Effect of ECAP process on deformability, microstructure and conductivity of AA5083 under thermal effect

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Abstract. An alternate method of alloying is to use extreme plastic deformation on commercially available AA5083 to generate an ultrafine-grained microstructure. The objective of this approach is to improve mechanical characteristics without sacrificing corrosion resistance and biocompatibility. Anisotropy in mechanical properties is introduced by plastic deformation leading to the production of a distinct texture. This is a crucial concept to understand in order to build and model structural devices and components from a perspective-based approach. The ultrafine-grained structure of AA5083, which was obtained by equal channel angular pressing, is examined in this work. Ex-situ and indirect in-situ thermal studies are used to supplement this investigation while the material is heat treated at different annealing temperatures. The results show that the elastic properties undergo very small change during the annealing process, in contrast to other parameters such as thermal expansion, internal friction, or hardness. The strong relationship between the elastic anisotropy and texture highlights the importance and possibilities of using texture into the design and customization of mechanical characteristics. Pure deforms plastically in order to improve mechanical qualities while maintaining biocompatibility and corrosion resistance. Analysis of the materials elastic inhomogeneity and crunchiness in detail. In comparison to other characteristics such as thermal expansion or hardness, the results demonstrate that elastic properties barely marginally change during annealing. The microstructure fragmentation had no influence on the conductivity of the AA5083, which oscillated at 18 MS/m after the ECAP procedure. According to the findings, all deformed specimens strain hardening exponent and capacity were lower than they were in their as-received state.

1 Introduction

A5083 is an example of a product that benefits considerably from the mechanical qualities made possible by ECAP. Many aspects including temperature of processing geometry of die,
the processing path and the back pressure have a significant impact on the characteristics of the titanium generated by ECAP. The developed nanocrystalline aluminium microstructure displays a robust texture. To uphold its favorable attributes especially concerning grain growth and fault annealing, it is essential to meticulously manage its thermal stability. Due to the suppression of dynamic recovery during processing, using room-temperature or even cryogenic ECAP can produce titanium with a small grain size and a high defect density. Its consequences for traditional characterization techniques include TEM, testing for uniaxial tension, compression and XRD [1].

The objective of this research is to utilize the identical material and offer a more foundational analysis of the elastic properties and anisotropy of titanium. These properties are intimately linked to the pronounced crystallographic texture. Microstructure thermal consistency in terms of grain development and hardness. Indirect in-situ measurements and thermomechanical analysis as well as the proposal offer annealing procedure and texture-based design for AA5083 functional parts.

Electron backscatter diffraction is a crucial tool used in SEM for general microstructure analysis of ECAPed. From a polished region on the sample this approach delivers comprehensive info on point-to-point orientation. It is possible to characterize features in small volumes due to their local nature, including specific grains and descriptions of anisotropy in the substance. It gets rid of the length change impact brought on by the linear thermal expansion of a reference sample. Even for materials with a restricted supply, it is possible to determine the geographic distributions of modulus of elasticity. The samples overall resonant response is consistent with the stated elastic characteristics, while the local assessment of indentation elastic moduli stems from micrometer-sized indents [2].

This research will shed new light on the interaction of elastic characteristics. The arrangement of elastic anisotropy concerning the texture is directly connected to the microstructure in the AA5083 as ECAPed material and after further annealing at increasing temperatures. The effective elastic modulus measurements obtained through microhardness will be compared to the determined full set of titanium elastic constants. The generated data set can be utilised to design, develop and fabricate highly textured titanium parts.

The firm texture of hexagonal materials might be viewed as a positive and desired asset rather than a disadvantage [3]. Metals having high ductility, electrical conductivity, strong thermal conductivity and low chemical affinity to air are required for application in modern electronic techniques. These alloys electrical conductivity is typically less than 25%. Refinement of the microstructure is one method for increasing the electrical conductivity and mechanical qualities of copper alloys. Techniques for extreme deformity of plastic that can result in materials with ultra-fine grains with a size range of 102–1020 nm have seen a significant increase in interest. Severe Plastic Deformation procedures which allow for significant plastic deformation are one of the fundamental ways to create this kind of structure. The primary presumption of SPD processes is that coarse-grained materials microstructure would be fragmented.

The dislocations produced in the material are dispersed at arbitrary levels of deformity of plastic. Increase in the degree of plastic deformation these flaws clump [4]. Dislocation cells and shear bands are specialized systems that only arise after reaching a certain strain value. When the SPD process causes more cumulative plastic deformation. As the misorientation angle rises and the distance between the grain boundaries narrows. Polymerization is the process by which such dislocations in the material are grouped together. The quantity of distortion that must be accumulated substantially higher than with traditional plastic working methods. Shear forces acting in the vicinity of intersecting channels cause the material to extremely high stress. In contrast to their coarse-grained counterparts the majority of the materials created by SPD techniques have special features. This is because
Grains are pulverized to a nano- or sub-micrometric scale during the ECAP process. This causes the material's strength to significantly rise while frequently allowing for enough plasticity. Contrary to typical plastic deformation technologies, intensive plastic deformation techniques are employed to alter the initial material rather than to shape it. The ultra-fine or nanometric microstructure develops from the coarse-grained microstructure [5]. The SPD process conditions have a major impact on the final grain size and type of nanostructures that are created. The beginning material's initial grain size and phase composition. The mechanical and physical properties of materials treated with SPD are superior to those of typical coarse-grained materials. Mechanical characteristics have improved up to eight times and 32-52% for alloys. When the structure is refined using SPD techniques a particular minimum grain size can be attained. This value is based on the materials propensity to eliminate deformation-induced flaws as well as the speed of the recovery and recrystallization processes. Thermal conductivity does not significantly decrease in conjunction with significant strengthening brought on by severe plastic deformation. AA5083 are of great interest because of their distinct magnetic characteristics and the potential for employing them as catalysts. The enhancement of magnetoresistance is most greatly aided by nickel alloying components. AA5083 with a low concentration in particular exhibits a considerable improvement in magnetoresistance. Nickel significantly affects the mechanical and physical characteristics of AA5083. Thermal and electrical conductivity decreases with increasing nickel concentration, although tensile strength, yield stress, hot-temperature resistance, solidus and liquidus temperatures and corrosion resistance rise. Because of this, it merits investigation [6]. They established that utilizing the ECAP method, severe plastic deformation causes the microstructure to significantly fragment after 12 ECAP cycles with a comparatively large quantitative proportion of HABs (> 58%). Intense plastic deformation also facilitates an equal precipitation distribution in the Cu matrix and permits control over the magnetic characteristics of these alloys. After extensive plastic deformation by HPT the AA5083 minimum grain size was 120 nm. The AA5083 microstructure and mechanical characteristics as they changed over time after being distorted by ECAP at various temperatures. Compared to samples deformed at ambient temperature, the samples deformed at cryogenic temperatures showed improved toughness. Due to dislocation subdivision and twin fragmentation, ECAP deformation drastically reduced the alloy's grain size. According to one research, the average grain size may be reduced to roughly 0.36 μm after 8 passes [7]. The impact of the second phase particle distribution on the commercial AA5083's hardness and electrical conductivity, effects of ECAP at 300°C and room temperature. After ECAP processing, the area proportion of coarse particles reduced according to microstructural characterization. The mechanical characteristics of AA5083 after being processed by ECAP and grain refining. Grain refinement was significantly improved with more passes, going from 16 μm in the initial sample to 310 nm after 4. The outcomes demonstrated that the microstructure has also undergone a significant amount of grain refining. When compared to the initial microstructure, the microhardness, ultimate tensile strength, and wear resistance were all much higher following the ECAP [8]. The purpose of this work is to evaluate how the deformability microstructure, microhardness and conductivity of the AA5083 are affected by various types of SPD in the modified T-ECAP method. Severe plastic deformation techniques because they are an efficient approach to enhance the mechanical properties of metals and alloys as well as their microstructures. One of the most popular and useful SPD approaches is equal channel angular pressing. ECAP has demonstrated a tremendous capacity for creating ultrafine grain materials, which is why materials scientists find them to be particularly appealing.
repeatedly pushing a sample into a die made of two channels with equal cross sections that intersect at an angle $\Phi$, a strong plastic strain is applied to the sample. The enhancement from precipitation strengthening and work hardening holds significant importance in bolstering the strength of these specific alloys. Most high strength materials of ECAP experience a declining ability for strain hardening, which leads to their appearance under tensile deformation conditions. As a result their homogenous deformation range is smaller and their extension to failure is often close to 6-12%. In Al alloys the dynamic recovery process proceeds quickly leading in a dislocation density that is early steady-state and does not support strain hardening [9]. Precipitates and the solute distribution they are connected with are thought to be the cause of this complexity. Dislocation's mean free route is altered by precipitation. It has been thoroughly studied how the precipitates affect the strain hardening behaviour of AA5083 under ageing or deformation circumstances. Metallic materials under stress, such as one variable simplicity or models that provide a more precise description of the microstructure. It is essential to improving their strain hardening capacity and dislocations climb during deformation is connected to the accelerated strain hardening rates of ultra-fine grained materials. The formation of dislocation cells and the accumulated strain serve as the progenitors of sub-grains, which are mostly created by low angle grain borders. Some researches claim that as a result of ongoing dynamic recrystallization events from high geometrically necessary dislocation densities [10]. LAGB can change into high angle grain boundaries [10]. The average distance between soft and hard patches during deformation distribution is related to gradients of plastic accommodation, which are associated with GNDs. A large capacity for storing additional GNDs can boost the ability to harden under pressure and subsequently encourage a healthy balance of strength and ductility.

2 Experimental Material and Procedure

Thermomechanical analyzers with quartz thermal chambers were used to monitor thermal expansion. A temperature rate of 3.4°C/min was used for the testing in the range of 40°C to 610°C, and subsequently from 610°C to 220°C. Both the ECAPed and non-ECAPed versions of the AA5083 were cut out, utilizing the same heating/cooling setup and thermomechanical analyzer. Plate samples were used for the static deflection measurements while being bent in three directions mentioned Figure 1.

Fig. 1. Schematic diagram of ECAP Die Geometry with Channel and Corner Angles [11].
These measurements sought to determine temperature-dependent deflection under a constant load (210 mN). Each sample had supports that were 7.2 mm apart. In order to test the elasticity at room temperature and the pulse-echo method in combination with resonant ultrasonic spectroscopy was used. From the samples that were previously utilised for EBSD, four cuboid samples were taken. The low pressure RUS chamber was then filled with the measured samples and composition of material mentioned in table 1.

Table 1. Aluminium alloy 5083 chemical composition

<table>
<thead>
<tr>
<th>Element</th>
<th>Si</th>
<th>Fe</th>
<th>Cu</th>
<th>Mn</th>
<th>Mg</th>
<th>Zn</th>
<th>Ti</th>
<th>Cr</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>% Present</td>
<td>0.4</td>
<td>0</td>
<td>0.4</td>
<td>0</td>
<td>0.1</td>
<td>0</td>
<td>0.4</td>
<td>0</td>
<td>0.25</td>
</tr>
<tr>
<td></td>
<td>0</td>
<td>0</td>
<td>0.1</td>
<td>0.15</td>
<td>0.05</td>
<td>4.0</td>
<td>0.1</td>
<td>4.9</td>
<td>4.0</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
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<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Their elasticity was vibrating due to a pumping. The coefficient of temperature evolution was identified. Due to the fact that the lowest frequency resonant modes mostly correlate to the shear deformation. The maximum level of precision was used to measure this shear coefficient.

The studies focus on AA5083 which was produced in a lab setting using an induction furnace. The finished alloy was forged into bars using a compressor hammer with a beater weight of 225 tonnes after being put into graphite ingot moulds with a diameter of 35 mm. At 910°C supersaturation was carried out, then water quenching. On the basis of the analysis of the three-component phase equilibrium system of copper with nickel and cobalt. The supersaturation temperature of the tested AA5083 was chosen. In an electric chamber furnace with a controller that allows temperature registration with a precision of 0.1°C, heating to the supersaturation temperature was conducted.

Fig. 2. Different passing route through ECAP Die.

In Figure 2 the ECAP method was used to treat the samples of this size to extreme plastic deformation following supersaturation. The samples were preheated to the procedure temperature prior to deformation.

Table 2. For AA5083 samples the ECAP procedure parameters.

<table>
<thead>
<tr>
<th>Sample</th>
<th>Angle of channel rotation</th>
<th>Cycles of ECAP in number</th>
<th>Temperature, T°C</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>45</td>
<td>1</td>
<td>210</td>
</tr>
<tr>
<td>2</td>
<td>45</td>
<td>1</td>
<td>110</td>
</tr>
<tr>
<td>3</td>
<td>45</td>
<td>1</td>
<td>100</td>
</tr>
<tr>
<td>4</td>
<td>45</td>
<td>1</td>
<td>010</td>
</tr>
</tbody>
</table>

Fig. 2. Different passing route through ECAP Die.
Different sample procedure parameters mentioned in table 2 utilising a die with a 90° channel intersection angle and a hydraulic press with a 1680 kN maximum permissible pressure and intense plastic deformation was carried out. The ability to obtain back pressure by die modification results in an increase in the squeezing force. This adjusted matrix deforms in a way that results in an equivalent deformation of $\varepsilon = 1.28$ each pass. Additionally, the use of a redesigned exit channel in the ECAP die results in turbulent material flow and makes it possible for more plastic strain to accumulate in a single process cycle [18].

By using method of route A the sample was pushed without rotation and the deform rate of sample of 2.2 mm/s. The heating mechanism in the hydraulic press that was employed in the ECAP procedure allowed the tool frame to heat the samples. The samples underwent annealing for 15 minutes. Analysis methods included backscattered electrons and secondary electrons utilising the EBSD method. On particular surfaces and along the scanning line and the misorientation angle of individual locations was determined. After the scanning, which was carried out with a step size of 0.2 m [19], a threshold angle of 30° was established in order to minimize the Orientation noise effects and the restrictions with regard to the angular resolution of the EBSD equipment. Grain sizes below a cutoff of four pixels were originally removed. The misorientation threshold angle of 5° was used to examine the average grain size. If there is more than 50° of misorientation between two nearby scan sites, they are thought to belong to two separate grains. Before and after the ECAP technique the cross-section of the specimen, microhardness was measured using the Vickers method with a stress of 9.81 N. The loading force lasted for 18 seconds. Eddy currents are used to measure electrical conductivity. By placing the measuring probe on the material being tested and measurement was completed [20].

The present AA5083 alloy was delivered in the form of 10 mm diameter extruded rods. Cylinders with an approximate length of 68 mm were formed from rods. At room temperature, deformation was carried out using a 0.04 m/s pressing speed. Lubricant was molybdenum disulfide (MoS$_2$).

3 Experimental Results

3.1 Stability of AA5083 prepared by room temperature

3.1.1 Microstructure texture and hardness during thermal stability

Large unrecrystallized patches are visible in both the ECAPed sample at room temperature as-processed and the sample after 2.2 hours of annealing at 457°C. Smaller equiaxed grains are present with them, The lack of mechanical twinning in the as-processed samples microstructure is probably pertaining to the ultra-fine grain size and the manufacturing process as twinning is suppressed by the small particle size.

Equiaxed grains with an typical grain diameter of 6.2 μm after the annealing at 610°C indicate a fully recrystallized microstructure. Further grain development is encouraged by the additional annealing temperature rise.

The EBSD measurements for this alloy at 910°C show no directly observed phase transition that is volumetrically incomplete. In contrast, the iron content in the work was almost at its maximum [22]. At areas with localised Fe concentration in the...
3.1.2 Anisotropy of elasticity

After annealing the specimens were annealed at various temperatures to observe velocity values varied by 1.2% or less in the right directions. Using a laser-based RUS, the annealed samples (heated to 457°C, 610°C, or 910°C) showed, there are over a hundred resonant frequencies associated with various modes of vibration. It possible to carry out the complete inversion of elastic coefficients [25]. Initially, it was believed that the samples had general (triclinic) symmetry. The examination of experimental modal shapes served as the foundation for the correct alignment of the Orientation of the coordinate system in relation to the sample coordinate system ED, ID, and TD. Due to the presence of multiple minima in the RUS inversion technique for this scenario, each corresponding to different sample corners as the center of the coordinate system, aligning these corners is essential to achieve overall triclinic symmetry. It was discovered that orthorhombic symmetry can reasonably mimic the measured triclinic elasticity (within experimental error). A weighted minimization was used to estimate the principal axis' orientation and the positions of nine independent elastic coefficients.

The combinations of prior measurements' accuracy from RUS measurements for triclinic elastic constants [26]. The elastic coefficients determined by RUS for the annealed samples reveal that, considering experimental uncertainties, the major symmetry directions in all three samples are almost the same as determined by RUS inversion for orthorhombic symmetry. However, the coordination axes ID, ED, and TD of the samples differ. Experimental results, the y axis, this is associated with the elastic coefficient governing longitudinal behavior of C11, all die close to ID [27]. The shearing that occurs perpendicular to the x axis, or in the y-z plane, has the softest shearing coefficient, which is C44. The orthorhombic symmetry's axes are marked to help you see the elastic anisotropy that RUS determined. As can be observed, all three measured samples had comparable Young's modulus distributions in terms of both directional dependency and the calculated modulus of elasticity. As a result, the elastic characteristics of the ECAPed AA5083 are not significantly affected by annealing at different temperatures, and the elastic anisotropy caused by ECAP is maintained throughout annealing. It is significant to note that no prior assumptions about the microstructure of the small samples at longitudinal wave velocities and density were made [29].
3.1.3 Effect of annealing

Table 3. RUS at 25°C determined the orthorhombic elastic coefficients of annealed materials. The shear elastic coefficients (C_{44}, C_{55}, C_{66}) standard deviations are within 0.25 GPa and 1.2 GPa for all other coefficients (C_{11}, C_{22}, C_{33}, C_{12}, C_{13}).

<table>
<thead>
<tr>
<th>C_{ij} [GPa]</th>
<th>457°C</th>
</tr>
</thead>
<tbody>
<tr>
<td>C_{11}</td>
<td>172.9</td>
</tr>
<tr>
<td>C_{22}</td>
<td>160.6</td>
</tr>
<tr>
<td>C_{33}</td>
<td>164.3</td>
</tr>
<tr>
<td>C_{44}</td>
<td>41.72</td>
</tr>
<tr>
<td>C_{55}</td>
<td>49.76</td>
</tr>
<tr>
<td>C_{66}</td>
<td>46.79</td>
</tr>
<tr>
<td>C_{12}</td>
<td>74.5</td>
</tr>
<tr>
<td>C_{13}</td>
<td>72.8</td>
</tr>
<tr>
<td>C_{23}</td>
<td>82.1</td>
</tr>
</tbody>
</table>

As is typical for metals, the C_{66} coefficient gradually drops as the temperature rises, neither its value nor the slope of the C_{66} dependency vs. temperature changing abruptly. The C_{66} logarithmically increasing with time at the dwell temperature of 457°C points to a diffusion-driven mechanism occurring at this temperature. When compared to the unannealed condition, the C_{66} progressively rises after cooling, peaking at room temperature after the first cycle. Processes involved in recovery are probably to blame. The temperature progression during the second cycle of annealing at the dwell temperature of 610°C is completely reversible and does not alter considerably. To exhibiting a logarithmic time dependency resembling the growth of the shear coefficient during the dwell duration at 457°C. As the imperfections gradually disappear, they have a significant impact on the measured internal friction in the ECAPed sample as well as its temperature vs. time dependence. Subsequent to the first annealing process carried out at low temperatures, internal friction shows that the ECAPed AA5083 sample may include some omega phase. In the unannealed ECAPed samples, the omega particles caused by stress were discovered using TEM. Given the significant shear deformation and comparatively low oxygen concentration, it was plausible for the omega phase to form during ECAP. Although the existence of the omega phase may have also affected the internal friction during the initial heating run, the substantial defect relaxation was primarily responsible for the high internal friction values that were observed. Internal friction associated with defect relaxation significantly decreased over the 2.2-hour heating and annealing processes at 457°C, during the first cycle's subsequent cooling phase. At room temperature, the thermal omega phase was relatively high; however, as the temperature increased and the omega particles disintegrated, the athermal omega phase rapidly reduced. Due to the formation of nanocrystalline precipitates and their variable stability, the amount of the omega phase is rather tiny. Internal friction significantly increases on heating above 460°C, or at the temperatures that were not achieved during the first cycle, and reaches a maximum at 500°C. The highest achievable damping capacity occurs at 540°C. It’s possible for the location of the peak to change at various temperatures due to the difference in frequency between the Resonant Ultrasound Spectroscopy measurements around 1.2 MHz and the internal friction measurements 1.2 Hz differing by six orders of magnitude.
friction values decrease dramatically after cooling, gradually falling to $2 \times 10^{-5}$. Omega phase has probably been suppressed to the point that it no longer has an impact on internal friction [33]. The observed slopes in the cooling curves correspond to activation energy values of approximately 0.8 eV for both recorded cycles. This indicates that internal friction in both measured cycles is influenced by thermally induced relaxation processes occurring at temperatures above $220^\circ C$. Notably, this value is on the same scale as the findings from Resonant Ultrasound Spectroscopy in composites composed of AA5083.

Fig. 3. Internal friction and shear coefficient temperature development [32].

The calculated value of 0.75 eV matched the dislocation slip activation energy. UFG AA5083 showed a comparable rise in damping capacity from $300^\circ C$. Therefore, it appears that the ECAPed AA5083 is undergoing a number of primary relaxation processes, taking into account the relaxation damping brought on by microstructural defects, the omega particle-generated low-temperature damping. When temperatures are greater and dislocations relax. Internal friction temperature dependency is measured, and the tested sample's microstructure progressively changes with time as well as with temperature [34]. Any coincidences during the first heating cycle may simply be coincidences because there are other activities that rise at higher temperatures.

3.1.4 Thermomechanical analysis

Since non-ECAP reference samples do not have any annealing defects, as might be expected, the greatest relative length difference is lower for this sample couple than for any other couple. In Figure 4, at least two separate annealing stages could be seen. First, when the temperature was lower, is about annealing vacancies and dislocations, the relaxation of grain boundaries, whereas grain increase can account for the subsequent significant change [35]. The static deflection at the site of application as a function of temperature after adjusting for the influence of the steel specimen holder for AA5083 samples that were ECAPed and not annealed ECAPed. The creep resistance of AA5083 that had been ECAP treated at RT was better than that of coarse-grained AA5083. No research has looked at ECAPed AA5083's creep resistance at temperatures above room temperature. Due to the inconsistent measurement temperature and the brief dwell time, the findings presented in Figure 4 may not accurately reflect a genuine creep deformation experiment [36]. This suggests that, for
temperatures exceeding 1200°C, the creep resistance of AA5083 samples subjected to ECAP may not be as strong as that of the non-ECAPed AA5083 specimen. The resistance to deflection is considerably improved for the annealed sample as a result of point defect annihilation prior to the test. Due of the great fault reduction and still unrecrystallized grains, 2.2 hours of heat treatment at 4570°C was specifically chosen. This means that the increased mechanical characteristics brought about by the annealed sample still undergoes ECAP processing while having increased deflection resistance [38].

4 Discussion

4.1 Analysis of thermal Expansion

4.1.1 Importance of Annealing and Thermal Stability

Fig. 4. Different relative length disparities affect temperature in both the heating and cooling processes [37].

![Graph](image_url)
4.1.2 Nanoindentation to Measure the Elastic Modulus in AA5083

Fig. 5. Schematic diagram relationship between number of friction and Misorientation angle [43].

Without making any prior assumptions about the symmetry of the material using measurements of the texture, the inverse estimation of elastic properties from the resonant spectra was performed. The simulation for the sample material AA5083 made use of the single-crystal elastic constants stated from. The simulation that the two poles would be distributed, with their centers skewed 65° of the TD direction.

4.1.2 Nanoindentation to Measure the Elastic Modulus in AA5083

All indentation studies supported the material under evaluation anisotropy in elastic-plastic properties of microhardness and indentation modulus values as well as the asymmetries of spherical indents. The nanoindentation findings yield a more narrow range when examining the difference in indentation modulus between the ED and ID orientations. The data of same pattern and are within the RUS limits. The modulus values from various directions are combined into an average in the triaxial stress state induced by the indentation process shown in Figure 6. The fact that the deformations in RUS are quite minor in comparison to the nanoindentation is more efficient [44]. The assumption of equal Poisson’s ratios in both investigated directions represents the second reason for the slight deviation from uniaxial values in nanoindentation. Nevertheless, the results of indentation suggest that it can still be employed to determine the elastic modulus and its anisotropy in ECAPed AA5083, despite these simplifications. However, when determining the indentation modulus of textured titanium, it is essential to consider multiple orientations. It is important to look into the possibilities of gaining a broad understanding of titanium’s anisotropy utilising a spherical indenter that produces a robust texture. By simply averaging the deformations of grains with varied orientations using weights provided by the texture, the sample deformation was achieved. The comparison of estimated sample averages from grain dilatations with experimental dilatations [46]. The observed good agreement between experimental and computed values lends credence to the notion that the sample texture and the anisotropy of dilatation are related.
4.2 Effect of ECAP process

4.2.1 Effect on Deformability

The AA5083 samples’ shape has changed as a result of the ECAP procedure; they are now noticeably twisted and have elongated ends. Sample No. 1 is to be compressed in the ECAP at a temperature of 22°C in accordance with the approved test plan. But during the test, the press’s maximum permissible pressure of 1020 MPa was attained. It was necessary to heat the sample to 200°C in order to complete one ECAP cycle since the sample’s limited plasticity at 20°C would have otherwise damaged the ECAP matrix. The ECAP matrix contained the heated sample at the time of heating. Figure 7 depicts the forces measured during sample number 1’s deformation at 25°C and 220°C. The maximum pressure dropped to almost 720 MPa after heating Figure 8. This made it feasible to evaluate how the temperature change impacts the AA5083 plasticity [47]. Due to the force used to squeeze the sample, which reached a stress of about, full extrusion of the AA5083 sample was not possible at room temperature. It served as a model for additional variations. Sample No. 2 was forced through the passage after being heated to 120°C. During the process, stress readings of roughly 820 MPa were achieved. Prior to the procedure, samples nos. 3 and 4 were heated sequentially to 220°C and 270°C. As seen in Figures 9 and 10, there was a clear correlation between the rise in process temperature and a fall in the stress value. 420 MPa, or 53% of the stress measured at 120°C, is the stress value of the sample that was deformed at 270°C.
4.2.2 Effect of Thermal Conductivity

The AA5083 exhibits the microstructural features of the deformed material following the ECAP procedure. When the shear bands generated, where the intense plastic deformation is located, were examined, it was found that the grains gradually fragmented. The possibility of dynamic recrystallization cannot be ruled out in the vicinity of shear bands. Additionally, the grains' lengthening due to the shear pressures acting on the sample in the vicinity of the intersection of the matrix channels can be seen. It was discovered that the degree of microstructure refinement depends on the temperature at which plastic deformation was conducted. The sample distorted at 250°C has an average grain size of 24 μm, whereas the average grain sizes for the samples deformed at temperature 100°C and 200°C are 19 μm and 21 μm, respectively.
Fig. 11. Microstructure of the AA5083 after the ECAP process at the temperature [53].

IPF-Z maps of the samples following supersaturation and plastic deformation are shown in Figure 11. Large equiaxial grains predominate in the structure after supersaturation, according to the examination of the produced EBSD maps. Annealing twins may be seen inside these grains, which have irregularly shaped borders. The GBs misorientation angle distribution histogram, which shows a significant proportion of high-angle boundaries. After supersaturation, the sample's average grain size is 7.2 μm (Table 3). The strong impact of plastic deformity on the microstructure of the AA5083 is confirmed by the crystallographic orientation measurements carried out using the backscattered electron diffraction method. The produced images demonstrate the distinction of grain size, demonstrating the material's structural variability [54]. According to the maps of grain boundary disorientation distribution, the microstructure is dominated by elongated grains after the ECAP process, in which deformation bands and low-energy configurations of low-angle borders are apparent.

The degree of microstructure fragmentation achieved is not considerably impacted by an increase in deformation temperature, the typical grain size ranges from 1.6 μm and 1.8 μm. On the other hand, plastic deformation increases the proportion of low-angle borders, which is supported by the quantitative information shown in Table 4 and the histograms of the distribution of grain boundary misorientation. They demonstrate that at 1000°C, nearly 75% of borders have low angles. As the temperature of plastic deformation grows from 120 to 220°C, the proportion of low-angle grain boundaries increases from 75% to 79%. When the temperature is raised further to 260°C, the fraction of low-angle grain boundaries marginally declines to 76%. The fraction of low-angle borders increased together with the temperature of the ECAP process. This can be explained by the fact that the temperature rises with the deformation, the ECAP process increases the recovery rate, which extinguishes dislocations. At the higher process temperature, the likelihood of dislocation absorption by the grain boundaries is also reduced [55]. The transition of the microstructure to an ultrafine-grained microstructure with a considerable fraction of high-angle barriers is slower at the higher temperature of the ECAP process. It is supported by data from the literature and the results of recent research.

Table 4. Summary of EBSD results [56]

<table>
<thead>
<tr>
<th>Grain size (diameter) [μm]</th>
<th>Low Angle Grain Boundaries, %Fraction</th>
<th>High Angle Grain Boundaries, %Fraction</th>
<th>Average Misorientation Angle, θAVA [0]</th>
</tr>
</thead>
<tbody>
<tr>
<td>Solution treated</td>
<td>7.5</td>
<td>74.0</td>
<td>50.0</td>
</tr>
<tr>
<td>ECAP at 1200°C</td>
<td>1.82</td>
<td>74.3</td>
<td>30.8</td>
</tr>
<tr>
<td>ECAP at 2200°C</td>
<td>1.86</td>
<td>79.5</td>
<td>24.8</td>
</tr>
<tr>
<td>ECAP at 2500°C</td>
<td>1.81</td>
<td>74.6</td>
<td>29.5</td>
</tr>
</tbody>
</table>
4.2.3 Microstructural Effect on Sample

The results of the microhardness tests shown in Figure 11 unequivocally demonstrate that modifications to the microstructure and the resulting refinement of grain improved the mechanical properties. According to the experimental results, the sample’s highest hardness is attained following the ECAP cycle at a temperature of 1200°C - 137 HV. In comparison to the undeformed sample, there is a 210% increase in hardness. For samples deformed at temperatures of 220 and 270°C, respectively, a slight drop in hardness to 128 and 122 HV is seen as the deformation temperature rises. The hardness distribution at the cross-sectional level showed significant variation regardless of the temperature used in the ECAP procedure. Inhomogeneous deformation and the development of shear bands in the material’s microstructure are to blame for the comparatively significant hardness spread. Similar results in AA5083, where a similar hardness distribution was seen after one ECAP cycle, show the presence of heterogeneity after a single pass [57].

Distribution of hardness along the samples’ cross-sectional planes. The simultaneous action of a number of strengthening mechanisms dislocation strengthening and strengthening from the substructure being the most significant is what leads to the observed increase in hardness. Similar grain sizes (a similar degree of microstructure refinement) were attained at the deformation temperature, going from about 9 m in the supersaturated condition to about 1.8 m after one cycle of the ECAP process, which produced an increase in hardness of around 210% for the sample deformed at 1200°C. The equilibrium between the annihilation of dislocations and formation in the ECAP process at low temperatures can account for such a considerable rise in hardness after the first pass. The degree of strain hardening, which is connected to the intense dislocation annihilation by thermally stimulated processes, is decreased by raising the temperature of the ECAP process.

The highest GNDs density on the maps is represented by the red “hot spots.” After one cycle of the ECAP procedure, similar values of a rise in microhardness were recorded for both alloys and pure copper. The results can be compared to data from the literature for pure copper conductivity (62.8 MS/m), which is four times greater than the tested AA5083 samples. The information in Table 3 leads to the conclusion that the conductivity of the tested alloy was not significantly affected by the plastic deformation that caused the microstructure to break. Copper’s mechanical characteristics are directly connected to its conductivity [58]. As hardness reduces, conductivity rises. For the alloy that was pressed at a lower temperature, more strengthening was noted. Because the processes of microstructure renewal are thermally stimulated, the increase in temperature was accompanied by a drop in hardness and a slight increase in conductivity. The sample distorted at the highest temperature showed the greatest increase in conductivity when compared to the sample following supersaturation.

Since beryllium is poisonous, it was left out of the smelting process.

4.3 Strain Hardening Behavior

4.3.1 Evolution of Microstructure
The microstructure of the first six ECAP passes can be related to the formation of new grain borders, which separate coarse grains into cell blocks, during the early stages of SPD. The majority of LAGBs are converted into HAGBs with deformations greater than six ECAP passes. Grain distortions contribute to an increase in the proportion of HAGBs with more substantial deformations in the analyzed deformation situations, where the grains evolve during deformation together with grain boundaries with HAGB features. Smaller grains are polished with time [59]. The evolution of grain size and grain borders can be better understood using various grain internal statistical approaches.

Due to the fact that damaged grains' dislocations result in cellular structures that cause local misorientations inside the grains of variable degrees, this observation was made. Recrystallized grains, on the other hand, have substantially lower dislocation densities and hence have less interior grain misorientations. The rise in deformation indicates a shift in GND values. The material's GND values as received range from 0.5 to 20. It shows a clear shift in the color of the grains between the first and fourth passes. The lowest and upper boundaries of the material distribution as received, where no discernible grain content is found, were taken into consideration when defining the threshold values in the current work to distinguish recrystallized from deformed grains. It is evident from the GND distributions. The distorted grains are represented by GND values greater than 2°. GND values under 0.8°, on the other hand, accurately depict the recrystallized state. A appears to depict various regions based on the GND values. With increasing ECAP passes, there is an increasing population of grains around GND values around 0.20. These grains match the tiny blue grains in the original grains following the SPD process in the shear direction. This behavior provides unmistakable proof that CDRX causes tiny grain microstructure renewal. A significant portion of the grains are also seen to be bent and substructured, particularly for materials processed by ECAP. These findings point to the alloy's potential for additional grain refinement, either through additional ECAP cycles or through the application of a different deformation method following ECAP.

The progression of microhardness with respect to ECAP passes. Up until the second ECAP pass, a sharp hardness increase can be seen at the start of the ECAP processing, which results in steady-state behavior. The equilibrium between grain refinement and accumulation annihilating dislocations during deformation causes the hardness saturation. Mg solute atoms' locked dislocations typically cause AA5083 to alter noticeably. A progressive increase in tensile passes is seen during succeeding ECAP passes. The substance then reaches a state of saturation. The elongation to failure in terms of material strength fell from 28% to roughly 18% following pass [60]. The GND analysis has previously shown that the initial elongation related with strain hardening capacity. The highest GND densities after one ECAP pass are visible inside the sheared grains, more precisely where sub-grains are present. This behavior is consistent with the larger percentage of LAGBs that are in this stage of deformation. Increasing strain causes GNDs to slightly rise along the same route as the material from a single ECAP pass, albeit with smaller grains. With high ECAP passes, the GNDs distribution curves shift in favor of greater values. The average GNDs density, however, remains rather constant and unaffected by the ECAP passes even after four passes. which aids in balancing the curvature brought on by the SPD. Given that GNDs are responsible for the subdivision of grains, the development of GNDs in ultrafine grain material means that there is a strong likelihood of further reduction in grain size.
4.3.2 Strain Hardening Behavior

A normalized parameter that measures the hardening capacity of ECAPed specimens at various passes. One can observe that the 0.526 (original specimen) went from 0.032 after one ECAP pass to 0.128 after twelve ECAP passes thanks to a progressive growth. The variation is consistent with the dislocation densities variation with the ECAP passes and the grain refinement variation. The more the capacity for strain hardening is lowered as a result of increasing dislocations.

To devoted particular attention to the behaviour of the ductility evolution following ECAP processing. The data between specimens with and without ECAPed. Although the strain hardening exponent values decline and fluctuate between 0.016 and 0.032 during ECAP processing up to the sixth ECAP pass. The \( n \) values thereafter significantly increase and range from 0.022 to 0.036 with more than five ECAP passes. The development of the microstructural properties can be used to explain the ongoing variations in the exponent \( n \) following ECAP processing. The strain hardening capability is influenced by the migration of dislocations within the grains. With more than 5 passes, the scenario shifts because the LAGB is transformed into HAGB is defined by a linear decline with flow stress.

The distorted substance quickly becomes harder. However, these materials are unable to maintain the hardening at high strains because of the limited space for the dislocation motion. Deformed causes early necking formation. However, values rise following the deformation process. The significant dynamic recovery rates attained during these stages of deformation account for the elongation improvement after four ECAP runs. The KME model produced the genuine stress-strain curve experimental and anticipated values for each of the examined situations.

When dislocations are poorly organized and do not create stable cell configurations, they are much easier to delete, generating an unpredictable dislocation structure [61]. The increasing levels that trigger recovery and cause material softening are linked to the ultrafine-grained alloy's early instability.

5 Summary

Anisotropy and thermal stability in UFG AA5083 produced by room temperature ECAP are characterized in depth in this work. It has been demonstrated that there is a strong correlation between the material's texture and the anisotropy of its elastic characteristics. Through the Resonant Ultrasound Spectroscopy technique, it is possible to experimentally ascertain the complete set of elastic constants or track the temperature-dependent changes in elastic moduli and internal friction. The detected elastic anisotropy is a consequence of the material's microstructure. Thermal Mechanical Analysis experiments, coupled with corresponding simulations have indicated that the primary factor responsible for the anisotropy in thermal expansion in UFG AA5083 is the material's texture. This anisotropy is mainly attributed to the texture itself, with only a minimal degree of expansion occurring in the basal planes during the heating process, and recovery processes do not significantly contribute to it.

Due to the significant extrusion force needed and the difficulty in obtaining the approximate 1020 MPa tension, AA5083 is difficult to deform at ambient temperature. The pressing force was clearly reduced when the process temperature rose from 120 to 270°C and the stress level dropped from 820 to 420 MPa. For the alloy that was compressed at a lower temperature, a stronger hardening is seen. The thermally activated process of microstructure recovery can be used to explain why the hardness decreased as the temperature rose. The measured copper alloy's conductivity, which fluctuated at a value of 18 MS/m, was unaffected greatly by the temperature of the ECAP process, which operated between 22 and 270°C. The transition from LAGBs to HAGBs which is fundamentally analogous to a continuous dynamic
Both yield and tensile strength increased dramatically as the number of ECAP passes increased, reaching roughly 354 MPa and 365 MPa after 5 passes, respectively. Recovery and ongoing recrystallization are the causes of the increase in ductility. The stored energy content has diminished due to the emergence of new, smaller grains and the subsequent energy recovery process. The slight uptick in stored energy observed at elevated deformation levels is a consequence of the generation of additional dislocations.

References


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