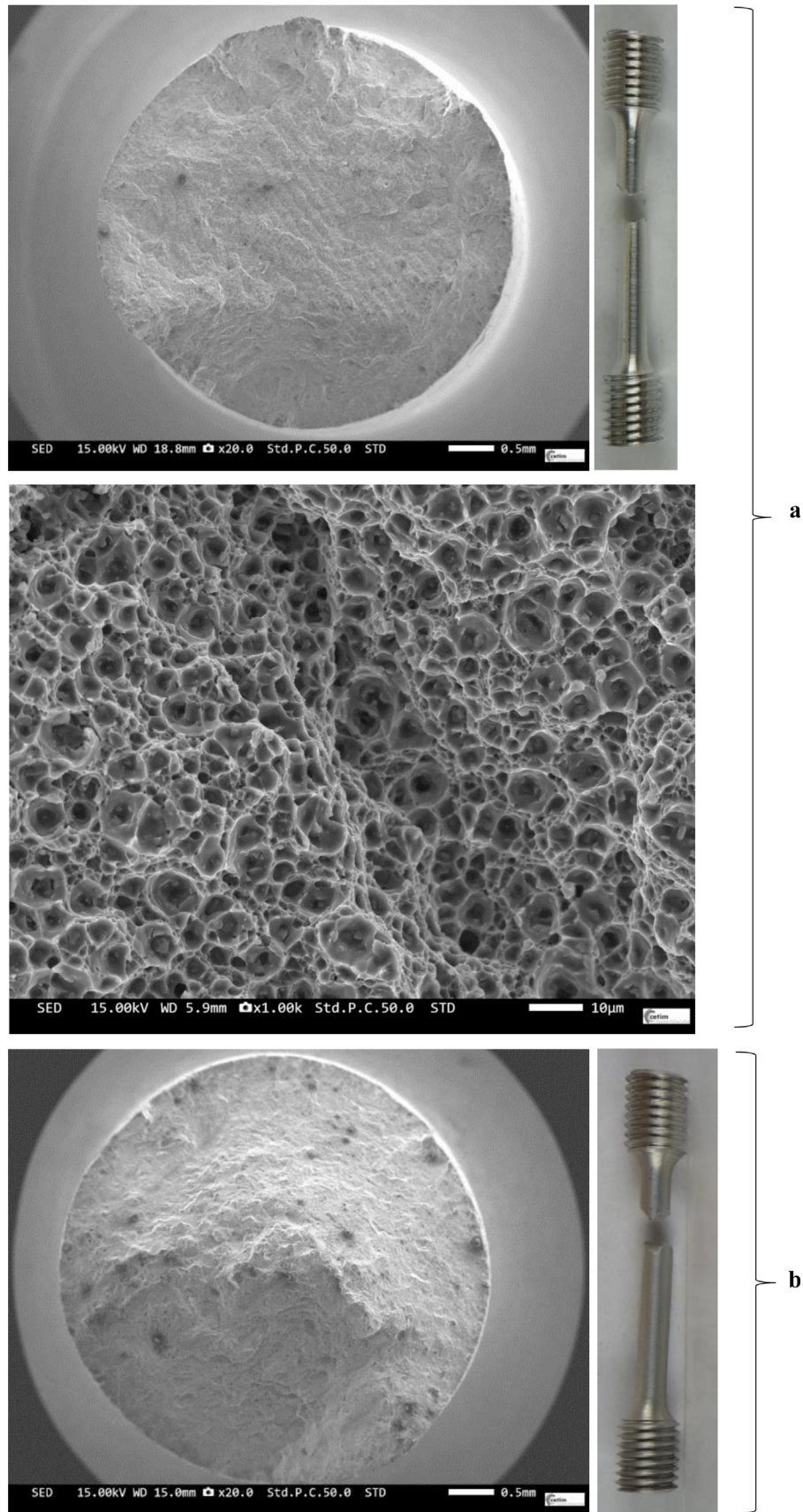


Fig. 10. Fractography – Tensile test pieces, T6 temper. (a) Z direction – (b) XY direction.

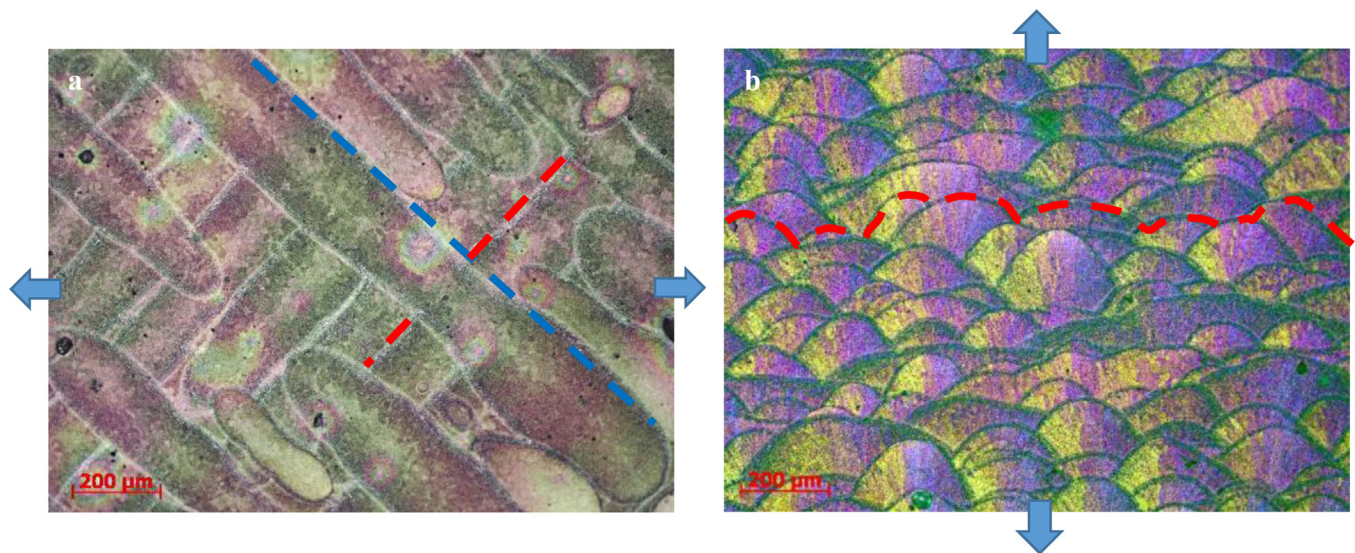


Fig. 11. Tensile stress on a tensile test piece. (a) XY plane – (b) Z direction.

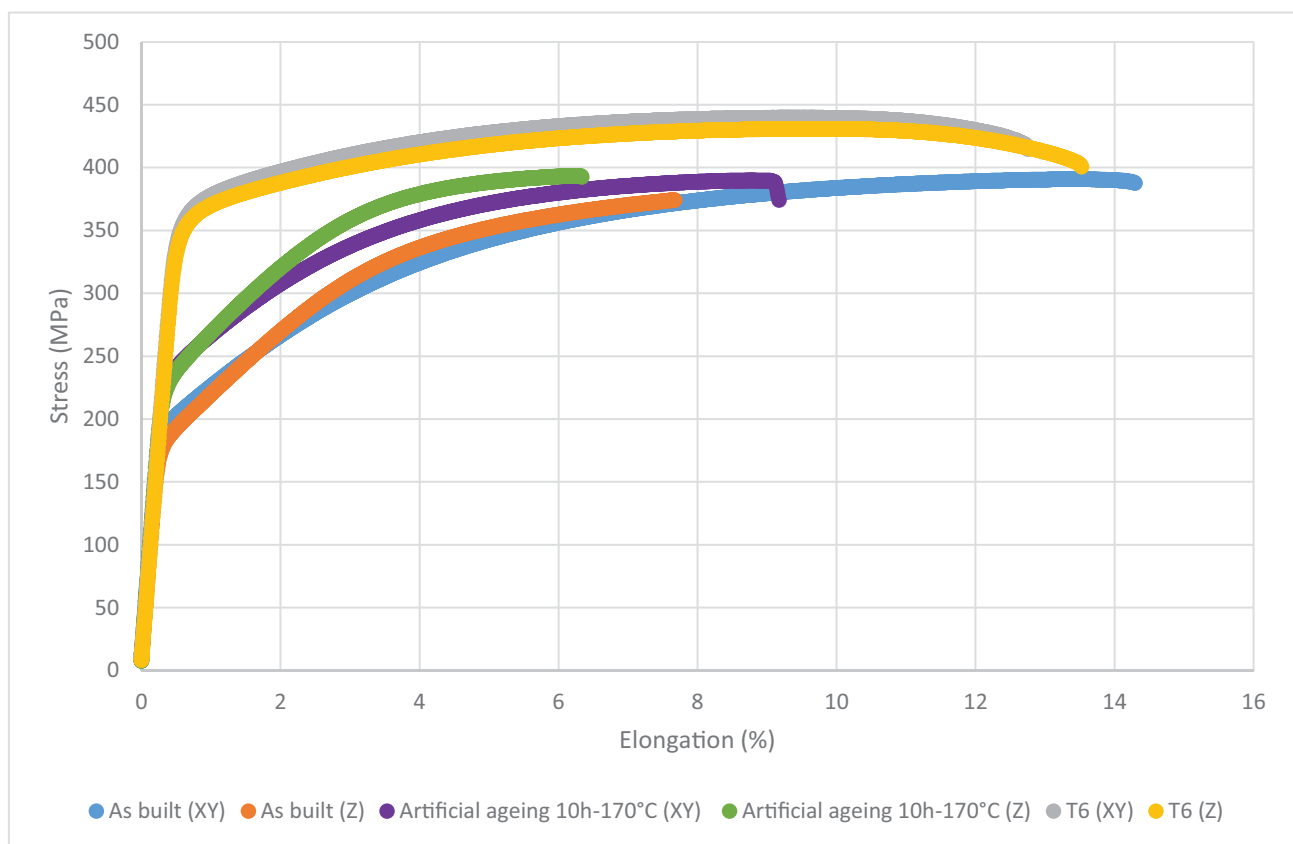


Fig. 12. Engineering stress-strain curves – AlSi5Cu3Mg alloy.

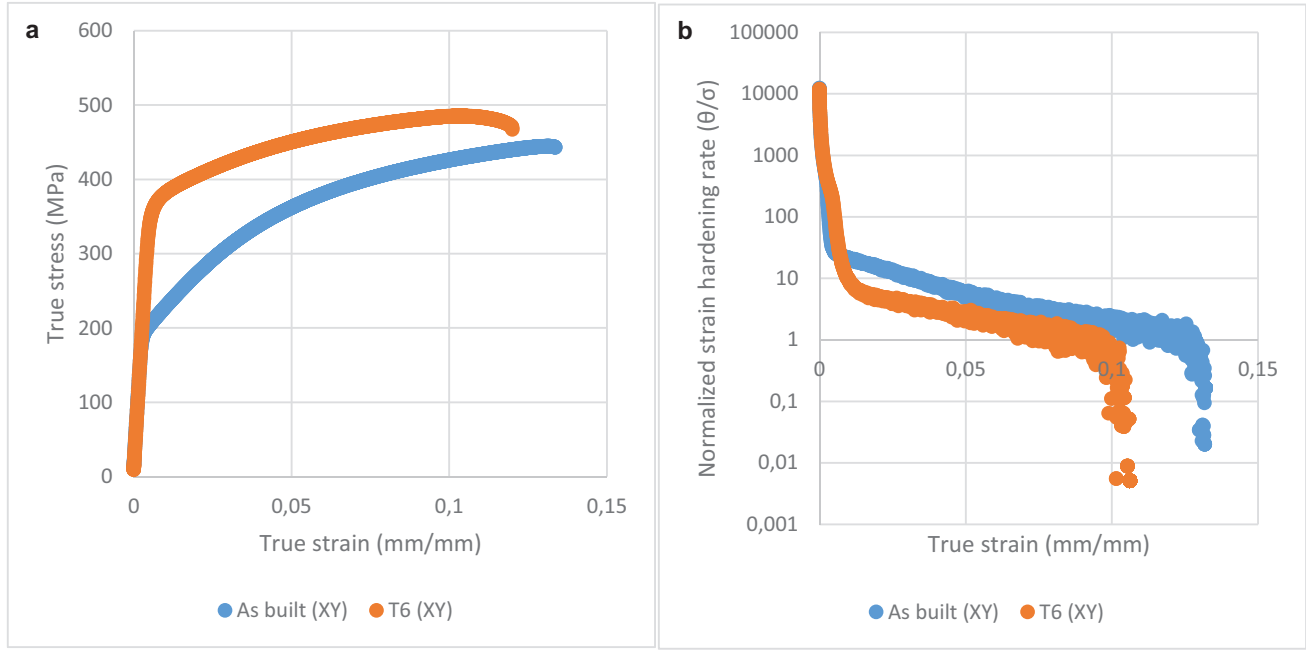


Fig. 13. (a) True stress-strain curves, as-built and T6 tempers – (b) Normalised strain hardening rate for the true stress-strain curves of (a).

Table 5. True uniform plastic strains – as-built and T6 tempers.

	As built (XY)	T6 (XY)
$\epsilon_{\text{uniform Considère}}$	0.125 (Fig. 17b)	0.09 (Fig. 17b)
$\epsilon_{\text{uniform Hollomon}} = n$	0.29 (in accordance with standard NF EN ISO 10275)	0.12 (in accordance with standard NF EN ISO 10275)
$\epsilon_{\text{uniform Voce}}$	0.114	0.08

The strain hardening rate θ is defined as the derivative of σ with respect to ϵ , as in the following equation:

$$\theta = \frac{d\sigma}{d\epsilon}, \quad (1)$$

where σ : true stress, ϵ : true strain.

The strain hardening rate is related to Considère's criterion, also called the maximum force criterion, which predicts the diffuse tensile necking strain, that is to say the limit of uniform elongation [53]. According to this criterion, necking appears when the tensile force F reaches a maximum value, which is expressed as follows:

$$\theta = \frac{d\sigma}{d\epsilon} = \sigma \quad \text{or} \quad \frac{\theta}{\sigma} = 1. \quad (2)$$

Figure 13 presents, for the as-built temper (XY) and the T6 temper (XY), the true stress-strain curves as well as the strain hardening rate normalised by the true stress ($\frac{\theta}{\sigma}$). Graphically, it is therefore possible to apply the

Considère's criterion to determine the beginning of the necking phenomenon and obtain the uniform elongation, in other words the true uniform plastic strain which can be named $\epsilon_{\text{uniform Considère}}$.

If the aluminium alloy obeys a Hollomon power law [54], then the true uniform plastic strain is equal to n strain hardening exponent, that is to say $\epsilon_{\text{uniform Hollomon}} = n$. However, many aluminium alloys rather follow a law initially introduced by Voce [55] that proposes a saturation of the true stress σ_s . In this case, the true uniform plastic strain follows the relation given below [52]:

$$\epsilon_{\text{uniform Voce}} = \frac{-1}{K} \ln \left[\frac{\sigma_s}{(K+1)(\sigma_s - \sigma_0)} \right], \quad (3)$$

where K , σ_s and σ_0 are the constants of Voce's behaviour law: $\sigma = \sigma_s - (\sigma_s - \sigma_0) \exp(-K\epsilon)$.

Table 5 compares the true uniform plastic strains obtained with the three methods: Considère, Hollomon (strain hardening exponent n), and Voce.

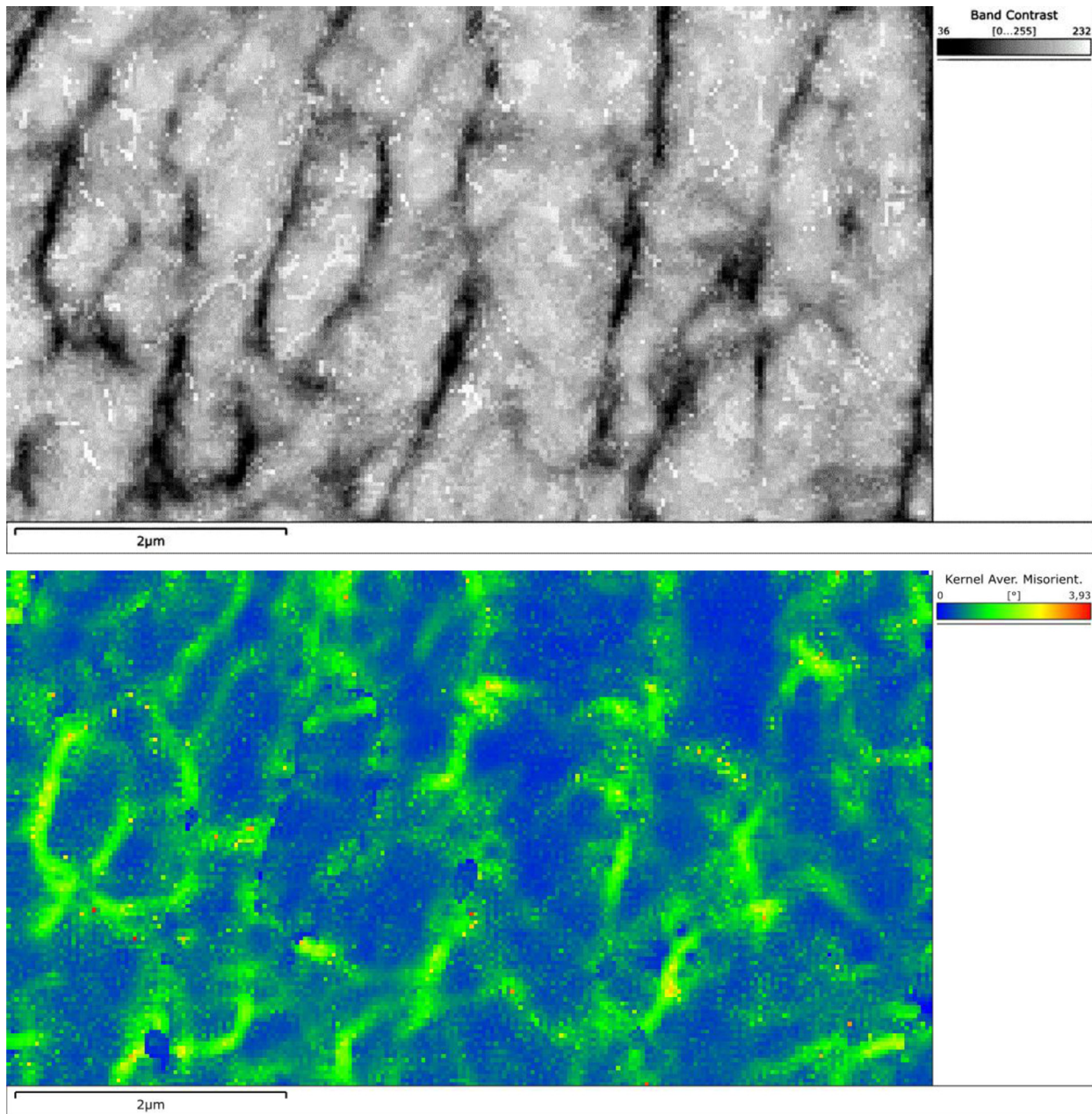


Fig. 14. HR EBSD map for the as-built temper. (a) Band contrast – (b) KAM.

We can see that, in both cases (As built and T6), the Hollomon method (strain hardening exponent n) gives a true uniform plastic strain that is rather different from the values given by the other two methods which, on the other hand, are rather similar. The method of Considère's criterion is the reference, because built directly by derivation of the true stress-strain curves (Fig. 13). Therefore, it appears that the LPBF-processed AlSi5-Cu3Mg alloy in the as-built temper and in the T6 temper follows a Voce's behaviour law. For this alloy, the strain hardening exponent n calculated based on standard NF EN ISO 10275 and which corresponds to the Hollomon method,

does not make it possible to determine the true uniform plastic strain, that is to say the beginning of the tensile necking.

On Figure 13b, we can see that the normalised strain hardening rate obtained in the case of the as-built temper (XY) is, over almost the entire true strain, higher than that obtained for the T6 temper, except at the beginning of the curve. As a matter of fact, if the yield strength is maximum for a fine dispersion of precipitates being 10 to 50 nm in size equivalent to a T6 temper (Figs. 5c, 12 and 13a), the strain hardening rate (as well as the normalised strain hardening rate) in this temper is rather low [49] (Fig. 13b). The

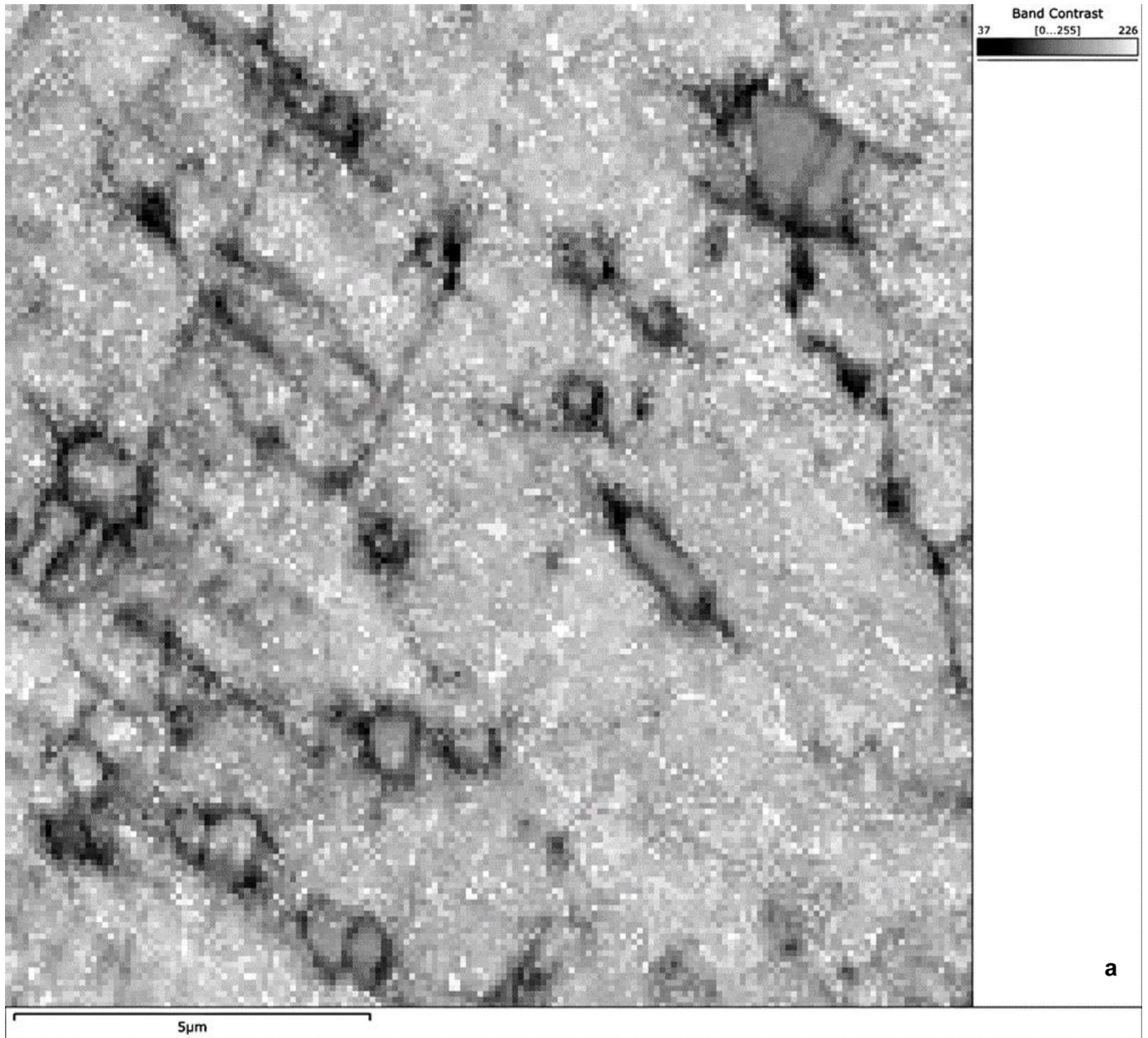


Fig. 15. HR EBSD map for the T6 temper. (a) Band contrast – (b) KAM.

maximum strain hardening rate is reached for particle sizes close to the sizes of the dislocation cells, i.e. approximately $1\ \mu\text{m}$ [49]. Thus, the explanation for the higher normalised strain hardening rate in the as-built temper than in the T6 temper lies in the very structure of this temper, that is to say the fine dendrites (Figs. 2c, 3a and 5a) which, precisely, have a size of approximately $1\ \mu\text{m}$ (between 0.8 and $1\ \mu\text{m}$).

An HR (high resolution) EBSD analysis was carried out on the tensile test pieces (XY) in as-built and T6 tempers. During this analysis, we particularly focused on the band contrast and the Kernel Average Misorientation (KAM) (Figs. 14 and 15). The band contrast highlights the grains of the analysed material but also some microstructures; the

KAM, which expresses the local misorientation, is related to the density of geometrically necessary dislocations ρ_{GND} by the following equation:

$$\rho_{GND} = \frac{\alpha \Delta\theta}{b \Delta x}. \quad (4)$$

This is Ashby's equation [56] which relates the curvature of the crystal lattice $\frac{\Delta\theta}{\Delta x}$ to the geometrically necessary dislocation density.

Where α : a constant between 1 and 5 (depending on the configuration); $\Delta\theta$: the misorientation of the lattice, that is to say the KAM; Δx : the size of the kernel; b : the Burgers vector.

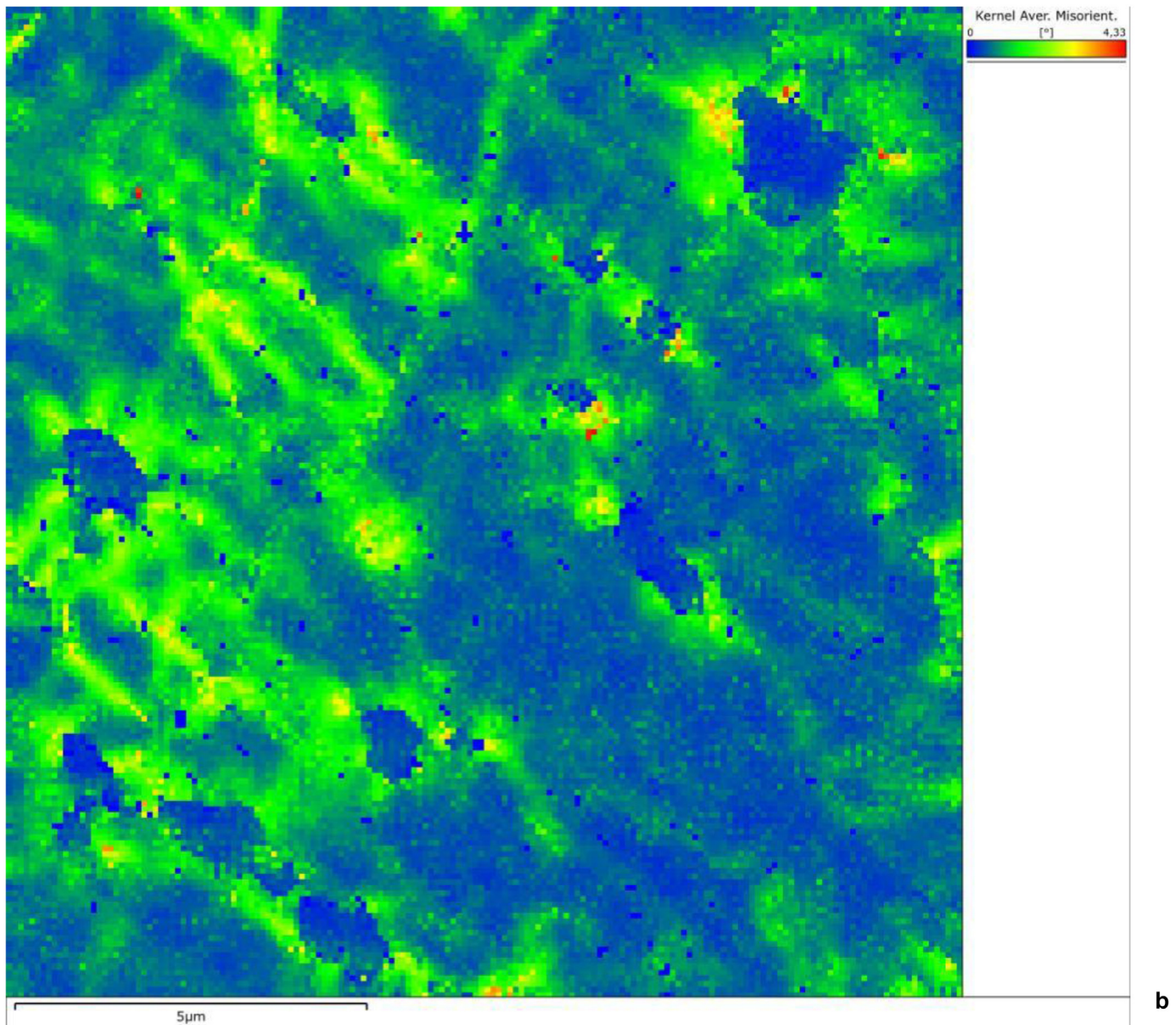


Fig. 15 (Continue).

Thus, one can understand the close relation between the KAM and the dislocations.

Both the band contrast and the KAM (Fig. 14) highlight the fact that the dislocations settle on the dendritic cells in the as-built temper, which explains a higher strain hardening rate for the as-built temper. We can also note that, in the T6 temper (Fig. 15), the dislocations are more diffuse, except at the edges of the silicon polyhedrons where significant concentration is visible. This point also makes it possible to understand that, in our case, the strain hardening rate is lower in the T6 temper than in the as-built temper, but not extremely low as stated in [49]. As a matter of fact, the particular structure of the LPBF-processed alloy in T6 temper shows the presence of silicon polyhedrons with sizes ranging between 0.5 and 4 μm (Fig. 4) which allow the dislocations to settle. In conclusion, it can be seen that the tensile behaviour of the AlSi5Cu3Mg alloy is closely linked to

its metallurgical temper. Thus, the alloy ranges from a problem of anisotropy on the elongation due to the manufacture (a certain “brittleness” at the interface between tracks), and of a hardening by work hardening due to the settling of the dislocations on fine dendrites (dendritic cells) in as built temper; to an isotropic behaviour with a perfectly elastic-plastic trend, with a hardening mainly based on the precipitation of the θ'' phase in the T6 temper.

3.5 Productivity and cost

The idea is to compare the productivity and the manufacturing cost of the AlSi7Mg0.6 alloy (a reference on the current LPBF market) and the AlSi5Cu3Mg alloy. For this purpose it is necessary to be able to estimate the manufacturing times with only the following information:

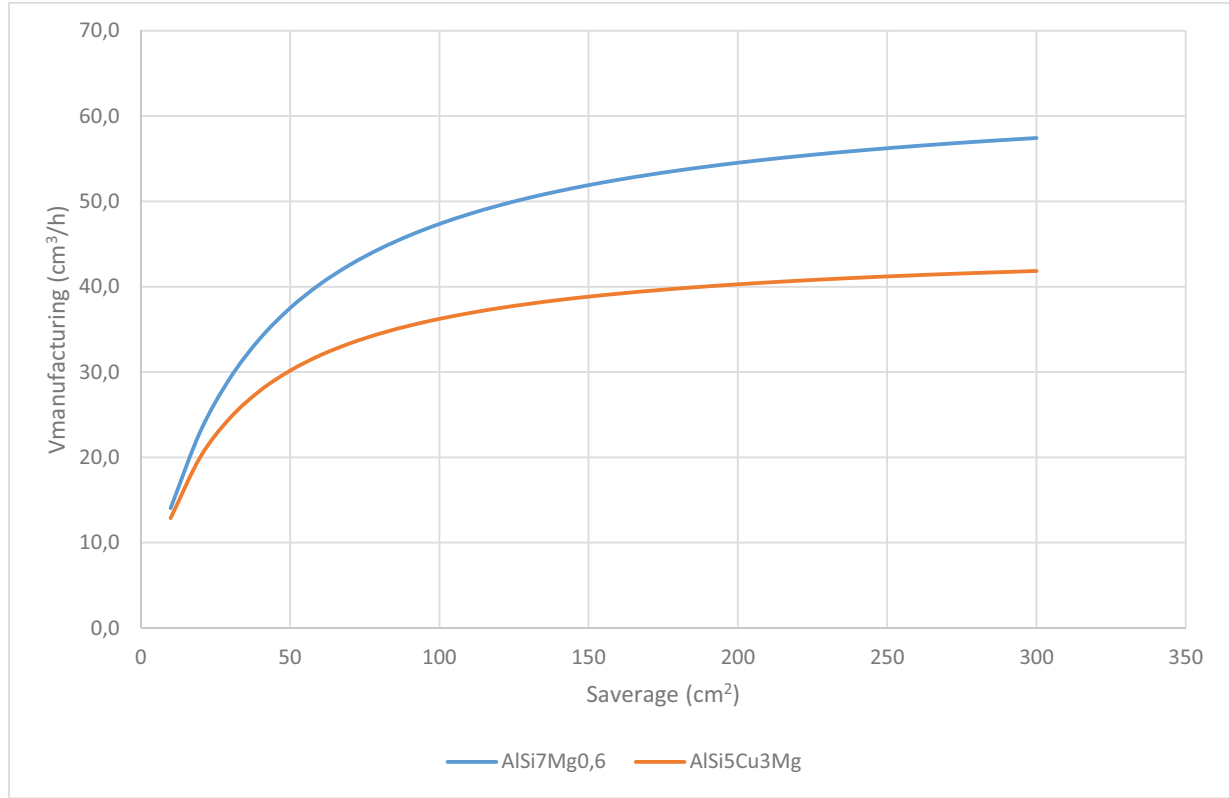


Fig. 16. Comparison of the manufacturing speeds of the AlSi7Mg0.6 and AlSi5Cu3Mg alloys.

the volume of parts, the build height and, of course, the parameters of the building machine. For this estimation, we use the method described by Pillot [1]. Thus, the build chamber heating and cooling times, the time to make the build supports, the contouring time and the time between two successive laser beams (skywriting) are not taken into account. The manufacturing time equation (6) can be broken down into two phases: layering and laser building.

$$t_{\text{manufacturing}} = t_{\text{layering}} + t_{\text{laser building}}. \quad (5)$$

Buchbinder [57] proposes the following equation (6) for the build rate:

$$V_{\text{laser building}} = vhe. \quad (6)$$

Therefore, we can estimate the laser building time and the layering time as follows:

$$t_{\text{laser building}} = \frac{\text{volume of the part}}{V_{\text{laser building}}}, \quad (7)$$

$$t_{\text{layering}} = \frac{\text{build height}}{e} t_{\text{for one layer}}. \quad (8)$$

For our study, we set the time for one layer to approximately 10 seconds (SLM280 machine).

Moreover, we set the average laser scanned surface area S_{average} per build layer to:

$$S_{\text{average}} = \frac{\text{volume of the part}}{\text{build height}}. \quad (9)$$

And therefore:

$$V_{\text{manufacturing}} = \frac{S_{\text{average}}e}{t_{\text{manufacturing}}} = \frac{1}{\frac{t_{\text{for one layer}}}{S_{\text{average}}e} + \frac{1}{V_{\text{laser building}}}}. \quad (10)$$

The manufacturing parameters for the AlSi5Cu3Mg alloy are indicated in Section 2.1. The manufacturing parameters for the AlSi7Mg0.6 alloy used on the same machine are as follows: $P = 650\text{W}$, $v = 2100\text{ mm/s}$, $h = 170\text{ }\mu\text{m}$ and $e = 50\text{ }\mu\text{m}$. For both the AlSi7Mg0.6 and AlSi5Cu3Mg alloys, the layering time is the same, as the time for one layer is identical: same machine (SLM280) and identical thickness layer e ($50\text{ }\mu\text{m}$).

Figure 16 presents the manufacturing speeds for both alloys, AlSi7Mg0.6 and AlSi5Cu3Mg. We can see that when the average surface area to be laser-scanned increases, the difference between the two alloys increases. Therefore, the AlSi7Mg0.6 alloy allows higher production rates than the AlSi5Cu3Mg alloy. This is mainly due to the laser building speed equation (6) which amounts to $45.4\text{ cm}^3/\text{h}$ for the AlSi5Cu3Mg alloy when it amounts to $64.3\text{ cm}^3/\text{h}$ for the AlSi7Mg0.6 alloy, which means a 40% faster speed.

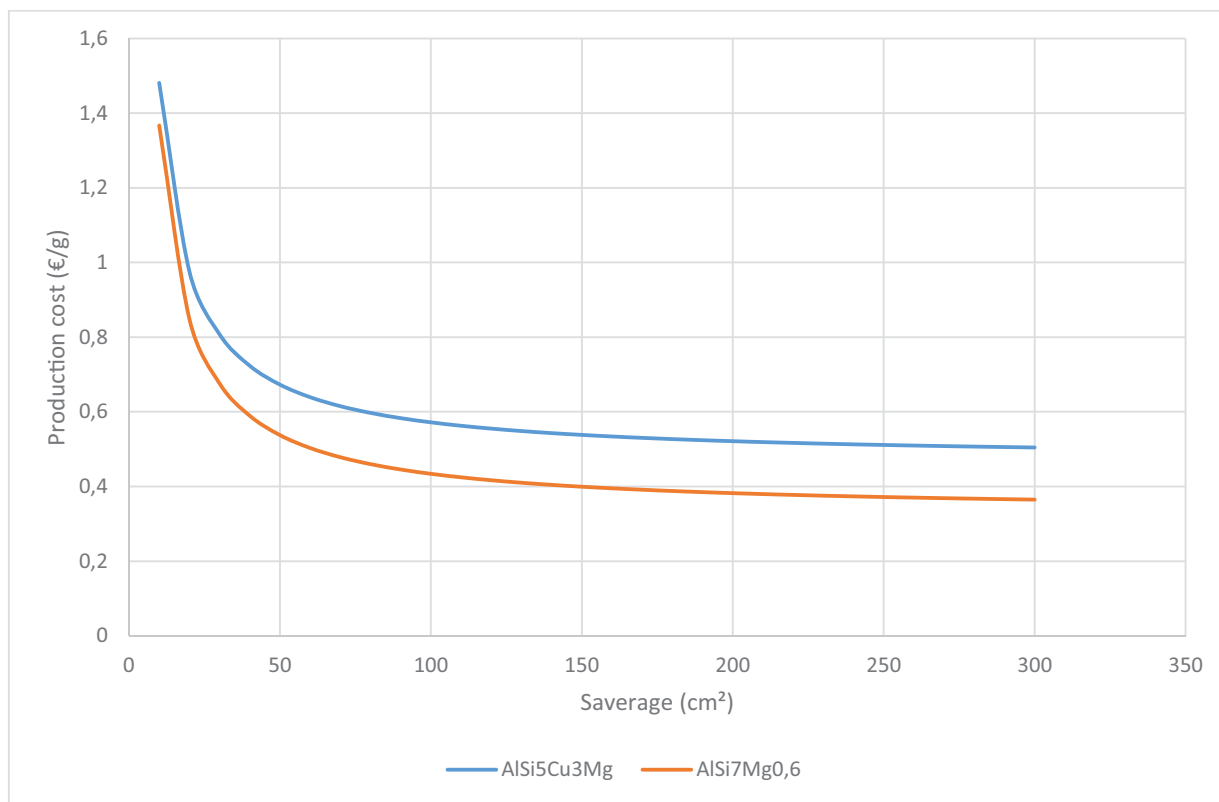


Fig. 17. Comparison of the production costs of the AlSi7Mg0.6 and AlSi5Cu3Mg alloys.

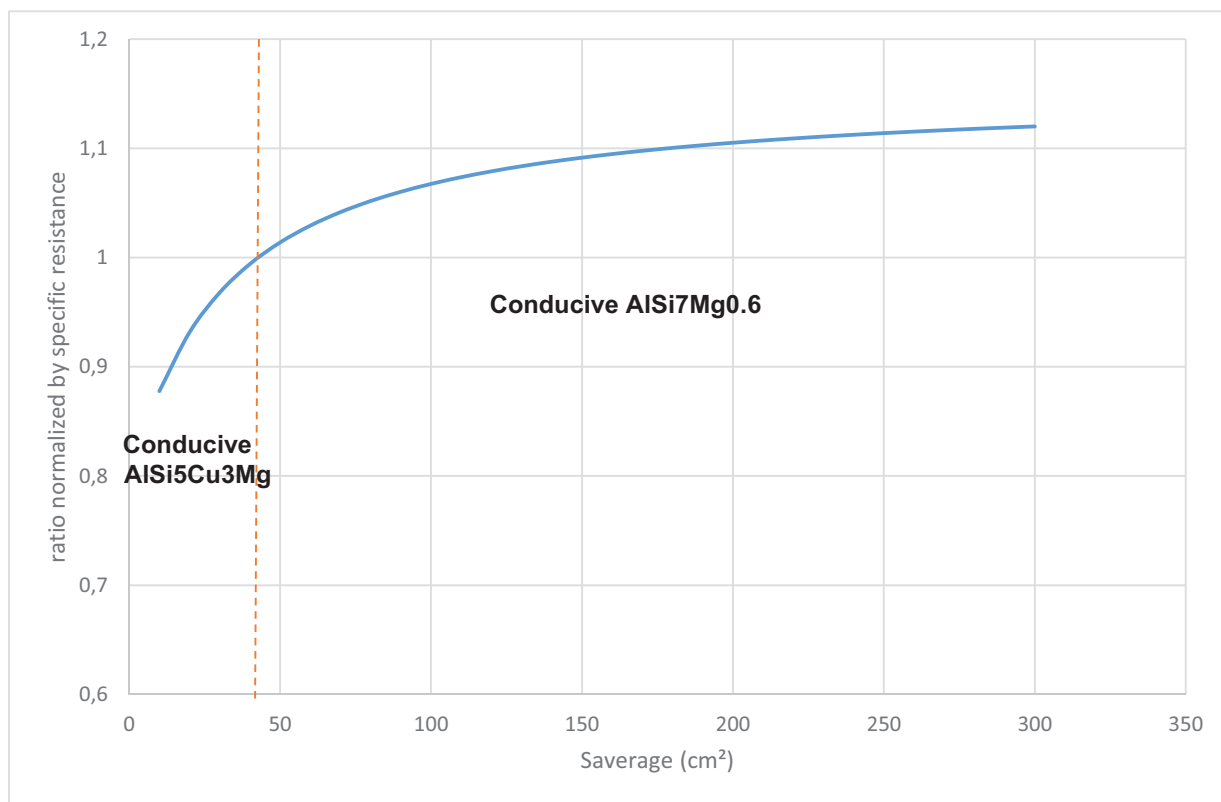


Fig. 18. "Profitability" region of the AlSi7Mg0.6 and AlSi5Cu3Mg alloys.

In order to estimate the manufacturing cost of these two alloys, we use a machine hourly rate of € 50 per hour [58] and a powder cost of € 70 per kg for the AlSi5Cu3Mg alloy (see Sect. 2.2) and € 40 per kg for the AlSi7Mg0.6 alloy.

With these elements, Figure 17 is obtained, which shows a very small cost difference between the two alloys for small average laser-scanned sections, that is to say below 20 cm². Moreover, for the two alloys, the production cost drops very quickly depending on the average laser-scanned surface area and eventually stabilises (asymptote) around € 0.5 per gram for the AlSi5Cu3Mg alloy and € 0.36 per gram for the AlSi7Mg0.6 alloy. This gives a difference in manufacturing cost of approximately € 0.14 per gram, i.e. a value which can become very substantial depending on the weight of the parts to be produced.

To go further, it is possible to define a “profitability” region for each alloy (AlSi7Mg0.6 and AlSi5Cu3Mg) depending on the average laser-scanned surface area, by qualifying Figure 17 with the specific resistance ($R_{p0.2}$ /density). Figure 18 presents the two “profitability” regions of the AlSi7Mg0.6 and AlSi5Cu3Mg alloys. It can be seen that the limit appears for an average laser-scanned surface area of approximately 42 cm². Below 42 cm², the AlSi5Cu3Mg alloy is more cost effective than the AlSi7Mg0.6 alloy.

4 Conclusion

After a description and a study of the microstructures and mechanical properties of the three tempers (as built, direct aging (10 hours at 170 °C) and T6), we can demonstrate the interest of the T6 temper for the LPBF-processed AlSi5Cu3Mg alloy. As a matter of fact, as we showed and unlike the AlSi7Mg0.6 and AlSi10Mg alloys, there is no change in porosity in the AlSi5Cu3Mg alloy after application of the T6 temper, due to a lower solution heat treatment temperature (510 °C). Therefore, the T6 temper has the following advantages:

- Good material health.
- Very good mechanical properties: high $R_{p0.2}$ and good elongation.
- Homogenous structure.
- No (or very little) anisotropy.
- No influence of the manufacturing parameters on the final mechanical properties: the solution heat treatment “cancels” the impacts, if any.
- Probability of low residual stresses: to be confirmed.

Nevertheless, it should be kept in mind that the T6 temper is more restrictive in terms of heat treatment (longer and more expensive) and that there is a risk of deformation of the parts (in particular thin and slender parts) during the water quenching operation.

In addition, this study has made it possible to explain the characteristics of the stress-strain curves of the three tempers through:

- The analysis of the fracture surfaces, which explains the anisotropy of the elongation in the Z direction for the as-built and direct aging tempers.
- The study of the strain hardening rate, by establishing a relationship with the microstructure and, in particular, the densities of type θ'' and/or θ' precipitates: demonstration of a high $R_{p0.2}$ value with a lower strain hardening rate for the T6 temper, unlike the as-built temper which features a high strain hardening rate, due to its dendritic structure, but a lower $R_{p0.2}$ value.
- The study (by EBSD) of the dislocation densities to explain the relationship between the fine dendritic structure and the high strain hardening rate for the as-built and direct aging tempers.

Finally, a last technical and economic point shows the “profitability” zone of the AlSi5Cu3Mg alloy in comparison to the AlSi7Mg0.6 alloy.

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