A high temperature nanoindentation study of Al-Cu wrought alloy

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Abstract

Aluminum-copper alloys are widely used because of their low density and good mechanical strength accomplished with precipitation hardening. The alloy Al-Cu-Mg-Pb (AA2030) has been investigated before and after aging, at room temperature and at high temperatures. The mechanical properties at room temperature have been studied by Brinell hardness tests. T4 and T6 stages of the alloy have been investigated by differential scanning calorimetry up to 450 °C, showing the precipitation of different clusters. High temperature nanoindentation has been used to characterize the mechanical properties up to 460 °C in order to obtain a better understanding of the local dominating deformation mechanism in the material. A continuous decrease in the hardness and young modulus was found with the increasing temperature. The strain rate sensitivity of the alloy increased with the temperature from 0.022 at RT, up to 0.16 at 460 °C. The activation volume was constant (around 31-41 b^3) up to 240 °C, beyond this point a large increase was observed up to 178 b^3 . The results were comparable with similar materials, and indicate thermally activated processes.

Keywords

Mechanical characterization, nanoindentation, aluminum alloys, age hardening,

1. Introduction

Lightweight materials solutions are of economic and environmental interest for the automotive and aerospace industry. Therefore, aluminum alloys are considered in new automotive and aviation developments due to their high strength-to-weight ratio, recyclability, good formability, and reasonable creep resistance [1]. To use these materials in structural applications, it is indispensable to evaluate the thermal and mechanical properties, such as, age-hardening, strain-rate sensitivity and activation volume, to understand the underlying mechanisms of material processing, especially for machining.

Free-machining alloys are materials characterized by their optimized machining characteristics. They are able to form short, easily disposable chips, and a high quality surface finish [2]. Standard aluminum machining alloys are found within the 2xxx and 6xxx series alloys which contain lead (Pb), and, sometimes, bismuth (Bi) in order to improve the free-cutting behavior. The microstructure of these systems have previously been characterized by various experimental methods such as optical microscopy, scanning/transmission electron microscopy, microradiography, electron probe micro analysis, x-ray diffraction [3–9], calorimetry [9–11], and positron annihilation [12–14]. It has been shown that the alloying system consists of (i) low-melting element inclusions, (ii) hardness increasing precipitates, and (iii) aluminides [9].

Differential scanning calorimetry (DSC) is a common technique to study the thermodynamics and kinetics of phase changes in materials and is particularly useful to investigate precipitation reactions. Nanoindentation at high temperature is a well-suited method to obtain mechanical properties with high lateral and depth resolution [15,16]. The rate dependent deformation mechanisms can be locally obtained by strain rate jump-tests, relaxation tests, or creep experiments [17-20]. There are several characteristic properties of a material that can be used as a description of the time and strain rate dependent mechanisms inside the material. To understand this mechanism is of practical interest, for instance in metal forming, where the material is subjected to low to moderate strain rates, and during collisions in automotive applications or during machining, where high strain rates occur [21]. The alloy investigated here is used as a free-cutting material.

The strain rate sensitivity (SRS), which controls the ductility of materials, and the activation volume (V^*), which is defined as the separation distance between points of dislocation intersection [21], are two of the important properties which are needed to characterize the materials' responses. In the last year, several publications were written to determine SRS at room temperature (RT) (25 °C) of nanocrystalline Al [22], ultrafine grain Al [23], and Al alloys [24-27]; however, few studies could be found on high temperature nanoindentation except AlSi alloy [28] and ultrafine grain Al [23], but none were found for Al-Cu-Mg alloys.

In this work, hardness, Young modulus, SRS, and V^* were determined by nanoindentation and creep tests at high temperatures in order to obtain a better understanding of the dominating deformation mechanism on a local level in the material.

In this study, the nanoindentation response of the named Al-Cu-Mg alloy at various testing temperatures is compared with both the macro hardness measurements and the thermoanalytical technique. The combination of DSC analysis, macroindentation, and nanoindentation should provide a more complete picture of the materials' behavior.

2. Materials and Methods

The materials used in this study are industrially produced T4 tempered rods of AA2030 and the lead-free AA2024 alloys. The average chemical composition of the stock material in rod shape was determined by optical emission spectroscopy (SPECTROMAXx from SPECTRO) and is shown together with the composition specifications of The Aluminum Association, Inc. [29] for named alloy in Table I.

The precipitation hardening process was studied on a macroscopic scale by various heat treatments. The material was solution heat treated (specimen size is 10 x 10 mm, thickness of 3 mm) at 490 °C, a common temperature for the industrial production of AA2xxx series alloys. This was followed by press quenching in water at RT and aged either at RT (in the case of T4) or artificially aged in a heating thermostat (LAUDA Eco E25G) at 180 °C (T6).

Brinell hardness measurements at RT were performed on those samples using an EMCO-Test M4 unit, with a hard metal ball of 2.5 mm diameter, and a force of 613.13 N with a 15 s loading time (HBW 2.5/62.5/15). Each hardness value reported in this work represents the mean value of at least four measurements. A maximum standard deviation of 2 HBW was achieved. To characterize the thermal behavior of T4 and T6 aged Al-Mg-Cu-Pb in comparison to Al-Mg-Cu

samples with a mass of 10 ± 0.1 mg were measured using a NETZSCH DSC 204 F1 Phoenix, scanning a temperature range of RT to 450°C at 15 K/min, under a nitrogen atmosphere with a 20 mL/min flow rate. The heat flow (DSC curve) was recorded simultaneously as a function of temperature. Both Brinell hardness and DSC measurements are focusing on the macro scale properties and samples over several grain and microstructural features.

High temperature nanoindentation measurements were carried out on a MicroMaterials NanoTest XP System (UK). The measurements at higher temperatures were performed using a heatable Berkovich cubic boron nitride (cBN) indenter tip. The area function was measured before and after measurements at elevated temperatures on a fused silica reference sample with current analyzing methods. The nanoindentation measurements were performed load controlled, with a load time of 10 s and an unload time of 5 s. A constant load of 12 mN was used for the indentation. The distance between the indents was set to 50 μ m. Small samples (50 mm x 50 mm x 3 mm) were solution heat treated at 490 °C, water quenched, and then mounted at the sample stage of the nanoindenter. The measurements were performed at RT, 180, 240, 400 and 460 °C with a rate of 1.6 °C per second to the target temperature at which point the sample was thermalized for 60 minutes to ensure an isothermal contact between sample surface and indenter tip [30].

At each temperature a minimum of eight indents for each test condition were made. High temperature experiments were carried out in an Argon (Ar) environment, where the maximum oxygen content was measured to be < 1 %. The thermal drift correction was carried out on a load level of 10 % of the maximum load, whereas the final 60% of the data in the dwell period for drift correction was used for the drift rate calculation. The drift rate did not exceed 0.40 nm/s. The hardness (H) and reduced modulus (E_r) as a function of temperature, obtained from the nanoindentation experiments, were calculated following the well-known Oliver and Pharr Method [31].

3. Results and Discussion

3.1. Aging of the samples and hardness testing

Figure 1 shows the change in hardness (measured with Brinell) of the Al-Cu-Mg-Pb alloy as a function of aging time, compared to AA2024 data from literature [32,33] at two different aging temperatures (RT and 180 °C). Since the literature data were available as Vickers hardness (HV)

and Rockwell hardness (HRB), their values were converted to Brinell hardness values according to EN ISO 18265:2003 Table F.5 [34].

Hardness values, measured after quenching at RT (addressed as T4 temper in this manuscript and referring to the heat treatment temper designations for aluminum and aluminum alloys in ANSI H35.1-2004), indicate a linear increase to peak hardness after a short incubation time within the first day (Figure 1a). This is followed by a long-time stable plateau of ~96 HBW for the Al-Cu-Mg-Pb alloy, compared to ~109 HBW of the reference AA2024 alloy [32].

The corresponding hardness behavior at the common artificial aging temperature of 180 °C, is plotted in Figure 1b and referred to in the following as ‘T6’ heat treatment. Figure 1b shows the typical three step hardening profile for both alloys, which is in agreement with previous work [17]. Here, the Al-Cu-Mg-Pb alloy reaches higher peak hardness values, ~125 HBW, compared to the natural ageing procedure. The reference material, AA2024, reaches even higher hardness levels, 133 HBW [33], at similar artificial ageing practice. It is well known that the main factors which influence the age hardening behavior are composition, plastic deformation and heat treatment technique [35]. Since the last two factors are the same for both alloys, the reason for the difference in the hardness level may only be found in their chemical composition.

According to Mondolfo [35] a change in the Cu:Mg ratio has limited effect within the compositional ratio range of 4:1 to 2:1. This ratio also applies to the alloy studied here. Since higher contents of Mg form a compound with Pb (Mg_2Pb) less Mg is available for the solid solution hardening response [35], so the Cu:Mg ratio is shifted to higher values, and the main hardening phases would become Al_2Cu [35]. This could be the reason for the different hardness values of the two compared alloys. Present authors have found similar behavior of Sn, forming Mg_2Sn particles in an Al-Cu machining alloy [7]. Other authors confirmed the occurrence of Mg_2Pb in machining aluminum alloys by XRD [36,37] and DSC [38].

3.2. DSC

DSC traces obtained at 15 Kmin⁻¹ from the investigated alloys in the RT and artificial temperature aged condition are shown in Figure 2. Figure 2a shows the DSC curves of the Al-Cu-Mg and the Al-Cu-Mg-Pb alloy at 15 Kmin⁻¹ heating rate obtained from fully naturally aged (T4) samples.

Herein five main effects, A–E, as suggested by Jena *et al.* [39] may be identified: (A) the small exothermic peak between 50 and 80 °C associated with Guinier Preston Bagaryatsky's (GPB) zone precipitation/atom clustering; (B) an endothermic effect between 160 and 180 °C, attributed to GPB zone/atom cluster dissolution; (C) an endothermic effect due to GPB2/S'' dissolution, between approximately 200 and 250 °C; (D) the large exothermic peak associated with S' (S) phase precipitation, between 250 and 350 °C; and (E) a broad endothermic effect due to the progressive S phase dissolution, at 340–500 °C. Where S''/S' are the metastable pre-cursors of the equilibrium Al₂CuMg phase (S). Similar results are proposed in other literature as well [40–42].

Comparison of the DSC curves obtained from T4 with those from T6 aged samples indicates that the larger endothermic peak (C) followed by a smaller exothermic peak (D) suggests that more precipitates pre-exist within the matrix of the T6 aged sample [43,44]. Figure 2b, showing the T6 aged samples, reveals that two exothermic peaks appear between 240 and 350 °C in both alloys. They are caused by the formation of two distinct types of S phase precipitates, Type I and Type II, respectively. The first peak (D1) is due to the formation of Type I precipitates and the second peak (D2) is due to the formation of Type II precipitates [40].

3.3. Mechanical behaviour at high temperature

3.3.1. Hardness and creep

A study on the mechanical properties of the AA2030 alloy, has been carried out by nanoindentation at high temperature. Figure 3a shows representative load-displacement curves obtained at different temperatures. All the indents were performed under the same conditions. As the test temperature is raised, the indentation depth increases for the same applied maximum load, indicating that the material softens with temperature. Similar curves showing the behavior of a similar Al alloy is shown in Chen *et al.* [28] at different temperatures, and similar curves at RT have been found experimentally and simulated in AA2024 alloy by Li *et al.* [45]. Indents performed at higher temperature sink deeper than indents at lower temperature.

Observing the part of the graph with a constant load, the significant differences of the creeping behavior at different temperatures can be seen. Figure 3b shows representative curves of nanoindentation depth from the creep test as a function of the creeping time. It can be seen how the tests carried out over a wide temperature range increase the creep rate (displacement of the indenter into the material). The tests carried out at 420 and 460 °C showed a significant creep of

600-800 nm over 200 s dwell time at constant load. No significant differences can be found between the curves below 420 °C testing temperature. For all the curves, it was observed that after an initial penetration of the indenter into the sample a short primary creep (Stage I) can be seen followed by a longer and steady state secondary creep stage (Stage II). The Stage I and II creep shown in Figure 3b will be explained in more detail later.

Since the indenter continues sinking into the material over the dwell period, one can calculate hardness associated with the depth change, as shown in Figure 4a. The Oliver and Pharr [31] method is used to calculate the hardness at each depth during the dwell time period. Table II shows the mechanical properties obtained by nanoindentation up to 460 °C. In Figure 4b the hardness values at the beginning (H_{ini}) and the end of the dwell period (H_{creep}) are shown. During typical RT indentation testing, one usually applies a 5 or 10 s dwell period, before unloading and indent depth evaluation, to allow the material to reach its maximum plastic deformation at a given load. Typically no significant difference is found between the hardness calculated from the beginning and end of the dwell period. However, at high temperatures where creep underneath the indenter is a concern it is appropriate to report both hardness numbers at the end and at the beginning of the dwell period [15].

The hardness and Young modulus of the Al-Cu-Mg-Pb alloy as a function of temperature (T) obtained from the nanoindentation experiments are shown in Figure 4b. The values of the hardness after the creep tests (H_{creep}) and the calculated hardness at the beginning (H_{ini}) of the creep are shown on the plot. The same trend is observed for both hardness values – a continuous decrease with the testing temperature. However, the percentage change on the hardness before and after the creep is significantly different at different temperatures (Table II). Meanwhile tests carried out up from RT to 240 °C show a decrease from 11 to 36 %; for the test carried out over 400 °C, values of 58-59 % are observed. This significant decrease in hardness is associated with the long creep observed in the tests at higher temperatures.

McLelland and Ishikawa studied the elastic properties of polycrystalline Al where a similar trend was found in the Young modulus and shear modulus [46]. The authors found a continuous decrease in the elastic properties up to 377 °C, while a larger decrease rate was observed at temperatures up to 600 °C. The initial hardness of the sample is 1.41 GPa at RT, and it decreases to 1.14 GPa at 180 °C and 0.82 GPa at 240 °C. The test carried out at 420 and 460 °C shows values of 0.26 and 0.19 GPa, respectively. Similar values have been found by Chen *et al.* of 1.45, 0.68, and 0.14 GPa for RT, 200 and 350 °C, respectively, for aluminum measured under similar

conditions [28]. Wheeler *et al.* have shown a decrease from 0.72 GPa at RT to 0.35 at 250 °C for ultrafine-grained (ufg) Al measured under similar conditions [23].

The Young modulus was also plotted in Figure 4b. A decrease from 84.23 to 18.6 GPa is observed in the range of measured temperatures. However, just a small decrease is observed from RT to 240 °C, with a significant decrease for the tests carried out at higher temperatures. Similar values of reduced modulus have been shown by Chen *et al.* [28] with a decrease from 84.6 to 27.5 GPa, with an increasing temperature from RT to 350 °C; and Maier *et al.* [19] with a decrease from 74.7 and 74.2 to 62 and 66 GPa for coarse-grained (cg) Al and ufg-Al up to 200 °C.

3.3.2. Strain rate sensitivity (SRS) and activation volume (V^*).

The SRS was determined by the constant load (CL) method following a description by Peykov *et al.* [18]. The SRS, also referred to as the m -value, is given by Eq. (1) as the ratio between the derivatives of natural logarithmic hardness ($\ln H$) and natural logarithmic strain rate over time ($\ln \dot{\epsilon}$), with the temperature (T) remaining constant:

$$m = \left(\frac{\partial \ln \sigma}{\partial \ln \dot{\epsilon}} \right)_T \approx \left(\frac{\partial \ln H}{\partial \ln \dot{\epsilon}} \right)_T \quad (1)$$

In Table II the SRS values for *Stage I and II* following the description by Peykov *et al.* [18] are shown. Both m_I and m_{II} show the same trend, a continuous increase from RT up to 460 °C. However, in this paper, we will only consider the values from m_I , after Peykov's conclusions. Figure 5 shows a plot of m_I obtained in this work as a function of the temperature, with other values obtained by a similar methodology on similar materials [18,19,23,25,26,30,47].

After comparing the SRS data obtained in this work with the results of Khan *et al.* [21] for a 2024 aluminum alloy, the authors observed no big differences between them. Even the comparison of current SRS results with slightly different alloy compositions, i.e. Romhanji *et al* [48], show that the values still lie in the same small region. Therefore, it is believed that the differences between the Pb-free 2024 and the Pb-containing 2030 aluminum alloy would not be seen by nanoindentation at high temperature.

Focusing only on the values at low testing temperatures (RT to 180 °C), the alloy exhibits low SRS values in the range of 0.022 to 0.036 for m_I . At 240 °C, a slight increase is found, up to

0.083, and beyond this point, an increase up to 0.16 is found. Interestingly, the SRS did not increase any more with the temperature after 400 °C. The values of SRS obtained in this work were compared with different grain size and composition aluminum (AA1050A [19,23,25], Al [18], AlZnMg [26], Al-30wt% Zn [30] nanocrystalline-Al [47]) obtained by similar nanoindentation methods (strain rate jump tests, constant strain rate, constant load). Face cubic centered nanocrystalline (nc) metals have been reported to possess m values that are up to an order of magnitude larger than those measured for cg-metals [19,47,49-50]. However, when the different values found in the literature by different methods are plotted together, the differences are not so clear.

The m -value at RT of this work (0.022) is very similar to the values obtained by most of the samples found in the literature (from 0.019 to 0.04). Zhang *et al.* suggested that the low value obtained for Al-5Zn-1.6Mg could be caused by the precipitation of different phases [26]. However, Zhao *et al.* [51] reported that the SRS was not enhanced by the introduction of precipitates in an Al7075 alloy. Chinh *et al.* [27] found an unusually high SRS of 0.22 for ufg Al-30 wt.% Zn, which was attributed to its high ductility.

Comparing the data obtained at high temperature from the work here, a good agreement with the data obtained by Wheeler *et al.* in ufg-Al [23] can be found up to 250°C. Recently, Khan *et al.* [21] showed the values of m for an Al 2024-T351 under uniaxial compression as a function of temperature, calculated at different strain rates between RT and 232 °C. The authors showed a threshold temperature (~ 100 °C), below which m was found to be close to zero. Above the threshold temperature, the value of m is controlled by two non-linear relationships, which are dependent on temperature and strain rate. The data presented here, and Wheeler's work [23] are in agreement with Khan's research, where a similar increase is found.

The small change in the SRS up to 180 °C could indicate a dislocation glide mechanism. After 180 °C, a big increase is observed in the SRS that could be associated with a change in the deformation mechanism associated with the thermally activated glide of dislocations. No data has been published by nanoindentation measurements beyond 250 °C on these materials. The data presented here could also be affected by potential precipitation processes, as observed by DSC [52]. This type of hardening obstacle, with small precipitates in the interior of the grain, is well described in Morris *et al.* [52]. The size of the precipitates increases with aging time, which decreases the total interface area [52]. Also, the precipitation of different phases or the coarsening of existing precipitates change the phase-matrix-interface structure, which could affect the

measurements. Charit and Mishra [53] reported a value for SRS of 0.17 for a similar 2024 Al at 430 °C using tensile test experiments, in agreement with the value of 0.16 found here for 460 °C.

The activation volume (V^*) was calculated using Eq. (2). The required parameters are the Taylor factor, the Boltzmann constant (k_B) and the absolute temperature (T). In order to illustrate the activation volume, it is often divided by the elemental atomic volume governing the characteristic deformation process, which is here assumed to be the cubed Burger's vector (b^3). Using a Burger's vector $b = 0.286$ nm of pure Al as a reference [54]:

$$V^* = \sqrt{3} \cdot k_B T \cdot \left(\frac{\partial \ln \dot{\epsilon}}{\partial \sigma} \right)_T \quad (2)$$

The activation volumes are shown in Table II and plotted in Figure 6. The activation volume shows a small increase, from $31 b^3$ at RT, to $41 b^3$ at 180 °C, and then a small decrease to $32 b^3$ at 240 °C. Beyond this temperature, a significant increase is observed to $112 b^3$ at 420 °C and $178 b^3$ at 460 °C. Conventional fcc metals have a large V^* (10^2 - $10^3 b^3$) [54]. The V^* for grain-boundary diffusion process is much lower, V^* (1 - $10 b^3$). The data obtained in this work are compared with similar works as shown in Figure 6. At RT, all the authors have found similar values; between 19.5 and $45 b^3$ for nc-Al [47] or ufg-Al [19,23,25], except Zhang *et al.* [25] for cg-Al-Zn-Mg alloy. Zhang *et al.* suggested that the activation volume of this alloys could be attributed to dislocation interactions (thermally activated dislocations), but the grain size of this material could be a fact to be considered [19].

Comparing the data obtained at high temperature, no big differences are found between the data from RT up to 240 °C. Wheeler *et al.* [23] showed a remaining constant activation volume from RT to 250 °C near $50 b^3$ for ufg-Al, measured under similar conditions, and in good agreement with uniaxial compression data [56]. Our study goes further than other nanoindentation studies, with tests up to 460 °C; a big increase is found on the activation volume up to $178 b^3$ with the increased temperature. The main mechanism at high temperatures is believed to follow the evolution of a dislocation structure and thermally activated process.

The SRS is important for knowing, not only in exemplum for the uniformity of sheet deformation, or for determining the temperature distribution in the billet during extrusion (the present Al-Cu-Mg-Pb alloy is mainly produced by extrusion molding), but also how the material responds during machining – surface deformation, chip microstructure, and burr formation are

main factors besides tool wear to evaluate the machinability [57]. In the present study, we approached this issue by measuring the mechanical properties, with special attention to the deformation behavior, by heating the material and performing the measurements. We found that both, the SRS and V^* increase continuously from RT to 460 °C, and therefore the deformation mechanism changes from a possible dislocation gliding and/or grain boundary diffusion processes, towards thermally activated dislocation processes. Furthermore, with increasing testing temperature the elastic properties and hardness decrease, while significant creep occurs.

The hardness measurements revealed that a softening of the material at elevated temperatures takes place. It is suspected to be the result of thermally activated processes. It is well known that the increase of temperature can amplify the thermal activation and reduces the short range thermal barriers to the motion of dislocations [21,52]. The thermal softening effect has been reported in various polycrystalline metals and alloys [58-60]. As result of the thermal softening of the Al alloy in the machining process the tool wear will be reduced [57].

The presented findings of the deformation properties, coupled with the microstructure evolution and suggested future practical machining experiments under varied conditions, may deliver valuable data for modeling metal cutting. Additionally, deformation behavior during high rate loads, strain hardening/softening, thermal softening as well as large variations in strain rate and temperatures may deliver promising results in describing machining processes [61].

4. Conclusions

Macroscopic Brinell hardness measurements after natural aging and after artificial aging, and nanonindentation up to 460°C on the widely used commercial alloy, wrought Al-3.38Cu-0.78Mg-1.0Pb, have been used to investigate the mechanical properties hardness, strain rate sensitivity, and activation volume. Furthermore, the thermoanalysis with DSC completed the picture of underlying precipitation mechanisms. The following conclusions can be drawn from the results obtained:

- The different hardening response of RT aged versus artificial temperature aged Al-Cu-Mg-Pb alloy was visualized by Brinell hardness and compared to hardness measurement results of the AA2024 alloy from literature – Peak hardness is reached faster with artificial aging. The Al-Cu-Mg-Pb alloy in both heat treatment cases reached not as high peak hardness values as the AA2024 alloy. This might be attributed to the altered Cu:Mg

ratio by adding lead into the composition: less Mg is available for its contribution to the hardness response.

- With thermal analysis (DSC measurements) it was possible to reveal the decomposition sequence of the Al-Cu-Mg-Pb alloy compared to the AA2024 reference.
- Al-Cu-Mg-Pb mechanical properties are strongly influenced by the testing temperature. With the increasing of the temperature up to 460 °C, the hardness and young modulus decrease, while m and V^* increase. A thermally activated process is suggested as the main mechanism of deformation.
- The biggest change to the mechanical properties of this alloy at high temperature is found beyond 240 °C, and is associated with the loss of the structure starting around 0.4 T/T_m . Due to the specific temperature dependent change of the strain rate sensitivity, the typical maximum operating temperature for this material is 250 °C.

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Figure Captions

Figure 1. Brinell hardness-time plots for the Al-Cu-Mg and Al-Cu-Mg-Pb alloys aged (a) at room temperature (T4) and (b) at 180°C (T6).

Figure 2. DSC thermograms for the Al-Cu-Mg (a) and the Al-Cu-Mg-Pb (b) alloy at 15 Kmin⁻¹ heating rate.

Figure 3. Representative nanoindentation load-displacement curves of Al-Cu-Mg-Pb (a) and displacement-time plot (b) obtained at different temperatures.

Figure 4. Al-Cu-Mg-Pb alloy a) Hardness measurement as a function of the creep time. b) Hardness values at the beginning (H_{ini}) and the end of the dwell period (H_{creep}) and Young modulus (E).

Figure 5. Strain rate sensitivity, m , as a function of the temperature of the Al-Cu-Mg-Pb alloy.

Figure 6. Activation volume, V^* , as a function of the temperature of Al-Cu-Mg-Pb.

Table I: Chemical composition of the investigated materials and their related composition specifications in weight percent [29]

Alloy	Cu	Mg	Mn	Ti	Si	Fe	Pb	Cu/Mg ratio
Al-Cu-Mg (AA2024)	4.31	1.46	0.54	0.022	0.116	0.178	–	1.13
Al-Cu-Mg-Pb (AA2030)	3.38	0.78	0.58	0.089	0.157	0.137	1.01	1.66
AA2024	3.8– 4.9	1.2– 1.8	0.3– 0.9	≤ 0.15	≤ 0.5	≤ 0.5	–	
AA2030	3.3– 4.5	0.5– 1.3	0.2– 1.0	≤ 0.2	≤ 0.8	≤ 0.7	0.8– 1.5	

Table II. Mechanical properties of the Al-Cu-Mg-Pb alloy obtained by nano indentation at varied testing temperatures: hardness after the creep tests (H_{creep}) and calculated hardness at the beginning (H_{ini}) of the creep, percentage of change in hardness during the creep, Young Modulus (E), strain rate sensitivity of Stage I (m_I) and II (m_{II}) and activation volume, V^* .

Temperat ure	H_{creep}	H_{ini}	Percentag e of change in H	E	m_I	m_{II}	V^*
	GPa		%	MPa			b^3
25	1.26 ± 0.06	1.41 ± 0.05	11.5	84.2 ± 3.0	0.022 ± 0.009	0.114 ± 0.093	31.2
180	0.90 ± 0.09	1.14 ± 0.08	21.4	68.6 ± 2.7	0.036 ± 0.011	0.185 ± 0.132	41.4
240	0.53 ± 0.07	0.82 ± 0.09	35.8	65.5 ± 4.8	0.082 ± 0.006	0.170 ± 0.042	32.2
400	0.11 ± 0.02	0.26 ± 0.05	57.8	28.6 ± 3.9	0.159 ± 0.035	0.226 ± 0.025	112.45
480	0.08 ± 0.01	0.19 ± 0.03	58.8	18.6 ± 2.9	0.158 ± 0.026	0.345 ± 0.093	178.10

Figure 1.

Figure 2.

Figure 3.

Figure 4

Figure 5

Figure 6