Strengthening study on 6082 Al alloy after combination of aging treatment and ECAP process

S. Dadbakhsh, A. Karimi Taheri, C.W. Smith

Abstract

Equal channel angular pressing (ECAP) was used before and after various aging treatments in order to strengthen a commercial 6082 Al alloy. Experiments were carried out to study the strengthening of the alloy due to pre and post-ECAP aging treatment. It was found that aging before and after ECAP processing is an effective method for strengthening of the alloy. An increase in both strength and ductility of the ECAPed specimen was achieved via appropriate post-aging treatment. This was in such a manner that for maximal strengthening, post-ECAP aging is best conducted at temperatures lower than those usually used for aging if prior work hardening is not undertaken. Pre-ECAP aging was also discussed in the light of dislocation density and work hardening.

Keywords:
Aluminum alloys
Equal channel angular processing
Aging
Dislocations
Mechanical characterization

1. Introduction

During the last two decades, severe plastic deformation (SPD) has been demonstrated as an effective approach to produce ultrafine grain (UFG) materials. Extensive research has been carried out to develop SPD techniques and to establish processing parameters to fabricate UFG metals and alloys with more desirable properties. Among different SPD techniques, equal channel angular pressing (ECAP), alternatively called equal channel angular extrusion (ECAE), has attracted the most attention since it is very effective in producing UFG materials with an adequate size for various structural applications [2–4]. Equal channel angular pressing (ECAP) is a process at which a billet of material is pressed through a die having two channels of equal cross-section intersecting at an angle [5]. Thus, the billet experiences simple shear without a change in cross-sectional area upon passage through such a die, and it is possible to repeat the process. It is particularly interesting because of development of near uniform, intensive and oriented simple shear in bulk billets [6,7].

Al–Mg–Si alloys including 6082 Al alloy are an important group of materials, widely used for demanding structural applications. Their age-hardening response can be very significant, leading to remarkable improvement of strength after an appropriate heat treatment [8,9]. Although in order to achieve optimum age hardening, the precipitation sequence of Al–Mg–Si alloys has been studied for many years, only in recent years has a satisfactory agreement on phase evolution during aging been attained [10–14]. For example, Edwards et al. [13] has developed the following sequences of precipitations in these alloys: Al (solid solution) → (clusters of Si atoms + clusters of Mg atoms) → (dissolution of Mg clusters) → (formation of Mg/Si co-clusters) → (small precipitates of unknown structure) → (B′ precipitates) → (B′ and B″ precipitates). It has been noted that B′ forms with B′′ and they have a similar structure but different sizes [13,15]. Also, it has been shown that the GP zone, B″, B′, B, and B″ precipitates have typical morphologies/sizes of near-spherical/1–2 nm, needles/up to 40 Å × 40 Å × 350 Å, ribbons/several μm long, plates or cubes/up to 10–20 μm, and ribbons/up to 1 μm long, respectively [15]. Among these, the B″ precipitates are considered to give the main strengthening contribution and hence they are mostly responsible for the maximum age-hardening effect [15,16]. The existence of a plateau in the curve of hardness versus aging time has been reported and attributed to two-step age-hardening response and the precipitation process of Al–Mg–Si(-Cu) alloys [16]. Recently, the present authors [17] have found that increasing the aging temperature as well as the predeformation converts the plateau to a minimum between the two peaks in the curve of hardness versus the aging time. Other researchers [18,19] have shown that a small amount of impurities such as Fe and Mn can cause the formation of a new phase in the microstructure in the form...
of quaternary Al–Fe–Mn–Si and thereby change the mechanical properties of the alloy.

In recent years, the need to improve the mechanical properties of light alloys has driven the attention of industry and research efforts towards two different, but parallel, fields. The first field is the process of age hardening of some alloys [10–14], and the second field is the development of new metal processing techniques such as ECAP to refine the microstructure and to improve the mechanical strength [1–3]. Some researchers [20–22] have combined these two fields to achieve even better properties. For example, Kim et al. [20] aged a 2024 Al alloy at 100 °C and 175 °C after 1 pass ECAP and reported an enhancement in both the strength and ductility.

While one of the major requirements for advanced materials is attaining very high strength with a reasonable level of ductility, there is no previous systematic study to assess the effects of pre- and post-ECAP aging treatment at various aging temperature on mechanical properties of Al–Mg–Si alloys. Moreover, the variation of microstructure and secondary precipitates due to aging treatment and ECAP process merits further investigation. Furthermore, assessment of dislocation density in pre-ECAP aging treatment has not been reported yet. The present study investigated the effective factors for enhancing the mechanical properties of an industrial AA6082 using combination of aging heat treatment and ECAP deformation. TEM inspections were carried out to identify the variation of microstructure due to aging after ECAP. Dislocation density of the aged and then deformed specimen was evaluated using an approach based on the Taylor equation [23] and a linear relationship [24] between the dislocation density and plastic strain.

2. Material and experimental details

AA6082 in the form of T6-extruded bar of 15 mm diameter was used in this study. The composition of the alloy in wt.% was 1.2 Mg, 0.82 Si, 0.5 Mn, 0.1 Cu, 0.2 Fe, and Al balance. The test specimens, cut from the bar, were prepared for use in 3 different aging treatments. These were: (i) quench aging treatment, i.e. solution treatment at 530 °C (2 h) → quenching to room temperature (RT) water → aging at 180 °C, (ii) pre-ECAP aging treatment, i.e. solution treatment at 530 °C (2 h) → quenching to RT water → aging at 180 °C → ECAP (1 pass), and (iii) post-ECAP aging treatment, i.e. solution treatment at 530 °C (2 h) → quenching to RT water → ECAP (1 pass) → aging at 100, 130, 160, 180, and 200 °C. These three different aging treatments form a systematic comparative group. Likewise the aging at five temperatures from 100 to 200 °C also form another systematic comparative group.

Polarized-light images of the alloy before and after ECAP, shown in Fig. 1, were acquired using a NEOPHOT 30 optical microscope. Fig. 1a is the microstructure in the solution treated condition (before ECAP) with equiaxed grains and an average grain size of 65 ± 15 μm, while Fig. 1b indicates the relatively elongated grain structure after 1 pass ECAP. ECAP was conducted at room temperature using a cylindrical die of 14 mm diameter, an internal angle of 120° between the channels, and a curvature angle of zero degree. This configuration induces an equivalent strain of approximately 0.67 after each pass through the die [5,25]. Molybdenum disulfide (MoS2) was used as a lubricant in the ECAP tests.

The specimens were tested using the Vickers hardness (20 kg load) and a mean of at least four hardness readings recorded. An Instron-Wolpert hardness testing machine was used. Tensile tests were carried out according to the E8 ASTM standard [26] on those aged specimens with the highest values of hardness. A crosshead speed of 5 mm/min was used in the tests.

The microstructures of the selected post-ECAP aged samples were examined using a PHILIPS CM200 transmission electron microscope (TEM) operated at 200 kV. Samples for TEM were cut along the longitudinal axis and ground to a thickness of 100 μm. Thin discs were punched and then electropolished using a solution of 33% HNO3 in methanol at −30 °C and 18 V. The quality of TEM pictures was enhanced using image processing software. Moreover, modeling the variation of dislocation density due to pre-ECAP aging of the alloy was carried out using the Taylor relationship [23] and the apparent linear relationship between dislocation density and plastic strain [24].

3. Results

3.1. Post-ECAP aging treatment

The hardness and aging time is shown in Fig. 2 for solution treated, ECAPed, and aged specimens at 100, 130, 160, 180, and 200 °C. As seen, the hardness of the alloy in solution treated condition increases by 45% after 1 pass ECAP, in agreement with findings of previous researchers [19,22]. Aging affects the hardness of the alloy in such a manner that, a rise, a fall, and a duplicate rise occur in the curve of hardness versus time, especially at lower temperatures of 100, 130, and 160 °C. This trend is less apparent in the curves relating to 180 and 200 °C.

The stress–strain curves of the solution treated and ECAPed samples with and without post-aging treatment at 100 °C are shown in Fig. 3a. The curve of the solution treated sample is shown for comparison. As observed, a significant strengthening appears after a
single pass, being in agreement with the hardness results plotted in Fig. 2. The yield strength (YS) and tensile strength (UTS) of the ECAPed sample are 350 MPa and 353 MPa, respectively, being 192% and 53% higher than the YS and UTS of the solution treated sample. The homogenous and fracture strain, on the other hand, decreases respectively from 36% and 42% for solution treated sample to 9.5% and 17% for as-ECAPed sample. It is interesting to note that when the ECAP process is combined with 6 and 30 h aging treatment at 100 °C (first and second peak in Fig. 2), a further increase in YS and UTS of the ECAPed sample is achieved confirming the hardness results presented in Fig. 2. These processes are also effective in improving the homogenous and fracture strain of the ECAPed material. The observed fracture shape of ECAPed tensile specimen and also of the specimen ECAPed and aged at 100 °C/30 h are shown in Fig. 3b and c, respectively, where the fracture surface of the ECAPed specimen exhibits a shear type rupture. Obviously, this kind of rupture is not similar to a brittle fracture (characterized by a separation normal to the tensile stress) or a ductile fracture (characterized by two separations oriented about 45° to the tensile stress). As seen from Fig. 3c, the shear type of fracture converts to be more cup-and-cone type due to post-aging treatment on ECAPed specimen, associated with increase in tensile ductility.

Fig. 4 indicates that aging following ECAP at 130 °C increases both the tensile strength and ductility. However, Fig. 5 shows that the post-ECAP aging at 160 °C decreases the yield strength. Also, as seen from this figure, the fracture strain of the specimen aged for 6 h after ECAP at 160 °C is less than that of the specimen aged...
A plot of hardness against aging time for solution treated material is shown in Fig. 6. It is apparent that during the early stages of quench aging, the alloy hardness increases from 85 HV to 120 HV. Then, it is followed by a minimum after about 12 h, and finally by a second increase to a hardness peak of 125 HV after about 24 h. It should be noted that this variation is correlated with the formation of clusters/GP zone, dissolution of clusters, and formation of $β''$, respectively [17]. Pre-ECAP aging treatment was assessed from solution treatment followed by aging at 180 °C and finally 1 pass of ECAP. The times of pre-aging were chosen as 3, 6, 12, 24, and 28 h corresponding to almost the half of aging time of the first peak, the aging time of the first peak, the minimum after the first peak, the second peak, and the time at which the alloy is overaged, respectively. As seen from Fig. 6, the hardness of ECAPed samples increased from 124 HV to 137 HV due to the 3 h pre-aging treatment. While development of hardness occurs most rapidly in 3 h, the 6h aging before ECAP exhibits a further increase in hardness. With an increase of pre-aging to 12 h the hardness decreases, while 24 h aging before ECAP causes a secondary increase in the hardness values. Finally, 28 h aging before ECAP decreases the hardness in comparison with the previous specimen.

The dependence of tensile strength and fracture strain with the time of pre-ECAP aging treatment at 180 °C is summarized in Fig. 7. Significant strengthening is evident in a single pass after 3 h aging, agreeing with the hardness result in Fig. 6. The UTS of the specimens aged 3 h before the ECAP process reached 397 MPa being 73% higher than that of the solution treated material (230 MPa), while fracture strain of the specimens aged 3 h before the ECAP process reduced. The increase of aging time before ECAP from 3 to 6 h has a negligible effect on strength and ductility. The 12 h pre-aging develops fracture strain but reduces the tensile strength, confirming the results presented in Fig. 6. Although the precipitates produced by 24 h pre-aging treatment before ECAP increase the tensile strength, but reduce the fracture strain. Finally, it can be seen that the 28 h aging before ECAP reduce the tensile strength and increase the fracture strain.

In conclusion, regarding the above results, the pre-ECAP aging treatment is slightly more effective in improving the strength of 6082 Al alloy compared with the post-ECAP aging treatment, but a higher ductility can be achieved by the latter treatment.

### 3.3. Transmission electron microscopy (TEM) investigation

TEM investigations were carried out in order to better understand the influence of the aging on microstructure affected by the ECAP and the hardening phases. Fig. 8a–c shows a series of TEM images of a post-ECAP aged (at 100 °C/6 h) material. Fig. 8a shows representative BF-TEM images of the samples, where two different shapes of particles, i.e. rounded and plate-like are present. These rounded and plates shapes belong to one type of rod-shaped precipitates, being mostly located on dislocation cell boundaries. The relatively large size of these precipitates indicates that they have not formed during this short aging time in this treatment, i.e. they remained in the matrix during solution treatment. Fig. 8b also shows some fragmented precipitates in the matrix. However, some very fine precipitates are apparent in the matrix as well. Fig. 8c demonstrates the microstructure of the sample including dislocation cells and precipitates. As is clear from this figure, dislocations accumulate often in thin boundary walls. On the other hand, some precipitates are found on these thin dislocation cell walls. Furthermore, it is apparent that dislocation cells have been textured and oriented after 1 pass of ECAP.

Fig. 9a–c shows a series of TEM images of a 12 h post-ECAP aged at 100 °C specimen. Fig. 9a shows some smaller precipitates compared to Fig. 8a and b, being distributed in the matrix. As seen in Fig. 9b, the size and wall thickness of dislocation cells are larger than Fig. 8c and orientation of elongated dislocation cells has changed. Fig. 9c shows some precipitates surrounded by dislocations, suggesting aging allows the precipitates to be released out from those areas.

### 3.4. Dislocation density in pre-ECAP aging treatment

After converting the experimental engineering stress–strain curves to true stress–strain curves (flow curves) up to the necking point [27], the variation of dislocation density with plastic strain for pre-ECAP aged specimens was calculated using the Taylor relation as:

$$\sigma = \sigma_0 + \alpha M G b \rho^{1/2}$$

where $\sigma$ is the flow stress, $\sigma_0$ the friction stress (assuming 25 MPa for AA6082), $\alpha$ a numerical constant ($\alpha = 0.33$), $G$ the shear modulus...
(G = 26 GPa for Al and its alloys), b the Burgers vector of dislocations (b = 0.286 nm), and M the Taylor factor (M = 3 for untextured polycrystalline materials) [23]. It should be noted that Gubicza et al. [23] has reported a good correlation of experimental results of dislocation density with the Taylor equation in both pure and solution treated metals and alloys which were subsequently ECAPed. The Taylor relation has also been used to study the strengthening and particle size effect in metal–matrix composites [28, 29] and to evaluate the dislocation density after cold rolling in Al–Mg–Cu–Mn alloys [30]. Here, the Taylor relation is used to understand the effect of particles on work hardening and dislocation density after ECAP. So, using the above values for AA6082, the variation of dislocation density with plastic deformation was calculated from the corresponding flow curves and is shown in Fig. 10. As seen, the dislocation density versus the plastic strain plot appears as a line (in all cases the $R^2 \geq 0.9$). This is in agreement with the following equation, being the relationship between the multiplication, creation, and unpinning of dislocations:

$$\rho = \rho_0 + C \varepsilon_p \alpha$$  (2)

where $\varepsilon_p$ is the plastic strain, $\rho_0$, $C$, and $\alpha$ the constants, and $\rho$ is the total dislocation density [24]. The agreement between the plots shown in Fig. 10 and the linear shape of Eq. (2) not only confirms the procedure developed from Taylor relation but
Fig. 9. TEM micrographs of the 1 pass ECAPed and aged specimen at 100 °C/12 h (first minimum in Fig. 1): (a) showing the typical morphology and size of particles distributed in the matrix and (b and c) microstructure and cell direction.

Also makes it possible to calculate the constants introduced in Eq. (2). Obviously, the product of $C \times \alpha$ is a constant, being the slope of dislocation density versus plastic strain. This parameter can be described as the multiplication rate of dislocation density with plastic strain or the increase in dislocation density per unit strain.

Fig. 10 shows the variation of dislocation density with increasing tensile plastic strain of pre-ECAP aged specimens at 180 °C for 0, 3, 6, 12, 24, and 28 h. As seen, $\rho_0$ is $23.4 \times 10^{14} \text{m}^{-2}$ and the multiplication rate of dislocation density with increasing plastic strain is $52 \times 10^{14} \text{m}^{-2}$ for the solution treated and then ECAPed material. Aging the specimen for 3 h before ECAP increases both the primary and multiplication rate of dislocation density to $27.2 \times 10^{14} \text{m}^{-2}$ and $68 \times 10^{14} \text{m}^{-2}$, respectively. The increase in primary and multiplication rate of dislocation density continues until 6 h aging before ECAP (Fig. 10c), while 12 h aging before ECAP has an inverse effect and decreases both the primary and multiplication rate of dislocation density as shown in Fig. 10d. Further development of aging treatment until 24 h before ECAP, increases again both the primary and multiplication rate of dislocation density. After 28 h aging (overaging condition) before ECAP, both the primary dislocation density and multiplication rate of dislocation density decreased.
4. Discussion

The present work has aimed to assess different methods of strengthening of 6082 Al alloy. The increase of hardness, yield strength (YS), and tensile strength (UTS) of the solution treated specimen after 1 pass ECAP are shown in Figs. 2 and 3. This increase of strength is accompanied with an orientation of solution treated microstructure due to ECAP process (see Fig. 1). These findings are in agreement with the reports of Kim et al. [22] and Cabibbo et al. [19]. Kim has attributed these increments in strength to the considerable substructure refinement occurring during intensive plastic deformation. In addition to substructure refinement, Cabibbo has claimed the hardness increase is mainly due to the work hardening of the alloy contributing to the grain refinement induced by ECAP. Cabibbo also presumed that the interaction of secondary phases (as hardening particles) with high-density dislocation system due to severe plastic deformation enhances the work hardening. Some of the relatively large secondary phase particles can be seen in Fig. 8 with the rounded and plate shapes representing rod-shaped precipitates. Considering the relatively large size of these precipitates, it is very unlikely for them to be formed in this aging treatment because the time scales are too short. In fact, their morphology and location on dislocation cell boundaries would suggest that these particles are often composed of Al–Fe–Mn–(Si) intermetallics [19] which remain unsolved after solution treatment [18,31] and show themselves as rod-shaped secondary particles. Thus, the particles are displaced by the ECAP process with the motion and accumulation of dislocations in cell boundaries. On the other hand, these particles assist work hardening by either pinning the dislocations or being cut to smaller particles [19]. Fig. 8b shows some fragmented precipitates confirming the cutting effect of ECAP process on particles, which would thus presumably increase the work hardening. It should be noted that although the work hardening increases the accumulated strains and subsequently strength of the material, this aspect leads to a reduction in the homogenous and fracture strain of ECAPed sample in comparison with solution treated specimen [32] (Fig. 3a). Moreover, there are some very fine precipitates in the matrix, assumed as Mg–Si precipitates due to their very small size. The resolution of TEM pictures (Figs. 8 and 9) was inadequate to characterize these precipitates accurately.

The fall and rise in the aging curves of Fig. 2 can be attributed to the formation of new phases with increasing the aging time. In Al–Mg–Si alloys, the first rise can be related to clustering of atoms, the fall between the peaks is associated with dissolution of clusters, and the second peak is correlated with formation of β′ phase [17]. Further development in aging reduces the strength because of formation of more stable phases such as β′ and β [16,17]. Furthermore, softening due to recovery reduces the effective strengthening with increasing the aging temperature. This occurs in such a manner that there is only one peak evident in the hardness curve of 180°C in post-ECAP aging treatment. The peak probably represents the influence of more stable precipitates forming at the second peak of lower temperatures, e.g. 100, 130, and 160°C (see Fig. 2). Cerri and Leo [33] has reported that in the severely deformed 6082 Al alloy modified by Zr, if the aging is performed at a relatively high temperature such as 170°C, the effect of recovery overcomes the strengthening associated with precipitation which is in agreement with the severe reduction of hardness of the post-ECAP aged specimen at 200°C (Fig. 2).

Referring to Fig. 3a, a combination of ECAP process with subsequent 6 and 30 h aging treatments at 100°C (first and second peak in Fig. 2), leads to an increase in YS and UTS as well as homogenous and fracture strain of the ECAPed sample. The increase of YS and UTS due to post-aging of ECAPed sample originates from precipitation strengthening, while the simultaneous increase of ductility with strength in aging is a rare phenomenon. This together increase of ductility and strength can be attributed to concurrent incidence of precipitation with internal stress relaxation. In other words, when hardening by aging dominates over the softening by relaxation of internal stress, an enhancement in both the strength and ductility of the ECAPed material is possible. In summary, a proper post-ECAP aging treatment imparts a high strength with a moderate level of ductility due to the annealing effect of aging after ECAP.

It is interesting to note that the changes in ductility are accompanied with changes in microstructure. In fact, work hardening and localization of strain and stress at the particles in the boundaries decrease ductility of the ECAPed specimen (Fig. 3a). This can be thought in terms of high accumulation of dislocations in the cell boundaries surrounding some precipitates, as seen in Fig. 8c. In contrast, increase in wall thickness of dislocation cells due to aging enhances the ductility (Fig. 9b). Furthermore, as observed from Fig. 9a, the new location of secondary phase and fragmented particles at the interior of grains leads to less strain and stress localization contributing to an enhancement in ductility. This new location of the precipitates due to aging can be explained through two possible hypotheses: (i) First hypothesis: during heat treatment some dislocation walls can disappear, thus leaving many precipitates in the cell interiors. (ii) Second hypothesis: the motion of dislocations, being due to the high gradient of dislocation density from walls to interior of grains, displaces the precipitates to...
interior of grains. These hypotheses would fit with Fig. 9c, showing some precipitates involved in dislocations. It seems that the aging development allows them to be released from those areas.

As seen in Fig. 3b, the fracture shape of ECAPed specimen exhibits a shear type rupture, oriented approximately 45° to the cross-section. The elongated dislocation cells (Fig. 8c) and grain structure (Fig. 1b) can be considered as a possible reason for this phenomenon which affects the mechanical properties in an anisotropic manner. In general, the tensile ductility will be lower in the transverse direction (normal to the dislocation cell) than parallel to dislocation cell [27]. In other words, in the ECAPed sample, because of elongated dislocation cells, shear rupture will more likely grow, as in Fig. 3b. On the other hand, increasing the aging time after ECAP changes the fracture to cup-and-cone type (Fig. 3c) from shear type (Fig. 3b). This can be associated with the changing in the orientation of elongated dislocation cells (compare Fig. 9b to Fig. 8c).

Fig. 5 shows that the post-ECAP aging at 160°C decreases the yield strength because of the domination of recovery over strengthening mechanisms. Also, the fracture strain of the 6 h aged specimen after ECAP is less than that of the 2 h aged specimen after ECAP, in contrast with the pattern seen in Figs. 3 and 4 for post-ECAP aged specimens at 100°C and 130°C. To explain this phenomenon, one can consider the following postulation – ordinary effect of aging, i.e. the removal of ductility due to pinning of dislocations, may in the case of post-ECAP aging treatment be overcome by the annealing effect of aging. It is therefore possible that in the first 2 h of post-ECAP aging treatment at 160°C, the fracture strain increases because of the removal of much of the internal stress, prompting subsequent reduction of ductility via dislocation pinning.

As seen in Fig. 6, pre-ECAP aging time of solution treated material leads to an oscillation of hardness. The oscillation of hardness can be explained by effect of particles on work hardening upon dislocation density. This is in such a manner that the first increase can be attributed to the fine primary precipitates (cluster of atoms and GP precipitates) produced by aging before ECAP [17]. It has been shown that coherent GP zones can produce a maximum opposing force to the motion of dislocations when dislocations cut through them [27,34]. This restricts the motion of dislocations, increases the primary dislocation density and the dislocation multiplication rate (see Fig. 10a–c), and consequently increases the strength after ECAP. However, an initial reduction of strength occurs with pre-aging of 12 h, when the hardness decreases due to dissolution of some clusters acting as reinforcement particles [17]. Therefore, it seems likely that the dissolution of clusters decreases both the primary and multiplication rate of dislocation density (Fig. 10d). A secondary rise in hardness appears when the alloy is aged for 24 h before ECAP, which can be attributed to effective β+ precipitates increasing both the primary and multiplication rate of dislocation density (Fig. 10e). However, crystals aged to peak hardness show a slight decrease in strength compared with GP precipitates, probably because dislocations are no longer cutting through particles to form well-defined slip bands. In fact, they are known to move around particles so as to by pass them [27]. A final fall occurs due to overaging before ECAP. This overaging condition leads to primary obstacles which are too large. In other words, it seems this condition allows dislocations to accumulate in the tangles around the large particles in the process of passing between them, facilitating slip on secondary slip systems [20,27]. This would reduce primary dislocation density and multiplication rate of dislocations as well [22] (Fig. 10f). These considerations are in agreement with the strength results shown in Fig. 7.

The comparison of hardness variations for pre- and post-ECAP aging treatments at the same temperature of 180°C illustrates that pre-ECAP aging at 180°C is effective in strengthening the alloy, while post-ECAP aging at this temperature leads to quick fall in hardness values due to active recovery mechanisms (see Figs. 2 and 6). In fact, the effective strengthening has been masked in the post-ECAP aging treatment at 180°C (Fig. 2). On the other hand, the hardness in post-ECAP aging treatment at 180°C reaches to its maximum faster than quench aging at the same temperature (see Figs. 2 and 6). Both of these phenomena can be attributed to the role of ECAP strain before aging. The ECAP strain produces new dislocations in the matrix which is the driving force for recovery and decreases the required relevant temperature [27]. Thus, a lower aging temperature should be utilized for effective strengthening in aging treatment after ECAP. Moreover, the dislocations act as short-circuit paths for the solutes and facilitate the atomic migrations. This in turn decreases the activation energy for the growth of precipitates and enhances the aging kinetics after ECAP [17,35], and subsequently decreases the time at which the maximum hardness is achieved. In summary, while the aging without or before ECAP can be carried out at 180°C or so, aging after ECAP should be carried out at lower temperatures in order for effective strengthening of the alloy which is particularly important for industrial application of the alloy.

5. Conclusions

- Combination of ECAP with aging treatments can effectively increase the strength of 6082 Al alloy.
- The ECAP process enhances the strength of the alloy, but it reduces the ductility. An increase in both strength and ductility was achieved via a suitable post-ECAP aging treatment which is unusual according to the literature. This result demonstrates a successful strategy to achieve high strength and a moderate level of ductility by a suitable post-ECAP aging treatment, suggesting a potential for ECAP processing of age-hardenable alloys.
- The ECAPed specimen fractured with a fracture surface typical of shear rupture under tensile load. This was attributed to elongated dislocation cells and grain structure. However, following aging treatment tended to change this shear appearance of the fracture surface. This change in appearance was accompanied with increasing ductility.
- Pre-ECAP aging treatment was found to be slightly more effective in improving the strength than that of post-ECAP aging treatment, while more ductility was achieved by post-ECAP aging treatment. The higher ductility obtained by aging after ECAP can be related to the annealing effects of the aging treatment.
- Effective strengthening by post-ECAP aging treatment is possible at aging temperatures lower than those usually used for quench aging. This can be particularly interesting for industrial processing.
- The dislocation Taylor model for determination of dislocation density in pre-ECAP aged alloy was applied in order to study the comparative effect of precipitates on work hardening and dislocation density. The results were consistent with the common understanding of work hardening after precipitate coarsening.

Acknowledgements

The authors would like to thank the Iran National Science Foundation (INSF) and the Sharif University of Technology for the financial support and the provision of the research facilities used in this work.

References

[26] ASTM: Annual Book of ASTM Standards; metals test methods and analytical procedures, Vol. 03.01, American Society for Testing and Materials, West Conshohocken, 2004, E 8M.