The Role of Mandrel Rotation during CSET Processing Demonstrated on a 3003 Aluminium Alloy

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Abstract: Recently, the complex shearing of extruded tube (CSET) technique was proposed representing a new way of processing tubes from a bulk billet as a combination of extrusion with two passes of equal channel angular processing and possible mandrel rotation. In the present paper, the influence of mandrel rotation on the final microstructure and mechanical properties of the tube fabricated of a 3003 aluminium alloy was investigated. Electron back scatter diffraction (EBSD) revealed differences in grain size and misorientation angles between tubes processed with stationary and rotating mandrel. Kernel average misorientation maps obtained from EBSD experiments and ASTAR analysis performed in transmission electron microscopy proved differences in the dislocation density and arrangement. The tube processed using a stationary mandrel showed a recovered microstructure with elongated grains separated by low-angle grain boundaries into subgrains. The microstructure of the tube processed with a rotating mandrel was dynamically recrystallized with the grain size in submicrometer range. Vickers microhardness measurements revealed only a 40% HV increase in the sample prepared using stationary mandrel as compared with the initial billet. The mandrel rotation resulted in a much higher HV increase up to 200% as a result of substantial grain refinement.

Keywords: tube forming; complex shearing of extruded tube; severe plastic deformation; microstructure; microhardness

1. Introduction

Methods like equal channel angular pressing (ECAP) [1] and high-pressure torsion (HPT) [2] are known as effective severe plastic deformation (SPD) methods resulting in significant increase of the strength especially due to grain refinement [3]. Grain size is namely one of the important microstructural features affecting mechanical and corrosion properties of materials [4–6]. The need for strength enhancement of tubular samples led to modification of these SPD methods. The tubular samples were filled with sand, rubber, or grease as mandrel to preserve the tubular shape of samples when pressed through traditional ECAP. Enhanced tensile strength and hardness were observed in these tubes [7,8]. Other modifications used a solid rigid mandrel, and the sample was passed through a channel between the hollow die and the cylindrical mandrel with a thinned zone where the billet is deformed. In tube channel pressing (TCP) [9], the material passes through a bottleneck region leading to shear and hoop straining. In related techniques as parallel tubular channel angular pressing (PTCAP) [10] and tubular channel angular pressing (TCAP) [11], the sample is strained by passing through a channel with two or three shear zones, respectively.

Other methods of tube processing were introduced as modifications of HPT exploiting deformation under high hydrostatic pressure. Tube high-pressure shearing (t-HPS)
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and high-pressure tube twisting (HPTT) [14,15] are typical representatives of these methods. In both methods, high hydrostatic pressure is applied on a tube positioned between inner and outer mandrels, each of which may rotate independently thus evoking shear deformation of the tube wall. Similarly, in the high-pressure tube shearing (HPTS) process, a tube has been drawn through a rotating die, and an internal rotating mandrel causes thinning of the tube wall [16]. Rotation and high hydrostatic pressure resulted in local shear deformation which resulted in a fine-grained and frequently gradient microstructure.

All the mentioned methods were developed to apply initially tubular samples. Recently, we developed the method called complex shearing of extruded tube (CSET) which is capable to prepare tubular sample with an ultra-fine-grained microstructure and enhanced strength in one step from a cylindrical bulk sample [17,18]. This method combines the tube extrusion ensued by two consecutive ECAP passes. Additionally, the mandrel can rotate and introduce further twist deformation into the tube due to friction between the rotating mandrel and produced tube. The role of individual partial deformation processes was described and discussed in [18]. The microstructural development during different stages of CSET along with mechanical properties was studied on a 3003 aluminium alloy. An overall grain refinement by 3 orders of magnitude was reported which led to remarkable hardness enhancement [18].

The present study focuses on the role of mandrel rotation during CSET. It compares the microstructure at different scales and the microhardness of CSET tubes fabricated from a 3003 aluminium alloy with and without mandrel rotation.

2. Materials and Methods

The 3003 aluminium alloy was chosen for application of CSET processing. The main advantage of this alloy is its good deformability even at room temperature (RT). The alloy was obtained by melting of Al, Mn, and Cu, all of high purity, in a Balzers VSG-02 vacuum furnace (Balzers, AG, Balzers, Liechtenstein) using graphite crucible and argon atmosphere. To remove internal stress and facilitate workability, the received cast sample was annealed for 4 h at 450 °C and cooled in the furnace [19]. The billets 55 mm long with a diameter of 11 mm were machined from the annealed ingot. The chemical composition of the alloy determined with an Ametek EDAX Orbis X-ray Fluorescence analyzer (Ametek GmbH, Unterschleissheim, Germany) was 0.12 wt.% of Cu, 1.2 wt.% of Mn and Al as balance.

CSET processing was applied to cylindrical bars at RT. The bars lubricated on the whole surface by GLEIT-µ HP 515 paste (Wessely GmbH, Korneuburg, Austria), were placed into the CSET device schematically shown in Figure 1. The bars were pressed by a plunger moving with the velocity of 0.2 mm s⁻¹. The bars of one set were pressed with a stationary mandrel (0 rpm), the bars of the other set were processed with a simultaneous mandrel rotation of 0.2 Hz (12 rpm). In both cases the tubes with an outer diameter of 26 mm and thickness of the wall of 2 mm were produced. The die as well as the mandrel were fabricated from Tenasteel® with the cone angle of 100°. All manufacture details are described in [17].

The microstructure of the initial bar was studied using light microscopy (LM; Zeiss Axio Observer D1m microscope (Carl Zeiss Microscopy GmbH, Jena, Germany)). The grain structure was visualized using Barker solution. The resolution of LM was not sufficient to reveal microstructure details in CSET processed samples. Scanning electron microscopy (SEM; FEI Quanta 3D FEG (FEI Czech Republic s.r.o., Brno, Czech Republic)) and transmission electron microscopy/scanning transmission electron microscopy (TEM, STEM; FEI Tecnai TF20 X-twin (FEI Czech Republic s.r.o., Brno, Czech Republic)) were used for the microstructure investigation of CSET processed tubes. Electron backscatter diffraction (EBSD) measurements were performed in SEM on samples which were polished electrolytically at −20 °C for 120 s with 10 V in a 10% solution of HClO₄ in ethanol. The step size of EBSD measurements was 50 nm. The grain size, grain orientation, and character of interfaces were evaluated. Additionally, the lattice distortion of the grains was determined using Kernel average misorientation (KAM) maps. TEM samples were made by ion milling.
using PIPS (Gatan Ametek, Inc., Pleasanton, CA, USA, 5 keV, 5° and 3°). Automated orientation mapping measurements were performed at TEM JEOL 2200FS equipped with “Spinning Star” electron precession with an ASTAR software package.

![Figure 1. Schematic illustration of individual stages in the CSET process. (a) start of CSET; (b) situation after extrusion and 2nd ECAP pass; (c) final stage of CSET.](image)

Measurements of the Vickers microhardness were performed using Struers Duramin 2 microhardness tester (Struers, Copenhagen, Denmark) under a load of 0.49 N (HV0.05). Twenty indents were performed at any locality, each indent was held for 10 s.

3. Results

The scheme of the CSET device and the operational steps are presented in Figure 1. During CSET processing, the cylindrical billet is inserted in the die above the mandrel (Figure 1a). Using the plunger, the billet is pressed against the mandrel with a defined translational velocity. In course of this pressing, the billet is first deformed by extrusion into a tube around the mandrel tip and then the material goes through two consecutive ECAP passes (Figure 1b). Additional straining can be achieved by a concurrent rotation of the mandrel. All deformation is introduced into the material in the area around the top part of the mandrel, denoted as main deformation zone (MDZ—see in the red inset in Figure 1b). Behind MDZ the mandrel diameter is reduced, and the tube is no more strained (Figure 1c).

The microstructural details of both CSET processed samples were studied at the center of the tube wall on three perpendicular planes (see Figure 2). The grains in the initial unprocessed bar were of millimeters length and of the width of few tenths of millimeters (Figure 3a). Figure 3 presents the grain structure on the PD plane of the samples processed by CSET with stationary (b) and rotating (c) mandrel using SEM-BSE contrast. Both samples exhibit a refined grain structure compared to the initial bar. The grain structure in the sample processed with stationary mandrel (Figure 3b) is coarser than that of the sample prepared using mandrel rotation (Figure 3c). The grain structure in the SEM-BSE contrast is revealed by channeling contrast, which is sensitive to small changes in orientation, thus Figure 3b,c forebodes the subgrain structure present in the PD of both processed samples.
Figure 2. Schematic depiction of the cross section of the CSET processed sample and definition of the directions: PD—pressing direction, ND—normal direction, RD—radial direction.

Figure 3. The grain structure (a) in the center of the initial bar (LM), and that of samples processed with (b) stationary, and (c) rotating mandrel on the PD plane, SEM-BSE.

The orientation image maps (OIMs) taken from 3 perpendicular directions, ND, RD, and PD, for the sample processed with stationary mandrel are presented in Figure 4.

The OIMs reveal that original large grains were highly deformed and split into fine regions of similar crystallographic orientation. The percentage of high-angle grain boundaries (HAGBs) with misorientations above 15° was below 50%. This result is clearly confirmed in Figure 4d–f showing OIMs taken from a much smaller area, i.e., with a higher resolution. These figures show that low-angle boundaries (LAGBs; marked in red) are more frequent than HAGBs (marked in blue). The microstructure is mostly composed of elongated subgrains inclined parallel to the shear direction during ECAP with a length of tens of micrometers and a width of units of micrometers.

The bottom Figure 4g–i show the same OIMs as Figure 4a–c, respectively, but the regions with misorientation larger than 5° are colorized by unique colors without any relationship to changing crystallographic orientations. The presence of small grains inside larger ones reveals that recrystallization process started at some places but volume fraction of the recrystallized fraction is small.

The same investigation was performed on the sample processed with rotating mandrel. OIMs in Figure 5 were taken from 3 perpendicular directions, ND, RD, and PD, the size of scanned area was the same as in case of Figure 4d–f.

On the ND plane, the grains were found to be elongated along the direction of the translational movement of the sample, PD. The mean grain sizes in PD and RD determined as 1.74 of the mean intercept [20] were close to 0.8 µm and 0.4 µm, respectively. A slightly lower grain aspect ratio was observed on the RD plane. These results show that the grains in the tube wall are prolonged in the axial direction and compressed especially in the radial
direction. Figure 5d–f colorized by unique color (considering a minimum misorientation of 5°) show that large grains observed in the original bar are split into a submicrometer grains or subgrains after processing with rotating mandrel. The proportion of HAGBs was 80–90% in all maps, i.e., most objects observed in Figure 5d–f are recrystallized grains.

To obtain information on the local strain distribution inside grains of the final tubes, KAM maps were evaluated from EBSD measurements. KAM describes the average misorientation of a selected point relatively to all neighboring points. The distance between neighboring points was selected 50 nm in our measurement. The KAM threshold value was chosen to be 5° to exclude large misorientations due to subgrain boundaries or LAGBs with higher misorientations (5 to 15°). The correlation between the local lattice distortion and the color in KAM maps is given in Figure 6.

![Figure 3](image-url)

**Figure 3.** The grain structure (a) in the center of the initial bar (LM), and that of samples processed with (b) stationary, and (c) rotating mandrel on the PD plane, SEM-BSE.

![Figure 4](image-url)

**Figure 4.** Orientation image maps taken on the ND (a,d,g), PD (b,e,h) and RD (c,f,i) planes of the sample processed with stationary mandrel. Panels (a–f) were colorized according to IPF (see orientation triangle), and panels (g–i) were colorized by unique color.
The OIMs reveal that original large grains were highly deformed and split into fine regions of similar crystallographic orientation. The percentage of high-angle grain boundaries (HAGBs) with misorientations above 15° was below 50%. This result is clearly confirmed in Figure 4d–f showing OIMs taken from a much smaller area, i.e., with a higher resolution. These figures show that low-angle boundaries (LAGBs; marked in red) are more frequent than HAGBs (marked in blue). The microstructure is mostly composed of elongated crystallized fraction is small.

To reveal the finest microstructural details of both samples TEM was carried out on foils cut parallel to the ND plane. Figure 7 shows the results of the ASTAR measurements investigated areas was estimated to be 5° to exclude large misorientations due to subgrain boundaries or LAGBs. The bottom Figure 4g–i show the same OIMs as Figure 4a–c, respectively, but the regions correspond to places with very low lattice distortion, i.e., with low density of lattice dislocations. KAM maps in Figure 6d–f (corresponding to OIMs in Figure 5a–c, respectively) show a more homogeneous distribution of the local strain in the sample processed with rotating mandrel. This corresponds well to the dynamically recrystallized character of the microstructure where dislocation walls or subgrain boundaries with small misorientations. KAM maps in Figure 6a–c (corresponding to OIMs in Figure 4d–f, respectively) reveal high contrast variation in the sample processed with stationary mandrel. Relatively large blue areas correspond to subgrain interiors. The blue areas are surrounded by yellow regions where the local strain is very high. These regions probably correspond to places with very low lattice distortion, i.e., with low density of lattice dislocations. The mean grain sizes in PD and RD determined from EBSD measurements. KAM describes the average misorientations of a selected point relatively to all neighboring points. The distance between neighboring points was selected 50 nm in our measurement. The KAM threshold value was chosen to be 5° to exclude large misorientations due to subgrain boundaries or LAGBs. The same investigation was performed on the sample processed with rotating mandrel. The proportion of HAGBs was 80–90%.

KAM maps evaluated from the OIMs corresponding to the tube processed with stationary mandrel (a–c) were colorized according to IPF (for color scheme, see the orientation triangle in Figure 4), and panels (d–f) were colorized by unique color.

Figure 5. Orientation image maps taken on the ND (a,d), PD (b,e) and RD (c,f) planes of the sample processed with rotating mandrel. Panels (a–c) were colorized according to IPF (for color scheme, see the orientation triangle in Figure 4), and panels (d–f) were colorized by unique color.

Color Coded Map Type: Kernel Average Misorientation

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Figure 6. KAM maps evaluated from the OIMs corresponding to the tube processed with stationary mandrel (a–c) and to the tube processed with rotating mandrel (d–f).
KAM maps in Figure 6a–c (corresponding to OIMs in Figure 4d–f, respectively) reveal high contrast variation in the sample processed with stationary mandrel. Relatively large blue areas correspond to places with very low lattice distortion, i.e., with low density of lattice dislocations. It can be supposed that these areas correspond to subgrain interiors. The blue areas are surrounded by yellow regions where the local strain is very high. These regions probably represent dislocation walls or subgrain boundaries with small misorientations. KAM maps in Figure 6d–f (corresponding to OIMs in Figure 5a–c, respectively) show a more homogeneous distribution of the local strain in the sample processed with rotating mandrel. This corresponds well to the dynamically recrystallized character of the microstructure where dislocations are distributed more homogeneously inside grain interiors and no dislocation walls or subgrain boundaries are formed. The GND density estimated as a mean value from the whole investigated areas was estimated to be $1 \times 10^{12} \text{m}^{-2}$ in the sample processed with stationary mandrel, whereas approximately $1.5 \times 10^{11} \text{m}^{-2}$ in the sample processed with rotating mandrel.

To reveal the finest microstructural details of both samples TEM was carried out on foils cut parallel to the ND plane. Figure 7 shows the results of the ASTAR measurements where the local lattice distortion was measured in a similar way as KAM in SEM.

![Figure 7. Lattice distortion for tubes processed with stationary (a) and rotating (c) mandrel and corresponding OIM maps colorized according to IPF for stationary (b) and rotating (d) mandrel.](image)

The results in Figure 7 agree well with previous results obtained from EBSD measurements. Larger subgrains of the size frequently exceeding 1 µm divided by very low-angle boundaries were found in the sample processed with stationary mandrel. The microstructure of the sample processed with rotating mandrel is composed of submicrometer grains divided predominantly by HAGBs. The grains are elongated, and the mean sizes evaluated by the same method as from EBSD are 800 nm and 400 nm, respectively, i.e., identical with data estimated from EBSD.

Figure 8 displays the real microstructures of both samples. Elongated subgrains with a length of about 1 µm were found in the sample processed with stationary mandrel (Figure 8a). The area denoted by a red circle in Figure 8b is shown as dark field image in Figure 8c formed by the selected diffracted beam on the inset in Figure 8b. It reveals that subgrains form grains with a size of several micrometers. The area denoted by a red square in Figure 8b was investigated at higher magnification (Figure 8d) and the presence
of dislocations predominantly arranged into dislocation walls and subgrain boundaries was observed.

Figure 8. TEM images of the microstructures of the samples processed with stationary mandrel (a–d) and of the sample processed with rotating mandrel (e–h).
TEM analysis of the microstructure of the sample processed with rotating mandrel documents clearly its submicrocrystalline character (Figure 8e). Selected area electron diffraction (inset in Figure 8e) was taken from the area denoted by a red circle and proves the random orientation of grains. The dark-field image taken using selected diffraction spot in SAED confirms the submicrometer grain size and non-correlated orientation of the grains (Figure 8f). The bright-field image taken at higher magnification confirms the presence of elongated grains with a length close to 500 nm (Figure 8g). Some grains were free of dislocations, whereas tangles of dislocations were also observed in some grains (Figure 8h).

To study the differences in the mechanical behavior of the samples processed with stationary and rotating mandrel, microhardness measurements were carried out. Figure 9 presents the comparison of measured average values of microhardness on different planes. The average microhardness of the sample processed with stationary mandrel was of around 56 ± 4 HV0.05 for all measured planes, which is about 40% higher than the value of 40 ± 1 HV0.05 measured on the annealed sample. Much higher values of the microhardness were measured for the sample processed with rotating mandrel. The highest value, 120 ± 3 HV0.05, was detected on the ND plane of the processed tube, whereas 98 ± 3 HV0.05 and 85 ± 3 HV0.05 were measured on the RD and PD planes of the same sample, respectively. In all cases, the microhardness increase is higher than 100% in comparison with the initial material.

![Figure 9. Evolution of microhardness during CSET with stationary and rotating mandrel.](image)

4. Discussion

CSET includes several SPD processes which represent a rather complicated straining route. The schematic detail of MDZ in Figure 10 shows that following SPD processes are applied consecutively:

(a) Extrusion (at position a),
(b) ECAP pass 90° No. 1 joined with circumferential strains, (b),
(c) Shearing of the vertical part of the tube resembling HPTT, (c),
(d) Shearing of the horizontal part resembling HPT, (d),
(e) ECAP pass 90° No. 2, (e).

Each of these processes contributes to the equivalent strain cumulated in the processed sample. It was shown in [18] that the tube extrusion and two ECAP passes, which are the only deformation processes during CSET with stationary mandrel, lead to an equivalent strain of approximately 4. The rotational movement of the mandrel results in additional shearing due to the friction between the rotating mandrel and the sample surface and
enhances the cumulative equivalent strain. Considering the geometry of the processed tube, the HPTT like deformation at the mandrel tip (the position c prior to the 1st ECAP pass) would lead to a strain value of 2.8, which is close to the values determined by finite element modelling of similar process—sideways extrusion—in [21]. A similar process at position c after the 2nd ECAP pass should yield a much higher strain value of 21.7 [22]. Both values were calculated for one turn in case of very high hydrostatic pressure which completely hinders any slip between the mandrel and tube wall. The HPT like process occurring at position d in the horizontal direction between both ECAP passes should introduce a total strain of 8.7 for one turn [2]. However, due to the setup of the CSET device no high hydrostatic pressure can be maintained during CSET processing to ensure the absence of slip between the material and mandrel. Therefore, only a fraction of the mandrel movement is transferred to the material, and the introduced strain will be much lower. Unfortunately, we are not able to measure experimentally the amount of rotation performed by the material itself, probably it will be only a part of one turn. Despite of this, a substantial role of the mandrel movement during CSET processing can be expected.

![Figure 10. Detail of MDZ (Figure 1b) presenting the location where individual SPD processes occur. For locations (a–e), see items (a–e) in the text (beginning of Discussion).](image)

It is well known that application of SPD leads to a significant grain refinement and, consequently, to an increase in strength [2]. The changes of the microstructure start with gradual increase in dislocation density followed by their arrangement into subgrain boundaries. With the ongoing straining, LAGBs transform to HAGBs, and the resulting microstructure can be considered as dynamically recrystallized with refined grains divided by HAGBs. Despite this general trend, it is not possible to find a unique correlation between the amount of introduced strain and the developed microstructure. This correlation depends both on the deformation conditions (especially on temperature) and on the material investigated.

Temperature is probably the most important parameter. SPD cannot be applied at too low temperatures because of limited plasticity of materials. On the other hand, high deformation temperature can accelerate restoration processes and lead to grain coarsening. Room temperature corresponds to the homologous temperature of 0.33\(T_{\text{melting}}\) for pure Al which seems to be a good compromise for SPD processing. Al exhibits a good plasticity at room temperature and is thus probably the most frequent material at which the influence of different severe plastic deformation processes on the microstructure was studied.

The correlation between the induced strain and microstructure evolution can be studied in the simplest way in ECAP where each pass brings the strain close to 1. The gradually increasing number of passes then multiplies this strain. The influence of up to 16 ECAP passes on microstructure was studied in commercial purity aluminium [23]. It was found that microstructure evolution (grain size, disorientation of neighboring grains) exhibits a saturation after 4th pass, the influence of further passes is negligible. The minimum grain size is slightly below 1 \(\mu\)m and the mean misorientation angle of grains

\[\begin{align*}
\mu_a & \rightarrow \mu_e,
\end{align*}\]
is slightly above 20°. There are two possible reasons which might explain this saturation. SPD can be accompanied by heat evolution and a temperature increase of 25 to 30° was measured during ECAP of pure aluminium [24]. The balance between the grain refinement due to straining and the grain growth due to the sample temperature rise associated with deformation, can result in a nearly constant grain size after high number of ECAP passes. Another explanation is based on the idea that the size of subgrains formed during deformation is determined by the applied stress. If the subgrains are larger than the actual grains, new subgrains cannot be formed and the actual grains cannot be further refined. The dislocations formed during straining are then absorbed by a huge number of grain boundaries and the microstructure does not change.

Alloying with appropriate elements can slow down the grain coarsening either due to hindering dislocation motion by solute atoms or due to pinning effect of second phase particles. This is the case of the Al3003 alloy used in our research. Experiments performed on the 3003 aluminium alloy processed by ECAP [25] revealed that the sample after four ECAP passes contained a mixture of elongated grains containing numerous LAGBs and small equiaxed grains with the average size close to 500 nm. The ratio of HAGBs was close to 60%. The increase of the number of ECAP passes to eight did not change the microstructure significantly, a relatively high ratio of LAGBs was still present [26].

As mentioned above, extrusion and two ECAP passes are the only SPD processes involved in the CSET with stationary mandrel. Although the introduced strain should be close to 4, i.e., comparable with that of the sample after four ECAP passes, the microstructure in the CSET processed tube is much less developed. It contains mostly elongated grains with LAGBs inside and only scarce fine recrystallized grains. This microstructure character, which was proved also by KAM and ASTAR analyses, resembles more to the sample after two ECAP passes. It can be concluded that the extrusion at the start of the CSET process has a small influence on the microstructure of the CSET processed tube and leads only to an increase in dislocation density.

The mandrel rotation during CSET leads to torsional straining of the material similar to that in HPTT and HPT like processes. The HPTT was shown to lead to a substantial grain refinement in a commercial purity Al, a half turn already led to a grain size of 650 nm [14]. During HPT, a single turn under a pressure of 1 GPa at RT was shown to lead to a grain size of 1 µm or slightly lower in a commercial purity Al sample [27]. These results show the ability of the torsional straining to reduce the grain size of the processed material. These HPTT and HPT like deformation processes were shown in [18] to lead to grain refinement in the CSET processed 3003 aluminium alloy even before the ECAP passes. These processes resulted in a substantial grain refinement also in the present research due to combined effect of the strain energy stored during severe plastic deformation and additional rotational deformation. A fine-grained microstructure with the grain size close to 500 nm and a high ratio of HAGBs were found in the center of the tube wall after CSET processing with rotating mandrel. The KAM and ASTAR analyses simultaneously with TEM observation showed that many recrystallized grains contained dislocations. The microstructure corresponds thus to a dynamically recrystallized state which undergoes further deformation. An expected fast absorption of lattice dislocations by HAGBs represents the main restoration process and results in a lower dislocation density inside grains as compared to the sample processed with stationary mandrel.

Severe plastic deformation leads generally to a substantial increase of the strength [3]. The experiment performed on Al of commercial purity [23] revealed an increase in microhardness from 20 to 50 HV already after 1st ECAP pass due to an increase in dislocation density. Following three ECAP passes resulted in increase of the microhardness to 60 HV and further ECAP passes (up to 16) did not influence its value more. This reflects the saturation of the microstructure evolution discussed above.

The values of the microhardness measured in tubes processed by CSET are higher than those for unprocessed material. The tube processed with stationary mandrel exhibits an increase of the values of microhardness by about 40% in comparison with the value
measured for the initial unprocessed sample. This value reflects the effect of the surviving coarse-grained microstructure, and the main strengthening is caused only by dislocations introduced during CSET processing. The additional HPTT-like and HPT-like deformation processes introduced due to the mandrel rotation resulted in an increase of the values of the microhardness by more than 100%. This strengthening is caused predominantly by a substantial grain refinement (more than 3 orders). KAM experiments showed a lower dislocation density in the tube processed with rotating mandrel, and therefore a lower contribution of dislocation strengthening can be expected in this sample. The measurements performed on different planes revealed microhardness anisotropy in the sample processed with rotating mandrel. This effect can be partially caused by a slightly different grain sizes along different direction and/or different textures at individual measured planes. However, no precise experiments were yet performed to prove this idea.

5. Conclusions

The conclusions of this paper can be summarized as follows:

- CSET method can produce strong tubular samples of the 3003 aluminium alloy;
- The microstructure of the sample processed by CSET using stationary mandrel is composed of large, elongated grains divided by low-angle boundaries into micrometer-sized subgrains. Only scarce fine recrystallized grains were observed;
- The mandrel rotation led to HPT- and HPTT-like deformation processes and enhanced the amount of introduced strain;
- The microstructure of the tube processed with rotating mandrel is ultra-fine grained with the grain size close to 500 nm. The ratio of high-angle boundaries exceeds 80%;
- KAM and ASTAR experiments confirmed that the microstructure of the tube processed with stationary mandrel is recovered with many dislocations arranged into dislocation walls or low-angle boundaries. The microstructure of the tube processed with rotating mandrel is dynamically recrystallized;
- The values of Vickers microhardness of the tube sample processed with stationary mandrel are increased by 40% in comparison with the initial bar. An increase of the values of the microhardness by about 200% was found in the tube processed with rotating mandrel. This increase is caused predominantly by the enormous grain refinement.

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