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In-situ EBSD investigation of orientation induced mechanical anisotropy in selective laser melted Al-Mn-Sc alloy

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ABSTRACT

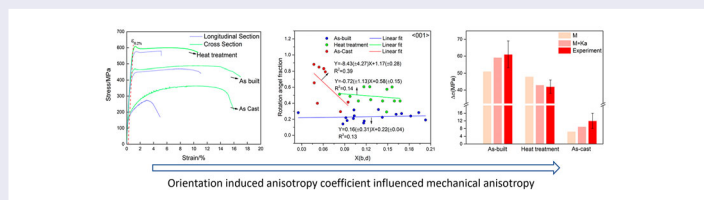
The mechanical anisotropy of selective laser melted (SLM) aluminum alloys resulting from heterogeneous grain structures is still not fully understood due to lacking study of grain orientation related mechanical response. Here, by introducing in-situ EBSD tensile method, the orientation induced anisotropy coefficient K_a related to the deformation-induced grain rotation was revealed by excluding effects of grain size and grain aspect ratio. Compared with traditional mechanical properties models by Taylor factor M , it was found that the orientation $\langle 001 \rangle$ in as-built and $\langle 102 \rangle$ in heat treatment contributed the most yield strength anisotropy, and introducing K_a predicts the mechanical anisotropy more accurately.

ARTICLE HISTORY

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KEYWORDS

Selective laser melting; mechanical anisotropy; Al-Mn-Sc alloys; in-situ EBSD; grain orientation



IMPACT STATEMENT

By introducing in-situ EBSD tensile method, we found the orientation induced anisotropy coefficient can reflect the mechanical anisotropy by excluding effects of grain size and grain aspect ratio.

1. Introduction

Selective Laser Melting (SLM) enables the direct forming of metal parts from computer-aided design (CAD) files using a high-energy laser beam to melt powder bed within a specified scan path [1,2]. Under the action of a large temperature gradient and inherent heat treatment process, the heterogeneity and anisotropy of microstructure and mechanical properties of SLM parts appear [3]. The anisotropy of alloys by additive manufacturing is an inherent characteristic of most metal alloys, such as steel, high entropy alloy, Ni-based alloy, and so on, because of rapid solidification [4–8], and the same is true for aluminum alloys [9].

The high-strength Al-Mn-Mg-Sc-Zr alloys are strengthened by solid solution strengthening owing to the supersaturated solution of Mn atoms and precipitation strengthening of nanophases after heat treatment,

whereas the anisotropy of microstructure and properties is inevitable [10,11]. It was reported that grain morphology and crystallographic texture affected the anisotropic behavior of additively manufactured metal alloy [4]. Different grain morphologies mean the distinction of the aspect ratio of grains and morphological heterogeneity affects anisotropic polycrystalline plasticity through size effects, which may affect the grain rotation during deformation. Accordingly, elucidating the correlation between grain size and grain aspect ratio and grain rotation will be an important direction in the study of anisotropy. The studies showed that additive manufacturing alloys present different Taylor factors M in different building directions owing to the difference in texture [6,12,13]. The Taylor factor is a measure of the tendency of a crystal to slip [14]. If the grain sizes and morphologies of different building directions are remarkably different,

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the building direction consisting of fine equiaxed grain with rich grain boundaries will be difficult to slip and rotate, which will also affect the Taylor factor. Therefore, the mechanical anisotropy can be determined by calculating the Taylor factor M for a given texture and grain shape, and the anisotropy trend can be predicted, however the strength anisotropy cannot be quantitatively calculated [15]. In this work, the in-situ EBSD tensile method was used to analyze the relationship between grain aspect ratio, grain size, grain orientation and mechanical anisotropy. By revealing the strengthening mechanism of the as-built, heat-treated, and as-cast Al-Mn-Sc alloy, the prediction model is optimized to accurately predict the mechanical anisotropy.

2. Experimental procedure

The powders with the chemical composition of Al-4.97 Mn-1.58 Mg-0.88 Sc-0.71 Zr-0.16 Fe-0.08 Si (wt. %) with the sizes of 20–53 μm were prepared by selective laser melting under a protective atmosphere of argon. The optimal processing parameter was a laser power of 370 W, a scanning speed of 1100 mm/s, a hatch distance of 0.12 mm, and layer thickness of 0.03 mm. A casting Al-Mn-Sc alloy with the same chemical composition as the SLMed Al-Mn-Sc alloy was melted at 780 °C and hot rolled from 10 mm to 0.5 mm at 420 °C. The SLMed Al-Mn-Sc alloy was directly heat-treated at 300 °C for 5 h, and then cooled in the air. The rolled Al-Mn-Sc alloy was treated with solid solution at 520 °C for 2 h and cooled by water, and then aged at 300 °C for 5 h.

In situ tensile specimens with nominal gauge dimensions of 7.2 mm (length) $\times$ 1.2 mm (width) $\times$ 1 mm (thickness) were placed on a commercial M-Test-1000 in-situ tensile testing module with 70 degrees tilt to perform in-situ EBSD tensile with a tensile speed of 0.1 mm/min. The EBSD was carried out with the accelerating voltage of 20 KV and the scanning step size of 0.2 μm . The inverse pole figure (IPF), grain size, and Euler angle (φ_1 , Φ , φ_2) were analyzed by TSL orientation imaging microscopy (OIM) analysis software. Through the rotation matrix $R_{0,i}$, the rotation angle (between step i and step j) defining the magnitude of lattice rotation is calculated by Euler Angle (φ_1 , Φ , φ_2) in equations 1–3 [16].

$$R_{0,j} = \begin{bmatrix} \cos\varphi_1\cos\varphi_2 - \sin\varphi_1\sin\varphi_2\cos\Phi & \sin\varphi_1\sin\Phi \\ \cos\varphi_2\sin\varphi_1 + \sin\varphi_2\cos\varphi_1\cos\Phi & \cos\varphi_2\sin\Phi \\ \sin\varphi_2\cos\varphi_1 - \cos\varphi_1\sin\varphi_2 & \sin\varphi_1\cos\Phi \\ -\sin\varphi_1\sin\varphi_2 + \cos\varphi_1\sin\varphi_2\cos\varphi_2\cos\Phi & \cos\varphi_2\cos\Phi \\ \cos\varphi_2\sin\Phi & \cos\Phi \end{bmatrix} \quad (1)$$

$$\Delta R_{ij} = R_{0,j}R_{0,i}^{-1} \quad (2)$$

$$\theta_{ij} = \left(\cos^{-1} \left(\frac{\Delta R_{ij11} + \Delta R_{ij22} + \Delta R_{ij33} - 1}{2} \right) \right) \quad (3)$$

The as-built, heat-treated and as-cast samples were electropolished with an electrolyte solution containing 30 Vol% nitric acid and 70 Vol% methanol at –25 °C and 16 V using MTP-1 twinjet. Bright field (BF) was conducted on an FEI Talos F200X at 200 kV, and high angle annular dark field (HAADF) was performed on an FEI Themis Z at 300 kV.

3. Results

3.1. Analysis of microstructure and property

The anisotropy of microstructure and mechanical properties derived from the epitaxial growth of the columnar grains is formed under the large temperature gradient and inherent heat treatment [17,18]. The as-built Al-Mn-Sc alloy presented noticeable microstructure and property anisotropy, as shown in Figure 1a, which was characterized by main columnar grains along the building direction (BD) and equiaxed crystals in the transverse direction (TD). The similar microstructure morphology was maintained for SLMed Al-Mn-Sc alloy after heat treatment at 300 °C for 5 h on account of high thermal stability, as shown in Figure 1b. The difference of microstructure in different directions leads to mechanical anisotropy. The yield strength difference between the building direction and transverse direction of the as-built Al-Mn-Sc alloy is about 60 MPa, and the strength difference between the two directions after heat treatment is about 42 MPa, as shown in Figure 1d. However, Figure 1c shows that the as-cast Al-Mn-Sc alloy present similar grain shape and size in two directions, and the corresponding strength difference of 13 MPa is also small, implying that the as-cast Al-Mn-Sc alloy exhibits mechanical isotropy. The texture has a certain influence on the anisotropy of additive manufacturing alloys [19]. Figure 1e shows the crystallographic texture of as-built Al-Mn-Sc alloy with respect to loading direction. The strong texture $\langle 001 \rangle$ is observed for as-built samples along BD and remains almost unchanged as the strain increases to 0.2%, while the samples along TD possess $\langle 102 \rangle$ and $\langle 112 \rangle$ textures, presenting transformation tendency from $\langle 102 \rangle$ to $\langle 112 \rangle$ when deformed to 0.2%. The heat-treated Al-Mn-Sc alloy has $\langle 113 \rangle$ texture along BD and $\langle 103 \rangle$ texture appears when deformed to 0.2%, while the samples along TD with $\langle 102 \rangle$ and $\langle 112 \rangle$ textures show a slight texture transition from $\langle 112 \rangle$ to $\langle 102 \rangle$ when deformed to 0.2%, as shown in Figure 1f. The as-cast

samples along RD present $\langle 133 \rangle$ texture and a weak $\langle 112 \rangle$ texture when deformed to 0.2%, whereas the as-cast samples along TD possess strong $\langle 001 \rangle$ and weak $\langle 133 \rangle$ textures, as shown in Figure 1 g. The

variation of texture during deformation is related to grain rotation, which will be disclosed by analyzing in-situ EBSD for as-built, heat-treated and as-cast Al-Mn-Sc alloys.

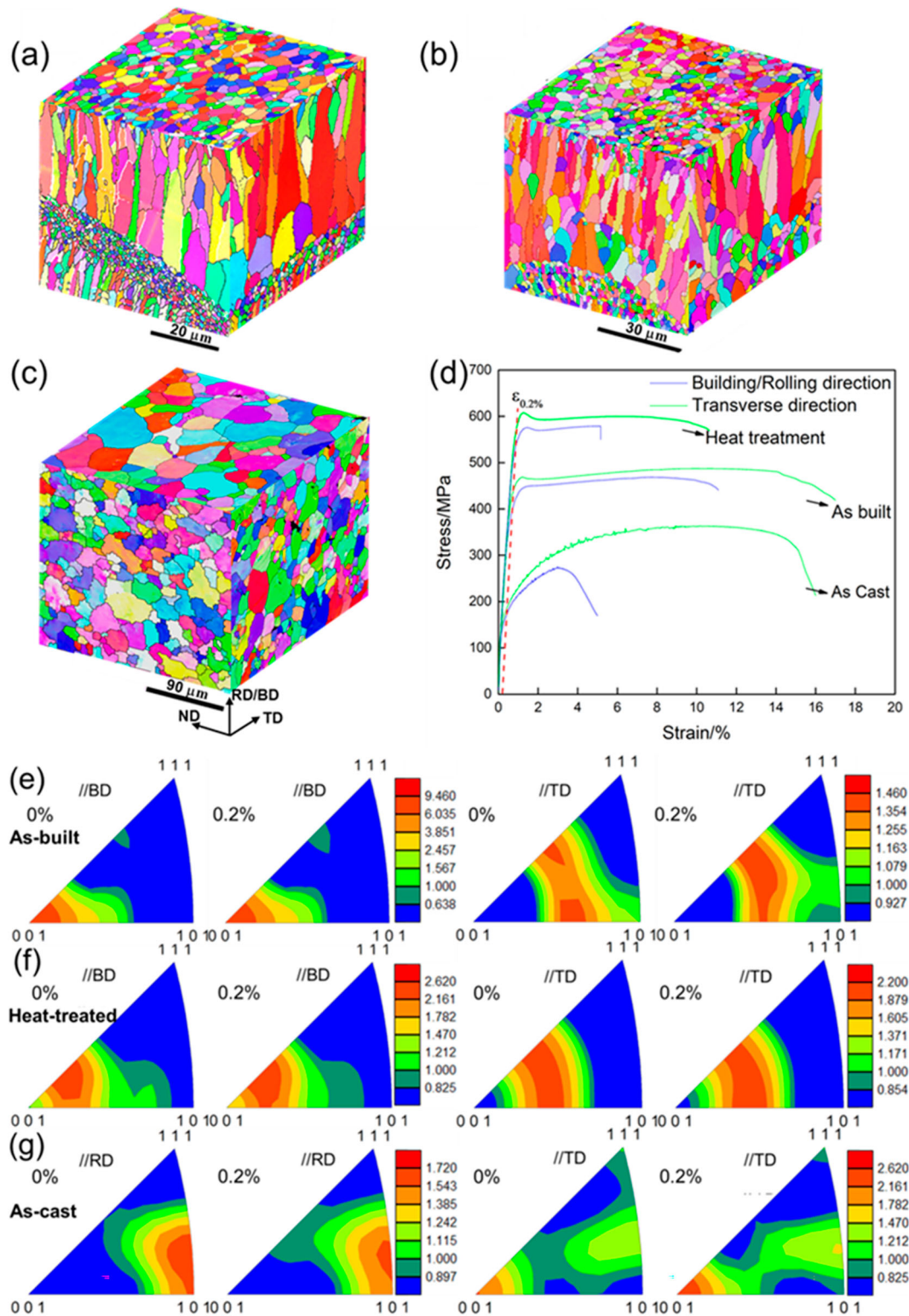


Figure 1. The inverse pole figures (IPF) from three-dimensional views of (a) as-built, (b) heat-treated and (c) as-cast Al-Mn-Sc alloy, (d) stress-strain curves and textures components of (e) as-built, (f) heat-treated and (g) as-cast Al-Mn-Sc alloy.

3.2. In-situ EBSD

Figure 2 demonstrated the microstructure evolution at the strain of 0% and 0.2%, grain orientation distribution,

and grain rotation of as-built Al-Mn-Sc alloys in building and transverse directions. It could be found that the building direction was dominated by columnar crystals, while the cross section mainly consisted of equiaxed

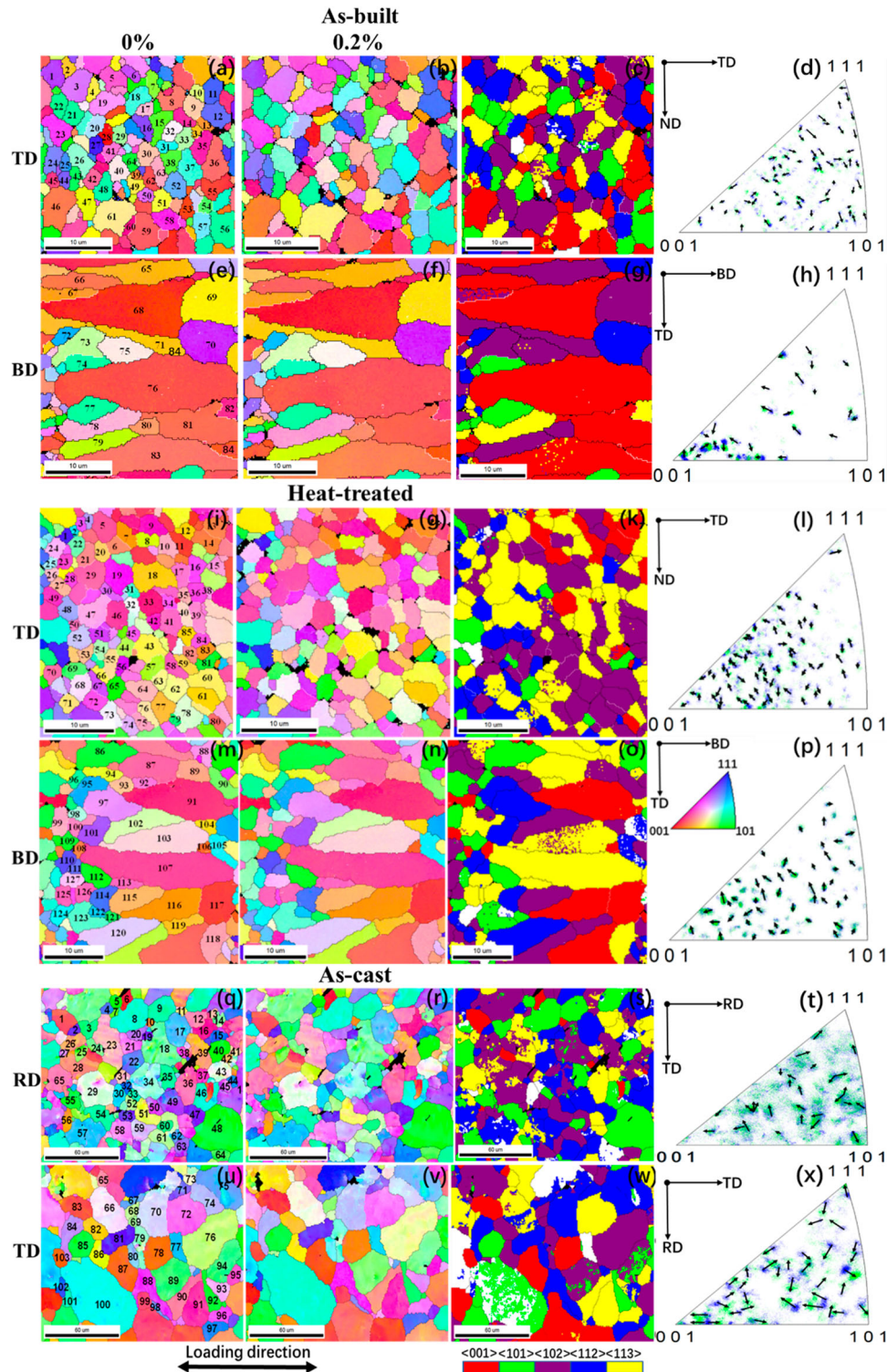


Figure 2. IPF images at the different strains, grain orientation distribution and grain rotation images of Al-Mn-Sc alloys in transverse direction (TD) and building direction (BD)/Rolling direction (RD). (a, e, i, m, q and u) IPF maps at the strain of 0%. (b, f, g, n, r and v) IPF maps at the strain of 0.2%. (c, g, k, o, s and w) Grain orientation distribution images. (d, h, l, p, t and x) Grain rotation images.

crystals, as shown in Figure 2a and 2e. To facilitate the research, all grains were classified into five orientations of $\langle 001 \rangle$, $\langle 101 \rangle$, $\langle 102 \rangle$, $\langle 112 \rangle$, and $\langle 113 \rangle$ with a tolerance of 15° , as shown in Figure 2c and 2g. Figure 2d and 2h illustrate the grain rotation in both directions at the strain of 0% and 0.2%. The blue and green in Figure 2d and 2h indicate the initial grains and the grains deformed by 0.2%, respectively, and the black arrows represent the lattice rotation. With an increase in strain, the $\langle 101 \rangle$ orientation almost always exhibits the smallest average rotation angle of 1.4° and 1.58° and the orientation $\langle 112 \rangle$ possesses a maximum average rotation angle of 1.81° and 1.75° in both directions, as shown in Table 1. The microstructure morphologies of SLMed Al-Mn-Sc alloy after heat treatment at 300°C for 5 h are similar to that of as-built Al-Mn-Sc alloy in both directions due to high thermal stability, as shown in Figure 2i and 2m. After deformation of 0.2%, grain rotation shown in Figure 2l and 2p occurred. It could be found that the orientation with minimum rotation angle was $\langle 001 \rangle$ in both directions, whereas the orientation with the largest rotation angle was $\langle 112 \rangle$ and $\langle 113 \rangle$ in the building and transverse direction. The minimum and maximum rotation angles of as-cast Al-Mn-Sc alloy are $\langle 001 \rangle$ and $\langle 101 \rangle$ orientations in the rolling direction, and $\langle 102 \rangle$ and $\langle 113 \rangle$ orientations in the transverse direction, respectively.

3.3. Analysis of precipitates

Due to the inherent heat treatment of SLMed Al-Mn-Sc alloy, massive precipitates will be formed at the inner and boundary of the grain, which mainly consists of Al-Mn phases and $\text{Al}_3(\text{Sc}, \text{Zr})$, as shown in Figure 3. The average radius of the precipitate Al-Mn is about 55.27 nm, while the large-sized $\text{Al}_3(\text{Sc}, \text{Zr})$ in Figure 3b formed by the agglomeration of nanoscale $\text{Al}_3(\text{Sc}, \text{Zr})$ particles shown in Figure 3c possesses the average radius of 26.13 nm. After heat treatment at 300°C for 5 h, the precipitated phase Al_6Mn with an average radius of 33.54 nm appeared at the inner and the boundary of the grains simultaneously, and a large number of diffusely distributed $\text{Al}_3(\text{Sc}, \text{Zr})$ with the average radius of 1.41 nm were formed within the grains, as shown in Figure 3f and 3 g. The as-cast Al-Mn-Sc alloy exhibited the lowest strength due to the extremely coarse Al_6Mn with an average radius of $1.62\ \mu\text{m}$ shown in Figure 3i and larger $\text{Al}_3(\text{Sc}, \text{Zr})$ with an average radius of 47.36 nm in Figure 3j. Figure 3k and 3i demonstrates that the existence of nano-precipitated phase cannot be found in as-cast Al-Mn-Sc even performing the same heat treatment.

Table 1. The fraction and grain rotation angle of three alloys for two directions.

Sample		Building direction/Rolling direction					Transverse direction				
		$\langle 001 \rangle$	$\langle 101 \rangle$	$\langle 112 \rangle$	$\langle 102 \rangle$	$\langle 113 \rangle$	$\langle 001 \rangle$	$\langle 101 \rangle$	$\langle 112 \rangle$	$\langle 102 \rangle$	$\langle 113 \rangle$
As-built	Fraction	0.546	0.050	0.085	0.261	0.053	0.032	0.127	0.258	0.282	0.220
	Average rotation angle/ $^\circ$	$1.52^{+0.48}_{-0.26}$	$1.40^{+0.69}_{-0.69}$	$1.81^{+0.40}_{-0.40}$	$1.51^{+0.72}_{-0.51}$	$1.76^{+0.10}_{-0.10}$	$1.71^{+0.55}_{-0.72}$	$1.58^{+0.54}_{-0.58}$	$1.75^{+0.92}_{-0.75}$	$1.59^{+1.26}_{-0.90}$	$1.56^{+1.74}_{-0.57}$
Heat treatment	Fraction	0.194	0.119	0.144	0.217	0.293	0.085	0.067	0.132	0.352	0.347
	Average rotation angle/ $^\circ$	$1.22^{+0.20}_{-0.22}$	$1.40^{+0.84}_{-0.40}$	$1.57^{+0.28}_{-0.27}$	$1.43^{+0.57}_{-0.43}$	$1.37^{+0.27}_{-0.37}$	$1.08^{+0.34}_{-0.40}$	$1.31^{+0.69}_{-0.34}$	$1.38^{+0.74}_{-0.45}$	$1.34^{+0.86}_{-0.34}$	$1.43^{+0.91}_{-0.69}$
As-cast	Fraction	0.027	0.231	0.260	0.269	0.161	0.142	0.148	0.201	0.227	0.169
	Average rotation angle/ $^\circ$	$1.32^{+0.16}_{-0.27}$	$1.64^{+0.69}_{-0.90}$	$1.43^{+0.81}_{-0.73}$	$1.49^{+1.00}_{-1.11}$	$1.48^{+0.75}_{-0.49}$	$2.87^{+0.29}_{-0.52}$	$2.88^{+0.36}_{-0.45}$	$2.82^{+0.62}_{-0.67}$	$2.70^{+0.80}_{-1.04}$	$2.89^{+0.69}_{-0.78}$

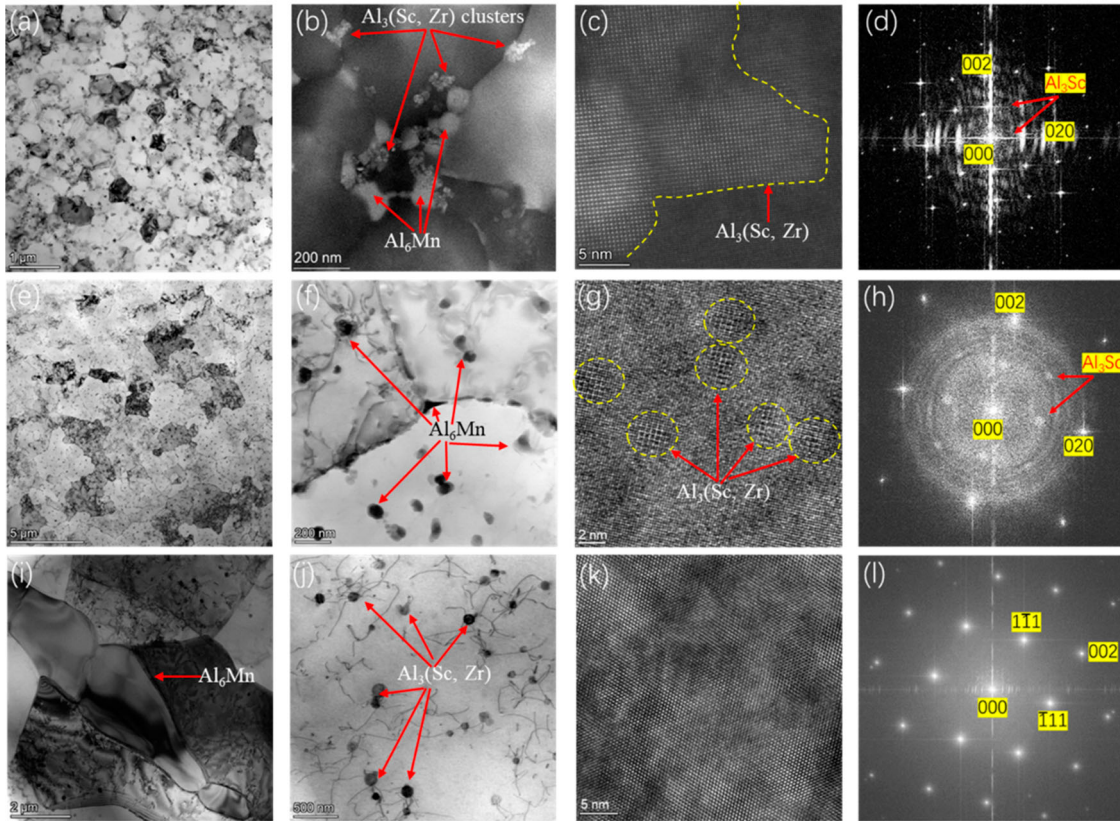


Figure 3. TEM images of as-built, heat-treated and as-cast Al-Mn-Sc alloy. (a) Low magnification images, (b) high magnification images, (c) HRTEM images, (d) FFT pattern of (c) of as-built Al-Mn-Sc alloy. (e) Low magnification images, (f) high magnification images, (g) HRTEM images, (h) FFT pattern of (g) of heat-treated Al-Mn-Sc alloy. (i) Low magnification images, (j) high magnification images, (k) HRTEM images, (l) FFT pattern of (k) of as-cast Al-Mn-Sc alloy.

4. Discussion

By comparing the deformation-induced grain rotations in the two directions for as-built and heat-treated Al-Mn-Sc alloy, it is found that the maximum angles of grain rotation are similar in the columnar grains, and the minimum rotation angles are almost in the orientation $\langle 101 \rangle$ and $\langle 001 \rangle$, implying that the deformation-induced grain rotation may be related to grain orientation. However, the mechanical anisotropy is also affected by grain morphology and grain size. Therefore, it is necessary to investigate the correlation between grain morphology, grain size and grain rotation in different orientations, so as to reveal orientation induced mechanical anisotropy more clearly. To unite grains in different directions and study the relationship of grain size and grain aspect ratio on anisotropy, the maximum rotation angle of grains in each orientation was defined as 1, and the rotation angle fraction was obtained by dividing all grains in the same orientation by the maximum rotation angle. The effects of grain size and grain aspect ratio on grain rotation are analyzed by Equations (4)

and (5) [20]:

$$X = \frac{1}{\sqrt{d}} \cdot \left(\frac{b + \frac{d}{2}}{\sqrt{(b + \frac{d}{2})^2 - (\frac{d}{2})^2}} - 1 \right) \quad (4)$$

$$R = K_a X + C \quad (5)$$

where d is the grain size of each grain, b is the minor axis of the aspect ratio, K_a is the slope of the linear fit, C is the intercept, and R is the fraction of the rotation angle. Based on equations 4 and 5, scatter-fitting graphs of the relationship between grain size, grain shape, and rotation angle for five orientations $\langle 001 \rangle$, $\langle 101 \rangle$, $\langle 102 \rangle$, $\langle 112 \rangle$, and $\langle 113 \rangle$ are obtained, as shown in Figure 4. The slope K_a reflecting the relationship between grain size, grain shape and grain rotation in different orientations was obtained. The difference in strength can be revealed by analyzing the strengthening mechanism, including solid solution

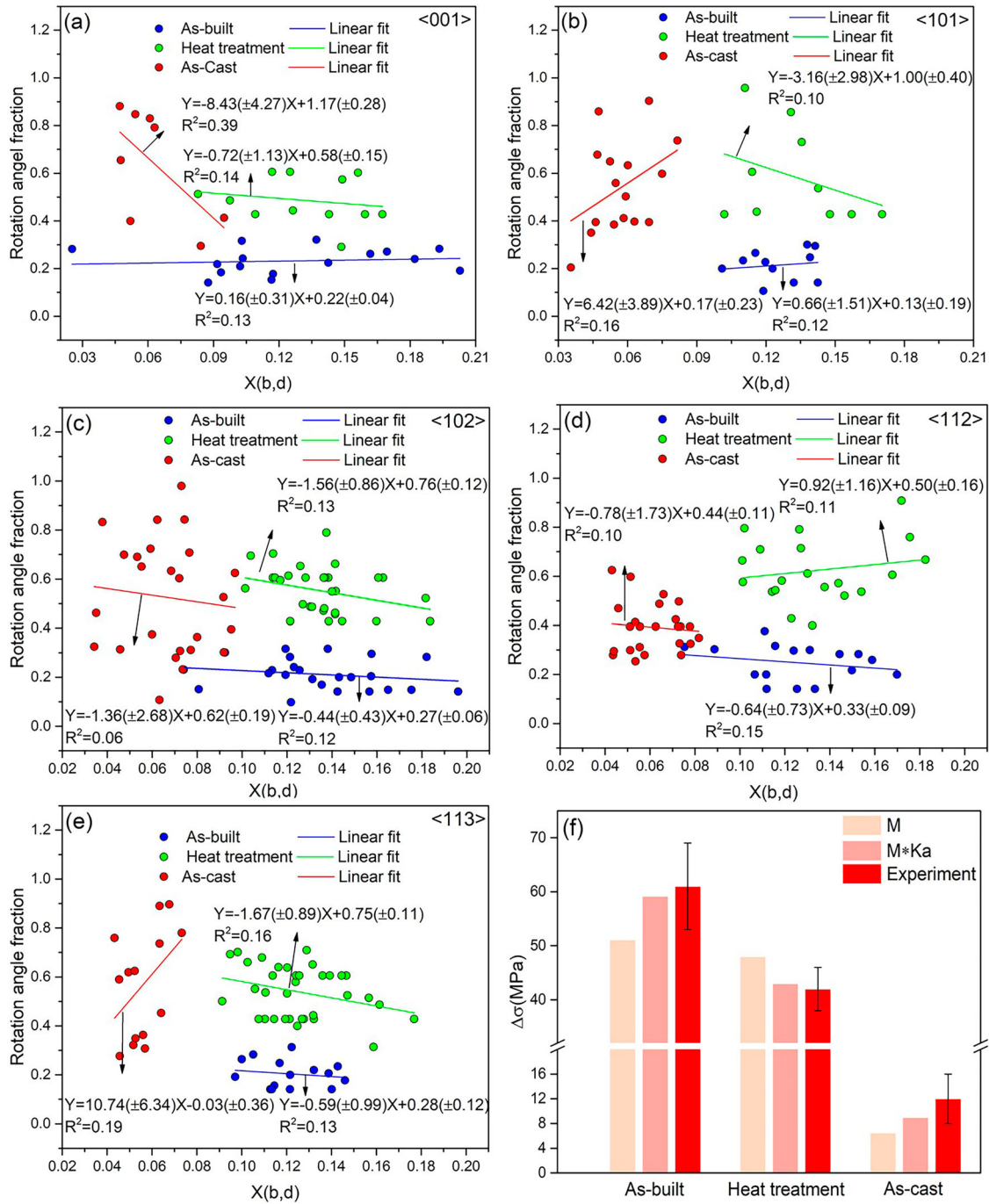


Figure 4. The relationship between rotation angle fraction and grain sizes and aspect ratios in (a) orientation < 001 >, (b) orientation < 101 >, (c) orientation < 102 >, (d) orientation < 112 >, (e) orientation < 113 >, and (f) the strength difference ($\Delta\sigma$) from the theoretical calculation with K_a and without K_a and the experiment.

strengthening, grain boundary strengthening and precipitation strengthening (σ_{ps}) [21]

$$\sigma = \sum_i H_i C_i + \sigma_0 + \alpha k d_f^{-\frac{1}{2}} + (1 - \alpha) k d_c^{-\frac{1}{2}} + \sigma_{ps} \quad (6)$$

where H_i is the solid solution strengthening constant of solution elements (mainly Mn and Mg, $H_{Mn} = 18.35$

MPa/wt% $^\alpha$, $H_{Mg} = 13.8$ MPa/wt% $^\alpha$), α_i is the power law coefficient ($\alpha_{Mn} = 0.9$, $\alpha_{Mg} = 1$) [22], and C_i is the concentration of solute, σ_0 is the friction stress of pure aluminum, α is the area fraction of fine crystals in the binding zone of 70%, k is the Hall-Petch coefficient of 0.17 MPa $m^{1/2}$ [23], d_f is the average grain size of fine crystals in the binding zone, and d_c is the grain size of the intra-layer.

Table 2. Relevant parameters and strength prediction of SLMed Al-Mn-Sc alloys (AB, HT, and AC represents as-built, heat-treated, and as-cast).

Direction Grain orientation	Building direction/Rolling direction					Transverse direction				
	< 001 >	< 101 >	< 112 >	< 102 >	< 113 >	< 001 >	< 101 >	< 112 >	< 102 >	< 113 >
f(AB)	0.546	0.050	0.085	0.261	0.053	0.186	0.136	0.178	0.258	0.207
f(HT)	0.194	0.119	0.144	0.217	0.293	0.085	0.067	0.132	0.352	0.347
f(AC)	0.027	0.231	0.26	0.269	0.161	0.142	0.148	0.201	0.227	0.169
M(AB)			2.61					3.07		
M(HT)			2.84					3.08		
M(AC)			3.15					3.03		
$\sigma_{0.2}$ (MPa) (AB)			390					450		
$\sigma_{0.2}$ (MPa) (HT)			540					582		
$\sigma_{0.2}$ (MPa) (AC)			179					170		

For precipitation strengthening (σ_{ps}), the order strengthening mechanism dominates when matrix dislocation shears the ordered precipitated particles with a small radius, while the precipitated particles with larger sizes are dominated by the Orowan mechanism [21]. Thus, the strength increment of the precipitate bypass is estimated using the Orowan mechanism for as-built and as-cast Al-Mn-Sc alloy, while the Orowan mechanism and order strengthening arising from dislocation shear are used to estimate the strength increment for heat-treated Al-Mn-Sc alloy. The coefficient K_a was introduced into the order strengthening and Orowan mechanism to obtain formulas 7 and 8 to evaluate the precipitation strengthening [24].

$$\sigma_{ord} = \sum (f_i K_{a_i}^n) M_i 0.81 \frac{\gamma_{APB}}{2b} \left(\frac{3\pi V_V}{8} \right)^{\frac{1}{2}} \quad (7)$$

$$\sigma_{Or} = \sum (f_i K_{a_i}^n) M_i \frac{0.4Gb \ln(2\sqrt{2/3}R/b)}{\pi \lambda \sqrt{1-\nu}} \quad (8)$$

where M is the average Taylor factor, γ_{APB} is the anti-phase boundary energy of precipitate (average value for Al_3Sc is 0.5 J m^{-2}), $b = 0.286 \text{ nm}$ is the magnitude of Burgers vector for Al, V_V is the volume fraction of the precipitates from TEM, $G = 25.4 \text{ GPa}$ is the shear modulus of aluminum, R is the average radius of precipitates, λ is the inter-precipitate spacing and $\nu = 0.345$ is Poisson's ratio for aluminum [10,21,24], f represents the proportion of the five grain orientations, n is 0.35.

The average Taylor factors M of three alloys in the building and transverse direction are different, as shown in Table 2. Based on Taylor factors M of two directions for three alloys, the mechanical anisotropy was calculated by introducing K_a . The estimated strengths of building and transverse direction of as-built Al-Mn-Sc are 390 and 450 MPa, those in the heat treatment are 540 and 582 MPa, and those in the as-cast Al-Mn-Sc are 170 and 179 MPa, which can well predict the actual tensile strength and mechanical anisotropy, as shown in Figure 1d. Figure 4f shows the comparison of the strength difference between

the direct calculation of the Taylor factor and the introduction of the anisotropy coefficient K_a and the experimental value, showing that mechanical anisotropy can be predicted more accurately by introducing the anisotropy coefficient K_a .

Figure 5 illustrates the mechanical anisotropy caused by each orientation after introducing the anisotropy coefficient K_a . The strength increment of the transverse direction of most grain orientations is higher than that of the building direction, which can be called positive mechanical anisotropy, while the < 001 > orientation presents negative mechanical anisotropy. The orientation with minimum mechanical anisotropy is < 102 > orientation and the maximum is < 001 >. After heat treatment, the mechanical anisotropy trend of < 001 >, < 102 > and < 113 > is the same as that of the as-built, while the orientation < 101 > presents negative mechanical anisotropy and < 112 > exhibits isotropy, which may be one of the reasons for the decrease of mechanical anisotropy after heat treatment. Therefore, by analyzing the orientation-induced anisotropy, it can be concluded that the mechanical anisotropy can be reduced by increasing the proportion of < 102 > orientations in the as-built Al-Mn-Sc alloy and producing more < 112 > orientations through the suitable heat treatment.

It can be found from Figures 4 and 5 that < 001 > and < 113 > orientations contribute the most mechanical anisotropy (32 and 30 MPa) to the as-built Al-Mn-Sc alloy, but the K_a of both orientations is the smallest (0.16 and 0.59), while the remaining orientations contribute less mechanical anisotropy (6–20 MPa) and larger K_a (0.44–0.66). For heat-treated Al-Mn-Sc alloy, < 001 > and < 102 > orientation orientations exhibit larger mechanical anisotropy (36 and 45 MPa) and smaller K_a (0.72 and 1.56), and the remaining orientations possess smaller mechanical anisotropy (0.3–22 MPa) and larger K_a (0.92–3.16). This means that there is an inverse relationship between K_a and orientation induced mechanical anisotropy, which can also be

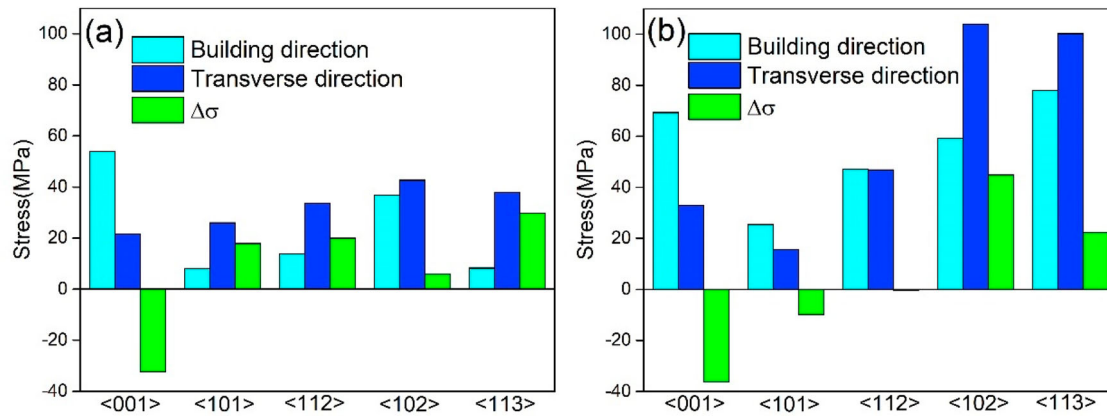


Figure 5. Mechanical anisotropy of five grain orientations for (a) as-built and (b) heat-treated Al-Mn-Sc alloy.

determined by the relationship between K_a and experimental mechanical anisotropy (Figure 1d). The K_a of the as-deposited, heat-treated and as-cast Al-Mn-Sc alloy is 0.15–1, 0.7–3.2, and mostly 3–10, while the as-built samples have the largest mechanical anisotropy, followed by heat-treated and as-cast samples (Figure 4f). Therefore, the meaning of the anisotropy coefficient K_a is that it can reflect the orientation-induced mechanical anisotropy.

5. Conclusion

By investigating mechanical anisotropy of Al-Mn-Sc alloy using in situ EBSD tensile, we found that the as-built Al-Mn-Sc alloy presented mechanical anisotropy due to epitaxial growth of columnar grains, which was weakened after heat treatment at 300 °C for 5 h, while the as-cast Al-Mn-Sc alloy showed isotropy after the same heat treatment. The deformation induced grain orientation play a critical role in mechanical anisotropy. Compared with the traditional model of Taylor factor M , the mechanical anisotropy of Al-Mn-Sc alloy of as-deposited, heat-treated, and as-cast Al-Mn-Sc alloy can be predicted more accurately by introducing coefficient K_a . By analyzing the orientation induced anisotropy, the orientation with minimum mechanical anisotropy is <102> and the maximum is <001> for as-built Al-Mn-Sc alloy. The anisotropy coefficient K_a related to grain orientation presents an inverse relationship with the mechanical anisotropy.

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Disclosure statement

No potential conflict of interest was reported by the author(s).

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