

Dynamic Deformation and Microstructural Characteristics of AA5083 Under the Influence of Electrochemical Behaviour Using Equal Channel Angular Pressing

Nagendra Singh^{1*}, Manoj Kumar Agrawal²

Abstract

Large-scale Equal channel angular pressing is a great engineering material that has more potential for AA5083 industrial applications. Uniaxial dynamic compressive tests across a temperature of 448 K were conducted for the ECAP technique AA5083 in order to create such a design framework. The microstructure was described using transmission electron microscopy and scanning electron microscope both before and after dynamic loading. The plastic flow behaviour of the AA5083 was described by a novel dynamic constitutive model that was built based on the experimental results. A particle swarm optimization method was used to determine the impacts of plastic strain, temperature and strain rate on the new model's input parameters. The strong agreement between model predictions and experimental findings suggests that the predictive constitutive model used today is very good at simulating the behaviour of dynamic deformation the large-scale ECAP method AA5083. The surface characteristics of the alloy were investigated using electrochemical impedance spectroscopy in the sodium chloride solution. The acquisition of the corrosion performance, scanning electron microscopy and Raman spectroscopy were performed. As the pass number rises, the alloy's microhardness and tensile strength get better. In ECAP technique the alloy's wear characteristics significantly improved along with its hardness and tensile strength. The AA5083 is taken into account in the current investigation both before and after equal channel angular pressing. Comparing the material's fracture locus with the initial material, the ECAP method revealed somewhat higher equivalent strain to fracture values.

Keywords: AA5083, Dynamic Response, ECAP, Thermal Activation, Raman Spectroscopy, Plasticity Methods, Plasticity Fracture.

INTRODUCTION

Due to its exceptional mechanical and physical features, including as its high specific strength, superior formability, and great corrosion resistance, AA5083 has been used extensively in the rail transit, automotive, nautical, and aerospace industries. The need for high-performance structural materials grows, a variety of innovative techniques have been devised to enhance the mechanical properties of a wide array of materials including AA5083. Equal channel angular pressing has gained recognition as a crucial and promising approach for refining the grain size and enhancing the strength of metals and alloys. According to sources, AA5083's mechanical characteristics can be further improved by Equal

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Received Date: January 06, 2024

Accepted Date: February 21, 2024

Published Date: April 16, 2024

Citation: Nagendra Singh, Manoj Kumar Agrawal. Dynamic Deformation and Microstructural Characteristics of AA5083 Under the Influence of Electrochemical Behaviour Using Equal Channel Angular Pressing. Journal of Polymer & Composites. 2024; 12(Special Issue 1): S95-S112

Channel Angular Pressing, which shows promise for expanded use in a variety of applications [1]. With the aim of enhancing predictive capabilities ensuring the accuracy of numerical simulations for deformation processes and indirectly promoting the broader engineering application of the ECAP technique on AA5083. It is essential to comprehend the fundamental constitutive relationships that describe the deformation behavior and this can be achieved through the formulation of a more precise constitutive model. A number of constitutive models have been created to describe how metallic materials behave when they deform plastically. To find that the aforementioned models are not able to predict the dynamic deformation behaviour of the given material with sufficient accuracy over a wide temperature range. This restriction results from their inability to take into consideration the phases that emerge in the microstructure and how they change as a result of dynamic loading and deformation processes [2]. As a result, some classic models have been updated, and more accurate models have been proposed to rectify this shortcoming. An example of such modifications is evident in the adaptation of the Zerilli-Armstrong (Z-A) model by Voyiadjis and Abed. This adjustment became necessary as the original model fell short in accurately characterizing dynamic plastic deformation at elevated temperatures and lacked the incorporation of the strain rate effect for thermally activated processes. The concept of mechanical threshold stress as an internal state variable. Such as the necessity for low-temperature mechanical experiments and the subsequent intricate process of parameter determination for the Mechanical Threshold Stress model.

The challenge in implementing numerical methods arises due to scale transitions and the utilization of implicit algorithms for models based on crystal plasticity. Therefore, in this study, the microstructure of AA5083 processed through the ECAP technique was initially characterized using scanning electron microscopy and transmission electron microscopy. Subsequently, dynamic uniaxial compression tests were conducted using a split Hopkinson pressure bar system across a broad temperature range. The microstructure of the material post-impact was analyzed using scanning electron microscopy and transmission electron microscopy. The investigated alloy can be used in various engineering applications where high strength is required [3]. To evaluate the effect of ECAP on AA5083, the microstructure, mechanical, and wear parameters of the alloy were examined both before and after the treatment. From these findings and thermal activation theory, a novel constitutive model was formulated to elucidate the mechanical behavior of the tested AA5083 material. The newly proposed constitutive model takes into account the initial microstructure and its transformation under varying dynamic loading conditions. Based on the experimental mechanical data, the parameters for the formulated model were determined using the Particle Swarm Optimization method.

High strength, outstanding weldability, and corrosion resistance are displayed by AA5083 in its ultra-fine-grained form. Consequently, there is a substantial demand for this material and it finds applications in various components of vehicle bodies, as well as in the manufacturing of armor plates ship bodies, and more. In response, researchers have undertaken thorough investigations to create Ultra-Fine-Grained materials by employing severe plastic deformation techniques. On the other hand, research on the corrosion properties of aluminium alloys that undergo equal-channel angular pressing is still ongoing. Pitting typically occurs when mechanical flaws, random local discontinuities, scratches, or the protective coating encircling second-phase particles are damaged. The literature addressing the corrosion of materials undergoing Severe Plastic Deformation indicates that certain studies propose a reduction in corrosion resistance for Ultra-Fine-Grained aluminum alloys, especially those processed by Equal Channel Angular Pressing.

MATERIALS AND EXPERIMENTAL PROCEDURE

Experimental Materials

Sample Preparation and Die Setup

Purchased was AA5083, which was in H34 condition. The alloy was machined to a diameter of 10 mm and a length of 50 mm in order to prepare the samples for ECAP. The alloy's chemical composition has been examined using the ARL spark analyzer, as shown in Table 1. With a maximum

magnesium content of 4.47%, it was demonstrated that the alloy was an AA5083. The tool steel ECAP die had two equal channels with a circular cross section that measured 10.1 mm in diameter. because it reduced some of the pressing load and produced strain homogeneity in the billet material. The channel intersection angle (ψ) of 105° had been chosen. In order to prevent bending-like distortion, which is more common with larger outer corner angles (ψ), a fillet radius of 4 mm has been chosen. In order to shield the samples' top surfaces from damage and fissures, a fillet with a radius of 1 mm was formed in the inner corner of the two crossing channels. This was done since the specimens were damaged by the sharp inner corner. To make it simple to remove the extruded specimen from the die, a split type die design has been adopted. Figure 1 displays a schematic of the ECAP die.

Iwahashi et al.'s equation (1) can be used to calculate the equivalent strain for N passes.

$$\varepsilon = N/\sqrt{3} \left[2\cot\left(\frac{\phi}{2} + \frac{\psi}{2}\right) + \psi\operatorname{cosec}\left(\frac{\phi}{2} + \frac{\psi}{2}\right) \right] \quad (1)$$

Where the number of passes N, channel intersection angle ϕ , and outer corner angle ψ are all equal to the equivalent plastic strain ε [6].

Table 1. Chemical Compositions of AA5083 in Weight Percent and Annealing Condition for 1 hr at 320°C Air Cool [4].

Alloy	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Al
5083	0.08	0.21	0.02	0.58	1.154	0.06	<0.01	-	bal

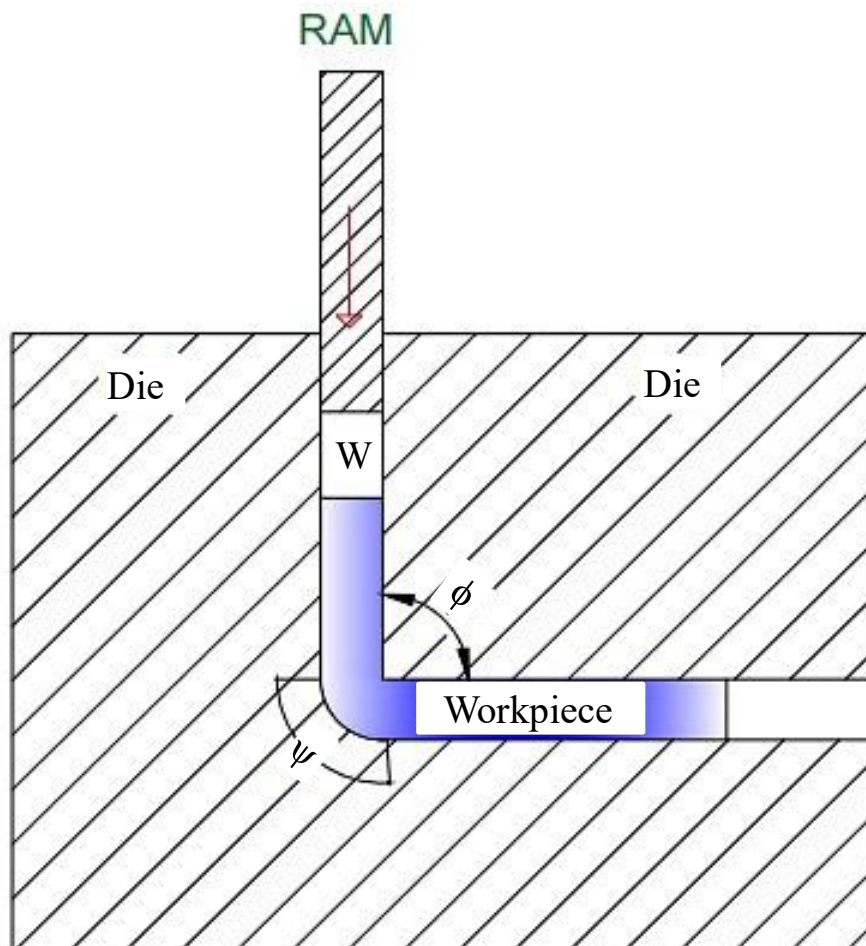


Figure 1. A Schematic of ECAP Die [5].

Using a hydraulic press with a 100 ton capacity, the prepared samples were forced through the ECAP die both at ambient temperature and at -150°C for cryogenic conditions. Lubricants were utilised to lower the friction between the samples and the die surface, including MoS_2 grease and SAE 68 oil. The rod that was used in this operation had a 50 mm length and a 10 mm diameter. The rod was then submerged in liquid nitrogen for approximately 60 seconds. The sample is then produced and quickly pressed into the ECAP entry channel. After each pass, the samples were rotated to a direction of 90° at route- B_C . Using a scanning electron microscope, microstructural analyses of the materials before and after ECAP were carried out. The microhardness of the samples along the transverse plane was measured using a Vickers microhardness tester both before and after ECAP with cryogenic treatment and without cryogenic treatment. Extensometer was used to measure the samples' tensile strength while they were at room temperature. The samples for the tensile test were machined with a 6 mm gauge diameter and 30 mm gauge length in the pressing direction. The 4 B_C route of large-scale ECAP was used to process the AA5083 at 280°C . In accordance with the route selection and processing conditions given in Figure 1, As seen in Figures 3 and 4, the material's microstructure was examined using transmission electron microscopy and field-emission scanning electron microscopy. On a desktop SHPB system, the dynamic uniaxial compression experiments were carried out]. The compression experiments were performed with nominally constant strain rates at temperatures ranging from 308 K to 588 K. The specimen was loaded in a direction that was perpendicular to the pressing direction along its short dimension, either 3.6 mm or 1.8 mm [7]. Separate quasi-static compression tests using a DNS100 universal testing machine outfitted with an electrothermal furnace were performed at nominal strain rates of 0.15 s^{-1} and temperatures ranging from 308 K to 738 K in order to determine the athermal stress parameters for the new constitutive model. Similar to dynamic testing, the specimen was loaded along an axis with a length dimension of 4.0 mm that was perpendicular to the direction of pressing shown in Figure 2. The universal testing machine's initial steel inserts were changed out for high-temperature inserts composed of an alumina ceramic. On the loading surfaces of the specimen, a lithium-based molybdenum disulfide lubricant was used to minimize friction.

Equal Channel Angular Pressing Technique of the AA5083 Alloy and Optical Microstructure

The samples were compressed using a hydraulic press with a 210-ton maximum load capacity for one to three passes at room temperature, A 0.8 mm/s pressing speed was used. For the ECAP technique of the materials, path C was chosen as the processing path depicted in Figure 3. The choice of the ECAP processing method was made in accordance with the available literature, which shows that ECAP via route C results in a homogeneous distribution of plastic strain in the processed material in Figure 4.

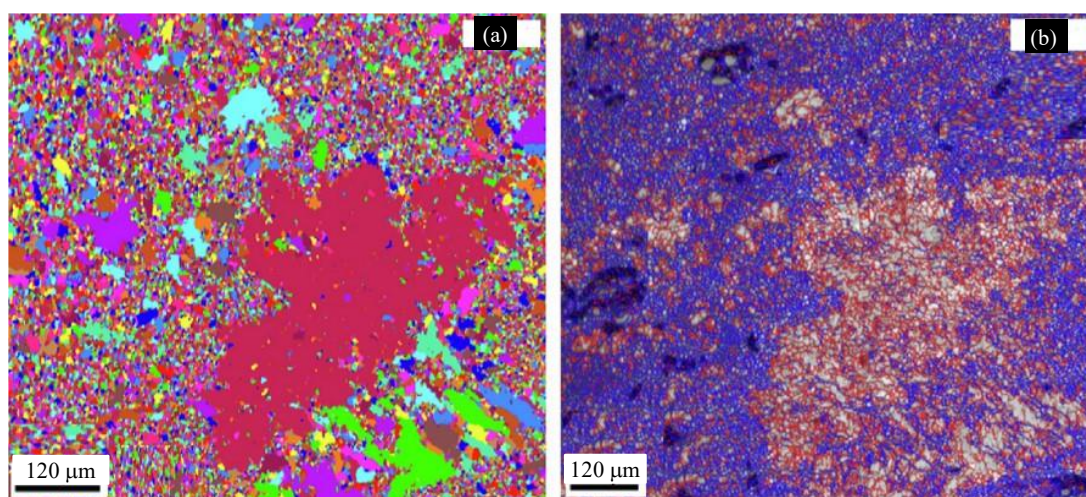


Figure 2. EBSD results of Microstructure of ECAP Technique AA5083 for (a) Grain Morphologies and (b) Grain Boundaries Before Impact at $120\text{ }\mu\text{m}$ [8].

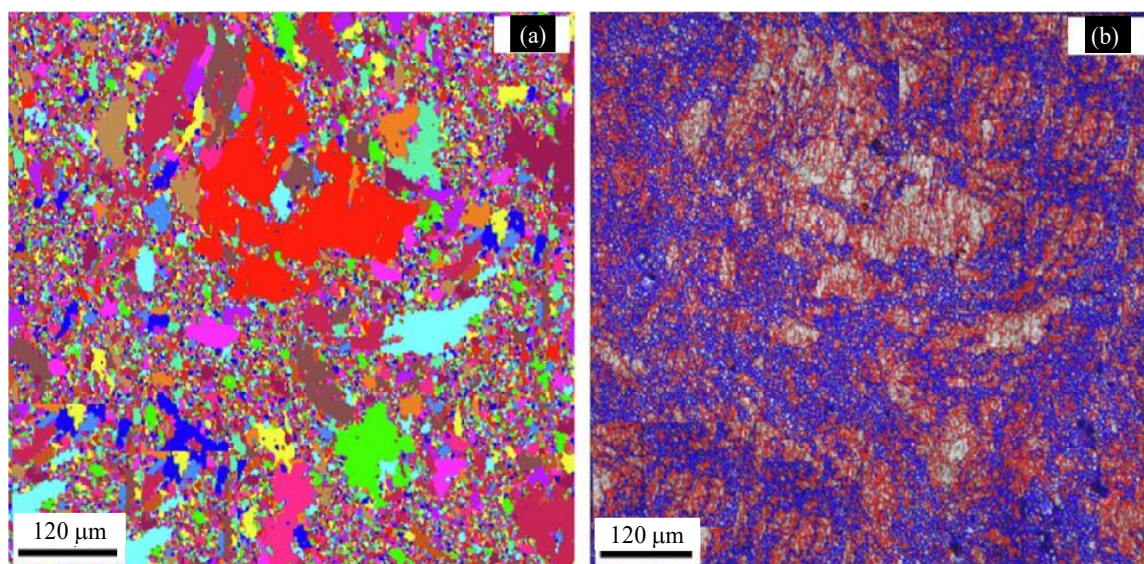


Figure 3. EBSD results of Microstructure of ECAP Technique AA5083 for (a) Grain Morphologies and (b) Grain Boundaries After Impact at 120 μm [8].

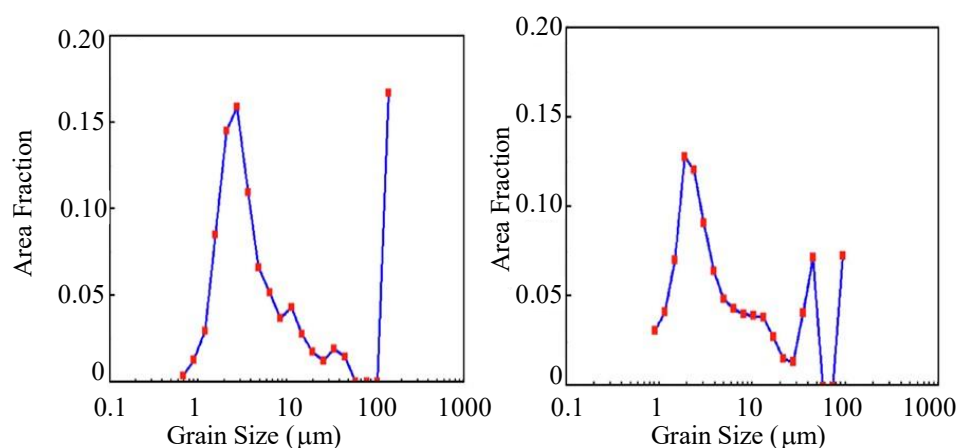


Figure 4. EBSD results of Microstructure of ECAP Technique AA5083 for Grain Size Distribution Before and After Impact [9].

Various grades of silicon carbide emery papers (180, 120, 1/0, 2/0, etc.) were used to polish the samples. The samples were also polished with cloth and abrasives such as powdered alumina. Prior to being cleaned with an ultrasonic cleaner, the samples were polished. Before being examined under a Leica optical microscope, the samples underwent a microstructural analysis in Figure 5.

Corrosion Test, Raman Spectroscopy and Scanning Electron Microscopic Analysis

The corrosion properties of samples from the ECAP were investigated. Every specimen underwent metallographic polishing. Using a potentiostat and potentiodynamic polarization in a 3.8% NaCl solution, the polished specimens were further evaluated to determine their corrosive qualities. Using commercial NaCl and distilled water, the necessary chemical solutions were made. It should be noted that the open-circuit potential test was carried out utilizing the aforementioned test solution for 1 hour and 15 minutes to obtain a consistent state before moving on to the actual potentiodynamic tests. A potentiodynamic analysis was then carried out. For these tests, a standard three-cathode cell with a saturated calomel anode and a graphite bar was utilised. In these investigations, a polished sample functioned as the working electrode, a SCE as the reference electrode, and graphite as the counter electrode. The broad frequency range of 120 kHz to 0.02 Hz was used for electrochemical impedance spectroscopy. A Jobin Yvon confocal micro-Raman spectrometer was used for the Raman

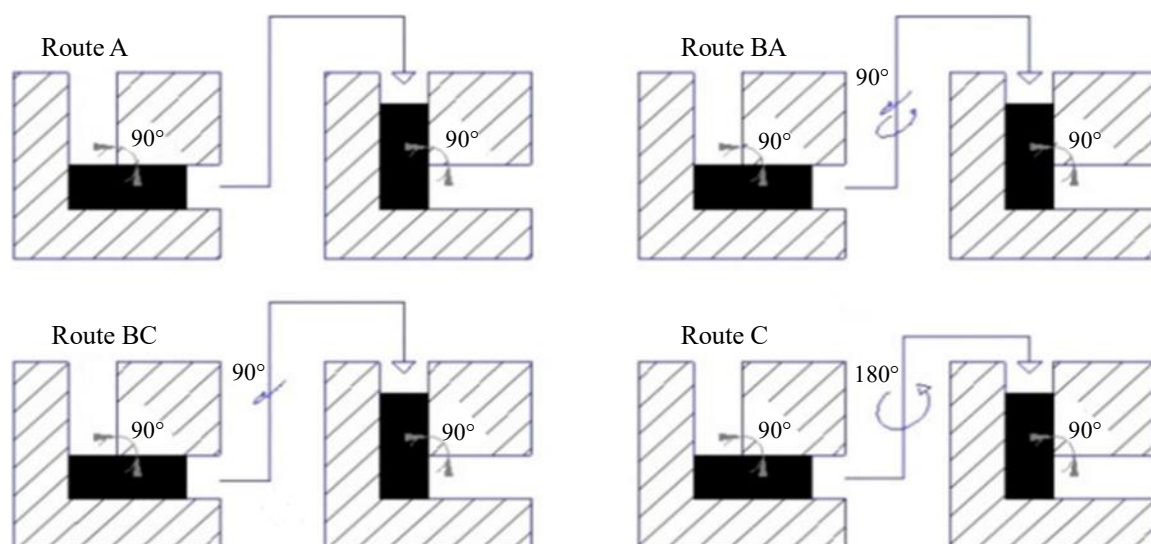


Figure 5. A Schematic Diagram with Different Route of Equal Channel Angular Pressing Die [10].

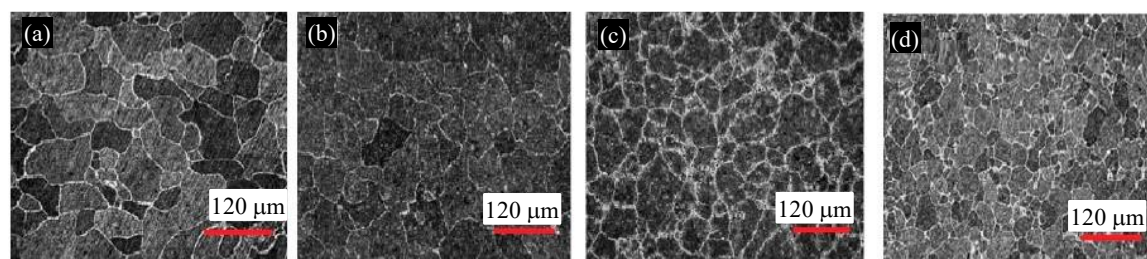


Figure 6. Microstructure of AA5083: (a) Annealed; ECAP following the First, Second, and Third Passes (b, c, d) [11].

spectroscopy. To determine how the chemical composition of the surface changed as a result of ECAP passes, Raman spectroscopy was used. The use of spectrometer backscattered geometry through a microscope objective was made. The Raman spectrometer contained 1850 grooves/mm of grating. The spectrometer was aligned using the 540 cm^{-1} Raman band of a silicon wafer. The repeatable tests used three points from each sample to record and collect all Raman spectra in the $220\text{ to }1200\text{ cm}^{-1}$ range Figure 6. Using a scanning electron microscope, the surface morphology of the equal-channel, angularly pressed, as-received AA5083 was investigated.

EXPERIMENTAL RESULTS

Microstructure

The distinctive grain structure at both high and low-angle grain boundaries, along with the accompanying distribution of grain sizes in AA5083, undergoes alteration through Equal Channel Angular Pressing (ECAP) both prior to and following dynamic compression deformation. The alloy has a substantial amount of fine, equiaxed grains with an average grain size of a few micrometers. In the ECAP process applied to AA5083, there are instances of larger unrefined grains, reaching diameters of up to ten micrometers. These grains possess low-angle grain boundaries, indicating that the dynamic recovery or recrystallization was not entirely accomplished during the multiple pressing passes. After dynamic plastic deformation, it was discovered that grains were marginally refined, with the average grain size decreasing from $4.12\text{ }\mu\text{m}$ to $3.76\text{ }\mu\text{m}$. The perimeter length of the low-angle GBs grows from 8.80 cm to 10.41 cm after dynamic compression, whereas the perimeter length of the high-angle GBs is essentially unaltered, 10.42 and 10.54 cm for before and after dynamic deformation, respectively. Additionally, it was discovered that the alloy has a significant amount of precipitates with erratic and plate-like forms. The Transmission Electron Microscopy (TEM) findings reveal the presence of precipitates in the material prior to dynamic compression. The central particle is

detailed further in Figure 6 as a bright-field image as a dark-field image at a higher magnification [12]. The dynamic compression at a temperature of 738 K and a strain rate of 3400 s⁻¹, the TEM-EDX elemental mapping images in the specimen. The findings from TEM-EDX affirm that these particles exhibit a high concentration of alloying elements such as Mg, Si, Fe, Mn, and Cr. This observation aligns with the outcomes reported in prior studies. Furthermore, it's evident that there exists a fractured precipitate in this sample following dynamic plastic deformation at an elevated temperature.

Mechanical Properties

Figure 7 illustrates the typical true stress versus true strain behavior across a temperature range spanning from 308 K to 588 K. The data is presented for three different strain rates: 520 s⁻¹, 3400 s⁻¹, and 24000 s⁻¹. Each of the blue, green, and red lines represents exemplary curves corresponding to temperatures of 308 K, 438 K, and 588 K, respectively. Compression tests were conducted under each condition, and each set of conditions was replicated at least three times. The obtained test results exhibited outstanding quality and reproducibility. Moreover, it's important to highlight to the reader that the constrained strain observed in the stress versus strain curves at 520 s⁻¹ and 3400 s⁻¹ resulted from the limitations imposed by the length of the bar utilized in the Split Hopkinson Pressure Bar (SHPB) test. The outcomes obtained at the strain rates of 520 s⁻¹ and 3400 s⁻¹ remain applicable for formulating the constitutive model. This is justified by the fact that the deformation of the samples under these conditions extends beyond the elastic limit and encompasses plastic flow. It can be noted that the ECAP technique AA5083 has a distinctly positive strain rate effect, meaning that the flow stress increases with increasing loading rate. The strain hardening and temperature softening effects are clearly seen in the real stress-true strain graphs, which is typical of FCC metals.

Assuming high temperatures and a strain rate of 0.15 s⁻¹, Figure 7 displays the quasi-static true stress vs. true strain graphs. The testing were terminated at a strain of 0.35 since more loading could severely distort the samples, which would lead the stress-strain result to be incorrect. Comparison of these curves reveals two salient traits. The first is that the capability for work hardening reduces as test temperature rises because the softening effect of rising temperature cancels out the work hardening effect. While the flow stress of the curve at 738 K doesn't vary much, the curve at room temperature shows a clear work hardening. The second is the significant effect of temperature softening on compression loading [14]. The curves show that as the temperature rises, the temperature-induced reduction in flow stress diminishes and approaches the athermal stress limit. We point out that the athermal parameters of the proposed constitutive model were determined using the true stress versus true strain curve at 738 K.

Constitutive Modeling

It is generally acknowledged that the flow stress of metals and alloys depends on their microstructural properties as well as the testing environment, particularly the temperature and strain rate. The flow stress (σ) can be determined as a function of the plastic strain ($\dot{\epsilon}$), the strain rate (ϵ) and the temperature (T), as shown by Eq. (1), and is additionally evaluated in relation to a specific plastic deformation level (ϵ) for FCC metals and microstructure (S) [15]:

$$\sigma = f(S, \dot{\epsilon}, T) = f(\epsilon, \dot{\epsilon}, T) \quad (2)$$

According to some studies, there are two main categories of barriers for dislocation motion: short-range and long-range barriers. The vast majority of fine precipitates in a range of sizes for the ECAP technique AA5083 are of the short-range barrier type. Thermal activation may be used to break down the short-range barriers; as such, strain rate and temperature are key variables. It is possible to disregard the long-range barriers' sensitivity to temperature and strain rate. As a result, the total of the atmospheric stress $\sigma_a(\epsilon)$ and the thermally activated stress ($\epsilon, \dot{\epsilon}, T$) can be used to express the flow stress, which represents the metallic material's resistance to dislocation motion in the plastic deformation zone:

$$\sigma = \sigma_a(\epsilon) + \sigma_{th}(\epsilon, \dot{\epsilon}, T) \quad (3)$$

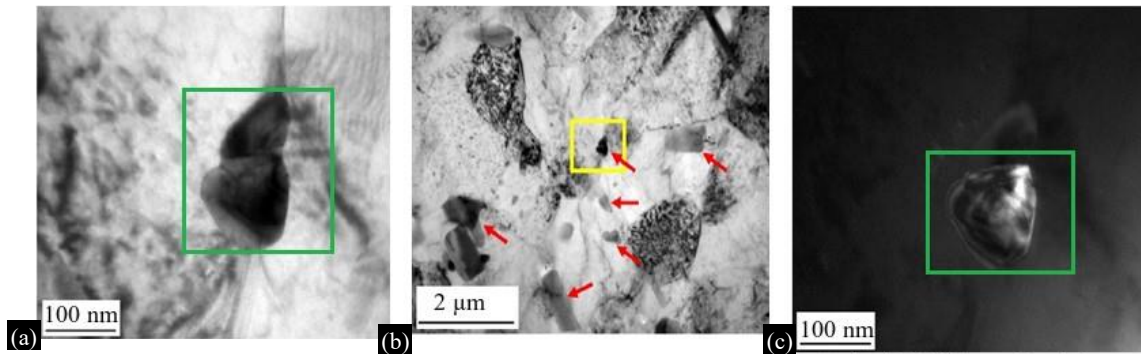


Figure 7. TEM Images of Precipitates of ECAP Technique AA5083 [13]. (a) The bright field image of precipitate (b) Low magnification TEM image of precipitates (c) The dark field image of precipitate.

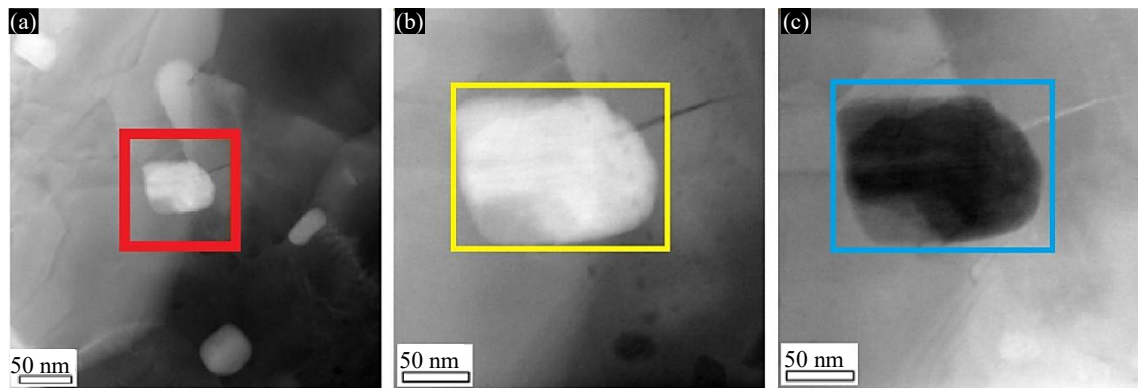


Figure 8. TEM EDX Elemental Mapping Images [17]. (a) TEM dark field image (b) Enlarged dark field image (c) Enlarged bright field image.

Athermal Stress

The athermal stress for a particular metallic material is typically assumed to depend only on the initial microstructure and to remain constant throughout the entire plastic deformation process, and is therefore defined in most models as [16]:

$$\sigma_a = \sigma_I + \sigma_S \quad (4)$$

The flow stress in Eq. (3) is caused by the initial imperfections or impurities, and the flow stress indicating size effects is caused by grain boundaries and twinning boundaries, which together cause a rise in stress. The Hall-Petch relationship can be used to directly represent the stress associated with GBs because it occurs in all polycrystalline metals and alloys. For FCC metals, deformation twinning has been seen during the plastic deformation process at high strain rates. Due of twinning's low temperature sensitivity, the stress connected to TBs was absorbed by the athermal stress. Furthermore, it has been demonstrated that the stress brought on by TBs follows a pattern like the Hall-Petch effect. As a result, the stress brought on by size effects is described as [17]:

$$\sigma_S = K_d d^{-1/2} + K_\lambda \lambda^{-1/2} \quad (5)$$

In Eq. (8), the microstructure coefficient denoted as K_d , linked to the Hall-Petch effect d , along with the mean diameter of initial grains λ , the mean boundary spacing between twins, and a coefficient K_λ indicating the connection between material strength and twin spacing. But snagging all four of these parameters directly shown in Figure 8.

The reciprocal of activation volume V^{-1} and grain size $d^{-1/2}$ in FCC metals has been observed to have a Hall-Petch-type relationship, or it varies linearly with grain size. Combining the thermal

activation analysis for plastic flow rate and the pile-up model of dislocation can explain the grain size dependency of the reciprocal of activation volume. Given that GBs and TBs both serve to restrict dislocation motion, Lu and Dao et al. In Figure 9 generalized a Hall-Petch-type connection to show V^{-1} how the activation volume's size depends on both the impacts of GBs and TBs as shown in Eq. (5) [19]:

$$(V^{-1} - V_0^{-1}) = K_d d^{-1/2} + K_\lambda \lambda^{-1/2} \quad (6)$$

where V_0 is the relationship between the intra-grain and intra-twin dislocation mechanisms and the physical activation volume shown in Figure 10. The link between the activation volume and plastic strain and the dislocation density, represented as Eqs. (6) and (7) [19],

$$(V^{-1} - V_0^{-1}) \sim \sqrt{\tilde{\rho} - \tilde{\rho}_0} \quad (7)$$

$$\tilde{\rho} = \tilde{\rho}_0 + \lambda_a [1 - \exp(-K_0 \epsilon)] \quad (8)$$

where $\tilde{\rho}$ is the initial density of the dislocations that interact with long-range obstacles $\tilde{\rho}_0$ is the far-field dynamic recovery coefficient shown in Figure 11 is a material constant related to the number of these dislocations K_a and is the density of the dislocations that interact with long-range barriers λ_a .

The above equations can be used to convert the athermal stress associated with the size effect into a function of plastic strain, as shown in Eq. 8 [21].

$$\sigma_s = A [1 - \exp(-K_a \epsilon)]^{1/2} \quad (9)$$

where A is a distinct stress metric that has no physical relevance in the real world.

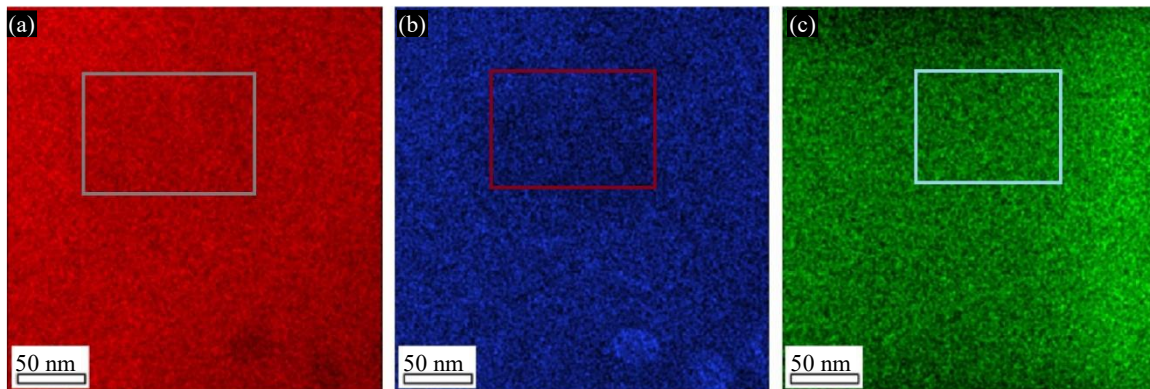


Figure 9. TEM EDX Elemental Mapping Images [18]. (a) Al mapping image (b) Mg mapping image (c) Si mapping image.

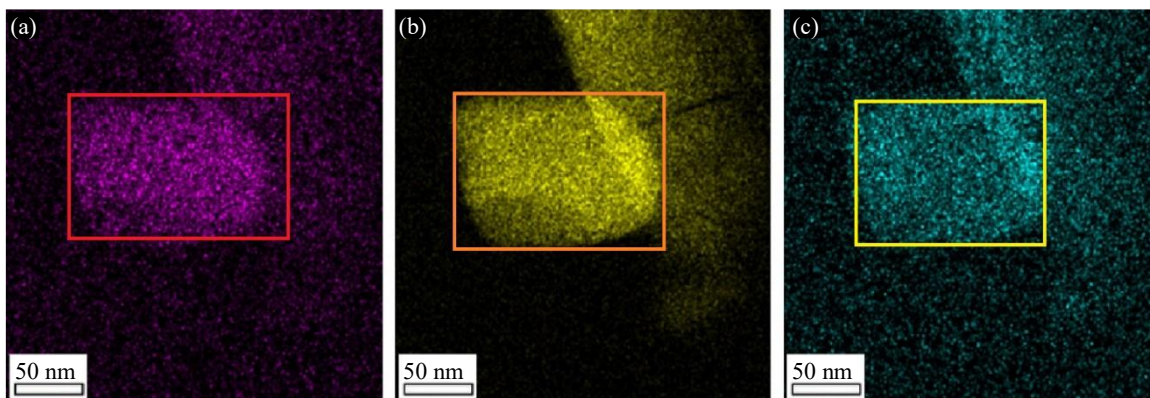


Figure 10. TEM EDX Elemental Mapping Images [18]. (a) Fe mapping image (b) Mn mapping image (c) Cr mapping image.

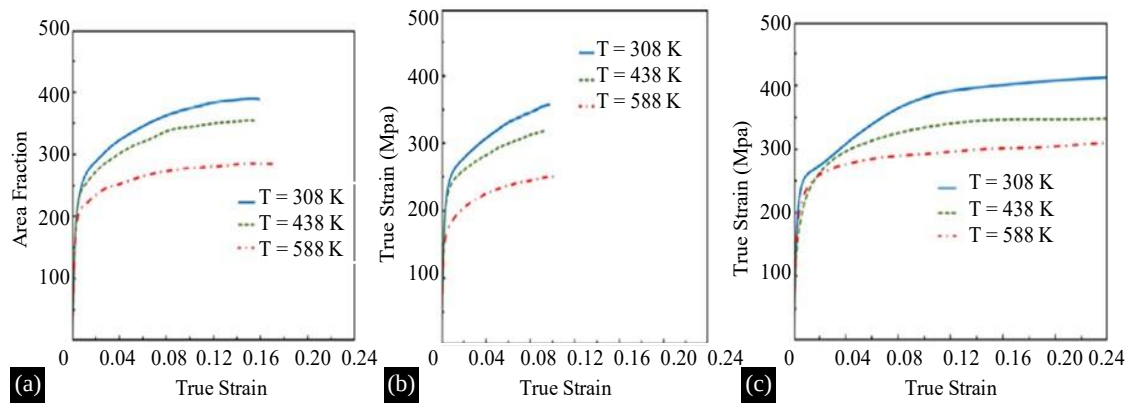


Figure 11. Representative True Stress versus True Strain Curves of Dynamic Uniaxial Compression Test at Strain Rate 520 s^{-1} , 3400 s^{-1} and 24000 s^{-1} [20].

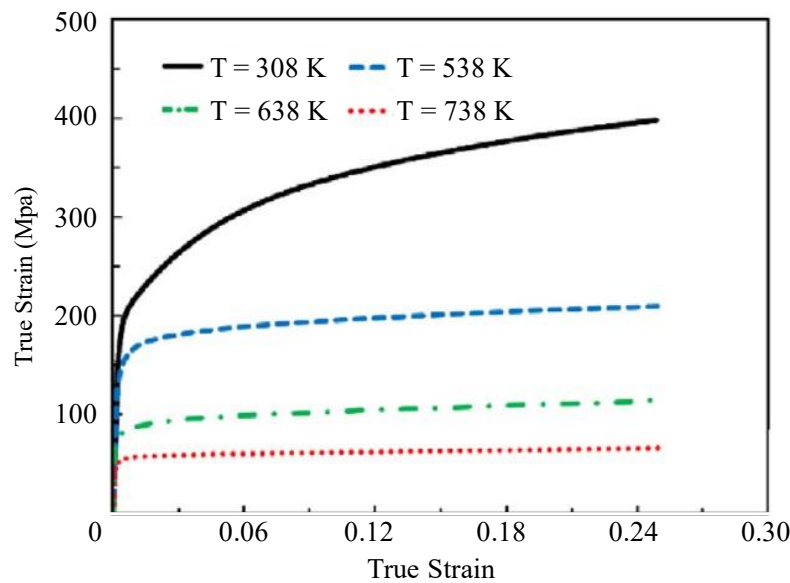


Figure 12. True Stress versus True Strain Curves of Quasi-static Compression Tests at Strain Rate of 0.01 s^{-1} [22].

So far, the strain-dependent stress component that the size effect previously contributed to athermal stress has replaced the challenging, expensive, and time-consuming microscopic characterization of grain size and twin spacing with straightforward shown in Figure 12, affordable mechanical testing and parameter fitting.

Thermally Activated Stress

An Arrhenius-type equation can be used to characterize the link between the shear strain rate and thermal activation energy, as indicated in Eq. (10) [23].

$$\dot{\gamma} = \dot{\gamma}_0 \exp\left(\frac{-\Delta G(\tau)}{kT}\right) \quad (10)$$

The activation energy, $\Delta G(\tau)$ is Boltzmann constant, $\dot{\gamma}$ shear strain rate, and reference shear strain rate $\dot{\gamma}_0$ are all given in Eq. (10).

$$\Delta G(\tau) = G_0 \left[1 - \left(\frac{\tau_{th}}{\hat{\tau}_{th}}\right)^p\right]^q \quad (11)$$

Within the parameters of Eq. (11), the thermally activated shear stress at any temperature τ_{th} and at absolute zero $\hat{\tau}_{th}$, which is intimately related with two shape constants $0 < p \leq 1$ and $1 < q \leq 2$,

determines the structure of resistance to dislocation motion resulting from short-range barriers. In the meantime, is a representation of the total energy required for dislocation motion is G_0 . The shear strain rate can be stated as by using Eqs. (10) and (11).

$$\tau_{th} = \hat{\sigma}_{th} \left[1 - \left(\frac{kT}{-G_0} \ln \frac{\dot{\gamma}}{\dot{\gamma}_0} \right)^{1/q} \right]^{1/p} \quad (12)$$

For the situation of uniaxial compression, Eq. (12) can be recast as by using the von Mises yield criterion:

$$\sigma_{th} = \hat{\sigma}_{th} \left\{ 1 - \left[-\beta T \left(\ln \frac{\dot{\epsilon}}{\dot{\epsilon}_0} \right)^{1/q} \right] \right\}^{1/p} \quad (13)$$

In Eq. (13), $\dot{\epsilon}_0$ is the reference strain rate, and stand in for the thermally activated stress at any temperature and at absolute zero, σ_{th} and $\hat{\sigma}_{th}$ respectively. Also, and is a material constant β and equal to k/G_0 . The expression for the thermally activated tension at absolute zero is [24]:

$$\hat{\sigma}_{th} = \bar{\sigma}_{th} \epsilon^n \exp \left(\alpha T \ln \left(\frac{\dot{\epsilon}}{\dot{\epsilon}_1} \right) \right) \quad (14)$$

where the saturation of strain hardening is n , the reference strain rate is $\dot{\epsilon}_1$, the reference thermal stress is $\bar{\sigma}_{th}$, and the strain hardening exponent α is a material constant. The thermally activated stress can therefore be represented as:

$$\hat{\sigma}_{th} = \bar{\sigma}_{th} \epsilon^n \exp \left(\alpha T \ln \left(\frac{\dot{\epsilon}}{\dot{\epsilon}_1} \right) \right) \left\{ 1 - \left[-\beta T \left(\ln \frac{\dot{\epsilon}}{\dot{\epsilon}_0} \right)^{1/q} \right] \right\}^{1/p} \quad (15)$$

The parameters of the test temperature should also be changed since, when the strain rate exceeds 12 s^{-1} , the adiabatic conditions that predominate during the plastic deformation of the specimens result in a non-negligible increase in actual temperature. As a result, the estimated real temperature is [25]:

$$T + T_0 + \int_0^\epsilon \frac{\eta}{\rho C_v} \sigma d\epsilon \quad (16)$$

the material's density ρ , which was determined to be 3.66 g/cm^3 for the ECAP technique AA5083 in the current investigation, C_v is temperature-dependent specific heat capacity at constant volume, the initial test temperature T_0 . This is quantifiable using conventional thermal analysis techniques, and which can roughly be written as $C_v = 0.785 \exp(T/3200)$ ($320 \text{ K} \leq T \leq 788$), where is the plastic-work-heat conversion coefficient, also known as the Taylor-Quinney factor, and is fixed at 1.2 for aluminium alloys.

Final Constitutive Model

The following thermal-activation-based constitutive equation was created for the large-scale ECAP approach AA5083 for dynamic deformation across a wide temperature range utilising Eqs. (9), (15), and (16).

$$\sigma = \sigma_1 + A[1 - \exp(-K_a \epsilon)]^{1/2} + \hat{\sigma}_{th} \epsilon^n \exp \left(\alpha T \ln \left(\frac{\dot{\epsilon}}{\dot{\epsilon}_1} \right) \right) \left\{ 1 - \left[-\beta T \left(\ln \frac{\dot{\epsilon}}{\dot{\epsilon}_0} \right)^{1/q} \right] \right\}^{1/p} \quad (17a)$$

$$T + T_0 + \int_0^\epsilon \frac{\eta}{\rho C_v} \sigma d\epsilon \quad (17b)$$

The preceding one-dimensional equation can be directly extended to three dimensions by relating the plastic strain and flow stress with the equivalent values of the strain or strain tensor, respectively. This yields a three-dimensional constitutive model. The fundamental presumption is that the material will act isotropically in all major directions.

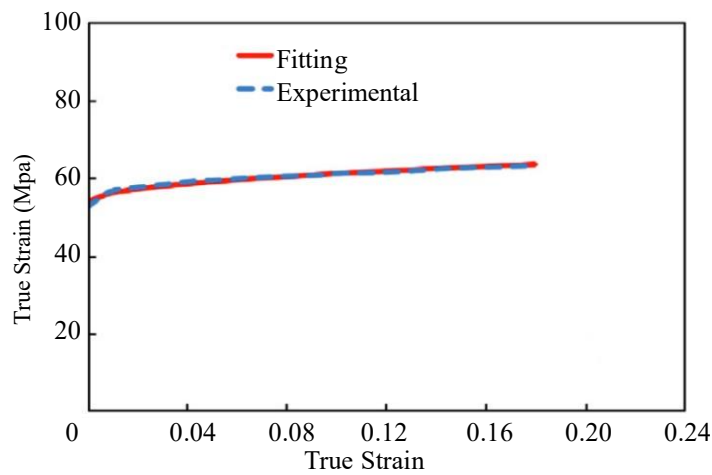


Figure 13. Determination of Parameters related to Athermal Stress [27].

Table 2. Input parameters that have been determined for use in current constitutive model for ECAP technique AA5083 [28]

Parameter	Value	Parameter	Value
η	1.2	n	0.1129
ρ	3.66 g/cm ³	$\dot{\epsilon}_0$	$10.89 \times 10^8 \text{ s}^{-1}$
σ_1	63.87 MPa	$\dot{\epsilon}_1$	$1.2 \times 10^9 \text{ s}^{-1}$
A	43.27 MPa	α	1.2×10^{-7}
K_a	0.80	β	1.18×10^{-5}
C_v	$0.775 \exp(T/3200) \text{ J/g} \cdot \text{K}$	p	0.1292
$\hat{\sigma}_{th}$	575.92 MPa	q	1.0

Parameter Determination

Overall, there are two groups that can be made up of all the variables in the established constitutive equation. The parameters that are simple to learn from experiments or the literature; these parameters include ρ , and C_v . Both parameters associated to the thermally activated stress ($\hat{\sigma}_{th}$, n , α , $\dot{\epsilon}_1$, β , $\dot{\epsilon}_0$, p , q) and those related to the athermal stress must (σ_1 , A , K_a) be found using nonlinear fitting. Determining the parameters in category 2 is therefore the main focus of this section. Regarding the three athermal stress parameters, as shown in Figure 13, they may be directly determined by applying a least-squares method based on the findings of uniaxial compression using the ECAP technique AA5083 at a temperature of 738 K and a strain rate of 0.01 s⁻¹. The three parameters were established as follows: $\sigma_1 = 63.87 \text{ MPa}$, $A = 43.27 \text{ MPa}$, $K_a = 0.8$. It is far more difficult to calculate the values of the parameters related to the thermal stress using non-linear fitting to guarantee global optimality. So, a MATLAB programme was created to calculate the parameters related to the thermal stress based on the particle swarm Optimization algorithm, a commonly used global optimization technique [26]. Figure 7 displayed the exact steps of this technique. Table 2 contains a list of the parameters. The results of the comparison between the experimental data of dynamic uniaxial compression at various temperatures and strain rates and those predicted by the current model and Z-A model are depicted in Figure 8 (a), (b), and (c). where the solid lines represent the experimental data and the solid and hollow symbols represent the results of fitting the Z-A constitutive model of the material using the PSO approach, as given in Eq. (18).

$$\sigma = 246.6 + 1577.8\epsilon^{1/2} \exp(-0.0055T + 0.000167T \ln \dot{\epsilon}) \quad (18)$$

According to the graph, the established model performs better than the Z-A model in accurately describing the dynamic plastic flow stress of the experimental materials over the whole range of test

temperatures and strain rates. The stress at 12% plastic strain was projected using the present constitutive model, which covers a wide range of temperatures 210 to 610 K and strain rates 120 s^{-1} to 25200 s^{-1} , as shown in Figure 13, to further illustrate the effect of strain rate and temperature on the plastic flow stress. Both strain rate hardening and temperature softening effects are taken into account by the current model. comparisons between the experimental data and the data generated by the current model and the Z-A model. a three-dimensional representation of the proposed constitutive model's capacity for prediction [29].

DISCUSSIONS

Microstructure and Grain Size Analysis

The alloy had coarse grains when it was first obtained. The microstructures demonstrated the possibility of the creation of finely grained, microscopically homogenous structures in the presence of substantial deformation. The measurement of grain size was conducted utilizing ImageJ software. The first sample was received with a particle size of $148 \mu\text{m}$. After the first pass, the grain size decreased to $105 \mu\text{m}$, the second pass to $68 \mu\text{m}$, and the third pass to $40 \mu\text{m}$. The process of grain refinement was a result of the deformation induced by the Equal Channel Angular Pressing.

The longitudinal section showed an increase in grain sizes whereas the cross-section showed a decrease in grain sizes. The tested samples' grains' lengthening and refining. Because the original grains in the sample separated into smaller subgrain groups after the first trip through the channel, a reduction in grain size was mostly observed after that. The first pass's dislocation augmentation speed was much higher than its dislocation obliteration speed because of the multi-pass ECAP and the low dislocation density. These types of material grains can essentially be purified. The grain was further polished during the course of multiple passes due to the expansion of the dislocation density and internal energy. After a few passes, the grain fineness gradually decreased due to the development of recrystallization and crystalline recovery, both of which were induced by the internal energy expansion [31]. For the various pressing routes, the specimen's grain refinement development process was once again modified. Route C was also exceptional in terms of the accumulation and equilibration of the dislocations. The analysis of the deformation and dislocation progression may very well demonstrate that the ECAP process can be used to obtain materials with a nanostructure. The process of refining grains can be described as consistently unique recrystallization and recovery. The grain refining process, as seen from the perspective of a microstructure analysis, is to oversee the lively evening beyond the age and the dislocations' deconstruction.

Electrochemical Behavior of AA5083

Aluminium develops a protective coating of AlOOH when it is exposed to water, but anodic breakdown products (AlOHCl and AlOHCl_2) appear when this passive layer is exposed to chloride ions. These products are prone to dissolving in water, which promotes the passive film's anodic disintegration. Figure 14 displays the polarisation curves of the ECAP and AA5083 samples in the 3.5% NaCl solution. Table 3 displays the data that Tafel's extrapolation method yielded from these curves.

The curve and the data unmistakably demonstrate that, as compared to the as-received sample, the corrosion resistance property (i_{corr} value) of the ECAP rises with increasing passes. Regarding the corrosion potential (E_{corr} value), there was no discernible pattern. In comparison to the ECAP sample, the as-received sample had the greatest value (E_{corr}), which is explicable by the micrograph shown in Figure 15. After each pass, this alloy's grain refining got finer. When grains are refined, more sites are available for the corrosion to start. In two different ways, grain refinement increases the potential for corrosion: (1) within the grain border, and (2) near the grain boundary. The outcomes unmistakably show that an increase in corrosion rate in the aluminium alloy is likely to occur with more ECAP passes as a result of improved grain structure. In the 3.8% NaCl solution, the Bode magnitude and Bode phase are plotted, respectively. The Bode magnitude diagram shows two distinct

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regions. The Bode magnitude plot shows consistent values $\log|Z|$ vs. in the lower and higher frequency $\log|F|$ portions, showing a reaction to the solution resistance. The middle-frequency spectra showed a linear slope of around 1 in the wide (0.21200.0 Hz) range. This is a typical reaction to the surface film's capacitive behaviour.

Table 3. The 3.8% NaCl Solution is the Consequence of Potentiodynamic Polarization [32].

Sample	i_{corr} (A/cm ²)	E_{corr} (V)
Annealed	1.2×10^{-6}	-4.29×10^{-1}
First pass	2.2×10^{-5}	-5.55×10^{-1}
Second pass	1.2×10^{-4}	-6.18×10^{-1}
Third pass	3.2×10^{-4}	-4.59×10^{-1}

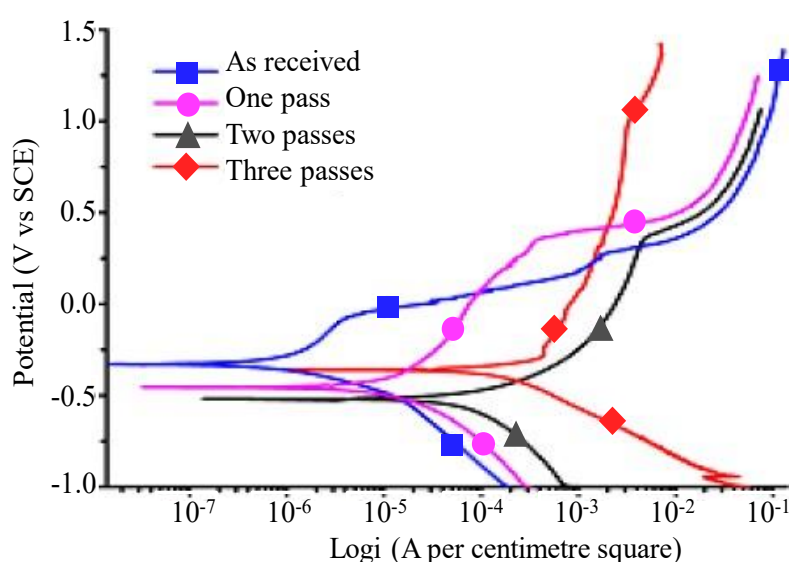


Figure 14. The AA5083's Potentiodynamic Polarisation Curve in a 3.8% NaCl Solution [30].

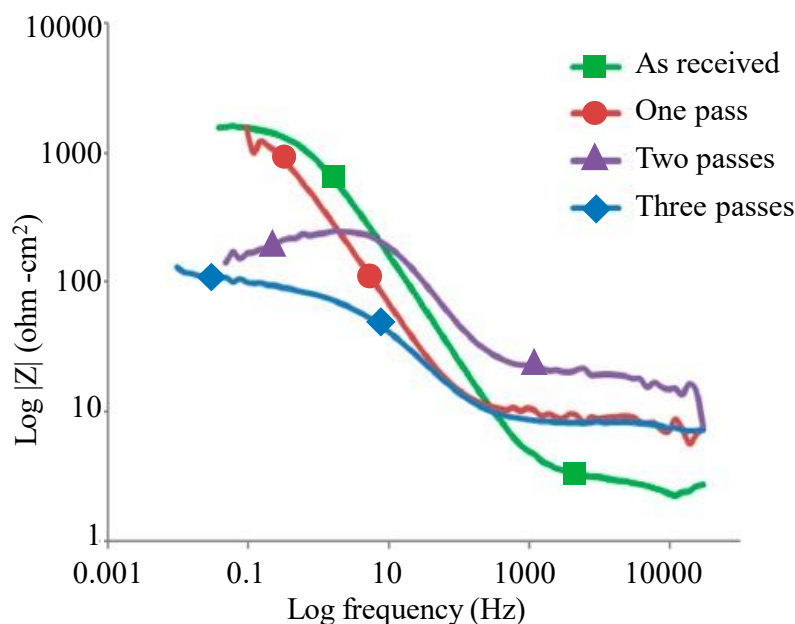


Figure 15. The AA5083 Electrochemical Impedance Spectroscopy Curve in a 3.8% NaCl Solution is Shown in [33].

The phase angle 70° in the Bode phase plot of the initially received sample moved towards in the intermediate frequency range. On the other hand, the phase angle steadily dropped throughout the first, second, and third passes, approaching 30° . This indicates a surface diffusion that impacted the solution resistance, leading to a subsequent decrease in corrosion resistance. The character of each Nyquist plot in Figure 16 displays a pattern that is essentially the same for each sample. We can therefore safely assume that the ECAP has no effect on the surface character. The Nyquist plots revealed two distinct time constants: (1) a high-frequency one linked to a well-defined double-layer constant phase element (Q_{dl}), and (2) a low-frequency counterpart associated with a significantly dissipated loop. In specific instances, an inductive loop characterized the behavior all around.

The same NaCl solution was used for these samples' EIS. The impact of grain refinement on corrosion resistance was also discovered in this article. The parameters that were extracted from the equivalent circuit were: solution resistance, charge transfer resistance, constant phase element, inductance element, and electrical inductance, in that sequence. The capacitive area in HF/IF was crucial to the establishment of a double layer capacitance at the surface of the Al alloy and the solution interface as well as the charge transfer resistance. The inductive loop was required for the production, adsorption, and desorption of corrosion products on the metal surface [35]. The pitting corrosion's early stage is indicated. The Al alloy's polarization resistance and corrosion resistance were inversely related. A few publications have discussed how variations in grain size caused by the prolonged forced strain caused by the ECAP affect corrosion resistance, as referenced.

Scanning Electron Microscopic Study

SEM measurements were performed on the potentiodynamically polarized annealed and ECAP samples in the 3.8 wt% NaCl solution. SEM micrographs in Figure 17 demonstrate plainly little corrosion-related degradation, particularly during the three passes of the ECAP sample test. Figure 18 shows the development of relatively small, semi-circular pits spread throughout the surface of the unaltered material. The ECAP approach led to an upsurge in pitting attacks. After three passes, a much larger and deeper rectangular pit appeared. This attack took place at the point where the Al lattice and particle interact.

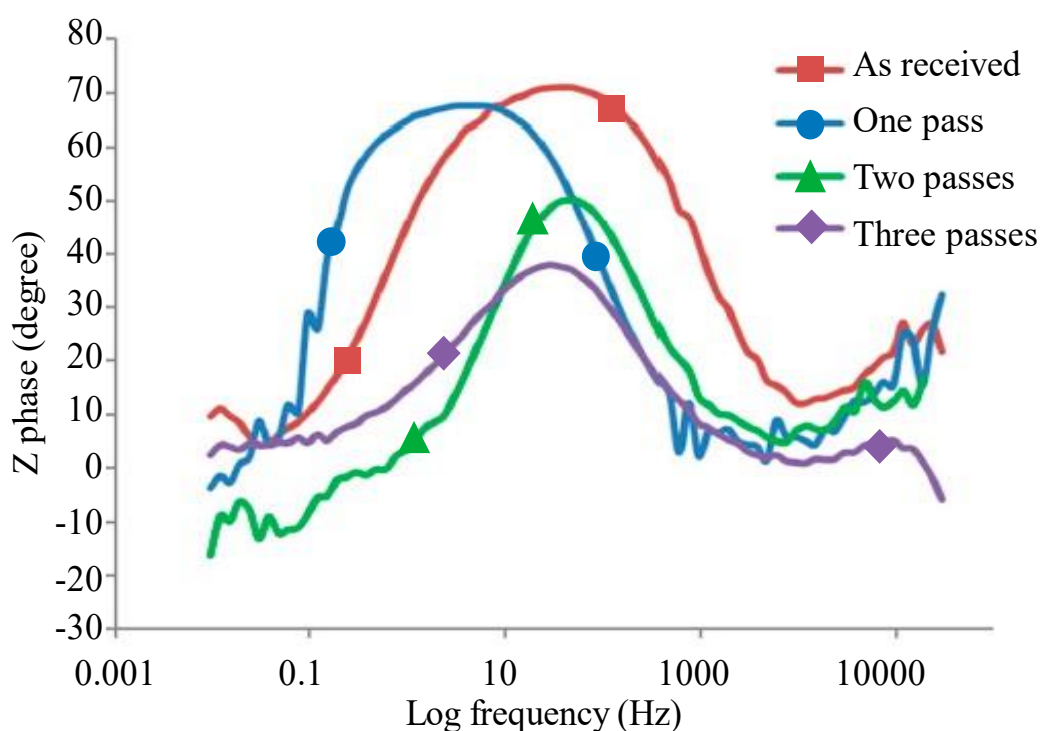


Figure 16. EIS Bode Angle Plot of the AA5083 in the 3.5% NaCl Solution [34].

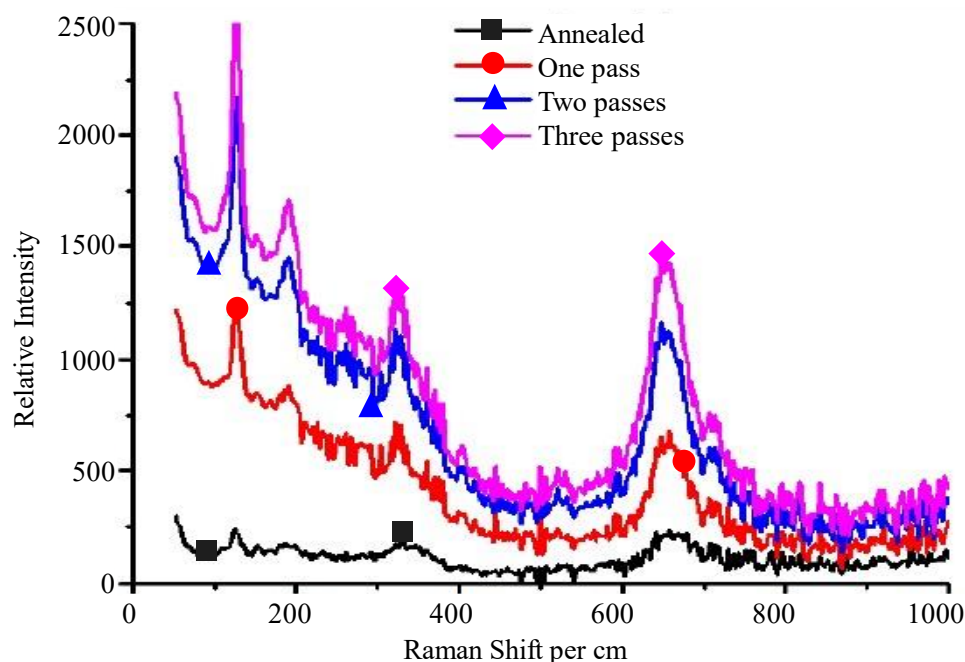


Figure 17. Raman Spectroscopy plot of the as-received AA5083 in Different Solution [36].

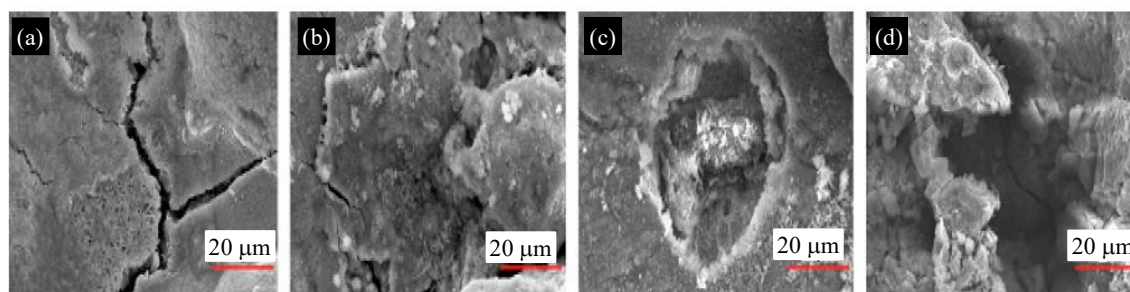


Figure 18. Under a Scanning Electron Microscope, the Corroded Surface of the AA5083 in 3.5% NaCl Solution was Captured in Pictures for the ECAP Samples following the (b) First, (c) Second, and (d) Third passes, as well as for the Annealed Samples (a) [37].

CONCLUSIONS

Utilizing thermal activation theory, a novel constitutive model was formulated to characterize the dynamic plastic deformation of AA5083 during the ECAP process. This model takes into account the interrelated effects of microstructure, temperature, and strain rate. The model's input parameters were identified using the Particle Swarm Optimization method, relying on experimental data obtained from uniaxial compression tests conducted on the experimental material across a broad range of temperatures and strain rates. When the newly created model's anticipated flow stresses were compared to the experimental findings, there was a good level of agreement. This illustrates that the proposed constitutive relationships in this study have a high level of applicability in replicating the dynamic plastic flow behavior of AA5083 during the ECAP process across temperatures ranging from 308 K to 588 K. The corrosion resistance of the ECAP samples revealed an opposite effect compared to the as-received sample, due to the grain refinement observed from the first to third passes. After the alloy was cryogenically processed using ECAP for two passes via route B_C, its original coarse grain structure was replaced with a highly refined and uniform microstructure. As the pass number increased, the microhardness of the alloy increased as well, peaking at 151 HV following cryogenic treatment as a result of the formation of fine grains. As the pass number increased, the alloy's tensile strength also saw a rise, albeit with a concurrent decrease in ductility attributed to the alloy's hardening.

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