Effect of particle size on the microstructure and tensile property of the Al-28.5Si alloy prepared by continuous powder extrusion

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Abstract
Continuous powder extrusion, a new powder consolidation technology, is used to process alloy powders that are difficult to form. In the paper, Al-28.5Si alloys were successfully prepared by continuous powder extrusion, and the influences of particle size on the microstructure and tensile property have been investigated. In the as-extruded rods, the primary Si phase is dispersed in the α-Al matrix. As the powder size reduces from more than 75 μm to less than 25 μm, the equivalent diameter of the primary Si particles consecutively decreases, but the shape factor increases. Comparing the Si particles in the alloy rod and the powder, it is found that the eutectic Si phase disappears during continuous extrusion. Primary Si particles change from coarse and irregular to round. Specifically, the ultimate tensile strength and elongation at break of the powder-extruded alloy further increase from 211 MPa to 266 MPa and 0.99% to 2.21%. Moreover, the fracture mechanism of alloy rods is the brittle fracture.

1. Introduction
Continuous powder extrusion is a new technology developed on the basis of extrusion technology of metal billet and polymer processing, which can obtain fine grain with uniform composition, hence avoid coarse grain and composition segregation easily appeared in melting casting processing [1]. It has low deformation resistance and high yield compared with melting casting processing, so it is widely used to process soft alloy [2]. The technical principle is that a squeeze groove wheel moves uninterruptedly, the powder moves to the upsetting zone under the driving of the friction force, forms a rod under the continuous extrusion deformation force generated by the rotational movement of the squeezing wheel in contact with the powder [3]. The process is continuous, utilizes standard equipment and does not require powder preheating or inert gas shrouding providing a footing for a true cost reduction in the product [4]. Continuous extrusion is a relatively complicated deformation process, during which rolling, upsetting, right-angle bending and extrusion deformation can occur. Both friction and plastic deformation during extrusion help to create high temperatures and pressures in the material. This leads to cold welding and consolidation of the powder particles [5]. For traditional hot extrusion, the preheating of the powder and the friction between the powder and the extrusion cylinder will consume a lot of energy. The coarsening of Si particles also occurred during the hot extrusion of powder. The extrusion temperature produced a considerable effect on the average diameter of Si particles, the average diameter increased with increasing extrusion temperature [6]. However, in the continuous extrusion process, the friction force that acts as useless work in traditional hot extrusion is ingeniously converted into a driving force and heating source during deformation, which saves energy greatly and improves material utilization. Furthermore, severe shear deformation that occurs during the process can contribute to fine microstructure and homogenous distribution.
of particles [7]. Continuous powder extrusion involves a sharp shear deformation, which can greatly refine grain size and eliminate the edges and corners of coarse particle, so as to improve the mechanical properties of the alloy.

Hypereutectic Al-Si alloy, as an electronic packaging material, is prospective because of the super performance of the weight, wear, corrosion and thermal behavior [8, 9]. The good comprehensive property of hypereutectic Al-Si alloy mainly comes from the excellent property of the Al matrix and Si phase, which can greatly enhance the mechanical and wear property, especially the thermal behavior [10]. Si can not only improve the strength and wear resistance of the alloy, but also reduce the thermal expansion coefficients (CTE) and density. The reason why the CTE of Al alloy can be reduced by adding Si is given by Zhang et al [11]. When the alloy is heated, compressive stress develops in Al and tensile stress develops in Si because of the mismatched CTEs of the two phases. As a result, the expansion of Al is restraint by Si. Si improves the properties of the alloy mainly by means of the supersolld strengthening of Si and the dispersion strengthening of the matrix formed by the formation of fine primary Si particle. Si has a high solid solubility in the Al matrix at high temperature, the solubility of Si in Al and the absence of interfacial reactions at high temperatures promote the consolidation process of the Al-Si alloys [12]. Additional, the high interface bonding strength between the Al matrix and Si particles is superior for good wettability. High silicon aluminum alloy has a series of excellent properties, however, which are tightly associated with the eutectic Si phase and primary Si phase. Such as traditional casting technology, which would lead to the primary Si phase and eutectic Si phase coarse and inhomogeneous, in turn cause poor mechanical properties [13]. Jiang et al demonstrated that the severe macrosegregation of the primary Si phase was mainly caused by fluid flow and temperature distribution in the hypereutectic Al-Si alloy [14]. Continuous powder extrusion eliminates the influence of fluid flow and temperature distribution. Cui et al reported that the primary Si phase will fracture and coarsen during the extrusion process [15]. The preferential precipitation of primary Si phase and subsequent epitaxial growth of the eutectic Si phase on it could contribute to the modification of the eutectic Si phase in the hypereutectic Al-Si alloy [16]. Zhao et al found that extrusion can eliminate coarse and acicular eutectic Si phase, so achieve an improvement of the mechanical performance [17].

The deformation of rapidly solidified hypereutectic Al-Si alloy powder is extremely difficult, which is significantly affected by the Al matrix and Si particle, especially the distribution, type and size of Si phase. Previous literature indicates that the deformation capacity is inhibited by the inhomogeneous distribution of the Si phase and the presence of dendritic eutectic phase [18]. So cause low densification of alloy powder. Gas-atomized Al-27Si alloy powder consists of α-Al phase, bulk β-Si and dendritic eutectic Si phase. With the decrease of particle size, the amount and size of bulk β-Si phase and dendrite eutectic Si phase decrease, and the bulk β-Si phase aggregates at the edge of powder particles [19]. Therefore, the forming of high silicon aluminum alloy powder is still a severe challenge.

Under the consideration of production cost and continuous preparation, continuous extrusion can improve the densification and mechanical properties of aluminum alloy products. In this paper, Al-28.5Si alloy rods were fabricated by continuous powder extrusion. Morphology and microstructure of atomized powder and alloy rods, microstructure evolution during extrusion and tensile property of the alloy rods are reported.

2. Materials and methods

In this experiment, Al-28.5Si alloy atomized powder was purchased from Tianjin Baienwei New Material Technology Co., Ltd.

The alloy powder obtained by standard sieving is divided into four different particle sizes, i.e., more than 75 μm, 75–45 μm, 45–25 μm and less than 25 μm. The extrusion die was heated to 400 °C for 30 min. There was no facility to preheat the wheel groove and it was left at room temperature during the process startup. The extrusion test was carried out on LJ350 continuous extruder. The powder was loaded into a hopper without preheating or inert gas shrouding and fed into the wheel groove. The extrusion resistance varies with the temperature of groove in different time. Therefore, The wheel speeds were chosen to cover the safe range of operating speeds of the machine. Under the friction between the powder and the wheel groove, the powder is sent to the die and extruded into a rod with a diameter of 8 mm. Afterwards, The extrusion rod is cooled in air. The process parameters of continuous extrusion were listed in table 1.

Alloy powder was mixed into pure Sn powder through a Mixing machine with a volume ratio of 1:3 and was compacted using a Tablet machine with three-stage of 10 MPa, 15 MPa and 20 MPa, holding time is 5 min, 5 min and 10 min. The microstructures of the powder samples, embedded in pure Sn powder, and extruded billets were researched by a VEGA3 TESCAN scanning electron microscope after etched with 5%HF (volume fraction) solution. Another etching test was performed on the powders and extruded alloys to study the
morphology of Si particles. The etchant solution is 50% HCl (volume fraction), and the etchant temperature is 323 K.

The tensile tests of the samples were conducted on a SHIMZU AG-X 100KN universal testing machine at a strain rate of \(1.3 \times 10^{-4} \text{ s}^{-1}\), and the extensometer was used. The fractures were examined using scanning electron microscope. The size of the Si particle in the powder and alloy rods was measured by the Image-Pro Plus analysis software. The cross-sections of alloy rods with the same particle size were measured for multiple Si particles (more than 100 particles) and then averaged.

3. Results and discussion

3.1. Morphology of powder

The morphology of powders with different particle sizes is shown in figure 1. The powder morphology is greatly affected by particle size. When the powder size is more than 75 \(\mu m\), as shown in figure 1(a), the powder morphology is irregular. There are many burrs, pits and satellite particles on the surface. As the particle size gradually decreases to less than 25 \(\mu m\), as shown in figure 1(d), the powder morphology is smoother and rounder, and particularly the burrs, pits and satellite particles are gradually eliminated as well.

The flatness of the surface of the powders is mainly determined by the rate of shrinkage during solidification. In the stage of atomization, the cooling rate of the large-size droplet is slow. Hence it shrinks violently during solidification resulting in an uneven surface. Small-size droplets solidified preferentially collide with large-size droplets during flight, and some adhere to the surface of large-size droplets to form satellite particles. Others

![Figure 1](image-url)

**Figure 1.** Different morphology and sizes of the powders with different particle sizes: (a) >75 \(\mu m\), (b) 75 – 45 \(\mu m\), (c) 45 – 25 \(\mu m\), (d) <25 \(\mu m\).
form burrs by friction with large-size droplets. However, small-size droplets with a fast cooling rate shrink less during solidification, hence form a smooth and nearly spherical surface.

For the large-size droplets, the cooling rate of the surface is fast, while the cooling rate of the center is slow. Therefore, the internal and external crystallization rates are different, which leads to the rough surface of the powder after solidification, and becomes rod-shaped and other irregular shapes. The formation of the spherical small-size powder is due to the high specific surface area, which causes the droplet as a whole to solidify rapidly. The shrinkage during solidification is small. Therefore, the surface of the powder is smooth.

The large-size droplets during the flight are impacted by the atomizing gas ejected at high speed, so the local surface is twisted and deformed. The form is maintained because of rapid solidification and thus the formation of burrs. The small-size droplets are rotated by the impact of high-speed gas, and the forces are relatively uniform in all directions. As a result, no burrs appear.

3.2. Microstructures of powders

The microstructures of alloy powder are exhibited in figure 2. Apparently, the primary Si particles are distributed non-uniformly in a separate powder, and the primary Si particles with sharp edges and corners are massive and polygonal. As a result, the stress is mainly concentrated at the sharp edges and corners, causing internal microcracks. However, the eutectic Si is uniformly distributed in the α-Al matrix.

Equivalent diameter is employed to evaluate the size of the primary Si phase, and it is defined as

\[ d = \sqrt{\frac{4A}{\pi}} \]  

Figure 2. Microstructures and distribution of the powders and eutectic Si phase: (a) >75 μm, (b) 75−45 μm, (c) 45−25 μm, (d) <25 μm.
In which A is the area of primary Si. In addition, the roundness of the particles is represented by the shape factor, which given by [20]:

\[
\text{Shape factor} = \frac{4\pi A}{p^2} \in [0, 1]
\]  

(3-2)

Where P is the perimeter of primary Si, 1 represents a perfect circle and 0 represents a line.

The equivalent diameter and shape factor of the primary Si particles of the alloy powder is shown in figure 3. As the powder size decreases, the equivalent diameter of the primary Si particles in power decreases, but the

Figure 3. Equivalent diameter and shape factor of Primary Si particles of alloy powders with different particle sizes.

Figure 4. Microstructure of alloy rods continuously extruded with powder of different particle sizes: (a) >75 μm, (b) 75~45 μm, (c) 45~25 μm, (d) <25 μm.
shape factor of the primary Si phase increases. When the particle size of atomized powder decreases from more than 75 μm to less than 25 μm, the equivalent diameter of primary Si particles decreases from 5.56 μm to 2.53 μm, shape factor increases from 0.73 to 0.85. The result indicates that the size of the primary Si is proportional to the droplet size and is closer to the sphere when decreasing. Decreasing the particle size not only resulted in decreasing the size of the Si phase, but also the evolution of morphology and distribution of the primary and eutectic Si phase. The solidification rate increased significantly with decreasing the particle size and resulted in the deviation of the microstructural characteristics [21].

3.3. Microstructure of alloy rods
The microstructure of the alloy rods continuously extruded with powder of different particle sizes is exhibited in figure 4. The primary Si particles are dispersed in the α-Al matrix, the edge and corner of primary Si particles are not obvious, the shape of the primary Si particles is round, it indicates that the edge and corner of primary Si particles are passivated during the continuous extrusion process. However, no eutectic Si in the alloy rods appears. Si phase usually enhances passivation on edges and corner during friction and can reduce the stress concentration, thereby increasing the strength between the Si and the matrix and improving the mechanical property [22].

Microstructure image of alloy rods by continuous extrusion with powder of different sizes is selected, and use Image-Pro Plus image processing software to separately calculate the equivalent diameter and shape factor of the primary Si particles.

The equivalent diameter and shape factor of primary Si particles with different particle sizes are shown in figure 5, which noted that the equivalent diameter of primary Si particles decreases as the powder size decreases. When the powder size is larger than 75 μm, the equivalent diameter is 4.58 μm. And the particle size is reduced to less than 25 μm, and the equivalent diameter is reduced to 2.48 μm. The shape factor increases as the particle size decrease. The smaller the powder size is, the closer the shape factor of the primary Si particles is to 1, showing that the Si particles are closer to spherical. Well known, the rounder Si particles lead to the better mechanical property of the Al-Si alloys. By comparing the change trends of equivalent diameter and shape factor, it can be preliminarily considered that the alloy prepared under four different powder particle sizes, of which the alloy rods prepared by a continuous extrusion of powder of less than 25 μm has the most superior performance. The diffusion of Si atoms will play an important role during the growth of primary Si and it will become much difficulty with increasing cooling rate. Thus, the size of the primary Si phase decreases significantly with the particle size [23].

Compared with figure 3, it can be noted that the Si particles more than 45μm is broken obviously after continuous extrusion. But the broken effect is not obvious in the particle size of less than 45μm. However, the shape factor of primary Si particles is larger than that of atomized powder.
3.4. Microstructure evolution during continuous extrusion

Figures 2 and 4 present the microstructure of the powders and rods. The phase composition of alloy powders includes irregular primary Si phase, eutectic Si phase and $\alpha$-Al matrix, the growth of eutectic Si phase occur on the surface of primary Si particles. While the alloy rods only consist of the primary Si phase and $\alpha$-Al matrix, the eutectic Si phase completely disappears. After the continuous extrusion, primary Si particles in the alloy powder change from coarse and irregular to round, primary Si particles are significantly refined.

The morphology of Si particles in the powder and rods is shown in figure 6. It can be seen that the powder contains coarse and irregular primary Si phase and non-equilibrium eutectic Si phase. Moreover, the edges and corners of the primary Si phase in alloy powder are very obvious. The coarse primary Si transforms into small regular and round Si particles, edges and corners disappear.

From the changes in figures 6(a) to (b), it can be noted that the continuous extrusion can effectively refine primary Si particles. The powder in the groove of the extrusion wheel rotates with the extrusion wheel. Friction and collision occur among the powders, which make the edges and corners of large and irregular primary Si particles smooth and even fracture to form new small Si particles. Secondly, pure shear deformation occurs at the corner of the plug, playing a major role in the slip of dislocations and the refinement of grains. The powder is further densely packed in the extrusion die cavity and then enters the die to be extruded into a rod. The metal flow direction and extrusion direction at the plug are $90^{\circ}$, it is an equal-angle angular extrusion, and sharp shear deformation significantly improves the refinement of primary Si particles.

It is the reason why eutectic Si disappears after continuous powder extrusion has been analyzed. It is considered that the continuous powder extrusion is a process of large plastic deformation, which can completely fracture the eutectic Si phase during the extrusion process. Cai et al found that the eutectic Si phase partially
The eutectic Si phase in the form of Si-Si clusters dissolves into the Al matrix when the temperature is more than 300 °C [24]. The dissolving of the Si phase generally starts at the edges and corners with the stress concentration. In continuous extrusion, the intense internal shear band and relatively high temperature (400 °C–550 °C) exist in the deformation zone [25]. So the dissolving rate of the broken eutectic Si phase is accelerated during continuous extrusion. The newly formed primary Si phase and growth of the pre-existing primary Si phase consume the Si-Si clusters. Eutectic Si is not

| Alloy       | Method                                      | Tensile strength/MPa | Elongation at break % | References |
|-------------|---------------------------------------------|----------------------|-----------------------|------------|
| Al-30Si     | Casting                                     | 48                   | —                     | [26]       |
| Al-30Si     | Modified by the in-situ synthesis of a novel 0.5%Al-P-O master alloy | 116                  | 0.97                  | [27]       |
| Al-30Si     | Modified by Al-P master alloy for three times | 128                  | —                     | [26]       |
| Al-30Si     | Vacuum hot-pressed sintering                | 137                  | —                     | [28]       |
| Al-30Si     | Pressure infiltration                       | 165                  | —                     | [29]       |
| Al-30Si     | Permanent magnet stirring                   | 160–168              | 6.7                   | [30]       |
| Al-27Si     | Spray forming + semi-solid extrusion        | 195                  | 6.7                   | [31]       |
| Al-30Si     | Thixotropic die casting + isothermal treatment | 206–220              | —                     | [32]       |
| Al-28.5Si   | Continuous powder extrusion                 | 211–266              | 0.99–2.21             | This study |

Figure 8. Fracture appearance of processed rods of particle sizes: (a) >75 μm, (b) 75~45 μm, (c) 45~25 μm, (d) < 25 μm.
easy to form under severe plastic deformation. Some eutectic Si phase is modified into finely dispersed particles mixed with primary Si phase so that it is impossible to distinguish eutectic Si from its morphology. Therefore, the eutectic Si phase is not observed in the rods.

3.5. Tensile property of alloy rods

The UTS and elongation at break of the extruded rods fabricated with different powder sizes are exhibited in figure 7, the tensile strength and elongation increase with the reduction of powder size, indicating the tensile property of alloy increases gradually as the powder size decreases. When the powder size is more than 75 μm, the tensile strength of the alloy rod is 211 Mpa, and the elongation at break is 0.99%; when powder size is reduced to less than 25 μm, the equivalent diameter of primary Si particles reduces, further increasing the ultimate tensile strength to 266 MPa. The shape factor of primary Si particles increases, so elongation at break is also further increased to 2.21%.

To refine primary Si particles so that achieving improvement of mechanical properties of hypereutectic Al-Si alloy, some other processing techniques and complex modification method for hypereutectic Al-Si alloy are summarized in table 2, UTS and elongation at break of the continuous powder extrusion rods are also exhibited. By comparing, the UTS of the alloy prepared by other processes is much lower than that of the continuous powder extrusion alloy of 266 MPa, which states better tensile property for hypereutectic Al-Si alloys can be obtained through continuous powder extrusion.

Figures 8(a)–(d) are fracture morphology of alloy rods with more than 75 μm, 75–45 μm, 45–25 μm and less than 25 μm, and there are a great number of cleavage platforms of different sizes in the fracture. Hence the fracture mechanism of the alloy is the brittle fracture. These cleavage platforms are formed by the brittle fracture of numerous Si particles. Therefore, the size of the cleavage platform is the same as that of Si particles. With the decrease of powder size, Si particles also decrease, which indicates that the size of the cleavage platform also decreases, as is shown in figure 8(d). At the same time, it also shows the obvious characteristics of the dimple. In fact, this is not a common dimple on the fracture surface of ductile materials. It is formed after the stripping of Si particles from the α-Al matrix. However, the number of dimples is very small, which indicates that the fracture of the alloy rods is mainly due to the brittle fracture of Si particles.

The micro-crack is the key to the fracture of alloy rods. Noted from figure 8 that fracture is mainly dominated by the fracture and the stripping of Si particles from the α-Al matrix. The stripping of Si particles from the Al matrix is not considered in this paper due to few dimples. It is considered that the alloy rods fracture is mostly caused by the fracture of Si particles.

Si particles can be approximately regarded as spherical inclusions existing in the hypereutectic Al-Si alloys. According to the inclusion theory, the total elastic energy \( W_1 \) of Si particles with radius \( R \) is given by [33]:

\[
W_1 = \frac{4}{3} \pi R^3 \omega
\]  

(3-3)

Where \( \omega \) is the elastic energy density of the Si particle, it can be determined by the stress field and the elastic modulus of each phase in the material. \( R \) is the radius of Si particles.

Yuan [33] believed that energy \( W_2 \) required for the fracture of Si particles with radius \( R \) is

\[
W_2 = 2 \pi R^2 \sigma_f
\]  

(3-4)

Where \( \sigma_f \) is the interface energy of the Si particle phase. Under the condition of complete brittle fracture, the critical condition of fracture is \( W_1 = W_2 \), that is

\[
\omega_c = \frac{3}{2} \frac{\sigma_f}{R_c}
\]  

(3-5)

Where \( \omega_c \) is the critical elastic energy density required for the fracture of Si particles, \( R_c \) is the critical radius required for the fracture of the Si particle.

Apparently, \( \omega_c \) is inversely proportional to the radius \( R \) of Si particles. That is to say, the Si particles are easier to be broken when the \( R \) is larger because the elastic energy density \( \omega \) is smaller. If the radius \( R \) of Si particles is larger than the critical radius \( R_c \), the elastic energy density \( \omega \) required for a fracture will be less than the critical elastic energy density \( \omega_c \), the material will fail prematurely. It can be noted from figure 5 that the radius \( R \) of Si particles decreases with the reduction of the powder size. Therefore, the tensile properties of the continuous powder extruded alloys are improved by reducing the powder size, as shown in figure 7.

Both global and local instability could control the final failure of an alloy. For the alloy rod with large particle size, the coarsened Si particles with low values of shape factor are easily assembled in the crystal boundaries [34]. During formation, the fractured Si particles tend to cluster, causing local instability and fracture. In addition, the ratio of tensile instability (\( \varepsilon_i \)) to fracture strain (\( \varepsilon_f \)) determines the plasticity of the alloy rods. If \( \frac{\varepsilon_i}{\varepsilon_f} > 1 \), the plasticity is lower, indicating that the failure of alloy rods is prior to the global instability. Furthermore, a stress concentration would occur at the cusp of the Si because of the discordant deformation between Al and Si. This is
easily causing the cracking of the Si particles \cite{35}. Whereas, for the alloy rod with small particle size, the fine and round Si particles with high values of shape factor are less vulnerable to damage, causing enhanced plasticity.

Conform extrusion process can significantly refine the grain and change the distribution of reinforcement phase. The strengthening effect of grain refinement is described by the Hall–Petch equation (3–6).

\[ \sigma_f = \sigma_i + k_d d_1^{1/2} \]  

(3–6)

where \(\sigma_f\) is the yield strength, \(\sigma_i\) is a constant, \(k_f\) is a material dependent constant, \(d_1\) is the grain diameter.

It is thus clear that the grain is refined and the strength is enhanced during extrusion. In the continuous extrusion of pure Al particles, Song et al. found that the strength of rod increases with the decrease of particle size. It is considered that the grain size decreases with the decrease of grain size \cite{36}. The performance of Al matrix increases with the decrease of particle size due to the fine grain strengthening and solid solution strengthening \cite{21}.

The strengthening in particle reinforced composites is related to the size and spacing of particles. Orowan strengthening plays a fundamental role in hypereutectic Al-Si alloy. According to the Orowan strengthening model \cite{37}, the yield strength of hypereutectic Al-28.5Si alloy is described by equation (3–7).

\[ \Delta\sigma_{Oro} = \frac{0.13b_u G_u}{\ln \left( \frac{1}{\sqrt{1 + \frac{1}{V_{Si}}} - 1} \right)} \frac{d_{Si}}{2h_u} \]  

(3–7)

where \(\Delta\sigma_{Oro}\) is the increase in yield strength by Orowan strengthening, \(b_u\) is the Burgers vector of the matrix Al, \(G_u\) is the shear modulus of the matrix Al, \(d_{Si}\) is the radius of Si particle, \(V_{Si}\) is the volume fraction of Si particle.

It can be seen from equation (3–7) that the diameter of Si particles is an important factor affecting the strengthening effect of Orowan mechanism. For Al-28.5Si alloy with a certain volume fraction of Si particles, the smaller the size of Si particles, the more obvious the strengthening effect of Orowan.

In addition, the thermal expansion coefficients of Si particles (7.6 × 10^-6 K^-1) and \(\alpha\)-Al matrix (23.5 × 10^-6 K^-1) at 300–373 K are quite different \cite{38}, which leads to the different thermal deformation between Si particles and \(\alpha\)-Al matrix. Therefore, the residual stress is generated around the Si particles. These residual stresses increase the dislocation density around the Si particles, which plays the role of thermal mismatch strengthening.

4. Conclusion

(1) The primary Si particles having passivated edges and corners are homogeneously distributed in the \(\alpha\)-Al matrix after continuous extrusion, and there are no eutectic Si phases in the processed products. The equivalent diameter and shape factor of the primary Si particles in the alloy rod are respectively proportional and inversely proportional to the powder size.

(2) The phase composition of powders includes irregular primary Si phase, eutectic Si phase and \(\alpha\)-Al matrix. After the continuous extrusion, eutectic Si phase completely disappears.

(3) Compared with the alloy of similar composition fabricated by other processes, the tensile property of the alloy rod fabricated by continuous powder extrusion has greatly been improved. Powder size plays an important role in the improvement of the tensile property. The particle size of the powder is more than 75 \(\mu\)m, and tensile strength is 211 MPa, the elongation is to 0.999%, the particle size of the powder is less than 25 \(\mu\)m, tensile strength is up to 266 MPa. The elongation is also to 2.21%.

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Data availability statement

The data that support the findings of this study are available upon reasonable request from the authors.
Declaration of competing interest

The authors declare that there is no conflict of interest.

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