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The punch speed influence on warm forming and springback of two Al-Mg-Si alloys

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Highlights

The influence of punch speed on warm forming under non-isothermal conditions was analyzed for heat-treatable aluminum alloys.

A higher punch speed minimizes the exposure time and the heat transfer with the tools in warm forming.

Dynamic precipitation (artificial aging) occurs in warm forming at low punch speeds for naturally aged alloys.

Formability and springback remained stable or improved with the increase in punch speed.

1 **Abstract**

2 The punch speed has a major influence on the success of sheet metal warm forming of aluminum alloys,
3 particularly in the part's final geometry and in-service behavior. However, this process parameter has not
4 often been analyzed in the literature. Additionally, in warm forming of heat-treatable aluminum alloys
5 under non-isothermal conditions no study has been reported concerning the influence of the punch speed
6 on springback. In the present work, two heat-treatable Al–Mg–Si alloys (EN AW 6016-T4 and EN AW
7 6061-T6) are studied between 22 °C and 200 °C, using tensile tests and cylindrical cup forming followed
8 by split ring (springback) tests. Monotonic tensile tests and stress relaxation tensile tests were performed
9 at strain rates from 2×10^{-4} to $2 \times 10^{-2} \text{ s}^{-1}$. Dynamic strain aging (Lüders bands) was observed at 22 °C after
10 each stress relaxation stage. When using low strain rates and stress relaxation stages at 200 °C, the
11 occurrence of dynamic precipitation for the EN AW 6016-T4 alloy becomes more evident due to the
12 higher test time. Cylindrical cup tests were performed for punch speeds from 0.1 to $10 \text{ mm} \cdot \text{s}^{-1}$. In warm
13 forming, a higher punch speed is advantageous since it minimizes the occurrence of dynamic precipitation
14 and the heat transfer between the sheet and the tools. Thus, formability and springback remained stable or
15 improved with the increase in punch speed. Additionally, a higher punch speed is advantageous in the
16 industrial field since it makes higher production rates possible with a better control of the temperature
17 difference between the tools.

18

19

20 **Keywords:** Al-Mg-Si alloys; Warm forming; Punch speed; Strain rate; Temperature gradient; Springback

21

1 Introduction

Current environmental issues demand higher sustainability, namely in the transport sector, which includes the reduction of fuel consumption and its emissions but also the use of recyclable materials. Therefore, the transport sector is examining new materials and manufacturing strategies that make it possible to reduce the weight of the components while increasing their strength and improving the vehicles' safety. The reduction of the weight of vehicles is crucial to reducing fuel consumption and its emissions. Aluminum alloys present good specific strength, high corrosion resistance and a high recycling rate, which has led to an increasing demand for these alloys concerning their application in lightweight vehicles [1]. Within aluminum alloys, the heat-treatable ones provide additional strength, by the precipitation hardening after forming operations, which is an advantage when compared to the non-heat-treatable aluminum alloys. Nevertheless, aluminum alloys present low formability and non-negligible springback in forming operations at room temperature (RT).

Warm forming has been shown to be an effective solution to improve formability for both non-heat-treatable and heat-treatable aluminum alloys. A warm temperature, below the recrystallization temperature of the material, increases the ability of the sheet metal to be formed into a desired shape without necking. The research on this topic has a high relevance to industry and has been focused on the: **design** of warm forming tools [2], development of alternative warm forming technologies [3], improvement of thermo-mechanical finite element analysis algorithms [4,5] and characterization and modeling of the mechanical behavior at warm temperatures [6–8]. Bolt et al. [9] compared the warm forming behavior of a non-heat-treatable (Al-Mg) with a heat-treatable alloy (Al-Mg-Si). The results indicate that, although both alloys showed significant improvement in their formability compared with that at RT, the non-heat-treatable alloy provides higher part depths than the heat-treatable one. Additionally, the studies on the warm forming conditions reported a reduction of the springback behavior for non-heat-treatable alloys, either when using the U-rail test by Moon et al. [10] or the split-ring test by Grèze et al. [11]. Recently, Simões et al. [12] have shown that warm conditions also reduce the springback for heat-treatable aluminum alloys, with the additional benefit of minimizing the variability induced by the natural aging in the forming operations of these alloys.

Warm forming can be performed under isothermal or non-isothermal (i.e. with a temperature gradient) conditions. Wilson [13] reviewed the warm forming of aluminum alloys concluding that

51 significant improvements in formability can be attained by reducing the flow strength of the material in
52 the drawing zone, relative to that in the stretch-formed zone by differential heating of the blank. In deep
53 drawing operations, this can be achieved by decreasing the punch temperature relative to that of the
54 die/blank-holder. Palumbo and Tricarico [14] adopted this approach to study a non-heat-treatable Al-Mg
55 alloy (EN AW 5754). They explain that the cooling action exerted by the punch and the softening of the
56 material at the heated flange leads to a reduction in the punch load and, consequently, to lower stress and
57 strain levels in the punch radius section. Thus, as shown by Panicker et al. [15] for the same alloy, the
58 non-isothermal warm forming is more effective in avoiding excessive thinning than isothermal warm
59 forming. On the contrary, Li and Ghosh [16] have shown that increasing the punch temperature, relative
60 to the one of the die/blank-holder, leads to an increase in the ratio of the material being stretched over the
61 punch relative to the material being drawn into the die.

62 The punch speed is an important process parameter in warm forming since it influences the
63 productivity. It is commonly accepted that the flow stress of aluminum alloys increases for higher
64 forming rates, particularly at higher temperatures Li and Ghosh [17]. This positive strain-rate sensitivity
65 counterbalances the decrease in the flow stress for higher temperatures. Shehata et al. [18] and Ayres and
66 Wenner [19] studied the influence of the punch speed in warm forming with a hemispherical punch, for
67 non-heat-treatable Al-Mg alloys, under isothermal conditions. Shehata et al. [18] concluded that the
68 influence of the punch speed may not be critical in the range from 0.02 to 0.42 millimeters per second
69 ($\text{mm}\cdot\text{s}^{-1}$). However, Ayres and Wenner [19] showed that formability decreases as the punch speed
70 increases in the range from 0.83 to $83.33 \text{ mm}\cdot\text{s}^{-1}$.

71 Under non-isothermal conditions, the temperature gradient is largely determined by the punch
72 speed. In fact, higher punch speeds lead to a lower exposure time and, consequently, minimize the heat
73 transfer between the tools and the blank. Palumbo and Tricarico [14] analyzed the influence of the punch
74 speed (range from 0.10 to $0.25 \text{ mm}\cdot\text{s}^{-1}$) on the evolution of the temperature difference between the
75 peripheral zone of the blank (flange heated zone at 200°C) and the blank center (cold punch zone at RT)
76 in the forming of a cylindrical cup. It was shown that the temperature difference in the blank between the
77 flange heated zone and the center progressively increases during the forming process since the contact
78 area of the blank with the cooled punch increases while its flange is kept heated. This temperature
79 difference slightly decreases with the increase in the punch speed resulting in lower formability. Naka and
80 Yoshida [20] analyzed the influence of the punch speed (from 0.003 to $8.33 \text{ mm}\cdot\text{s}^{-1}$) for temperatures

81 between RT and 180 °C for the EN AW 5083 alloy, concluding that for values lower than $\approx 2.00 \text{ mm}\cdot\text{s}^{-1}$
82 the changes in the limiting drawing ratio are negligible. Whatever the temperature gradient considered,
83 the studies for non-heat-treatable aluminum alloys state that the formability decreases as the punch speed
84 increases.

85 On the other hand, Ghosh et al. [21] studied the influence of the punch speed during warm
86 forming at 250 °C of two heat-treatable Al-Mg-Si alloys. From 1.00 to 2.00 $\text{mm}\cdot\text{s}^{-1}$, the punch speed had
87 no effect on the force – displacement curve. However, the punch force presents a relevant increase for a
88 sufficiently lower (i.e. 0.183 $\text{mm}\cdot\text{s}^{-1}$) punch speed, since the initial soft material gradually becomes harder
89 and approaches the T6 condition (i.e. a change of the initial heat treatment occurs as a result of the longer
90 exposure time at 250 °C). In this context, Palumbo et al. [22] studied the influence of the pressure rate
91 during the warm hydroforming of an Al-Mg-Si alloy, in a temperature range from RT to 350 °C. The best
92 improvement in formability was achieved at 200 °C using the higher pressure rate of 25 $\text{bar}\cdot\text{s}^{-1}$, to
93 minimize the exposure to the temperature employed. For an exposure time longer than 400 s, at
94 temperatures equal to or higher than 250 °C, there is an increase in the yield stress and tensile strength
95 and reduced ductility.

96 Besides good formability, to ensure a successful forming operation, it is also necessary to
97 guarantee a satisfactory in-service performance of the part. Thus, the part should maintain acceptable
98 strength and ductility after forming, or else, additional treatments have to be performed to attain the
99 required properties. Li and Ghosh [16] noted a high decrease in the post-forming yield stress, as
100 compared with the yield stress before the forming operation, when the temperature increases from 200 to
101 350 °C, for both non-heat and heat-treatable alloys. This decrease in the post-forming yield stress occurs
102 linked with the increase in the tensile elongation. Fan et al. [23] investigated the warm forming and post-
103 forming mechanical properties of a cylindrical cup for a heat-treatable Al-Mg-Si alloy in a temperature
104 range from 25 to 300 °C, under isothermal conditions. Enhanced formability and post-forming hardness
105 could be obtained simultaneously between 150 and 200 °C; however, both declined as the temperature
106 increases to 300 °C.

107 The influence of warm forming on the springback is usually evaluated using the cylindrical cup
108 test followed by split-ring test or the U-rail. Using the split-ring test, under isothermal conditions with a
109 punch speed of 0.5 $\text{mm}\cdot\text{s}^{-1}$, a decrease in springback for forming temperatures equal to or above 150 °C

110 was observed by Grèze et al. [11] for a non-heat-treatable EN AW 5754-O alloy. A similar conclusion
111 was observed under non-isothermal conditions by Laurent et al. [6] for the EN AW 5754-O alloy and by
112 Simões et al. [12] for two Al-Mg-Si alloys (EN AW 6016-T4 and EN AW 6061-T6), both using a punch
113 speed of $1 \text{ mm}\cdot\text{s}^{-1}$. Kim and Koç [24] performed a numerical analysis of the U-rail test for an EN AW
114 5754-O alloy. These authors compare the isothermal with non-isothermal heating conditions for a punch
115 speed of $10 \text{ mm}\cdot\text{s}^{-1}$ and conclude that both conditions reduce springback. However, isothermal conditions
116 presented the highest springback reduction. The influence of the punch speed (from 1 to $20 \text{ mm}\cdot\text{s}^{-1}$) on
117 springback was also analyzed by the same authors, which concluded that it increases as the punch speed
118 increases, in both isothermal and non-isothermal conditions. This indicates that, in general, a slow
119 forming rate is preferable in warm forming to increase formability and improve the dimensional accuracy
120 of the parts formed. Similar conclusions were reached by Moon et al [10] in the experimental analysis of
121 a non-heat-treatable EN AW 1050 alloy under non-isothermal conditions in the punch speed range from 1
122 to $10 \text{ mm}\cdot\text{s}^{-1}$. Moreover, according to Moon et al. [10], the variations of the heat transfer between the
123 tools and the blank, caused by punch speed changes, are the probable cause of some springback
124 variations.

125 The aim of this paper is to analyze the influence of the punch speed on the springback behavior
126 and formability of heat-treatable alloys under non-isothermal warm forming conditions. In fact, only a
127 few studies exist in the literature about the warm forming of heat-treatable alloys, but none considers the
128 springback analysis. In this work two Al-Mg-Si alloys were selected, EN AW 6016-T4 and EN AW
129 6061-T6. The paper is organized as follows. Section 2 presents these alloys and the experimental
130 characterization of their thermo-mechanical behavior at RT and 200°C . The cylindrical cup tests are
131 presented in Section 3 for different punch speeds. The experimental results analyzed are the temperature
132 evolution over time, the punch force evolution with its displacement, the thickness distribution along the
133 cup wall and the springback. For the EN AW 6016-T4 alloy, the influence of warm forming on the
134 natural aging, at different punch speeds, is also evaluated. Finally, the analysis and discussion are
135 presented in Section 4 and the main conclusions are summarized in Section 5.

136 **2 Materials and thermo-mechanical behavior**

137 The EN AW 6016-T4 alloy is commonly used for skin applications while the EN AW 6061-T6
138 alloy is frequently used for structural components. The alloys with the T4 designation are more ductile

139 than the T6 ones. Figure 1 presents the differences between these heat treatments in the Al-Mg-Si alloys,
 140 correlating the production sequence with the precipitation one, as reviewed in Engler and Hirsch [25]. As
 141 shown in Figure 1 a), after the rolling operations, the alloys were submitted to solution heat treatment
 142 (SHT) that consists in a heating phase and a holding stage between 500 to 540 °C, followed by a fast
 143 quenching to RT. The aim of the holding time at solvus temperature is to dissolve the Mg-Si precipitates
 144 in the aluminum matrix, whereas the fast quenching to RT makes it possible to maintain the alloying
 145 elements dissolved in the matrix, leading to a supersaturated solid solution (SSSS), see Figure 1 b). This
 146 SSSS state is the driving force for precipitation in Al-Mg-Si alloys. Consequently, precipitation hardening
 147 occurs during storage at RT, known as natural aging, up to stabilization, which is designated by T4.
 148 Artificial aging (the T6 heat treatment) is promoted in a new holding stage at a temperature range of
 149 between 170 to 200 °C, see Figure 1 a), for between 20 to 30 minutes. During this stage, the strength
 150 increases until a maximum value, linked to the precipitation of β'' needles, see Figure 1 b). The extended
 151 holding at the artificial aging temperature leads to a decrease in strength, known as over-aging.

Figure 1 – a) Thermal cycle of heat treatment and production; b) Precipitation sequence in Al-Mg-Si alloys [26].

152
 153 The temperature range for the warm forming of Al-Mg-Si alloys is similar to the one used in
 154 their T6 heat treatment, as shown in Figure 1. Therefore, artificial aging or even over-aging can occur
 155 during the warm forming process. In this sense, according to the precipitation sequence shown in
 156 Figure 1 b), an alloy, initially in the T4 condition, may evolve to the T6 condition; while an alloy initially
 157 in T6 condition may evolve to the condition of over-aging. The precipitation kinetics increases as the
 158 temperature increases and then decreases as the precipitation evolves. Thus, the driving force for
 159 precipitation is higher in the T4 condition than in the T6 condition. Moreover, the precipitation is a

diffusion phenomenon, which is time dependent. The precipitation hardening that occurs due to temperature and time is called static aging. However, the plastic deformation leads to dislocation-assisted precipitation, known as dynamic aging or dynamic precipitation (hardening). As shown by Cai et al. [27], the occurrence of plastic deformation greatly reduces the exposure time required for precipitation hardening at warm temperatures, when compared to that associated to static aging. Kolar et al. [28] studied the influence of pre-strain at RT followed by precipitation hardening at 190 °C (sequential mode) and warm forming at 190 °C (simultaneous mode) on the material strength. They concluded that the simultaneous mode leads to higher in-service strength than the sequential mode. However, according to Fan et al. [23], warm forming at temperatures higher than 200 °C can contribute to reducing post-forming mechanical properties and the in-service behavior of aluminum components. Thus, in the present work, the analysis of results takes into account the exposure time, the temperature and the strain rate, which are related to the punch speed. As highlighted in the literature review performed, the influence of temperature on the precipitation sequence and kinetics as well as the connection between microstructural changes and the mechanical behavior are well known for Al-Mg-Si alloys. This knowledge is used to discuss the results following a macroscopic approach using uniaxial tensile tests, cylindrical cup tests and split ring tests.

The correlation between the punch speed and the strain rate imposed on the material was studied using finite element analysis by Palumbo and Tricarico [14] and Neto et al. [8] in the warm forming of the cylindrical cup test. Both studies analyzed a non-heat-treatable Al-Mg alloy under non-isothermal conditions, with tool geometry identical to the one used in this study. Neto et al. [8] analyzed three different temperatures (RT, 150 and 240 °C) and concluded that the strain rate in a point initially located in the flange (5 mm from the perimeter) is independent of the test temperature. For a punch speed of 5 mm·s⁻¹, a point sliding over the die (flange region) suffers a plastic strain rate increase from about 10⁻² to 10⁻¹ s⁻¹. This value of the plastic strain rate is kept almost constant during its sliding over the die radius (up to 18 mm of punch displacement). The plastic strain rate suddenly decreases when the point reaches the cup wall. Palumbo and Tricarico [14] studied two punch speeds of 0.016 and 1 mm·s⁻¹ for four points, located at: (i) the bottom of the cup, (ii) the cup radius and two points in (iii) the flange zone. This study shows that after 12.5 mm of punch displacement, the points located at the bottom of the cup and in the radius present no plastic deformation. The higher strain rate was obtained for the point located at the cup radius, which presents the greater reduction in thickness. This strain rate was comprised in a range

190 between 10^{-4} and 10^{-3} s^{-1} , for a punch speed of $0.016 \text{ mm}\cdot\text{s}^{-1}$, and between 10^{-2} and 10^{-1} s^{-1} , for a punch
191 speed of $1 \text{ mm}\cdot\text{s}^{-1}$. For the points located in the flange zone, the strain rate was comprised in a range
192 between 2×10^{-4} and $6\times 10^{-4} \text{ s}^{-1}$ for a punch speed of $0.016 \text{ mm}\cdot\text{s}^{-1}$ and, 1×10^{-2} and $3\times 10^{-2} \text{ s}^{-1}$ for a punch
193 speed of $1 \text{ mm}\cdot\text{s}^{-1}$. In brief, the increase in strain rate seems to be proportional to the increase in punch
194 speed, i.e. by doubling the punch speed the strain rate in the flange zone also doubles. Based on these
195 results, the punch speeds between 0.1 and $10 \text{ mm}\cdot\text{s}^{-1}$ can be related to maximum strain rate values in the
196 range 10^{-3} and 10^{-1} s^{-1} , respectively. The influence of the temperature and strain rate on the mechanical
197 behavior is first analyzed using uniaxial tensile tests.

198 2.1 Materials

199 The EN AW 6016-T4 alloy was provided by Constellium along with the date of the SHT;
200 therefore, the storage time at RT is counted after this date for this alloy. The EN AW 6061-T6 alloy was
201 acquired on the retail market and no information was provided about the date of the SHT and T6 heat
202 treatments; therefore, the storage time at RT is counted taking its reception in the laboratory as its
203 reference. In a previous work by Simões et al. [12], it was shown that the mechanical properties of the EN
204 AW 6016-T4 alloy change during storage at RT, becoming relatively stable after 7 months. Therefore, in
205 order to better understand the effect of natural aging on warm forming, the tests at different punch speeds
206 were performed considering the natural aging times of 1 and 18 months for the EN AW 6016-T4 alloy.
207 The mechanical properties of the EN AW 6061-T6 remained relatively stable during the storage period,
208 thus a natural aging analysis was not performed. The temperature for the warm forming was chosen based
209 on the literature which often refers to 200°C as a temperature that improves formability while
210 maintaining good post-forming mechanical properties.

211 Table 1 presents the mechanical properties obtained for both alloys, from a tensile test at RT at a
212 strain rate of $2\times 10^{-3} \text{ s}^{-1}$, with the specimens oriented along the rolling direction (RD). The definitions used
213 in this work are according to the ISO 6892-1:2009 [29], i.e. R_m tensile strength; $R_{p0.2}$ proof strength at
214 0.2% of the extensometer gauge length; A_{gt} percentage of total elongation at maximum force; n_{4-6} strain
215 hardening exponent between 4 and 6 % of plastic elongation; and n_{10-15} strain hardening exponent
216 between 10 and 15 % of plastic elongation. The thickness of both alloys is approximately 1 mm, with the
217 average value of fifty measurements also presented in Table 1. Note that for the EN AW 6016-T4, the
218 increase of the natural aging time leads to the decrease of the strain hardening exponent. However, this is

219 accompanied by an increase of the strength coefficient, as previously reported by Zhang et al. [30] and
 220 Zhong et al. [31].

221

Table 1 – Mechanical properties at RT of EN AW 6016-T4 and EN AW 6061-T6 alloys along RD [12].

	Aging time	$R_{p0.2}$	R_m	A_{gt}	n_{4-6}	n_{10-15}	Thickness
EN AW 6016-T4	1 month	101 MPa	215 MPa	23.9 %	0.32	0.26	1.05 mm
	18 months	130 MPa	247 MPa	22.8 %	0.28	0.24	
EN AW 6061-T6	–	270 MPa	325 MPa	12%	0.14	–	0.98 mm

222

223 2.2 Thermo-mechanical behavior in uniaxial tension

224 To better understand the thermo-mechanical behavior, uniaxial tensile tests were performed at
 225 RT ($\approx 22^\circ\text{C}$) and 200°C , at strain rates of 2×10^{-4} , 2×10^{-3} and $2 \times 10^{-2} \text{ s}^{-1}$, using a Gleeble 3500 machine
 226 (Dynamic Systems Inc.). The specimens have the traditional dog-bone shape with a gauge section of
 227 40 mm length and 10 mm width and were cut with the length aligned with the RD using wire cut
 228 electrical discharge machining (EDM). The Gleeble device uses a high-speed heating method by direct
 229 resistance (Joule effect), enabling a heating time of 20 s. The test temperature was controlled by a type K
 230 thermocouple fixed at the center of the specimen. This heating procedure promotes a temperature gradient
 231 along the specimen's length, with a higher temperature and symmetry at the center of the specimen. As
 232 shown by Coër et al. [32], this temperature gradient can promote heterogeneous mechanical properties
 233 along the length of the specimen, resulting in a higher deformation at its center. Nonetheless, at 200°C ,
 234 the temperature difference between the center of the specimen and a point located 1.5 mm from it does
 235 not exceed 0.4°C . Accordingly, the true strain is computed using only a rectangular gauge measurement
 236 area which is 6 mm wide and 3 mm long, at the center of the specimen, as detailed in Coër et al. [32]. The
 237 strain field was acquired by digital image correlation (DIC) with the Aramis-4M (GOM mbH) system.
 238 The tensile test results are presented as true stress – true strain curves, the true stress is determined using
 239 the true strain obtained by DIC at the center of the specimen.

240 The uniaxial tensile tests were performed considering two loading conditions: a) monotonic
 241 loading up to rupture; b) monotonic loading interrupted by relaxation stages of 60 s, in a cycle up to
 242 rupture, named stress relaxation tests. These stress relaxation tests were performed considering two
 243 different conditions: b.i) a single strain rate of $2 \times 10^{-3} \text{ s}^{-1}$ was used in all loading stages; b.ii) two strain

244 rates were used alternately, such that, the test starts with a strain rate of $2 \times 10^{-4} \text{ s}^{-1}$, followed by a stress
245 relaxation stage, then a load stage at $2 \times 10^{-2} \text{ s}^{-1}$, followed by a stress relaxation stage, and so on, up to
246 rupture. All the stress – strain curves are presented up to the maximal load, except in the cases
247 corresponding to the b.ii) condition. The goal of the stress relaxation tests is to study the possibility of the
248 occurrence of dynamic precipitation under warm forming conditions.

249 **2.2.1 Uniaxial tensile tests on the EN AW 6016-T4 alloy**

250 The specimens used for the EN AW 6016-T4 alloy had a natural aging time of 1 month. Figure 2
251 presents the stress – strain curves for EN AW 6016-T4, performed at RT and 200 °C, for three different
252 strain rates: 2×10^{-4} , 2×10^{-3} and $2 \times 10^{-2} \text{ s}^{-1}$, according to condition a). Globally, the alloy presents negligible
253 strain-rate sensitivity at RT and remarkable strain-rate sensitivity at 200 °C. At both temperatures, the
254 increase in the strain rate causes a slight increase in the $R_{p0.2}$ while the hardening rate decreases. It results
255 in a slightly positive strain-rate sensitivity for the initial values of plastic strain and negative strain-rate
256 sensitivity at higher values of plastic strain. This behavior is particularly notable at 200 °C, where the test
257 performed at $2 \times 10^{-4} \text{ s}^{-1}$ presents the lowest yield stress and the highest hardening rate that saturates after a
258 strain of 0.05, resulting in a lower elongation. A similar hardening behavior was reported by Ghosh [33]
259 for this alloy, based on tensile tests performed at 250 °C, but at a strain rate of $2 \times 10^{-3} \text{ s}^{-1}$. This indicates
260 that at a higher temperature the dynamic precipitation hardening occurs for a shorter exposure time. On
261 the other hand, the tests performed at strain rates of 2×10^{-3} and $2 \times 10^{-2} \text{ s}^{-1}$ present a smooth evolution of
262 the hardening. Nevertheless, the hardening rate is lower at 2×10^{-2} than at $2 \times 10^{-3} \text{ s}^{-1}$. Thus, although the test
263 performed at $2 \times 10^{-2} \text{ s}^{-1}$ presents a higher $R_{p0.2}$, the value of R_m is lower than the one attained at $2 \times 10^{-3} \text{ s}^{-1}$.
264 The engineering stress – strain curves presented by Laurent et al. [34] highlight that the tests performed at
265 strain rates of 2×10^{-3} and $2 \times 10^{-2} \text{ s}^{-1}$ show an increase of the total elongation at fracture with the
266 temperature increase, i.e. the ductility enhancement is achieved primarily by the post-uniform elongation.
267 However, the test performed at $2 \times 10^{-4} \text{ s}^{-1}$ and 200 °C shows a clear reduction of the total elongation at
268 fracture, even when compared with the test performed at RT.

Figure 2 – Influence of the strain rate on the stress – strain curves obtained from uniaxial tensile tests performed at 22 and 200 °C, for EN AW 6016-T4 at 1 month of natural aging. The times presented at the end of the curve correspond to the test duration until attaining the R_m instant. For the test at a strain rate of $2 \times 10^{-4} \text{ s}^{-1}$ other relevant instants are marked.

269

Figure 3 – Influence of the strain rate on the strain distribution along the length of the specimen. The strain profiles are taken at four instants and correspond to the tests at 200 °C presented in Figure 2.

270

271 Figure 3 presents the strain distribution along the length of the specimen, measured for a section

272 at the middle of the specimen's width (purple dash line in the detail of the specimen), for the tests

273 performed at 200 °C. The strain profile is presented for four instants: $\varepsilon = 0.02$; 0.06 ; R_m and at rupture. It

can be observed that the strain profile presents a different evolution with the strain rates of 2×10^{-3} and $2 \times 10^{-2} \text{ s}^{-1}$, the strain profile always presents preferential strain localization at the center of the specimen, whatever the strain value. This strain localization is linked with the temperature gradient along the specimen's length, which presents a higher temperature at its center. On the other hand, for a strain rate of $2 \times 10^{-4} \text{ s}^{-1}$, and in the strain interval between 0.02 and 0.08, the specimen deforms preferentially in the regions adjacent to its center (see supplementary material, video 1). This seems to be linked with the occurrence of dynamic precipitation hardening at the center of the specimen, due to its higher exposure time to the same temperature, caused by the slow strain rate. This leads to the increase in strength at the center of the specimen promoting the deformation of the adjacent regions, which are softer. However, after $\varepsilon = 0.08$ and up to rupture, the specimen also preferentially deforms at its center, indicating the occurrence of dynamic precipitation in all the specimen's gauge area. Moreover, at 2×10^{-3} and $2 \times 10^{-2} \text{ s}^{-1}$ and in a strain interval of up to 0.06, both tests present a similar stress – strain curve (see Figure 2) as well as similar strain profiles (see Figure 3). However, afterwards, the test at a strain rate of $2 \times 10^{-3} \text{ s}^{-1}$ presents a higher stress value for the same strain value. In fact, as shown in Figure 3, at $2 \times 10^{-2} \text{ s}^{-1}$ the deformation is more localized at the center of the specimen, while at $2 \times 10^{-3} \text{ s}^{-1}$ it is more distributed along the length. This may also be linked with the occurrence of dynamic precipitation hardening at $2 \times 10^{-3} \text{ s}^{-1}$, which increases the material strength at the center, due to its higher exposure time, and causes the material to deform in the adjacent regions that are at lower temperatures.

Stress relaxation tests were performed to improve the knowledge concerning the occurrence of dynamic precipitation hardening and the influence of the exposure time (test duration). Figure 4 a) and b) show the results for the conditions b.i) and b.ii), respectively. In both figures, the curves of the monotonic tests presented in Figure 2 are added for comparison. Figure 5 a) and b) present the stress – time curves corresponding to the stress – strain curves in Figure 4 a) and b), respectively.

At RT, for both conditions b.i) and b.ii), the results show that the stress relaxation curves match the monotonic tensile curves, see Figure 4 a) and b). These results highlight the fact that, at RT, the stress relaxation stage at single and multiple strain rates has no significant contribution to the strain hardening behavior. However, for both conditions, after each relaxation stage and upon reloading, the stress value suddenly increases attaining a local maximum and stabilizes to a plateau. After each stress-plateau, the formation and propagation of Lüders bands (dynamic strain aging mechanism) was observed by digital image correlation, as shown in the detail in Figure 5 a) (see supplementary material, video 2). According

304 to Picu et al. [35], this is caused by the diffusion of Mg and Si atoms towards dislocations, originating
305 Mg-Si precipitates. Such precipitates block the dislocation movements in the subsequent loading stages,
306 promoting an increase in stress to overcome them and, consequently, Lüders bands occur.

307 At 200 °C, the stress relaxation stage has a non-negligible influence on the mechanical behavior
308 of the material. As shown in Figure 4 a), after each stress relaxation stage, the stress-strain curve offsets
309 upwards compared to the monotonic curve, while presenting a similar hardening rate. Such behavior
310 suggests that during each stress relaxation stage some precipitation phenomenon occurs causing the
311 strengthening of the alloy. Concerning the stress relaxation tests performed with two strain rates at
312 200 °C, see Figure 4 b), the stress relaxation curve overlaps the equivalent monotonic one in all the
313 periods performed with a strain rate value of $2 \times 10^{-4} \text{ s}^{-1}$ and diverges when the strain rate value of $2 \times 10^{-2} \text{ s}^{-1}$
314 is used. These results show that the EN AW 6016-T4 alloy has a positive strain-rate sensitivity of at
315 200 °C, since higher stress values are associated with higher strain rate values.

a) Stress relaxation tests at a single strain rate

b) Stress relaxation tests at multiple strain rates

Figure 4 – Influence of the strain rate on the stress – strain curves obtained from the stress relaxation tensile tests, for EN AW 6016-T4 at 1 month of natural aging.

316

317 As shown in Figure 5 a), the exposure time of the stress relaxation test with a single strain rate of
 318 $2 \times 10^{-3} \text{ s}^{-1}$ is about three times higher than the equivalent monotonic test, limiting any comparison to only
 319 the first load cycle. The same comment is valid for the two tests performed at RT shown in Figure 5 b).
 320 On the other hand, both tests performed at 200 °C have a similar exposure time, which results in the
 321 overlap of the stress – strain curves shown in Figure 4 b). It is interesting to note that the monotonic
 322 tensile test at $2 \times 10^{-4} \text{ s}^{-1}$ and the stress relaxation tensile test with two strain rate values, despite the

different load sequence, present a similar stress – strain behavior when the specimen is loaded for the same instant of time at the same strain rate. It highlights the fact that in warm forming of the EN AW 6016-T4 alloy, the control of the exposure time is crucial for determining the mechanical behavior. In fact, for a similar exposure time the material presents a positive strain-rate sensitivity (see Figure 4 b)), but for different exposure times, leading to different rates of dynamic precipitation, a negative strain-rate sensitivity can be observed (Figure 2).

a) Stress relaxation tests at a single strain rate

b) Stress relaxation tests at multiple strain rates

Figure 5 – Influence of the strain rate on the stress – time curves obtained from the stress relaxation tensile tests, for EN AW 6016-T4 at 1 month of natural aging.

2.2.2 Uniaxial tensile tests on the EN AW 6061-T6 alloy

Figure 6 presents the stress – strain curves obtained from the uniaxial tensile tests performed at RT and 200 °C, for EN AW 6061-T6. The monotonic tensile tests were performed for three strain rates: 2×10^{-4} , 2×10^{-3} and $2 \times 10^{-2} \text{ s}^{-1}$. At RT, a stress relaxation test was also performed considering a constant strain rate of $2 \times 10^{-3} \text{ s}^{-1}$ for each load cycle. The alloy presents a negligible strain-rate sensitivity at RT. The stress relaxation test at RT indicates no changes in the mechanical behavior and a stress jump associated with the presence of Lüders bands, similar to that observed for the EN AW 6016-T4 alloy.

At 200 °C, a stress relaxation test was also performed with condition b.ii). At this temperature, this alloy presents a positive strain-rate sensitivity. The hardening rate seems to be independent of the strain rate since the monotonic and the stress relaxation tests show a similar hardening behavior. Moreover, a similar hardening behavior in function of the strain rate was observed by Ghosh [33], for monotonic tensile tests performed on the same alloy, at 250 °C with strain rates of $2 \times 10^{-2} \text{ s}^{-1}$ and $2 \times 10^{-1} \text{ s}^{-1}$. Lastly, no relevant mechanical behavior change was noticed at 200 °C for this alloy as a result of the stress relaxation stages, indicating that no significant microstructural change occurs, i.e. there is no influence of the exposure time.

Figure 6 – Influence of the strain rate on the stress – strain curves obtained from stress relaxation tensile tests, for EN AW 6061-T6.

3 Cylindrical cup forming tests

In the present section, the forming behavior of the two alloys selected is evaluated at RT and 200 °C, considering different punch speeds. The forming behavior of the EN AW 6016-T4 alloy was analyzed at two natural aging times: 1 and 18 months; while for the EN AW 6061-T6 alloy the forming tests were only performed considering 18 months after its reception. At RT, the punch speeds analyzed were 1 and 10 mm·s⁻¹, while at 200 °C three values were considered: 0.1, 1 and 10 mm·s⁻¹. The lower and higher punch speeds were imposed by the characteristics of the device. As previously mentioned, this punch speed range corresponds to the maximum strain rate values in the range 10⁻³ and 10⁻¹ s⁻¹, respectively.

3.1 Experimental procedure

The cylindrical cup test consists in forming a cup from a circular shaped metallic sheet, the blank, in a forming die by the mechanical action of a punch. A blank-holder tool is used to avoid wrinkles and to control the material flow into the die. The dispositive used was a Zwick/Roell Amsler BUP200 sheet metal testing machine, using tools specifically adapted for warm forming. The details about the warm forming tools are given in Coër [36] and Laurent et al. [6]. The circular blank has a diameter of 60 mm and was extracted from the sheet by squaring shear using a specific blade tool in the Zwick BUP200 dispositive.

Figure 7 – Deep drawing of a cylindrical cup: a) Heating the die and blank holder to 200 °C, punch kept at RT by water cooling system; b) Blank heating; c) Cup drawing; d) Ironing; e) Ejection step.

364 A schematic illustration of the procedure used in the warm forming of the cylindrical cup is
365 presented in Figure 7. The procedure starts with the heating of the die and the blank-holder by internal
366 electrical heating rods (Figure 7 a)), until the defined test temperature. The blank is lubricated with “Jelt
367 Grease 5411 aerosol 95cSt” and is positioned in the blank-holder (Figure 7 b)) to be heated. A reduced
368 heating time of approximately 60 s was selected to minimize the occurrence of microstructural changes.
369 During the forming, the punch is water cooled to maintain its temperature close to RT. The water flow
370 rate imposed within the cooling system was equal in all the tests. The forming operation presented
371 involves two distinct phases: (i) the drawing, as schematically shown in Figure 7 c); and the (ii) the
372 ironing, resulting from the fact that the thickening of the flange (see Figure 7 d)) is higher than the gap
373 between the die and the punch. When the punch reaches the displacement set it stops and the ejector is
374 actuated to remove the cup from the die cavity (Figure 7 e)). The clamping force of 6 kN was used in all
375 tests.

376 The dimensions of the tools and the blank are presented in Figure 8, as well as the position of the
377 Type K thermocouples. A thermocouple is placed on the die radius (labeled TC Die); another is right
378 below the ejector (labeled TC Punch) in the punch; and in the blank-holder the thermocouple is in the
379 flange zone (labeled TC B-H). On the blank, the thermocouples are initially placed at the center (labeled
380 TC Blank 1) and at 15 mm (limit of the punch radius) from the center of the blank (labeled TC Blank 2).
381 The thermocouples have a fixed position in the tools, thus, the temperature values correspond to a local
382 measure. On the other hand, the position of the blank thermocouples evolves relative to the tools during
383 the warm forming test, meaning that the temperature values measured are a function of the punch
384 displacement. Therefore, the analysis of the evolution of the temperature throughout the drawing test
385 must be performed considering the evolution of the relative position of the thermocouples. At the end of
386 the forming operation, the thermocouple TC Blank 1 is still positioned in the center of the bottom of the
387 cup while TC Blank 2 is located in the radius zone of the cup, which are zones submitted to low levels of
388 plastic strain.

Figure 8 – Tool dimensions (in mm) and detail of the position of the thermocouples.

The following variables were recorded in each test as a function of time: the punch force, the punch displacement, the blank-holder force and the temperature. For each condition, a minimum of three reproducible tests was performed. The reproducibility was validated by two conditions: (i) the average scatter of the punch force, which was less than ± 0.1 kN for the same displacement; and (ii) the difference in heating time was always inferior to ± 15 s. It is known that Type K thermocouples can present temperature measurement errors outside the standard limits, due to the drift in the Seebeck coefficient [37,38]. Nevertheless, under the previously specified conditions, the temperature results were always reproducible with temperature differences inferior to 2°C . Moreover, for the same test conditions, the temperature evolutions were similar for the alloy EN AW 6016-T4 for both natural aging periods (1 and 18 months) and for the alloy EN AW 6061-T6. Therefore, the results shown are only for one representative test for each condition.

The thickness of the cup and height were measured as shown in Figure 9 a). This was performed using a 3D measurement machine “Brown & Sharpe Mfg. Co.” model “MicroXcel PFX-454”. The coordinate measurements are performed along the interior and the exterior sides of the cup, in a normal direction to each surface, such that the difference between them corresponds to the thickness value. The measurements are performed at each 45° from 0° to 315° RD. To minimize measurement errors resulting from axisymmetric deviations, the thickness values presented correspond to the average of the vertical axis symmetries for the cup. According to Coër [36], the accuracy of the thickness measurement is a function of: the accuracy of the 3D measurement machine ($\pm 3 \mu\text{m}$), the control of the positioning of the

cup in the measurement device ($\pm 15 \mu\text{m}$), and of angular deviations in the RD direction ($\pm 5^\circ$). The cup height was measured along its edge, at each 5° . The results of the cup dimensions were highly reproducible within the same test conditions.

A springback analysis was performed using the split-ring test. In this test a ring is cut from the cup wall (trimming) and afterwards it is opened along the RD (split-ring). As shown by Grèze et al. [11], the difference between the ring's diameter before and after splitting (ring opening) is an indirect measure of the springback phenomenon, i.e. of the residual circumferential stresses present in the cup. As shown in Figure 9 b), three rings were cut from the cup wall, each with a height of 3 mm. This enables a more thorough analysis of the springback behavior since the magnitude of the ring opening is dictated by its vertical position along the cup wall, as previously shown by Xia et al. [39]. EDM was used to cut and open the rings. The extent of the ring opening was measured using a microscope to ensure a high degree of accuracy and the value measured was corrected taking into account the thickness of the EDM cut (0.3 mm).

Figure 9 – a) Cup dimensions; b) Split-ring springback test.

422

423 3.2 Results of the forming process

424 3.2.1 Temperature evolution

425 Figure 10 a) and b) present the temperature evolution of the blank and the tools, as a function of
 426 the punch displacement, respectively, for the three punch speeds. The results correspond to the EN AW
 427 6016-T4 alloy at 18 months of natural aging. Although not shown here, the same temperature evolution

occurs for the EN AW 6061-T6 alloy as a function of the punch speed. Moreover, whatever the punch speed or the alloy under study, the heating stage is similar for all tests.

a) Temperature evolution of the blank

b) Temperature evolution of the die and punch

Figure 10 – The temperature evolution as a function of the displacement of the punch, at 200 °C for the tested punch speeds of 0.1, 1 and 10 mm·s⁻¹.

430

431 The temperature evolution of the blank and the tools is highly dependent on the punch speed
 432 since the amount of heat transferred depends of the forming time, which is in a range between
 433 approximately 300 s (for 0.1 mm·s⁻¹) and 3 s (for 10 mm·s⁻¹). Globally, the blank's temperature is higher
 434 for the higher punch speeds. As shown in Figure 10 a), the shape of the temperature evolution curves is

quite similar for the punch speeds of 0.1 and 1 mm·s⁻¹. In the beginning of the forming stage, the temperature at the center of the blank (TC Blank 1) is slightly lower than at the area close to the flange zone (TC Blank 2) since there is some heat transfer by convection between the blank and the punch during the heating stage. In the drawing phase, the blank temperature (TC Blank 1 and TC Blank 2) is mostly coincident for both punch speeds. Nevertheless, it should be mentioned that the TC Die shows lower temperature values for the punch speed of 0.1 mm·s⁻¹, while the TC Punch presents higher ones, see Figure 10 b). In the ironing phase, the temperature in the tools and in the blank is lower at the punch speed of 0.1 mm·s⁻¹. Concerning the temperature evolution for a punch speed of 10 mm·s⁻¹, the TC Die presents a slight temperature decrease and the TC Punch a slight temperature increase. The temperature evolution in the TC Blank 1 and the TC Blank 2 presents an almost linear decrease, with a higher rate from the beginning of the test up to the instant of the maximum punch force. Thereafter, the temperature of the TC Blank 1 and the TC Blank 2 decreases linearly, but at a lower rate.

3.2.2 The punch force analysis

The influence of the punch speed and temperature on the punch force evolution, as a function of its displacement, is shown in Figure 11. Globally, the punch force decreases as the temperature increases from RT to 200 °C, which is linked to the softening behavior resulting from the temperature increase. Regarding the influence of the punch speed on the evolution of the punch force, globally the results show that at RT it has a negligible effect, while at 200 °C it is noticeable. This is valid for both alloys. Finally, for the EN AW 6016-T4 alloy, the natural aging (from 1 to 18 months) leads to higher punch forces at RT. However, at 200 °C the natural aging influence is negligible, whatever punch speed used, see Figure 11 a).

At 200 °C, both alloys show a slightly higher punch force with a higher punch speed for the first millimeters of the punch displacement. However, on attaining the plateau of maximal force, the highest punch force occurs for the lowest punch speed. For the EN AW 6016-T4, this inversion of the evolution of the punch force occurs at 10 mm of punch displacement, at about 90, 9 and 0.9 s for 0.1, 1 and 10 mm·s⁻¹ of punch speed, respectively. Moreover, for the punch speed of 0.1 mm·s⁻¹ an unusual increase in the drawing force is observed between 10 and 15 mm of punch displacement, which is analyzed in the discussion. For the EN AW 6061-T6 alloy, the inversion of the punch force occurs at 14 mm of punch displacement, at about 150, 15 and 1.5 s for 0.1, 1 and 10 mm·s⁻¹ of punch speed, respectively.

a) EN AW 6016-T4

b) EN AW 6061-T6

Figure 11 – The punch force evolution as a function of punch displacement.

3.2.3 The thickness analysis

The thickness evolution along the curvilinear distance from the center of the cup is presented in Figure 12. The thickness is presented only along RD since the analysis of the other profiles lead to similar conclusions. For the EN AW 6016-T4 alloy, the results presented in Figure 12 correspond to the tests performed only at 18 months of natural aging, since the natural aging effect does not induce significant dimensional differences, as shown in Simões et al. [12].

Both alloys present a similar trend for the thickness evolution, which is independent of the test temperature and the punch speed, see Figure 12. The thickness profile is constant at the bottom of the cup and close to the initial thickness value. The lower thickness values occur at the entrance to the punch radius (section b to c) and at the transition between the punch radius and the cup wall (ring 3 zone). After that, the thickness increases along the cup wall but attains a constant final value, due to the ironing phase. Comparing both alloys, it is visible that at the bottom of the cup and radius zone, the EN AW 6061-T6 alloy presents lower thickness values, which are related to the difference in the initial thickness values, see Table 1. However, the shape of the thickness profile is independent of the initial thickness value. Globally, for the same location in the cup, the thickness values are higher for the cups deep drawn at 200 °C. Regarding the punch speed, it has a negligible influence on the thickness of the cup at RT, while at 200 °C small differences in thickness are noted at the cup radius zone for the EN AW 6016-T4 alloy and at the ironed zone for both alloys. In these zones, globally, the thickness is higher for the cups performed at higher punch speeds. Although not shown here, the results confirm that there is a relation between the average cup height and the thickness distribution, i.e. greater heights occur associated with a lower average thickness distribution, as previously reported by Simões et al. [12]. Nevertheless, the punch speed induces only slight differences in the cup height.

Figure 12 – The influence of punch speed and temperature on the thickness evolution along the RD.

3.2.4 Springback analysis: the split-ring test

The measurements of the ring openings as a function of the punch speed are shown in Figure 13. Three rings were cut from the cup wall at various heights, as presented in Figure 9 b). For the same temperature and punch speed conditions, the EN AW 6061-T6 alloy always presents higher springback values than the EN AW 6016-T4 alloy due to its higher yield stress. In fact, it is known that, springback increases with the increase in the **flow** stress, assuming the same Young modulus value. At RT, for both alloys the punch speed has a negligible effect on the opening of the ring. While at 200 °C, for the EN AW 6016-T4 alloy, the ring opening decreases linearly with the increase in logarithmic punch speed. Moreover, at 200 °C the ring openings are very similar for the naturally aged material of 1 and 18 months. For the EN AW 6061-T6 alloy, the punch speed has a negligible effect on the ring opening at 200 °C.

Regarding the influence of the axial position of the ring in the cup wall, it may be seen that the ring opening decreases as the distance from the bottom of the cup increases. This trend occurs for both alloys and is independent of the punch speed and the forming temperature. According to Simões et al. [40], who analyzed the residual stress distributions for this benchmark test numerically, the ironing of the vertical wall changes the distribution of the residual axial and circumferential stresses along the cup wall, even for relatively small ironing strains. This alters the trend observed for the ring opening, when rings are cut at different heights. In fact, since the ironing phase imposes high contact forces on the cup wall, which reduces the tangential stress, ring 1 presents a lower springback variation for the various test conditions. The fact that ring 1 presents a negligible springback variation with the increase in punch speed, seems to be connected to the reduction in the gradient of the circumferential stress component throughout the thickness induced by the ironing phase.

a) EN AW 6016-T4

b) EN AW 6061-T6

Figure 13 – The ring opening as a function of the punch speed presented on a logarithmic scale. The green, blue and red symbols correspond to rings 1, 2 and 3, respectively. The colored lines correspond to tests performed at 200 °C while the black lines are used for tests at RT.

510

511 4 Analysis of results and discussion

512 The results presented in the previous subsection show that formability and springback remain
 513 stable or improve with the increase in punch speed. Nonetheless, as previously mentioned in the
 514 introduction, it has also been reported that the increase in the punch speed leads to a decrease in
 515 formability, for non-heat-treatable aluminum alloys. This was observed both under isothermal conditions

516 by Ayres and Wenner [19] and non-isothermal conditions by Naka and Yoshida [20]. Additionally, Moon
517 et al. [10] and Kim and Koç [24] reported an increase in springback with the increase in punch speed,
518 both under isothermal and non-isothermal conditions. On the other hand, Ghosh et al. [21] and Palumbo
519 et al. [22] have shown that for Al-Mg-Si alloys a low forming speed may be counterproductive in
520 formability due to dynamic precipitation. Thus, in the case of heat-treatable alloys, an analysis of
521 formability and springback results needs to consider the range of exposure time (heating and forming).

522 In warm forming under non-isothermal conditions, the temperature evolution depends on the
523 punch speed since this process parameter dictates the forming time. Figure 14 presents the temperature
524 evolution of the tools and blanks as a function of time for the three punch speeds. Concerning the
525 TC B-H, the temperature decreases smoothly as time increases, remaining mostly above 190 °C. The
526 temperature of the TC Punch is constant in the beginning (approximately 1 s) and afterwards it increases
527 as time increases. However, at punch speeds of 0.1 and 1 mm·s⁻¹, the temperature of the TC Punch
528 decreases during the ironing phase, highlighting the capability of the cooling system to dissipate heat for
529 higher forming times. The evolution of the temperature of the TC Die is more complex. For the three
530 punch speeds, the temperature of the TC Die presents an initial first period with a slight decrease,
531 followed by a faster decrease until attaining a plateau. This plateau corresponds to the instant that the die
532 and the punch shoulder radii are completely formed in the part, i.e. the instant of maximal punch force,
533 between 10 and 15 mm of punch displacement, see Figure 11. The high contact surface between the die
534 and the blank contributes to a stabilization of the punch and the die temperature, since the heat flows from
535 the flange zone, through the blank, and is dissipated by the punch cooling system, promoting a thermal
536 equilibrium. Afterwards, the temperature of the TC Die abruptly decreases during the ironing phase and
537 increases when the die loses contact with the blank. In fact, according to Chang et al. [41], a higher
538 contact pressure causes more deformation of the asperities at the interface (reducing the blank's surface
539 roughness), so the actual contact area enlarges and the thermal contact resistance decreases, resulting in
540 an increase in the interfacial heat transfer coefficient.

Figure 14 – Temperature evolution as a function of time (on a logarithmic scale), in the warm forming of a cylindrical cup at 200 °C. Tests performed on the EN AW 6016-T4 alloy, natural aged of 18 months.

541

542 The temperature difference between the TC Die and the TC Punch decreases as the punch speed
 543 decreases (see Figure 10 b) and Figure 14)). These results are in accordance with those presented by
 544 Ma et al. [42], reporting that a high punch speed can keep heat transfer between the tools at a minimum
 545 level during the hot forming process. In fact, during the forming process, the heat transfer rate between
 546 the cold punch and the heated flange zone is a function of the temperature difference between the tools
 547 and is also proportional to the time. Thus, the improvement in formability, related with the temperature
 548 difference, described by Palumbo and Tricarico [14], decreases as the punch speed decreases.

549 Figure 14 also highlights the blank exposure time, which is particularly important to analyze the
 550 occurrence of dynamic precipitation. At a punch speed of 10 mm·s⁻¹, the average blank temperature is at
 551 its highest, but for the shortest exposure time. For the other two punch speeds, the temperature of the
 552 TC Blank 1 is always under 150 °C; while the TC Blank 2 presents a temperature under 150 °C after 21.1
 553 and 2.2 s, for 0.1 and 1 mm·s⁻¹, respectively. Whatever the punch speed adopted, both thermocouples
 554 have a temperature below 100 °C after 20 mm of punch displacement (see Figure 10 a)), which
 555 corresponds to approximately one third of the forming time.

556 The punch speed directly influences the strain rate and, consequently, the mechanical behavior of
 557 the material. Despite their negligible strain-rate sensitivity at RT, both alloys present a positive strain-rate
 558 sensitivity in warm conditions during tensile tests, see Figure 4 b) and Figure 6. During warm cup

forming (see Figure 11), for the two alloys, this positive strain-rate sensitivity prevails until approximately 9 mm of punch displacement, resulting in higher punch forces linked with higher punch speeds. In fact, during this period the blank presents a large contact area with the heated flange, indicating that it keeps a relatively stable high temperature, whatever the punch speed (see Figure 14). Subsequently, from 9 to 19 mm of punch displacement, the blank progressively loses contact with the flange zone and its temperature is mainly imposed by the temperature difference between the die and the punch. During this period, the punch force is similar for each punch speed, with the exception of the test performed with the EN AW 6016-T4 alloy, with a punch speed of 0.1 mm s^{-1} , as already mentioned in section 3.2.2 (Figure 11 a)). Globally, the higher average temperature of the blank at the punch speed of 10 mm s^{-1} (see Figure 14) may contribute to reducing the forming force, since the material becomes softer as the temperature increases. However, the positive strain-rate sensitivity leads to the increase in the forming force, since the material becomes harder as the strain rate increases. For the lower punch speeds, the lower average temperature of the blank may contribute to the strengthening of the material, increasing the forming force. However, the forming force may reduce due to the lower strain rate attained by the material. In fact, it seems that the positive strain-rate sensitivity at higher punch speeds is counterbalanced by the hardening of the material due to the lower temperatures at the lower punch speed. Afterwards, during the ironing phase, a higher punch force is obtained with the lower punch speed for the two materials. This is related to the lower gap between the die and the punch that causes a harder ironing phase. In fact, the temperature dictates the thermal dilatation of the tools. As shown in Figure 14, the decrease in the punch speed causes a reduction in the temperature difference between the die and the punch, causing a reduction in the gap between these tools and leading to higher punch forces.

Regarding the tests performed at the punch speed of 0.1 mm s^{-1} for the EN AW 6016-T4 alloy, an unexpected increase in the punch force occurs between 10 to 16 mm of punch displacement (see Figure 11 a)), which is not observed for the other tests. Ghosh et al. [21] also observed a similar behavior in the warm forming at $250 \text{ }^{\circ}\text{C}$ of the same alloy, for a decrease in the punch speed from 1 to 0.18 mm s^{-1} . According to the authors, these observations can be explained in terms of dynamic precipitation, which causes the material to approach the T6 condition. For the test under analysis, the punch displacement between 10 to 16 mm corresponds to a time of about 100 s. Thus, the increase in the punch force can be understood with the analysis of the stress – strain curves, at a strain rate of $2 \times 10^{-4} \text{ s}^{-1}$ (see Figure 2 and Figure 4), where it is observed that under uniaxial tensile stress conditions an exposure time of about 80 s

589 is enough to promote the beginning of dynamic precipitation hardening. The influence of the dynamic
 590 precipitation on the material's hardening behavior is particularly evident in the stress relaxation test, see
 591 Figure 4. According to Cai et al. [27] and Kolar et al. [28], who studied the combined effect of
 592 deformation and precipitation on the mechanical properties of Al-Mg-Si alloys, the dislocations
 593 introduced by plastic deformation promote the growth of precipitates by a dislocation-assisted diffusion
 594 mechanism, reducing the time to reach the peak strength, as compared to the one associated with static
 595 aging. It should be mentioned that the range of the exposure time (heating and forming) of the cylindrical
 596 cup tests for the punch speeds used, matches the ones of the tensile tests for the strain rates selected. In
 597 fact, the warm forming of the cylindrical cup at a punch speed of $0.1 \text{ mm}\cdot\text{s}^{-1}$ requires a heating and
 598 forming time of approximately 60 and 300 s, respectively. Accordingly, the tensile test at the strain rate of
 599 $2\times 10^{-4} \text{ s}^{-1}$ requires 20 s for heating and 450 s to attain the R_m (see Figure 2). Although these exposure
 600 times may seem low when compared with the ones commonly used to perform the T6 heat treatment, the
 601 dislocation-assisted precipitation clearly reduces the time required for the precipitation hardening to
 602 occur. For the punch speeds of 1 and $10 \text{ mm}\cdot\text{s}^{-1}$, the forming time is approximately 30 and 3 s,
 603 respectively. Accordingly, the tensile tests for a strain rate of $2\times 10^{-3} \text{ s}^{-1}$ and $2\times 10^{-2} \text{ s}^{-1}$ require 65.6 and
 604 5.48 s to attain the R_m , respectively. The change in the strain-rate sensitivity observed between the tensile
 605 tests at $2\times 10^{-3} \text{ s}^{-1}$ and $2\times 10^{-2} \text{ s}^{-1}$ indicates that the dynamic precipitation also occurs at $2\times 10^{-3} \text{ s}^{-1}$ (see
 606 Figure 2). This change starts to occur for a test duration of approximately 30 s. Thus, although its effect is
 607 not so visible in the evolution of the punch force, the results indicate that dynamic precipitation hardening
 608 can also occur for the punch speed of $1 \text{ mm}\cdot\text{s}^{-1}$ since, as previously mentioned, the average temperature of
 609 the blank during the deep drawing stage is always between 200 and 150°C (see Figure 14).

610 The influence of temperature and punch speed on the cup thickness is noticeable in two distinct
 611 zones: (i) the cup radius zone; and (ii) the ironed zone (see Figure 12). According to Palumbo and
 612 Tricarico [14], the difference in thickness in the cup radius zone is linked to the temperature gradient of
 613 the tools. In fact, the cooling action exerted by the punch produces a strengthening effect on the material
 614 located at the bottom of the cup and in the radius zone. The softening of the material at the flange reduces
 615 the punch load required to attain a certain displacement and, consequently, the stress and strain levels also
 616 reduce in the cup radius zone. These effects can explain the slight reduction in the thickness at the cup
 617 radius as the punch speed decreases, observed at 200°C for the EN AW 6016-T4 alloy, see Figure 12.
 618 Regarding the ironed zone, as shown by Simões et al. [12], the temperature increase leads to the thermal

dilation of the tools and, consequently, to an increase in the gap between the die and the punch. Since it is the gap between the die and the punch that determines the final thickness, its increase explains the higher thickness values observed in the ironed zone of the warm formed cups. Additionally, a greater gap results in a softer ironing phase with lower punch forces (see Figure 11) and, consequently, lower stress and strain levels in the cup radius zone, which also contributes to increasing the thickness in the cup radius zone.

Concerning the influence of punch speed on the springback, at RT both alloys show a negligible variation of the opening with the increase in punch speed (see Figure 13), as a result of their negligible strain-rate sensitivity at RT (see Figure 2 and Figure 6). At 200 °C, the EN AW 6061-T6 alloy also presents a negligible variation in springback with the increase in punch speed. This can be related with the behavior observed in the stress relaxation tests (see Figure 6), characterized by a sudden stress drop upon unloading, followed by a stress value upon reloading which is consistent with the one attained in the monotonic test performed at the same strain rate. Although the increase of the punch speed leads to a higher strain rate for a point initially located in the flange, which is kept almost constant during its sliding over the die radius, it suddenly decreases when the point reaches the cup wall until attaining a null value [8]. This results in a stress distribution in the cup wall that tends to be similar, whatever the punch speed. Therefore, although this alloy has a positive strain-rate sensitivity (see Figure 6), there is no clear increase in springback with the increase in the punch speed. Another reasonable explanation for this may be linked with the reduction in the temperature of the blank and the increase in the ironing forces for lower punch speeds. In fact, the increase in the material strength caused by the positive strain-rate sensitivity, at higher punch speeds, may be counterbalanced by the higher material strength caused by the lower temperature at lower punch speeds (see Figure 14). On the other hand, at 200 °C, the EN AW 6016-T4 alloy shows an increase in springback as the punch speed decreases. Although both alloys present a positive strain-rate sensitivity, for the EN AW 6016-T4 alloy the springback behavior is mainly dictated by the occurrence of dynamic precipitation hardening, which leads to higher material strength and, consequently, higher springback.

5 Conclusions

The results presented clearly show that under non-isothermal conditions, for heat-treatable alloys such as the Al-Mg-Si alloys, the formability and springback remain stable (for the EN AW 6061-T6 alloy) or decrease (for the EN AW 6016-T4 alloy) as the punch speed increases. Nevertheless, previous results presented in the literature for non-heat-treatable aluminum alloys report a decrease in formability and a springback increase with the increase in the punch speed, both under isothermal and non-isothermal conditions. This behavior corresponds to the one expected for materials presenting a positive strain-rate sensitivity, as for the alloys under analysis.

The present study highlights the fact that the influence of the punch speed can be divided in its impact on the behavior of the material and on the temperature gradient in the blank. Globally, the heat transfers that occur during the process are mostly dependent on the forming time. The higher punch speed reduces the exposure time and consequently reduces the heat exchanges. Thus, at higher punch speed, the temperature values of the tools remain nearly constant during the warm forming. On the other hand, at lower punch speeds, the temperature of the blank decreases and consequently the strength of the material increases, which can reduce the effect of the material positive strain-rate sensitivity. Additionally, a low punch speed contributes to decreasing the advantage of the temperature gradient in the blank, which may result in more thinning. This may explain why the formability and springback remained stable for the EN AW 6061-T6 alloy. For the EN AW 6016-T4 alloy, a high punch speed value also avoids dynamic precipitation that causes additional hardening, reducing the formability and increasing the springback. In this sense, a high punch speed is advantageous in the warm forming of heat-treatable alloys under non-isothermal conditions, because it allows a better control of the process, including on the temperature difference between the tools.

Previous results highlighted the potential of warm forming of heat-treatable alloys, under non-isothermal conditions in a temperature range from 200 to 250 °C, to increase the formability, reduce the springback, and minimize the influence of natural aging in sheet metal forming parts variability [12]. The present study underlines the need to consider the influence of the strain rate in the design of warm forming process, particularly when considering heat-treatable alloys, since it also affects the exposure time. This allows exploiting the advantages of warm forming using a higher punch speed, even for alloys

674 with a positive strain rate sensitivity, which is advantageous in the industrial field since it makes higher
675 production rates possible.

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692 **References**

- 693 [1] Hirsch J. Aluminium in innovative light-weight car design. *Mater Trans* 2011;52:818–824.
- 694 [2] Harrison NR, Friedman PA, Pan J. Warm forming die design, Part III: Design and validation of a
695 warm forming die. *J Manuf Process* 2015;20:356–66. doi:10.1016/j.jmapro.2015.01.006.
- 696 [3] Pourboghrat F, Venkatesan S, Carsley JE. LDR and hydroforming limit for deep drawing of
697 AA5754 aluminum sheet. *J Manuf Process* 2013;15:600–15. doi:10.1016/j.jmapro.2013.04.003.
- 698 [4] Harrison NR, Rubek V, Friedman PA. Warm forming die design, part I: Experimental validation of
699 a novel thermal finite element modeling code. *J Manuf Process* 2013;15:263–72.
700 doi:10.1016/j.jmapro.2013.01.001.
- 701 [5] J. Martins, D.M. Neto, J.L. Alves, M.C. Oliveira, H. Laurent, A. Andrade-Campos, et al. A new
702 staggered algorithm for thermomechanical coupled problems. *Int J Solids Struct* 2017;122–123:42–
703 58. doi:10.1016/j.ijsolstr.2017.06.002.
- 704 [6] Laurent H, Coër J, Manach PY, Oliveira MC, Menezes LF. Experimental and numerical studies on
705 the warm deep drawing of an Al–Mg alloy. *Int J Mech Sci* 2015;93:59–72.
706 doi:10.1016/j.ijmecsci.2015.01.009.
- 707 [7] Abovyan T, Kridli GT, Friedman PA, Ayoub G. Formability prediction of aluminum sheet alloys
708 under isothermal forming conditions. *J Manuf Process* 2015;20:406–13.
709 doi:10.1016/j.jmapro.2014.08.003.
- 710 [8] Neto DM, Martins JMP, Cunha PM, Alves JL, Oliveira MC, Laurent H, et al. Thermo-mechanical
711 finite element analysis of the AA5086 alloy under warm forming conditions. *Int J Solids Struct*
712 2017. doi:10.1016/j.ijsolstr.2017.06.011.
- 713 [9] Bolt PJ, Lamboo NAPM, Rozier PJCM. Feasibility of warm drawing of aluminium products. *J*
714 *Mater Process Technol* 2001;115:118–121.
- 715 [10] Moon YH, Kang SS, Cho JR, Kim TG. Effect of tool temperature on the reduction of the
716 springback of aluminum sheets. *J Mater Process Technol* 2003;132:365–8. doi:10.1016/S0924-
717 0136(02)00925-1.
- 718 [11] Grèze R, Manach PY, Laurent H, Thuillier S, Menezes LF. Influence of the temperature on residual
719 stresses and springback effect in an Aluminium alloy. *Int J Mech Sci* 2010;52:1094–100.
720 doi:10.1016/j.ijmecsci.2010.04.008.
- 721 [12] Simões VM, Laurent H, Oliveira MC, Menezes LF. The influence of warm forming in natural
722 aging and springback of Al–Mg–Si alloys. *Int J Mater Form* 2018. doi:10.1007/s12289-018-1406-7.
- 723 [13] Wilson DV. Aluminium versus steel in the family car — the formability factor. *J Mech Work*
724 *Technol* 1988;16:257–77. doi:10.1016/0378-3804(88)90055-1.
- 725 [14] Palumbo G, Tricarico L. Numerical and experimental investigations on the Warm Deep Drawing
726 process of circular aluminum alloy specimens. *J Mater Process Technol* 2007;184:115–23.
727 doi:10.1016/j.jmatprotec.2006.11.024.
- 728 [15] Panicker SS, Singh HG, Panda SK, Dashwood R. Characterization of Tensile Properties, Limiting
729 Strains, and Deep Drawing Behavior of AA5754-H22 Sheet at Elevated Temperature. *J Mater Eng*
730 *Perform* 2015;24:4267–82. doi:10.1007/s11665-015-1740-6.
- 731 [16] Li D, Ghosh AK. Biaxial warm forming behavior of aluminum sheet alloys. *J Mater Process*
732 *Technol* 2004;145:281–93. doi:10.1016/j.jmatprotec.2003.07.003.
- 733 [17] Li D, Ghosh A. Tensile deformation behavior of aluminum alloys at warm forming temperatures.
734 *Mater Sci Eng A* 2003;352:279–86. doi:10.1016/S0921-5093(02)00915-2.
- 735 [18] Shehata F, Painter MJ, Pearce R. Warm forming of aluminium/magnesium alloy sheet. *J Mech*
736 *Work Technol* 1978;2:279–90. doi:10.1016/0378-3804(78)90023-2.
- 737 [19] Ayres RA, Wenner ML. Strain and strain-rate hardening effects in punch stretching of 5182-0
738 aluminum at elevated temperatures. *Metall Trans A* 1979;10:41–6. doi:10.1007/BF02686404.
- 739 [20] Naka T, Yoshida F. Deep drawability of type 5083 aluminium–magnesium alloy sheet under
740 various conditions of temperature and forming speed. *J Mater Process Technol* 1999;89–90:19–23.
741 doi:10.1016/S0924-0136(99)00057-6.
- 742 [21] Ghosh M, Miroux A, Werkhoven RJ, Bolt PJ, Kestens LAI. Warm deep-drawing and post drawing
743 analysis of two Al–Mg–Si alloys. *J Mater Process Technol* 2014;214:756–66.
744 doi:10.1016/j.jmatprotec.2013.10.020.
- 745 [22] Palumbo G, Piccininni A, Guglielmi P, Di Michele G. Warm HydroForming of the heat treatable
746 aluminium alloy AC170PX. *J Manuf Process* 2015;20, Part 1:24–32.
747 doi:10.1016/j.jmapro.2015.09.012.

- [23] Fan X, He Z, Zhou W, Yuan S. Formability and strengthening mechanism of solution treated Al–Mg–Si alloy sheet under hot stamping conditions. *J Mater Process Technol* 2016;228:179–85. doi:10.1016/j.jmatprotec.2015.10.016.
- [24] Kim HS, Koç M. Numerical investigations on springback characteristics of aluminum sheet metal alloys in warm forming conditions. *J Mater Process Technol* 2008;204:370–83. doi:10.1016/j.jmatprotec.2007.11.059.
- [25] Engler O, Hirsch J. Texture control by thermomechanical processing of AA6xxx Al–Mg–Si sheet alloys for automotive applications—a review. *Mater Sci Eng A* 2002;336:249–62. doi:10.1016/S0921-5093(01)01968-2.
- [26] Banhart J, Chang CST, Liang Z, Wanderka N, Lay MDH, Hill AJ. Natural Aging in Al-Mg-Si Alloys - A Process of Unexpected Complexity. *Adv Eng Mater* 2010;12:559–71. doi:10.1002/adem.201000041.
- [27] Cai M, Field DP, Lorimer GW. A systematic comparison of static and dynamic ageing of two Al–Mg–Si alloys. *Mater Sci Eng A* 2004;373:65–71. doi:10.1016/j.msea.2003.12.035.
- [28] Kolar M, Pedersen KO, Gulbrandsen-Dahl S, Marthinsen K. Combined effect of deformation and artificial aging on mechanical properties of Al–Mg–Si Alloy. *Trans Nonferrous Met Soc China* 2012;22:1824–30. doi:10.1016/S1003-6326(11)61393-9.
- [29] ISO. ISO 6892-1:2009 Metallic materials - Tensile testing - Part 1: Method of test at room temperature. Geneva, Switzerland: International Organization for Standardization; 2009.
- [30] Zhang P, Li Z, Liu B, Ding W. Tensile Properties and Deformation Behaviors of a New Aluminum Alloy for High Pressure Die Casting. *J Mater Sci Technol* 2016. doi:10.1016/j.jmst.2016.02.013.
- [31] Zhong H, Rometsch P, Cao L, Guo F, Muddle B. Tensile properties and work hardening behaviour of alloy 6016 in naturally aged and pre-aged conditions. *Proc 12th Int Conf Alum. Alloys* Yokohama, Tokyo: Japan Institute of Light Metals; 2010, p. 2203–8.
- [32] Coër J, Bernard C, Laurent H, Andrade-Campos A, Thuillier S. The Effect of Temperature on Anisotropy Properties of an Aluminium Alloy. *Exp Mech* 2011;51:1185–95. doi:10.1007/s11340-010-9415-6.
- [33] Ghosh M. Microstructurally Controlled Mechanical Properties of Al-Mg-Si Alloys for Warm Forming Applications. PhD-thesis. Delft University of Technology, 2011.
- [34] Laurent H, Simões VM, Oliveira MC, Menezes LF. The influence of warm forming conditions on the natural aging and springback of a 6016-T4 aluminum alloy. *IOP Conf Ser Mater Sci Eng* 2018;418:012020. doi:10.1088/1757-899X/418/1/012020.
- [35] Picu RC, Ozturk F, Esener E, Li R. Aluminum Alloys with Identical Plastic Flow and Different Strain Rate Sensitivity. *Metall Mater Trans A* 2010;41:3358–64. doi:10.1007/s11661-010-0423-z.
- [36] Coër J. Mise en forme par emboutissage en température d'un alliage d'aluminium AA5754-O. PhD-thesis. Université de Bretagne Sud, 2013.
- [37] Webster ES. Seebeck Changes Due to Residual Cold-Work and Reversible Effects in Type K Bare-Wire Thermocouples. *Int J Thermophys* 2017;38:135. doi:10.1007/s10765-017-2269-8.
- [38] Greenen AD, Webster ES. Thermal Recovery from Cold-Working in Type K Bare-Wire Thermocouples. *Int J Thermophys* 2017;38:179. doi:10.1007/s10765-017-2316-5.
- [39] Xia ZC, Miller CE, Ren F. Springback Behavior of AA6111-T4 with Split-Ring Test. In: Ghosh S, Castro JC, Lee JK, editors. *Proc 8th Int Conf Numer. Methods Ind. Form. Process*. Columb. OH, vol. 712, New York: American Institute of Physics; 2004, p. 934–9. doi:10.1063/1.1766647.
- [40] Simões VM, Oliveira MC, Neto DM, Cunha PM, Laurent H, Alves JL, et al. Numerical study of springback using the split-ring test: influence of the clearance between the die and the punch. *Int J Mater Form* 2018;11:325–337. doi:10.1007/s12289-017-1351-x.
- [41] Chang Y, Tang X, Zhao K, Hu P, Wu Y. Investigation of the factors influencing the interfacial heat transfer coefficient in hot stamping. *J Mater Process Technol* 2016;228:25–33. doi:10.1016/j.jmatprotec.2014.10.008.
- [42] Ma W, Wang B, Fu L, Zhou J, Huang M. Influence of process parameters on deep drawing of AA6111 aluminum alloy at elevated temperatures. *J Cent South Univ* 2015;22:1167–74. doi:10.1007/s11771-015-2630-7.