Kovove Mater. 49 2011 37–42
DOI: 10.4149/km_2011_1_37

Investigation of microstructure and strength of AZ80 magnesium alloy by ECAP and aging treatment

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Received 20 July 2010, received in revised form 10 August 2010, accepted 10 August 2010

Abstract

Grain refinement by equal channel angular pressing (ECAP) and aging was combined to strengthen a wrought Mg alloy AZ80. The resulting mechanical properties and microstructure were examined by uniaxial tensile test, optical microscope (OM) and transmission electronic microscope (TEM). It is found that ECAP for 8 passes at 633 K followed by aging treatment at 423 K is the optimum condition to strengthen AZ80 alloy. Microstructure analysis reveals that many discontinuous precipitates appear at grain boundaries, accompanied by a few continuous precipitates in grain interior. The strengthening effect of this combination method is not quite remarkable in comparison with simple fine-grain strengthening by ECAP without aging, which is probably due to the contradiction between the low temperature required for grain refinement by ECAP and the high temperature needed for solid solution during ECAP.

Key words: magnesium alloy, ECAP, aging, precipitate, strengthening

1. Introduction

As the lightest structural metallic material, magnesium alloy has wide prospects in applications for high specific strength, good damping properties [1–3]. However, poor ductility and low strength restrict its wide applications in structural field. It has been well established that both fine-grain strengthening based on Hall-Petch relation [4, 5] and precipitate strengthening [6–8] can remarkably improve mechanical properties of metallic materials. Recently, the strengthening of magnesium alloys has been investigated by a number of studies using fine-grain strengthening by equal channel angular pressing (ECAP) [9–12] and precipitate strengthening by aging treatment [13–15] independently. However, research combining the above two strengthening approaches for magnesium alloys has been rarely reported. In order to strengthen magnesium alloys more significantly, a commercial Mg-Al-Zn alloy, AZ80, which can be strengthened by heat treatment, was selected for investigation. The AZ80 Mg alloy was first ECAP processed at elevated temperatures to refine the microstructure and produce supersaturated solid solution simultaneously, then directly aged at a lower temperature to superimpose precipitate strengthening on fine-grain strengthening to realize the maximum strengthening effect. Based on the resulting mechanical properties, OM and TEM were employed to observe the microstructure change and precipitation characteristics for better understanding of the complex strengthening effect of AZ80 Mg alloy.

2. Experimental material and procedure

The as-received material is a commercial cast AZ80 Mg alloy with the chemical composition listed in Table 1. To avoid the negative influence of segregation in the as-cast microstructure (Fig. 1a) on the following ECAP, homogenization treatment in vacuum furnace was conducted at 693 K for 16 h before ECAP. The resulting microstructure without obvious segregation is shown in Fig. 1b.

To reduce the grain size, AZ80 rods after homogenization were ECAP processed by route A, namely no rotation around the rod axis after each pass. The samples were processed to 1P, 4P, 6P and 8P, respect-

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Table 1. Chemical composition of AZ80 Mg alloy (wt.%)

Al	Zn	Mn	Si	Cu	Mg
7.80–9.20	0.20–0.80	0.12–0.50	0.10	0.05	Bal.

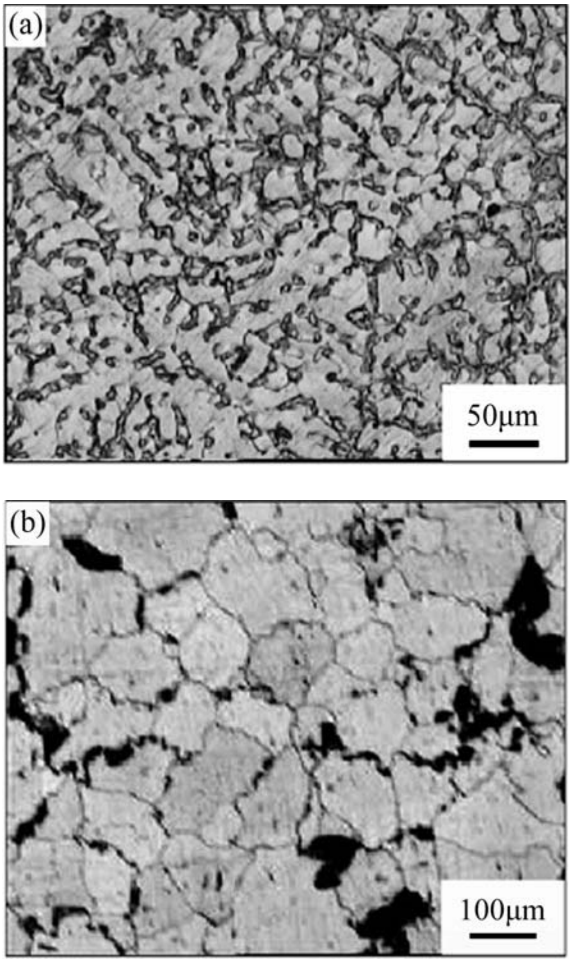


Fig. 1. Microstructure of AZ80 Mg alloy before and after homogenization heat treatment: (a) as-cast microstructure with segregation before heat treatment, (b) homogenized microstructure after heat treatment.

ively, with a pressing speed of 1 mm s^{-1} in ECAP. To investigate the influence of ECAP temperature, ECAP were carried out at 553 K and 633 K, after which aging treatment was conducted at 423 K for 4 h, 8 h, 12 h, 16 h and 20 h, respectively. To evaluate the strengthening effect, uniaxial tensile test was carried out at room temperature (RT) at an initial strain rate of $1.25 \times 10^{-3}\ s^{-1}$ to examine the mechanical properties of AZ80 alloy after ECAP and aging treatment. The tension direction is set parallel to the outlet direction of ECAP. Optical microscope (OM) and transmission electronic microscope (TEM) were used to observe the metallographic microstructure and distribution of second phase particles.

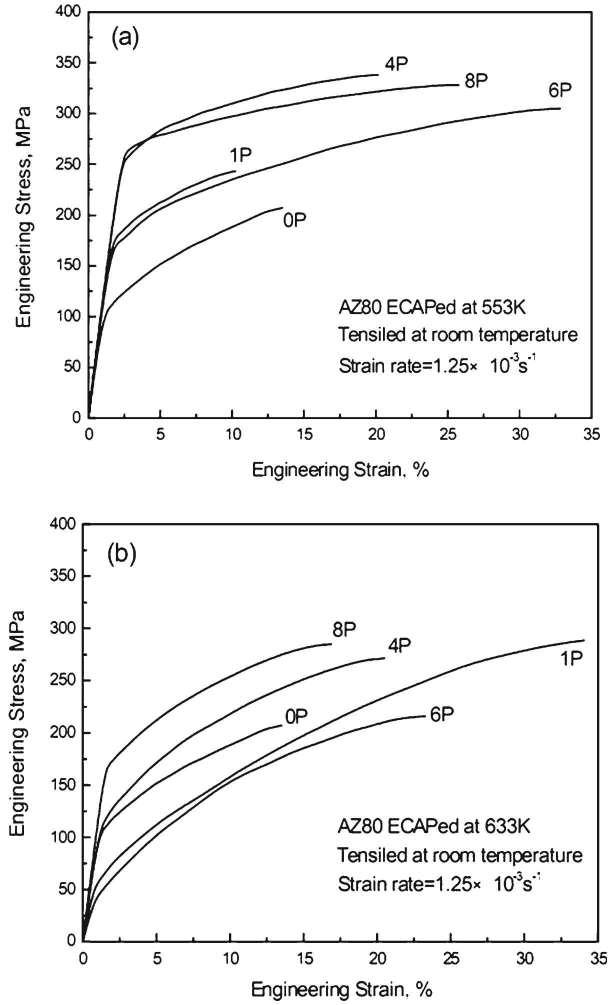
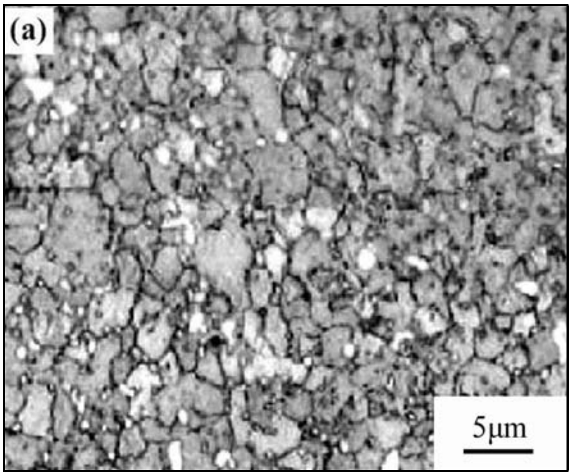


Fig. 2. Tensile curves of AZ80 alloy ECAPed at different temperatures: (a) 553 K, (b) 633 K.

3. Results and discussion

3.1. Mechanical properties after ECAP

Engineering stress-strain curves of AZ80 alloy ECAP processed to various passes at different temperatures are shown in Fig. 2. It is seen that, for ECAP samples processed at 553 K, tensile strength first increases obviously from ECAP 0P to 4P (Fig. 2a), then decreases obviously at 6P and increases again to a level similar to that of 4P. After 8 passes of ECAP, the ultimate tensile strength (UTS) is 329 MPa, 58.7 % higher than that of 207 MPa in the homogenized state before ECAP (0P). The obvious decrease of flow stress at 6P might have resulted from complicated effects of ECAP strain and texture, similar phenomenon was also ob-



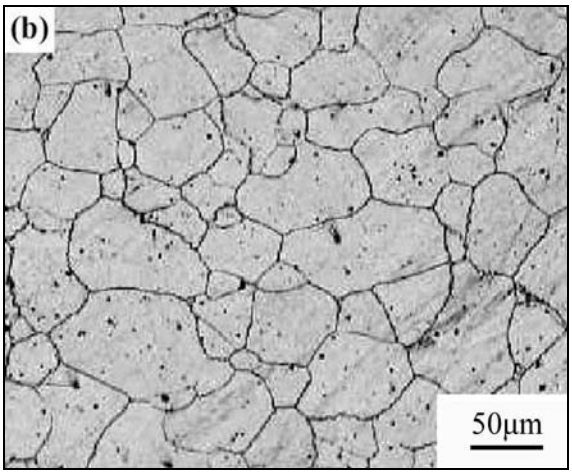


Fig. 3. Microstructure after ECAP 8 passes at different temperatures: (a) 553 K, (b) 633 K.

served in AZ31 and AZ61 [16, 17]. For ECAP samples processed at a higher temperature of 633 K (Fig. 2b), the flow stress after 8 passes of ECAP increases to 285 MPa, which is 37.6 % higher than 207 MPa of 0P. At the same time, the ductility is apparently enhanced after ECAP, especially for ECAP 1P. It is also noted that, the flow stress of the sample of ECAP 0P is higher than 1P and 6P, which again probably results from the complicated effects of ECAP strain and texture. It is clear from Fig. 2a that for solely fine-grain strengthening, ECAP at a lower temperature of 553 K has a better result.

3.2. Microstructure after ECAP

To investigate the fine-grain strengthening from a microscopic view, the microstructure after ECAP 8 passes at different temperatures was observed by optical microscope, as shown in Fig. 3. It is noticed that apparent refinement occurs both at 553 K and 633 K, but the grain-size reduction depends remarkably on ECAP temperature. The lower is the temperature of ECAP, the better effect of grain refinement. The as-homogenized microstructure with an average grain size of 155 μm (Fig. 3a) was refined to 3.5 μm after 8 passes of ECAP at 553 K (Fig. 3a), and to 65 μm after 8 passes of ECAP at 633 K (Fig. 3a).

At the same time, the generated microstructure after ECAP at 633 K exhibits equaxial grain-shape and obvious absence of second-phase, implying remarkable occurrence of dynamic recrystallization (DRX) and generation of supersaturated solid solution.

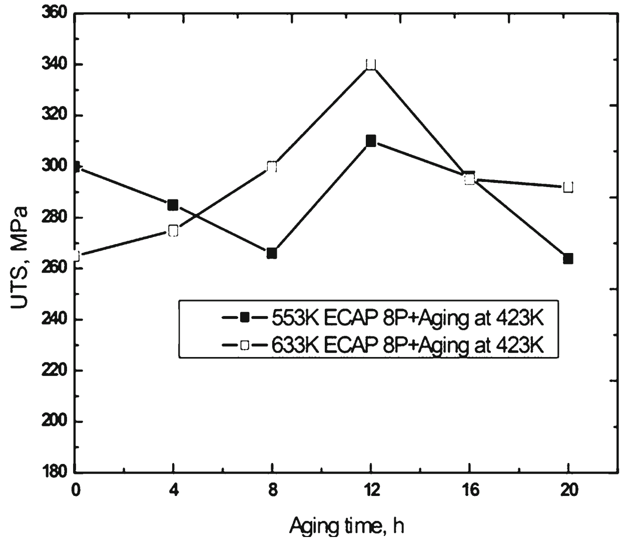


Fig. 4. Relation between aging time and UTS after ECAP 8 passes at different ECAP temperatures.

3.3. Mechanical properties after ECAP + aging treatment

Figure 4 illustrates the relation between aging time and UTS of AZ80 alloy after ECAP 8 passes at different temperatures. It is noticed that the UTS after ECAP 8 passes at 553 K followed by aging at 423 K for 12 h is 310 MPa, which is 10 MPa lower than 320 MPa before aging treatment. While the UTS of samples with ECAP 8 passes at 633 K followed by aging at 423 K for 12 h is 340 MPa, which is 75 MPa higher than that before aging treatment. Therefore, the condition of ECAP at 633 K + aging at 423 K is better than ECAP at 553 K + aging at 423 K to strengthen AZ80 Mg alloy for the combination method.

It is interesting to note that the change of UTS after ECAP 8 passes at 553 K with aging time below 8 h differs from that after ECAP 8 passes at 633 K, with the former decreasing with aging time while the latter increasing reversely. When continuously increasing aging time, their UTS show the similar feature, with the maximum values obtained at

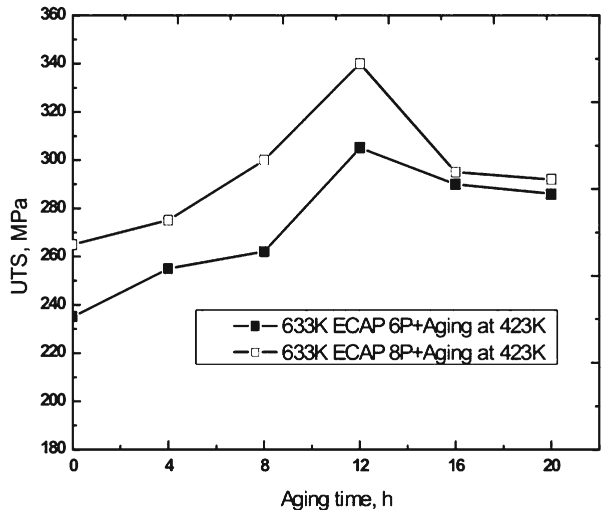


Fig. 5. Relation between aging time and UTS at different ECAP passes at 633 K.

12 h followed by remarkable decrease due to over-aging. The reason for this different trend can be explained as follows. During ECAP at 553 K, precipitation may also well progress under the effect of both temperature and deformation, which exhaust aging hardening potential by producing coarse grains at a much higher temperature use for conventional aging hardening of the alloy. Thus during subsequent aging treatment, over-aging appears which leads to softening. Consequently, the UTS decreases with aging time. But for ECAP at 633 K, since the solid solution degree is relatively high, the subsequent aging treatment resulted in remarkable increase of UTS with aging time. With the extension of aging time, the proportion of precipitates in AZ80 after ECAP 8 passes at 633 K increases remarkably, showing obvious increase of UTS.

Figure 5 shows the relation between aging time and UTS of AZ80 alloy after different ECAP passes at 633 K. It is found that, for both cases of 6 passes and 8 passes, the UTS reach the maximum values after aging at 423 K for 12 h, implying a peak of strengthening by aging treatment. Moreover, the maximum UTS of ECAP 8 passes + aging at 423 K is apparently higher than that of ECAP 6 passes + aging at 423 K, which might be ascribed to more remarkable refinement and precipitation with increasing ECAP passes.

3.4. Microstructure after ECAP + aging treatment

Figure 6 presents the microstructures after ECAP different passes at 633 K + aging at 423 K for 12 h. Note that, many lamellar second-phase precipitates appear along grain boundaries and grow into these grains. This is a kind of discontinuous precipitation that initiates favorably at grain boundaries with higher energy and more blocked dislocations.

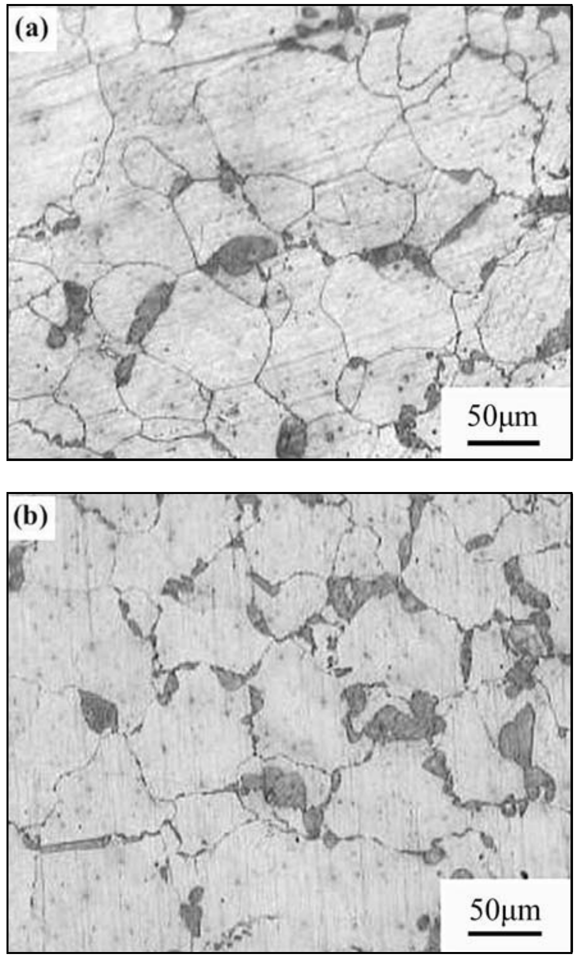


Fig. 6. Microstructure after ECAP different passes at 633 K + aging at 423 K for 12 h: (a) 6 passes, (b) 8 passes.

In addition, with increasing ECAP passes, the fraction of discontinuous precipitates along grain boundaries increases accordingly. Since the shearing deformation becomes more severe with increasing ECAP passes, the stored distortion energy increases correspondingly. Consequently, the discontinuous precipitating is energetically promoted along grain boundaries. Considering the strong hindering effect of discontinuous precipitates on dislocation movement of matrix phase, the increased discontinuous precipitates should be partially responsible for the higher UTS of sample with ECAP 8 passes + aging than that of ECAP 6 passes + aging in Fig. 5. Further, TEM observation showed two kinds of continuous precipitates after ECAP 8P at 633 K + aging at 423 K for 12 h, as indicated in Fig. 7.

The ribbon-like precipitates and short pole-like ones are shown in Fig. 7a and 7b, respectively. As reported in references [7] and [8], most continuous β precipitates $Mg_{17}Al_{12}$ in Mg alloys are parallel to the

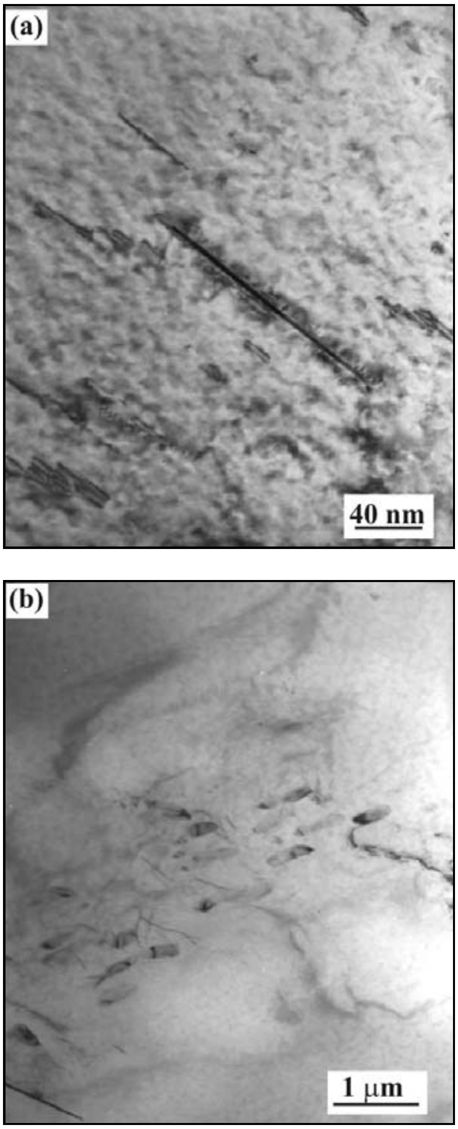


Fig. 7. TEM images of continuous precipitates in grains after ECAP 8 passes at 633 K + aging at 423 K for 12 h: (a) ribbon-like continuous precipitate, (b) short pole-like continuous precipitate.

basal plane of αMg matrix. Therefore, the continuous β precipitates have little hindering effect on dislocation movement and contribute little to the strengthening of AZ80 Mg alloy. Under general aging conditions, discontinuous precipitation is readily to occur first and nucleate at grain boundaries, then after a period of time, continuous precipitation can take place in grain interior. In addition, aging temperature can exert significant influence on precipitation mode [18]. The higher is the aging temperature, the earlier the continuous precipitation occurs and the larger the fraction of continuous precipitates is. Consequently, in the case of ECAP 8 passes at 633 K + aging at 423 K for 12 h, the discontinuous precipitation occurs along grain boundaries followed by continuous precipitation in grains interior, as noticed in Fig. 7.

From above results, it is indicated that the proposed combination method can strengthen AZ80 Mg alloy to a certain extent, but the strengthening effect is not remarkable enough compared with simple fine-grained strengthening by ECAP. This is because that the fine-grain strengthening is realized through DRX in ECAP, which requires low ECAP temperature to suppress grain growth. In contrast, the followed aging treatment requires high ECAP temperature to form supersaturated solid solution and high concentration of vacancy for precipitation strengthening effect. This contradiction is responsible for the unsatisfactory strengthening effect of the combination of ECAP and aging treatment. It is suggested that the two strengthening approaches may be combined effectively if suitable alloying element is added to Mg alloy to considerably suppress grain growth during ECAP at solid solution temperature.

4. Conclusions

By analyzing the microstructure and strength of AZ80 magnesium alloy after different ECAP passes and aging time, we got the following conclusions:

(1) At the ECAP temperature of 553 K, the combined strengthening effect of AZ80 alloy after ECAP 8 passes + aging at 423 K is not remarkably enough. This might be resulted by the low ECAP temperature which is unfavorable for sufficient solid solution for precipitation process of AZ80 Mg alloy.

(2) The combined strengthening effect of ECAP 8 passes at 633 K + aging at 423 K is better than ECAP 8 passes at 553 K + aging at 423 K. When the aging time is 12 h, the UTS reaches 340 MPa, which is 75 MPa higher than that after ECAP 8 passes at 633 K. This shows the combination method can strengthen AZ80 alloy remarkably to a certain extent.

(3) The UTS after ECAP at 633 K + aging at 423 K reaches a maximum value after aging for 12 h, which is not a monotonic function of aging time. Many lamellar discontinuous precipitates appear at grain boundaries after aging treatment, setting strong hindrance against dislocation movement of matrix, resulting good strengthening effect.

(4) The combined strengthening effect of ECAP 8 passes at 633 K + aging at 423 K is not remarkable enough compared with solely fine-grain strengthening by ECAP at 553 K. This is because of the contradiction between the low ECAP temperature necessary for grain refinement and the high temperature necessary for sufficient solid solution before aging.

Acknowledgements

This work was financially supported by the National Key Fundamental Research Project of China, the National Nature Science Foundation of China under Grant No. 50474028, the Postdoctoral Science Foundation of China under Grant No. 20080431102 and the Postdoctoral Foundation of Jiangsu Province under Grant No. 0801022B.

References

[1] AVEDESIAN, M. M.—BAKER, H.: Magnesium and Magnesium Alloys. In: ASM Specialty Handbook. OH, Mater. Park, ASM International 1999.
[2] MORDIKE, B. L.—EBERT, T.: Mater. Sci. Eng. A, *302*, 2001, p. 37. doi:10.1016/S0921-5093(00)01351-4
[3] FRIEDRICH, H.—SHUMANN, S.: J. Mater. Proc. Tech., *117*, 2001, p. 276. doi:10.1016/S0924-0136(01)00780-4
[4] HALL, E. O.: Proc. Phys. Soc. B, *64*, 1951, p. 747. doi:10.1088/0370-1301/64/9/303
[5] PETCH, N. J.: J. Iron Steel Inst., *174*, 1953, p. 25.
[6] CLARK, J. B.: Acta Metall., *16*, 1968, p. 141. doi:10.1016/0001-6160(68)90109-0
[7] CRAWLEY, A. F.—LAGOWSKI, B.: Metall. Trans., *5*, 1974, p. 557.
[8] NIE, J. F.—MUDDLE, B. C.: Acta Mater., *56*, 2008, p. 3490. doi:10.1016/j.actamat.2008.03.028
[9] WANG, J. T.—LIANG, M. X.: Mater. Sci. Forum, *25*, 2005, p. 370.
[10] FIGUEIREDO, R. B.—SZÁRAZ, Z.—TROJANOVÁ, Z.—LUKÁČ, P.—LANGDON, T. G.: Scripta Mater., *63*, 2010, p. 504. doi:10.1016/j.scriptamat.2010.05.016
[11] DING, R. G.—CHUNG, C. W.—CHIU, Y. L.: Mater. Sci. Eng. A, *527*, 2010, p. 3777. doi:10.1016/j.msea.2010.02.030
[12] YOSHIDA, Y.—ARAI, K.—ITOH, S.: Sci. Tech. Adv. Mater., *6*, 2005, p. 185. doi:10.1016/j.stam.2004.11.011
[13] WATANABE, H.—MORIWAKI, K.: Mater. Trans., *44*, 2003, p. 775. doi:10.2320/matertrans.44.775
[14] NIE, J. F.: Scripta Mater., *48*, 2003, p. 1009. doi:10.1016/S1359-6462(02)00497-9
[15] ZHOU, J. P.—ZHAO, D. S.—WANG, R. H.: Mater. Letters, *61*, 2007, p. 4707. doi:10.1016/j.matlet.2007.03.013
[16] KIM, W. J.—AN, C. W.—KIM, Y. S.—HONG, S. I.: Scripta Mater., *47*, 2002, p. 39. doi:10.1016/S1359-6462(02)00094-5
[17] MUKAI, T.—YAMANOI, M.—WATANABE, H.—HIGASHI, K.: Scripta Mater., *45*, 2001, p. 89. doi:10.1016/S1359-6462(01)00996-4
[18] CELOTTO, S.: Acta Mater., *48*, 2000, p. 1775. doi:10.1016/S1359-6454(00)00004-5