Selective laser melted Al-7Si-0.6Mg alloy with in-situ precipitation via platform heating for residual strain removal

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HIGHLIGHTS
- Heating build platform during selective laser melting avoids further heat treatment.
- The new method results in better mechanical properties than its counterpart.
- Residual strain after selective laser melting results in super solid solubility of Si.
- Precipitation of Si due to heating build platform relieves residual strain.

GRAPHICAL ABSTRACT

ABSTRACT

A series of complex post-build heat treatment procedures are usually inevitable for selective laser melted (SLMed) Al-Si-Mg alloys, aiming to remove significant process-induced residual strains. In this work, we propose a new method to avoid the post-build heat treatment without compromising the mechanical properties in SLMed Al alloy A357. By heating up the build platform (up to 90 °C) during SLM fabrication or applying post heat treatment after SLM, the evolution of microstructure and mechanical properties for Al-7Si-0.6Mg alloy are compared at different conditions. Characterisations using scanning electron microscopy (SEM), X-ray Diffraction (XRD), transmission electron microscopy (TEM), and hardness tests suggest heating up the build platform is more effective in alleviating residual strain and resulting in uniform precipitates in the Al matrix. Tensile properties imply that SLMed A357 fabricated with a heated substrate possesses improved mechanical properties compared to SLMed A357 build on room temperature platforms and even with post-build heat treatment.

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1. Introduction

Al alloys are mainly used for the production of parts requiring high thermal conductivity and reasonably good mechanical properties in the combination with lower costs compared to titanium and nickel alloys in a selective laser melting (SLM) manufacturing field [1–5].
SLMed Al—Si materials possess an ultrafine microstructure generated by a fast cooling rate \([1,6]\), and the effects of various silicon content on densification, mechanical and thermal properties of selective laser melted Al—Si binary alloys have been investigated in details before \([7]\). In addition, a desirable microstructure can be obtained within a much lower temperature and shorter solution heat treatment time than what ASM T6 recommends \([4,8]\), \(<2\) h at around 540 °C for instance \([9,10]\). On the other hand, aluminium alloy A357 with silicon and magnesium as the major alloying elements is regarded as one of the generic class of heat treatable alloys by precipitation hardening \([11,12]\). Following various heat treatment, Al-Si-Mg alloys typically undergo an artificial ageing treatment to obtain increased strength by facilitating the precipitation of various metastable hardening phases, such as silicon particles with random crystalllographic orientations, \(\beta^\prime\), and \(B^\prime\) \([12]\). Therefore, in the real engineering world during the last few years, a series of post processing procedures suitable for selective laser melted Al-Si-Mg alloys have been developed \([9,13]\). These procedures are mainly design in order to relieve the high process-induced residual strains, so that large aerospace engine components are easily removed from SLM build platform without forming internal cracking leading to specimen failure, and at the same time to achieve the maximum tensile properties.

Currently, treatment methods have to be applied on selective laser melted Al samples after laser fabrication and involve a series of complex heat treatment steps including stress relief, sample removal, solution treatment, and artificial ageing \([9,13]\). Although the mechanical property issues have been solved, the whole post treatment can be quite insufficient and less cost effective. In this regard, some previous work has systematically demonstrated the effects of substrate preheating during SLM of aluminium components on the component distortion \([14]\). Via the measurement of the twin cantilever spreading, an appropriate preheating temperature of 250 °C was determined to avoid sample distortion and cracking, independent of the twin cantilever test geometry. However, although process reliability in this situation can be ensured, tensile properties of SLMed Al-Si-Mg alloy sample manufactured at 200 °C are largely reduced \([1]\).

For SLMed materials, there is a typical difference in the microstructure and resulting properties at two orientations, known as anisotropy \([15–18]\). In general, samples with a layer orientation parallel to the tensile direction (horizontal test samples) show higher breaking elongation than those built with a layer orientation perpendicular to the applied load (vertical test samples) \((T)\) \([1,2,17,19–23]\). In comparison, the tensile strengths in two directions are almost the same. Moreover, the anisotropy on ductility and almost isotropy on tensile strengths cannot be remarkably changed by various post-build heat treatment methods \([1,2,17,19–23]\). Although there are quite obvious anisotropic properties for SLMed Al alloys, overall good tensile properties at different fabrication conditions can still be obtained \((T)\) \([4,21,24–26]\), which are better than conventional cast alloys \([2,20,21,24,25,27–29]\).

There is a considerable amount of previous work about various Al—Si alloys manufactured by SLM and processed by heat treatment \([1,2,17,19–23,26]\), and about the relationship of platform temperatures the resultant microstructure. Si particle morphology at various conditions in particular, as well as fractography and mechanical property evolution \([18,34–36]\). However, the specific influence of those procedures on the residual stresses and precipitation sequences of SLMed Al-Si-Mg alloys have never been studied before. Also, the influence of various processing conditions on the ductility anisotropy of SLMed A357 still requires detailed explanation. Therefore, the aim of this work is to recommend a novel and simple method, omitting the complex post-build heat treatment procedures currently applied on SLMed Al alloy A357, relieving the high internal stresses within as-built components while maintaining their superior mechanical properties obtained after SLM.

### 2. Experimental methods

The material used in this study was gas atomised Al alloy A357 powder, and the main chemical compositions of the material used was determined in Table 2. The particular SLM process in this project was conducted by a Concept X-Line 1000R powder-bed machine using the optimised processing parameters: 400 W (laser power), 1100 mm/s (laser scan speed), 0.04 mm (layer thickness), and 0.15 mm (hatch distance). Various platform temperatures of 35 °C (Condition A and B) and around 80 °C (Condition C) were also applied during SLM. Some samples were also stored for 5 years at room temperature (Condition B) or heat treated at the currently optimised parameter (Condition D, which was also built at 35 °C): stress relief (SR) at 300 °C for 2 h with air cooling, solution treatment (SHT) at 543 °C for 1.5 h with polymer quench, and artificial ageing (AA) at 160 °C for 8 h with air cooling, to meet up the mechanical property standards \([9,37]\). These four conditions are summarised in Table 3. The scan strategy for SLM and the measurement of tensile properties for SLMed samples were illustrated in an early study \([1]\). The tensile tests were conducted using samples built with the tensile axis parallel to the build direction (BD) or z plane, defined as vertical test samples, and those perpendicular to the BD or XY plane, defined as horizontal test samples (Fig. 1).

Direct ageing (DA) (without prior solution treatment) was conducted in an oil bath for as- SLMed samples from Condition A to C at 165 °C. The ageing time applied was 2, 4, 8, 16, and 30 min, and 1, 2, 4, 8, and 16 h, followed by TRQ water quench again. After heat treatment, Vickers hardness was measured using a macro hardness tester with a load of 1 kg. Seven indentations were performed at each heat treatment condition, with one maximum and one minimum hardness value removed afterwards. The mean values of the rest 5 measurements per sample were taken to give an average data point and the corresponding error bars.

After being sliced, ground, and polished along both horizontal and vertical direction, 1% hydrofluoric (HF) acid was then used as the etchant for about 15 s to reveal the microstructures. A JEOL 7001F SEM operating at 15 keV and Aztec and HKL channel 5 software were used for microanalysis. X-rays diffraction (XRD) patterns of the SLMed samples at all conditions were recorded at room temperature, using a Philips PW1729 X-ray generator with Cu Kα (λ = 1.5418 Å) radiation, and a scan rate of \(1^\circ/min\) was applied with the 20 angle varying from 20° to 80°. According to Bragg’s law, the interplanar distance \(d = \frac{n\lambda}{2\sin\theta}\) can be obtained. Then the real lattice spacing of Al FCC system can be calculated through \(a = d/\sqrt{h^2 + k^2 + l^2}\), where \(h, k, l\) and \(a\) are the Miller indices of a lattice plane. Here we assume that the unit cell of Al is still cubic even with crystallographic-dependent strains, which is reasonable for the qualitative estimation of residual crystallographic-dependent strains. According to the change of real lattice parameters at each plane, the influence of the fabrication method (SLM) on the formation of internal strains within the Al samples can be quantified and evaluated. TEM samples were prepared by a precision ion polishing

<table>
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<td>The tensile data for the as- SLMAl—Si alloys manufactured in different directions, and the typical property ranges of cast Al—Si alloys ([2,17,19–21,24,25,28–32]).</td>
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<td>SLMed Al alloy tensile properties</td>
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<td>Tensile strength (MPa)</td>
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energy is expressed as $E_{\text{Strained Al}}$ and 1 Si atom was relaxed with supercell shape. The modified supercell shape were free in other directions. The stretched case was used to represent the Al matrix with 3% tensile strain. The formation energy of relaxed supercell with strain is expressed as $E_{\text{Strained Al}}$. The relaxed supercell with strain (96 Al atoms) was modified by replacing one Al atom by a Si atom, to calculate the formation energy of one Si solute atom in the strained Al matrix. The modified supercell (95 Al atoms and 1 Si atom) was relaxed with supercell shape fixed, and its formation energy is expressed as $E_{\text{Strained Al-Si}}$. The formation energy of Si solutes in the strained Al matrix is expressed as:

$$E_{\text{Si-formation}} = E_{\text{Strained Al}} - E_{\text{Strained Al-Si}}$$

Other calculations were carried out similarly with different strains along (100)$_{Al}$ or (111)$_{Al}$ directions.

3. Results
3.1. Solute-atom induced lattice distortion and crystallographic-dependent residual strain identification by XRD

The lattice parameters of Al matrix were calculated using the peak positions of different lattice planes on the XRD patterns, as shown in Fig. 2. Same to an annealed high purity Al (4 N) sample, the lattice parameters for Condition B sample calculated from different lattice planes (e.g., (111) and (200)) almost keep the same (near zero slopes), as illustrated by the almost horizontal lines in red and green. In this regard, there should be no process-induced crystallographic-dependent residual strains in Condition B samples. Also, no vertical position shift for Condition B sample is observed compared to the high purity Al, indicating no solute-atom induced lattice distortion (shrink/expansion) as well. Therefore, it is worth noticing that a 5-year storage at room temperature seems to remove both the lattice distortion due to solute atoms and crystallographic-dependent residual strains.

However, the lattice parameters of the as-build A357 with a platform temperature of 35 °C differs a lot when calculated with different lattice planes (both the largest slope and vertical position shift in Condition A sample), and this suggests remarkable process-induced crystallographic-dependent residual strains and solute atoms induced lattice distortion are both present [12,38]. For samples fabricated at

$$
\begin{array}{|c|c|c|c|c|c|c|}
\hline
\text{Condition} & \text{As-SLMed Al sample built at 35 °C} & \text{As-SLMed Al sample built at 80 °C (±10 °C)} \\
\hline
\text{A} & \text{As-SLMed Al sample built at 35 °C} & \text{As-SLMed Al sample built at 80 °C (±10 °C)} \\
\text{B} & \text{As-SLMed Al sample built at 35 °C and stored for 5 years at T\text{room}} & \\
\text{C} & \text{As-SLMed Al sample built at 80 °C (±10 °C)} & \\
\text{D} & \text{Post heat treated: SR + SHT + AA as discussed above [9]} & \\
\hline
\end{array}
$$

Condition C, the solute atoms induced lattice distortion is only slightly higher than that for Condition B samples at two build directions. The complex heat treatment in Condition D is initially designed to remove the crystallographic-dependent residual strains that form during the rapid solidification in SLM process [8,13]. However, there are still high process-induced crystallographic-dependent strains remained in Condition D than in Condition B. These comparisons suggest that the complex heat treatment is not the best choice for lowering crystallographic-dependent residual strains in SLMed Al-Si-Mg alloys though it is widely accepted [9].

The Si concentration in the Al matrix at different conditions can be estimated by comparing the lattice parameters to the high purity Al at the same lattice plane (Fig. 2). A smaller lattice parameter means a high Si concentration in the Al matrix because Si atoms have smaller radius than Al atoms and thus would shrink the Al lattice parameter. Since there are crystallographic-dependent residual strain interferences, the comparison should only be made between lattice parameters calculated from the same lattice planes.

$$
\begin{array}{|c|c|c|c|c|c|c|c|}
\hline
\text{Sample} & \text{Composition specification} & \text{Alloy element} & \text{Si} & \text{Mg} & \text{Fe} & \text{Ti} & \text{Zn} & \text{Cu} & \text{Bal.} \\
\hline
\text{SLMed bulk A357 sample} & 6.5-7.5 & 0.45-0.7 & 0.19 & 0.2 & 0.07 & 0.05 & Bal. \\
\text{Composition specification for standard A357 powder} & & & & & & & & & \\
\hline
\end{array}
$$

Fig. 1. Schematics of selective laser melted Al tensile samples built at two directions based on ASTM E8 standard.

Fig. 2. Al lattice parameters at various conditions, indicating solute-atom induced lattice distortion and crystallographic-dependent residual strains in SLMed Al samples. An inclined line suggests crystallographic-dependent residual strains in the Al samples, comparing to the near horizontal line in the high purity Al samples after annealing. These AI samples were characterised in two directions: X-rays scanning planes parallel with laser scanning planes (in solid lines) and X-rays scanning planes normal to laser scanning planes (in dashed lines).
Here the sample in Condition A has the smallest lattice parameter and therefore the highest Si concentration in the Al matrix. This has been verified by early EDS analysis which shows 6 wt% Si in the Al matrix [13]. Both 5-year storage at Troom and post-build heat treatment can reduce the Si concentration in the Al matrix. Due to a high supersaturated Si content in the as-built condition after high temperature laser melting and subsequent fast cooling [1,8,9,13], natural ageing gradually happens even without artificial ageing. Therefore, for the SLMed Al sample stored for 5 years (Condition B), the high internal strain is expected to be relieved by precipitation of Si-containing nano-particles, which will be further discussed in the following sessions. The reasons for both the low crystallographic-dependent residual strains and Si concentration in Condition C sample will also be discussed later.

3.2. The morphology of eutectic Si particles formed during solidification

Based on detailed analysis in previous studies, the directional microstructure and anisotropic tensile properties are quite obvious for both the as-SLMed and heat treated A357 samples [1,9,13]. A very fine microstructure generated by a fast cooling rate in the SLM process can be divided into several different zones with varied sub-micron sized Si existed inside, as illustrated in Condition A in Fig. 3, also observed by others [1,2,39]. For samples built at Condition B with a same substrate temperature but stored for 5 years, the Si particles remain unchanged as expected as no heat flow applied, with the fine, coarse, and heat affected zones all clearly distinguished (Fig. 4).

The microstructure of SLMed A357 is remarkably changed after complex post-build heat treatment (condition D) as shown in Fig. 6. Due to the microstructure coarsening after solution heat treatment in Condition D, the Si particles increase by roughly 50 times in size [8,13] compared to Condition A, B and C (Fig. 5). The 3-zone features that form during solidification completely disappears after the complex heat treatment for both horizontal and vertical directions.

3.3. Tensile properties

Tensile curves of A357 samples at various conditions are shown in Figs. 7 and 8 together with the changes in tensile anisotropy (Fig. 8a) for comparison, where the anisotropy is regarded as none when it is equal to 1. As the substrate temperature increases from 35 °C (Condition A and B) to 80 °C (Condition C), there is a slight growth in the ductility (Fig. 8b), and the tensile stresses are quite similar (Fig. 8c and d). For Condition B sample, in particular, the ductility obviously decreases, possibly due to the natural ageing effect which leads to nano-precipitate strengthening Al matrix [13]. Moreover, the ductility of Condition C sample is as high as that of Condition D sample. Although a higher substrate temperature slightly lowers the yield strength, the corresponding UTS in Condition C is much higher (~90 MPa) than those samples processed by the complex heat treatment in Condition D (Fig. 8).

The anisotropy in the tensile stresses is the lowest for Condition C sample. It also decreases by applying post heat treatment on the SLMed samples. Furthermore, the anisotropy in ductility witnesses some variation at different conditions, but all remain relatively high within the 1.5–2 region (Fig. 8a). Still, post heat treatments do not lead to the lowest anisotropy. Therefore, interesting questions arise again: is the current complex post heat treatment truly “optimised” to achieve the best properties? If not, is it still worth conducting? The samples in Condition C and D seem to have a homogenous microstructure compared to condition A and B. However, the anisotropy regarding ductility is not significant reduced. Therefore, microstructural homogenisation seems to be insufficient to remove the anisotropy.

3.4. Tensile fractography

In general, the fractography of the SLMed Al alloy samples agrees well with the tensile properties, and the changes in fracture surfaces (Fig. 9) correspond to the ductility difference (Fig. 8b). For instance,
there are more equiaxed dimples formed in samples in the horizontal samples (Fig. 9a, c, e, g) than in the vertical samples (Fig. 9b, d, f, h). In particular, quite obvious cleavage planes can be observed in the vertical samples. Therefore, the lower ductility along the vertical direction than the horizontal direction for all the samples is expected to be correlated with these brittle features, instead of the microstructural features of Si particles as shown above.

The complex post-build heat treatment remarkably changes the morphology of sample porosity under Condition D, where the porosity dimension is much larger than other samples in Fig. 9(a–f), indicating that gas porosity expansion tends to occur during high temperature solution treatment [9,37]. Also, the quantity of porosities are much higher on the vertical fracture surface (Fig. 9g) than the horizontal one (Fig. 9h) for heat treated samples. Furthermore, although Condition C sample shows a slightly higher ductility than Condition A and B samples, not much difference in terms of defect features can be identified on the fracture surfaces.

3.5. Hardness tests

The direct ageing responses for samples built at Condition A-C are exhibited in Fig. 10. For Condition A and B samples, a maximum peak hardness of up to 150 HV is achieved after direct ageing for 2 h at 165 ℃. These two curves also witness a steady increase and then an obvious drop. For Condition C sample, however, the hardness is quite steady during the whole ageing period. Therefore, Condition A and B samples are likely to have remarkable precipitation hardening up to 2 h, while Condition C sample has no recognisable precipitation hardening effects. Also, Condition B sample possesses a higher hardness during the whole ageing process compared to Condition A sample, which may suggest that natural ageing has strengthened the sample to some extent. These phenomena will be supported by TEM results as followed.

3.6. TEM analysis

For a better understanding of the ageing response of the samples built at different conditions, TEM images are investigated (Fig. 11). The Al matrix in Condition A sample is clear of any nanoparticles. Si particles are only observed along low-angle grain boundaries (Fig. 11a) [12]. In contrast, the Al matrix in the sample built under Condition B is decorated by nano-particles (Fig. 11b) [12]. Apart from the nano-precipitates, there are also big particles which are definitely identified as Si phases based on the diffraction rings in the SAED pattern in Condition C (Fig. 11c) [12].

3.7. Si solute formation energy

Fig. 12 displays the effects of strains on the formation energy of one Si solute atom in the Al matrix, calculated from DFT. Strains are applied along either \( \langle 111 \rangle _{\alpha} \) or \( \langle 100 \rangle _{\alpha} \) direction along which a strong compression strain is observed in the SLMed A357 under Condition A (see Fig. 2). It is worth noting that the formation energy of one Si solute is lowered with decreased tensile strains and increased compressive strains in the Al matrix. This means that the Si solutes in the Al matrix becomes more energetically favourable as the sample is compressively strained along \( \langle 111 \rangle _{\alpha} \) or \( \langle 100 \rangle _{\alpha} \).

4. Discussion

4.1. Super solid solubility of Si in the Al matrix

As-SLMed A357 samples generally exhibit a strong process-induced residual strain and a very high solid solubility of Si in the Al matrix due to the rapid solidification. As shown in Fig. 2, a stronger crystallographic-dependent residual strain is always coupled with a
higher solid solubility of Si. Since the Si solid solubility can be up to 6 wt % in sample under Condition A [13], both direct ageing and solid solution treatment can decrease the Si concentration in the Al matrix (to a level lower than 1.6 wt%, the equilibrium Si solubility). As suggested in our recent paper, both direct ageing at 165 °C and solid solution treatment at roughly 540 °C can remarkably reduce the crystallographic-dependent residual strains in SLMed A357 [12]. Therefore, these results seem to suggest that the crystallographic-dependent residual strains increase the solid solubility of Si.

In high purity Al and Condition B Al—Si sample, the four lattice parameters from four lattice planes are almost the same, suggesting no crystallographic-dependent residual strains. If we compare these two samples with other samples, we can find that the lattice parameter lines for other samples change in two aspects. On the one hand, the vertical positions for Condition A, C and D samples are lower, indicating the lattice parameter shrinkage induced by Si solute atoms. On the other hand, if there is only Si solute induced lattice shrinkage, the lattice parameters calculated from different lattice planes in a sample (A, C, and D) should be almost the same, as for high purity Al and condition B sample. However, here we observe clear lattice parameter deviation from a horizontal line for samples A, C and D. This indicates the FCC lattice of Al has been distorted to different extent along different crystallographic directions.

The Si solute formation energy trend (Fig. 12) is consistent with our above XRD experimental observation (Fig. 2): a lower formation energy corresponding to a higher Si solid solubility. In this regard, the supersaturated solid solubility of Si in SLMed A357 Al matrix is caused by the strong crystallographic-dependent residual strains in the Al matrix. It should be noted that the super solubility of Si has been attributed to the rapid solidification during SLM. In fact, however, the rapid solidification itself cannot induce super solubility of Si in the Al matrix. This can be supported by previous work on Al—Si alloys prepared by melt

![Fig. 5. SEM micrographs showing the Si morphology for Condition C sample in the (a) (c) horizontal direction and (b) (d) vertical direction. Three different zones can be distinguished at lower magnifications: 1. fine region, 2. coarse region, 3. heat affected zone (HAZ) only in image (b), the fine regions are also illustrated at higher magnifications in image (c) and (d).](image)

![Fig. 6. SEM micrographs showing the coarsening of Si particles for Condition D sample in the (a) horizontal direction and (b) vertical direction.](image)
spinning at a cooling rate much higher than that in SLM [40], where the Si solubility in the Al matrix is still close to 1.6 wt% [1,13].

4.2. Precipitation and solute-atom induced lattice distortion and residual strain relief

Since crystallographic-dependent residual strain is always coupled with Si super solid solubility (>1.6 wt%) in the Al matrix, the crystallographic-dependent residual strains can be removed by the formation of Si-enriched precipitates [12]. The most common way to form Si-enriched precipitates in SLMed A357 is by applying post-build heat treatment. As shown in Fig. 2, this method can reduce the residual strains formed during SLM. However, this method seems not to be the most cost-effective and the residual strain removal is not thorough. Similarly, direct ageing (DA) at 165 °C for 2 h after SLM can also effectively remove the residual strains, which has been discussed in our recent report [12]. But DA highly reduces the SLMed alloy’s ductility to a very low level [13]. 5-year storage at room temperature can result in natural ageing, which brings in nano-precipitates in the Al matrix as shown in Fig. 11b. However, this method is not realistic for industrial applications. Furthermore, since the Si content in the 5-year storage sample is comparatively lower and quite close to the equilibrium state (Fig. 2), the precipitates observed in Fig. 11b are believed to be Si-rich precipitates.

Furthermore, for Condition C sample with very low Si concentration and residual strains (Fig. 2), abundant Si particles with a diamond structure and random crystallographic orientations are readily formed after SLM (Fig. 11c). In the A357 sample build at Condition C, the residual strains in the Al matrix are among the lowest (Fig. 2). This means that heating platform up slightly is the best way to remove residual strains by generating precipitates compared to complex post-build heat treatment or long time room temperature storage.

4.3. In-situ precipitation during SLM: a simple answer to the questions

The A357 sample built at Condition C (with in-situ precipitation) shows an improved ductility, which is expected to be contributed by a homogenous microstructure of Si particles with a small size (~50 nm in Fig. 11). However, the anisotropy on ductility is still consistently high in Condition C sample (Fig. 8a). The authors believe that anisotropy is attributed to more serious porosities along the vertical fracture than the horizontal fracture. In fact, the anisotropy on ductility is always present in the investigated conditions (Fig. 8a) and in the literature (Table 1). The ductility of an individual sample depends on whether the failure is controlled by the exhaustion of work hardening and the consequent inability of the material to accommodate further deformation or whether defects circumscribe work hardening [24]. However, to the best knowledge of the authors, there have been no effective ways to remove the anisotropy of ductility in SLMed A357 so far. Therefore, further studies about how to remove porosities or alleviate the detrimental effects of porosities are still required.

The A357 sample under Condition C also shows a higher UTS and slight lower yield strength compared to samples under other 3 conditions. The higher ultimate tensile strength is expected to be correlated.
to the more uniformly distributed eutectic Si particles along Al grain boundaries and the high density of Si precipitates in the Al matrix (see Figs. 5 and 10) [1,10,13]. However, the morphology changes of eutectic Si from almost fibrous along Al grain boundaries to isolated particles are also expected to be responsible for the lowered tensile strength. This is because harder particles in a continuous form is well known for strengthening materials [41,42]. Although the yield strength in Condition C is slightly lower (~30 MPa) than all other conditions in both horizontal and vertical directions, it is still higher than the cast counterparts (Table 1).

Fig. 9. SEM micrographs showing fracture surfaces of tensile samples built at (a) (b) Condition A, (c) (d) Condition B, (e) (f) Condition C, (g) (h) Condition D at two different directions, where (a, c, e, g) are horizontal tensile samples and (b, d, f, h) are vertical tensile sample.
The hardness curves shown in Fig. 10 suggest the A357 under Condition C cannot be further ageing hardened. Moreover, the hardness in this condition is always much lower than other 2 conditions, even though the corresponding tensile strengths for each sample discussed above only show a small difference. This means hardness is not necessarily associated to strengths for SLMed Al-Si alloys, and the decreased hardness is due to the loss of solid solution strengthening effects [12,13]. However, since Al alloy A357 is not generally used for the purpose of wear resistance, the lower hardness here in Condition C sample is not a serious problem for industrial applications.

Comparison, we proposed that the unprecedented Si solubility in the Al matrix of SLMed Al\textsubscript{Si} alloy is due to high crystallographic-dependent residual strains. To validate this, we performed density functional theory (DFT) calculations. The DFT calculations suggest strains along different crystallographic directions of Al can lower the formation

![Fig. 10. Vickers hardness (HV) curves for directly aged SLMed Al A357 built at different conditions at 165 °C.](image)

![Fig. 12. DFT calculated formation energy of one Si solute atom in the strained Al matrix. The Al matrix is strained along (111)\textsubscript{Al} or (100)\textsubscript{Al} direction to different extent. A negative strain means compression, while a positive one represents tension.](image)

![Fig. 11. BF TEM images with OA inserted showing the microstructure of SLMed Al samples along the \langle 100 \rangle Al direction in Condition A and B, and along the \langle 110 \rangle Al direction in Condition C. The diffraction ring with 0.31 nm distance in the real space from the transmitted dot is responsible for the diffraction from \{111\} planes of Si particles.](image)
energy of Si solutes in the Al matrix. In others words, a crystallographic-dependent residual strain in the Al matrix could promote the solid solubility of Si. The authors believe that these analyses are solid and comprehensive. We have explained Si solutes can precipitate out during SLM by just heating up build platforms. The critical thing is Si concentration is already much higher than the equilibrium solid solubility of Si, which make the alloy ready for direct ageing without further solid solution treatment. Other heat treatment strategies for SLMed Al—Si alloy just follow the protocol for casting Al—Si alloy with a preceding stress relief stage. Here we in fact point out the heat treatment process for Al-Si-Mg alloys in terms of the total method is much better than other post processing conditions currently of mechanical properties. From the authors’ perspectives, this new ple and elegant way to produce a fabricated sample with very low resid-

sual properties at the same time for SLMed Al alloy A357. The residual strains have been determined based on the changes in Al lattice parameters. In this work, a novel and simple method by heating up the build platform has been proposed to replace a series of complex post-build heat treatment currently inevitable for selective laser melted Al-Si-Mg al-

loys. This method would still relieve the high internal strains within as-built components while maintaining the superior mechanical properties at the same time for SLMed Al alloy A357.

2. The ultrafine eutectic Si particles are originally uniformly distributed along Al grain boundaries in the as-SLMed Al357 samples. After build-
ing on a heated platform, the Si particle morphology changes from almost fibrous along grain boundaries to isolated particles, but still possession a nano-size and high number density in the Al matrix. After high temperature solution heat treatment, however, nano-
sized Si particles severely coarsen around the high angle Al grain boundaries.

3. SLMed Al357 fabricated with a heated substrate possesses improved mechanical properties and possibly decreased anisotropy in tensile stresses, comparable or even better than that post-build heat treated samples. However, the influence of various processing conditions on alleviating the ductility anisotropy of SLMed Al357 is quite limited.

4. DFT calculations suggest that crystallographic-dependent strains lower the formation energy of Si solutes in the Al matrix, implying that the high solid solubility of Si (up to 5–6 wt%) in the SLMed Al—Si alloy is due to the existence of residual crystallographic-dependent strains.

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