Enhancing Mechanical Properties of Mg–6Zn Alloy by Deformation-Induced Nanoprecipitation

Kai Yan¹ · Huan Liu² · Xiaowei Xue³ · Jing Bai⁴ · Honghui Chen¹ · Shuangquan Fang¹ · Jingjing Liu¹

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Abstract

We presented the solution of deformation-induced precipitation after homogenization to enhance the mechanical properties of Mg–6Zn alloys. The results show that the improved strategy exhibits more effective strengthening role than grain refinement methods based on low-temperature severe plastic deformation under the same strain. The low-temperature deformation with larger extrusion ratio results in massive nano-sized precipitates and excellent mechanical properties with the yield strength of 355 MPa and the ultimate tensile strength of 405 MPa. The increased mechanical properties are strong and tough enough to resist the stress and not be worn away when the alloy nail penetrates through the pig thigh bone, potentially extending more orthopedic surgery applications for Mg–Zn alloys.

Keywords Mg–6Zn alloy · Equal channel angular pressing · Deformation-induced precipitation · Severe plastic deformation · Mechanical property

1 Introduction

Extensive interest in degradable metal for implants has led to numerous notable literature and important progress in this field, especially the Mg–Zn binary alloys have attracted great attention as a biodegradable material due to their good biocompatibility and good biosafety [1–5]. With respect to the applications of biodegradable implants, mechanical and corrosion properties of a material are the most important [6]. As a proper biocompatible element, Zn could improve corrosion resistance of Mg alloys by elevating the corrosion

potential, promoting the intact corrosion film and elevating the charge transfer resistance of magnesium [7–9]. Taking into account Zn concentrations over solubility limit (approximately 6.2 wt% in Mg–Zn binary alloy), corrosion properties are deteriorated due to the formation of secondary phases and the consequent galvanic couple effect, which are responsible for production of excessive hydrogen gas [10–13]. In addition, the mechanical properties of Mg–Zn alloys with the Zn concentration ranging from 1.0 to 6.0 wt% increased with Zn content [6, 7, 14]. Therefore, on the whole, Mg–6Zn alloy shows advantages over its potential rivals in the application of biomedical implants.

In the last two decades, an enormous amount of research effort had gone into ultrafine-grained Mg alloys. Especially, the severe plastic deformation (SPD) technologies were introduced to refine microstructure, imposing large deformation to Mg alloys at temperature as low as possible and reducing the grain size to the micron and even submicron scale [15–18]. However, these technologies only can improve the ultimate tensile strength of Mg–Zn series alloys to at most 370 MPa [19–27]. Recently, some exceptional high strengths, such as about 390 MPa in yield strength (YS) and 440 MPa in ultimate tensile strength (UTS), have been achieved in deformation processed ZK60 Mg alloys [28, 29]. These materials are characterized by larger grain sizes of about 7 μm , and their high strengths are attributed to the

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✉ Kai Yan
yankai@yzu.edu.cn

✉ Huan Liu
liuhuanseu@hhu.edu.cn

¹ College of Mechanical Engineering, Yangzhou University, Yangzhou 225127, China

² College of Mechanics and Materials, Hohai University, Nanjing 211100, China

³ Northern Jiangsu People's Hospital Affiliated to Yangzhou University, Yangzhou 225001, China

⁴ School of Materials Science and Engineering, Southeast University, Nanjing 211189, China

nano-scaled precipitates, the formation of substructures and the modification of non-basal texture. The common point to the emerging technologies is that the low-temperature deformation at the last procedure induces the nanoprecipitation as a thermal–mechanical aging treatment. Before that all the alloys have been fully homogenized to improve precipitation effect. From this, it indicates that we might employ a new strategy based on the deformation-induced precipitation followed by fully homogenization to develop the ultra-high strength Mg–6Zn alloy.

Since the metal bars are the necessary materials for orthopedic devices such as Kirschner wire and tapping bone nail, in this study, we integrated the forming process and designed a comparative experiment to investigate its influence on microstructures and mechanical properties of Mg–6Zn alloy based on the two strengthening strategies, i.e., fine grain strengthening and deformation-induced precipitation strengthening. Moreover, the mechanism of deformation-induced precipitation strengthening for Mg–6Zn alloy was discussed.

2 Experimental

The experimental materials used for this investigation were the ϕ 15-mm-extrusion rod (sample EX) with a nominal composition of 6 wt% Zn and Mg balance. The samples were processed by six passes of equal channel angular pressing (ECAP) with inner angle of 90° and route B_C at a pressing speed of 1.5 mm/s and then were die extruded to thin rods for use in Kirschner wire. The temperature of extrusion should be as low as possible to ensure the high mechanical properties of the resulting rods. The Mg–6Zn rods were processed by two processing routes. The route A represents the strengthening strategy toward microstructure refinement, the samples were ECAP processed at the lower temperature of 160°C for 6 passes (sample A1) and then extruded into ϕ 2.6 mm rods at 180°C with the extrusion ratio of about 33 (sample A2). The route B represents the strategy toward deformation-induced precipitation followed by complete homogenization. The samples were ECAP processed at a higher temperature of 320°C for 6 passes (sample B1) and then extruded into ϕ 2.6 mm rods at 180°C with the extrusion ratio of about 33 (sample B2). Thus, it can be seen that

the two routes have nearly identical processes except the temperatures of ECAP processing (Table 1).

Microstructural observations were conducted by an optical microscope (OM, Leica DMI3000M), a field-emission scanning electron microscope (SEM, Hitachi S-4800II) and a transmission electron microscope (TEM, Tecnai G2). The OM and SEM samples were cut along transverse cross section, grinded with a series of SiC sandpaper and diamond polishing agent, and then etched in a 2% aqueous nitric acid solution. The misorientation angle distribution and texture examination were carried out on an electron backscatter diffraction system (EBSD, Oxford instruments) with channel 5 data acquisition and analysis software. The samples prepared for EBSD were grinded and polished by ion-beam. According to ASTM E8-11 [30], room temperature tensile tests of the samples were carried out by an CMT5105 electronic universal testing machine at a speed of 2 mm/min. Specimens of sample EX, A1 and B1 for tensile tests had a gauge length of 30 mm and cross section of ϕ 6 mm, and sample A2 and B2 for tensile tests had a gauge length of 10 mm and cross section of ϕ 2 mm. Ultimate tensile strength (UTS), yield strength (YS) and elongation were obtained based on the average value of three tests. Hardness test was performed using an Hv-1000z Vickers microhardness tester with a load of 100 gf and a dwell time of 10 s. Hardness was obtained based on the average value of five tests. Finally, we proposed a bone-penetration test to simulate the surgical environment of orthopedic devices. The rods of sample A2 and sample B2 were processed into Kirschner wires with sharp three-arris edge as test materials and then were driven by an electronic hand drill to drill into an adult pig thigh bone.

3 Results

3.1 Evolution of the Intermetallic Phase Distribution

As shown in Fig. 1, the distribution of intermetallic phase was characterized by SEM micrography, and it can be seen that the microstructure of the as-extruded Mg–Zn alloy consists of primary α -Mg matrix, eutectic structure and fine secondary phase particles (Fig. 1a). Processed by the route A, eutectic structure is distorted and broken after six passes

Table 1 Processing state of Mg–6Zn samples

Sample code	Processing state	Diameter (mm)
EX	Casting and extrusion	15
A1	EX, ECAPed for 6 passes at 160°C	15
A2	EX, ECAPed for 6 passes at 160°C and extruded at 180°C	2.6
B1	EX, ECAPed for 6 passes at 320°C	15
B2	EX, ECAPed for 6 passes at 320°C and extruded at 180°C	2.6

Fig. 1 SEM images of the tested Mg–6Zn alloy: **a** EX, **b** A1, **c** A2, **d** B1, **e** B2

ECAP at 160 °C (shown in Fig. 1b) and then the secondary phase is transformed into the clear banded structure distributed along the extruded direction after the post-ECAP extrusion at 180 °C (shown in Fig. 1c). By contrast, the six-pass ECAP at 320 °C results in fully solid solution with the exception of small minority undissolved secondary phase (shown in Fig. 1d), and the subsequent extrusion at 180 °C causes the massive and uniform precipitation of fine secondary phases (shown in Fig. 1e).

The morphology of precipitates is further observed by TEM micrography, as shown in Fig. 2. The broken eutectic structure after lower temperature ECAP is consisted of irregular-shaped secondary phase particles with the size of

about 1 μm, and then turns into the band structure after the post-ECAP extrusion, as shown in Fig. 2a and b. There are small number of particles distributed sparsely in the vicinity of the banded structure in the sample A2, which have spherical shape and the average diameter of about 20 nm, as shown in Fig. 2c. By comparison, the TEM micrographs of A1 and A2, it can be seen that the subsequent extrusion causes the morphology changing of the secondary phase particles and induces the formation of clear banded structure but has little effect in precipitation within the matrix. However, high-temperature ECAP can eliminate completely the segregation of Zn element. Figure 2d shows the microstructure of regular-arranged rods array mixed up with interspersed

Fig. 2 TEM images of the tested Mg–6Zn alloy: **a** A1, **b**, **c** A2, **d** B1, **e**, **f** B2

blocky particles, very similar to the typical microstructure of Mg–Zn alloys after solid solution treatment. These rod-like precipitates with the length of 100–200 nm were generally

identified as the β' -Mg₄Zn₇ phase in the previous studies, and the blocky particles were regarded as β'' -MgZn₂ phase [31–33]. Furthermore, as shown in Fig. 2e and f, the rod-like

precipitates observed in the sample B1 completely disappear and are replaced by spherical precipitates with the diameter from 10 to 50 nm after the subsequent extrusion. Although these nanoprecipitates are too small to be identified by the selected area electron diffraction (SAED), the high-resolution TEM image (the inset in Fig. 2e) indicates that the nanoparticles possess the structure of β'' -MgZn₂ phase, showing the (313) plane of MgZn₂ phase with the interplanar spacing of 1.15 nm. Moreover, there are some sub-grains existing within the α -Mg grains with the size of about 200–300 nm from the larger-magnification micrograph, as shown in Fig. 2f. It is apparent that the sample B2 has much higher uniformity and more nanoparticles than the sample A2. In addition, it is noted that no dislocation is clearly observed in the TEM images of both the samples A2 and B2.

3.2 Difference of Grain Size, Dislocation Density and Texture

In addition to intermetallic phases, other microstructural characteristics such as grain size, dislocation density and texture are also essential to investigate the difference of effects between two technology routes. EBSD is performed on the transversal surface of the tested samples A2 and B2, where only α -Mg grains are marked. For each sample, at least 140 α -Mg grains are analysed to achieve credible results with statistical significance, as shown in Fig. 3.

The grain orientation mapping indicates, firstly, the great difference of grain size in two samples, as shown in Fig. 3a and c. The corresponding grain size statistics show that the average grain size of 2.3 μ m in sample A2 is far less than that of 10.8 μ m in sample B2, as shown in Fig. 3b and d, which is mainly because of the temperature difference during the ECAP process. Secondly, both A2 and B2 samples have uniform and equiaxial grains, as well as straight grain boundaries. It means that the fully dynamical recrystallization has happened after the last extrusion at 180 °C, and the integration of ECAP and extrusion can provide enough strain to eliminate the inhomogeneity of grain size [34, 35].

Figure 3e and g show the local misorientation mapping with different processing routes. In these large area mapping (LAM) maps, the black and light-green lines present high-angular grain boundaries (HAGBs) and low-angular grain boundaries (LAGBs), respectively. According to the kernel average misorientation (KAM) method, the local crystal orientations could be used to measure the geometrical necessary dislocations (GND). However, the very low densities of LAGBs are observed in the samples A2 and B2 and the statistics histograms show that the difference in the distribution of LAGBs between A2 and B2 is very small, as shown in Fig. 3f and h. This suggests that there is very low density of dislocation in both samples, which

also can be verified from the former TEM images. Figure 3i and j show the pole figures, and the extrusion direction (ED) is vertical to the plane of projection in figures. Both samples show the same texture components, similar to the $\{11\text{--}20\} \langle 0001 \rangle$ fiber texture and the texture intensity of A2 is 10.77, higher than that of 9.51 in B2.

Generally, the dislocation density and texture intensity of the sample A2 are very close the level of B2. The primary microstructural differences between A2 and B2 are the distribution of solute atom and the grain size and other factors are neglectable.

3.3 Mechanical Properties

Figure 4a shows the yield strength (YS), ultimate tensile strength (UTS) and elongation (EL.) of tested Mg–6Zn alloy. Processed through route A, the YS and UTS after low-temperature ECAP for six passes are improved to 275 MPa and 340 MPa, respectively. The subsequent extrusion at 180 °C has slight negative impact on both the strengths and the ductility for that processed via route A. The YS, UTS and elongation of sample A2 decrease to 200 MPa, 330 MPa and 10.5%, respectively. By contrast, the high-temperature ECAP results in the decreasing of strengths, but then the strengths dramatically increase after subsequent extrusion. The YS of 355 MPa and the UTS of 405 MPa, as well as the elongation of 17.5% are obtained in sample B2. Figure 4b shows the Vickers hardness of the tested Mg–6Zn alloys. The hardness of the Mg–6Zn alloy processed by route A is gradually improved and the sample A2 has the second highest hardness of 87.9. The highest hardness of 106.5 is obtained in the sample B2. According to the results of tension test at room temperature and hardness test, it is feasible that the ultrahigh-strength Mg–6Zn alloy is able to be prepared by the route B, the strengthening strategy for Mg–Zn alloy we conceived earlier.

3.4 Bone-penetration test

The bone-penetration test was conducted to assess the ability of resisting the deformation now that the ultrahigh strength has been achieved in Mg–6Zn alloy sample B2. The thin rods of sample A2 and sample B2 were machined into Kirschner wire with a sharp trigone end, as shown in Fig. 5a. The result of drilling is displayed in Fig. 5b. The Kirschner wire fabricated from sample B2 can drill successfully into the adult pig thigh bone but that of A2 fails. It can be seen that the end of B2 remains sharp after drilling test, as shown in Fig. 5c, while the end of A2 is distorted and blunted, as shown in Fig. 5d.

Fig. 3 EBSD analysis results of samples A2 and B2. **a, b** Grain orientation mapping and grain size histogram of A2, **c, d** grain orientation mapping and grain size histogram of B2, **e, f** local misorientation mapping and statistics of A2, **g, h** local misorientation mapping and statistics of B2, **i** pole figures of A2, **j** pole figures of B2

4 Discussion

4.1 Precipitation Behavior

The important and noteworthy microstructural characterization in the sample B2 was homogenization and high-density nanoprecipitation of MgZn_2 phase. The fully homogenization is the prerequisite for high-density nanoprecipitation. The high-temperature SPD could clearly accelerate homogenization and promote the single-phase structure in Mg–6Zn alloy when the temperature is up to the melting point of the intermetallic phase (about 325 °C) [36, 37]. After that, the increasing supersaturation of the solute resulted in the spontaneous precipitation of massive directional aligned rod-like nano-sized Mg_4Zn_7 phase from the single-phase supersaturate solid solution. The rod-like intermetallic phase plays a key role in inducing nano-sized spherical precipitates during the extrusion process because the long-rod particle might break into short-rod particles during extrusion and turn into spherical particles later, and the solution diffusion is involved possibly in their spheroidization and transformation [26, 38]. Another study revealed that the rod nanoparticles could also directly turn into spherical in ZK60 magnesium alloy as the aspect ratio decreased at very high strain rate [38]. Although the dynamic transformation mechanism is ambiguous, the rod-like particles are certainly a benefit for the quick and massive nanoprecipitation as precursors under subsequent large deformation at low temperatures. However, high-temperature ECAP processing cannot suppress the growth of α -Mg matrix grains as it dissolves secondary phase and inevitably leads to larger grains in the final thin bars.

In contrast, the low-temperature ECAP and extrusion processing are effective to produce ultrafine grained microstructure. However, it could not reduce the dimension of the second-phase particles and eliminate the segregation of the solute atoms. For instance, large bulk secondary phase changed into the clear banded structures after the ECAP process at 160 °C in this study, which results in the lower solid solution of zinc in α -Mg matrix than that processed by high-temperature ECAP and extrusion. Therefore, these are only a small number of precipitates existing in the matrix and very few are observed in the region far away from the banded structures after the subsequent extrusion.

4.2 Comparison of Mechanical Properties with Literature Results

Grain refinement is considered as one of the most effective methods to strengthen magnesium alloys and this is principal because the slope k in the Hall–Petch equation ($\sigma_y = \sigma_0 + k d^{-1/2}$) can be particularly high. The k coefficient in Mg alloys was estimated to amount to above 200 MPa (μm)^{1/2} in literature [39–41]. Moreover, Caceres et al. [42] found that the k value was strongly associated with the Zn content and it could reach 660 MPa (μm)^{1/2} for the Mg–2.3at.%Zn alloy (approximate to Mg–6 wt%Zn) when the grain size was over 50 μm . This is probably a good explanation of why the researchers were keen on refining grains in Mg alloys.

Table 2 summarizes the data on mechanical properties and microstructural characteristics of high strength Mg–6Zn based alloys after different kinds of thermo-mechanical processing. It suggests the consistent evidence of differentiated results from the two strategies. The combination of ECAP and extrusion is effective to produce the high-strength Mg alloy bar for now, such as route A and route B in this study, but the technical details decide the mechanism and its strengthening effects. The data listed in Table 2 show that limiting strengths in Mg–6Zn alloy processed by grain refinement processing are about 310 MPa in YS or 360 MPa in UTS. However, the route B in this study can easily make the strengths exceed the strength limit. What matters for the strengthening of Mg–Zn alloys is a large number of nanoprecipitates rather than ultrafine grains whether the nanoprecipitates are introduced by T6 heat treatment or low-temperature deformation [28, 29].

The hardening effect of grain refinement comes from that the grain boundaries obstacle against the moving of dislocations. However, the grain refinement strengthening of magnesium alloys is restricted by their lower stacking fault energy. On the one hand, the lower stacking fault energy results in the dislocations piled on grain boundary to be easily to recovery by cross-slip. On the other hand, the dimension of dynamic recrystallization grain is correlated with the size of sub-grain, which is depended on the width of extended dislocation. According to the records listed in Table 2, the minimum average grain size is 0.6 μm , which ought to have reached or been close to the limit of grain refinement.

Instead, the hardening effect of the precipitation comes from the particle's obstacle against the dislocations. The Orowan increment in critical resolved shear stress (CRSS) produced by the need for dislocations to by-pass these obstacles is given as [43]:

$$\Delta\tau = \frac{Gb}{2\pi\lambda\sqrt{1-\nu}} \ln \frac{d_p}{b}, \quad (1)$$

where $\Delta\tau$ is the increment in CRSS due to dispersion strengthening, G the shear modulus of the magnesium

Fig. 4 Histograms of tensile mechanical properties **a**, hardness **b** of the tested alloys

Fig. 5 Photos of the bone-penetration test: **a** the undrilled Kirschner wire, **b** the result after drilling, **c** the drilled Kirschner wire from sample B2, **d** the drilled Kirschner wire from sample A2

matrix phase, b the magnitude of the Burgers vector of the slip dislocations, ν the Poisson's ratio, λ the effective planar inter-obstacle spacing, and d_p the mean planar diameter of

the point obstacles. Nie et al. [43] developed an Orowan equation for spherical precipitates, the effective planar inter-precipitate spacing, λ , is calculated as:

$$\lambda = \left(\frac{0.779}{\sqrt{f}} - 0.785 \right) d_t, \quad (2)$$

where f is the volume fraction of precipitates, d_t is the diameter of spherical precipitates. The increment in YS due to Orowan strengthening, $\Delta\sigma$, is then obtained from the following relation:

$$\Delta\sigma = M\Delta\tau, \quad (3)$$

where M is Taylor factor, equals to 4.5 for the randomly textured Mg. Substituting Eqs. (1) and (2) into Eq. (3), and presuming the other parameters are constants, the relationship of $\Delta\sigma$ and the mean planar diameter of the point obstacles d_p are given as:

$$\Delta\sigma \in \frac{\ln(d_p)}{d_p}. \quad (4)$$

This relational expression shows that the $\Delta\sigma$ will be increased by about 11 times as the d_p is reduced by one order of magnitude when the d_p is less than one micrometer. Thus, the dispersion strengthening mechanism provides an opportunity to significantly improve the strengths to a surprising level when the specially designed technique is used to manufacture a tremendous number of nanoprecipitates.

4.3 Correlation Between Mechanical Properties and Microstructure

The strength improvement in sample B2 is a crucial progress for the orthopedic application of Mg–6Zn alloy in case of no successful bone-penetration test on Mg–Zn alloys before. More importantly, the sample processed by grain refinement, such as the sample A2 in this study, cannot pass the penetration test. It is apparent that the strength improvement can only be attributed to high-density nano precipitates because the present results illustrate that the texture, work hardening and grain size in sample B2 have a less strengthening effect than that in sample A2.

The strengths and hardness of the B2 Kirschner wire are improved to above a certain threshold so that the thin rod could withstand high stress during the drilling process. In contrast, the sample processed by the common strategies (i.e., the route A) is vulnerable to the bone even though its strength has been increased substantially. In light of this, it can be considered as a crucial progress in the aspect of mechanical properties of Mg–Zn alloys.

The average grains size depended strongly on the extrusion temperature during the multi-pass ECAP processing

Table 2 Mechanical properties and average grain diameters of the Mg–6Zn based alloys after different kinds of thermo-mechanical processing

	YS (MPa)	UTS (MPa)	El. (%)	Eq.dia.* (μm)	Material	Process	Refs.
1	273	300	30	0.6	ZK60	ECAP, 4p at 523 K + , 4p at 473 K + 4p at 423 K	[19]
2	270	326	5.4	1	ZK60	Rolling, 4p at 623 K, 4.3–1 mm	[20]
3	175	266	31.9	0.8	ZK60	ECAP, 4p at 513 K + 4p at 453 K	[21]
4	310	351	17.1	1.7	ZK60	Direct extrusion +ECAP, 2p at 623 K	[22]
5	215	289	37.8	1.4	ZK60	Cyclic extrusion–compression, 4p at 503 K	[23]
6	286	338	30	1.4	ZK60	High-ratio differential speed rolling, 1p at 473 K	[24]
7	260	371	18.5	1	ZK60	ECAP, 4p at 523 K + 2p at 473 K, + 2p at 423 K	[25]
8	258	359	20.1	1.2	Mg–5Zn–1Mn	HSRR, 1p at 573 K, 10–2 mm	[26]
9	275	340	15	1.9	Mg–6Zn	ECAP, 6p at 433 K	This study
10	190	330	10.5	2.3	Mg–6Zn	ECAP, 6p at 433 K + extrusion, at 453 K	This study
11	355	405	17.5	10.8	Mg–6Zn	ECAP, 6p at 593 K + extrusion, at 453 K	This study
12	390	440	8.3	–	ZK60	Twin roll cast + rolling, 6p at 573 K, 3.2–0.5 mm	[28]
13	333	420	16.1	6.8	ZK60	Twin roll cast + rolling, 6p at 573 K, 3.2–0.5 mm + T6, 3 h at 648 K and 10 h at 448 K	[28]
14	396	440	9	7	ZK60	ECAP, 12p 573 K + rolling, total reduction 30% at room temperature	[29]

*Eq.dia. is the abbreviation of equivalent grain diameter

[37]. The six passes ECAP at 160 °C produced the ultrafine grains by dynamic recrystallization, as well as strong texture and high-density dislocations. However, the follow-up extrusion at 180 °C inevitably caused the loss of mechanical properties of the ultrafine-grained sample A2 because the elevating of deformation temperature lowered these effects. On contrary, the high-temperature ECAP tended to construct a single-phase supersaturate solid solution microstructure and the subsequent extrusion dramatically improved the mechanical properties in the sample B2 by causing the dynamic precipitation of nano-size phase. By comparing the route A and route B, the difference between two strategies can be boiled down to that they strengthen the Mg–Zn alloy in two very different ways, and the latter way is undoubtable more effective and more promising.

It is worth exploring why the grain refinement cannot have the same strengthening effects with nanoprecipitation as the results shown in this study. Generally, shape, size and distribution of precipitates have dramatic effects on the strengthening. The extensive researches reported that the ultrafine-grained Mg–6Zn or ZK60 alloys have spherical or rod nanoprecipitates, but the segregation in grain boundaries and nonuniform distribution of the particle phase can also be observed. In addition, it may also be attributed to the reduced dispersion strengthening caused by the coarsening of precipitates during the multiple-pass plastic deformation. The Gibbs–Thomson effect illustrates that the nanoprecipitate is thermodynamically unstable due to the dependence of the corrected solubility limit of the solute in the matrix on the radius of precipitation particle [44, 45]. It means that the smaller precipitates prefer to dissolve into the matrix

and then the solute atoms diffuse and segregate on the surface of the bigger precipitates or on the grain boundaries. Moreover, it will be accelerated by deformation and elevated temperatures. As a consequence, it can be concluded that the mechanical properties can get great improvement only when the latter extrusion after full homogenization is carried out at a higher strain rate and a lower temperature.

Based on above discussions, it is considered for Mg–Zn alloys that promoting the microstructure of nanoprecipitation followed by fully homogenization is a more effective strengthening method than merely grain refinement. In other words, the combination technology of deformation-induced homogenization treatment and low-temperature large deformation processing is more competitive and productive than the previous desultory attempts for the development of high-mechanical properties Mg–Zn alloys.

5 Conclusions

In this study, a comparison with the two processing routes based on different strategies shows vastly difference of microstructures and mechanical properties in Mg–6Zn alloys and results confirm that thermal–mechanical treatment has great potentiality in strengthening Mg–Zn alloys. The following conclusions can be drawn:

- (1) The improved strategy based on deformation-induced nanoprecipitation (six-pass ECAP at 320 °C and extrusion at 180 °C) provided the yield strength of 355 MPa and the ultimate tensile strength of 405 MPa and the

elongation of 17.5%, as well as the Vickers hardness of 106.5, which are much higher than the sample processed by comparative strategy.

- (2) The ultrahigh mechanical properties were principally depended on a large number of uniformly distributed MgZn₂ phase precipitates with the diameter 10 nm to 50 nm instead of grain refinement, strong textured microstructure and working hardening.
- (3) The improvement of mechanical properties is a crucial factor for Mg–6Zn alloy Kirschner wire penetrating through the adult pig thigh bone, which signs the breakthrough of high strength Mg–Zn alloy orthopedic devices.

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