Influence of thermally activated processes on the deformation behavior during low temperature ECAP

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Abstract. High strength aluminum alloys are generally hard to deform. Therefore, the application of conventional severe plastic deformation methods to generate ultrafine-grained microstructures and to further increase strength is considerably limited. In this study, we consider low temperature deformation in a custom-built, cooled equal channel angular pressing (ECAP) tool (internal angle 90°) as an alternative approach to severely plastically deform a 7075 aluminum alloy. To document the maximum improvement of mechanical properties, these alloys are initially deformed from a solid solution heat-treated condition. We characterize the mechanical behavior and the microstructure of the coarse grained initial material at different low temperatures, and we analyze how a tendency for the PLC effect and the strain-hardening rate affect the formability during subsequent severe plastic deformation at low temperatures. We then discuss how the deformation temperature and velocity influence the occurrence of PLC effects and the homogeneity of the deformed ECAP billets. Besides the mechanical properties and these microstructural changes, we discuss technologically relevant processing parameters (such as pressing forces) and practical limitations, as well as changes in fracture behavior of the low temperature deformed materials as a function of deformation temperature.

1. Introduction

Severe plastic deformation (SPD) methods allow generating ultra–fine grained microstructures. One of the most interesting approaches to obtain high-strength materials in combination with good ductility is equal–channel angular pressing (ECAP). Enhanced properties such as very high strength can be provided by grain refinement into the sub micrometer range by deformation in the solution heat-treated condition, in combination with a suitable (post-ECAP) heat treatment. The reduction of grain size in SPD processes is based on the formation of low–angle grain boundaries in regions of high dislocation densities, which are then further transformed into high angle grain boundaries. While ECAP of many metals and alloys is well understood today, there are some difficulties with ECAP processing of high–strength, age hardening aluminum alloys such as AA7075: Because of the poor formability, the material tends to shear localization and cracking when ECAP is performed at RT. Higher deformation temperatures (120 °C) do in principle allow a successful production of homogenous billets. However, because of dynamic recovery processes and the formation of precipitates during the deformation, the mechanical properties of the resulting materials are hardly improved compared to the conventional peak aged condition.
One very promising approach to improve the mechanical behavior of hard-to-deform fcc metals is deformation at low temperatures. It is well known that forming at low temperatures (or even in the cryogenic regime) results in high dislocation densities because dynamic recovery processes are suppressed. This leads to an increase of strain hardening capability, which in turn impedes strain localization and crack formation. One important microstructural aspect is that, because of the limited mobility of defects at lower temperatures, cross slip processes are more strongly hindered in the cryogenic regime. This also results in reduced annihilation processes of dislocations and in more strongly pronounced strain hardening stages III and V. As long as additional dislocations can be stored, the material exhibits strain hardening, which is related to stable plastic deformation. Another limiting factor for ductility is dynamic strain aging, which can be simply related to the interaction of dislocations with solute atoms. Materials that are prone to the nucleation of deformation bands (which are observed as serrated flow in the stress–strain curves and can be classified in different types: most well-known: types A, B, C) are generally more difficult to deform homogeneously. This so-called Portevin–Le Chatelier effect (PLC effect) by its very nature is a thermally activated process and therefore strongly depends on temperature and strain rate. Both – an increased strain hardening rate and a reduced (or suppressed) PLC effect – can be related to a better formability of fcc metals at low temperatures.

The most widely established process that directly exploits the beneficial effects of forming at lower temperatures is cryogenic rolling. The main advantage of cryogenic rolling is that it is a relatively simple procedure that can be performed in conventional rolling mills. In contrast to the accumulative roll bonding process, the deformation takes place in small incremental steps to insure a limited increase of temperature. Various research groups have uses rolling mills for forming at cryogenic temperatures in constant rolling direction [1–6]. The billets were typically cooled in liquid nitrogen, then transferred to the rolling mill, and subsequently deformed. This procedure of cooling and subsequent deformation has to be repeated for each deformation step. The relatively high velocity during the rolling process leads to an increase of the billet temperature, which negatively influences the homogeneity of the billet and the resulting mechanical properties. Nevertheless, by accumulating deformation in small incremental steps, true plastic strains about three have been achieved in aluminum alloys [3, 7]. A suitable heat treatment after the deformation results in a strong increase of the initial flow stress in combination with a somewhat increased ductility. Moreover, the increased dislocation density results in a pronounced acceleration of precipitation kinetics, which shifts aging time to much lower values.

The conventional process (constant rolling direction), however, adversely affects the useable residual volume of the billets: “uniaxial” rolling always results in very thin plates, which is a strongly limiting parameter for the use in technological applications. Other approaches for SPD of hard-to-deform age hardening aluminum alloys include multidirectional forging [8] and ECAP [9–13] at low temperatures. The main advantage of these processes is that the deformation conditions (dimensions of the billets, strain paths) can be carefully controlled. Especially ECAP is characterized by these positive features. Due to the constant cross section of the ECAP tool, and because the velocity and backpressure during the ECAP process can be adjusted, the deformation conditions can be exactly adapted to suit a particular alloy and large strains can be accumulated in a large useable (fully deformed) fraction of the total billet volume.

In the present study, the concept of ECAP at low temperatures is discussed in detail. We characterize the microstructure and the mechanical behavior of a coarse grained AA7075 alloy at different low temperatures. Furthermore we discuss how a tendency for the PLC effect and the strain hardening rate affect the formability during severe plastic deformation at low temperatures. We then discuss how the deformation temperature, pressing speed and backpressure influence the occurrence of PLC effects and the homogeneity of the deformed ECAP billets. Furthermore we characterize technologically relevant processing parameters and practical limitations, as well as changes in fracture behavior of the low temperature deformed materials as a function of the deformation temperature.
2. Experimental methods

For our investigations, the aluminum alloy 7075 in commercial purity (nominal chemical composition 5.9 Zn–2.3 Mg–1.3 Cu–0.2 Fe–0.1 Si wt.–%) as an extruded bar with a cross-section of 15 x 15 mm² was used. The material was solution heat treated at 475 °C for 2.25 h and then quenched to RT in water. To suppress natural aging processes, all specimens were immediately stored in a freezer after solutionizing. For mechanical characterization, tensile tests were performed in the solution heat-treated condition under quasi–static loading with an initial strain rate of $10^{-3}$ s⁻¹ (free gauge length and diameter of the cylindrical samples: 10.5 mm and 3.56 mm, respectively) at different temperatures (RT, -60 °C, -120 °C and -196 °C). Furthermore, we performed jump-tests (uniaxial tensile testing with a sudden change of strain rate of about one order of magnitude) in the range from $10^{-4}$ s⁻¹ to $10^{-1}$ s⁻¹ to characterize the strain rate sensitivity as a function of testing temperature. For statistically relevant data, three samples for each condition were tested. The investigations were performed in a mechanical tensile testing machine (Zwick/Roell) with a maximum force of 20 kN. To determine the strain values, digital image correlation systems (DIC) were used for RT measurements. Low temperature conditions were generated by different experimental methods: Testing at -60 °C was performed in a climate chamber installed at the testing machine. For mechanical testing at -120 °C and -196 °C, liquid nitrogen was filled in a special double–ring cooling device placed around the tensile specimen. Strain values at low temperatures were determined from the crosshead displacement data.

The material was deformed at -60 °C in a custom-designed ECAP tool that can be cooled to low temperatures and that includes moveable walls to minimize friction. More detailed information on the ECAP die can be taken from our earlier publication [9]. The equivalent plastic strain per ECAP pass was 1.1 [14]. ECAP was performed with varying pressing speeds (between 10 and 50 mm/min⁻¹); a backpressure of 80 bar was applied to prevent shear localization or cracking. To further reduce friction, the billets were covered with a lubricant (molybdenum disulphide, Aero Shell Grease 33MS) prior to forming.

The ECAP deformed billets and the tensile specimens were analyzed by optical microscopy (on the longitudinal plane) to characterize strain localization events due to the PLC effect. The samples were embedded in synthetic resin under vacuum at RT to avoid recrystallization and aging processes during sample preparation. After grinding and polishing, the samples were etched for 40 s at RT in an acidic solution (Graff-Sargent: 84 ml H₂O; 15.5 ml HNO₃; 0.5 ml HF and 3 g CrO₃). Optical micrographs were taken in phase contrast mode. The investigations of the fracture surfaces were performed with a field emission scanning electron microscope (Zeiss NEON40). The acceleration voltage was 5 kV with an operating distance of 4 mm. Besides the detectors for secondary and back-scattered electrons, an in–lens detector for enhanced contrast was used.

3. Results and discussion

As reported in our earlier publications [9, 15], deforming the solution heat-treated condition by SPD processes provides some advantages for potential improvements of the mechanical properties compared to ECAP of the conventional peak-aged condition (i.e., solution heat treated and aged to maximum hardness, T6). The solutionized alloys are free of precipitates and exhibit an improved ductility. This results in a much higher formability which is mandatory for SPD processes. Large amounts of plastic deformation lead to the formation of dislocation networks and (sub–)grain boundaries, which then also offers a possibility for the nucleation of precipitates. The concentration of dislocations is increased by increasing plastic strains. This provides a higher number of nucleation sites, which allows a more homogeneous formation of finely distributed precipitates during subsequent artificial aging. Furthermore the aging times are significantly shortened by the accelerated precipitation kinetics in ultra–fine grained microstructures [2, 3, 5].

Due to the high amount of foreign alloying elements in the investigated aluminum alloy, there is a strong susceptibility to deformation localization by the PLC effect [16].

Figure 1 shows the mechanical properties of AA7075 in the annealed condition as a function of testing temperature. The engineering stress–strain curve measured at RT clearly shows serrated flow and thus indicates occurrence of the PLC effect. A reduction of the testing temperature to -60 °C results in an
increase of ductility (elongation to failure) of about 35%. Lowering the temperature to -120 °C further increases ductility by ten percent and at -196 °C, elongation to failure is twice as high as at RT. Moreover, the strain range where the discontinuous flow is observed becomes much smaller at lower temperatures. While some effects of the PLC effect can still be observed at -60 °C, this behavior no occurs at -120 °C or at -196 °C. The uniform elongation is almost constant from RT down to -120 °C, but it exhibits a significant increase when tensile testing is performed at -196 °C. Tensile strength (as expressed by 0.2 % yield stress) also markedly increases from -120 °C to -196 °C, which can be related to less thermally activated cross slip of screw dislocations, which is the primary deformation mode in aluminum alloys [2, 3, 9, 17].

![Figure 1](image)

**Figure 1.** Engineering stress–strain curves of AA7075 under quasi–static loading at different testing temperatures. A reduction of testing temperature also leads to less serrated flow.

In figure 2, we present the strain hardening rate $\theta$ as a function of true stress as determined from the stress-strain curves measured at different temperatures. The raw data in figure 2a exhibit some scatter which results from the numerical calculation of the first derivative of the stress-strain data and which is particularly pronounced when serrated flow occurred. For a more precise presentation, these curves were therefore corrected using a smoothing function (figure 2b). While a decrease of testing temperature clearly leads to an improved ductility (see figure 1), the (almost) constant strain hardening rates between RT and -120 °C do not directly indicate a better formability at lower temperatures. Only when the testing temperature is reduced to -196 °C, the reduction of thermal activation results in a significantly higher strain hardening rate. From the data presented in figure 2, it is difficult to distinguish between strain hardening stages III and IV. This indicates that the characteristic microstructural processes – dynamic recovery (stage III) and interaction of an increasing amount of dislocations with (sub-)grain boundaries (stage IV) – take place simultaneously. At RT and at -60 °C, the tensile samples failed without reaching the conventional Considère criterion for the onset of necking ($\theta \geq \sigma_W$) [18], whereas at -120 °C and -196 °C, strain hardening stage V was reached and $\theta$ values fell below the Considère limit. This indicates that the sudden failure without significant necking (at RT and -60 °C) and the occurrence of a high strain hardening rate are related to dynamic strain aging. When local deformation in a PLC band exceeds the Considère limit, apparently premature failure of the material is observed from a macroscopic point of view.
Figure 2: a) Full data and b) smoothed data of strain hardening rate as a function of true stress of AA7075 at different testing temperatures. The strain hardening rate curves are quite similar from RT down to -120 °C. Lower testing temperatures (cryogenic conditions: -196 °C) result in a marked increase of strain-hardening rate.

These results on the temperature-dependent mechanical properties of the age hardening aluminum alloy illustrate the important influence of the PLC effect. The formability of this alloy can be strongly increased by the suppression of dynamic strain aging at low temperatures. To further illustrate the increase of ductility with a decrease of deformation temperature from RT to -196 °C, optical micrographs (from the center and at the fracture surface of the tensile specimens) are shown in figure 3. Traces of deformation bands are apparent (from the deformation and the local displacement of large impurities that are present throughout this alloy and that are quite visible in differential interference contrast) at an angle of about 45° with respect to the loading direction (from left to right in all micrographs). These bands were also visible on the sample surfaces after testing at RT. In figure 3, these regions, where local shear deformation obviously exceeds the nominal one, have been artificially brightened for better visibility. Interestingly, in all cases where such deformation bands were present, fracture surfaces were oriented in parallel: because locally the critical shear stress was exceeded, shear failure occurred without necking at RT (figure 3b). Furthermore, parts of severed PLC bands were detected at the fracture surfaces, which confirmed the failure behavior within one band. This behavior is in a good agreement with a previous report by Chung et al., who described the local shear band formation as the primary failure mechanism of AA7075 at RT, [19].
Figure 3. Optical micrographs in differential interference contrast of the solution heat-treated condition of AA7075. The micrographs were taken in the central section of the tensile specimens (orientation of the polished sections shown here: parallel to the deformation axis, which is from left to right in all micrographs) and in the region of fracture. PLC bands can be identified by analyzing the local displacements of the coarse impurities; the corresponding regions have been artificially brightened. a) and b) deformation at RT; c) and d) deformation at -60 °C; e) and f) deformation at -196 °C. Lower testing temperatures result in a transition from brittle (RT) to ductile (-196 °C) fracture.
Figure 4. SEM micrograph of the fracture surface in solution heat treated condition of AA7075 with a transition from smooth shear fracture a) and b) at RT; to a mixture between brittle and ductile fracture c) and d) at -60 °C; to ductile fracture with dimples on the fracture surface e) and f) at -196 °C.

After deformation at -60 °C, PLC bands were no longer visible on the surface of the samples even though the microstructure showed a (somewhat lower) number of bands (figure 3c). Deformation at -60 °C results in smooth necking and the formation of shear fracture surfaces with a slightly elevated roughness (figure 3d). Crack nucleation of the samples occurred simultaneously along several PLC bands. This results from the overall reduction of the PLC effect and an increased ductility. Despite a significant reduction of serrated flow, the stress-strain behavior was considerably influenced by localization events (figure 1). This leads to an irregular microstructure and a premature failure along the deformation bands. Finally, cryogenic conditions lead to a distinct transition from brittle to ductile fracture; shear band fracture is suppressed (figure 3f), which agrees with literature [19]. At -196 °C the
PLC effect does no longer occur and deformation bands were not observed (figure 3e). These observations clearly demonstrate that suppressing dynamic strain aging (by strongly reducing thermal energy and thus retarding thermal activation), leads to homogenously deformed microstructures.

The fracture surfaces of the tensile specimens were also investigated in greater detail by SEM, figure 4. The fracture surface of the tensile specimens that were tested at RT indicates a shear fracture behavior (figure 4a and b, taken at different magnifications). A smooth fracture surface was formed by deformation along a PLC band. At -60 °C (figures 4c and d), dynamic strain aging was reduced and a partial transition from brittle to ductile fracture was observed. Typical dimple structures can be observed on the fracture surfaces. At -196 °C, the fracture surface contains fine dimples and indicates a completely ductile fracture (figure 4e and f), which agrees well with the strongly increased ductility observed in the corresponding stress-strain curves.

Figure 5. a) Influence of aging time (10 min to 240 min) on the onset of serrated flow at RT. The first occurrence of PLC bands is indicated by arrows. Longer aging times shift these to higher strains. The flow stress can be increased by increasing the aging time. b) Critical stresses for the onset of PLC effects as a function of testing temperature and for different natural aging times.

Besides the deformation temperature, one further important parameter that has an effect on the occurrence of PLC effects is the formation of precipitates after solutionizing. Depending on the aging time, these precipitates can already be formed at RT. The early formation of Guinier-Preston (GP) zones is strongly promoted by the high amount of copper in AA7075, which can lead to an immediate nucleation of GP zones after the solution heat treatment at RT. The influence of natural aging on the occurrence of PLC effects can be analyzed by considering tensile tests (performed at RT) of differently aged samples (figure 5a): All samples were naturally aged at RT for various times (10, 30, 60 and 240 min). With increasing aging time, strength increases with a simultaneous decrease of ductility; both can be related to the formation of GP zones. In addition it is obvious that natural aging has a strong influence on the occurrence of the PLC effect: The initial points of serrated flow (i.e., the first points where serrated flow could be clearly identified in the stress-strain curves; these points are marked by arrows in figure 5a) are shifted to higher stress/strain values with increasing aging time, which is in a good agreement with literature [20]. Natural aging for 10, 30 and 60 min first results in the presence of PLC type A. Afterwards, the curves exhibit a PLC type C character, including a subsequent crack initiation and failure of the specimens. After natural aging for 240 min, the fracture was exactly initiated when the first PLC bands occurred. A further increase of aging time resulted in a sufficient formation of GP zones, which suppress strain localization because a higher amount of GP zones leads to a decreased number of point defects [20].
Thermal activation also plays an important role in determining the occurrence of PLC bands; this is obvious from figure 5b, which summarizes the initial stress values of serrated flow as a function of testing temperature and for samples aged naturally for different times. As known from the literature the maximum susceptibility to PLC effects is about one third of the melting temperature [21]. Especially for the AA7075 alloy our investigations indicate a critical value of about -20 °C, which is in good agreement with the literature. Below RT, all samples showed the PLC type A. Even without natural aging, tensile testing below -60 °C was not affected by PLC effect. At this temperature and strain rate (1·10\(^{-3}\) s\(^{-1}\)), diffusive processes are slowed down enough to fully prevent dynamic strain aging.

Dynamic strain aging, as a thermally activated process, depends on the deformation temperature as well as on the strain rate. In order to further characterize the influence of PLC effects on the strain rate sensitivity, we also consider the results of the jump-tests. Figure 6 shows the stress–strain behavior of samples deformed at RT (figure 6a) and at -196 °C (figure 6b), respectively; both samples were first solution heat-treated and then naturally aged for 10 min. At RT, an increase of strain rate leads to a stress drop (by a factor of \(\Delta \sigma\)), which can be described as a negative strain rate sensitivity. This stress drop is directly related to the PLC effect (i.e., formation of a PLC band), and it is further increased at higher strain rates. In contrast, the amount of \(\Delta \sigma\) remains constant (and positive) for every change in strain rate at -196 °C (figure 6b). At this low temperature, we clearly observe positive strain rate sensitivity. These results fully agree with our previous observations (figures 1, 5) on the effect of testing temperature and strain rate on the occurrence of PLC bands: localization of deformation is associated with insufficient hardening and/or negative strain rate sensitivity. By lowering the testing temperature, both hardening and strain rate sensitivity can be increased and therefore the PLC effect can be suppressed, which in turn is favorable with respect to ductility and failure during tensile testing (figures 3 and 4).

![Figure 6. Influence of testing temperature and strain rate on the mechanical behavior under tensile loading at a) RT and b) -196 °C. The change of strain rate (by one decade in each jump) is marked by dotted lines. The yield stress decreases (by \(\Delta \sigma\)) with increasing strain rate at RT. Because of the suppression of dynamic strain aging, this effect is reversed at low temperatures.](image_url)

Based on our fundamental analysis of the effect of testing temperature on mechanical behavior and, particularly, on the occurrence of localized deformation related to PLC bands, we now discuss, from an engineering point of view, the influence of thermally activated processes on the possibility to manufacture homogenously deformed billets by ECAP at low temperatures. In figure 7, we present pressing forces that were recorded during ECAP at -60 °C as a function of the displacement of the plunger (and of the billet while passing through the ECAP die, respectively) with backpressure (80 bar, as controlled in the hydraulic cylinder; note that the actual mechanical compressive stress at the billet is different from this value: 176 MPa), or without backpressure, and with different pressing speeds range-
ing from 10 to 50 mm/min. The data in figure 7a correspond to measurements performed while ECAP processing the AA7075 alloy in solution heat-treated condition, while the data in figure 7b were measured, as a reference material, using an AA6082 alloy in peak aged condition. Because of the chemical composition and because we used the peak aged condition of the 6xxx alloy, no PLC effects could occur during deformation. A comparison of the pressing force data for both alloys demonstrates that fluctuations in the pressing forces (as observed for the AA7075 alloy) cannot be simply assigned to stick–slip effects (and hence friction within the ECAP die), but clearly must be related to shear localization in the ECAP shear plane when deforming this alloy. The strain rate in the ECAP shear plane for our setup (cross section: 15 x 15 mm²) is in the range from $10^{-2}$ s$^{-1}$ to $10^{-1}$ s$^{-1}$ (for the pressing speeds considered here), which allows a careful comparison with the data from our jump-tests. Because of the multi-axial stress state associated with the ECAP process, the results given in figure 6 cannot be directly compared. However, the principal trends in terms of strain rate and testing temperature can be correlated with the processing data to estimate limitations and useful temperature/ strain rate regimes for low temperature ECAP.

Figure 7. Pressing forces during the ECAP process (first ECAP pass) with and without backpressure depend on the deformation speed: a) AA7075 in solution heat-treated condition and b) AA6082 in peak aged condition (T6). Fluctuations of the pressing force can be related to dynamic strain aging in AA7075.

Figure 7 shows that, for both materials, the pressing forces are higher when an active backpressure is applied and lower when no backpressure is used. In the ECAP experiments on AA6082, the pressing forces were hardly affected by the deformation velocity (figure 7b). Because of the coarse grained condition deformed during this first ECAP pass, the material was not prone to shear localization. In contrast, the pressing forces for the deformation of AA7075 (figure 7a) did depend on the deformation velocity, as is expected when dynamic strain aging dominates the deformation behavior. When (without backpressure) the deformation velocity was increased from 10 mm/min to 30 mm/min (which is equivalent to a shear zone strain rate of about $10^{-2}$ s$^{-1}$), pressing forces continuously increased. At a velocity of 40 mm/min, the pressing forces were nearly the same as at 30 mm/min, but the amount of serrated flow (as evidenced by the fluctuations in the pressing force curves) was considerably decreased. At 50 mm/min (corresponding to a strain rate in the shear zone of about $10^{-1}$ s$^{-1}$), both the amount of serrated flow and the pressing forces decreased, which agrees well with the reduction of the flow stress and the reduction of dynamic strain aging observed at the corresponding strain rates in the jump-tests (figure 6a).
Figure 8: Optical micrographs (differential interference contrast) of the ECAP billets deformed at -60 °C at a) 1 mm/min, c) 20 mm/min, e) 50 mm/min, and with an additional backpressure (80 bar) at the similar velocities b), d) and f). The displacements of the coarse impurity phases indicate shear localization.

Similar observations hold when one considers the measurements during ECAP with an additional backpressure; because of the multi-axial stress-state in the shear zone, the maximum susceptibility for the PLC effect was shifted to a slightly lower deformation velocity (40 mm/min). This analysis of pressing forces as a function of deformation velocity allowed to define useful deformation parameters for successful ECAP processing at low temperatures: a relatively low temperature (-60 °C) with a medium strain rate (10⁻² s⁻¹) provides optimum conditions for a suppression of the PLC effect.

Finally, the occurrence of shear bands – as directly related to the deformation velocity – can be confirmed from optical micrographs of the ECAP billets (figure 8; all billets were deformed at -60 °C). After ECAP processing at 1 mm/min, a high amount of (relatively broad) shear bands was observed.
(figure 8a). By applying an additional backpressure, the number of these bands can be clearly reduced (figure 8b). Increasing the deformation velocity to 20 mm/min resulted in a strongly decreased amount of shear bands (figure 8c) and backpressure leads to smaller widths (figure 8d). The strongly increased amount of serrated flow at 50 mm/min documented in figure 7a fully agrees with the corresponding microstructure (figure 8e). This velocity also has a negative effect on the production of homogeneously deformed billets, which can be inferred from the wavy displacement of the coarse impurities. While this phenomenon can be compensated by using an active backpressure, the number of shear bands cannot be significantly decreased (figure 8f). The key result of these first microstructural observations on ECAP billets deformed at low temperatures is that billet homogeneity and the material’s tendency to form shear bands can be directly related to the uniaxial mechanical properties studied in greater depth in this paper.

4. Summary and conclusions
We have studied the influence of thermally activated processes, i.e., the role of strain rate and temperature, on the workability and mechanical properties of the high strength aluminum alloy 7075. Due to the increased strain hardening capability of this alloy, low deformation temperatures result in an increased ductility and improved workability. Moreover, we studied how these improvements benefit from the reduction of dynamic strain aging by suppressing the PLC effect at low temperatures. A deformation temperature of -60 °C leads to a strongly decreased amount of shear bands (resulting from PLC effects) and to a sufficient ductility. Our results also show that natural aging at room temperature has a strong influence on the occurrence of dynamic strain aging. By increasing the natural aging time, PLC effects can be reduced. This effect is related to the formation of GP zones, which shift the strength of the material to higher values and lead to increased pressing forces during the ECAP process. To verify our observations from uniaxial testing at different temperatures and strain rates, ECAP was performed with varied deformation velocities at -60 °C. Processing at 20 mm/min, with the use of an active backpressure, was found to be appropriate to produce homogenously deformed billets.

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