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ARTICLE

Microstructure and Tensile Properties of Solutionized Mg-3.52Sn-3.32Al Alloy Deformed by Equal Channel Angular Pressing

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Abstract: The solutionized Mg-3.52Sn-3.32Al alloy was processed by equal channel angular pressing (ECAP) via route Bc for 1, 4, and 8 passes. The microstructure and phase composition of the alloy were analyzed by optical microscope, scanning electron microscopy, transmission electron microscopy and X-ray diffraction, and the room-temperature mechanical properties were measured. The results show that many fine Mg_2Sn particles and a few $Mg_{17}Al_{12}$ phases precipitate in the alloy after ECAP. The mechanical properties first increase and then gradually decrease with increasing the extrusion passes. The alloy possesses better mechanical properties after ECAP for four passes, and the ultimate tensile strength, elongation and hardness $HV_{9.8}$ increase to 250 MPa, 20.5% and 613 MPa, respectively, which are increased by 43.7%, 105% and 26.9% compared with that of the solutionized alloy, respectively. The room temperature fractograph of the ECAP processed alloy is a ductile-fractured morphology under tensile conditions. The mechanical properties of the ECAP processed Mg alloy depend on the grain size, precipitate and texture.

Key words: magnesium-tin-aluminum alloy; equal channel angular pressing; microstructure; mechanical property

Due to the low density, high specific strength and specific stiffness, high moduli of elasticity, excellent damping capacity, good dimensional stability, good machinability and outstanding anti-electromagnetic interference capacity, magnesium alloys have been found to be widely used in 3C (computer, communication and consumer electronic) industry and automobile industry[1]. At present, the main problems needed to be solved in the application of magnesium alloys are low ductility and poor thermal properties. Therefore, most magnesium alloys cannot be used for components working at high temperatures for a long period of time because the strength and creep resistance decrease with increasing temperature considerably. So far, the magnesium alloys with excellent elevated temperature properties are Mg-RE-based alloys. For example, working temperature of the earlier developed Mg-Y-Nd-based alloys, such as WE43 and WE54 magnesium alloys, can reach 573 K. Mordike[2] has shown that the creep resistance of Mg alloys containing a high percentage of gadolinium is better than that of WE43 alloy even at a high temperature of 573 and 623 K. However, compared with other alloys, the Mg alloys containing rare earth are expensive in price, so the Mg alloys without rare earth have been a hot-spot in the development of heat-resistant magnesium alloys.

In recent years, Mg-Sn based alloys have attracted signi-

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ficant interest due to their excellent performance [3,4]. As freezing over a narrow temperature range, Mg-Sn based alloys have a small porosity and hot tearing tendency. The solid solubility of tin in magnesium varies significantly with temperature, and thus the magnesium alloys doped with tin can be strengthened by aging strengthening. Furthermore, the melting point of Mg_2Sn, $Mg_{17}Al_{12}$ and $MgZn_2$ is 1043.5, 710 and 863 K, respectively, and as strengthening phases, the thermal stability of Mg_2Sn is higher than that of $Mg_{17}Al_{12}$ and $MgZn_2$. Previous investigations have shown that Mg_2Sn compound has a stable structure that does not transform with rising temperature in the temperature range of 373~473 K[5]. Among Mg-Sn binary alloys, Mg-3.5Sn alloy exhibits excellent mechanical properties at room and elevated temperatures. As aluminum is added to these alloys, the dentritic grains of α-Mg are refined, and a small amount of fine-grained $Mg_{17}Al_{12}$ phases are precipitated. As a result, the mechanical properties of Mg-Sn-Al alloys at both room and elevated temperatures are improved. Wang[6] investigated the effect of heat treatment on the microstructure and mechanical properties of vacuum die-casting Mg-7Al-2Sn alloy. The results showed that a solution treatment was carried out at 673 K for 20 h, and most of the $Mg_{17}Al_{12}$ phase was taken into solution. The precipitation strengthening effect of particle Mg_2Sn phases is more apparent due to the dissolution of $Mg_{17}Al_{12}$. Whereas, the strength and ductility of the alloy are reduced because of the discontinuous precipitation of brittle $Mg_{17}Al_{12}$ phases and the coarsening of Mg_2Sn phase after aging at 473 K for 15 h.

As an innovative technique, the severe plastic deformation, especially equal channel angular pressing (ECAP), has attracted considerable attention. ECAP can not only remove cast microstructure deteriorations and effectively refine the matrix structure, but also decompose large reinforcements and disperse them in matrix, and thus the mechanical properties can be improved[4,6-8]. It was found that the shear planes are perpendicular to each other in two adjacent passes of ECAP via route Ba or Bc[9]. These two routes favor fragmentation of large grains, so the grain refinement is very effective. In particular, as a widely applied method of severe plastic deformation, ECAP is a favorable way to acquire the most uniform accumulated strain in alloy bars by route Bc.

Zheng and Wang et al[7] investigated the microstructure and elevated-temperature mechanical properties of Mg-15Al-xNd alloys that have undergone ECAP for four passes following route Bc. The results indicate that after ECAP, the average grain size is reduced from 30~40 μm to 2~3 μm and the β-$Mg_{17}Al_{12}$ phase is significantly refined. The grain refinement strengthening and dispersion strengthening are the main mechanisms for the improvement of the properties, which benefits from the pinning effect of the refined dystectic Al-Nd phases at grain boundaries. ECAP was successfully performed on Mg-3Sn-0.5Mn alloy by Pan et al[4] using route Bc. The results showed that the alloy with ultrafined grain microstructure can be obtained by ECAP process. After 4 passes of ECAP at 320 °C, the average grain size of the alloy reached 1.65 μm, and the local grain size reached 0.8 μm. The grain refinement was caused by severe plastic deformation and recrystallization. The elongation increased from 22.2% to 58.6%, and the tensile strength decreased from 242.8 MPa to 195.6 MPa after 4 passes of deformation. The mechanical tests showed an improvement in the elongation after 4 passes, and it was about 3 times higher than that of as-extruded state, whereas the tensile strength decreased slightly. The mechanical property of the alloy after ECAP was affected by the texture orientation and grain strengthening.

Mostaed et al investigated the effect of ECAP processing on mechanical properties of pure magnesium and ZK60 alloy[10]. They argued that ECAP can result in a remarkable grain refinement and texture modification, which leads to adequate mechanical properties with about 100% improvement in elongation to failure while keeping relatively high tensile strength. The original extrusion fiber texture with planes oriented parallel to extrusion direction was gradually undermined during ECAP process and eventually replaced by a newly stronger texture component with considerably higher intensity, coinciding with ECAP shear plane.

In order to improve mechanical properties of Mg-Sn-Al alloys and understand the effect of ECAP, the Mg-3.52Sn-3.32Al alloy was processed by ECAP via route Bc, and microstructures as well as mechanical properties of the alloy processed were investigated in this paper.

1 Experiment

1.1 Alloy preparation

Mg-3.52Sn-3.32Al magnesium alloy (wt%) was melted by commercial pure magnesium, pure tin and pure aluminum in a mild steel crucible using an electric resistance furnace. Pure magnesium was placed in the crucible and melted first. After that, proper amounts of the pure tin and pure aluminum were added when the melt was at 1003 K. During melting, the melt was manually stirred every 10 min for 3 min to obtain a liquid of essentially uniform composition. A refining agent (a mixture of NaCl, KCl, $MgCl_2$, etc) was then added to the melt and stirred for 2 min. After that, the melt was heated to 1023 K and stewed for 20 min, and then it was cooled down in the crucible in air. Eventually, the melt was poured into a cast iron mold preheated to 523 K to produce an ingot immediately after dross was skimmed off at 988 K. During melting and pouring processes, the melt was protected by a mixed gas atmosphere of ulphur hexafluoride (SF_6, 0.2 vol%) and carbon dioxide (CO_2, Bal.).

The ingot was put into a heat-treatment furnace and coated with fine graphite powder, and then it was solution treated at 723 K for 24 h. The solutionized ingot was machined into round bars with 16.5 mm in diameter and 150 mm in length, and then the bars were processed into rods with 14 mm in diameter by ECAP with an internal angle of 90° between the vertical and horizontal channels and a curvature angle of 30° via route Bc (between every two adjacent passes, the bars were rotated 90° clockwise around their longitudinal axis) for 1, 4 and 8 passes. Fig.1 shows the schematic diagram of ECAP process. The channel of the die is circular with diameters of 17.0 and 16.5 mm at the points of entrance and exit, respectively. After the bars were heated at 603 K for 30 min, ECAP was carried out at an extrusion speed of 15 mm/min using a molybdenum disulfide grease as a lubricant. Then the bars were released intermediately from the die to cool in air.

1.2 Microstructure observation & property measurement

The samples used for microstructure observation were polished and etched with a 4% nital solution, and the morphology was observed by an optical microscope (OM, Olympus GX71) and a scanning electron microscope (SEM, JSM-6390A). The fracture morphologies were observed by the SEM. The phase composition was analyzed by energy dispersive X-ray spectroscopy (EDS) under the SEM and X-ray diffraction (XRD, Shimadzu-XRD7000). Morphological characteristic and distribution of precipitates were analyzed using a transmission electron microscope (TEM, JEM-3010).

Microhardness was measured by a Vickers hardness tester at 2 N load and 20 s dwell time. The hardness of the alloy was measured at 10 positions under the same conditions. The average values from ten tests were reported. The tensile mechanical properties were tested at a crosshead speed of 2 mm/min at room temperature, and the values reported in this paper were averages from three tests.

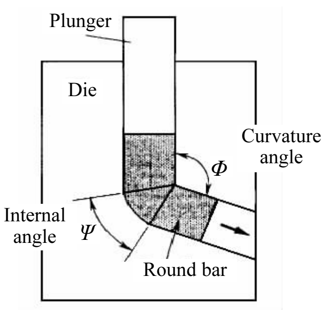


Fig.1 Principle diagram of ECAP

2 Results

2.1 Microstructure and phase constitution

Fig.2 shows the microstructure of the solutionized Mg-3.52Sn-3.32Al alloy subjected to different ECAP passes at lower magnification. After ECAP deformation for 1 pass, a streamline formed because grain elongation and refinement occurred by shear deformation. The microstructure in this case was due to the twist rotation and slip of the fragmented dendrite arms of α-Mg under extrusion pressure[11]. Dark regions are fine second-phase precipitates distributed along the extrusion streamlines. After 4 passes, more secondary particles precipitated and were homogeneously distributed. After 8 passes, the alloy was subjected to shear deformation two times along each direction in three dimensions. As a result, no streamlines in the microstructure could be seen.

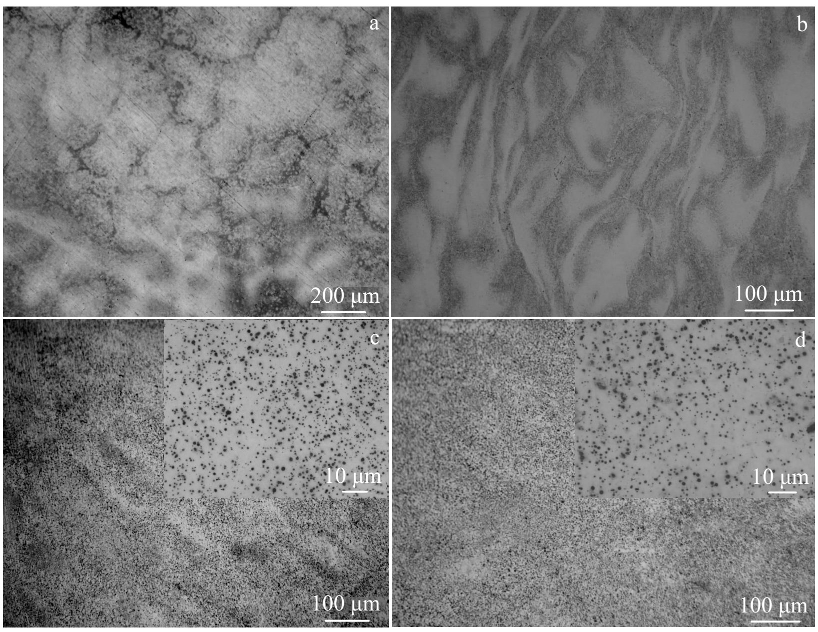


Fig.2 Optical microstructures of the Mg-3.52Sn-3.32Al alloy subjected to different ECAP passes (insets show the corresponding enlarged parts of the microstructures): (a) solid solution, (b) 1 pass, (c) 4 passes, and (d) 8 passes

In order for clearer observation, SEM was used to examine the microstructures. From the high magnification SEM photograph (Fig.3a) of the solutionized Mg-3.52Sn-3.32Al alloy, it can be seen that the microstructure is homogeneous, and the metallographic examination does not show the presence of the second-phase particles. This indicates that both Sn and Al may be dissolved in α-Mg matrix. Micron-size short rod-like compounds appeared after 1 pass (Fig.3b). After 4 passes, more intermetallic phases precipitated and their rod-like morphology became nearly round shape (Fig.3c). In addition, one can see that the number and size of precipitates gradually increase with increasing the number of ECAP passes (Fig.3d).

Fig.4 shows the XRD patterns of the solutionized Mg-3.52Sn-3.32Al alloy before and after ECAP processing. Evidently, only α-Mg solid solution exists in the solutionized alloy. After the alloy was deformed by ECAP, Mg_2Sn precipitates appeared in the microstructure. With the increase of extrusion passes, the intensity reflection of α-Mg changed, such as (100), (002) and (110)-type planes. This implies that ECAP deformation leads to microstructure orientation and formation of new texture orientations, i.e. cylinder orientation decreases, whereas basal and pyramidal orientations increase. The increased basal and pyramidal orientations are favorable for the activation of slip system and improvement of plasticity.

Fig.5 shows the SEM image and EDS analysis of the solutionized alloy subjected to 4 passes of ECAP. The microstructures are characterized by a large number of bright granular second phases (indicated by arrow 2) and a very few

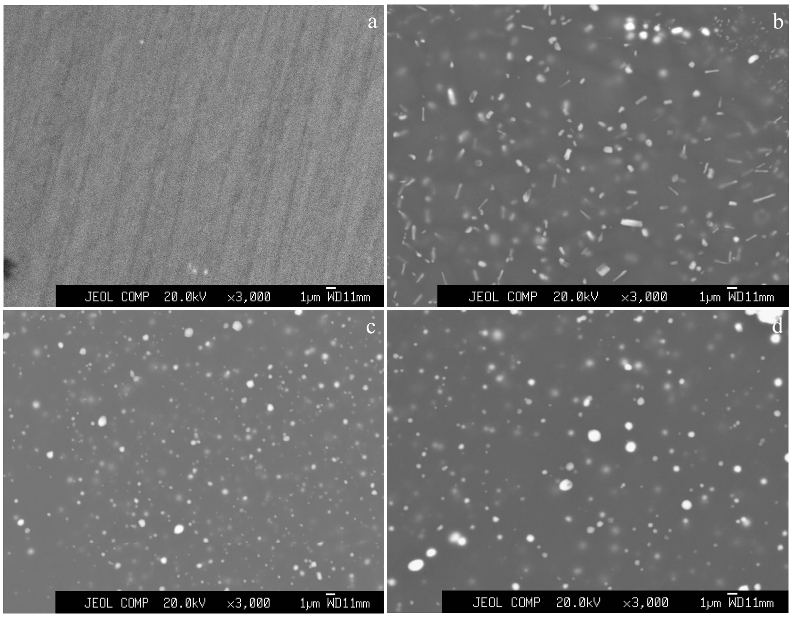


Fig.3 SEM micrographs of the Mg-3.52Sn-3.32Al alloy subjected to different ECAP passes: (a) solid solution, (b) 1 pass, (c) 4 passes, and (d) 8 passes

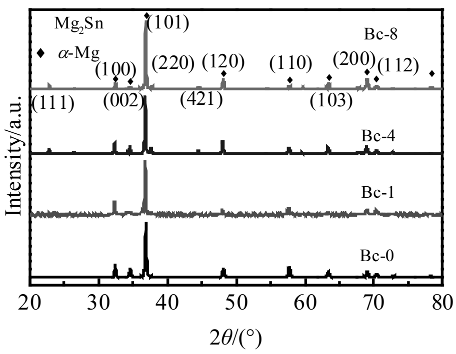
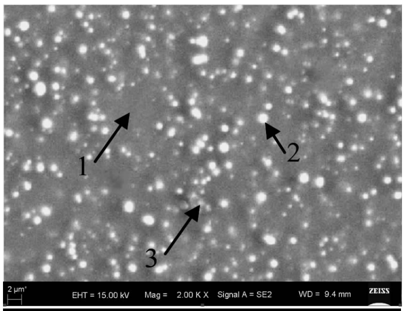


Fig.4 XRD patterns of the Mg-3.52Sn-3.32Al alloy subjected to different ECAP passes

grey blocky second phases (indicated by arrow 3) dispersed in the matrix (indicated by arrow 1). The EDS analysis shows that the matrix is α-Mg solid solution containing a small amount of tin and aluminum. The bright granular second phase contains high percentages of magnesium and tin as well as a small amount of aluminum. This result reveals that the bright granular second phase is Mg_2Sn compound.

Fig.6 shows the SEM image, TEM micrograph and SAED pattern of second phase of the solutionized alloy subjected to different passes of ECAP. It can be seen that after ECAP for 1 pass, the grain size of the alloy is refined, and a large amount of small dispersed second phases appear (Fig.6a). After ECAP for 4 passes, further grain refinement is achieved. Grain boundaries and second phases can be observed clearly by TEM (Fig.6b). It was confirmed by SAED pattern and XRD



Point	Mass fraction/%		
	Mg	Al	Sn
1	94.64	2.13	3.23
2	56.81	1.86	41.33
3	57.77	40.05	2.18

Fig.5 SEM image and EDS analysis of the Mg-3.52Sn-3.32Al alloy subjected to 4 pass ECAP

analysis that the bright granular second phases dispersed in the matrix are Mg_2Sn compounds.

2.2 Mechanical properties

The tensile stress-strain curves for the ECAP processed Mg-3.52Sn-3.32Al alloy is shown in Fig.7a. During the initial elastic stage, the stress increases linearly with strain. Beyond the elastic stage, the tensile stress increases with increasing the degree of deformation due to work hardening. The dependence of the tensile strength, elongation and hardness on the number of ECAP passes for the alloy is shown in Fig.7b and 7c.

Fig.7b shows that the tensile strength and elongation of solutionized Mg-3.52Sn-3.32Al alloy are 174 MPa and 10.0%, respectively. The tensile mechanical properties are improved remarkably by ECAP. With the increase of ECAP passes, the mechanical properties increase first and then decrease gradually. The mechanical properties are most desirable for the four passes ECAP processed alloy, and the tensile strength, elongation and hardness $HV_{9.8}$ reach 250 MPa, 20.5% and 613 MPa, respectively, increased by 43.7%, 105% and 26.9% compared with the solutionized state, respectively. Whereas, after eight passes of ECAP, the tensile strength, elongation and hardness $HV_{9.8}$ decrease to 225 MPa, 17% and 544 MPa, respectively.

Hardness is a measure of the resistance to localized plastic deformation. The hardness of the 4 passes ECAP processed alloy is the highest, which indicates that the alloy has the optimum resistance to plastic deformation. In general, elongation decreases as the tensile strength increases. However, the results of the tensile stress-strain test (Fig.7b) reveal that the elongation and tensile strength of the 1 and 4 pass ECAP processed alloys increase simultaneously. This implies that ECAP can improve the ductility and tensile properties simultaneously.

2.3 Fatigue fractograph

Fig.8 shows the SEM fracture morphologies of the solutionized Mg-3.52Sn-3.32Al alloy subjected to different passes of ECAP under room temperature tensile conditions. One can see that the tensile fracture surfaces of the solutionized alloy separate between grains; a large number of the cleavage surfaces, tearing ridges and a few dimples exist on the fracture surfaces (Fig.8a). This fracture morphology indicates that the tensile failure of the solutionized alloy appears to be a characteristic of quasi-cleavage fracture. After 1 pass of ECAP process, the number of cleavage surfaces decreases, whereas the number of dimples increases obviously, and second-phase particles are distributed in the dimples (Fig.8b); this result shows that the ductility of the alloy is enhanced. After 4 passes, no cleavage steps were observed on the tensile fracture surface, whereas a small number of second-phase particles and a large amount of uniformly distributed small dimples can be seen clearly (Fig.8c); these features indicate that the tensile failure appears to be a characteristic of ductile fracture. After 8 passes, the size and depth of dimples are different, and a small amount of cleavage steps exist on the fracture surface (Fig.8d). It can also be noticed that some of dimples are elongated and second-phase particles are distributed within dimples. Consequently, the eight passes ECAP processed alloy exhibits a ductile failure mode.

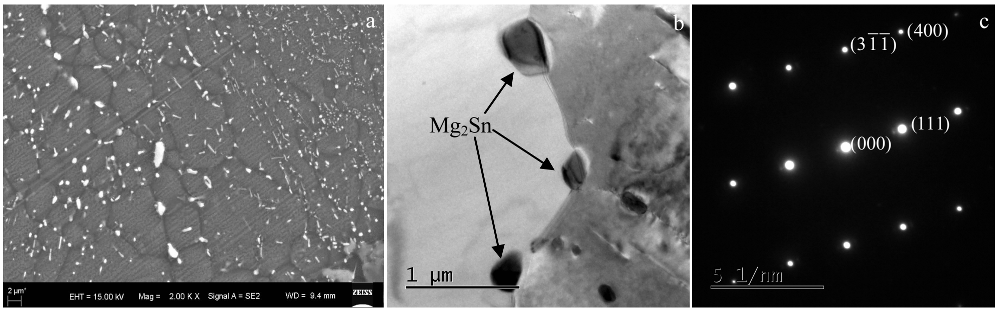


Fig.6 SEM (a) and TEM (b) images of the Mg-3.52Sn-3.32Al alloy subjected to different ECAP passes and corresponding SAED pattern (c) of the second phase: (a) 1 pass, (b, c) 4 passes

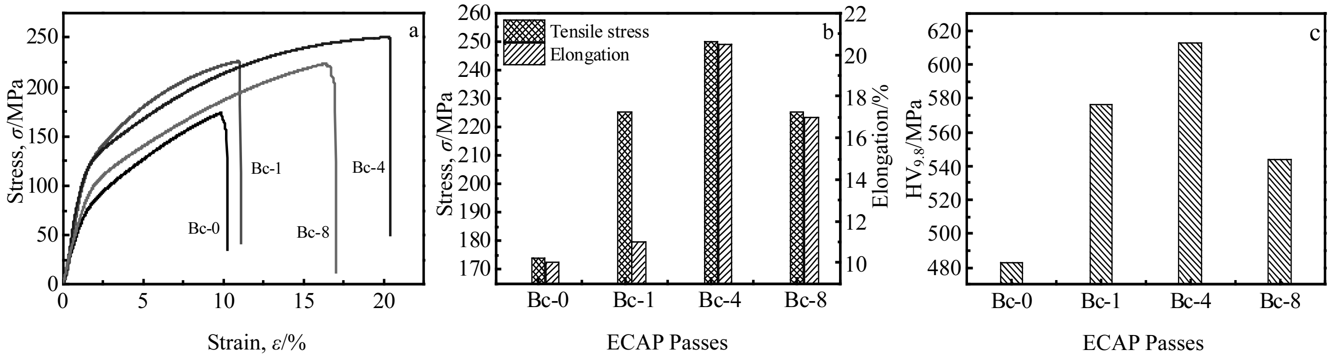


Fig.7 Tensile stress-strain curve (a), strength & elongation (b), and Vickers hardness (c) for the Mg-3.52Sn-3.32Al alloy subjected to different ECAP passes

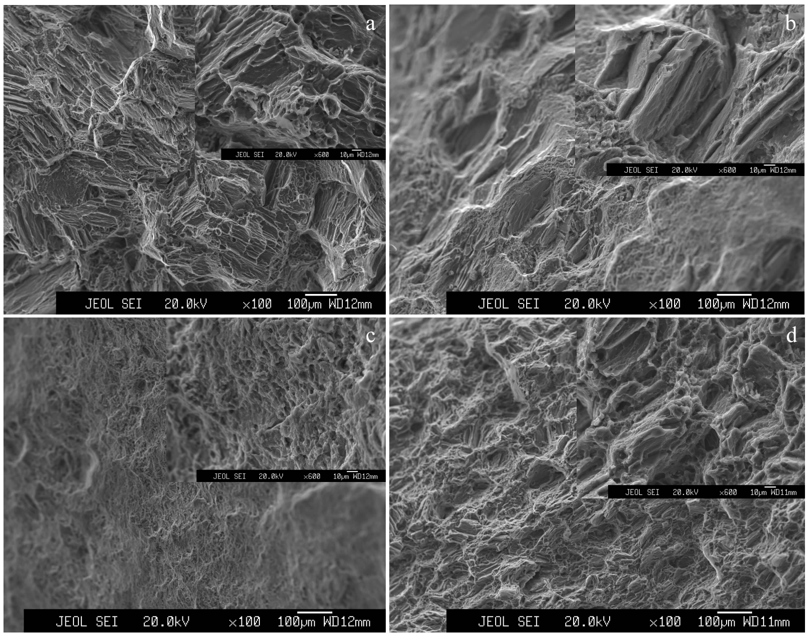


Fig.8 Fracture morphologies of the Mg-3.52Sn-3.32Al alloy subjected to different ECAP passes (the insets in the upper right corners are the high magnification fracture structure): (a) solid solution, (b) 1 pass, (c) 4 passes, and (d) 8 passes

3 Discussion

3.1 Phase composition

According to the Mg-Sn binary phase diagram[12], tin can react with magnesium to form only one kind of intermetallic compound, i.e. Mg_2Sn. It can be seen from the region on the magnesium-rich side of binary phase diagram of Mg-Al that dendrites of α-Mg solid solution form first[12]. As the temperature decreases, $Mg_{17}Al_{12}$ begins to precipitate from the α-Mg phase. As a result, the as-cast Mg-Al alloy consists of dendritic α-Mg phase and $Mg_{17}Al_{12}$ precipitate phase. According to the Al-Sn binary phase diagram, tin can not react with α-Mg to form any compound[12,13].

For the Mg-3.52Sn-3.32Al alloy, the host atom is magnesium whose mass fraction is as high as 93%. Consequently, the reactions only occur between Mg and Sn as well as Mg and Al. According to the binary phase diagrams, the solubility of elements Al and Sn in Mg at 603 K is 8% and 2.5%, respectively [12]. Previous studies have showed that the addition of Sn increases the solubility of element Al in Mg[14]. Recent researches have showed that dispersed Mg_2Sn phase can inhibit the precipitation of divorced eutectic β-$Mg_{17}Al_{12}$ phases[15]. Thus the amount of $Mg_{17}Al_{12}$ precipitates is few in the alloy. In addition, it is well-known that the formation of compound between elements depends on their electronegativity difference[16]. Larger electronegativity difference tends to make the binding force stronger and therefore it is easier to form a compound. The electronegativities of magnesium, tin and aluminum are 1.31, 1.96 and 1.61, respectively [17]. So the formation of compound between magnesium and tin may be easier. This prediction is consistent with the results from XRD and SEM (Fig.3 and Fig.5). According to the Mg-Sn-Al ternary phase diagram[12], primary α-Mg initially forms during the solidification process of the liquid Mg-3.52Sn-3.32Al alloy, and the further decrease in temperature causes a rejection of alloying elements from the α-Mg dendrite at liquid/solid interface. As a result, the solute concentration in the liquid at the solid/liquid interface increases gradually. As the temperature is lowered to 701 K, a ternary eutectic reaction ($L \rightarrow \alpha\text{-}Mg+Mg_2Sn+Mg_{17}Al_{12}$) occurs. Thus the cast alloy consists of three phases, i.e. α-Mg, Mg_2Sn and $Mg_{17}Al_{12}$.

When the cast Mg-3.52Sn-3.32Al alloy was subjected to a solution treatment at 723 K for 24 h, Mg_2Sn and $Mg_{17}Al_{12}$ phases were all dissolved in the α-Mg matrix. During ECAP process, approximate pure shear produced severe plastic deformation and grain fragmentation with the result that the dynamic recrystallization of α-Mg favorably occurred, and consequently the grains were refined. In addition, a significant amount of strain-induced lattice defects (e.g. dislocations and subboundaries) were also induced in grains by deformation. These defects can provide favorable sites for the precipitation of second phases and enhance the diffusion of alloying elements[18~22]. Consequently, dispersive Mg_2Sn and $Mg_{17}Al_{12}$ phases precipitated homogeneously during ECAP process.

However, X-ray technique did not detect $Mg_{17}Al_{12}$ phases (Fig.4). This phenomenon can be attributed to the trace amount of $Mg_{17}Al_{12}$ phase which is beyond the detection limit of the XRD patterns. In addition, because $Mg_{17}Al_{12}$ phase is so fine that the EDS measurement inevitably contains matrix elements, the EDS analysis of the $Mg_{17}Al_{12}$ phase showed that there is a small amount of Sn element (Table 1).

With the increase in extrusion pass, the accumulated strain increases gradually, and the grain fragmentation is strengthened by shear deformation. As a result, the grains are further refined and uniformized accordingly; whereas, the second phase coarsens correspondingly though the phase composition is unchanged.

3.2 Variation of mechanical properties

Only a single α-Mg phase solid solution is present in the solutionized Mg-3.52Sn-3.32Al alloy, and its ambient tensile mechanical properties are limited because only tin and aluminium atoms strengthen the alloy by solution strengthening.

During ECAP, dislocation movement occurred along the basal slip system and non-basal slip systems in response to an applied shear stress. When the dislocation moved to the prior grain boundaries and second phases, its displacement was restricted, which resulted in the dislocation pinning and pile-up. Subsequent incorporation and rearrangement of dislocations led to the creation of numerous sub-grains. The sub-boundaries could transform to high-angle grain boundaries by dislocations absorption and an increase in orientation differences between adjacent grains. So, with the increase of ECAP passes, the grain size of the alloy decreased. After extrusion, the ECAP processed alloy was released intermediately from the die to cool in air, and thus there was no time for grain growth to occur, so the refined grain structure was retained.

In addition, the interaction between dislocations and dispersive precipitates may impede the motion of dislocations and effectively block slips and thus strengthen the resistance to plastic deformation. Hence, we can conclude that the high strength of the ECAP processed Mg-3.52Sn-3.32Al alloy can be attributed to fine-grain strengthening and dispersion strengthening.

Because uniform plastic deformation and grain rotation easily occur in a fine-grained alloy, a higher degree of plastic deformation can be sustained. Furthermore, fine grains can distribute the stress concentration and strain concentration, and delay or even hinder the initiation of crack. Consequently, the crack is more difficult to propagate through the fine grain structure than through coarse grain structure, so the fine-grained alloy exhibits high ductility and elongation to failure.

However, although the accumulated stain increased after eight passes, the grain of the alloy was no longer refined, and inversely the ultimate strength fell somewhat.

Previous studies also have shown that the strength of magnesium alloys decreases even though the grains are refined by severe plastic deformation[23,24].

3.3 Effect of ECAP on mechanical properties

The mechanical properties of ECAP processed alloy have been found to depend on the grain size, precipitates and texture[22]. The correlation between strength and grain size is in accordance with the widely quoted Hall-Petch formula. According to the formula, it is known that the strength of the materials usually increases with the decrease of the average size of grains. Whereas, there is a reverse Hall-Petch relationship between the strength and the grain size of the ultra-fine grained Mg alloy[4]. The texture orientation affects the slip system and the resolved shear stress, and has an significant effect on mechanic property of magnesium alloy[4,23,25]. Kang et al[23] reported that shear texture was formed during ECAP process. The texture intensity increases with increasing the number of ECAP passes.

ECAP deformation led to change in the texture orientations of Mg-3.52Sn-3.32Al alloy. The mechanical property of the alloy after ECAP was affected by the texture orientation and grain strengthening. It is well known that grain refinement is an effective approach to increase the mechanical properties, whereas texture softening has an adverse effect on strength of an alloy. The new texture orientations induced by ECAP in the alloy are favorable for the activation of slip system and plastic slip, which in turn causes the improvement of plasticity and an adverse effect on strength. Grain refinement is more significant than texture softening at low ECAP passes; however, texture softening is more significant than grain refinement at high ECAP passes. As a result, the ultimate strength of the ECAP processed Mg-3.52Sn-3.32Al alloy decreased somewhat after multiple passes.

Additionally, the dispersive fine precipitate is an effective impediment to the movement of dislocation and is beneficial for the improvement of mechanical properties. It should be noted that when the alloy was subjected to high number of ECAP passes, the holding-time and extrusion time were prolonged and precipitates coarsened gradually as the passes increased. Coarse dispersoids tend to cause weakening of alloy and lead to a reduction in strength and ductility.

Under these effects, the tensile strength of the ECAP processed Mg-3.52Sn-3.32Al alloy for eight passes is lower than that for four passes. However, as the grain size of the ECAP processed alloy is significantly refined and the dislocation densities are low now, the deformation mechanism of the alloy may transform from prior dislocation slip to grain boundary sliding, so the alloy still shows high elongation.

As mentioned above, after ECAP at low number of passes, the strength of the magnesium alloy is improved due to the enhancement of fine grain strengthening and precipitate strengthening. Whereas, after ECAP at high number of passes, the strength decreases, which can be attributed to the combined deleterious effect of texture softening and precipitates coarsening.

4 Conclusions

1) After ECAP, a large number of fine, granular and uniformly distributed Mg_2Sn phases and a few $Mg_{17}Al_{12}$ phases precipitate from the solutionized Mg-3.52Sn-3.32Al alloy.

2) As the number of ECAP passes increases, the mechanical properties of the solutionized alloy increase to a maximum at 4 passes and then decrease. The 4 pass ECAP processed alloy shows excellent mechanical properties, and the tensile strength, elongation and hardness $HV_{9.8}$ reach 250 MPa, 20.5% and 613 MPa, respectively, increased by 43.7%, 105% and 26.9% compared with the solutionized alloy, respectively.

3) The mechanical properties of the ECAP processed Mg alloy are dependent on the grain size, precipitate and texture. The fracture mechanism of the solutionized alloy is quasi-cleavage fracture, while the fracture form of the ECAP processed alloy is ductile fracture.

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等通道转角挤压 Mg-3.52Sn-3.32Al 合金的组织与力学性能

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摘　要：采用等通道转角挤压（ECAP）Bc 路径对固溶态 Mg-3.52Sn-3.32Al 合金分别挤压 1、4 和 8 道次。利用光学显微镜、扫描电子显微镜、透射电子显微镜和 X 射线衍射仪分析合金的组织和相组成，并测试了其室温拉伸性能。结果表明，经 ECAP 挤压后，固溶态合金组织中析出大量细小的 Mg_2Sn 相和极少量的 $Mg_{17}Al_{12}$ 相。随挤压道次增加，合金的综合力学性能先提高后降低。经 4 道次挤压后，合金的综合拉伸性能相对较佳，抗拉强度、伸长率和硬度 $HV_{9.8}$ 分别达到 250 MPa、20.5%和 613 MPa，较未 ECAP 时分别提高 43.7%、105%和 26.9%。经 ECAP 挤压的合金室温拉伸断口均呈韧性断裂。等通道转角挤压 Mg-3.52Sn-3.32Al 合金的力学性能受晶粒尺寸、析出相以及组织织构的共同影响。

关键词：镁-锡-铝合金；等通道转角挤压；组织；力学性能

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