

Understanding the effect of tailored heat treatment on Zr-modified Al7075 fabricated by laser powder bed fusion

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ABSTRACT

Despite high-strength aluminium alloys now being widely processed by laser powder bed fusion (L-PBF), a specifically designed heat treatment (HT) for L-PBF Al7075 has yet to be identified and the microstructural features responsible for strengthening remain to be fully understood. In this study, fully dense Al7075+1.8%Zr was processed without hot cracking. High temperature heat treatments between 350°C and 525°C were adopted, attaining the highest hardness (162HV) at 425°C followed by an artificial ageing (AA). This heat treatment produced a yield strength of 515 MPa and an elongation at fracture of 12%. To understand the effect of this tailored low-temperature HT, comparisons were made with a classical T6 HT (1 h at 470°C followed by AA). Transmission electron microscopy analyses revealed a higher volume fraction of Al_3Zr $L1_2$ secondary precipitates with the 425°C HT than with the 470°C HT. These hardening precipitates were quantified as the main contributor to strengthening, accounting for 207 MPa. The 425°C HT prevented the over-ageing Zr precipitates generated by the 470°C HT, while inhibiting recrystallization. η' precipitates played a minor role in the tensile properties as they were nearly absent from the microstructure due to the excessive evaporation of Mg and Zn during L-PBF.

1. Introduction

Laser powder bed fusion (L-PBF) is the leading additive manufacturing technology for Al alloys. This manufacturing process is capable of producing complex, detailed parts that offer considerable potential, particularly for the aerospace industry [1–4]. The high cooling rates generated by L-PBF (10^6 – 10^7 K/s for Al alloys [5]) and the complexity of the process, arising from the numerous processing parameters, limit the currently available commercial material palette [1,3].

High-strength Al alloys suffer from two inherent difficulties that hinder their processability by L-PBF. First, the preferential evaporation of alloying elements such as Zn and Mg occurs due to the high temperatures attained during the process (up to 1750°C for Al alloys [5]). These two elements possess high vapour pressures, causing their preferential evaporation [6]. Second, high-strength Al alloys experience hot cracking during L-PBF due to the extreme cooling rates involved in the process,

combined with the wide solidification interval (60 °C for Al7075) characteristic of these alloys [2,3].

Solutions to hot cracking are widely discussed in literature. The seminal work of Martin et al. [7] first demonstrated the suppression of hot cracking through Zr addition in L-PBF Al7075. The underlying mechanism involves grain refinement. Al_3Zr sub-micron precipitates form early during cooling, enabling Al grains nucleate preferentially on these heterogeneous sites [7]. The newly formed fine equiaxed grain structure is free of hot cracking. The fine grains improve liquid feeding at the end of solidification and promote a uniform distribution of thermal strains [8]. High-strength Al alloys are frequently alloyed with a combination with Zr to enable successful processing by L-PBF [9–24]. Additionally to grain refining primary Al_3Zr precipitates, Zr forms secondary Al_3Zr nanometric hardening precipitates. Due to the high L-PBF cooling rates, part of the Zr atoms are trapped in solid solution during the process. Thus, when a HT is applied as a post-treatment, Al_3Zr hardening pre-

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precipitates are formed, enhancing the yield strength of the material. These precipitates are observed in the literature for various L-PBF Al alloys using TEM [15,17,18,20,24] or using atom probe tomography [10,24]. Michi et al. [24] have analysed the secondary Al_3Zr precipitates for an Al-Cu-Mn-Zr alloy processed by L-PBF. After 6 h at 400°C, these precipitates, exhibiting a mean radius of 1.84 nm, have generated a precipitate strengthening of approximately 190 MPa.

$Ti/Ti(B_2)$ is another grain refiner that is widely reported in the literature and exhibits the same grain-refining characteristics as Zr, namely a low lattice mismatch with Al and the early nucleation of grain-refining precipitates. However, Ti is not reported to possess the ability to form nano-sized, efficient hardening precipitates, unlike Zr [25–36]. Ti only forms Al_3Ti precipitates, typically a few hundred nanometers in size, which act as grain refiners but are ineffective hardening precipitates.

Different grain refiners can be combined to achieve higher mechanical strength or more effective grain refinement. The most widely used combination of grain refiners is Sc with Zr, which is present in the most common high-strength L-PBF commercial alloy to date (Scalmalloy[®], developed by Airbus) [37–39]. Like Zr, Sc-Zr combination, in addition to its grain-refining effect, strengthens the alloy through controlled precipitation. This co-addition has thus been employed in various alloys in addition to Scalmalloy[®], including Al-Zn-Mg-Cu [40], Al-Zn-Mg [41], Al-Mn-Mg [42–44], Al-Mg-Mn-(Cu) [45–51] and Al-Cu-Li [52]. With the addition of Sc and Zr, these alloys have been successfully processed with high yield strengths, achieved through three distinct strengthening contributions: Sc-Zr secondary precipitation induced by heat treatment, grain boundary strengthening from the primary Sc and Zr precipitates, and the intrinsic strengthening provided by the base alloy system (for instance, $MgZn_2\eta'$ precipitates in Al7xxx alloys). This intrinsic strengthening may arise from either precipitation strengthening or solid solution strengthening. However, the main drawback of using Sc is its scarcity and significantly high cost, which presents an economic barrier [53]. The mechanical properties of these alloys will be analysed in details in subsection 4.2 when discussing the strengthening mechanisms of L-PBF AlZnMgCuScZr alloy fabricated by Zhu et al. [40]. More exotic grain-refiner co-additions, such as Er-Zr [54] and Sc-Zr-Hf [55] have been explored. Since Hf is highly expensive and both Hf and Er are listed as critical materials by the European Union [56], they are not the focus of the present study.

Another widely discussed approach in the literature to avoid hot cracking during L-PBF processing of Al alloys involves modifying the alloy composition to promote the formation of a high volume fraction of eutectic constituents. Si is the most commonly used alloying addition for such applications [19,32,57–59]. Ce and La are less frequently used but pose a similar effect [60,61]. However, eutectic alloys exhibit inferior mechanical properties compared to Al alloys strengthened through grain refiners. Therefore, this hot cracking mitigation strategy is not explored further in the study.

For this study, Zr is added to Al7075 to suppress hot cracking. Previous studies have demonstrated that Zr effectively eliminates hot cracking in Al7075 processed via L-PBF. Moreover, although achieving high densities for this alloy is typically challenging, samples with a relative density of 99.9% have been successfully fabricated using an unconventional pre-sintering scanning strategy [62]. Beyond optimising its density, post-processing heat treatment (HT) requires investigation for Al7075 processed via L-PBF. Currently, no HT protocol has been specifically tailored to the unique microstructure of high-strength Al7075 alloy modified with grain refiners to realise their full strengthening potential.

A systematic analysis of various HT temperatures, for solution HT or direct ageing, has been conducted only for Al-Mg-Mn-Sc-Zr alloys [37,47–49]. The HT temperature that maximises hardness has been reported as approximately 350°C. However, Al-Mg alloys are inherently strengthened through solid solution rather precipitation. Contrarily, most high-strength Al alloys are precipitation-hardened and therefore require HT optimised for controlled precipitation of both the grain-refining elements (such as Sc or Zr) and the inherent alloying elements

(such as Zn and Mg for Al7075). Moreover, the co-addition of Sc and Zr responds differently to heat treatment (HT) compared to Zr alone. Sc and Zr interact during the formation of grain refiner precipitates at the beginning of solidification [63–65], and Sc has a higher diffusion coefficient than Zr [66], affecting the behaviour of Sc trapped in solid solution during L-PBF and modifying the hardening precipitation kinetics. Therefore, even though systematic studies on the effect of HT on the Sc-Zr combination have been conducted, the behaviour of Zr alone under HT remains to be fully understood. Understanding effects of HT on microstructure necessitates comparative studies under different HT conditions using scanning electron microscopy (SEM) and transmission electron microscopy (TEM). For instance, Schuster et al. examined an Al-Cu-Mg-Zr alloy [17], Barode et al. investigated the A20X alloy [26] and Buttard et al. studied an Al-Mn-Ni-Cu-Zr alloy [10]. In the case of Al7075, only a few studies have used TEM to analyse the precipitation state following a conventional T6 HT, without including HT optimisation [18,22,27]. Li et al. [67] analysed the effect of direct ageing at 120°C on an Al-Zn-Mg-Cu-Ta alloy using both TEM and microhardness measurements. Similarly, Li et al. [68] studied an Al-Zn-Mg-Cu alloy with TiB_2 addition. Zhu et al. [40] compared the effect of HT at 465°C and 490°C, with and without artificial ageing (AA), on nanoscale precipitates using TEM and on the tensile properties of an Al-Zn-Mg-Cu-Sc-Zr alloy. In all these studies, either the grain refiner is expected to behave differently from Zr (being Ti-based [26,27,68] or a Sc-Zr combination [40]), or no exploration of various temperatures for the heat treatment (HT) is performed [10,17,18,22]. Thus, to the best of the authors' knowledge, no systematic investigation of various HT temperatures for L-PBF Al7075 with Zr addition has been reported to date, clearly highlighting the lack of identification in the literature of a tailored heat treatment that considers both intrinsic $MgZn_2$ precipitation and Zr precipitation.

Hence, building upon the process optimisation conducted in our previous research [62], the objective of this study is to investigate the effect of a tailored HT on the microstructure and static mechanical properties of a Zr-modified Al7075 produced via L-PBF. First, a broad screening of high temperature HT enables the identification of a tailored HT condition. Then, an extensive comparison using TEM with a conventional T6 HT is conducted to comprehensively assess the influence of the tailored HT on both the microstructure and the tensile behaviour of the Zr-modified Al7075 alloy.

2. Materials and methods

2.1. Powder feedstock

Pre alloyed Al7075 powder (Carpenter Additive, UK) and 2 wt.% of ZrH_2 (Nanografi, Turkey) were mixed in a Turbula mixer for a minimum of 6 hours. The particle size distribution (PSD) was measured using a Coulter LS 100Q laser diffraction analyser. The Al7075 and ZrH_2 powders were the same as those used in our previous study [62], with PSD values of respectively $d_{10} = 21\text{ }\mu\text{m}$, $d_{50} = 38\text{ }\mu\text{m}$ and $d_{90} = 63\text{ }\mu\text{m}$ for Al7075, and $d_{10} = 3\text{ }\mu\text{m}$, $d_{50} = 14\text{ }\mu\text{m}$ and $d_{90} = 36\text{ }\mu\text{m}$ for ZrH_2 (Fig. A.1). After mixing, the powder morphology analysed by SEM shows large particles of ZrH_2 between Al7075 particles as well as small ZrH_2 particles grafted at the surface of Al7075 powder particles (Fig. A.2).

Table 1 presents the results of inductively coupled plasma optical emission spectroscopy (ICP-OES) performed on the powder mixture using an Agilent Technologies 5100 ICP-OES instrument. Accordingly, the alloy investigated in this study is referred as Al7075+1.8%Zr.

2.2. L-PBF process

Al7075+1.8%Zr was processed using a ProX[®]DMP 200 system (3D Systems Inc., USA) with the following laser specifications: 1075 nm wavelength, 70 μm spot size, and 273 W maximum power. A non-heated 141 mm x 141 mm Al5Mg(wt% Mg) build plate was used. The O_2 content was maintained below 500 ppm by purging with Ar gas.

All characterisation described in subsection 2.3 was conducted on rectangular parallelepiped specimen of dimensions 10 mm x 10 mm x 5 mm. Cylindrical horizontally oriented tensile specimens were machined from 6 mm x 60 mm x 6 mm parallelepiped samples. All samples were processed using a laser power of 164 W, a scanning speed of 400 mm/s, a hatching space of 50 µm and a layer thickness of 30 µm. An unconventional scanning strategy (pre-sintering) was adopted. In this approach, each layer was scanned twice: the first scan was performed using half the laser power intended for the second scan of the same layer [69]. A detailed description of the scanning strategy and the optimisation of L-PBF parameters of the Al7075+1.8%Zr used in this study has been provided by Nothomb et al. [62].

2.3. Characterisation techniques

The fine microstructure of the specimens was analysed using a field emission scanning electron microscope (SEM)(ZEISS Ultra 55), primarily employing a backscattered electron detector to reveal the precipitates present in L-PBF Al7075+1.8%Zr. Energy-dispersive X-ray spectroscopy (EDX) was conducted on some selected samples using a Bruker XFlash 5010 detector, with an acceleration voltage of 15 kV applied for all EDX analyses. Electron backscattered diffraction (EBSD) analyses were performed using a Helios NanoLab 600i FEG-SEM equipped with an HKL EBSD detector and Aztec acquisition software. EBSD data were processed using Channel 5.0 software. An acceleration voltage of 25 keV and a beam current of 5.5 nA were used for the measurements. For the low and high-magnification maps, step sizes of 0.5 µm and 0.1 µm were used, respectively. Grain size was measured based on grain boundaries with a misorientation greater than 15°. Areas of 180000 µm² and 7500 µm² were analysed for the low and high-magnification maps, respectively. These two types of maps enabled the analysis of both large columnar grains and melt pools morphology, along with fine equiaxed grains. Given the irregular nature of the microstructures, grains were classified into three populations: small equiaxed, medium equiaxed and columnar. Small equiaxed grains were defined as having an equivalent diameter smaller than 2 µm ($d_{eq} = \sqrt{4S/\pi}$). Medium-sized grains were larger than 2 µm and not columnar, while columnar grains were defined as has having an equivalent diameter greater than 3 µm and an aspect ratio larger than 2 ($AR = \text{major axis}/\text{minor axis}$) [70]. This classification in three different populations is purely geometric and allows to identify clearly grain size changes in the microstructure. Additionally, bi-modal melt pool morphology exhibits two different zones: its core and its borders. The core is referred as coarse grain zone (CGZ) since, in as-built (AB) condition, no Al_3Zr grain refining precipitate was formed, creating mainly columnar grains. The borders of the melt pool are referred as fine grain zone (FGZ) since in AB condition, Al_3Zr grain refining sub-micrometric precipitates were formed, creating mainly small equiaxed grains as explained and shown experimentally in subsection 3.3. To reflect to the highly inhomogeneous grain structure, a surface-weighted average grain size was used instead of a simple arithmetic mean. This surface-weighted average grain size ($d_{surf\ weighted}$) was computed by multiplying the mean diameter each grain population by its surface fraction ($d_{surf\ weighted} = f_{col} \cdot d_{col} + f_{med} \cdot d_{med} + (1 - f_{col} - f_{med}) \cdot d_{fine}$, where f_{col} and f_{med} the surface fraction of columnar and medium grains respectively and d_{col} , d_{med} and d_{fine} the mean diameter of columnar, medium and fine grains respectively). Samples for microstructural characterisation were mechanically polished using diamond sprays with a minimum particle size of 1 µm, followed by final polishing with a 0.04 µm colloidal silica suspension.

High-resolution microstructural analysis and precipitate characterisation were conducted using transmission electron microscope with aberration correction and a monochromatic source (TEM FEI TITAN Themis 300). The samples were mechanically polished to a thickness between 32 µm and 44 µm and, subsequently electro-polished in a solution composed of 70% methanol and 30% nitric acid. Current (100-170 mA) and tension (12.5 V) were applied at a temperature ranging from

-12°C and -1.6°C. Final polishing was performed using a Precision Ion Polishing System (PIPS).

To identify a tailored HT, microhardness measurements were conducted to determine the condition producing the highest hardness. These measurements were performed using a DuraScan G5 indenter (EmcoTest, Austria), employing Vickers indentation method with an applied load of 0.3 kg (HV0.3) and a dwell time of 10 seconds. The average hardness was calculated from nine indentations per condition. The high temperature heat treatments (> 350°C for this study) were conducted in a Nabertherm furnace, and the process was followed by water quenching. Ageing HT (120°C) was applied in a Binder furnace. No protective atmosphere was used during either HT.

Cylindrical tensile samples were machined from the L-PBF samples described in subsection 2.2, using both electrical discharge machining and turning. Their geometry complied with the ASTM-E8/E8MA-15a standard, exhibiting a total length of 60 mm, a diameter of 5 mm, and a reduced section of 4 mm. A detailed schematic is presented in Fig. A.3. Tensile testing was conducted using a Zwick/Roell 50 kN testing machine operating in position control mode. A crosshead speed of 1 mm/min and a gauge length of 20 mm were considered. The reported tensile properties corresponded to the mean values obtained from three tests per condition.

3. Results

3.1. Chemical composition

Table 1 presents ICP-OES results obtained from a representative sample processed using the optimised parameters described in subsection 2.2. The analysis reveals losses of 27% and 37% in Mg and Zn, respectively, during the L-PBF process. These losses are calculated based on the composition of the specific powder batch used to fabricate the analysed sample (see Table A.1 for the reference powder composition). The final concentrations of 1.76% Mg and 3.32% Zn place the produced sample outside the standard composition range for Al7075 (2.1-2.9% Mg and 5.1-6.1% Zn [71]).

3.2. Heat treatment selection

Optimisation of the L-PBF process was conducted as detailed in a previous study [62]. A density of 99.9% was achieved using an unconventional pre-sintering scanning strategy, and hot cracking was entirely suppressed. A summary of the effect of pre-sintering on the density of Al7075+1.8%Zr is shown in Fig. A.4. The first main objective of this study involved identifying a HT that enabled Al7075+1.8%Zr to attain the highest possible strength. HT required optimisation for the unique L-PBF microstructure and altered alloy composition, while accounting for the presence of Zr. The standard treatment for wrought Al7075 typically comprises a high temperature HT (between 465 and 490°C), followed by a low temperature artificial ageing (AA) (24 h at 120°C) [72]. First, microhardness measurements were performed to identify the best first step high temperature heat treatment (HT1). This step was followed by tensile testing, prior to any microstructural characterisation. These procedures enabled the selection of the two most relevant HT conditions for comparison, ultimately fulfilling the primary aim of this study: understanding the effect of an optimised HT.

Fig. 1(a) presents the microhardness values of Al7075+1.8%Zr samples subjected to various HT1 between 350°C and 530°C, followed by AA (24 h at 120°C). The HT1 duration was fixed at one hour. A maximum hardness of 162 HV was observed for the sample treated at 425°C, followed by AA. Consequently, 425°C was selected as the optimised HT1 temperature. This optimised HT was used for comparative studies with the 470 °C HT1 typically used for wrought Al7075 [73,74].

Fig. 1(b) presents the tensile tests data obtained for Al7075+1.8%Zr. Only one representative curve is shown for each condition. The average mechanical properties, based on the mean of three tests, are illustrated

Table 1

ICP-OES measurement performed on the Al7075 and ZrH_2 mixture powder feedstock and on a L-PBF sample processed with the optimised set of parameters. Reproduced from earlier work [62].

		[%wt]	Al	Cr	Cu	Fe	Mg	Si	Zn	Zr
Powder	Mean		bal	0.2	1.53	0.08	2.39	0.03	5.49	1.79
	St. Dev.		-	0.00	0.06	0.01	0.03	0.01	0.21	0.08
L-PBF sample	Mean		bal	0.22	1.60	0.08	1.76	0.03	3.32	1.82

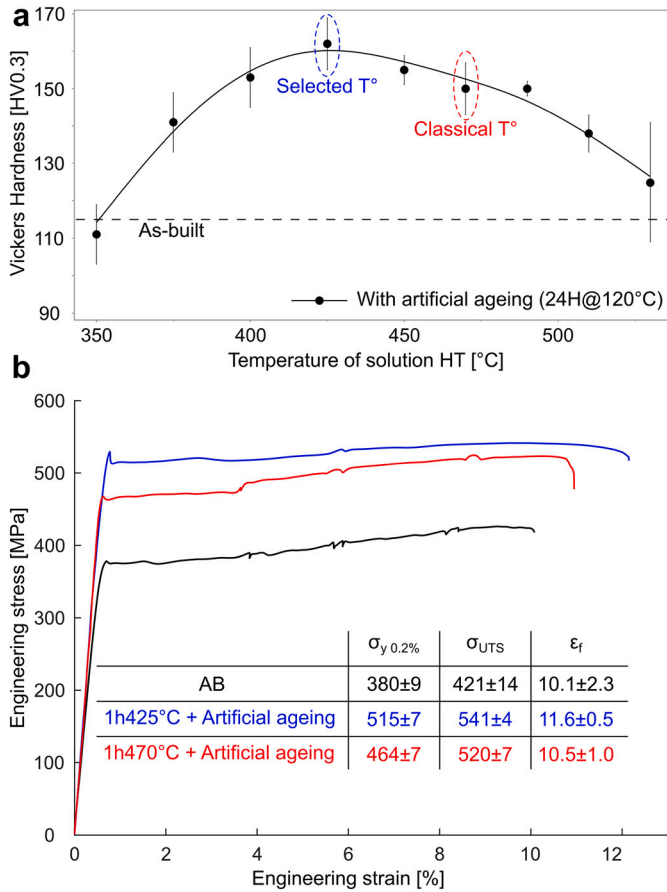


Fig. 1. (a) Vickers microhardness measurements for several 1 h first step heat treatments followed by artificial ageing (24 h at 120°C). (b) Tensile curves of the AB state (black), 1h425°C + artificial ageing HT (blue) and 1h470°C + artificial ageing HT (red).

in Fig. 1(b). All individual tensile curves used to calculate the mean values are depicted in Fig. B.1. The as-built (AB) condition exhibited lower strength and ductility compared to the heat treated samples. HT improved the 0.2% yield strength (YS) by 130 MPa under the optimised HT1 condition of 425°C followed by AA, attaining a YS of 515 MPa. In addition to this high strength, the elongation at fracture remained at 11.6%, which was slightly higher than that of the AB condition). The elongation at fracture remained relatively constant across all HT conditions. The work-hardening for the specimen heat-treated at 470°C was higher than that of the 425°C HT condition (56 MPa compared to 26 MPa). The effect of AA was analysed by comparing the obtained results with those of naturally aged samples (aged for at least for 7 days), as shown in Fig. B.1. AA significantly increased strength for both HT1 conditions: +23 MPa for 425°C and +31 MPa for 470°C. A more detailed discussion on the effect of AA is provided in Appendix B. Finally, serrations were observed in the plastic deformation regime. These serrations were more pronounced for in the AB condition than in the

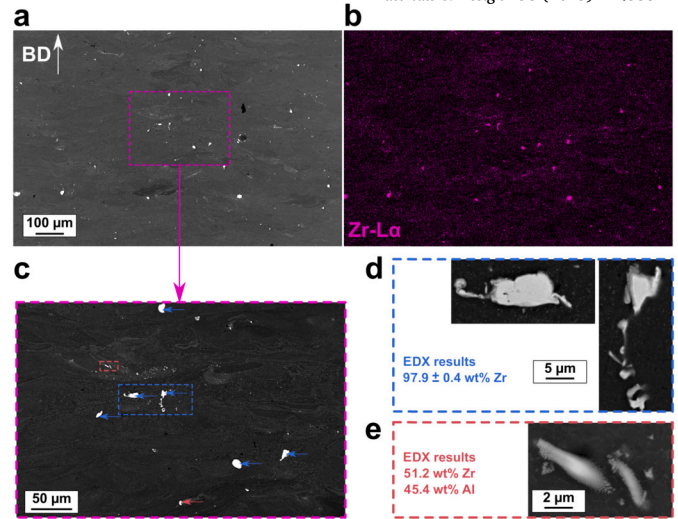


Fig. 2. (a) and (c) low resolution SEM images of AB Al7075+1.8%Zr showing the distribution of Zr rich particles inside the microstructure. (d) and (e), high resolution SEM image of Zr particles (d) or Al-Zr large precipitates (e). (b) EDX mapping identifying the Zr rich zones of (a). The arrows on (c) are showing where EDX was performed to obtain the composition of white particles. The blue ones refer to Zr-rich particles while the red one refers to Al-Zr large precipitates.

HT conditions. When AA was applied, these serrations were less pronounced compared to those observed after HT1, followed by natural ageing (Fig. B.1).

3.3. Coarse microstructure

Fig. 2 presents low-magnification SEM micrographs of Al7075+1.8%Zr in AB condition, revealing Zr-rich particles dispersed throughout the microstructure. Fig. 2(b) highlights the Zr-rich particles in purple in the EDX map. At high magnification, two types of particles were identified by EDX at the referenced locations, as shown in Fig. 2(c). Fig. 2(d) provides a magnified view of the predominantly larger particles, identified as nearly pure Zr. Such particles corresponded to ZrH_2 powder particles that were too coarse for complete melting. Fig. 2(e) shows the second type of particles: smaller, more sparsely distributed (approximately 5 µm in size). Based on the EDX results, they may correspond to large Al-Zr precipitates or partially melted ZrH_2 particles, but further analyses are needed to precisely identify them. Such occurrences of Zr-rich species have been reported in the literature following L-PBF processing of a powder mix composed of Al7017, 3 wt% Zr, and 0.5 wt% TiC. Mehta et al. [23] observed two types of Zr inclusions in the microstructure: unmelted Zr powder particles and partially melted Zr particles that exhibited various Al-Zr phases at the interface between the Zr particle and the Al matrix, making their observations similar to those in this work. They also identified these Zr-rich phases as preferential sites for crack propagation using in-situ tensile testing within a synchrotron. Thus, the observed Zr-rich particles could play a role in crack propagation in the Al7075 + 1.8% Zr alloy studied here. The analysis of this impact lies beyond the scope of the present study.

To thoroughly analyse the complex microstructure of L-PBF Al7075+1.8%Zr, higher magnification SEM micrographs were required. Fig. 3 presents a set of SEM images at various magnifications for three different conditions: AB, HT1 at 425°C and HT1 at 470°C, both followed by AA. As mentioned previously, these two HT conditions were selected because HT1 at 425°C generated the highest hardness (Fig. 1(a)) and HT1 at 470°C is the reference temperature for wrought Al7075 T6 HT.

Fig. 3(a1–c1) presents a classical bimodal microstructure. The melt pool organisation in all these samples was irregular, with melt pools varying in shape and depth, despite being expected to have a constant geometry, since the L-PBF parameters were held constant (see also

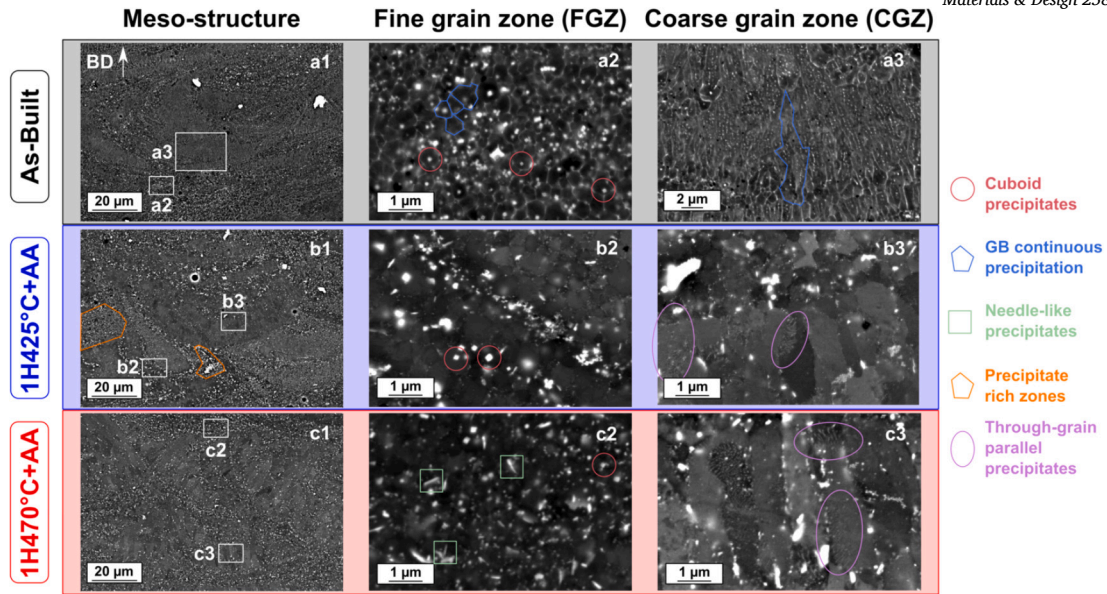


Fig. 3. SEM micrographs made using the backscattered electron detector showing Al7075+1.8%Zr in three different conditions: (a1) to (a3) in AB condition, (b1) to (b3) after undergoing a HT1 of 1 h at 425°C followed by AA at 120°C for 24 h and (c1) to (c3) after the same HT but changing the HT1 temperature from 425°C to 470°C.

Fig. 4). These irregularities are caused by balling and by spatters that disrupt the continuity of the powder bed. This phenomenon is clearly visible on the last layer of the Al7075 + 1.8% Zr samples, which exhibit significant surface roughness (see Fig. A.4(c)). It is this excessive roughness of each layer that induces the formation of irregular melt pools. To thoroughly characterise this bimodal microstructure, both the core, referred to as the coarse grain zone (CGZ), and the borders, referred to as the fine grain zone (FGZ), of the melt pool were analysed for each condition. For the AB condition, the FGZ small grains were decorated with a fully continuous network of precipitates identified in blue in Fig. 3(a2). Small sub-micron cuboidal precipitates were preferentially located at the centres of the equiaxed grains as circled in red on Fig. 3(a2). These well-known precipitates comprised Al_3Zr $L1_2$, which were responsible for grain refinement upon the addition of Zr to the alloy [12,17,70]. The characteristics were confirmed by TEM analyses (Fig. C.1). In the CGZ, columnar grains were also decorated with a continuous network of precipitates, as shown in Fig. 3(a3). Further, another population of precipitates located inside those grains identified using TEM (Fig. C.1).

For the HT conditions (Fig. 3(b1-b3,c1-c3)), the first notable phenomenon to notice involved the dissolution of the continuous precipitate network, previously highlighted in blue for AB condition (Fig. 3(a2,a3)). The sub-micron cuboidal Al_3Zr $L1_2$ precipitates, highlighted in red, remained observable at the centres of the equiaxed grains, as shown Fig. 3(b2,c2). However, for the 470°C condition, these precipitates were sparsely distributed and significantly difficult to identify than in AB condition. This lower density of precipitates is explained exploiting TEM analyses in subsection 3.5. Continuing on the comparison between the two HT conditions, parallel thin precipitates traversing the full width of certain grains were observed in the CGZ of both conditions, marked by purple ellipses on Fig. 3(b3,c3). Furthermore, the precipitates previously present within the columnar grains were no longer observed after HT. These parallel thin precipitates were difficult to distinguish in SEM images and were better characterised using TEM.

The primary difference between the two HT conditions was in the FGZ. Specifically, a new population of precipitates was observed in the 425°C sample. These precipitates were significantly small for identification using SEM (Fig. 3(b2)). Conversely, the 470°C sample exhibited another population of precipitates, partially comprising needle-like features, identified in green in Fig. 3(c3). Additionally, some irregu-

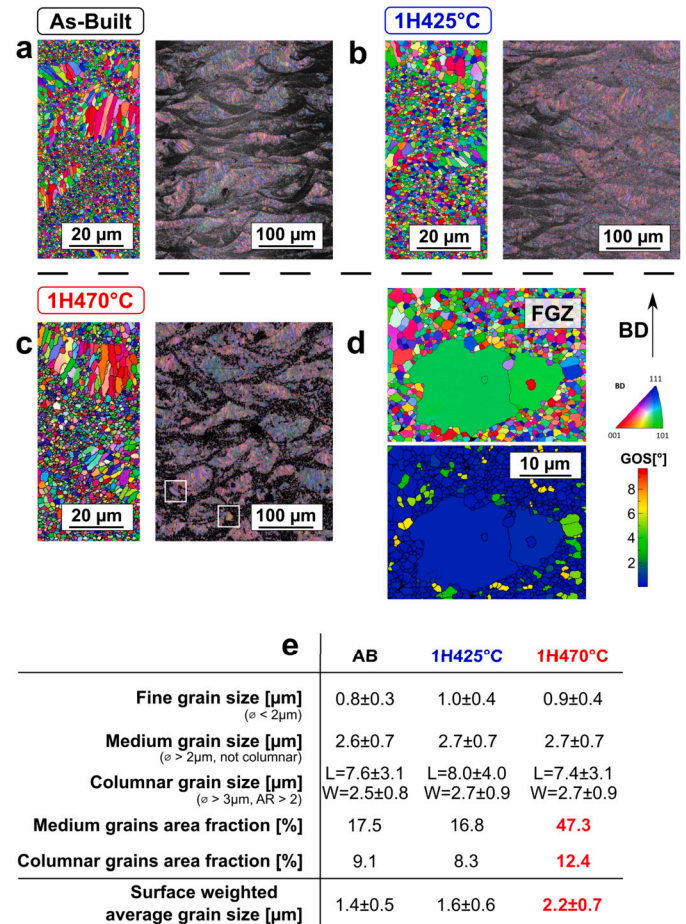


Fig. 4. Detailed high magnification IPF in the building direction EBSD map and a low magnification large Euler angle EBSD map of (a) as-built Al7075+1.8%Zr, (b) after a 1 h at 425°C HT and (c) after a 1 h at 470°C HT. (d) High resolution EBSD IPF map and grain orientation spread (GOS) maps of the 1h470°C sample exhibiting unusual grain growth in the fine grain zone. (e) Grain statistics based on maps from (a), (b) and (c). The as-built EBSD mapping (a) was already used in a previous work [62].

lar zones, highlighted in orange in Fig. 3(b1), were observed for both HT conditions, as shown in Supplementary Fig. A.5, displaying a different microstructure to that in Fig. 3(b2). These regions were enriched in precipitates, comprising a high surface fraction of the new HT-induced precipitate population, but with a locally elevated density. They exhibit a clearly uneven distribution of Zr due to the initial mixing of the powders. In some of these regions, located near large Zr-rich particles (previously identified in Fig. 2(d,e)), a high density of Al_3Zr $L1_2$ was observed (Supplementary Fig. A.5(a3,b3)). In these specific locations, the Al_3Zr $L1_2$ precipitates adopted a flower-like morphology, which is characteristic of a high local concentration of Zr in the alloy [13]. In the present alloy, such precipitates were only observed locally. All precipitates not identifiable by SEM are further characterised using TEM in subsection 3.5.

3.4. Grain structure

To characterise the effect of HT on the grain structure, EBSD analyses were performed for AB, 1h425°C and 1h470°C conditions. Fig. 4(a-c) presents high and a low magnification EBSD maps, providing sufficient statistical representation of both columnar and small equiaxed grains. These maps revealed that, under all three conditions, the bottom of the melt pool is composed of fine grains. Therefore, the applied HTs did not affect the overall bimodal grain distribution.

Due to the highly heterogeneous grain structure, the grains were categorised into three distinct populations: fine, medium and columnar grains (subsection 2.3). Fig. 4(e) shows no significant effect of HT on the grain size of each population. Only a relatively minor increase in fine-grain size was observed following both HT: the fine grains in AB condition exhibited a mean diameter of 0.8 μm , which increased to 1 μm and 0.9 μm after 1 h at 425°C and 1 h at 470°C, respectively. Notably, no specific crystallographic texture was observed in the AB condition. The absence of texture remained consistent after both HTs. The inverse pole figures confirming this lack of textures are presented in Fig. A.6.

The main effect of HT was observed by analysing the area fractions of the medium and columnar grains. For the HT at 425°C, the area fraction remained approximately the same as in the AB condition: typically 17% of the area for medium grains and 9% for columnar grains. Conversely, the HT at 470°C resulted in a significant increase in the area fractions of the medium and columnar grains, reaching 47% and 12%, respectively. These trends were corroborated by the evolution of the surface-weighted average grain size. The initial value of 1.4 μm in the AB condition increased slightly to 1.6 μm after 1 h at 425°C, and further increased to 2.2 μm after 1 h at 470°C. Fig. 4(d) presents a magnified view of a large grain located in the FGZ of the 1 h at 470°C sample. The presence of such a large grain located in the FGZ was specific to the 470°C condition. No such grains were observed in the FGZ for the AB and 425°C conditions. This phenomenon corresponded to the overall grain growth tendency, which was only evident following the 470°C HT. This large grain exhibited a grain orientation spread (GOS) below 2°, indicative of a recrystallized grain [75,76].

3.5. Coarse precipitation

A schematic summarising the complex microstructural analysis based on the following TEM results is provided in section 4 (Fig. 9). The schematic is intended to facilitate the understanding of the relatively complex descriptions that follow.

Since the AB state was not the primary focus of this study and has already been extensively documented in the literature, only a low-resolution TEM analysis at was conducted, as discussed in Appendix C. In summary, the grain boundary (GB) continuous network of precipitates (observed in Fig. 3(a1-a3)) was characterised as rich in Cu, Mg and Zn with some Fe segregation. This observation was consistent with previous reports for L-PBF Al7xxx series alloys [18,27,35,40,68,77,78]. Squared cuboid precipitates were identified as Al_3Zr $L1_2$, as shown in

Fig. C.1(b). Finally, a higher Zr concentration in solid solution is measured in the CGZ compared to the FGZ: 0.9 at.% compared to 0.6 at.%, respectively (Table A.2).

Fig. 5 presents TEM micrographs of the optimised HT condition (1 h at 425°C + AA). As observed in the SEM images in Fig. 3, the continuous Cu, Mg and Zn-rich phase initially present at the grain boundaries has been entirely dissolved in both the CGZ and the FGZ. Al_3Zr $L1_2$ remained within the fine grains even after the 1h425°C + AA HT, as shown in Fig. 5(b,c). A magnified view of one of these precipitates is presented in Fig. A.7. It shows a segregation of both Zn and Si within the Al_3Zr $L1_2$ precipitate. This phenomenon was beyond the scope of this study and was not further analysed. Fig. 5(a,b) presents EDS maps highlighting the presence of Cu and Fe-rich plate-shape precipitates located at GB, corresponding to previous Fe segregation location observed in the AB condition, as shown in Fig. C.1. EDS quantification of these precipitates revealed 4 at.% of Cu and 1.7 at.% of Fe, which approximately matched the stoichiometry of the Al_7Cu_2Fe phase. To confirm this identification, an EDS line scan was conducted on the magnified Al_7Cu_2Fe precipitate shown in Fig. 5(c). Fig. 5(d) shows Cu and Fe content of 10 at.% and 4.5 at.%, respectively. Notably, Al_7Cu_2Fe precipitates of similar morphology and location have been reported for heat-treated Al-Zn-Mg-Cu alloys produced by L-PBF [40,68]. Two additional, smaller second phases were detected, as shown in Fig. 5(c). First, sparse Mg oxides with diameters below 30 nm were observed. These oxides originated from the powder surface or were formed during the L-PBF process. No further analysis on these oxides were conducted in this study. Second, a Cr and Mg-rich phase was observed, as depicted in Fig. 5(c). These dispersoids exhibited a mean diameter of 33 ± 8 nm and occupied 1.4% of the analysed volume, assuming the TEM sample comprised a constant thickness of 50 nm (275 dispersoids analysed spread over 6 different areas). The particles were identified as $Al_{18}Mg_3Cr_2$ dispersoids. This is a common phase observed in 7xxx aluminium alloys with size, shape and relative atomic ratio matching this work [79–81]. $Al_{18}Mg_3Cr_2$ was the only Al-Mg-Cr ternary phase reported in the literature for such elemental ratios. This phase was identified experimentally using diffusion couple experiments and thermodynamic simulations [82]. Jiang et al. [35] observed $Al_{18}Mg_3Cr_2$ dispersoids in L-PBF 7075. They reported a 50 MPa contribution to YS. Therefore, these $Al_{18}Mg_3Cr_2$ dispersoids were considered during the yield strength prediction, as discussed in subsection 4.1.

Fig. 6 presents TEM micrographs similar to those in Fig. 5, but corresponding to the second, less hardening, HT examined in this study, namely 1 h at 470°C followed by AA. Semi-quantitative EDS results for the different phases observed are provided in Table A.5. Al_7Cu_2Fe precipitates, Mg oxides and $Al_{18}Mg_3Cr_2$ dispersoids were observed after this HT. However, the $Al_{18}Mg_3Cr_2$ exhibited a larger mean diameter and volume fraction: 56 ± 15 nm (70% larger than for 1h425°C + AA condition) and a volume fraction of 2.9% (+1.5% relative to 1 h 425°C + AA condition) (176 dispersoids analysed spread over 4 different areas). The main difference compared to the 1h425°C+AA condition is the presence of needle-shaped Zr-rich precipitates, as shown in Fig. 6(b). This morphology is characteristic of $D0_{23}$ Al_3Zr precipitates, as reported for heat-treated Al L-PBF alloys containing Zr. Al_3Zr precipitates have been identified in the literature using SEM [12,17,83], X-ray diffraction [10,83], or TEM [84] in Zr-modified Al alloys. These stable $D0_{23}$ Al_3Zr precipitates require a sufficiently high heat treatment temperature to form, through transformation from the initially metastable $L1_2$ Al_3Zr , as demonstrated for an Al–0.1 at.% Zr alloy [85,86]. This transformation from the metastable to the stable form of Al_3Zr explains the lower density of $L1_2$ Al_3Zr precipitates observed for the 470°C condition, as shown in Fig. 3. Si segregation, similar to that in Fig. A.7, is observed in Fig. 6(b) and in the EDS line scan of Fig. 6(c). These $D0_{23}$ Al_3Zr precipitates corresponded to the needle-shaped precipitates reported in the SEM analyses, as shown in Fig. 3(c2) (and in Fig. A.5(b1-b3)). Despite some of the $L1_2$ Al_3Zr precipitates being transformed into $D0_{23}$ Al_3Zr , some $L1_2$ precipitates remained in the microstructure after 1 h

1H425°C+AA

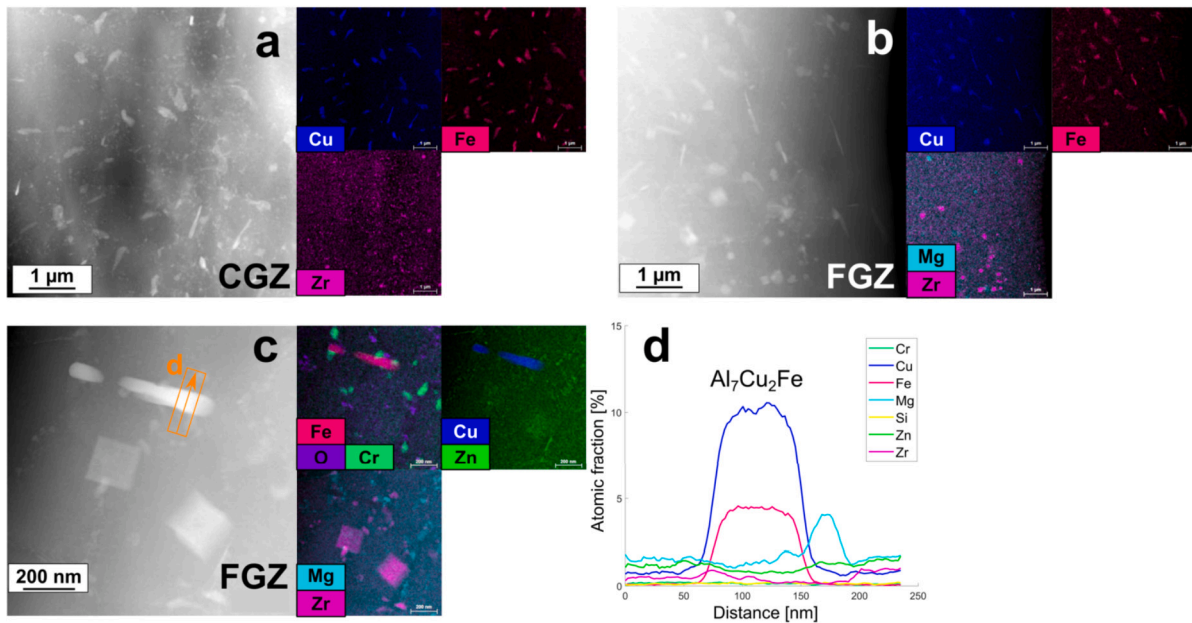


Fig. 5. Low magnification TEM analysis of the global microstructure of L-PBF Al7075+1.8%Zr after 1 h at 425°C followed by artificial ageing. (a) HAADF-STEM micrographs of the coarse grain zone (CGZ) and (b) of the fine grain zone (FGZ). (c) Zoomed view of the FGZ showing both Al_3Zr $L1_2$ grain refiner precipitates and Cu, Fe-rich needle like precipitates. STEM-EDS mappings of the different precipitates are also available (X-ray emission lines: Cu K_{α} , Cr K_{α} , Fe K_{α} , Mg K_{α} , O K_{α} , Zn K_{α} , Zr K_{α}). (d) EDS line scan of a plate shaped Cu, Fe-rich (highlighted in orange on (c)) exhibiting the atomic ratio of Al_7Cu_2Fe . STEM-EDS mappings of additional elements for (a) and (b) are available in Fig. A.9.

1H470°C+AA

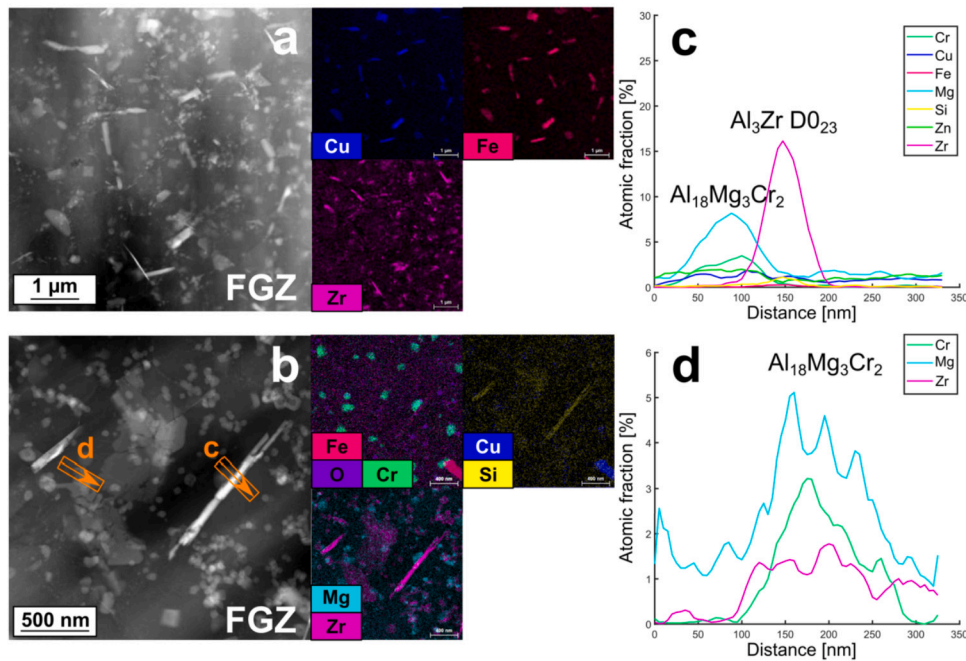


Fig. 6. Low magnification TEM analysis of the global microstructure of L-PBF Al7075+1.8%Zr after 1 h at 470°C followed by artificial ageing. (a) HAADF-STEM micrograph of the fine grain zone (FGZ). (b) Zoomed view of the FGZ showing Al_3Zr $D0_{23}$ precipitates, a Zr-rich phase at grain boundaries and $Al_{18}Mg_3Cr_2$ dispersoids. STEM-EDS mappings of the different precipitates are also available (X-ray emission lines: Cu K_{α} , Cr K_{α} , Fe K_{α} , Mg K_{α} , O K_{α} , Si K_{α} , Zr K_{α}). (c) and (d) EDS line scans of needle shaped Al_3Zr $D0_{23}$ and $Al_{18}Mg_3Cr_2$ dispersoids (highlighted in orange on (b)). STEM-EDS mappings of additional elements for (a) are available in Fig. A.9.

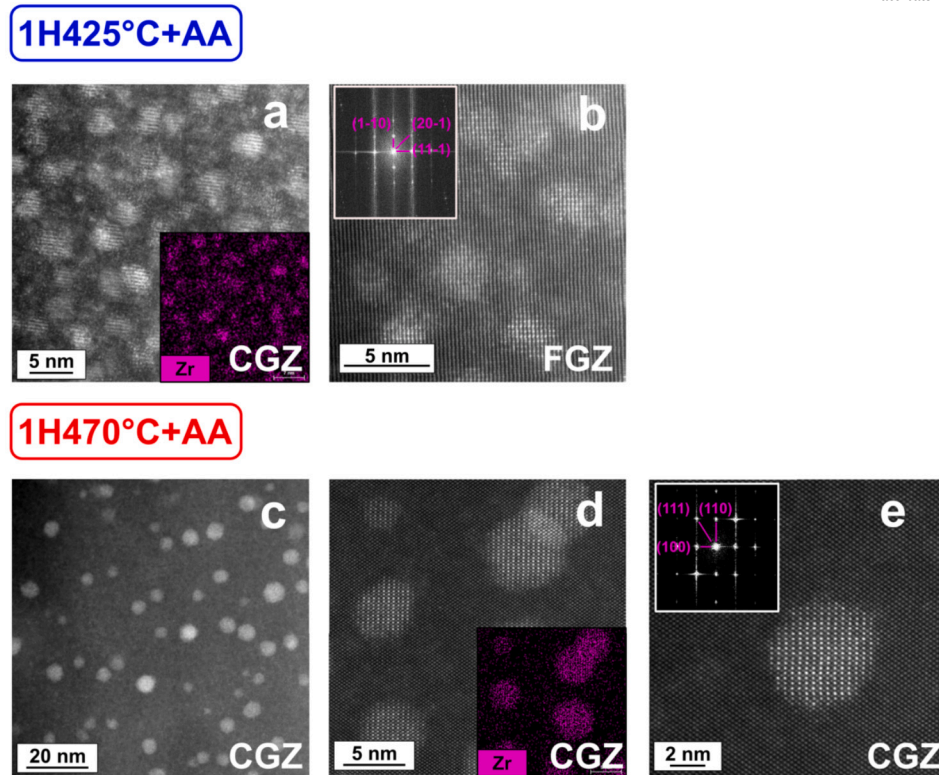


Fig. 7. Hardening Al_3Zr $L1_2$ precipitates comparison of L-PBF Al7075+1.8%Zr after two different heat treatments. (a) and (b) High resolution TEM micrographs for the 1h425°C followed by artificial ageing condition: (a) coarse grain zone (CGZ) and (b) fine grain zone (FGZ). (b) includes the diffraction pattern of the coherent $L1_2$ Al_3Zr precipitate along the $[1\ 1\ 2]$ zone axis. (c), (d) and (e) High resolution TEM micrographs for the 1h470°C followed by artificial ageing condition in the coarse grain zone. (e) includes the diffraction pattern of the coherent $L1_2$ Al_3Zr precipitate along the $[0\ -1\ 1]$ zone axis. STEM-EDS mappings of (a) and (d) are provided as inset (X-ray emission line: Zr K_α).

at 470°C followed by AA. Finally, some Zr rich phases were observed at GB. These phases were not observed after the 1h425°C + AA HT.

At a relatively low resolution for TEM, an additional Zr-rich phase was observed for both HT in the CGZ, as shown in supplementary Fig. D.1. This Zr-rich phase comprised parallel precipitation lines that spanned across an entire grain. The diffraction pattern of one line Zr-rich precipitate corresponded to Al_3Zr $D0_{23}$ with a $[2\ -2\ -1]$ zone axis and diffraction spots corresponding to the reflection of axes (110), (204) and (1-14). This phase has been observed for L-PBF Al alloys in existing literature [17,84]. The observed discontinuous precipitation is detailed in Appendix D. However, this phase was not significantly important in this study, as it was only scarcely observed (Fig. 3(b3,c3)) and did not influence the mechanical properties of Al7075+1.8%Zr.

3.6. Fine hardening precipitation

Using HR-STEM micrographs, another type of Zr precipitate with a diameter below 10 nm was observed, as shown in Fig. 7. For both HT, the precipitate were identified as $L1_2$ Al_3Zr secondary precipitates. Secondary precipitates were formed by HT of the Zr supersaturated solid solution, resulting from the high cooling rate of the L-PBF process. For the sample after 1 h at 425°C followed by AA, the diffraction pattern of corresponded to zone axis $[1\ 1\ 2]$ of $L1_2$ Al_3Zr (parallel to the $[1\ 1\ 2]$ zone axis of α Al) with the diffraction spots corresponding to reflection of axis (1-10), (20-1) and (11-1), as shown in Fig. 7(b). For the sample after 1 h at 470°C followed by AA, the diffraction pattern corresponded to zone axis $[0\ -1\ 1]$ of $L1_2$ Al_3Zr (parallel to the $[0\ -1\ 1]$ zone axis of α Al) with the diffraction spots corresponding to reflection of axes (100), (111) and (110), as shown in Fig. 7(e). This identification of $L1_2$ Al_3Zr hardening precipitates corresponded to that reported in the literature for L-PBF high-strength Al alloys with Zr addition [10,15,18,20,24,84].

The mean radius of these nanometric $L1_2$ Al_3Zr precipitates (excluding the large grain refining Al_3Zr observed in Fig. 5 and Fig. 6) was obtained by image analysis, and their volume fraction was computed assuming the TEM sample exhibited a constant thickness of 50 nm. For the 1h425°C + AA sample, a mean radius of 1.7 ± 0.2 nm and a volume fraction of 1.24% were obtained for the CGZ (250 precipitates analysed spread over 4 different areas). For the FGZ, the $L1_2$ Al_3Zr were smaller with a mean radius of 1.3 ± 0.2 nm and occupied lower volume fraction, estimated at 0.65% (97 precipitates analysed spread over 5 different areas). This difference was correlated to the higher Zr concentration in solid solution for the CGZ (Table A.2). For the 1h470°C + AA sample, only the CGZ was observed in the TEM images. For this zone, the mean radius was estimated at 2.3 ± 0.5 nm and the volume fraction was measured at 0.83% (78 precipitates analysed spread over 4 different areas). Compared to the $L1_2$ Al_3Zr of the CGZ in the 1h425°C + AA HT condition, an increase of 35% in the radius and a decrease of 33% in the volume fraction were recorded. These Zr-rich phases were over-aged with the higher temperature HT. Indeed, Al_3ZrL1_2 precipitates are only metastable, when subjected to higher HT temperature, they tended to form more stable Zr-rich phases, such as the Al_3ZrD0_{23} and the Zr-rich phase at the GB observed after the 470°C HT (Fig. 6(b)). Zr being captured by these Zr-rich phases only observed for the 470 °HT, it explains that a higher volume fraction of nanometric $L1_2$ Al_3Zr precipitates was observed after the lower temperature HT.

Fig. 8 shows Zn-rich precipitates typically reported in Al7075 alloy. Fig. 8(a) presents η' precipitates located at the GB of the FGZ for the 425°C HT. These Zn-rich precipitates are illustrated in Fig. 5 and Fig. A.7. They were identified by the $MgZn_2$ ratio sampled by the two EDS line scans, as shown in Fig. 8(c,d). Fig. 8(b) illustrates a similar $MgZn_2$ precipitate, but this precipitate was identified as an η' specie. The diffraction pattern demonstrated a $[0\ 1\ 1]$ zone axis of the Al matrix, and

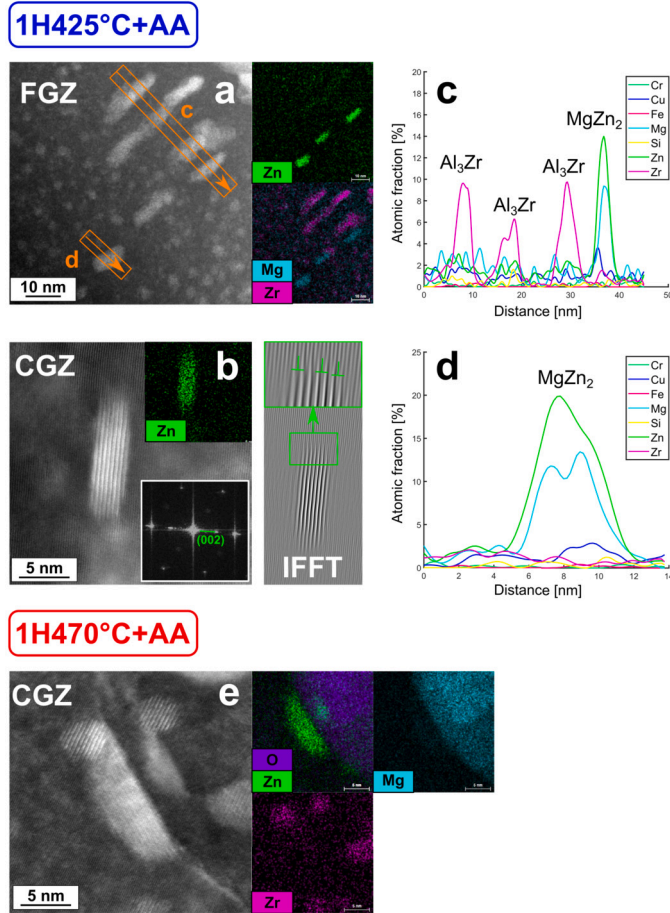


Fig. 8. $MgZn_2$ precipitates of L-PBF Al7075+1.8%Zr after two different heat treatments. (a) and (b) High resolution TEM micrographs for the 1h425°C followed by artificial ageing condition: (a) fine grain zone (FGZ) and (b) coarse grain zone (CGZ). (b) includes the diffraction pattern of the semi-coherent η' $MgZn_2$ precipitate along the [0 1 1] zone axis of Al and shows the inverse fast Fourier transform for this η' precipitate evidencing some misfit dislocations. (c) and (d) EDS line scans of Al_3Zr DO_{23} and $MgZn_2$ at GB (highlighted in orange on (b)). (e) Heterogeneous nucleation of a $MgZn_2$ on a Mg oxide in the CGZ of the 1h470°C+AA sample. STEM-EDS mappings of the different precipitates are also available (X-ray emission lines: Mg K_{α} , O K_{α} , Zn K_{α} , Zr K_{α}).

the diffraction spot corresponded to the reflection (002) of the $MgZn_2$ precipitate. The orientation relationship $(1-11)_{Al}/(002)_{MgZn_2}$ was consistent with those discussed in existing literature [87]. Furthermore, the inverse fast Fourier transform (IFFT) image (Fig. 8(b)) revealed a semi-coherency between the $MgZn_2$ precipitate and the Al matrix, which was characteristic of the η' precipitate [88–90]. Such η' precipitates have been observed for L-PBF Al7075 [40,68]. However, they have been observed in a much higher density than reported in this study. These Zn-rich precipitates were difficult to locate, as they were scarcely present. Only isolated η' precipitates (as in Fig. 8(b)), some located at GB (as in Fig. 8(a)) or those heterogeneously grown on other species (Fig. 8(e)) could be identified. Other examples of this heterogeneous growth of $MgZn_2$ are presented on Fig. A.8. The exact mechanism of this heterogeneous nucleation was not analysed in details and remains to be fully understood. No homogeneous distribution of η' was observed within the grains, despite more than 10 areas analysed of the TEM samples for both HT conditions. Finally, no major difference between the $MgZn_2$ precipitates distribution for the 425°C condition and for the 470°C condition was observed, as the precipitates were formed during the AA. The reasons leading to the scarcity of these $MgZn_2$ precipitates were discussed in subsection 4.1.4.

4. Discussion

Fig. 9 summarises the microstructure observed for the AB, 1h425°C + AA and 1h470°C + AA conditions. The primary tensile behaviours observed in Fig. 1(b) are also reminded on Fig. 9. Consequently, this scheme established the relationship between the tensile behaviour of the analysed samples and their specific microstructure.

4.1. Yield strength

This section discusses the differences observed between the YS of Al7075 +1.8%Zr generated using the two different HT1 temperatures. Assuming that each hardening mechanism acted independently of others and that the flow stress of the alloy was estimated as the sum of separate contributions from each active hardening mechanism [91], YS was estimated as the sum of four main contributions: grain boundary strengthening (σ_{GB}), back-stress strengthening (σ_{BS}), solid solution strengthening (σ_{SS}) and precipitate strengthening (σ_{PS}). Therefore,

$$\sigma_{YS} = \sigma_{GB} + \sigma_{BS} + \sigma_{SS} + \sigma_{PS} \quad (1)$$

To evaluate these three contributions, EBSD and TEM results were required to quantify the key microstructural features of the 1h425°C + AA and the 1h470°C + AA samples. The AB state was not analysed in this study because no HR-TEM analyses were conducted for this condition. The higher and more interesting mechanical properties of the HT state did not require consideration of the AB state. All material parameters and constants required to estimate the four different contributions of YS were extracted from existing literature, as summarised in Table 2.

4.1.1. Grain boundary strengthening

Grain boundary strengthening was estimated using the Hall Petch relationship [40,44,47,54]:

$$\sigma_{GB} = \sigma_0 + kd^{-0.5} \quad (2)$$

where σ_0 and k are provided in Table 2. d corresponds to the average grain size that can be identified using the EBSD data, as depicted in Fig. 4. However, since the grain structure of Al7075+1.8%Zr was highly heterogeneous, the following expression was used to consider the surface fraction of each population of grain:

$$\sigma_{GB \text{ global}} = \sigma_{GB \text{ fine}} \cdot f_{\text{fine}} + \sigma_{GB \text{ medium}} \cdot f_{\text{medium}} + \sigma_{GB \text{ col}} \cdot f_{\text{col}} \quad (3)$$

where f_{fine} , f_{medium} and f_{col} denote the respective area fraction of the fine, medium and columnar grains. The formula was used under the condition that (i) each of the three regions was in a state of uniaxial stress aligned perpendicular to the surface along which area fractions were measured, and (ii) that the three regions yielded simultaneously. GB strengthening of 171^{+39}_{-22} and 149^{+32}_{-18} MPa for the 1h425°C + AA and the 1h470°C + AA conditions were recorded, respectively. The smaller value of σ_{GB} reflected the recrystallization generated by the 1h470°C + AA HT owing to the increase in medium and columnar grains fractions (Fig. 4(e)).

4.1.2. Back-stress strengthening

Back-stress strengthening is not an usual contribution to the yield strength and is only considered for alloys exhibiting specific microstructures. Back stress strengthening originates from inhomogeneities in deformation between two different phases. An excessive strain in one phase compared to another induces a pile up of geometrically necessary dislocations (GND) to accommodate the generated strain gradient in the microstructure [97–101]. Therefore, it is necessary to identify the origin of back-stress strengthening for Al7075+1.8% Zr.

First, back stress can be generated by Orowan looping around non-shearable precipitates as depicted in literature for Al7075 [102, 103]. In this study, it was demonstrated that the hardening precip-

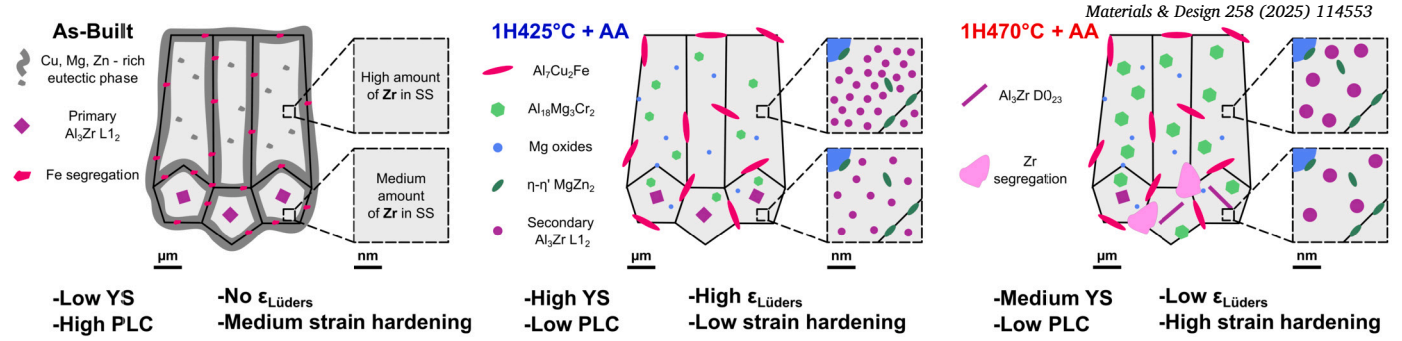


Fig. 9. Schematic representation of the microstructure of the three main conditions analyses in this study: AB, 1h425°C + AA and 1h470°C + AA. It highlights the precipitates nature and location for each condition at two different scales. Note that the scheme is only roughly scaled. The main tensile characteristics are also reminded below each schematic (YS = yield strength, PLC = Portevin Le Chatelier effect, $\epsilon_{Lüders}$ = Lüders strain, $\Delta\sigma$ = Ultimate tensile strength - YS).

Table 2

Description, values, units and references for all the symbols used for the quantification of the various strengthening mechanism.

Symbol	Name	Value	Unit	References
σ_0	Friction stress of Pure Al	20	MPa	[40,44,54]
k	Hall-Petch coefficient	0.17	$MPa \cdot \sqrt{m}$	[44,47,54]
M	Average matrix orientation factor (Al)	3.06	-	[20,40,44,47,54,61,86,92–95]
ν	Poisson's ratio (Al)	0.33	-	[20,40,61,94]
b	Magnitude of Burgers vector (Al)	0.286	nm	[20,40,44,47,54,86,92–95]
G_{Al}	Shear modulus (Al)	25.4	GPa	[20,44,47,54,86,92,93]
G_{Al_3Zr}	Shear modulus (Al_3Zr)	65.3	GPa	[96]
α_e	constant	2.6	-	[47,61,93–95]
m	constant	0.85	-	[20,44,47,61,93–95]
γ_{APB}	Anti-phase boundary energy (Al_3Zr)	0.5	J/m^2	[47,93,95]
n	Number of dislocations in a pile-up	3	-	[97]

itates of Al7075+1.8% Zr for both HT conditions are shearable (see subsection 4.1.4). Thus this hypothesis can be rejected.

Back-stress strengthening is also experienced by dual-phase steel. Both phases exhibit different deformation response due to grain size differences [98,99] or due to a difference in their crystallographic nature [100]. In this study, two different zones were observed: fine grain zone (FGZ) and coarse grain zone (CGZ). However, no difference of strains between these two zones were observed in a previous work using in-situ tensile tests combined with micro-DIC, for the same alloy as present study [104]. This observation was also confirmed by Buttard et al. [11] for L-PBF Al-4Mn-3Ni-2Cu-1Zr aluminium alloy. Consequently, this hypothesis is not valid for the present study.

Grain boundaries can also generate a heterogeneous deformation by deforming more than grain interior, generating GND to accommodate this strain mismatch. The generated GNDs make the grain boundaries appear more resistance to deformation, inducing back-stress strengthening. This phenomenon has been observed in the literature for steels [105] and for ultra fine grains aluminium alloys (grain diameter smaller than 0.7 μ m) [97]. Moreover, for the specific case of Al7075 alloy, the precipitate free zone present at grain boundaries has been identified as responsible for this back-stress strengthening [101,106]. This grain boundaries back stress strengthening has been reported in the literature to increase with decreasing grain size [97,105]. This hypothesis is valid for the present study with the grain size after HT being is only 1.6 μ m for the 1h425°C and 2.2 μ m for the 1h470°C with the presence of zones with equiaxed grains smaller than 1 μ m (see Fig. 4). The back-stress strengthening generated by grain boundaries can thus be estimated using the following relationship [97,105]:

$$\sigma_{BS} = M \frac{G_{Al} b}{d} n \quad (4)$$

where M , G_{Al} , b and n are provided in Table 2. d , like for Hall-Petch Equation (2), corresponds to the average grain size. Like for grain boundary strengthening, to take into account the heterogeneity of the grain structure, a similar equation to Equation (3) was used. It results in a contribution from back-stress strengthening of 55^{+33}_{-15} and 42^{+26}_{-12} MPa for the 1h425°C + AA and the 1h470°C + AA conditions, respectively. The smaller value of σ_{BS} reflects the recrystallisation generated by the 1h470°C + AA HT, evidenced by the increase of medium and columnar grains fraction for this condition (Fig. 4(e)).

To corroborate the hypothesis of back-stress presence in Al7075+1.8% Zr, the shape of the tensile curve (Fig. 1(b)) exhibited some similarities with various materials experiencing back stress reported in the literature. The specific and unusual yield drop just before entering in the plasticity regime observed for L-PBF Al7075+1.8% Zr is a characteristic feature of the tensile curve when samples are experiencing back-stress strengthening. The same feature has been observed for dual phase steels [98–100] and for Al alloys with ultra fine grains [97,107].

4.1.3. Solid solution strengthening

Solid solution strengthening arose from the distortion of the lattice generated by atoms in solid solution and from their interactions with dislocations. For this study, the effect of the three main alloying elements of Al7075 was considered, i.e. Cu, Mg and Zn. The amount of Zr in solid solution was considered equal to zero due to the high density of secondary Al_3Zr L1₂ hardening precipitates shown on Fig. 5 and Fig. 6. It was measured at 0.15 at.%, as shown in Fig. 6(d), having a negligible contribution to the YS. Moreover, the contribution of Mg was neglected, the concentration of Mg in solid solution obtained by EDX

ranging from 1.7 at.% to 1.9 at.% for both HT conditions, the Mg concentration was too large to be correct. The ICP analysis (Table 1) showed a total amount of Mg after L-PBF of 1.76 wt.% (2 at.%). Moreover, since Mg was already present in several Mg oxides and $Al_{18}Mg_3Cr_2$ dispersoids (Fig. 5(c), Fig. 6(b)), Mg should not be present in solid solution after HT. Thus, only the contribution of Cu and Zn will thus be summed following the expression proposed by Uesugi et al. [108]:

$$\sigma_{SS} = M \left(\frac{3}{8} \right)^{2/3} \left(\frac{1+\nu}{1-\nu} \right)^{4/3} \left(\frac{w}{b} \right)^{1/3} G_{Al} \epsilon_{SS}^{4/3} c^{2/3} \quad (5)$$

where M , ν , b , and G_{Al} are the average orientation (Taylor) factor of Al, the Poisson's ratio of Al, the magnitude of Al Burgers vector, and the shear modulus of Al (Table 2), respectively. Further, $w = 5b$ [108]. ϵ_{SS} denotes the lattice strain generated by an atom in solid solution expressed in %. This strain was equal to -1.94% and -0.38% for Cu and Zn, respectively, as obtained by Uesugi et al. [108] using first-principle calculations. Herein, c denotes the concentration in at.% of the solute in solid solution. The value of c was measured by STEM-EDS in the matrix, and the precise values are provided in Table A.3 and Table A.5 for the two HT. The σ_{SS} value of 38 and 53 MPa were obtained for the 1h425°C + AA and the 1h470°C + AA conditions, respectively.

4.1.4. Precipitate strengthening

Precipitation strengthening is of paramount importance for L-PBF Al7075+1.8%Zr as the Al_3Zr $L1_2$ precipitates exhibit mean radii between 1.3 and 2.3 nm, depending on the HT condition (Fig. 7). In addition, the $Al_{18}Mg_3Cr_2$ dispersoids can strengthen the Al7075+1.8%Zr. Depending on the size of the precipitate, two different types of interaction between dislocations and precipitates can occur: at small sizes, the dislocations shear precipitates while at larger sizes, the dislocations form Orowan loops around the precipitates [24,93–95,109]. Concerning shear strengthening, σ_{shear} is driven by three different contributions:

- Coherency strengthening (σ_{coh}) generated by the stress field interaction of the dislocations and the precipitates [63,110]. Here, σ_{coh} is defined by the following equation [47,61,93–95]:

$$\sigma_{coh} = M \alpha_e (G_{Al} \epsilon_c)^{1.5} \sqrt{\frac{R\phi}{0.5G_{Al}b}} \quad (6)$$

where α_e is provided in Table 2. ϵ_c is the room temperature lattice mismatch between the precipitate and the matrix. The value can be calculated using $\epsilon_c = \frac{2}{3} \frac{a - a_0}{a_0}$ [20,61,93] with a_0 as the lattice parameter of FCC Al (0.405 nm [20,85,86]) and a as the lattice parameter of the precipitate (0.408 nm for Al_3Zr [20,85]). R denotes the mean precipitate radius, and ϕ is the volume fraction of these precipitates.

- Modulus strengthening (σ_{mod}) arises from the difference in shear modulus of the precipitate and the matrix [63,110]. Herein, σ_{mod} can be defined by the following equation [47,61,93–95]:

$$\sigma_{mod} = 0.0055 M b \Delta G^{\frac{3}{2}} \left(\frac{R}{b} \right)^{\frac{3m}{2}-1} \sqrt{\frac{\phi}{0.5G_{Al}b^2}} \quad (7)$$

where ΔG is the shear modulus mismatch between FCC Al and Al_3Zr [44,47,95]. Since $G_{Al_3Zr} = 65.3$ GPa [96], $\Delta G = 39.9$ GPa.

- Order strengthening (σ_{order}) occurs due to the formation of antiphase boundaries (APBs) under precipitate shearing by the dislocation. The value of σ_{order} can be determined using the following equation [24,93–95,109]:

$$\sigma_{order} = 0.81 M \frac{\gamma_{APB}}{2b} \sqrt{\frac{3\pi\phi}{8}} \quad (8)$$

where γ_{APB} is provided in Table 2. Unlike σ_{mod} and σ_{coh} , σ_{order} only depends on the volume fraction of precipitates and not on their radius.

Some discrepancies in the literature on the selection of the dominant phenomenon between coherency, modulus and order strengthening. In this study, the maximum between σ_{order} and $\sigma_{mod} + \sigma_{coh}$ is selected [24,93–95,109]:

$$\sigma_{shear} = \max(\sigma_{order}, \sigma_{mod} + \sigma_{coh}) \quad (9)$$

Both σ_{mod} and σ_{coh} occur before the shearing of the precipitate, while σ_{order} occurs during the shearing of the precipitate.

The Orowan looping strengthening, dominant at larger precipitate radii, can be expressed as [24,95]:

$$\sigma_{Orowan} = M \frac{0.4G_{Al}b}{\pi\sqrt{1-\nu}} \frac{\ln(\frac{2\bar{R}}{b})}{\lambda} \quad (10)$$

where $\bar{R} = \sqrt{2/3}R$ is the mean radius of a random plane cross-section of the precipitate, assumed spherical [94]. λ is the inter-precipitate distance and can be approximated by [20,47,92,93,95]:

$$\lambda = \left(\left(\frac{3\pi}{4\phi} \right)^{0.5} - 1.64 \right) R \quad (11)$$

Precipitate strengthening can be computed for both Al_3Zr $L1_2$ hardening precipitates (σ_{Al_3Zr}) and $Al_{18}Mg_3Cr_2$ dispersoids ($\sigma_{Al_{18}Mg_3Cr_2}$). Both $\sigma_{shearing}$ and σ_{Orowan} are analysed for Al_3Zr precipitates, while, owing to their large size and incoherent interface with the Al matrix [80], only σ_{Orowan} is computed for $Al_{18}Mg_3Cr_2$ dispersoids. Since the amount of Zr available for precipitation trapped in solid solution after L-PBF is higher in the CGZ than in the FGZ (Table A.2), the volume density and size of the precipitates vary depending on these two zones (Fig. 5). Thus, exploiting the surface fraction of medium and columnar grains obtained in Fig. 4, σ_{Al_3Zr} is expressed as:

$$\sigma_{Al_3Zr} = \sigma_{Al_3Zr CGZ} \cdot (f_{medium AB} + f_{col AB}) + \sigma_{Al_3Zr FGZ} \cdot f_{fine AB} \quad (12)$$

where $f_{fine AB}$, $f_{medium AB}$ and $f_{col AB}$ denoted the surface fraction of the fine, medium and columnar grains for the AB condition, respectively. This formula can be used under the same assumptions as discussed in subsection 4.1.1 (both regions are in uniaxial stress and yield at the same time). The surface fractions of the AB condition are selected because the amount of Zr trapped in solid solution depends on the amount of medium and columnar grains generated after L-PBF and measured for the AB state. Subsequent recrystallization or grain growth should not impact the amount of precipitates generated by the two HT. No data on the precipitation in the FGZ of the 1h470°C + AA condition are available in this study. Therefore, the same ratio between the radius and volume fraction of σ_{Al_3Zr} $L1_2$ precipitates between the CGZ and the FGZ of the 1h425°C + AA condition is applied on the CGZ data of the 1h470°C + AA condition. Some uncertainty in the precise result of σ_{Al_3Zr} for this condition can be observed. However, a comparative study between the two HT is possible.

A σ_{Al_3Zr} of 209 MPa and a $\sigma_{Al_{18}Mg_3Cr_2}$ of 48 MPa were obtained for the 1h425°C + AA condition. In comparison, a σ_{Al_3Zr} of 172 MPa and a $\sigma_{Al_{18}Mg_3Cr_2}$ of 48 MPa were obtained for the 1h470°C + AA condition. No difference was observed for the $\sigma_{Al_{18}Mg_3Cr_2}$ precipitates between the two HT. The smaller size of dispersoids for the 1h425°C + AA HT was compensated by their lower volume fraction. Jiang et al. [35] obtained a similar value of $\sigma_{Al_{18}Mg_3Cr_2}$ for L-PBF Al–Zn–Mg–Cu+TiH₂ alloy.

The contribution of Al_3Zr $L1_2$ precipitate strengthening was 18% lower for the 1h470°C + AA condition than for the 1h425°C + AA condition. Despite the size of the precipitates being slightly more efficient for the higher temperature HT, the difference in volume fraction served as the determining factor of that significant difference. Fig. 10(a) shows that after the 470°C HT, σ_{Al_3Zr} $L1_2$ precipitates are in the shearing domain where $\sigma_{mod} + \sigma_{coh}$ dominate. Thus, their size is closer to the size leading to optimum strength (the intersection between $\sigma_{mod} + \sigma_{coh}$ and σ_{Orowan}) than the Al_3Zr $L1_2$ after the 425°C HT. The latter are still in the σ_{order} part of the shearing domain. The volume fraction

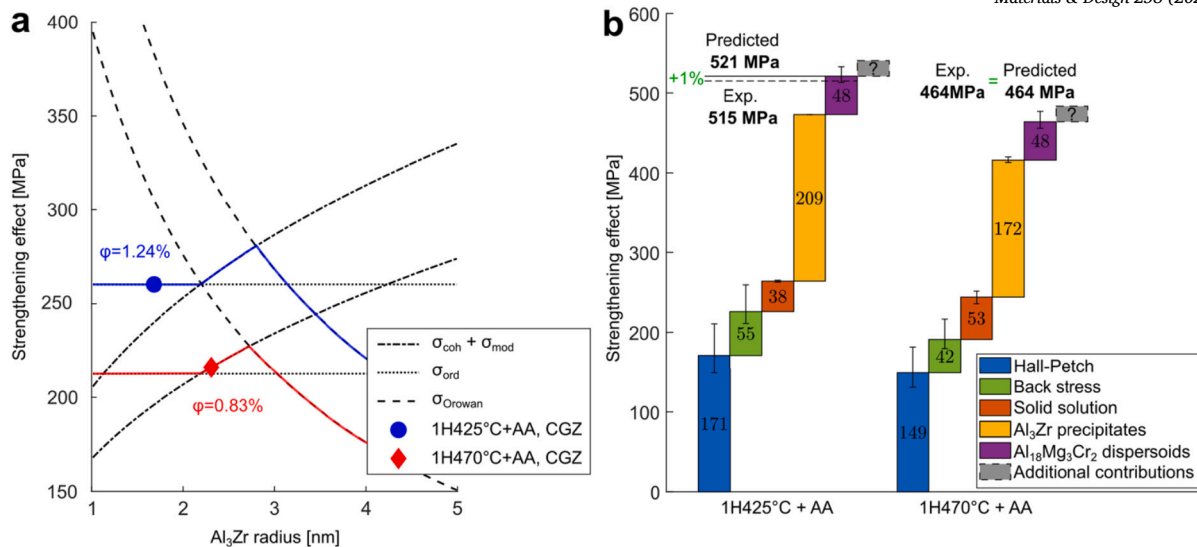


Fig. 10. (a) Strengthening effect of σ_{Al_3Zr} $L1_2$ secondary precipitation for the coarse grain zone (CGZ) for both 1h425°C + AA HT (1.24% volume fraction, in blue) and 1h470°C + AA HT (0.83% volume fraction, in red). It shows the hardening mechanism in action depending on the precipitate radius. The blue square and the red diamond represent the actual value obtained for the 1h425°C and 1h470°C condition, respectively. (b) Summary of the different strengthening contributions computed for both 1h425°C + AA HT and 1h470°C + AA HT.

difference was attributed to the global over-ageing generated by the 1h470°C HT. In addition to generation of larger secondary Al_3Zr $L1_2$ hardening precipitates, Zr segregation at GB (see Fig. 6(b)), and transformation of several primary Al_3Zr $L1_2$ grain refiner precipitates into Al_3Zr $D0_{23}$ (see Fig. 6 and Fig. A.5) were observed. Moreover, the 1h470°C HT generates larger $\sigma_{Al_{18}Mg_3Cr_2}$ than the 1h425°C HT, corroborating hypothesis of global over-ageing. This value of σ_{Al_3Zr} for the 1h425°C + AA condition was higher than that reported in the literature [10,20,24,54]. Existing studies have reported a lower volume fraction of $Al_3(Zr, (Sc), (Er))$, varying between 0.7 and 0.8%. However, Bayoumy et al. [43] obtained higher values of σ_{Al_3Zr} between 258 and 314 MPa with volume fraction between 1.16 and 1.68% (respectively in the FGZ or the CGZ) for an L-PBF Al-Mn-Mg-Sc-Zr alloy.

Fig. 10(b) summarises the effect of each strengthening mechanism on the overall YS. The superior Al_3Zr $L1_2$ precipitates strengthening and grain boundary strengthening generated by the 1h425°C + AA HT compared to that by the 1h470°C + AA HT was observed. Moreover, the limited grain growth improved σ_{GB} and σ_{BS} for the 1h425°C condition. The remaining contributions played minor roles in the YS difference generated by the two HT. The experimental behaviour accurately described by the analytical models detailed in this section with only 1% over estimation for the 425°C condition and the same computed yield strength for the 470°C + AA condition. This finding underscores the superiority of the 425°C HT, which is more efficient in forming a large volume fraction of hardening Al_3Zr $L1_2$ nanometric precipitates while avoiding excessive grain growth.

One other important strengthening contribution to detail is σ_{MgZn_2} . The strengthening effect of the conventional η' precipitates is the principal contribution to Al7075 high strength as widely described in literature [40,90,94,111,112]. Despite this fact, no significant amount of these Zn-rich hardening precipitates were identified in L-PBF Al7075+1.8%Zr alloy, thus their contribution to the yield strength was neglected. Regardless of being unusual, this lack of precipitates was explained by several observations. First, less Zn and Mg were available for $MgZn_2$ formation since 27% and 37% of the initial amount of Mg and Zn present in the powder were evaporated during L-PBF, as shown by ICP (Table 1). Second, as already mentioned in subsection 4.1.3, Mg already formed several Mg oxides and $Al_{18}Mg_3Cr_2$ dispersoids (Fig. 5, Fig. 6). Thus, these species prevented Mg from forming a significant amount of homogeneous η' precipitation. To summarise, the loss of Zn and Mg by evaporation during the L-PBF combined with the presence

of $Al_{18}Mg_3Cr_2$ and Mg oxides explained the scarcity of η' precipitates for both HT conditions. Moreover, despite σ_{MgZn_2} being absent for the computation the yield strength, Fig. 10(b) highlights the high coherence between the computed yield strength value and the experimental one.

Forest dislocation strengthening is also worth mentioning in this section. Indeed, if dislocations are initially present in the microstructure, they will hinder the movement of additional dislocations and thus strengthen the material [33,60,90,94,113–115]. However, in this study, this contribution was not analysed due to its relatively low importance, Al7075+1.8% Zr alloy being studied after HT. Consequently, the majority of dislocations that could have been generated by thermal residual stresses following the L-PBF process were restored by the HT. No additional strain was applied on the sample after HT. Therefore, the dislocations density initially present in the undeformed state of the sample can be neglected.

Despite being of low values, these two contributions could have a slight impact on the yield strength. The total value of the yield strength computed using the various analytical models could thus be slightly higher than provided in Fig. 10. This uncertainty is represented on the same figure by a grey box.

4.2. Tensile behaviour understanding and comparison with literature

The tensile behaviour shown in Fig. 1(b) is discussed further. The presence of serrations in the tensile curves can be identified as Portevin-Le Chatelier (PLC) effect, also known as dynamic strain ageing. It arises from the interaction between mobile dislocations and solute atoms [116]. In this study, these serrations are more pronounced in the AB state and after 1h470°C followed by natural ageing (Fig. B.1). Due to the high cooling rate during the L-PBF process, a large amount of Mg, Zn, Cu and even Zr are trapped in solid solution. Since PLC effect is driven by the diffusion of solute atoms, Zr atoms in solid solution should not affect the PLC effect. Zr diffusion coefficient is several orders of magnitude smaller than that of Mg, Zn and Cu (approximately $10^{-13} m^2/s$ for Mg, Zn and Cu compared to $3.2 \cdot 10^{-18} m^2/s$ for Zr at 500°C [66]). However, the high density of dislocations generated by the high cooling rate could facilitate Zr diffusion and allows Zr to play a minor role in the generation of PLC effect.

A yield plateau is observed for all HT conditions. This yield plateau can be called the Lüders plateau and usually results from the propagation of deformation bands just after yielding. They are normally caused

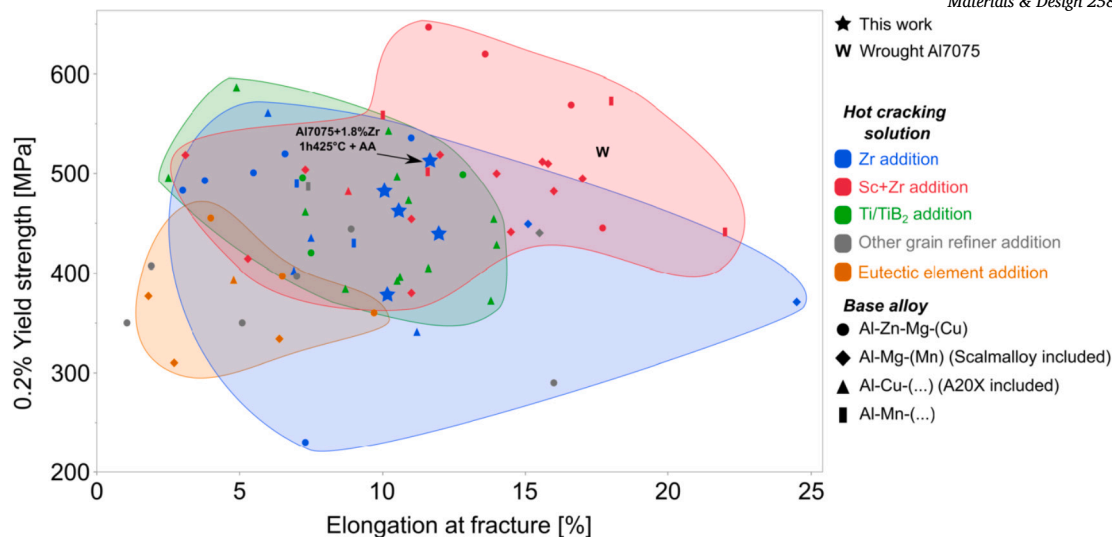


Fig. 11. 0.2% yield strength versus elongation at fracture for high-strength aluminium alloys processed by L-PBF. The star-shaped points correspond to the tensile properties listed in Fig. 1(b) while the 'W'-shaped point corresponds to a T6-state wrought Al7075. All the other data points are extracted from works between 2020 and 2024. Each colour corresponds to the type of hot cracking mitigation used: blue for Zr addition [9,11,13,15,16,21,22,24], red for Sc+Zr addition [38–40,42,43,45–50,52], green for Ti or TiB_2 addition [25–30,33–35,124], orange for the use of a high amount of eutectic phase (with Si addition for example) [19,32,58–61,125] and grey for other grain refiners [23,54,55,67,126,127]. The shape of the data point corresponds to the base alloy processed: disks for Al-Zn-Mg-(Cu) alloys (mainly 7xxx series), diamonds for Al-Mg-(Mn) (mainly Scalmalloy), triangles for Al-Cu-(...) alloys (mainly 2xxx series) and vertical rectangles for Al-Mn-(...) alloys.

by the Cottrell atmosphere of solute atoms hindering the initial movement of dislocations. Such a constant flow stress plateau is common for Al-Mg [117–119] alloys and for some steels [120–122]. In addition, similar yield plateaus were observed for high-strength Al alloys in various studies [11,13,40,43].

In this study, the Lüders strain changes depending on the HT (even when identification is difficult due to the low strain hardening): $\epsilon_{Lüders}$ for 1h470°C + NA $\approx 2.5\text{--}3\% < \epsilon_{Lüders}$ for 1h470°C + AA $\approx 3.5\% < \epsilon_{Lüders}$ for 1h425°C + NA $\approx 4\text{--}5\% < \epsilon_{Lüders}$ for 1h425°C + AA $\approx 4.5\text{--}5\%$. No large stress plateaus can be observed for the AB state, although the condition contains the highest amount of atoms in solid solution (due to the fast cooling rate of the L-PBF process) as evidenced by the presence of PLC effect. Therefore, the origin of this stress plateau is not linked to interactions between solute atoms and dislocations. Such a hypothesis has been formulated by Buttard et al. [11], who observed a similar stress plateau for Al-4Mn-3Ni-2Cu-1Zr L-PBF alloy in the heat treated state and not in the AB state. Additionally, they performed macroscopic digital image correlation (DIC) tensile tests and did not observe any deformation bands propagation, usually typical of the Lüders plateau. Bayoumy et al. [43] demonstrated that the yield plateau could be generated by the fine-grain zone, which they tested independently of the coarse grain zone of an Al-Mn-Sn L-PBF alloy. The theory was supported by the micro-DIC results obtained by Buttard et al. [11]. A detailed investigation of this phenomenon can be explored in the future. However, the subject lies beyond the scope of the present study.

Finally, the strain hardening is analysed. A large difference is observed between the 425°C and the 470°C samples. The globally low strain hardening observed for heat treated conditions can be attributed to the size of the precipitates. The Al_3Zr $L1_2$ secondary precipitates radius is in the range of shearing interactions with dislocations (Fig. 10(a)). Shearing of precipitates is detrimental for the strain hardening capacity of the alloy. When the radius of a precipitate is large, Orowan looping of dislocations around the precipitate becomes dominant. This mechanism leads to an improved dislocation storage capacity, thus improving strain hardening [123].

The origin of the tensile properties for the L-PBF Al7075+1.8%Zr has been discussed and the results obtained in this study can be compared to those in existing literature. Fig. 11 shows YS versus elongation at fracture for several existing studies. The mechanical performances of

high-strength L-PBF alloys are compared with respect to their hot cracking mitigation strategy and to the base alloy selected. As discussed in section 1, two main strategies are used to avoid hot cracking: the use of grain refiners or the use of a high amount of eutectic phase. Notably, the eutectic method leads to lower mechanical performances (data marked as orange in Fig. 11). To improve the Al7075+1.8%Zr alloy analysed in this study, alloys exhibiting higher mechanical performances in the literature should be analysed.

By considering first Zr-doped alloys (data points in blue in Fig. 11), only one study obtained a higher YS with similar elongation at fracture than the Al7075+1.8%Zr of this study. Martin et al. [22] analysed L-PBF Al7075 containing between 1.62 and 1.83% of Zr, in the same composition range as the present study. For their best case, they used a cold, mechanically derived powder functionalised with Zr by electrostatic assembly. They obtained a YS of 543 MPa and an elongation at fracture of 12% after hot isostatic pressing (HIP) (2 h at 400°C) followed by a conventional T6 HT (1 h 480°C + AA). Despite showing similar mechanical performance to the alloy of this study, they used HIP, which is an additional post-processing step that was not used in this study. Assuming that their powder functionalisation allowed avoidance of Zr unmelted particles (Fig. 2), some Zr was freed to either refine more grains or serve as hardening precipitates after HT. Moreover, they managed to obtain less Zn and Mg evaporation than in the present work (using cold mechanically derived powder instead of gas atomised ones), making Zn and Mg more available to serve as hardening precipitates. Therefore, they obtained similar properties as in the present work without tailoring their HT. In the future, improvement in the Zr addition method for avoiding trapping Zr trapped unmelted particles should be explored. An alternative processing path can be to involve using pre-alloyed powder.

Second, the high-strength Al alloys with a combination of Zr and Sc as grain refiners exhibit high strength combined with high ductility, as marked in red in Fig. 11. The origin of these superior mechanical properties should be analysed to determine the requirement of Sc addition to obtain higher mechanical properties. Zhu et al. [40] used L-PBF Al7075, initially doped with Zn (6.78 wt.% of Zn in the initial powder), 0.45 wt.% of Sc, and 0.25 wt.% of Zr. This Zn addition allowed the final alloy to contain 5.4 wt.% of Zn. After a T6 HT with a high temperature HT of 465°C for 20 minutes, YS of 647 MPa and elongation at fracture of 11.6% were obtained. This high value of YS was mainly ob-

tained due to the dual-nanoprecipitation of both secondary $Al_3(Sc, Zr) L1_2$ and η' leading to a strengthening effect of 412 MPa. Moreover, Sc prevented over-ageing of the $Al_3(Sc, Zr)$ precipitates, as $Al_3Zr D0_{23}$ precipitates or (Sc, Zr) segregation at GB were evidenced after HT, unlike as observed in this study (Fig. 6). Therefore, Sc is not essential for fabricating Al7075+1.8%Zr in this study, as over-ageing is suppressed using a tailored HT at 425°C. To attain even higher values of YS, a compensation of Zn and Mg loss can combine η' hardening precipitates with the obtained $Al_3Zr L1_2$.

5. Conclusions

In this study, a tailored HT composed of a first step high temperature HT for one hour at 425°C followed by a classical AA of 120°C for 24 h was carefully analysed. This tailored HT was compared to a conventional T6 HT, with a first step HT performed at 470°C followed by the same AA. The specific effect of the low temperature HT is unveiled.

1. The 425°C HT allowed the Al7075+1.8%Zr L-PBF alloy to attain a YS of 515 MPa while maintaining an effective elongation at fracture of 12%. The HT outperforms the 464 MPa YS and 10.5% elongation at fracture generated by the 470°C HT. This low temperature tailored HT prevents excessive grain growth, maintaining the surface weighted grain size at 1.6 μm compared to 2.2 μm for the high temperature HT.
2. The difference in YS between the two conditions was quantified by exploiting the TEM and EBSD analyses combined with analytical models to decorelate the effect of the various microstructural features experimentally observed:
 - The main contribution to YS comes from the hardening precipitation of secondary $Al_3Zr L1_2$ precipitates formed during the first step high temperature heat treatment. The choice of a relatively low 425°C HT prevents Zr over-ageing and optimises the hardening precipitation, generating a 209 MPa contribution compared to only 172 MPa for the 1h470°C condition.
 - Grain boundary strengthening is the second largest contribution to YS. The 425°C HT prevents excessive grain growth generating higher value of σ_{GB} . Moreover, back stresses generated at GB are identified as an additional source of strengthening, being slightly higher with the smaller grain size of the 425°C condition.
 - The contribution of η' $MgZn_2$ hardening precipitates is neglected, given their very low amount within the microstructure. This lack of homogeneous precipitation is explained by Zn and Mg evaporation during the L-PBF process and by a competitive use of Mg to form Mg oxides and $Al_{18}Mg_3Cr_2$ dispersoids.
3. Based on an extensive literature comparison, the addition of Sc is not essential in obtaining interesting mechanical properties for L-PBF Al7075 alloy. By controlling Zr precipitation with a tailored HT, the potential of Zr to form hardening precipitates rivals with other high-strength L-PBF alloys.

Some prospects to continue improving the mechanical performances of Al7075+1.8%Zr would be to compensate the loss of Zn and Mg. It would allow to form a significant amount of η' precipitates and to further improve the YS. Improving the addition of Zr to the Al7075, e.g. by Zr plasma implementation or pre-alloying would avoid the coarse Zr particles in the microstructure. These coarse Zr particles act as damage sources and prevent the Zr to form either grain refining or hardening precipitates.

CRediT authorship contribution statement

Nicolas Nothomb: Writing – original draft, Visualization, Validation, Project administration, Methodology, Investigation, Funding acquisition, Conceptualization. **Ignacio Rodriguez-Barber:** Visualization,

Investigation. **Marie-Noëlle Avettand-Fénoël:** Writing – review & editing, Visualization, Validation, Resources, Methodology, Investigation, Conceptualization. **María Teresa Pérez Prado:** Writing – review & editing, Resources, Methodology, Conceptualization. **Maya Marinova:** Investigation. **Aude Simar:** Writing – review & editing, Supervision, Resources, Project administration, Methodology, Funding acquisition, Conceptualization.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Appendix A. Supplementary material

Supplementary material related to this article can be found online at <https://doi.org/10.1016/j.matdes.2025.114553>.

Data availability

Data will be made available on request.

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