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Characterization of deformation behaviour of aluminium alloy AA6082 processed by HPT

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Abstract

In this study, ultrafine grain structure evolution during high pressure torsion (HPT) of commercial aluminium alloy AA6082 at an elevated temperature is presented. Two different initial structural states of the alloy were prepared by thermal treatment. The dependence of the progress of microstructure refinement on shear strain was investigated by TEM observation of thin foils. The impact of various amounts of strain ($\varepsilon_{\mathrm{ef}}$) was analysed with respect to the increased temperature of deformation. Microhardness data measured across the deformed discs show scatter. Observation of microstructure revealed that an ultrafine grain (UFG) structure formed in the deformed disc as early as the end of the first turn, regardless of the initial structure of alloy resulting from the prior thermal treatment. The heterogeneity of UFG deformed structure in the deformed discs is consistent with the scatter in microhardness values. By increasing the strain level through adding turns ($N = 2, 4, 6$), the UFG structure was homogenized in the deformed discs. The effect of the increase in deformation temperature became more evident at higher straining when dynamic recrystallization modified the UFG structure locally. In the specimens prepared by two-stage thermal treatment (quenching and ageing) prior to torsion deformation, the growth of new grains was inhibited and the UFG microstructure was more stable. Tensile strength values suggest that strengthening was partially relaxed by local recrystallization. Torque vs. time plots reveal that the torque required to deform the sample was increasing until the completion of the first turn and then remained stable or even decreased slightly.

Key words: 6082 Al alloy, thermal treatment, structure modification, HPT severe plastic deformation, ultrafine grain structure, mechanical properties, torque

1. Introduction

Severe plastic deformation processing (SPD) of metals and alloys can lead to grain refinement and finally to the formation of a ultrafine grain structure. During the last decade, bulk nanostructured materials or materials with submicron structure produced by severe plastic deformation have been studied, as well as their microstructure development and its relation to mechanical properties. There are various SPD techniques (e.g. ECAP, ARB, HPT and CGP), which are capable of producing ultrafine grain (UFG) materials with submicrometre or even with nanometre grain size [1, 2].

The HPT process (high pressure torsion) uses discs as input stock. Recently, a ring-sample has been used [3]. When compared to other SPD processes, the HPT technique offers a number of advantages, as stated in [4]. In this process, shear strain is proportional to the distance from the disc centre, which means inhomogeneous microstructure develops across the diameter of the disc. Among all available severe plastic deformation techniques, HPT represents a simple and effective method, which produces reproducible and well defined samples. The high pressure applied prevents

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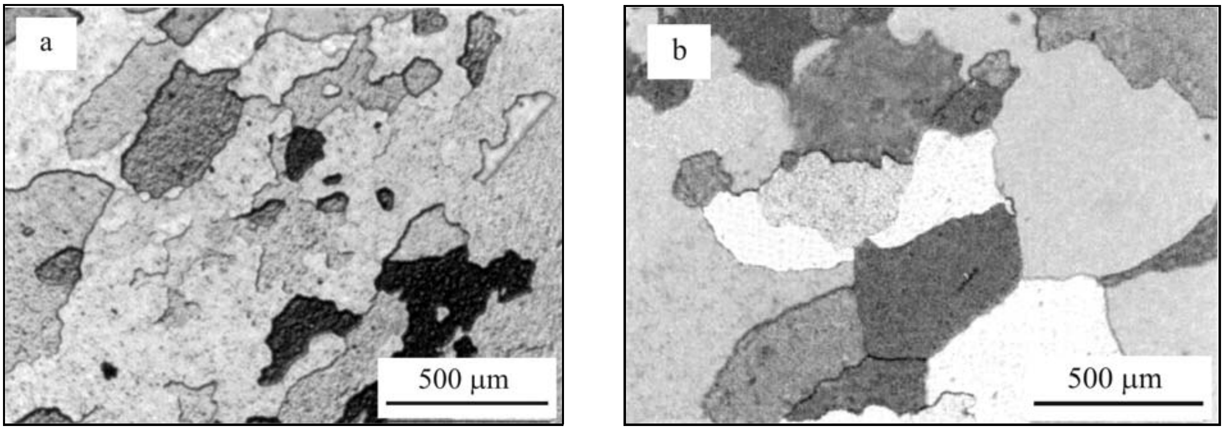

Fig. 1. Micrographs of initial microstructure of the alloy: a) as-cast (S1 state); b) after solutioning and ageing (S2, S3 states).

the sample from fracturing. Quantitative parameters describing the sample response during the torsion deformation can be correlated to the resulting microstructure.

The wide range of results available for various materials processed using HPT confirmed microstructure refinement, strength and hardness saturation taking place with increasing strain [5, 6]. There are differences between the microstructure development in pure metals, multiphase steels and alloys [7, 8].

In this study, discs of aluminium alloy AA6082 (EN AW6082), having different initial structure due to various heat treatment applied, were processed using HPT at elevated temperature. Microstructure refinement was analysed, taking into account the differences between the initial microstructures obtained by heat treating the alloy. The evolution of ultrafine grain microstructure was examined to find its relationship with hardness, torque and effective strain levels.

2. Material and experimental procedure

EN AW 6082 commercial aluminium alloy was supplied in the form of rods made by casting and extrusion. Their diameter was 20 mm. Prior to the HPT process, two additional states of the alloy were prepared by thermal treatment. The initial states of the alloy were as follows:

S1 – initial microstructure of as-received alloy produced by casting and extrusion;

S2 – the state S1 modified by annealing at 540 ℃ for 1.5 h, followed by water cooling (*solutioning + quenching*);

S3 – the state S2 + ageing at 160 ℃ for 12 h (*precipitation-hardened microstructure*).

The aim of the thermal treatment was to prepare different initial structures for the severe plastic deformation experiment. Specimens upon annealing (solutioning) and quenching (S2 state) were expected to undergo only strain hardening of the solid solution during SPD processing. Specimens prepared by additional long-term ageing (S3 state) leading to formation of precipitates in the matrix were intended to undergo a maximum precipitation hardening. The purpose of the precipitates was to modify the process of formation of the deformed microstructure during the torsion deformation process and to influence the deformation behaviour of tensile test specimens (contributing to the work hardening effect of the HPT process).

Microstructures of the initial alloy states are presented in Fig. 1a,b. The as-cast and extruded microstructure (S1) (Fig. 1a) was already altered by the solutioning step alone (S2), which produced equiaxed grains. However, no detectable changes in the microstructure were observed after additional ageing (S3) carried out after solutioning, despite rather long ageing time. The received microstructure resulted from solutioning and quenching thermal treatment (S2) and further modified by long term ageing (S2 + S3) was then quite coarse grained and is documented in Fig. 1b. Any morphological differences in structure were not observed when alloy experienced additional long term ageing.

HPT samples were prepared from the only thermally treated rods (S2, S3). The discs had a thickness of 0.92–0.95 mm and a diameter of 8 mm. HPT was conducted using the torsion facility shown in Fig. 2. It consists of upper and lower anvils with a shallow hole of 8 mm diameter and 0.8 mm depth in its centre. Various levels of strain were introduced in the specimens by torsion load through up to 6 revolutions. Each sample was placed in the hole and the lower anvil was rotated with respect to the upper one at a temperature of 350 ℃ and a rotation speed of 0.4 rpm under a pressure of 4 GPa. The rotation was terminated after N turns, where $N = 1$, 2, 4, or 6. The equivalent strain (von Mises strain) $\varepsilon_{\mathrm{eq}}$ as a function of the number of turns N was calculated according to the relationship:

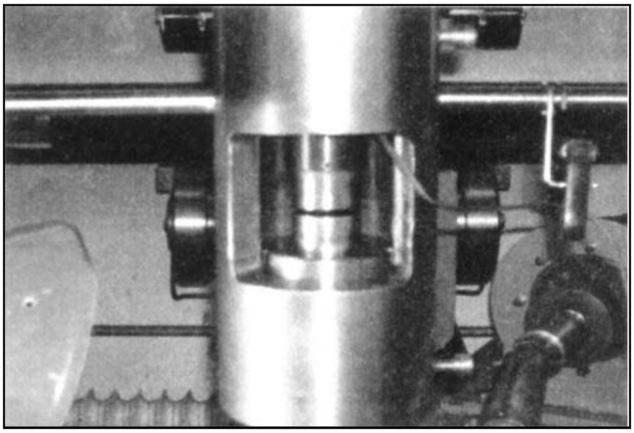

Fig. 2. HPT deformation facility.

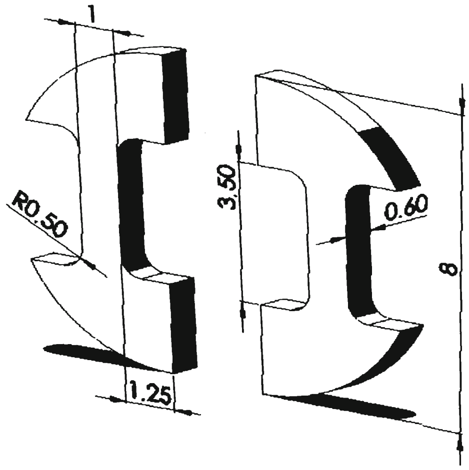

Fig. 3. Specimens used for tensile tests. The dimensions are in mm.

$$\varepsilon_{\mathrm{eq}} = 2\pi Nr/t\sqrt{3}, \qquad (1)$$

where r and t are the disc radius and disc thickness, respectively. The effective strains corresponding to $N = 1$, 2, 4 and 6 turns were $\varepsilon_{\mathrm{eq}} \sim 15$, 30, 60 and 90, respectively.

Changes in mechanical properties with increasing strain (number of turns) were explored by microhardness measurement, room-temperature tensile tests on sub-size specimens cut from outside the centre of discs (in radial direction) and by recording the torque. Vickers microhardness was measured at three points of the disc (at opposite disc edges L and R and in the disc centre C) after grinding with emery (SiC) paper with P1200 grit. A load of 30 N (HV3) was applied and the average value was calculated from 4 measurements at approximately identical distance from the centre of the disc. Microhardness measurement (kp mm^{-2}) at the disc edges and in the disc centre upon different amounts of shear deformation introduced was an effective technique to correlate microhardness values with mechanical strength [7, 8].

In order to measure mechanical strength, sub-size specimens from outside the disc centre were machined (Fig. 3). Two tensile specimens were prepared from each deformed disc sample.

The recorded torque value includes, in addition to the torque required for deforming the sample, a contribution from formation of the burr [9]. This contribution appears to be difficult to evaluate.

Transmission electron microscopy (TEM) was employed to examine the ultrafine grain microstructure and to correlate it with the number of turns and the corresponding effective strain $\varepsilon_{\mathrm{eq}}$. Discs with 3 mm diameter were cut from the centre and the peripheral area of the deformed disc. After thinning and polishing the discs, TEM micrographs were taken using JEOL JEM 2000FX operating at 200 kV.

3. Results and discussion

High pressure torsion processing of AA6082 aluminium alloy requires large effective strain $\varepsilon_{\mathrm{eq}}$ to refine its microstructure to submicron size. In general, the value of 15 could be sufficient for light metals. The equivalent von Mises strain as a function of the number of turns n is given earlier in this paper. There is a problem arising from HPT processing: the strain non-uniformity within the disc due to $\varepsilon_{\mathrm{eq}}$ value growing with the distance from the specimen centre. This results in non-uniform microstructure formation, when lower effective strain is applied. HPT processing thus requires higher effective strain to produce uniform ultrafine microstructure all across deformed disc. Electron microscopy analysis of thin foils prepared from locations in deformed discs (at their edge and in the centre) subjected to different levels of equivalent strain $\varepsilon_{\mathrm{eq}}$ depending on the number of turns revealed the microstructure variation across the discs.

3.1. Microstructure

The UFG microstructure was examined in order to evaluate the impact of the difference between the alloy initial microstructures prepared by thermal treatment, to assess the impact of effective strain on microstructure refinement and its homogeneity across the deformed discs. TEM micrographs in Fig. 4 (state S2) and Fig. 5 (state S2 and S3) show the UFG structure development across the deformed discs, upon different thermal treatments of the alloy prior to HPT processing, and upon exposure of different effective strain $\varepsilon_{\mathrm{ef}}$ corresponding to number of turns performed.

TEM micrographs clearly prove the development and presence of ultrafine grain structure at the disc

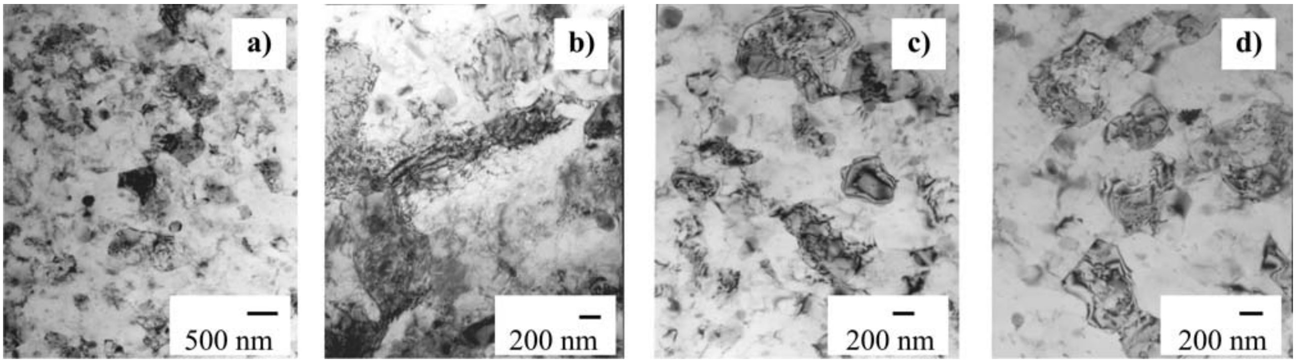

Fig. 4. TEM micrographs of HPT-processed alloy (initial microstructure state S2), in the disc centre upon the first turn $N = 1$ (a, c) and in the disc centre upon six turns $N = 6$ (b, d).

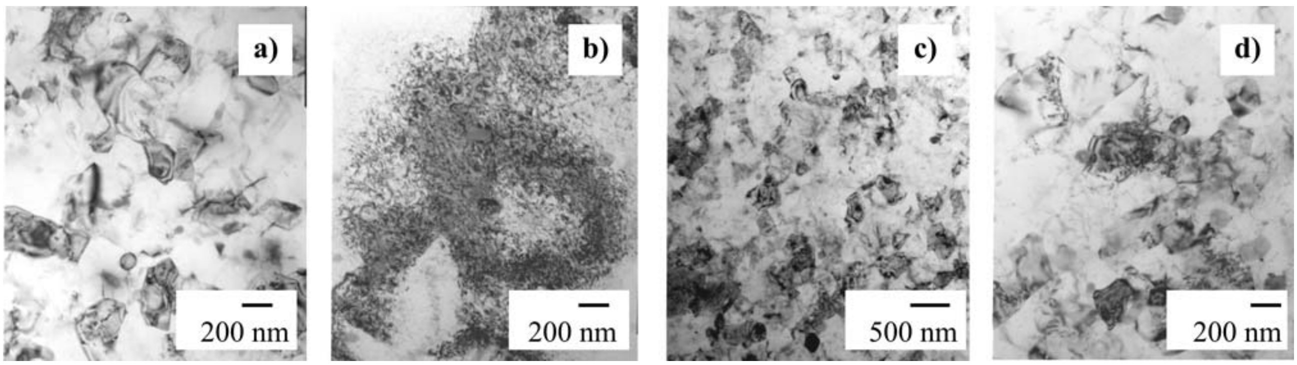

Fig. 5. TEM micrographs of HPT-processed alloy (initial microstructure S3) at the disc edge upon one turn $N = 1$ (a, c) and in the centre upon six turns $N = 6$ (b, d).

edge upon finishing the first turn ($N = 1$), regardless the initial microstructure of the alloy, which resulted from different thermal treatment applied, as shown in Fig. 4a (state S2) and Fig. 5a (state S2 and S3). Microstructure characteristics of ultrafine grain structure formed at the discs edge for both structural states are similar, no matter what the initial structural state of alloy was. Refinement of grains to about 500 nm in size, having high angle grain boundaries, was observed for both initial structural states of alloy. However in samples, performing the first turn, there is present considerable microstructure non-uniformity across the disc between its centre and edge. A dense dislocation network formed in the centre of the discs, regardless of the previous thermal treatment of the alloy, Fig. 4b, Fig. 5b. The state of microstructure suggests that the strain in the disc centre is not large enough to produce new grains with high angle boundaries. The tangled dislocation networks and scattered locations with recovered dislocations feature result from partial recovery and local reordering of dislocations towards subgrains, Fig. 4b, Fig. 5b. The structure results confirmed that the strain at the end of the first turn was not sufficient to refine the microstructure uniformly throughout the deformed disc. Microstructure observation yielded no evidence that the increased temperature of deformation modified in any way or enhanced the growth of newly formed fine grains. On the other hand, this increased temperature modified in some way the formation of deformed microstructure across the deformed disc by supporting dislocation rearrangement prior to recovery.

Upon $N = 6$ turns ($\varepsilon_{\mathrm{eq}} \sim 90$), the initial equiaxed coarse grained structure was significantly modified and ultrafine grained microstructure formed in almost the entire disc, except for some areas, where residue of dislocation tangles and recovered islands of subgrains in ultrafine grain structure were accidentally found. This is alike for both initial conditions of the alloy, as shown in Fig. 5. The size of individual new grains ranged from 200 nm to 1 μm. In the central part of the deformed disc (initial S2 condition), areas with partially recovered dislocation structure were distributed randomly, as in Fig. 5b. This fact may be attributed to the lower strain in the centre of the disc, where the driving force was insufficient for new fine grains to form. Due to the lower driving force, the less deformed structure in this central area underwent only partial and local recovery upon processing at elevated temperature.

Table 1. The results of Vickers microhardness as a function of the number of turns and location of measurement

Alloy state	Turns (N)	Initial thickness of disc (mm)	Final thickness of disc (mm)	HV3		
				L	C	R
S_2	1	0.92	0.64	83	112	83
S_2	2	0.94	0.65	83	111	82
S_2	4	0.95	0.63	110	118	108
S_2	6	0.95	0.61	102	110	102
S_3	1	0.94	0.65	116	120	105
S_3	2	0.94	0.65	105	113	105
S_3	4	0.95	0.61	108	116	109
S_3	6	0.95	0.63	101	112	101

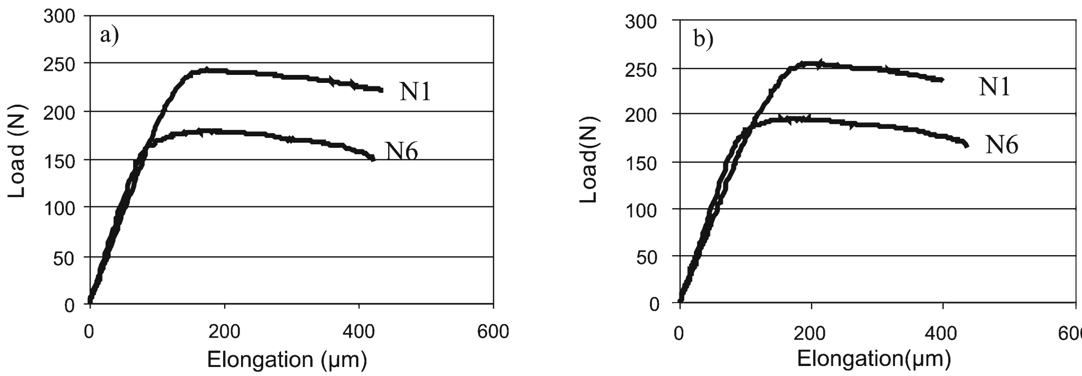

Fig. 6. Tensile test plot of load vs. elongation: a) state S2, $N = 1$, $N = 6$ turns; b) state S3, $N = 1$, $N = 6$ turns.

3.2. Mechanical properties

3.2.1. Hardness

Measurement of Vickers microhardness was used to explore the strengthening (work hardening) across the disc: on the left (L) and right sides (R) and in the centre (C). Hardness (HV3) for both alloy states and for a given number of turns is shown in Table 1. Hardness values (effect of strain hardening) depend on the initial microstructure. The decrease in hardness suggests that softening took place, regardless of the location (edge/centre of disc) upon six turns in both states S2 and S3 (Table 1). However, in the solutioned and aged alloy (S3 samples), gradual decrease in hardness was found upon an even lower number of turns, that is for $N = 2$, 4, and 6. This softening is in contradiction to the expected effects of ageing. It may be attributed due to local dynamic recovery and recrystallization of the deformed structure, which might have contributed to the softening of the refined microstructure. Ultrafine grain features, particularly at disc edges, support this assumption of selective growth of fine grains, as shown in Fig. 4c,d. Signs of recovery of ultrafine microstructure were found upon $N = 6$ turns, regardless of the initial microstructure, Fig. 4. The softening effect was also apparent in the tensile test records regardless of the initial state of the alloy, where a decline in strength occurred.

3.2.2. Tensile deformation behaviour

Load vs. elongation plots for the lowest and highest strain levels ($N = 1$ and $N = 6$) are shown in Fig. 6. Specimens of the alloy upon different thermal treatments (S2 and S3) showed similar strength and ductility values. The ageing step increased the strength by a small amount only. Additional precipitation may have occurred at the forming temperature of 350 °C. Its contribution to strength and ductility was, however, apparently equally small. Strength values found by tensile test were changing with the number of turns. Strength decreased with increasing number of turns, regardless of the initial heat treatment. Metallographic observation suggests that softening is due to the processing conditions (temperature and strain level), which caused ultrafine grain structure to form, but also initiated recovery and recrystallization. Local growth of fine grains on the disc periphery and dislocation density decline within the grains were observed. The decrease in strength values and the presence of a short strain hardening phase in the deformation records at the highest strain levels ($N = 4$ and $N = 6$ turns) are then probably due to recovery process,

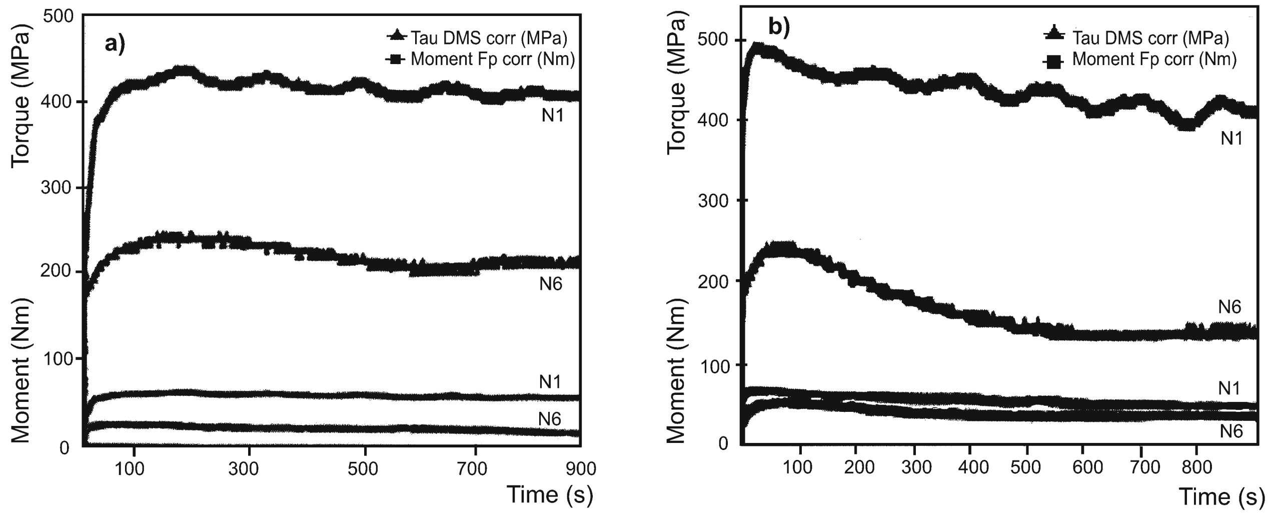

Fig. 7. Torque curves for processing with $N = 1$ and $N = 6$ turns and microstructure states S2 in a) and S3 in b).

which could have led to grain growth. The softening, which took place in both structural states of the alloy, disproves the efficiency of ageing for matrix strengthening by Mg_2Si (β') precipitation, although the proposed heat treatment was aiming at such type of strengthening.

3.2.3. Torsion deformation behaviour

The impact of the introduced strain on mechanical properties of the material can be explored through measuring the torque in the process. In addition to the torque required for deforming the disc, M_d, the measured torque contains a contribution from the torque required for burr forming M_b. From the torque value the shear stress can be then calculated.

The HPT deformation method proved effective in introducing large shear deformation and in refining the coarse grain Al alloy structure. Selected torque vs. time plots (one turn took 150 s) recorded at 350 °C for quenched (S2) and quenched and aged (S3) alloy states are presented in Fig. 7a,b, respectively. In these plots, intensive strain hardening section is followed by a steady state deformation stage. As the records show, the magnitude of torque was lower but the strengthening effect lasted longer in the alloy where only solutioning was performed. The highest torque and greater strengthening prior to saturation were recorded for the quenched and aged state (S3), as shown in Fig. 7b. The strain (stress) necessary to reach this steady state torsion deformation depends on the material condition (microstructure). The deformation behaviour of the material in both initial states appears to be very similar, regardless of the initial microstructure. Upon higher number of turns, the softening was more evident. A possible reason for this could be the progress of on-line recovery and recrystallization process, due to the quite high HPT deformation temperature, Fig. 4. At the same time, the contribution of precipitation of Mg_2Si (β') particles to controlling the grain refinement and recovery processes does not appear significant, since the characteristics of ultrafine grain microstructure appear rather similar in the two groups of specimens with different initial conditions obtained by heat treatment. There was one difference, however, which was the increase in torque at the end of the first turn in processing the S3 material (see Fig. 7b). This might be related to the effects of precipitation, more intensive strengthening and thermal stability of the ultrafine microstructure. At the same time, the continuous decrease in torque upon higher number of turns does not seem to indicate any effect of the precipitation on the deformed ultrafine microstructure. Both groups of samples (obtained by different initial heat treatment procedures) showed effects of recovery and recrystallization of the deformed microstructure. Along with the grain refinement, the mechanical strength increases until saturation point, and then starts to decrease, no matter what the initial structure was upon thermal treatment.

4. Summary and conclusions

Severe plastic deformation processing of Al alloy AA6082 was performed to study the influence of large shear deformation on its microstructure development and mechanical properties. Two initial microstructure states of the alloy were prepared to explore the impact of secondary phase particles. The thermal processing was effective in obtaining different initial structures of the alloy.

At the same time, the contribution of precipitation of Mg_2Si (β') particles to controlling the grain refinement and recovery processes does not seem significant, since the characteristics of ultrafine microstructure appear rather similar in the two groups of specimens with different initial conditions obtained by heat treatment. There was one difference, however, which was the increase in torque at the end of the first turn in processing the S3 material. This might be related to the effects of precipitation, more intensive strengthening and thermal stability of the ultrafine microstructure. At the same time, the continuous decrease in torque upon higher number of turns does not seem to show any effect of the precipitation on the deformed ultrafine microstructure. Both groups of samples (obtained by different initial heat treatment procedures) showed effects of recovery and recrystallization of the deformed microstructure.

Strengthening intensity and mechanical properties were higher in those alloy samples where precipitation process was introduced at alloy treatment.

The required torque was higher but less stable for samples upon the precipitation treatment. With the progress of microstructure refinement, the mechanical strength increases until it reaches a saturation point. Then it starts to decrease, regardless of the initial microstructure of the alloy. The effect of softening was observed in both alloy states in the form of a short work hardening stage in the tensile plots.

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