



Experimental study on Nb₃Cr₂Ti_{0.9}W_{0.4}Fe₆ high entropy alloys with high strength and good ductility by laser powder bed fusion

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ABSTRACT

High entropy alloys (HEAs) have been widely used in aerospace complex parts due to their unique mechanical properties. HEAs with body-centered cubic (BCC) structure have high strength and low plasticity. Due to the synergistic effect of solid solution strengthening and lower electron concentration, high entropy alloys containing refractory elements tend to have considerable strength and plasticity. A novel high strength Nb₃Cr₂Ti_{0.9}W_{0.4}Fe₆ HEA was developed in this work. The spherical powders were prepared by the dual process of mechanical alloying and high temperature remelting spheroidization (DP) since such technology has combined effect of mechanical milling (plastic deformation) and temperature lifting (atomic diffusion) to achieve the composition homogenization. At the same time, the powder prepared by gas atomization (GA) method is compared, and the printed parts are compared under the same parameters of laser powder bed fusion (LPBF). The results indicate that DP-SLM has higher microhardness and wear resistance than GA-SLM. The yield strength (YS), ultimate tensile strength (UTS) and average elongation (ε) of the samples measured by DP-SLM were 953 MPa, 1159 MPa and 14.8 %, respectively. The average elongation (ε) is 25 % higher than GA-SLM. The hot compression test of Nb₃Cr₂Ti_{0.9}W_{0.4}Fe₆ was carried out with Gleeble-3800 thermal simulation machine. The thermal processing diagram of the alloy under different stress variables was drawn by establishing the thermal deformation constitutive equation. The DP-SLM material recovers when the deformation temperature is 1000 °C with deformation rate stays at 0.01 s⁻¹. Nb₃Cr₂Ti_{0.9}W_{0.4}Fe₆ alloy processed at 1050–1150 °C and a strain rate of 0.005–0.1 s⁻¹ shows high power dissipation efficiency (up to 37 %) and better dynamic recrystallization structure. Nb₃Cr₂Ti_{0.9}W_{0.4}Fe₆ can be considered widely used in the engine combustion chamber, guide, turbine blade and turbo disk four hot end components. The BCC-HEA containing refractory element can also realize the synergistic increase of strength and toughness through reasonable design.

1. Introduction

The high entropy alloys (HEAs) have exhibit high strength over a wide temperature range and under a large strain state when its structure contains body-centered cubic (BCC). At the beginning, refractory high entropy alloys (RHEAs) are focused from 3d transition metals for high temperature applications with single-phase BCC structure [1,2]. However, their plastic deformation mode is quite different from that of metals with face-centered cubic (FCC) structure. This is mainly due to their crystallographic lattice characteristics and high lattice friction [3,

4]. Due to the synergistic effect of solid solution strengthening and lower valence electron concentration, RHEAs often tend to exhibit comparable strength and plasticity with BCC structure [5]. Therefore, the HEAs with BCC structure can also achieve the synergistic increase of the strength and toughness by reasonable design. There are many ways to prepare refractory high entropy alloys, but additive manufacturing is believed to be a good choice. As one of the most promising and hot metal additive manufacturing technologies, laser powder bed fusion (LPBF) can realize the rapid formation of complex components effectively [6]. It has been proved that such technology shows a remarkable effect on enhancing the

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mechanical properties on high entropy alloy complex parts [7].

The composition uniformity of alloy powder has great influence on the performance of molded parts in LPBF. Especially with refractory elements, the prepared powders are prone to component segregation if it is not handled properly. Nevertheless, the spherical powders required for additive manufacturing are usually prepared by traditional gas atomization method (GA). The advantages of GA are high yield, wide variety of powder and low cost. But it is easy to introduce foreign impurities into the powder prepared. Meanwhile, the alloy powder containing high melting point elements (Nb, Cr, W, etc) is prone to severe segregation of components, heavily influencing the performance of printing parts. And the powder prepared by this technology contains more hollow powder. These bottlenecks limit its further development. Due to equipment temperature and process flow, it is difficult to ensure uniform composition of alloy ingot containing refractory elements. As a result, the resulting ingot contains refractory elements attached to the grain boundaries (partially or unalloyed) in the form of a single substance. Therefore, the alloy ingot has component segregation. Therefore, the spherical powders produced by GA are difficult to prepare RHEAs and ensure uniform composition [6–8]. How to control alloy powders to make them have the characteristics of high density, high sphericity and uniform composition has become the latest concern in the field of additive manufacturing. [9–11].

Based on the disadvantages of the traditional process of preparing refractory alloy powder, this paper provides an innovative idea. The RHEA spherical powders with uniform composition were prepared by mechanical alloying (MA) and High temperature remelting spherification technology (HTRS). MA technology achieves the composition uniformity of RHEA powder through the combined action of mechanical milling (plastic deformation) and temperature raising (atomic diffusion) [11–13]. But the shape of these powders is generally polygonal granular, and the fluidity is poor [14–17]. In this paper, we are inspired by the idea of Temble's droplet experiment: The single particle of powder is melted by plasma melt-blowing and then cooled to obtain spherical powder directly. Thus, the HTRS technology has been invented and verified as a wise way to solve the issue. In conclusion, MA can promote the alloying of refractory elements, while HTRS could change the powder morphology (spheroidization) [18,19]. The process obtained by combining these two processes is called dual process (DP). Additive manufacturing has the characteristics of personalized printing and customization. The production of alloy powder with special requirements is low, and the general alloy on the market is generally difficult to be used directly. In order to adapt to the trend of individualization and customization, it is very important to design and develop special alloy and its preparation process in small batch. In addition, the self-developed technology of this project can significantly reduce the cost of powder production. This is an important way to promote the wider application of additive manufacturing.

In the previous research, our group has researched Ni-Cr-W-Fe-Ti high-entropy alloys, proving that our design concept is feasible [20]. The purpose of this paper is to improve the strength and plasticity of RHEAs with BCC structure through composition design control. It is believed that Nb can refine grain, improve wear resistance and tensile strength of alloy. In addition, Nb has high temperature strength and fatigue resistance. Thus, we have designed newly $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA and it is prepared by LPBF technology with the aim of developing inexpensive non-equal atomic ratio HEA on high mechanical properties. According to the slow diffusion effect of HEAs, these are beneficial to inhibit the precipitation of intermetallic compounds [21]. The HEAs prepared by this process have the characteristics of good metallurgical bonding with the matrix, controllable thickness and uniform microstructure. In addition, the homogenization and related mechanical mechanism of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ have been studied by means of experiments and finite element simulation. The cost-effective HEA-SLM with exceptional properties can be applied to other BCC-HEAs to enhance their mechanical properties, leading to prolong the life of high

temperature complex parts in aerospace field. This research will significantly enhance the economy and dependability of novel material design. $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ can be widely used in the engine combustion chamber, guide, turbine blade and turbo disk four hot end components. In addition, it is also used for casing, ring, afterburner and tail nozzle and other components. And under the self-developed pulverizing process, the cost can be reduced by more than 20 %.

2. Material and methods

Table 1 displayed the parameter characteristics of HEA components in the experiment. $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA powders were prepared by mechanical alloying process. After the alloyed powder was dried sufficiently, high temperature remelting equipment was used to prepare spherical powder, provided by Shaanxi Zhiqi Kaiwu Co. Table 2 exhibits related parameters. The printing parameters were set: laser power reached up to 550 W with 3200 mm/s scanning speed. laser wavelength is 1.06 μm , the thickness of each layer is 0.02 mm and spot diameter is 0.1 mm. 316 steel was regarded as the basic materials (BM) and its composition is shown in Table 3.

The phases of the printing parts were analyzed by X-ray (D8 Advance, Bruker AXS, Germany, $\lambda=0.15406\text{ nm}$). The microstructure and powder morphology were characterized by Transmission Electron Microscope (TEM, Talos F200x, FEI, America), Scanning Electron Microscope (SEM, Regulus8230, Hitachi, Japan), EDS (SEM, Regulus 8230, Hitachi, Japan). The HEA samples were tested by Electron back scatter diffractometer (D8 Advance, EBSD, Fig. 1a). The ML-300 pin-disk wear test machine (Beijing) was used for a two-body wear test, with a flat cylinder measuring $\phi 5\text{ mm} \times 30\text{ mm}$. The rotation rate was 108 r min^{-1} and the slip rate was set as 10 mm r^{-1} . The MFT-EC4000 reciprocating tribometer was utilized for all sliding tests with a 8 mm diameter ball being loaded onto the RHEA SLM sample. The load used was 15 N with testing time 2400 s. The average sliding velocity was 20 mm/s. The microhardness was analyzed by the micro Vickers tester on the longitudinal section. The size of the tensile sample was prepared in accordance with standard (GB/T 228–2002). ABAQUS software [12] was utilized to perform a finite element analysis of stress distribution, leading to the establishment of a sliding test model. Fig. 1b describes the schematic diagram of sliding test simulation. To improve the accuracy, components were carefully selected for simulation (Fig. 1c) [22]. C3D8R cells were used to divide the entire model into 70,848 hexahedral cells (mesh independence analysis doubled the number of grids, resulting in an error of less than 1 %). The GCr15 ball was regarded as a discrete rigid body [23]. After the mesh-independent analysis, the error is less than 1 % when the number of meshes is doubled. Simulation parameters are as follows: the load was 15 N and 2 N, respectively. The sliding speed was set to be 20 mm/s.

3. Results and discussion on powder production

3.1. Alloying degree analysis of alloyed powder

Generally speaking, the particle size of the alloy powder has been determined after the preparation of MA. The main reasons for affecting the particle size of the powder are milling time and rod material ratio. Under the milling process, the fragments and the flake ductility powder are welded together to form composite lamellar structure and the layer spacing decreases. The powder surface curvature increases, the diffusion distance decreases, and the diffusion is easier to carry out. After continued milling, the powder particles were further refined and the size was uneven. This is due to the continuous impact, deformation and shear of the composite layered structure of cold welding.

The high temperature remelting technique is inspired by Temble's droplet experiment. Spherical powder was obtained by melting single particle through falling tube melting and cooling. Fig. 2 shows the SEM images of spherical powders prepared at melting temperatures of

Table 1

Parameter characteristics of components in high entropy alloys.

Element	Atomic radius/ Å	Crystal structure	Electro-negativity	Melting point/ °C	VEC	Density/ (g/cm ³)
Cr	1.2491	BCC		1857		7.19
Nb	1.4672	BCC	1.70	2468	6	8.57
Fe	1.2412	BCC	1.60	1538	5	7.86
Ti	1.4615	HCP	1.83	1666	8	4.54
W	1.3670	BCC	1.51	3410	4	19.35
			2.36		6	

Table 2

Experimental parameters.

Parameter	Value
Temperature K	1500/1550/1600
Particle feed rate g·min ⁻¹	1.238/1.866/2.124/2.578
Protecting gas flow rate ml·min ⁻¹	120 (Ar)

Table 3

The chemical composition of 316 L substrate.

element	Ni	Cr	Mo	S	P
mass (%)	10.0 ~14.0	16.0 ~18.5	2.0 ~3.0	≤ 0.03	≤ 0.035

1500 K, 1550 K and 1600 K, respectively. As shown in Fig. 2a, the powder obtained at 1500 K has a very low degree of spheroidization (12.5 %). Nevertheless, most of are unmelted MA layered powders (alloy powders are not spheroidized). Due to the insufficient thermal field temperature in the remelting process, it is difficult for the powder to completely melt in a limited time. At 1600 K, the sphericity of the powder is greatly improved (97.7 %), and the powder surface is uniform (See Fig. 2f). For comparison, Fig. 2d shows the spherical powder obtained by GA process. The observation shows that the spheroidization rate of the powder is high and the surface of the powder is rough. Since the elements in the design of HEA composition contain refractory elements, GA method has some problems such as hollow powder and poor apparent quality for alloyed powder. Therefore, there is no doubt that DP method shows advantages on the powder sphericity with high quality. In the GA preparation process, the end face of the metal base metal is melted at high temperature to form a large molten pool and then thrown out. The HEA with refractory element W is difficult to segregate. This also reflects the superiority of DP method in the milling process.

3.2. Phase structure

The XRD patterns were tested for the powder and sample on Nb₃Cr₂Ti_{0.9}W_{0.4}Fe₆ under GA and DP process (shown in Fig. 3). According to the structure of the elements in Table 1 and the HEA characteristics forming a single or simple structural phase, the phase are the same in the powder and its corresponding sample: a single body-centered cubic (BCC) solid solution. Since supersaturated solid solutions in meta-stable state often have more stable BCC phase structure after scanning by high energy laser beam, it is not easy to undergo phase transformation.

Compared with the GA-SLM, the angle of the DP-SLM BCC phase (110) peak moves to the right according to the Bragg equation, corresponding to the expansion of the lattice constant. Since the refractory element W, Nb are affected by high temperature in the previous DP process, the atomic diffusion is more intense. This makes the W atoms better dissolved in the solid solution, leading to stronger lattice distortion [24-25]. Meanwhile, the enthalpy of mixing between elements directly affects stability of HEAs and the phase structure as well as is regarded as an indispensable reference for HEA composition design (shown in Table 4). In this paper, the absolute enthalpy of mixing between Nb and other elements is the largest, indicating that the binding force of Nb and other elements is weak. Therefore, it is easy to segregate during solidification [26]. In addition, the radius of the Nb atom is the largest in this component. Its distribution in the solid solution has the most intense effect, which is easy to cause lattice distortion.

3.3. Microstructure

The cross-section microstructure of Nb₃Cr₂Ti_{0.9}W_{0.4}Fe₆ HEA is shown in Figs. 4a-4b. Figs. 4c-4d shows the longitudinal sections of the DP and GA sample, respectively. The results represent that the microstructure exists in the form of equiaxed crystals by both GA and DP method. In order to better observe the experimental results, the spherical powder used in the experiment was screened after the particle size phase. Therefore, the grain size of the molded parts obtained under the

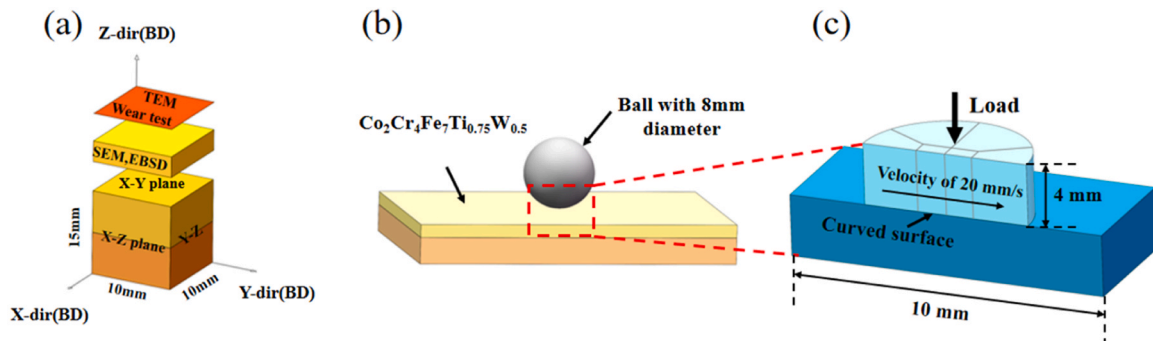


Fig. 1. (a) SLM diagram for characterization analysis; (b) The schematic diagram of the sliding test simulation; (c) FEA model of contact area.

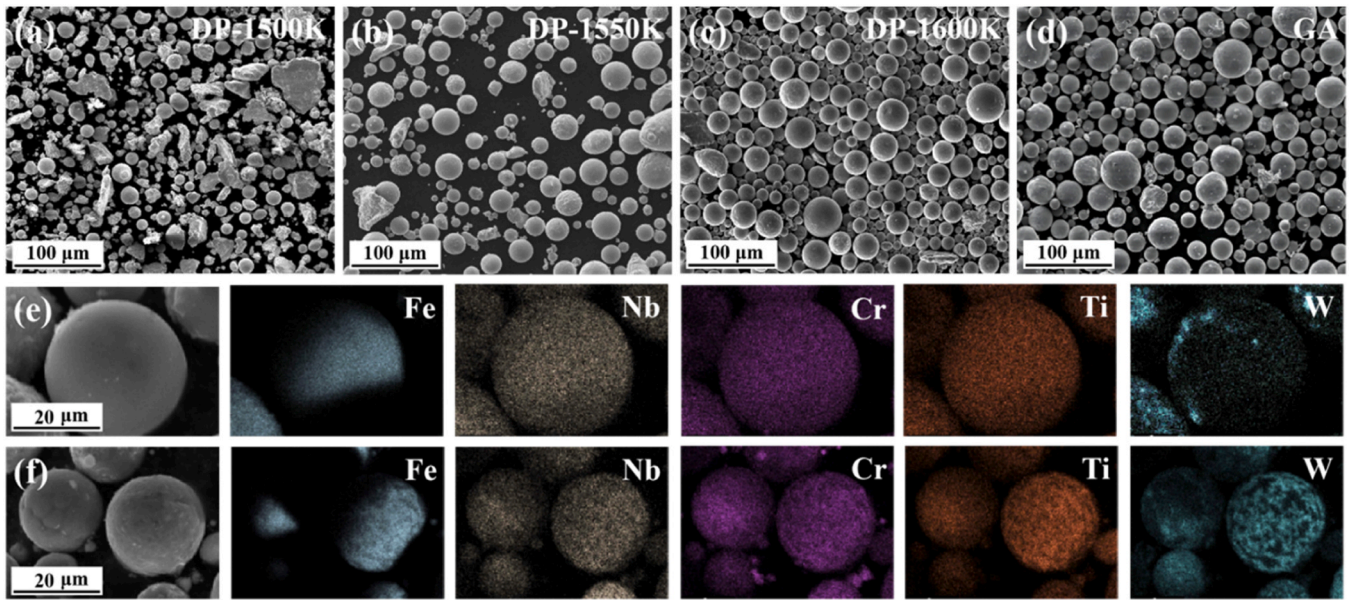


Fig. 2. (a)-(d) SEM images of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ spherical powders prepared at different melting temperatures; (e) EDS of spherical powder (GA); (f) EDS of spherical powder (DP).

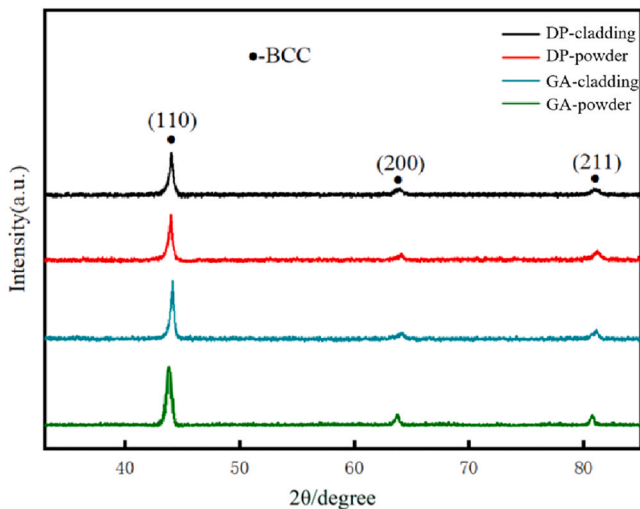


Fig. 3. The XRD pattern of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$.

Table 4

Enthalpy of mixing of each element atom pair /(kJ/mol) [35].

Mixing enthalpy	Ti	Cr	W	Fe	Nb
Ti	–	–7	–6	0	2
Cr	–7	–	1	–1	–7
W	–6	1	–	16	–8
Fe	0	–1	0	–	–16
Nb	2	–7	–8	–16	–

same process parameters is almost the same. The findings illustrated in Fig. 4c and Fig. 4d suggest that the produced parts are effectively bonded to the substrate, and no cracks are present. The design of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ high entropy alloy is proved to be reasonable.

Fig. 5 describes the SEM and EDS diagram of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ in two different SLMs (GA and DP). Characterization tests show that the elements Nb, W, Cr, Ti and Fe in the high entropy alloy have high alloying degree and uniform element distribution. EDS showed that

there was no segregation in the structured components, especially in the refractory elements. The MA technology enables solid alloying between insoluble elements. By mechanical alloying, refractory elements can be promoted to enter the solid solution to achieve supersaturated solution [27–28]. In addition, the spheroidization process undergoes high temperatures. This further promotes the diffusion of refractory elements.

In Fig. 6, TEM was used for microscopic characterization on $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA in BCC solution. Fig. 4a describes the bright field image of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$. Figs. 6b–6c represent region 1–2, with region 1 along axis [013], region 2 along axis [–111], respectively. By calculating the distance between TEM surfaces, it is confirmed that $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ has BCC structure. The results are consistent with those obtained by XRD.

Fig. 7a shows the splicing metallographic diagram of DP-SLM from bottom to top. Fig. 7b depicts the bottom, middle and top layers of DP-SLM through EBSD images, and Fig. 7a is marked in red. Fig. 7c represents the orientation imaging microscope (OIM). Fig. 7d shows the HAGBs (see red line) and LAGBs (see blue line) statistics for the regions shown in Fig. 7c (A_1), (B_1), and (C_1). The LAGBs accounted for 8.6 %, 10.7 % and 15.3 %, respectively. Fig. 7e depicts the statistical results for different regional granularity. The $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA specimens are metastable structures with non-equilibrium solidification. This microstructure characteristic is related to the layer-by-layer cumulative manufacturing method in SLM rapid preparation of additive specimens [29–32]. The small angle grain boundary distribution of the alloy is caused by rapid heating and rapid solidification. The lap area between the adjacent melting channel and the adjacent forming layer will be heated repeatedly under laser irradiation, which causes part of it to recrystallize and gradually change from small-angle grain boundaries to large-angle grain boundaries [33–36]. In addition, both grain growth and grain boundary flattening can reduce the grain boundary area and thus reduce the total grain boundary energy. Atoms at grain boundaries have high kinetic energy and also deviate from the equilibrium position. Due to the existence of defects such as holes, impurity atoms and dislocations at the grain boundaries, the diffusion rate of atoms at the grain boundaries is greater than that at the grain boundaries. Therefore, the grain boundary energy is higher than the energy within the grain boundary. Since the new phase has higher energy and atomic activity, it tends to nucleate at the grain boundary. Consequently, solute atoms spontaneously segregate to these boundaries during their migration.

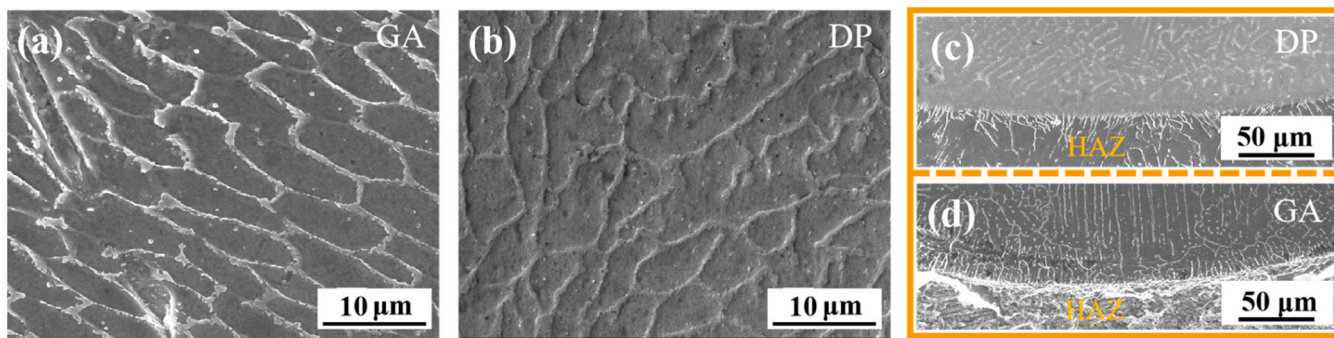


Fig. 4. The cross-sectional microstructure of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA.

From the longitudinal cross-section in Fig. 7c (A_1), (B_1), and (C_1), it can be seen that the adjacent molten pools are closely superimposed with uniform and fine dendrite structure inside the molten pools. In the molten pool of the additive manufacturing printed samples, the columnar crystals transform into equiaxed crystals [37]. The equiaxed crystals are finer, being beneficial to prevent printing cracking.

In addition, the bottom region between the base material and the bottom of the additive is mainly composed of equiaxed crystals in Fig. 7c. But the grain size is not the same. Moreover, the grain size statistics in Fig. 7e can further illustrate the situation. This phenomenon is related to the crystal solidification rate and the temperature gradient at different positions in the fuse. In the process of SLM preparation of high-entropy alloy additive specimen, high energy laser irradiation and rapid cooling in a very short time will lead to the formation of fine grains, while the large equiaxed crystals will be further formed due to the large temperature gradient and heat flow direction. In the process of SLM melting powder, there is a larger temperature gradient in the upper and lower edges of the melting pool, giving priority to rapid solidification. When the laser scans the next layer of powder, the heat will rapidly diffuse and remelt the solidified surface, and the precipitates uniformly distributed on the BCC solid solution in the remelt zone will provide new heterogeneous nucleation points, and the alloy will quickly nucleate into equiaxed crystals. At the same time, the original equiaxed grains will show ductile growth characteristics along the forming mode and penetrate multiple forming layers. In addition, the middle and top structures are mainly composed of fine equiaxed crystals. This is due to the fast scanning speed and short solidification time of high entropy alloy forming in the process of high energy laser: the grain nucleation is less than the growth, and the growth becomes a fine structure, thus affecting the alloy properties.

3.4. Microhardness and wear resistance

In Fig. 8a, the results of the microhardness test demonstrate the newly designed $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA show a much higher average microhardness than the substrate, up to 672 $\text{HV}_{0.2}$ (DP-SLM). The experimental results show that the alloy powder can complete the inter-diffusion of atoms under the action of high energy laser beam, being conducive to the formation of fine grains and complete solid solutions of HEA. The solid solution strengthening and fine crystal strengthening can significantly improve the microhardness while the rapid cooling and solidification process can refine the grain of HEA-SLM. Fig. 8b describes the relationship between friction coefficient of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA and time (All the experiments have been repeated 3 times). As friction progresses, the contact area grows in a continuous manner. This leads to an increase in the coefficient of friction. After entering the stable wear zone, the stable friction coefficients of DP-SLM and GA-SLM are 0.08 and 0.09, respectively. The stable friction coefficient of the substrate (316 L) is 0.13. Fig. 8c describes the wear weight loss of HEA-SLM (DP, GA and 316 L) when loaded with 15 N on 500 mesh Al_2O_3 sandpaper.

It is easy to find that the wear weight loss of the $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA-SLM is lower than that of the 316 L stainless steel substrate as a whole. At this time, the increase of the hardness of the molten layer is mainly due to the intersolubility of atoms with different sizes, which causes lattice distortion and produces the effect of solid solution strengthening. In addition, previous studies have also shown that the W addition could stimulate grain refinement [38]. Thus, the wear resistance of the $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA is improved under the two strengthening mechanisms. It is worth noting that the W element diffuses more evenly under the DP process. Compared with GA, it also causes more lattice distortion. This also explains why the friction coefficient of DP-SLM is lower than GA-SLM.

Figs. 9a–9c show the wear morphology of the HEA under 15 N load under the substrate and different milling processes. At a 15 N load, the 316 L substrate has a highly abrasive surface, stratified and prone to falling objects, resulting in a high wear rate. A plastic deformation zone with uniform stress is formed during friction. This is due to the deep depth of the abrasive particles or micro-convex embedded in $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ during friction, causing serious damage to the surface, and the edge of the furrow has formed a jagged edge due to micro-cutting [39]. The microstructure of DP wear surface is equiaxed. Furthermore, the fine furrows also show that the surface has undergone fine abrasive wear. The appearance of the microstructure proves the low degree of wear.

In order to better observe the morphology distribution after wear, three-dimensional laser scanning microscope was used to detect the 3D morphology of wear samples (Figs. 9d–9f). Protrusions are particles projected or deposited by the broken layer. At this time, It indicates that the surface microhardness stays in a low value, and the Al_2O_3 particles can easily penetrate the metal surface. This explains the mass loss and surface roughness of 316 L. As the overall hardness of newly designed $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ specimens shows high value, the wear degree of wear specimens is parallel groove. In addition, the grooves on the specimen surface were narrowed under the DP process [40–42]. As a result, GA-SLM shows a mixed wear mechanism: both micro-ploughing and micro-cutting. As for DP-SLM, it contains the only micro-cutting.

3.5. Analysis of plastic deformation during friction

Fig. 10 shows the equivalent plastic strain cloud diagram of the high entropy alloy under different milling processes when the ball just touches the surface under 15 N load. The surfaces of GA-SLM (Fig. 10a) and DP-SLM (Fig. 10b) undergo plastic deformation under 15 N loading. The plastic deformation region is ellipsoidal. The secondary surface's plastic strain rate is reduced as strength increases and the two HEAs have a maximum equivalent plastic strain of 0.273 % and 0.178 %, respectively. The degree of downward sag of plastic strain is also different. The depressions of GA-SLM and DP-SLM are 0.087 μm and 0.045 μm , respectively. It can be seen that the plastic deformation will have a certain influence on the test results when measuring the wear

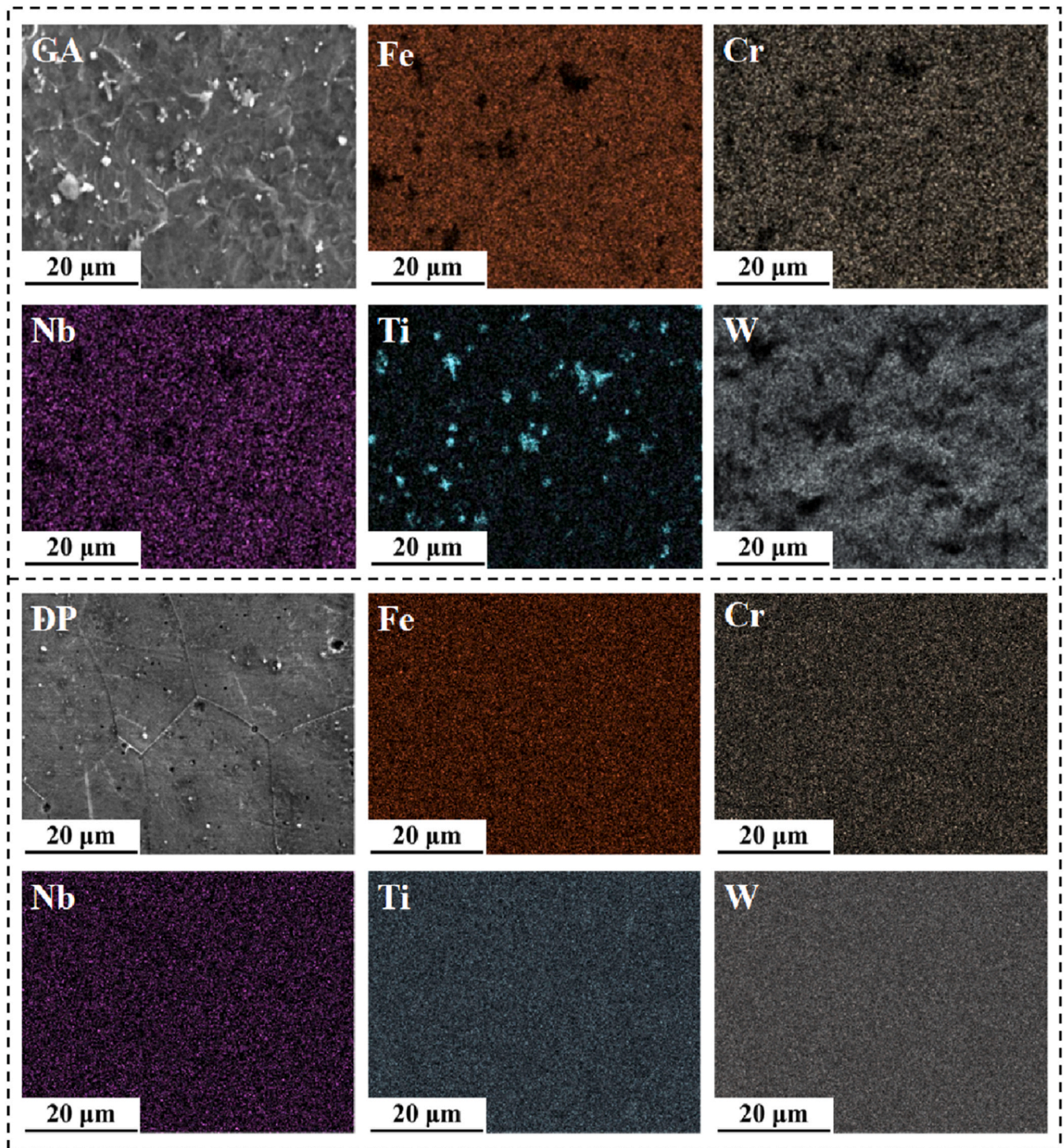


Fig. 5. SEM and EDS of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA by DP.

contour of high entropy alloy. Since the amount of depression is too small, the result can be ignored. Sliding of the HEA is demonstrated by Figs. 10c-10d, which displays both surface and sub-surface plastic deformation. GA and DP alloy SLMs are particularly prone to plastic deformation on the subsurface. This plastic deformation of the surface can increase the direct contact between the HEA and the outermost surface, making the surface of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA after sliding similar to work hardening and thus increasing its wear resistance. Similarly, severe plastic deformation can cause the surface to fracture and form pits. The maximum plastic strain is completely located on the

specimen surface. The variation of maximum plastic strain is similar to that of maximum Mises stress. That is, with the increase of the load, it gradually changes from the sub-surface to the surface. The plastic deformation of the subsurface will result in work hardening of the material. The increase in the strength of the plastic deformation layer can prevent the surface crack from spreading inward. The thickness of the plastic deformation layer should be shown in the color area Fig. 10. In addition to the work hardening of the surface and the suppression of cracks, the large amount of plastic deformation also has a side that is not conducive to frictional properties. When the plastic deformation exceeds

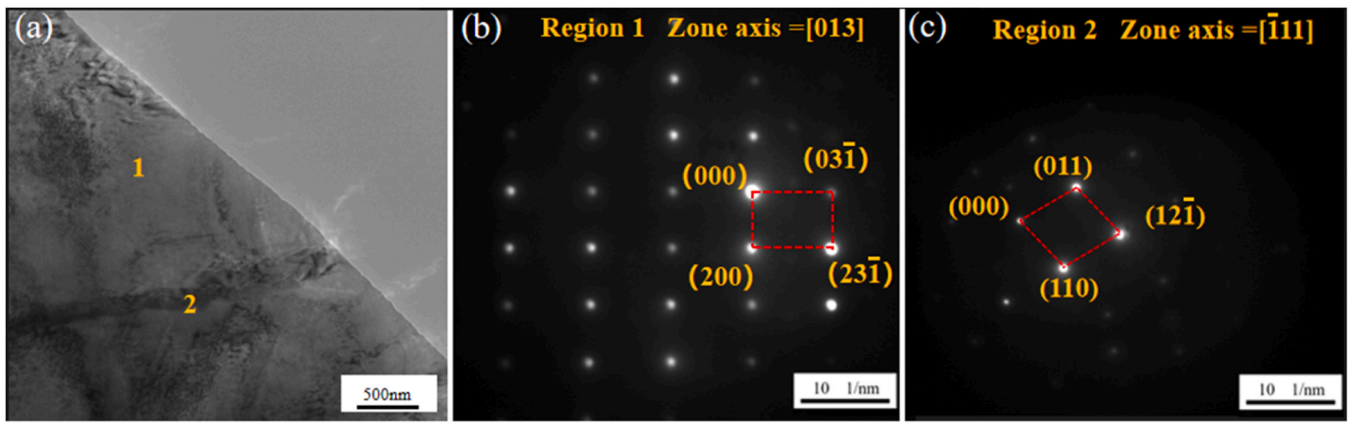


Fig. 6. TEM of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA (by DP).

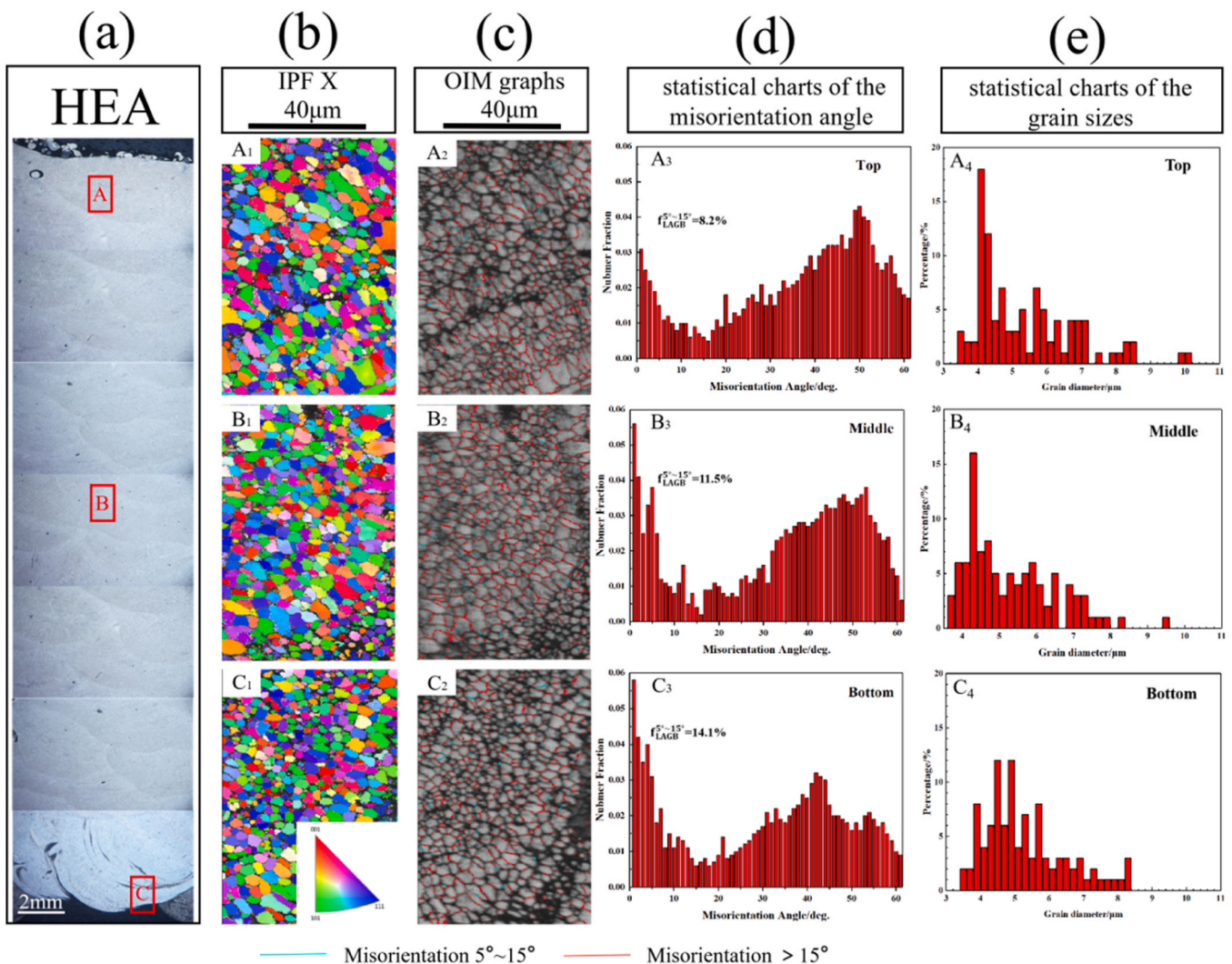


Fig. 7. (a) The spliced metallographic picture from DP-SLM, (b) EBSD, (c) OIM, (d) Statistical charts of the misorientation angle, (e) Statistical charts of the grain size.

a certain range, the material on the surface of the wear will fracture and fail, and then fall off from the wear surface, thus intensifying the wear of the surface.

These fractured scraps and pits are not conducive to the friction and wear reduction of high entropy alloys. For comparison, Figs. 10e–10f

describes the plastic deformation of the surface and sub-surface of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ at the beginning of sliding under 2 N load. Unlike the contact case, the maximum plastic strain of GA during friction is located on the surface. With a sufficiently high hardness, no considerable plastic strain is still present during the friction of DP.

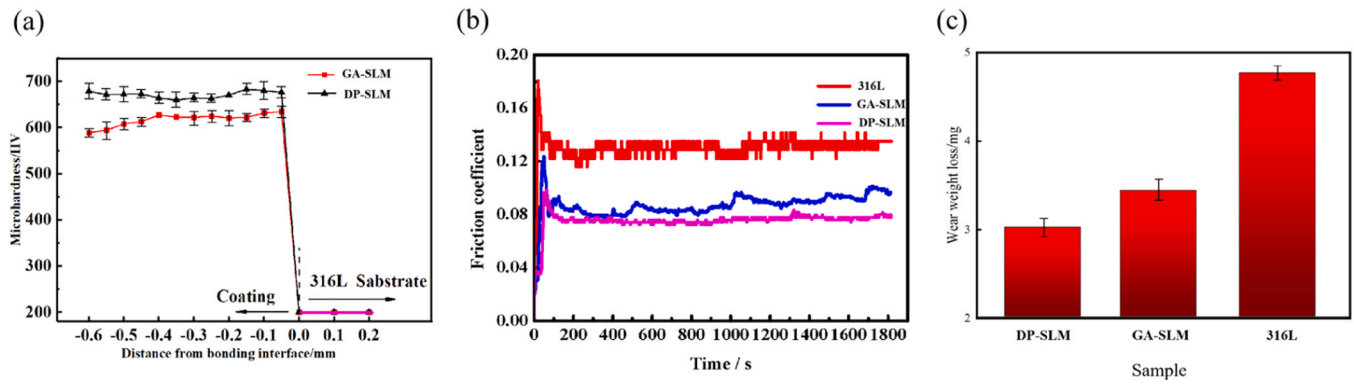


Fig. 8. (a) The microhardness; (b) Friction coefficient of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ with the time; (c) The wear weight loss of different samples.

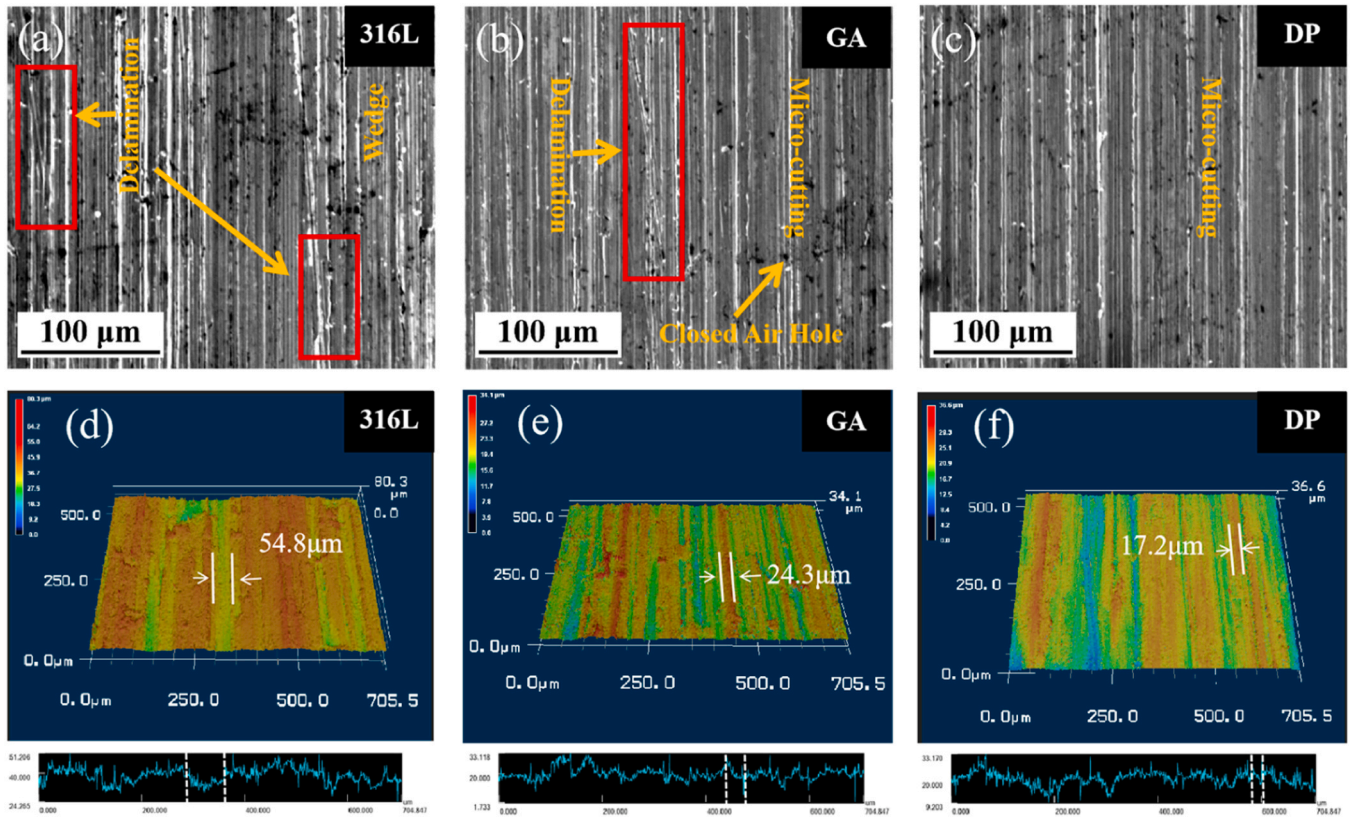


Fig. 9. (a)-(c) The worn samples tested by SEM, (d)-(f) The three-dimensional morphology of the wear sample.

As friction progresses, the surface of the ball becomes more uneven, thus the simulation results only reflect this [43]. The milling ball's surface roughness increases with the progression of friction. At this time, the reciprocating movement of the micro-convex body in the wear marks produces furrows, destroying the initial appearance of the wear marks. The likelihood of plastic distortion on the HEA with greater hardness is reduced when it is exposed to a smooth micro-convex body, thus resulting in a thinner width of the SLM in the wear marks than that of a HEA alloy with low hardness [44].

To verify the simulation results, Fig. 11 shows enlarged images of GA and DP wear marks under different loads. Furrows appeared on the surface of GA-SLM by enlarging, and the width of furrows increased significantly. These furrows are caused by the reciprocating movement of oxide debris [45]. Under 2 N load, GA-SLM plastic deformation is observed in Fig. 11a. However, this plastic deformation is mainly at the start of friction and has little bearing on the wear outcome. Subsequent

to the surface plastic deformation, the contact area between the pressure drop zone and the ball increases, and the contact pressure consequently diminishes. Subsequently, plastic distortion will result in the fortification of the surface handling, and the fortified surface will also be able to withstand plastic distortion. Under 2 N load, the abrasive surface of DP-SLM shows little delamination with micro-cutting mechanism, which proves that the wear degree is low.

When the load reaches up to 15 N, the wear mechanism of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ and GCr15 was still mainly abrasive wear, and the oxides were mainly distributed in the furrow. In addition to obvious abrasive wear, the wear mechanism of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ also includes fatigue wear. Fig. 11c shows that grid fatigue cracks appear in the GA region during the initial wear phase (The test temperature is 800°C). Cracks occur in the process of reciprocating friction, expanding along the direction perpendicular to the friction. With the expansion of cracks, the granular state of the surface is gradually shed, and the shed chips

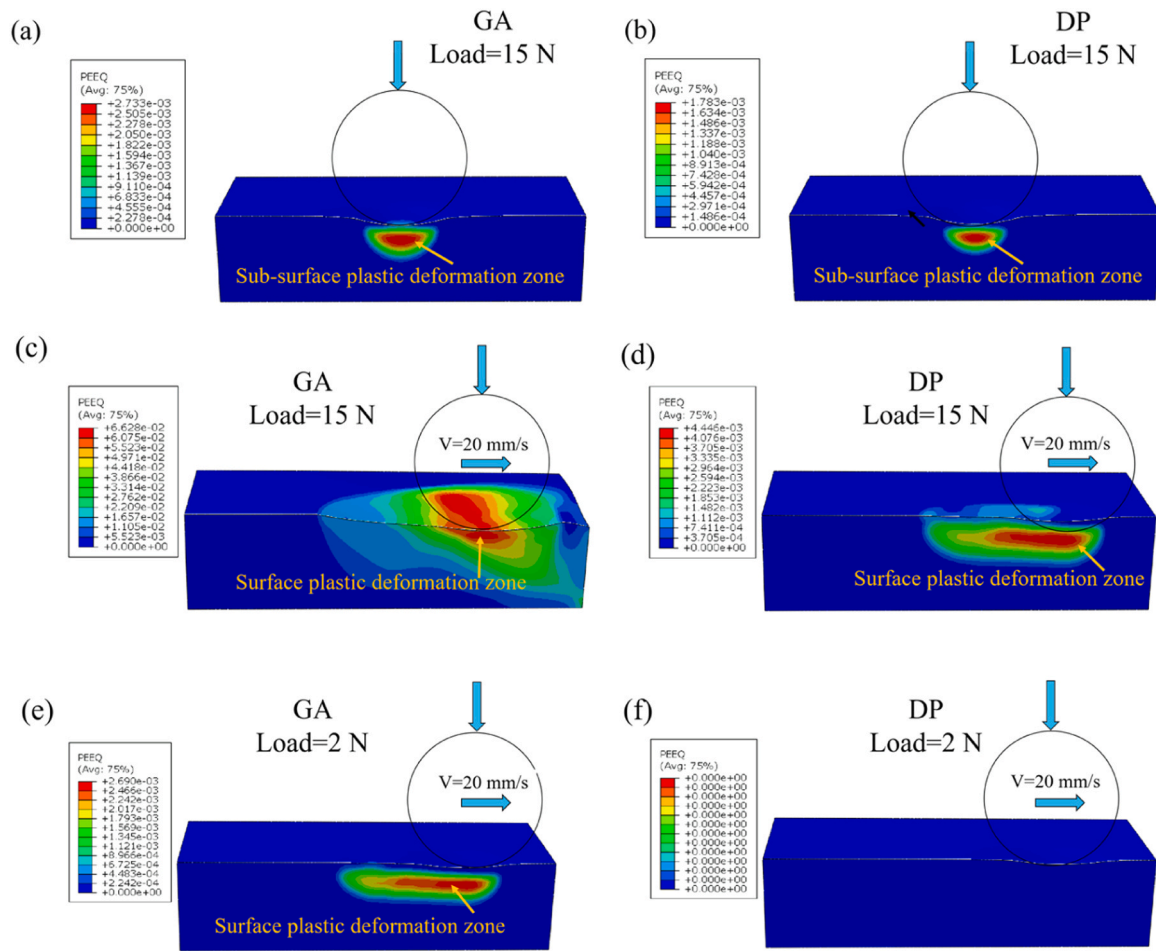


Fig. 10. (a)-(b) PEEQ of alloy section in contact process under 15 N load; (c)-(d) The stress distribution of von Mises stress (Load of 15 N, Velocity of 20 mm/s); (e)-(f) The stress distribution of von Mises stress (Load of 2 N, Velocity of 20 mm/s).

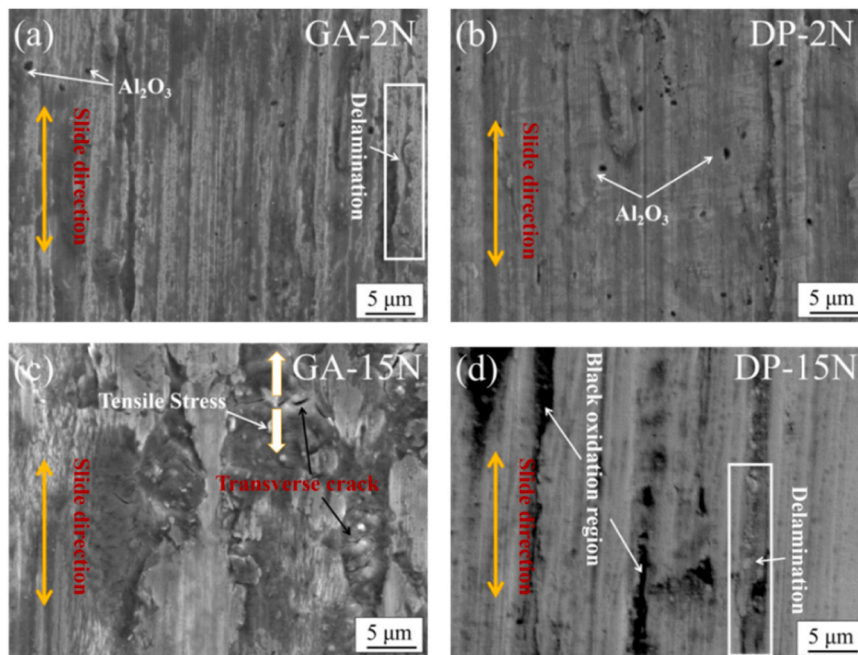


Fig. 11. Abrasion characteristics of Nb₃Cr₂Ti_{0.9}W_{0.4}Fe₆ SLM (GA and DP process) and GCr15 ball under different load.

accumulate in the pits and are gradually oxidized. As shown in Fig. 11d, the black region of DP is the oxide film. The pits are caused by the repeated friction of GCr15 pellets on the surface of the high entropy alloy. The ball surface is very rough during actual friction process. When the micro-convex is sharp, the surface of the high entropy alloy will have serious plastic deformation and form furrows. There are oxides formed by plastic deformation in furrows. When the micro-convex body is relatively smooth, the DP surface will undergo plastic deformation within a certain width, thus forming an oxide film. When the pressing amount of the micro convex is small, the DP surface will have no plastic deformation, only causing the passivation film wear. The DP-SLM exhibits excellent tribological properties due to the combined action of oxides in furrows, oxide film in wear marks and regenerated passivation film.

3.6. Mechanical properties

In HEAs, the uneven distribution of chemical components causes the edge dislocation to be subjected to strong pinning forces at the nano-scale. Instead of moving in a smooth, fast way, it moves in a slow, discontinuous way. In addition, the fluctuation of chemical composition will promote spontaneous nucleation of spiral dislocation twists [49]. This has a significant pinning effect on the kink expansion, resulting in the speed of the spiral dislocation being limited not by the kink nucleation rate but by the kink expansion rate. Therefore, the motion of blade and screw dislocations in BCC HEAs is dominated by special activation behavior. The two kinds of dislocations need to overcome similar local barriers and show similar dynamic characteristics, being significantly different from traditional BCC metals.

In the previous study, Zhen et al. developed $\text{CoCr}_{2.5}\text{FeNi}_2\text{TiW}_{0.5}$ high entropy alloy additive manufacturing forming parts [50]. Its yield strength (YS) was 581 MPa, ultimate tensile strength (UTS) reached up to 893 MPa and total elongation (ϵ) can be obtained to 9.9 %. In this paper, the authors have also tested the mechanical properties of newly designed HEA printed parts under two kinds of process (DP and GA). (See Fig. 12). The results are shown in Table 5. The mean extended YS and UTS of DP-SLM are slightly lower than GA-SLM. The yield strength (YS) of DP-SLM is 953 MPa and UTS is 1159 MPa. As for GA-SLM samples, their YS and UTS are 984 MPa and 1176 MPa, respectively. The average ϵ of DP-SLM (13.4 %) is 25 % higher than that of the GA-SLM (10.7 %). According to the previous analysis, the improvement of mechanical properties of high entropy alloys is related to the fine crystal strengthening mechanism [46,50].

As described in Fig. 4, $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA-SLM structures are mainly composed of equiaxial crystals (both in GA and DP method). The increase of grain boundary region is often accompanied by the increase of nucleation rate of grain boundary precipitates and the decrease of grain boundary precipitates size. The high entropy alloy with BCC as the

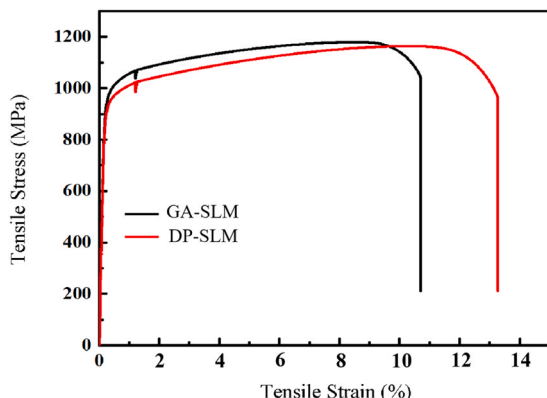


Fig. 12. The mechanical properties of GA-SLM and DP-SLM alloy.

Table 5

Mechanical properties of different HEA alloys at room temperature.

Name	Refs.	σ_y / MPa	σ_{UTS} / MPa	ϵ / %
$\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$	This work (GA)	984	1176	10.7
$\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$	This work (DP)	953	1159	13.4
$\text{CoCr}_{2.5}\text{FeNi}_2\text{TiW}_{0.5}$	[50]	581	893	9.9
$\text{CoCrFeNiNb}_{0.412}$	[21]	637	1004	1.3
$\text{CoCrFeNiNb}_{0.155}$	[21]	321	744	23.3
$\text{CrFeNi}_{2.2}\text{Al}_{0.8}$	[53]	479	956	12.7
$\text{CrFeNi}_{2.2}\text{Al}_{0.9}$	[53]	610	1173	9.1
$\text{Al}_{16}\text{Co}_{21}\text{Cr}_{21}\text{Fe}_{21}\text{Ni}_{21}$	[54]	1207	1398	1
$\text{Al}_{13}\text{Co}_{21.74}\text{Cr}_{21.74}\text{Fe}_{21.74}\text{Ni}_{21.74}$	[54]	412	932	21
$\text{Al}_{13}\text{Co}_{21}\text{Cr}_{21}\text{Fe}_{21}\text{Ni}_{21}\text{Ti}_3$	[54]	1098	1415	5

main structure is easy to cause internal defects to hinder the dislocation slip and make the alloy material plastic deformation to increase the strength. At the same time, plasticity and toughness are greatly reduced. Due to the synergistic effect of solid solution strengthening and lower valence electron concentration, RHEAs (Nb, Ti, W) often tend to exhibit comparable strength and plasticity with BCC structure [49]. Therefore, the HEAs with BCC structure can also achieve the synergistic increase of the strength and toughness by reasonable design.

In Fig. 13, it is believed that our present work $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ (DP) has shown high YS and UTS with appreciable ductility when compared to other HEAs [21,52–54]. Compared previous studies in Table 5, the newly designed $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA is a single phase BCC solid solution. XRD and TEM showed that no intermetallic compounds were precipitated. This reduces the brittleness of the material. In addition, we have made innovations in process methods. Through the high-temperature powder remelting, the composition homogenization reduces the deformation resistance, promotes the plastic improvement, and makes the plasticity and strength of the material increase synergically.

The micromorphologies have been scanned prepared by different processes in Fig. 14. It is obvious that GA-SLM has cracks (Fig. 13a, red dash circle). Since the spherical powder particle composition is not uniform, the forming part will have intercrystalline segregation, affecting the high temperature service performance. This eventually cause a decline in the mechanical properties of the HEA alloy [48], resulting in cracks in the formed parts. For DP-SLM, the resulting crystals are equiaxial (Fig. 13b). The fracture morphology is shown in Fig. 13c (GA-SLM) and Fig. 13d (DP-SLM), respectively. High-density tear pits were found at the fracture of SLM samples (Fig. 13d). It is evaluated as a typical ductile fracture. Pores can be observed on the surface of Fig. 13c (yellow dash circle). This is because the laser power is small or the scanning speed is too fast, resulting in insufficient energy that makes the high melting point refractory element W insufficiently melted. The spherical particles obtained by GA method have the possibility of hollow balls (a problem that has always existed in this technology). This makes it possible to have trapped gases in the powder bed during printing. These gases fail to escape in time during the melting process and form small pores during rapid solidification. In the process of crystal growth, the external force acts, which makes the pores become larger and form holes.

Due to the increase of material the height of the liquid metal in the process of producing too cold. The gas is discharged from the molten pool in the fast melting and fast condensing process. As the viscosity of the molten liquid decreases, the liquid metal is hard to enter the pores, forming the pores.

3.7. True stress-strain curves in high temperature

The rheological behavior of superalloys can be described by the macroscopic phenomenological constitutive equation of Arrhenius model [47]. It is considered that the thermal deformation behavior of

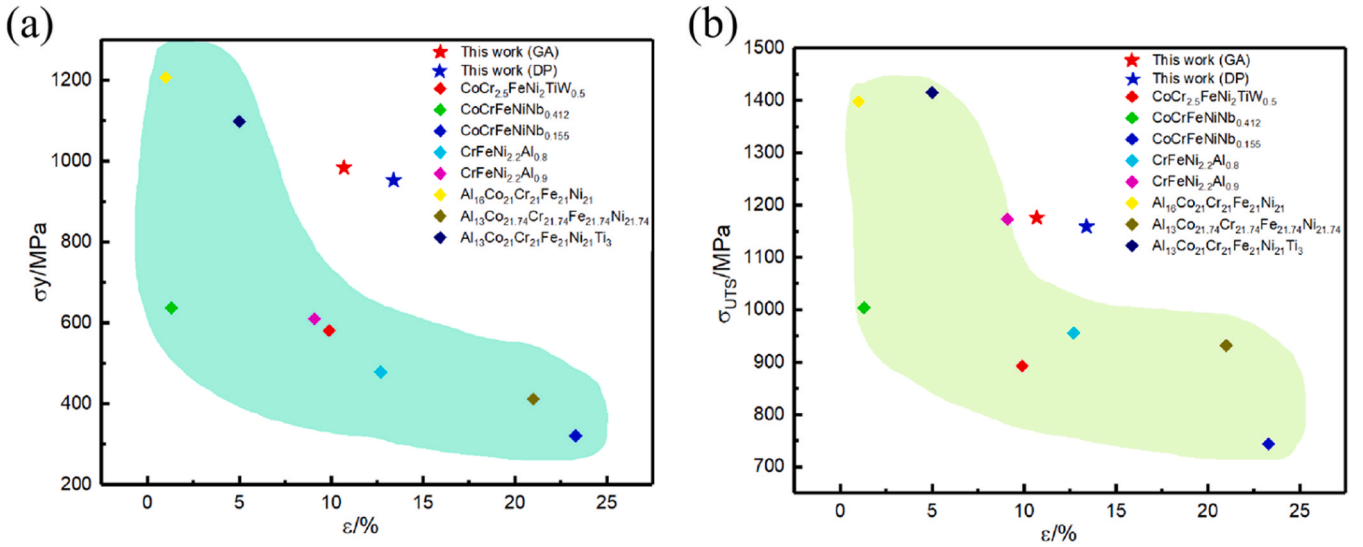


Fig. 13. Summary of reported mechanical performances of HEAs [21,52–54] and collected data in this work.

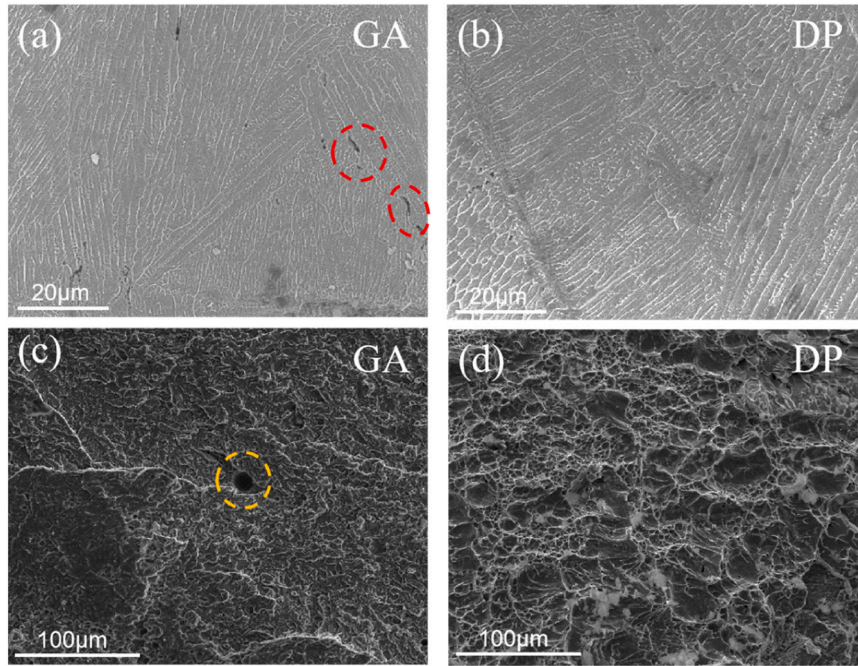


Fig. 14. The fracture surfaces by different manufacturing processes: (a) and (c) GA-SLM, (b) and (d) DP-SLM.

superalloys is a thermal activation process [48]. The rheological behavior of the alloy at high temperature is described by the relation between peak stress σ and deformation temperature T and deformation rate $\dot{\epsilon}$. Formula (1) [47] is applicable to describe low stress conditions ($\alpha\sigma < 0.8$), formula (2) is applicable to describe high stress conditions ($\alpha\sigma > 1.2$), and formula (3) is applicable to any stress conditions [47].

$$\dot{\epsilon} = A_1 \sigma^{n_1} \exp[-Q/(RT)] \quad (1)$$

$$\dot{\epsilon} = A_1 \exp(\beta\sigma) \exp[-Q/(RT)] \quad (2)$$

$$\dot{\epsilon} = A[\sinh(\alpha\sigma)]^n \exp[-Q/(RT)] \quad (3)$$

Where, R is the ideal gas constant, Q is the thermal deformation activation energy of the alloy, where A_1 , A_2 , A , n , n_1 , α and β are related constants, and $\alpha = \beta/n_1$.

Taking logarithms of both sides of (1), (2) and (3) respectively, the following equation can be obtained:

$$\ln \dot{\epsilon} = \ln A_1 - Q/RT + n_1 \ln \sigma \quad (4)$$

$$\ln \dot{\epsilon} = \ln A_2 - Q/RT + \beta\sigma \quad (5)$$

$$\ln \dot{\epsilon} = \ln A - Q/RT + n \ln[\sinh(\alpha\sigma)] \quad (6)$$

From formulas (4) and (5) n_1 and β are obtained according to the linear relationship between $\ln\sigma_p - \ln\dot{\epsilon}$ and $\sigma_p - \ln\dot{\epsilon}$, $n_1 = 4.175128$, $\beta = 0.03533$, and $\alpha = \beta/n_1 = 0.006105$. The above results are brought into formula (6), and the n value is calculated as 4.166304 by linear regression. At the same time, R is the ideal gas constant with the value of $8.314 \text{ J}\cdot\text{mol}^{-1}\cdot\text{K}^{-1}$. The $1000/T - \ln[\sinh(\alpha\sigma)]$ relation gives the value of Q to be 464.712 kJ/mol . Fig. 15 represents the relationship between peak stress of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ and strain rate and temperature.

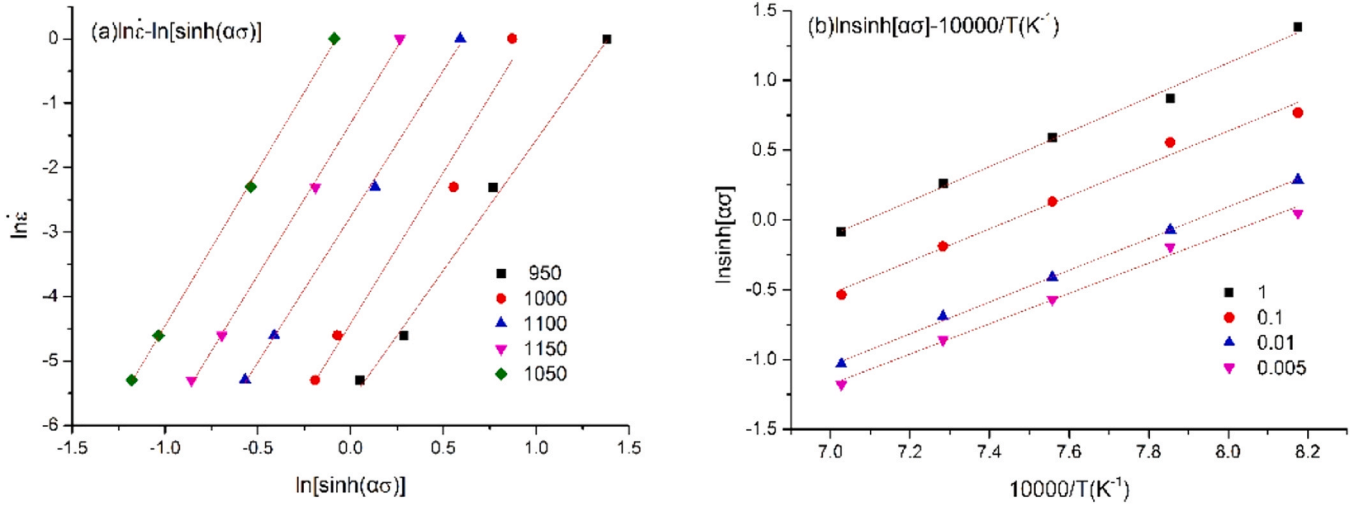


Fig. 15. Relationship between peak stress of strain rate and temperature (a) $\ln \dot{\epsilon} - \ln[\sinh(\alpha\sigma)]$; (b) $\ln[\sinh(\alpha\sigma)] - 10000/T$; .

Zener and Hollomon use the parameter Z to represent the coefficient between the deformation temperature and the deformation rate [5, 46]:

$$Z = \dot{\epsilon} \exp[Q/(RT)] = A[\sinh h(\alpha\sigma)]^n \quad (7)$$

The formula is deformed as follows:

$$\ln Z = \ln A + n \ln[\sinh h(\alpha\sigma)] \quad (8)$$

The value of $\ln A$ can be obtained from the intercept of the $\ln[\sinh h(\alpha\sigma)] - \ln Z$ curve, Then $\ln A = 48.9833$ and $A = 4.00377 \times 10^{16}$ (Seen in Fig. 16)

In summary, the constitutive equation of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ is as follows:

$$\dot{\epsilon} = 4.00377 \times 10^{16} [\sinh h(0.00615\sigma)]^{4.175128} \exp[463664/(RT)] \quad (9)$$

For better research on $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA DP-SLM in high temperature, Fig. 17 shows the true stress-strain curves at the experimental temperature of 1000 ~ 1150 °C along with the strain rates of 0.005 s⁻¹, 0.01 s⁻¹, 0.1 s⁻¹ and 1 s⁻¹, respectively. In the initial stage of hot compression deformation, the dynamic recovery of the alloy cannot completely eliminate the work hardening caused by hot compression. The cumulative distortion energy inside the grain enhances with the

increasing strain [46]. The work hardening plays a dominant role in the deformation process. The DP-SLM recovers when the deformation temperature is 1000 °C and the deformation rate stays at 0.01 s⁻¹. The softening and work hardening induced by this process reach a dynamic equilibrium state, and the stress-strain curve enters a constant state for stable deformation.

The increase of the stress variable will induce the generation of dynamic recrystallization, thus causing softening. Dynamic recrystallization acts as the main softening mechanism during the hot compression of super-alloys. When the stress-strain curve reaches the equilibrium state, the peak value of the curve appears, such as deformation temperature of 1100 °C and 1150 °C with deformation rate of 0.01 s⁻¹. The softening caused by dynamic recrystallization plays a leading role in the deformation mechanism. After the peak value, the stress decreases with the increasing strain. When the softening caused by work hardening and dynamic recrystallization reached dynamic equilibrium, the stress-strain curve showed a stable plateau.

3.8. Hot working diagram of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$

In the hot working process of a metal, workability indicates the ability of a metal to undergo plastic deformation without damage. It is an important index to judge the forming properties of metals. There are two main factors affecting machinability. On the one hand, it is determined by the inherent properties of the material, such as chemical composition and organizational form. The other part is determined by processing parameters, such as strain rate, deformation temperature, stress state, deformation amount, etc. [51]. Hot working diagram can objectively describe the high temperature deformation behavior of materials, including the microscopic deformation mechanism and microstructure evolution of materials in the process of high temperature deformation, and is often used in the prediction and optimization of metal hot working technology [52]. When the material is thermal deformed, the total power P consumed is composed of the consumption G generated by plastic deformation and the dissipation covariance J caused by the microstructure evolution, which can be expressed by the formula (10) as follows:

$$P = \sigma \cdot \dot{\epsilon} = G + J = \int_0^{\dot{\epsilon}} \sigma d\dot{\epsilon} + \int_0^{\sigma} \dot{\epsilon} d\sigma \quad (10)$$

The ratio of J to G is used to define the strain rate sensitivity index m , as shown in formula (11):

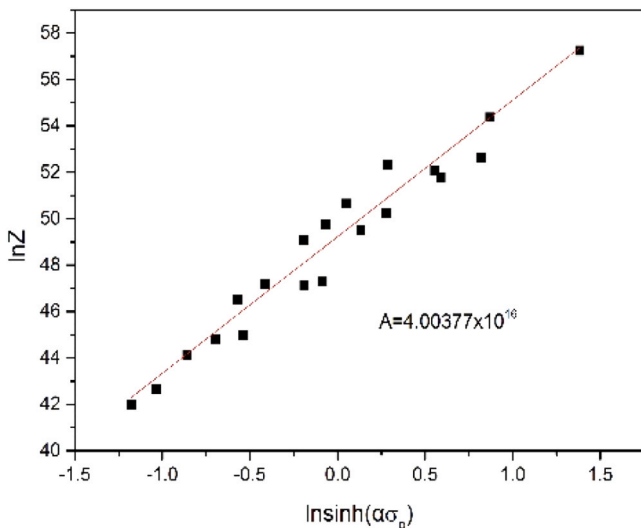


Fig. 16. Relationship between peak stress and Z -parameter.

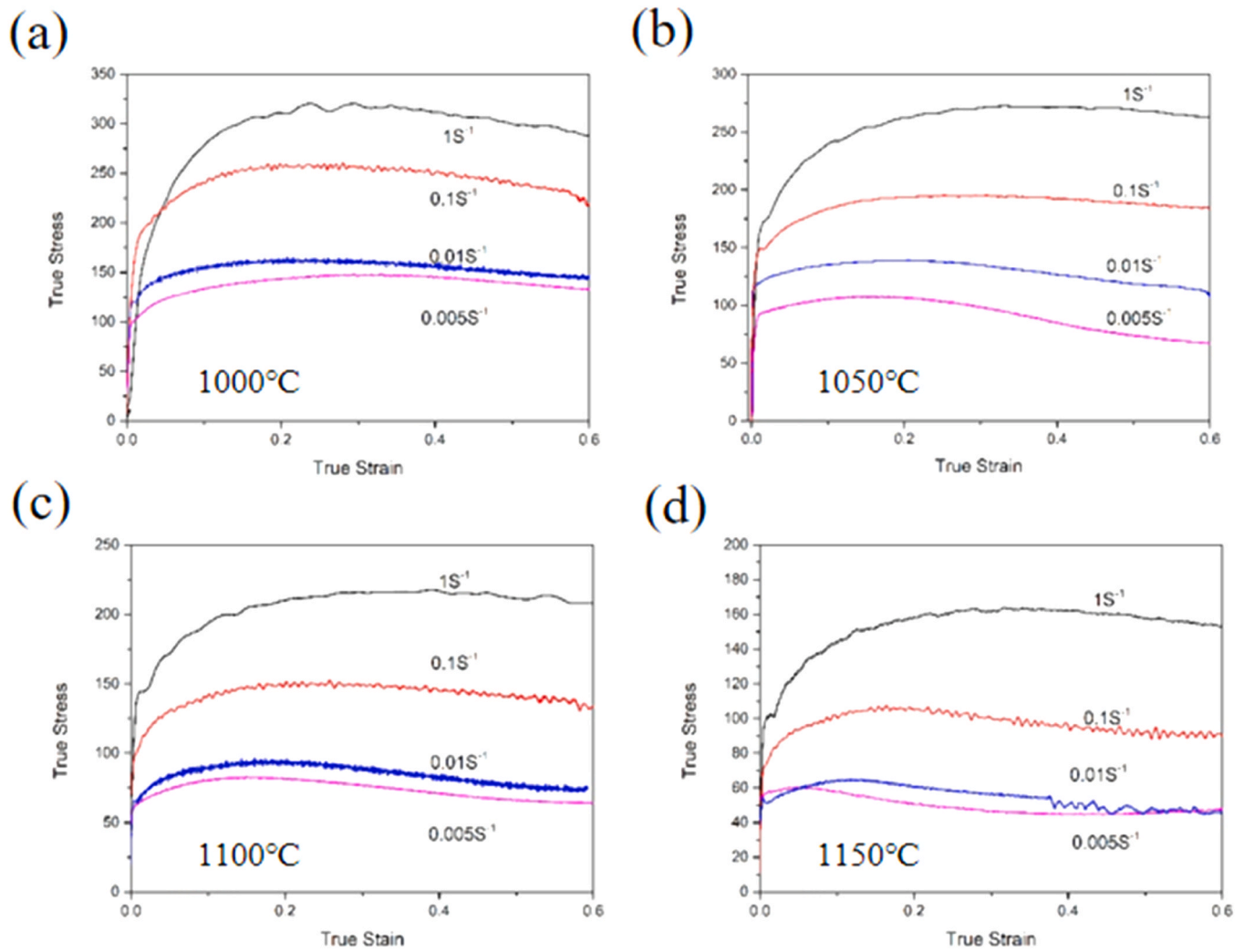


Fig. 17. True stress-strain curves of $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ at different deformation temperatures and deformation rates (a) 1000 °C; (b) 1050 °C; (c) 1100 °C; (d) 1150 °C.

$$m = \frac{dJ}{dG} = \frac{\dot{\epsilon} d\sigma}{\sigma d\dot{\epsilon}} = \left(\frac{\partial(\lg \sigma)}{\partial(\lg \dot{\epsilon})} \right)_{T, \dot{\epsilon}} \quad (11)$$

The power dissipation coefficient η is used to represent the ratio of the power dissipation covariance to the ideal linear power dissipation covariance during thermal deformation of the material (Formula 12).

The larger the dissipation coefficient η is, the more likely dynamic recrystallization will occur, which is more conducive to the improvement of the structure in the process of hot working [53]. At the same time, the rheological instability criterion is obtained according to the large strain plastic deformation maximum principle (Formula 13).

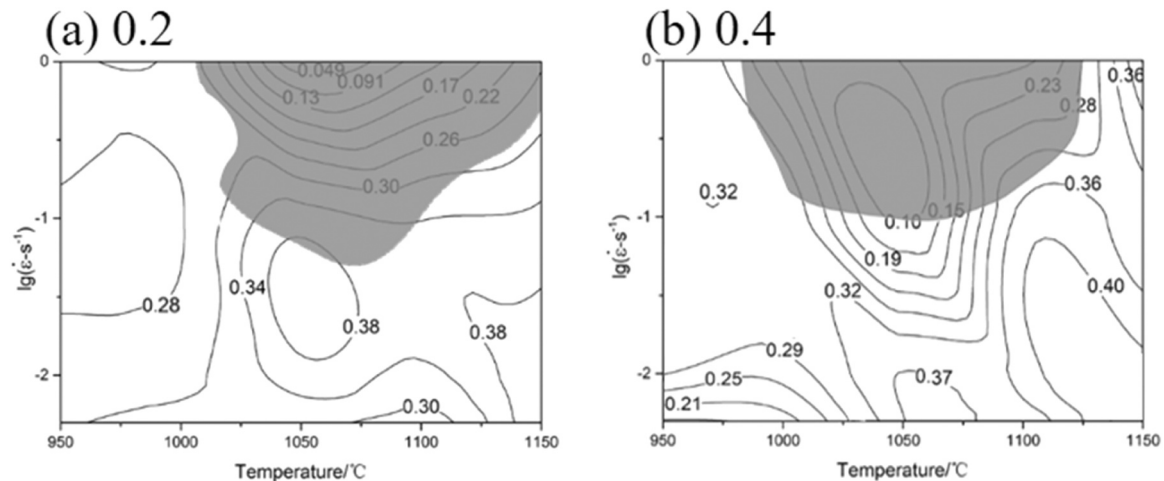


Fig. 18. Hot processing maps $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ in different strains (a) 0.2; (b) 0.4.

$\xi(\dot{\epsilon})$ is a negative region of instability. In the process of thermal deformation, instability phenomena such as local flow, adiabatic shear zone, cavity and cracking are easy to occur in the instability region.

$$\eta = \frac{J}{J_{\max}} = \frac{2 \int_0^{\sigma} \dot{\epsilon} d\sigma}{\sigma \cdot \dot{\epsilon}} = \frac{2m}{m+1} \quad (12)$$

$$\xi(\dot{\epsilon}) = -\frac{\partial \lg\left(\frac{m}{m+1}\right)}{\partial \lg \dot{\epsilon}} + m < 0 \quad (13)$$

In Fig. 18, the curve represents the isoline of the dissipation covariate η , and the numbers represent the dissipation coefficient. The shaded part in the hot working diagram is the instability region of the material. As can be seen from the figure, with the increase of material stress variables, the safe processing range of material processing is also increasing. Higher η values can be obtained by controlling deformation rate from 0.1 s^{-1} – 0.005 s^{-1} and deformation temperature from 1050 to 1150 °C. That is, the material can have a high degree of recrystallization.

When the strain rate is 1 s^{-1} and the processing temperature is 950°C (as shown in Fig. 19a), the alloy is in the deformation instability zone with η value lower than 0.12. It can be observed that grains are obviously elongated without obvious dynamic recrystallization, and the structure is mainly composed of coarse original grains [53]. Further reducing the corresponding strain rate and increasing the deformation temperature will promote the recrystallization tendency. Fig. 19b has describes the "necklace" structure caused by insufficient dynamic recrystallization, in which the small grains after recrystallization surround and intertwine the large grains. This is due to the high deformation energy at the grain boundary, so the small dynamic recrystallization grains preferentially nucleate on the original deformed grains. This phenomenon is caused by incomplete recrystallization process [54]. With the further increase of deformation temperature, a complete dynamic recrystallization grain can be obtained (Fig. 19c), and the η value reaches a peak of 0.37. Further increasing the deformation temperature and decreasing the strain rate will not cause much influence on the power dissipation (Fig. 19d).

Therefore, $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ is in the rheological instability region and η value is small when the processing rate is lower than 1050 °C and greater than 1 s^{-1} . Therefore, according to the $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$

thermal processing diagram, in order to obtain the structure with a high degree of recrystallization, it is recommended that the temperature range of the processing parameters be selected be 1050–1150 °C, and the deformation rate be selected to be 0.005 – 0.1 s^{-1} .

4. Conclusion

This paper presents a newly designed $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ HEA powder with uniform composition, prepared via a dual-process method (DP) combining mechanical alloying and high-temperature remelting spheroidization. The BCC-HEA containing refractory element can also realize the synergistic increase of strength and toughness through reasonable design. The main conclusions were made as follows:

- (1) HEA-SLM are composed of a single BCC solid solution with uniform composition. This is mainly due to the mechanical alloying technology advantage. It combined effect of mechanical milling (plastic deformation) and temperature lifting (atomic diffusion) to achieve the composition homogenization. Such technology can promote the solid alloying of immiscible systems and further expand the solid solubility of insoluble elements in immiscible matrix.
- (2) High friction coefficient can lead to considerable surface plastic deformation, thereby reducing the tribological properties of the material. According to the maximum principal stress distribution, the GCr15 ball is subjected to compressive stress at the front and tensile stress at the back during the sliding process. This proves that the stress behind the ball is similar to the stress in uniaxial tension. The cracking and falling off of GA-SLM surface with high friction coefficient are related to cyclic tensile stress.
- (3) The yield strength (YS), ultimate tensile strength (UTS) and average elongation (ϵ) samples were 953 MPa, 1159 MPa and 14.8 % respectively by the DP-SLM. An increase in the homogenization of the elemental composition can hinder the spontaneous nucleation of the helical dislocation distortion. The resistance of the dislocation during motion is reduced. This allows the dislocation to move more freely, thereby increasing the plastic deformation capacity of the alloy.

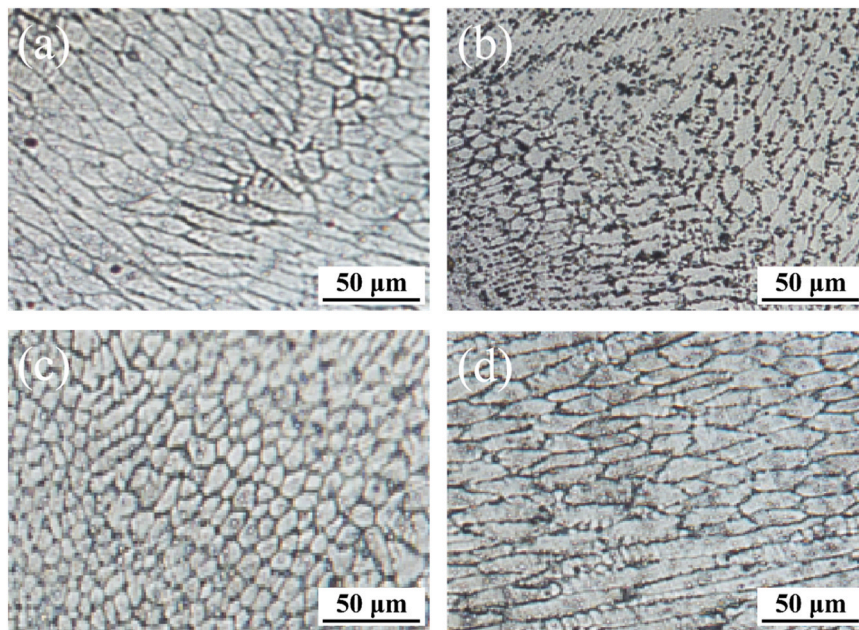


Fig. 19. microstructure of hot deformed $\text{Nb}_3\text{Cr}_2\text{Ti}_{0.9}\text{W}_{0.4}\text{Fe}_6$ in different conditions: (a) 950°C , 1 s^{-1} ; (b) 1000°C , 0.1 s^{-1} ; (c) 1100°C , 0.1 s^{-1} ; (d) 1150°C , 0.005 s^{-1} ; .

CRediT authorship contribution statement

Wang Jingqi: Investigation. **Lin Jiahao:** Investigation. **Gu Zhen:** Writing – original draft, Funding acquisition, Data curation. **Li Jinting:** Data curation. **Yuan Hongjie:** Investigation. **Zhang Shiwei:** Methodology. **Fan Chuanwei:** Software. **Shan Zhiwei:** Writing – review & editing, Visualization. **Luo Jun:** Visualization, Supervision. **Xi Shengqi:** Writing – review & editing, Supervision.

Declaration of Competing Interest

The authors declared that they have no conflicts of interest to this work. We declare that we have no financial and personal relationships with other people or organizations that can inappropriately influence our work, there is no professional or other personal interest of any nature or kind in any product, service and/or company that could be construed as influencing the position presented in, or the review of, the manuscript entitled.

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