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Hot deformation behavior of Al-Zn-Mg-Cu-Zr aluminum alloys during compression at elevated temperature

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Abstract: The hot compression tests of Al-Zn-Mg-Cu-Zr aluminum alloys (7056 alloy and 7150 alloy) were performed in a temperature range from 300 to 450 °C and at strain rate range from 0.01 to 10 s^{-1}. The results show that the true stress—true strain curves exhibit a peak stress at a critical strain, then the flow stresses decrease monotonically until high strains, showing a dynamic flow softening. The peak stresses depend on the temperature compensated strain rate, which can be represented by the Zener-Hollomon parameter Z in the hyperbolic-sine equation with hot deformation activation energy of 244.64 kJ/mol for 7056 alloy and 229.75 kJ/mol for 7150 alloy, respectively, while the peak stresses for the former are lower than those for the latter under the similar compression condition. The deformed microstructures consist of a great amount of precipitates within subgrains in the elongated grains at high Z value and exhibit well formed subgrains in the recrystallized grains at low Z value. The smaller subgrains and greater density of fine precipitates in 7150 alloy are responsible for the high peak stresses because of the substructural strengthening and precipitating hardening compared with 7056 alloy.

Key words: Al-Zn-Mg-Cu-Zr aluminum alloys; flow stress; dynamic recrystallization; dynamic precipitation

1 Introduction

The 7000 series alloys, based on the Al-Zn-Mg system, have a combination of high strength and fracture toughness, as well as resistance to stress corrosion cracking, which renders them very useful in aircraft and aerospace industry applications. Increased strength of these alloys was anticipated by increasing Zn/Mg ratio and Cu concentration, especially by increasing Zn/Mg ratio, as these are the principal basis of precipitation strengthening. Minor addition of Zr could inhibit recrystallization during thermo-mechanical processing, but possibly lowers the hot workability at the same time[1−3]. Workability is usually defined as the amount of deformation that materials can undergo without cracking and reach desirable deformed microstructures at a given temperature and strain rate. Improving workability means increasing the processing ability and improving properties of the materials, which probably can be achieved by adopting optimum processing parameters. However, these difficulties usually arise due to the lack of knowledge of the hot deformation behavior of specific materials under the conditions encountered in the processing, such as extrusion, forging and rolling.

Early studies on the hot deformation of Al-Zn-Mg aluminum alloys clearly demonstrated that the flow stress diminishes with increasing deformation temperature and decreasing strain rate, and the flow curves which decline as the deformation temperature rises are marked by a peak and softening. As a consequence, subgrain structure develops inside elongated grains as a result of dynamic recovery (DRV). CERRI et al[4] made a comparative study on the hot workability of 7012 and 7075 alloys after different pretreatments, indicating that the peak flow stress of various treated Al-Zn-Mg-Cu alloys is related to the strain rate by the hyperbolic sine equation; the activation energy for precipitated alloys and over-aged materials is 141−162 kJ/mol for 7012 alloy and 143−156 kJ/mol for 7075 alloy, similar to that of the bulk self-diffusion energy of pure aluminum, 142 kJ/mol. For the solution-treated materials, the values are 200−230 kJ/mol for 7012 alloy and 300−400 kJ/mol for 7075 alloy, respectively. The flow behavior is the result of not only dynamic precipitation (DPN), but also dynamic recrystallization (DRX) during hot deformation of 7012 and 7075 aluminum alloys. HU et al[5] demonstrated

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that the stress level of 7050 aluminum alloy can be presented by a Zener-Hollomon parameter in a hyperbolic sine-type equation with the hot activation energy of 256.6 kJ/mol, and the soften mechanism of the alloy during high temperature deformation transforms from dynamic recovery to continuous dynamic recrystallization with decreasing Z value. JIN et al[6] reported that the peak stress of 7150 aluminum alloy can be represented by a Zener-Hollomon parameter in the hyperbolic-sine equation with hot deformation activation energy of 229.75 kJ/mol, and the main reason for the flow softening transforms from dynamic recovery and recrystallization at low Z value to the dynamic precipitates and successive dynamic particles coarsening at high Z value.

7056 aluminum alloy contains higher level of Zn/Mg ratio and alloying contents than those of 7150 alloy and other Al-Zn-Mg-Cu alloys mentioned above. The purpose of the present work is to determine the differences between the hot workability of 7056 and 7150 alloys.

2 Experimental

The experiments were carried out on two Al-Zn-Mg-Cu-Zr alloys (7056 alloy and 7150 alloy), whose main chemical compositions are given in Table 1. Cylindrical samples with size of d10 mm×15 mm were machined from commercially semi-continuous chill cast billets of 180 mm thickness and subsequently were homogenized soaking at 460 °C for 24 h for 7056 alloy and at 465 °C for 24 h for 7150 alloy, followed by water quenching. Convex depressions of 0.2 mm depth were machined on both ends of the samples in order to maintain the lubricant of graphite mixed with machine oil during compression tests. Compression tests were carried out on a computer servo-controlled Gleeble 1500 system at strain rates of 0.01, 0.1, 1 and 10 s^{-1} and deformation temperatures of 300, 350, 400 and 450 °C, respectively. The samples were heated to deformation temperature at a heating rate of 10 °C/s and held for 180 s by thermocoupled-feedback-controlled AC current before compression, and then were deformed to half of their original height and water quenched immediately. The deformed structures were sectioned parallelly to the compression axis along the direction of centerline and prepared by the conventional methods for micro-structural observations on an MM−6 metallographic microscope (OM) and an H-800 transmission electron microscope (TEM).

Table 1 Chemical compositions of alloys used in tests (mass fraction, %)

Alloy	Zn	Mg	Cu	Zr	Al
7056	8.98	1.54	1.45	0.12	Bal.
7150	6.29	2.33	2.10	0.09	Bal.

3 Results and discussion

3.1 Flow stress of Al-Zn-Mg-Cu-Zr alloys

A series typical true stress—true strain curves obtained during hot compression of 7056 and 7150 alloys at strain rate of 0.01−10 s^{-1} and deformation temperature of 300−450 °C are shown in Fig.1. It can been seen that the true stress—true strain curves exhibit a peak stress at a certain strain, followed by dynamic flow softening until the end of compression. The flow softening is probably subjected to the dynamic recovery and recrystallization[5, 7], as well as dynamic coarsening of precipitates during hot deformation of precipitation hardening aluminum alloys[4, 6]. Moreover, the peak stress and flow stress increase with increasing strain rate and decreasing deformation temperature. The strain corresponding to the peak stress increases with increasing strain rate and/or with decreasing deformation temperature due to the high hardening rate at initial deformation stages.

Tables 2 and 3 respectively show the peak stress values of 7056 alloy and 7150 alloy at different deformation temperatures and strain rates. It can be found that the peak stress values of 7056 alloy are lower than those of 7150 alloy with the similar deformation.

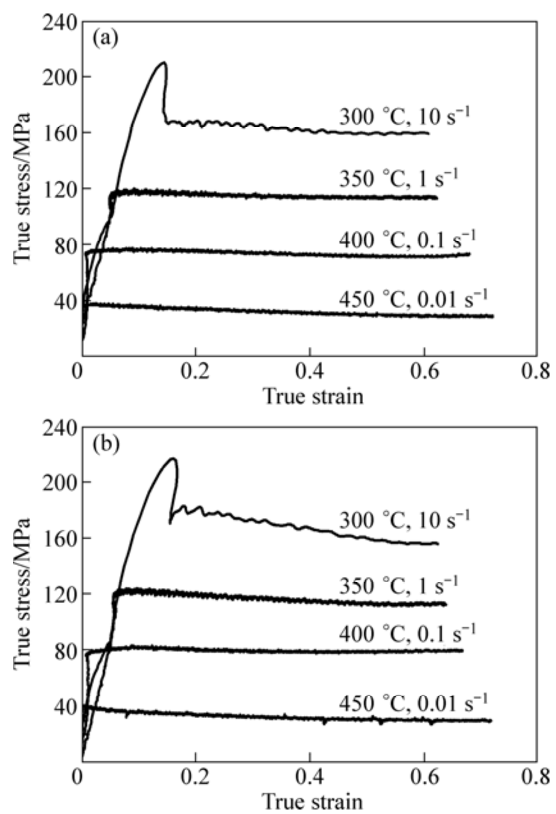

Fig.1 True stress—true strain curves during hot compression of Al-Zn-Mg-Cu-Zr alloys: (a) 7056 alloy; (b) 7150 alloy

Table 2 Peak stress of 7056 alloy at different temperatures and strain rates

Strain rate/s^{-1}	Peak stress/MPa			
	300 °C	350 °C	400 °C	450 °C
0.01	153.22	86.71	58.47	37.61
0.1	167.67	108.24	76.84	61.15
1	176.95	118.38	96.18	77.23
10	208.24	159.63	127.52	116.75

Table 3 Peak stress of 7150 alloy at different temperatures and strain rates

Strain rate/s^{-1}	Peak stress/MPa			
	300 °C	350 °C	400 °C	450 °C
0.01	157.19	99.24	70.53	55.84
0.1	170.73	115.8	83.26	70.53
1	182.79	123.83	105.08	90.53
10	216.01	167.29	127.57	117.12

The actual softening mechanisms and the difference of the peak stress level will be discussed in detail in the following section by comparison analysis with hot deformation activation energy calculations and microstructural observations.

In hot deformation of metallic materials, it is commonly accepted that the relationship between the peak stress or steady state stress, strain rate and temperature can be expressed as[4, 8−10]:

$$Z = \dot{\varepsilon} \exp\left(\frac{Q}{RT}\right) = f_1(\sigma) = A_1 \sigma^{n_1} \tag{1}$$

$$Z = f_2(\sigma) = A_2 \exp(\beta\sigma) \tag{2}$$

where A_1, A_2, n_1 and β are constants; Z is the Zener-Hollomon parameter; Q is the activation energy for hot deformation and R is the gas constant. The power law Eq.(1) and the exponent-type Eq.(2) break at a high stress and a low stress, respectively. The hyperbolic sine type Eq.(3), proposed by SELLARS and McTEGART[11] is a more general form suitable for stresses over a wide range.

$$Z = f(\sigma) = A(\sinh \alpha\sigma)^n \tag{3}$$

where A and n are constants; α is the stress multiplier and also the additional adjustable parameter and it is calculated as:

$$\alpha = \frac{\beta}{n} \approx \frac{\beta}{n_1} \tag{4}$$

It is easy to obtain values of n_1 or n and β by means of linear regression through Eq.(1) and Eq.(2), respectively. The value of α is taken as a first approximation in Eq.(4) and the relationship of Eq.(3) is derived to obtain a new value of n, which is then iterated to obtain the optimum values of α, n, A and Q[8, 10]. The material constants of the alloys obtained from the experimental data are shown in Table 4, which can be used to quantificationally describe the relationship of the flow stress and temperature and strain rate by a Zener-Hollomon parameter in the hyperbolic sine equation with the related coefficient of 0.972 1 for 7056 alloy and 0.977 8 for 7150 alloy, respectively.

Table 4 Material constants of alloys used in hyperbolic sine equation

Alloy	A/s^{-1}	α/MPa^{-1}	n	Q/(kJ·mol^{-1})
7056	8.238×10^{16}	0.015 061	6.347 71	244.64
7150	4.161×10^{14}	0.019 56	5.143 36	229.75

The hot deformation activation energy Q is an important physical parameter serving as an indicator of deformation difficulty degree in plasticity deformation. The values of Q for 7056 alloy and 7150 alloy are 244.64 and 229.75 kJ/mol, respectively, which are higher than those of the precipitated and over aged 7012 alloy (141−162 kJ/mol), similar to that of the bulk self-diffusion of pure aluminum (142 kJ/mol), and close to the values of the solution-treated 7012 alloy (200−230 kJ/mol)[4], as well as the as-quenched 7050 aluminum alloy (256.6 kJ/mol)[5]. Generally, higher deformation activation energy is found in hot deformation of the solution treated aluminum alloys. It is known that a stationary dislocation in a solid solution alloy can be pinned by solute atoms if they are able to diffuse to the dislocation, and hence an additional increment of stress will be required to free the dislocation from its energetically favorable position. In addition, any dislocation moving through a solid solution will encounter friction drag thus raise the energy required for the movement. Hence, any increase of foreign atoms held in solution state will increase the activation energy. In this case, it can be deduced that higher Q value of 7056 alloy than 7150 alloy may be associated with increasing alloying contents of Zn, Mg and Cu held in solution by the homogenization sequence. At the same time, recovery processes are also hindered by higher solute atom vacancy binding energy which effectively reduces the number of vacancies available for dislocation climbing whilst the misfit strain effectively raises the dislocation density, contributing to an increase in hot deformation activation energy for 7056 alloy.

3.2 Microstructural evolution

Figures 2 and 3 show the representative optical deformed microstructures of 7056 alloy and 7150 alloy, respectively. The microstructure is composed of elongated and newly refined grains separated by moderately high-angle boundaries, indicating that

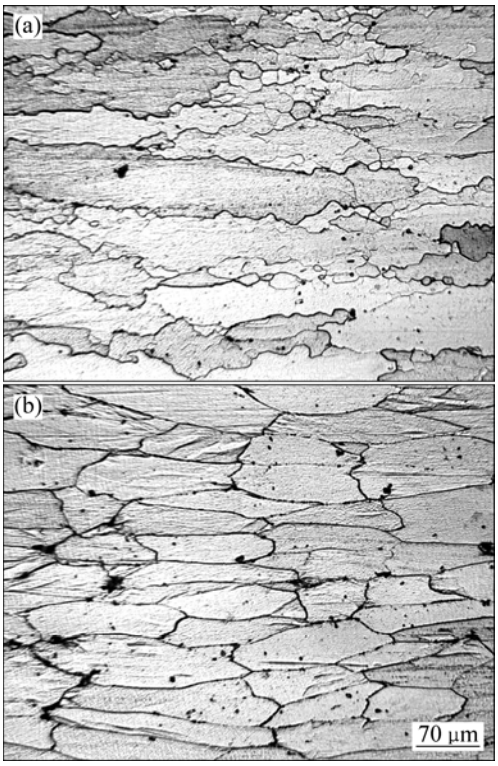

Fig.2 Optical microstructures of 7056 alloy deformed at 450 °C, 0.01 s^{-1} (a) and 300 °C, 10 s^{-1} (b)

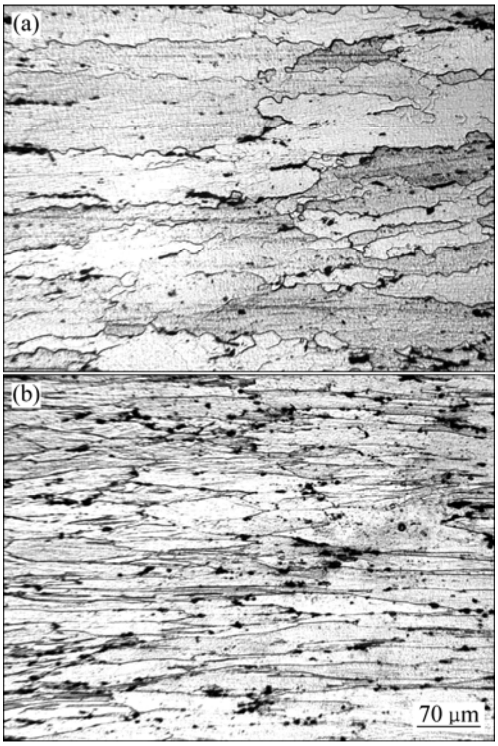

Fig.3 Optical microstructures of 7150 alloy deformed at 450 °C, 0.01 s^{-1} (a) and 300 °C, 10 s^{-1} (b)

dynamic recrystallization (DRX) occurs in evidence during hot compression deformation and the DRX grain size is dependent sensitively on deformation temperature and/or strain rate, also on Zener-Hollomon parameter *Z*. Decreasing *Z* value, that is increasing deformation temperature or decreasing strain rate, leads to more adequate proceeding of DRX and coarser recrystallized grains. The difference is that the recrystallized grains observed in 7150 alloy are finer compared with those in 7056 alloy. During hot deformation of aluminum alloys at high temperature, sufficient migration of atoms and dislocations causes merging of some grains, and low-angle grain boundaries transform into high-angle grain boundaries through absorbing dislocations. Subsequently, the recrystallized grains grow rapidly through the migration of dislocations and high-angle grain boundaries. Grain boundary sliding (GBS) takes place along the initial grain boundaries, resulting in inhomogeneous strain distribution and then frequent development of microshear bands in grain interiors[6−7, 12]. Subdivision of original grains into misoriented small domains leads to the average grain size evolved which is similar to the minimum size of regions fragmented by microshear bands. The grain refinement caused by a deformation-induced continuous reaction is essentially similar to continuous dynamic recrystallization (CDRX) [7, 12]. CDRX caused by gradual subgrain growth leads to the formation of high-angle boundary migration. Typically, the mechanisms of CDRX nucleation appear likely that some subgrains coarsen through mergering other adjacent subgrains to produce high-angle boundaries[13−16]. Subgrain boundary migration occurs when particle coarsening leaves some subgrain boundaries weakly restrained. Alternatively, elevated temperature straining may provide added driving pressure for boundary migration to accelerate continuous recrystallization[13]. Indeed, it was reported that aluminum alloys containing high density of Al_3Zr or Al_3Sc dispersoids or containing high density of several types of fine particles act as sufficient boundary drag force to prevent the nucleation of discontinuous recrystallization and hence accelerate continuous recrystallization[13−15]. As a result, it can be concluded that the grain refinement during hot deformation of Al-Zn-Mg-Cu-Zr alloys occurs with the main type of the CDRX.

TEM images show that the subgrains exhibit high-angle sub-boundaries with a small quantity of dislocations in the specimen deformed at 450 °C and strain rate of 0.01 s^{-1}, which is typical of all low *Z* value deformation. With lower *Z* value, the subgrains increase in diameter and more equiaxed subgrains appear with more clearly defined sub-boundaries containing dislocations in an orderly fashion, as shown in Figs.4(a) and 5(a). On the contrary, under the condition of high *Z* values, the subgrains are elongated with thick walls of very tangled dislocations. Moreover, large number of dynamic precipitates and coalesced particles appear in the subgrain interiors when being deformed at 300 °C and strain rate of 10 s^{-1}, indicating that the dynamic precipitates and successive dynamic particles coarsening have occurred during hot compression of Al-Zn-Mg-Cu-Zr alloys at low deformation temperature and high strain rate, as shown in Figs.4(b) and 5(b). The soluble atoms and uniform dispersion of fine particles interact with the dislocations, which reduces dynamic recovery and recrystallization, hence, the higher strain energy may speed up dynamic precipitation and result in the flow stress fluctuation at low deformation temperature and high strain rates. The subgrains remain well formed with clean high-angle boundaries and coarse precipitates, which are distributed in the grain interior, extremely similar to that produced by solution-treatment followed by ageing of Al-Zn-Mg-Cu alloys[17−18].

Figures 4 and 5 also demonstrate the differences in size of the subgrains and precipitates in 7056 alloy and 7150 alloy. The smaller subgrains and greater density of precipitates exhibit in 7150 alloy are responsible for the high peak stresses because the peak stress is found to depend inversely on the subgrain size. CERRI et al[4] made a comparative study on the hot workability of 7012 and 7075 alloys after different pretreatments, indicating that the flow strength of 7012 alloy is generally greater than that of 7075 alloy, but the activation energy of 7075 alloy is higher than that of 7012 alloy in the solution-treated condition.

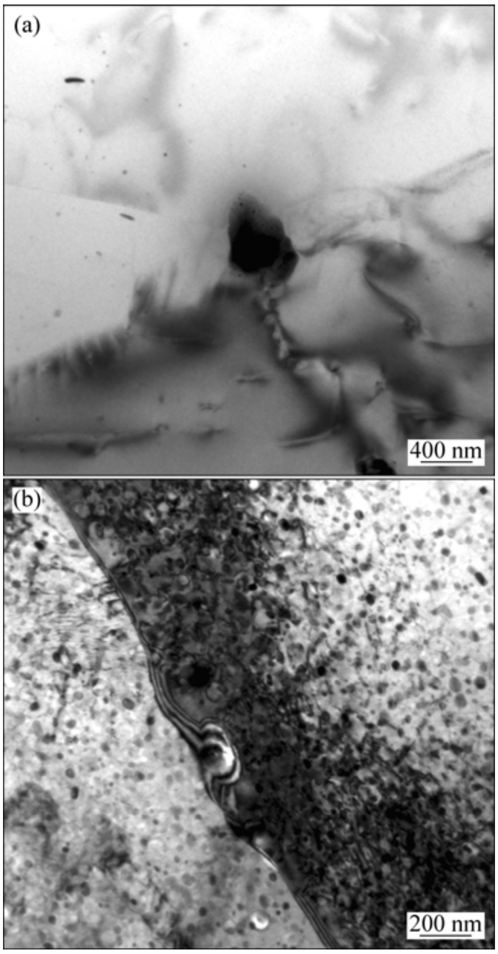

Fig.4 TEM images of 7056 alloy deformed at 450 °C, 0.01 s^{-1} (a) and 300 °C, 10 s^{-1} (b)

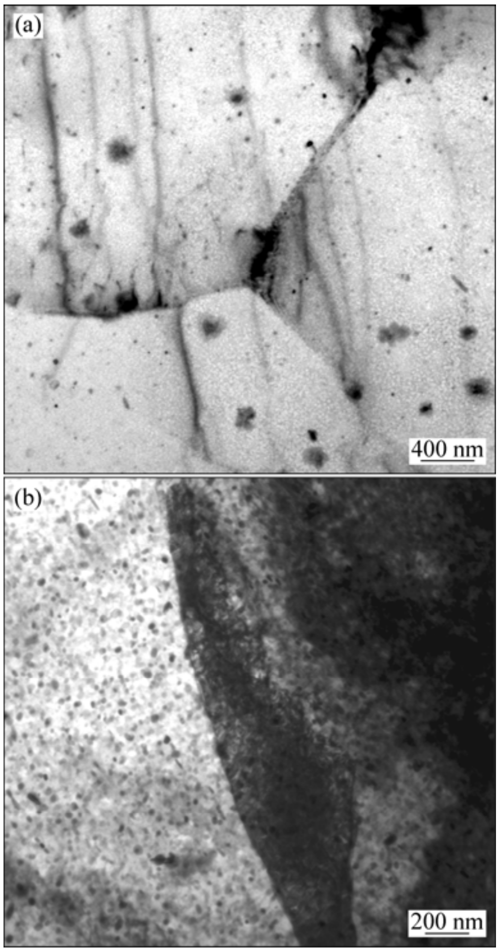

Fig.5 TEM images of 7150 alloy deformed at 450 °C, 0.01 s^{-1} (a) and 300 °C, 10 s^{-1} (b)

4 Conclusions

1) The true stress—true strain curves of 7056 and 7150 alloys exhibit a peak stress at a critical strain, after which the flow stresses decrease monotonically until high strains, showing a dynamic flow softening. The peak stresses depend on the temperature compensated strain rate, which can be represented by the Zener-Hollomon parameter Z in the hyperbolic-sine equation with hot deformation activation energy of 244.64 kJ/mol for 7056 alloy and 229.75 kJ/mol for 7150 alloy, while the peak stresses for 7056 alloy are lower than those for 7150 alloy under similar compression condition.

2) The deformed microstructures of 7056 and 7150 alloys consist of a great amount of precipitates within

subgrains in the elongated grains at high Z value, and exhibit well formed subgrains in the recrystallized grains at low Z value. The smaller subgrains and greater density of fine precipitates in 7150 alloy are responsible for the high peak stresses because of the more substructural strengthening and precipitating hardening compared with 7056 alloy.

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Al-Zn-Mg-Cu-Zr 铝合金的高温热压缩变形行为

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摘　要：在温度为 300−450 °C 和应变速率为 0.01−10 s^{-1} 的变形条件下，对 Al-Zn-Mg-Cu-Zr 合金(7056 和 7150 铝合金)进行热压缩实验。结果表明：在一定的应变峰值出现后，流动应力随应变增加单调下降，呈现出流动软化。峰值应力取决于温度补偿应变速率 Z 的大小，可用包含 Zener-Hollomon 参数的双曲正弦关系来描述合金热流变行为。7056 合金的变形激活能为 244.64 kJ/mol，而 7150 合金的为 229.75 kJ/mol；在同样的变形条件下，前者的峰值应力却低于后者。在高 Z 值条件下，在延长晶粒的亚晶粒中存在大量析出物；而在低 Z 值条件下，再结晶化的晶粒内出现完整的亚晶。7150 合金中存在细小亚晶和大量析出物，由于亚结构强化和析出硬化造成其峰值应力比 7056 合金高。

关键词：Al-Zn-Mg-Cu-Zr 合金；热压缩变形；流变应力；动态再结晶；动态析出

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