

Article

Effects of ECAP Processing Temperature and Number of Passes on the Mechanical Behavior of CuAg0.1 Alloy

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Abstract

This study investigates the effect of equal channel angular pressing (ECAP) temperature and pass number on the microstructural evolution and hardness response of a CuAg0.1 alloy. ECAP was performed by using route B_c at room temperature and 100 °C for up to six passes. X-ray diffraction results showed that no additional crystalline phase was detected after ECAP, while peak broadening and relative intensity changes indicated deformation-induced lattice distortion, defect accumulation, and a tendency toward preferred orientation. Optical microscopy revealed progressive microstructural subdivision with increasing pass number, particularly under room-temperature processing. Vickers microhardness increased sharply after the first ECAP pass and approached a saturation-like regime after approximately four passes. Room-temperature ECAP produced higher hardness and hardness-derived estimated strength values than processing at 100 °C, indicating more effective defect storage and suppressed recovery. In contrast, the lower hardening response at 100 °C is attributed mainly to thermally assisted recovery and dislocation rearrangement. The section-dependent hardness response suggests ECAP-induced anisotropy was associated with the imposed shear deformation and preferred crystallographic orientation. Within the investigated processing, four-pass room-temperature ECAP provided the most effective hardness-based strengthening response.

Keywords: CuAg0.1 alloy; equal channel angular pressing (ECAP); severe plastic deformation (SPD); grain refinement; sustainable materials



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1. Introduction

Severe plastic deformation (SPD) techniques are widely used to produce bulk metallic materials with refined microstructures and improved mechanical properties without substantially changing the overall geometry of the workpiece [1–3]. Among these techniques, equal channel angular pressing (ECAP) is one of the most effective methods because it introduces high accumulated shear strain through repeated pressing in an angular die while preserving the cross-sectional dimensions of the sample [4–6]. The microstructural response during ECAP is governed by several processing parameters, including die geometry, processing route, frictional conditions, pressing temperature, and number of passes [7–10].

These parameters control strain distribution, dislocation accumulation, substructure formation, grain-boundary evolution, and the resulting mechanical response of the processed material [11].

Copper and copper-based alloys are of continuing technological importance because of their high electrical and thermal conductivity, good ductility, corrosion resistance, and recyclability [12,13]. These characteristics make them essential materials for power electronics, electrical connectors, transportation systems, conductive components, and energy-related applications [6,14]. However, the relatively low strength of high-conductivity copper alloys limits their use in applications requiring mechanical durability. Therefore, strengthening strategies that improve hardness and strength are of considerable interest. In this context, SPD processing may offer a route because it can enhance mechanical performance mainly through deformation-induced defect accumulation and microstructural refinement rather than through high alloying additions or extensive heat treatment. The increase in strength of CuAg0.1 alloys containing only a small amount of Ag can be achieved primarily through dislocation accumulation, substructural refinement, and grain boundary formation caused by severe deformation. Despite this technological relevance, systematic studies on ECAP processing of commercial CuAg0.1 alloy remain limited compared with studies on pure Cu, Al alloys, Mg alloys, and higher-Ag Cu–Ag systems [15,16].

Processing temperature is another critical factor in ECAP because it directly affects the competition between deformation-induced hardening and recovery-assisted softening. At room temperature, recovery is relatively limited, and a larger portion of the imposed shear strain can be stored as dislocations, lattice distortion, residual stress, and deformation-induced substructures. Conversely, even a moderate increase in processing temperature may activate partial dislocation rearrangement, residual-stress relaxation, limited subgrain coalescence, and recovery-related softening. Most ECAP studies focus either on room-temperature deformation or on higher-temperature processing, where recovery and recrystallization effects become more pronounced. The low warm-processing regime around 100 °C has received less attention for dilute Cu–Ag alloys, although it may provide useful insight into the early influence of thermal activation on defect storage and microstructural stability [17,18].

Another important issue in ECAP-processed materials is anisotropy. Prior works report texture-driven directional differences in hardness and mechanical response in Mg- and Cu-based materials [15,19–21]. The intense simple shear imposed at the die intersection and the sample rotation sequence used in route B_c can promote crystallographic reorientation, deformation texture, and direction-dependent mechanical response. Such anisotropy is particularly important for conductive structural components where the loading direction and service orientation may influence performance. Although texture-related directional differences have been reported in several ECAP-processed FCC and HCP materials, the extent to which section-dependent hardness and XRD response develop in CuAg0.1 alloy has not been sufficiently clarified.

Accordingly, the present study investigates the effects of ECAP processing temperature and pass number on the microstructural evolution and hardness response of a commercial CuAg0.1 alloy. ECAP was performed using route B_c at room temperature and 100 °C for up to six passes. The study evaluates the XRD response, optical microstructure, Vickers microhardness, and hardness-derived estimated strength values on ED-parallel and ED-perpendicular sections. The novelty of this work lies in the systematic comparison of room-temperature and low warm-temperature ECAP for a CuAg0.1 alloy, the assessment of pass-dependent hardening and saturation behavior, and the evaluation of section-dependent anisotropy under identical processing conditions. The study contributes to a clearer understanding of the processing–microstructure–hardness relationship

in low-Ag, Cu–Ag systems and supports the potential use of ECAP as a resource-efficient strengthening route for conductive copper alloys.

2. Materials and Methods

2.1. ECAP Samples and Experimental Setup

The material used in this study was a commercial CuAg0.1 alloy supplied by Er-Bakır Electrolytic Copper Products Co. (Denizli, Türkiye) in the form of vertically continuous-cast wire rods. Cylindrical samples with a diameter of 12 mm and a length of 35 mm were machined from the as-received rods for ECAP processing. The initial mechanical and electrical properties of the supplied CuAg0.1 alloy are given in Table 1.

Table 1. The properties of CuAg0.1 alloy.

Property	Value	Unit
Single Wire Elongation	47.4	%
Wire Strength Yield	165.9	N/mm ²
Wire Resistance	0.0172	Ohm.mm ² /m
Electrical Conductivity	58.16	m/Ohm.mm ²

The ECAP experiments were carried out using a 120-ton hydraulic press (TurkYus, Bursa, Türkiye) equipped with a heated die block. The die was manufactured from AISI H13 tool steel (DIN 1.2344) to provide sufficient mechanical strength and dimensional stability during repeated pressing. The ECAP die consisted of two intersecting circular channels with a diameter of 12 mm, corresponding to the sample diameter. The internal channel angle was $\Phi = 120^\circ$, and the outer curvature angle was $\Psi = 20^\circ$. The inner and outer corner radii at the channel intersection were $r = 2$ mm and $R = 4$ mm, respectively. A plasma-nitrided steel punch was used to minimize wear under repeated loading. The ECAP die block used in the experiments is shown in Figure 1.

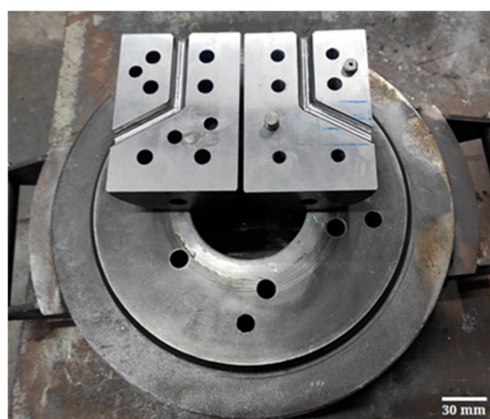


Figure 1. Die block used in the ECAP process (İzmir Katip Çelebi University Mechanical Engineering Laboratory).

Based on the ECAP die geometry, the equivalent strain imposed per pass was calculated using the conventional ECAP strain equation [22]

$$\varepsilon_N = \frac{N}{\sqrt{3}} \left[2 \cot \left(\frac{\Phi + \Psi}{2} \right) + \Psi \csc \left(\frac{\Phi + \Psi}{2} \right) \right] \quad (1)$$

where N is the number of ECAP passes, Φ is the internal channel angle, and Ψ is the outer curvature angle. For the present die geometry, the equivalent strain introduced in a single

pass was approximately 0.63. Therefore, the accumulated equivalent strain after N passes was estimated as $\epsilon_N \approx 0.63 N$. This indicates that the total imposed plastic strain increased progressively with increasing ECAP pass number.

2.2. ECAP Processing and Sample Sectioning

Prior to ECAP processing, the die channels and sample surfaces were lubricated with a high-temperature graphene-containing lubricant to reduce friction, minimize wear, and improve the reprocessability of the material pass during pressing. Cylindrical CuAg0.1 samples with a diameter of 12 mm and a length of 35 mm were processed under two deformation conditions: room temperature and a nominal die-set temperature of 100 °C. The room-temperature condition was performed without external heating. For the 100 °C condition, the die block was heated to the target temperature using the integrated heating system before pressing, and the sample was inserted into the preheated die channel prior to deformation.

ECAP was performed under an applied pressure of 60 bar and at a constant pressing speed of 0.2 mm/s for all processing conditions. The samples were subjected to up to six ECAP passes. Route B_c considered the optimal route in the literature was selected throughout the study; accordingly, each sample was rotated by 90° in the same direction around its longitudinal axis after each pass before reinsertion into the die [23,24]. This method ensures increased homogeneity of multi-directional shear and strain in FCC metallic materials during the multi-stage ECAP process. The route B_c rotation sequence is schematically shown in Figure 2, and the CuAg0.1 samples obtained after different ECAP passes are shown in Figure 3.

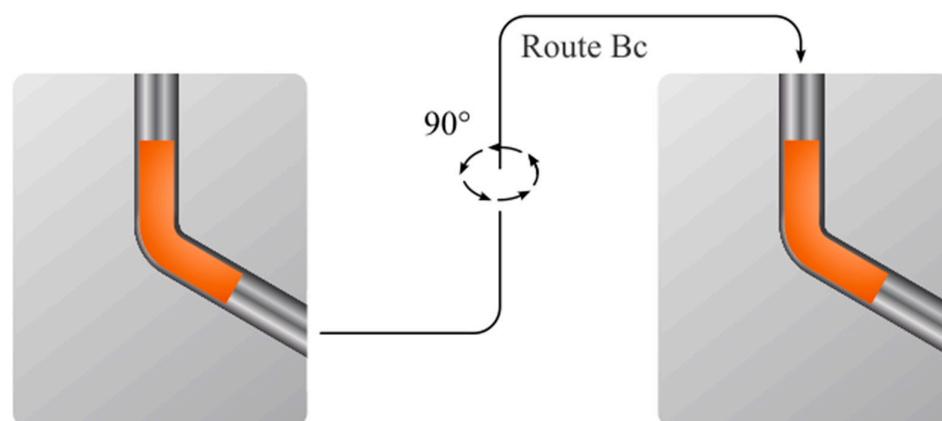


Figure 2. B_c route diagram of ECAP process.

The unpassed sample in Figure 3a retains its original cylindrical form, reflecting the undeformed condition of the alloy. After ECAP processing, the samples in Figure 3b,c display noticeable surface distortions and visible shear marks produced by the high strain imposed during pressing. These photographic images provide a clear visual comparison of how processing affects the overall deformation of the samples.

The 100 °C nominal die-set temperature condition was selected as a low warm-ECAP temperature rather than as a temperature intended to induce extensive recrystallization. The purpose of this condition was to examine whether limited thermal activation during ECAP affects the balance between deformation-induced defect accumulation and recovery-assisted softening in the CuAg0.1 alloy. The same heating procedure was applied consistently for all 100 °C ECAP passes to maintain comparable processing conditions.

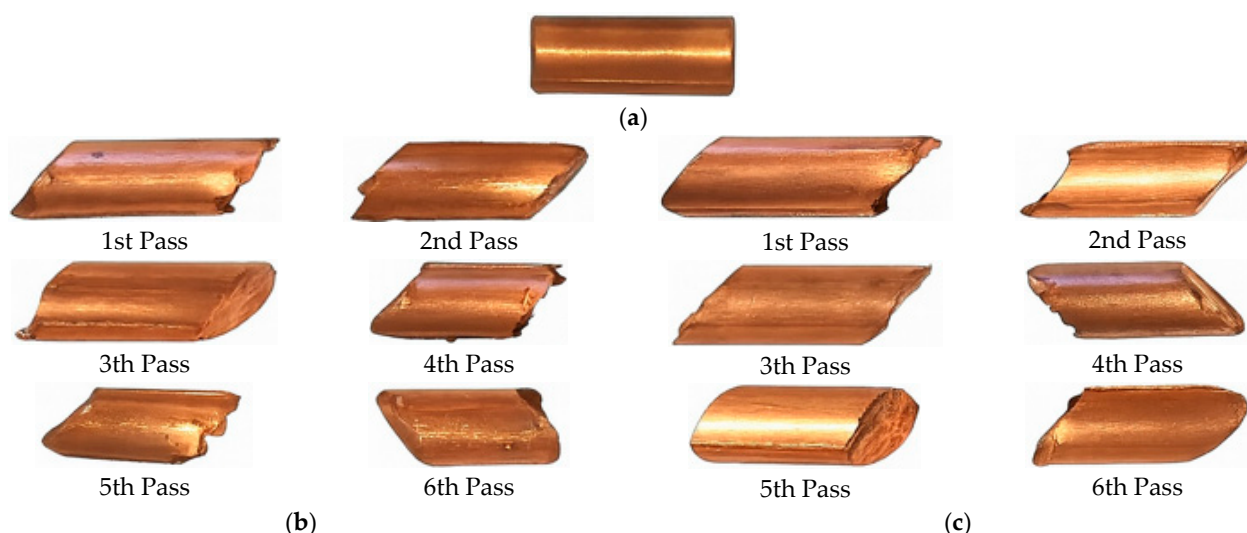


Figure 3. CuAg0.1 samples: (a) Unpassed sample, (b) ECAP processed samples at room temperature, and (c) at 100 °C.

After ECAP processing, the unprocessed and ECAP-processed samples were sectioned according to the conventional directional nomenclature used for ECAP-processed materials. The pressing direction of the ECAP sample is defined as the extrusion direction (ED), also referred to as the longitudinal (lateral) direction. The transverse direction (TD) is defined as the direction perpendicular to ED within the transverse (vertical) section, and direction parallel to ED within the longitudinal section. The nomenclature of the processed samples is summarized in Table 2.

Table 2. Nomenclature of the samples.

Sample	Abbreviation
Raw Material	Unpassed
at Room Temperature, Transverse Section, # of Pass 1-6	RTV1-6
at Room Temperature, Longitudinal Section, # of Pass 1-6	RTL1-6
at 100 °C, Transverse Section, # of Pass 1-6	100V1-6
at 100 °C, Longitudinal Section, # of Pass 1-6	100L1-6

The sectioned samples were mounted, ground, polished, cleaned with alcohol, and chemically etched for microstructural characterization. Two etching solutions were used: a 50 vol.% HNO₃–50 vol.% distilled water solution and Keller’s reagent composed of 2 mL HF, 3 mL HCl, 5 mL HNO₃, and 90 mL H₂O.

2.3. Characterization Methods

2.3.1. X-Ray Diffraction (XRD) Analysis

X-ray diffraction (XRD) analysis was performed to examine the crystallographic response of the unprocessed and ECAP-processed CuAg0.1 samples. The measurements were carried out using a Panalytical Empyrean X-ray diffractometer (Malvern Panalytical, Almelo, The Netherlands) with CuK α radiation ($\lambda = 1.5406 \text{ \AA}$) operated at 40 kV. The diffraction patterns were collected over a 2θ range of 40–100° in order to identify the main FCC Cu-based reflections and to evaluate the changes induced by ECAP processing.

The XRD analysis was used primarily to determine whether any additional crystalline phase was formed after ECAP and to evaluate deformation-induced changes in peak position, peak broadening, and relative peak intensity. Particular attention was given to the FCC reflections, including the (111), (200), (220), (311), and (222) planes. Changes in

peak broadening were interpreted as qualitative indicators of lattice distortion. Relative peak-intensity changes, especially the variation in the (111) reflection, were used to discuss the possible development of preferred crystallographic orientation during ECAP.

2.3.2. Optical Microscopy

Microstructural changes associated with the number of ECAP passes were examined by capturing optical microscope images (ZEISS Smartzoom 5, Carl Zeiss AG, Jena, Germany) of the polished and etched surfaces of the bakelite-mounted samples at 100×, 200×, and 500× magnification.

2.3.3. Hardness

The Vickers hardness values of the CuAg0.1 samples were measured on polished longitudinal and transverse sections using an Emco-Test™ DuraScan (Emco-Test Prüfmaschinen GmbH, Kuchl, Austria) hardness tester. A load of 1 kgf was applied for each indentation with a dwell time of 10 s. For each processing condition and section orientation, three indentations were performed at different locations on the polished surface. Adjacent indentations were separated by at least 2 mm to avoid overlap between the plastically deformed zones around neighboring indents and to minimize measurement interaction. The hardness results are reported as mean ± standard deviation.

Because the dimensions of the ECAP-processed samples were insufficient for machining standard ISO/ASTM-compliant tensile specimens, direct tensile testing was not performed in the present study. Therefore, the yield strength and ultimate tensile strength values were estimated from the measured Vickers hardness values using empirical hardness–strength correlations reported for copper alloys by Krishna et al. [25]. These estimates were calculated only to provide a comparative representation of the hardness-based strengthening trend as a function of ECAP pass number, processing temperature, and section orientation.

The hardness-derived estimated ultimate tensile strength and yield strength were calculated using the following equations

$$UTS_{est}[\text{MPa}] = 3.353 \times HV \quad (2)$$

$$YS_{est}[\text{MPa}] = 2.874 \times HV \quad (3)$$

where UTS_{est} is the estimated ultimate tensile strength, YS_{est} is the estimated yield strength.

The standard deviations of the hardness-derived estimated strength values were calculated from the standard deviations of the corresponding hardness values using the same empirical coefficients.

It should be emphasized that the estimated strength values are not direct tensile test results and should not be interpreted as a substitute for standard tensile characterization. In severely plastically deformed materials, hardness–strength relationships may be influenced by anisotropy, residual stress, strain-hardening capacity, deformation localization, dislocation density, and texture. Therefore, the hardness-derived strength values are used only as empirical indicators of the relative strengthening trend.

3. Results and Discussion

3.1. X-Ray Diffraction (XRD) Analysis

Figure 4 shows the XRD patterns of the unprocessed and ECAP-processed CuAg0.1 samples after different pass numbers at room temperature and 100 °C. The diffraction patterns of all samples exhibit the characteristic reflections of the Cu-based FCC structure, mainly corresponding to the (111), (200), (220), (311), and (222) crystallographic planes. No

additional crystalline phase was detected after ECAP processing under either temperature condition. This result should not be interpreted as evidence that a deformation-induced phase transformation was expected in the present CuAg0.1 alloy. Rather, it confirms that the observed changes in the XRD response are mainly associated with deformation-induced lattice distortion, defect accumulation, residual-stress evolution, peak broadening, and preferred orientation tendency rather than the formation of a new crystalline phase.

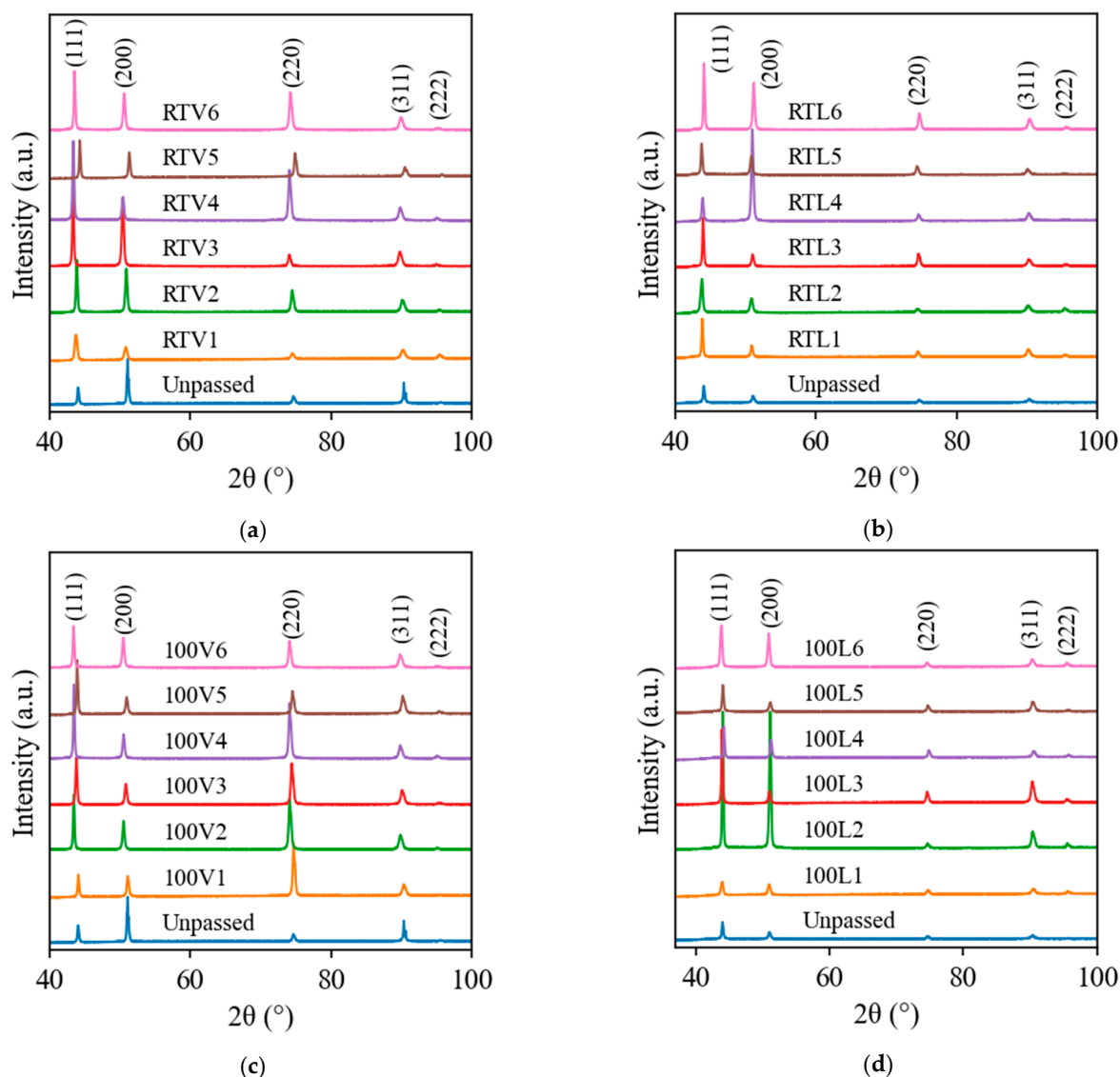


Figure 4. XRD analysis results: (a) Transverse section at room temperature, (b) Longitudinal section at room temperature, (c) Transverse section at 100 $^{\circ}$ C, and (d) Longitudinal section at 100 $^{\circ}$ C.

The most relevant changes in the XRD patterns are related to peak broadening, slight peak shifts, and variations in relative peak intensity. With increasing ECAP pass number, the diffraction peaks generally became broader, particularly for the room-temperature-processed samples. This broadening can be associated with the accumulation of lattice defects, increased microstrain, residual stress, and the reduction in coherent diffraction domains caused by severe shear deformation [26]. Therefore, the broadening observed in the present study could be interpreted as a qualitative or semi-quantitative indicator of deformation-induced defect accumulation rather than as a direct measurement of grain size or dislocation density.

A distinct change in the relative intensity of the FCC reflections was also observed after ECAP processing. In particular, the relative increase in the (111) reflection suggests the development of a {111}-related preferred orientation tendency during severe shear deformation. This interpretation is reasonable for FCC Cu-based alloys because plastic deformation occurs primarily through {111}<110> slip systems. The route B_c rotation sequence may further contribute to crystallographic reorientation because the sample is rotated by 90° in the same direction between successive passes, thereby changing the shear plane orientation during multi-pass processing. The increase in the (111) peak intensity may be interpreted only as qualitative evidence of preferred orientation tendency rather than as a fully quantified texture component.

The temperature-dependent XRD response indicates that room-temperature (Figure 4a,b) ECAP promoted stronger deformation-induced lattice distortion than ECAP at 100 °C. The room-temperature-processed samples exhibited more pronounced peak broadening with increasing pass number, which is consistent with higher defect storage and more limited recovery. In contrast, the samples processed at 100 °C (Figure 4c,d) showed a less monotonic evolution. This behavior can be attributed to the competition between deformation-induced defect accumulation and thermally assisted recovery [27]. Although 100 °C is not sufficiently high to confirm extensive dynamic recrystallization in the present CuAg0.1 alloy, it may promote partial dislocation rearrangement, defect annihilation, residual-stress relaxation, and limited substructural recovery. These mechanisms can reduce the net lattice distortion and explain the weaker hardening response observed at 100 °C compared with room-temperature processing.

The XRD results are discussed in combination with optical microscopy and Vickers hardness measurements to support the interpretation of deformation-induced microstructural evolution. The XRD patterns were primarily used to identify phase constitution and assess changes in relative peak intensities associated with deformation-induced preferred orientation or texture development during ECAP.

3.2. Microhardness and Hardness-Derived Estimated Strength

Figure 5 shows the Vickers hardness evolution of the CuAg0.1 alloy as a function of ECAP pass number, processing temperature, and section orientation. The unprocessed alloy exhibited relatively low hardness values, consistent with its undeformed initial condition. After the first ECAP pass, a pronounced hardness increase was observed under both processing temperatures. At room temperature, the hardness increased from approximately 53 HV to values above 120 HV after the first pass, corresponding to an improvement of more than 125%. This sharp increase indicates that severe shear deformation introduced by ECAP effectively promoted defect accumulation, lattice distortion, and deformation-induced microstructural subdivision in the CuAg0.1 alloy [28].

The hardness increase continued with increasing pass number, but the hardening increment gradually decreased and approached a saturation-like regime after approximately the fourth pass. This behavior is typical of severely plastically deformed metallic materials, where further strain becomes progressively less effective in increasing hardness [29,30]. During the early ECAP passes, intense shear deformation promotes rapid dislocation multiplication, dislocation interaction, and the formation of deformation-induced cells and substructures [31]. At higher accumulated strain, however, the net increase in stored defect density becomes limited because dislocations rearrange into lower-energy configurations, accumulate at cell/subgrain boundaries, and are partly annihilated through dynamic recovery. Therefore, the hardness plateau after the fourth pass can be interpreted as the result of a dynamic balance between defect generation, defect rearrangement, and recovery-assisted processes.

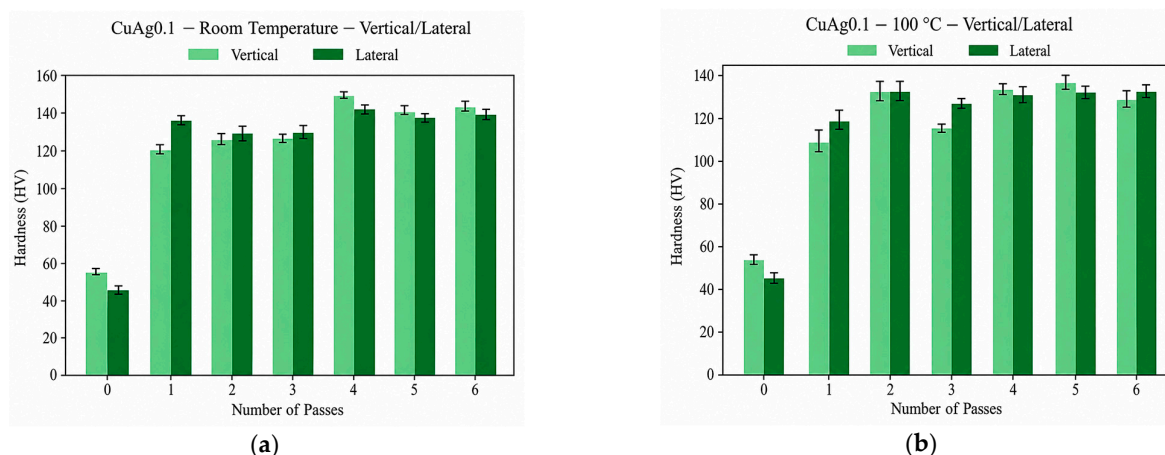


Figure 5. Vickers hardness values of CuAg0.1 samples processed by ECAP: (a) at room-temperature and (b) at 100 °C.

The average hardness of the unprocessed CuAg0.1 alloy, calculated from the longitudinal and transverse sections, was approximately 49.14 HV. After four ECAP passes at room temperature, the average hardness increased to approximately 142.34 HV, corresponding to an increase of about 190%. For the samples processed at 100 °C, the average hardness after four passes was approximately 131 HV, corresponding to an increase of about 167%. These results indicate that most of the achievable hardness improvement under the present ECAP conditions was obtained within the first four passes, whereas additional passes provided only limited incremental hardening.

The room-temperature-processed samples exhibited higher hardness values than the samples processed at 100 °C. This difference can be attributed to the stronger suppression of recovery at room temperature, which allows a larger fraction of the imposed shear strain to be stored as dislocations, lattice distortion and deformation-induced substructures. In contrast, processing at 100 °C promotes thermally assisted dislocation rearrangement, partial defect annihilation, and limited substructural recovery. These mechanisms reduce the net stored defect density and partially counteract the hardening effect of SPD. Therefore, the lower hardness values obtained at 100 °C are interpreted mainly in terms of recovery-assisted softening under the warm-processing condition.

Figure 6 presents the hardness-derived estimated yield strength and ultimate tensile strength values calculated from the measured Vickers hardness values. Since these values were obtained using empirical hardness–strength correlations, they follow the same pass-dependent trend as the hardness results. Accordingly, the estimated strength values increased sharply after the first ECAP pass and approached a saturation-like trend after approximately four passes. The room-temperature ECAP samples showed higher estimated strength values than the 100 °C samples, reflecting their higher measured hardness and stronger defect-storage capability.

Section-dependent differences were observed in both hardness and hardness-derived estimated strength values. These differences suggest the development of ECAP-induced anisotropy associated with severe shear deformation, section orientation, and possible preferred crystallographic orientation. This interpretation is consistent with the XRD results, where relative intensity changes in the crystallographic patterns of the FCC structure indicated a preferred orientation tendency.

Within the investigated processing, room-temperature ECAP for four passes can be considered the most effective condition in terms of hardness-based strengthening and processing efficiency. This condition produced most of the attainable hardness improvement while additional passes resulted in only limited further hardening.

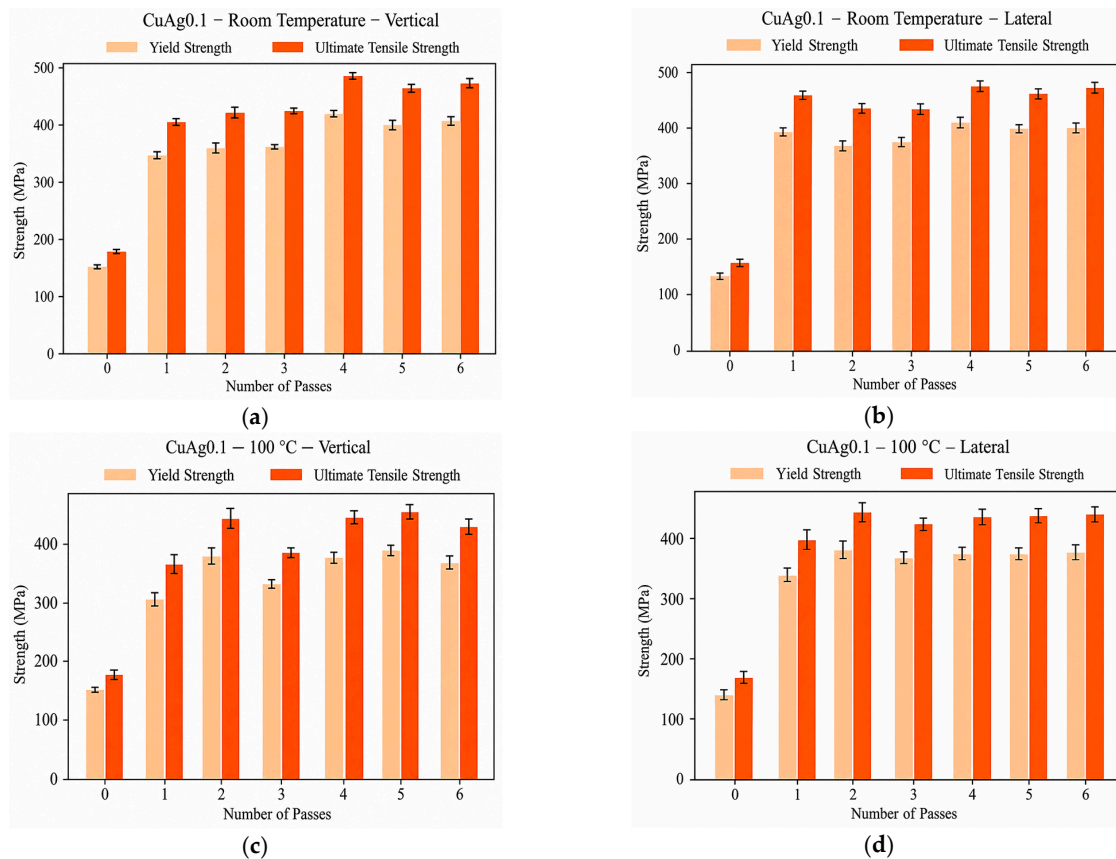


Figure 6. Hardness-derived estimated strength values from hardness results: (a) Transverse section and (b) Longitudinal section at room temperature; (c) Transverse section and (d) Longitudinal section at 100 °C.

3.3. Optical Microscopy

Figures 7 and 8 present the optical micrographs of the unprocessed and ECAP-processed CuAg0.1 samples. Figure 7 shows the initial microstructure of the unprocessed alloy on the longitudinal and transverse sections. The unprocessed sample exhibits a coarse and relatively heterogeneous grain morphology, which is consistent with the cast and undeformed initial condition of the alloy. The optical micrographs indicate the presence of coarse grains together with smaller grains within the observed field [22]. This initial morphology provides a reference for evaluating the deformation-induced microstructural changes after ECAP.

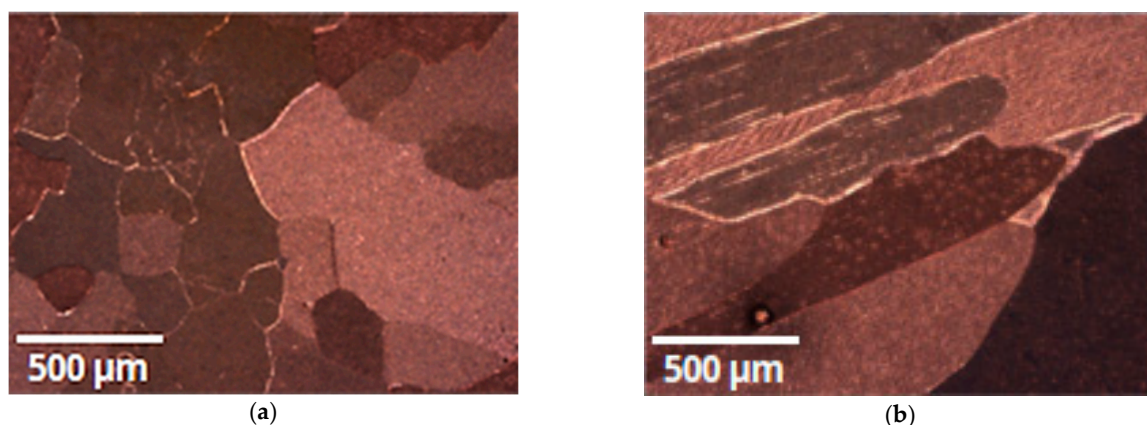


Figure 7. Optical microstructure images (200× magnification) of unprocessed CuAg0.1 sample: (a) Transverse section, and (b) Longitudinal section.

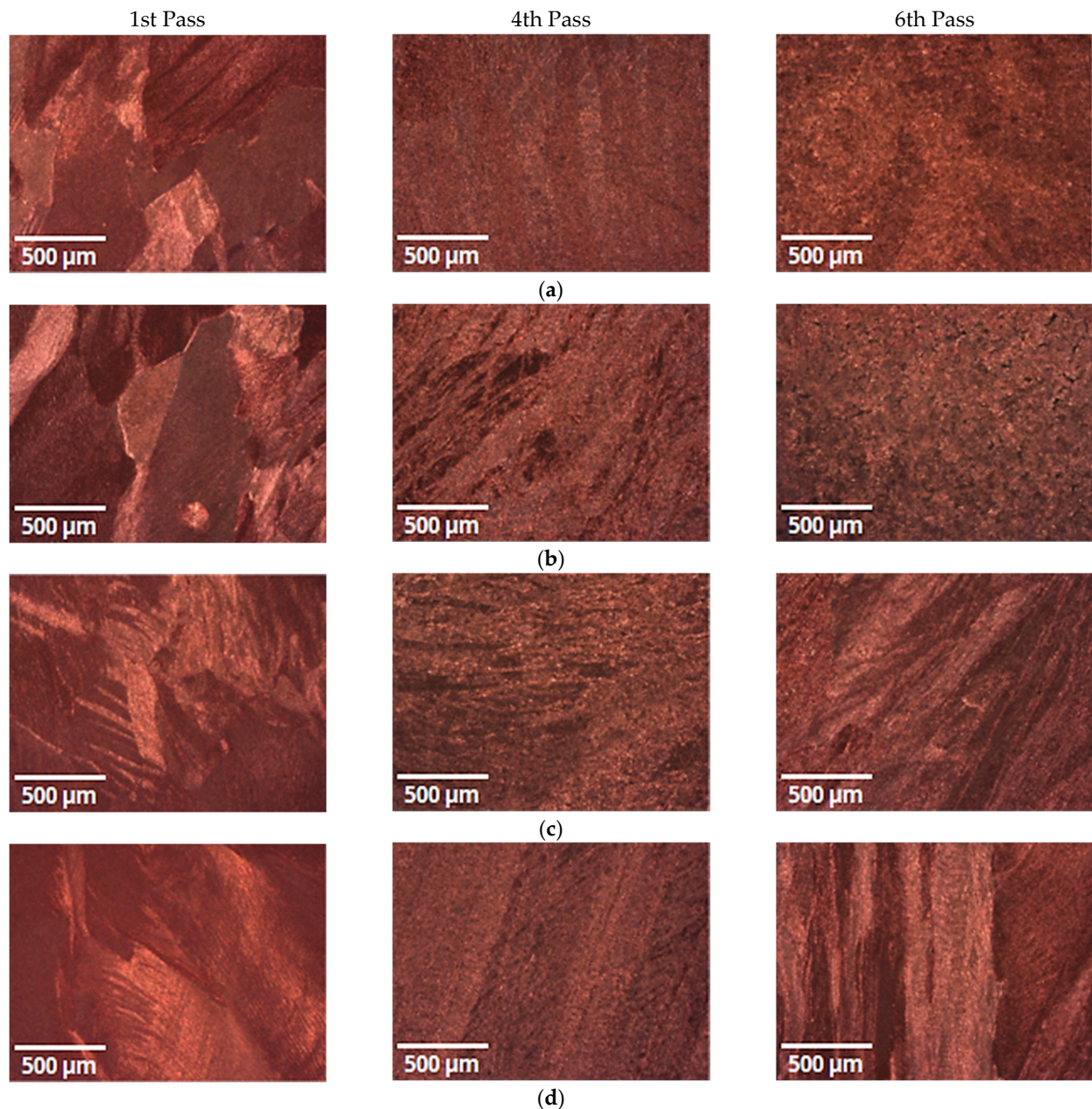


Figure 8. Optical micrographs of ECAP-processed CuAg0.1 alloy after the 1st, 4th, and 6th passes at room temperature and 100 °C: (a) Transverse and (b) Longitudinal sections at room temperature; (c) Transverse and (d) Longitudinal sections at 100 °C.

Figure 8 shows the optical microstructures of the CuAg0.1 alloy after the 1st, 4th, and 6th ECAP passes at room temperature and 100 °C. After the first pass, elongated deformation features and shear-band-like regions become visible. These features are consistent with the severe shear deformation imposed during ECAP. With increasing pass number, the microstructure exhibits progressive subdivision of the initially coarse grains into more elongated and fragmented regions. This evolution is more evident in the room-temperature-processed samples, where the deformation contrast becomes stronger and the microstructure appears more highly subdivided after multiple passes [32].

For the room-temperature ECAP condition, the optical micrographs show pronounced deformation bands and a more fragmented morphology as the number of passes increases. This observation is consistent with the higher hardness values measured for the room-temperature-processed samples. The stronger microstructural contrast suggests that a larger

fraction of the imposed strain was stored in the material as deformation-induced substructure and lattice distortion [33,34]. However, optical microscopy does not provide sufficient resolution to directly quantify dislocation density, subgrain size, or grain-boundary misorientation. Therefore, the observed morphological changes are interpreted as qualitative evidence of deformation-induced microstructural subdivision rather than as direct proof of high-angle grain-boundary formation or dislocation-cell development [35].

The samples processed at 100 °C show a comparatively less pronounced deformation contrast than the room-temperature-processed samples. The deformation bands and fragmented regions are still visible, but the microstructure appears somewhat smoother and less sharply subdivided. This difference is consistent with the lower hardness response obtained at 100 °C. Since 100 °C is a low warm-processing temperature for CuAg0.1 alloy, the observed response is attributed primarily to thermally assisted recovery processes, such as partial dislocation rearrangement [35,36].

Section-dependent differences are also observed between the longitudinal and transverse sections. These differences reflect the non-uniform shear deformation path imposed by ECAP and are consistent with the section-dependent hardness response reported in this study. They may also be related to the preferred orientation tendency suggested by the XRD relative peak-intensity changes.

Overall, the optical microscopy observations support the conclusion that ECAP induces progressive microstructural subdivision in CuAg0.1 alloy and that this effect is more pronounced under room-temperature processing than at 100 °C. These observations are consistent with the XRD peak-broadening trends and Vickers hardness results.

3.4. Industrial Relevance

The present results indicate that room-temperature ECAP can substantially improve the hardness response of CuAg0.1 alloy without changing the overall sample geometry. This finding is relevant for high-conductivity Cu-based components where improved surface hardness and mechanical resistance are required together with dimensional stability. Potential application areas include electrical connectors, contact pins, terminals, relay and switch elements, conductive spring contacts, and fine conductive wire products. In such components, enhanced hardness may contribute to improved resistance against local plastic deformation, handling damage, and contact-related mechanical degradation [37,38].

From an industrial perspective, the four-pass room-temperature ECAP condition appears to be the most effective processing condition within the investigated parameter window. This condition provided most of the achievable hardness increase, whereas additional passes produced only limited incremental hardening. Therefore, four-pass ECAP may offer a more efficient balance between strengthening response and processing effort. However, this conclusion is based primarily on Vickers hardness and hardness-derived estimated strength values. It should not be interpreted as a complete optimization of service performance.

4. Conclusions

This study investigated the effect of ECAP processing temperature and pass number on the XRD response, microstructural evolution, and Vickers microhardness behavior of CuAg0.1 alloy. The results show that ECAP is an effective severe plastic deformation route for improving the hardness response of this CuAg0.1 alloy, particularly when processing is performed at room temperature. XRD analysis showed that no new crystalline phase was detected after ECAP processing, indicating that the Cu-based FCC phase was retained under the investigated conditions. This result is not interpreted as evidence of an expected phase transformation, but rather confirms that the observed changes are

mainly related to deformation-induced lattice strain, peak broadening, defect accumulation and preferred orientation tendency. The relative increase in the FCC (111) peak intensity suggests the development of a {111}-related preferred orientation tendency during ECAP. Room-temperature ECAP produced stronger peak broadening and higher hardness values than ECAP at 100 °C, indicating more effective defect storage and limited recovery. In contrast, processing at 100 °C resulted in a lower net hardening response, attributed to thermally assisted recovery, dislocation rearrangement, and partial dislocation annihilation. The Vickers microhardness increased markedly after the initial ECAP passes and approached saturation after approximately four passes. Therefore, among the investigated conditions, room-temperature ECAP up to four passes provided the most favorable hardness-based strengthening response. Overall, ECAP shows potential as a resource-efficient processing route for CuAg0.1 alloy. The present study provides a qualitative interpretation of the process–microstructure–property relationship based on XRD, optical microscopy, and Vickers microhardness measurements. To further validate the proposed strengthening mechanisms and microstructural evolution, future work should include direct tensile testing, electrical conductivity measurements, EBSD/TEM characterization, and thermal-stability evaluation, thereby establishing a more comprehensive and quantitative understanding of the relationships among ECAP processing parameters, microstructural evolution, and the mechanical performance of the CuAg0.1 alloy.

Author Contributions: Conceptualization, E.A. and M.Ö.; methodology, E.A., M.R., O.A. and M.S.; investigation, E.A., M.R., M.Ö. and O.A.; resources, M.R. and M.Ö.; data curation, M.R. and M.Ö.; writing—original draft preparation, E.A. and M.S.; writing—review and editing, M.Ö. and M.S.; visualization, M.Ö. and M.R.; supervision, E.A. and M.S. All authors have read and agreed to the published version of the manuscript.

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Data Availability Statement: The data that support the findings of this study are available from the corresponding author upon reasonable request.

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Conflicts of Interest: Orhan Akyüz was employed by the company Er-Bakır Electrolytic Copper Products Co. The remaining authors declare that the research was conducted in the absence of any commercial or financial relationships that could be construed as a potential conflict of interest.

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