Formability of a high-strength Al–Mg6.8 type alloy sheet

E. ROMHANJI, M. POPOVIĆ, D. GLIŠIĆ, V. MILENKOVIĆ
Faculty of Technology and Metallurgy, University of Belgrade, Karnegijeva 4, POB 494 Belgrade 11001, Yugoslavia
E-mail: TMF@ELAB.TMF.BG.AC.YU

Room temperature formability testing was performed on an AlMg6.8 type alloy sheet with a fully recrystallized structure (average grain diameter $\sim 18\ \mu m$) and after partial annealing with a retained deformed structure. The yield strengths attained after full recrystallization and after partial annealing, were 175 and 283 MPa respectively. Such an increase in strength is followed by formability degradation, maximized around the plain strain state to either 42%, as obtained using the limiting dome height test (LDH), or 35% after using forming limit curves (FLC). A comparison with known high-strength formable alloys has shown that the tested alloy in the recrystallized condition has a better stretch formability (at the same or even higher yield stress level), while in the unrecrystallized-partially annealed condition it has a lower formability, limiting its application to moderate forming requirements for very high-strength parts. © *1998 Chapman & Hall*

1. Introduction

Research in the development of high strength-to-weight ratio materials is of great interest to the aerospace and automotive industries. A number of aluminium alloys have been investigated as potential replacements for steel in the manufacture of car bodies [1–3] and aircraft parts [4–6] due to the potentially significant weight savings at the correct strength levels. It is worth noting some other advantages of aluminium such as a high corrosion resistance [1, 7] and a good weldability [1]. These motives fuelled studies into the optimization of known systems and also the development of new alloys to meet the basic requirements of a high strength and a good formability. Systems under particular consideration include those based on the strengthening effect of magnesium addition (the 5000 series of non-heat-treatable alloys), and Al–Cu, Al–Mg–Si–Cu and Al–Mg–Si alloys (the 2000 and 6000 series) which strengthen samples by precipitation of particles (heat treatable alloys). Recent developments include an Al–Mg–Mn alloy (AA5654 grade) of autobody stamping grade, with promising stretch formability [8], and a new series of 6000 alloys that have a promising combination of strength and formability properties, attaining the desired strength level even after a paint-bake [9].

In non-heat-treatable Al–Mg based alloys, in addition to the strengthening effect of magnesium, annealing conditions, i.e., recovery degrees can also be exploited to make partially hardened tempers starting from cold worked material and softening to the required properties [10].

The present paper presents results concerning the formability analysis of Al–Mg6.8 type alloy samples with in one case a recrystallized structure, and in another case a partially annealed sample that had a partly retained work-hardened strength.

2. Experimental procedures

2.1. Materials

The as-received Al–Mg sheet was 3.0 mm thick, in the annealed condition, with a chemical composition as listed in Table I. The grain size was about 35 μm which was homogeneously distributed.

The as-received material was further cold rolled from 3.0 to 0.9 mm (70%), and some samples partially annealed at 260 °C for 3 h (close to the H26 temper, designated as series B) and 320 °C for 3 h, attaining a fully recrystallized condition (designated as series A) in an inert gas atmosphere. Thus, the B series samples retained the original pancake structure while the A series samples underwent recrystallization with an average grain diameter of 18 μm.

2.2. Testing

2.2.1. Tensile test

Tensile tests were performed at a crosshead speed of $10\ \mathrm{mm\,min^{-1}}$ using ASTM sheet specimens with a gauge length of 25 mm and a 6.25 mm width, oriented at 0, 45, and 90° to the rolling direction. The instantaneous strain hardening exponent $n' = d(\ln \sigma)/d(\ln \varepsilon)$ was calculated using the best fit of stress–strain (σ–ε) values.

2.2.2. Forming limits

Gridded rectangular blanks of various widths (from 20–150 mm) were firmly clamped in the long

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TABLE I Chemical composition [mass %]

Mg	Mn	Si	Fe	Zn	Ti	Cu	Pb	Cr	Ni	Al
6.8	0.5	0.1	0.2	0.03	0.054	0.001	0.002	0.001	0.005	balance

TABLE II Tensile properties

Material		Yield strength, YS (MPa)		Ultimate tensile strength, UTS (MPa)		ε_t (%)		n	
	0°	177.2		350.1		24.6		0.229	
A	45°	177.0	175.4*	344.0	344.5*	25.4	24.6*	0.231	0.227*
	90°	170.3		339.9		23.2		0.218	
	0°	277.5		433.2		12.6		0.105	
B	45°	280.0	283.0*	415.5	423.2*	13.8	13.7*	0.141	0.125*
	90°	294.5		428.5		14.6		0.115	

*Average value: $X_{AV} = (X_0 + 2X_{45} + X_{90})/4$; n-terminal values ($n \equiv n'_{\text{max. load}}$).

direction, and stretched in an Erichsen sheet metal testing machine over a 75 mm diameter, unlubricated, hemispherical punch as proposed by Ghosh [11]. For every specimen, the dome height at maximum load and the minor strain, ε_2, in the necked region were measured. The limiting dome height (LDH) values were normalized with respect to the punch radius, and plotted against the respective minor strains (ε_2). Forming limit curves (FLC) were evaluated by plotting the combination of major and minor strains obtained from the same specimens. The grid circles selected to measure the strain are those situated in the region of fracture, region of localized necking and a region of uniform deformation. Additional stretching was performed on full sized samples using a combination of polyethylene sheet and mineral oil used as a lubricant, in order to extend the equibiaxial part of the FLC. Finally, the FLC were drawn following the procedure proposed by Hecker [12], taking the necking values as being the forming limit.

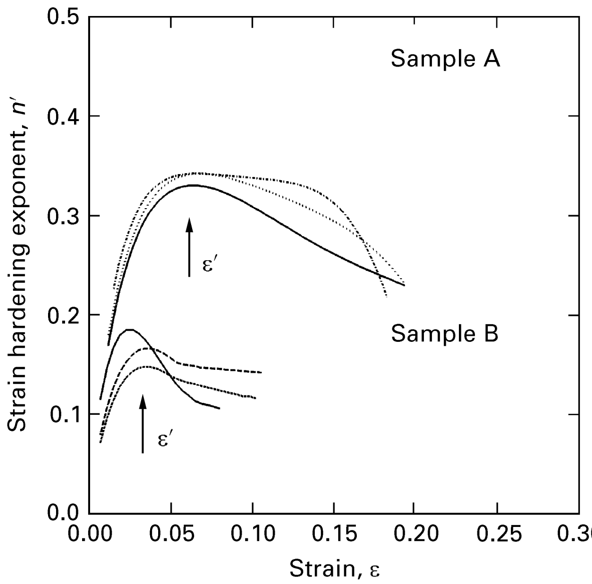

Figure 1 Instantaneous strain hardening exponent (n') variation during straining of A and B sheet samples. For sample A the data were taken at: (—) 0°, (- - - -) 45° and (· · · ·) 90° and for sample B at: (—) 0°, (- - - -) 45° and (· · · ·) 90° to the rolling direction.

3. Results

3.1. Tensile properties

The mechanical properties of the tested materials, listed in Table II differ a great deal, depending on the achieved degree of recovery. The partially annealed samples (B) have a considerably higher strength (the yield strength (YS) increases by about 38% and ultimate tensile strength (UTS) by about 18%) and a lower ductility as compared to the recrystallized samples (total elongation, ε_t, was lowered by about 40%). The Lüder's deformation was eliminated in the case of the partially annealed material, while it reached ~1% in the case of the A samples that had a fully recrystallized structure.

For both the A and B conditions, the instantaneous strain hardening exponent, n', varied during straining (Fig. 1). In both cases the value of n' quickly increased at low strains (approaching the maximum at $\varepsilon' \approx 0.06$ for sample A and $\varepsilon' \approx 0.025$ for sample B), but further straining of the sample resulted in a decrease in this property indicating a gradual exhausting of the hardening ability. The overall n' level, or the terminal values n (Table II), clearly reveal the superior hardening ability of the recrystallized A samples.

3.2. Forming limits

The normalized limiting dome height (LDH/R) versus critical minor strain (ε_2) curves plotted in Fig. 2 show a considerably higher ductility for the recrystallized material over the entire range of strain states tested. The maximum difference of 42% in the LDH appeared for $\varepsilon_2 = 0$. Similar results are obtained using the FLC criterion (a maximum difference of 35% in the radial peak strain, ε_1, had a value in the range around $\varepsilon_2 = 0$, Fig. 3).

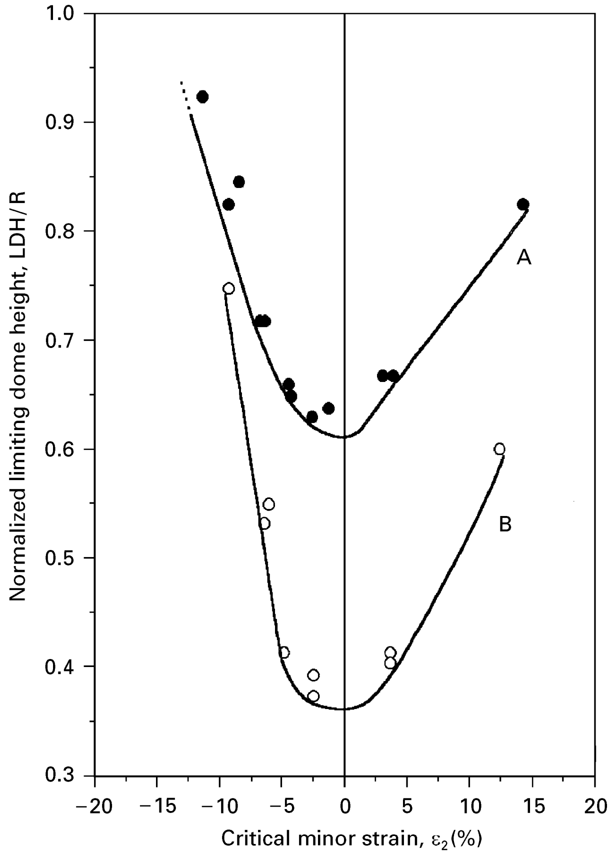

Figure 2 Normalized limiting dome height (LDH/R) versus critical minor strain ε_2 for 0.9 mm thick (○) A and (●) B sheet samples.

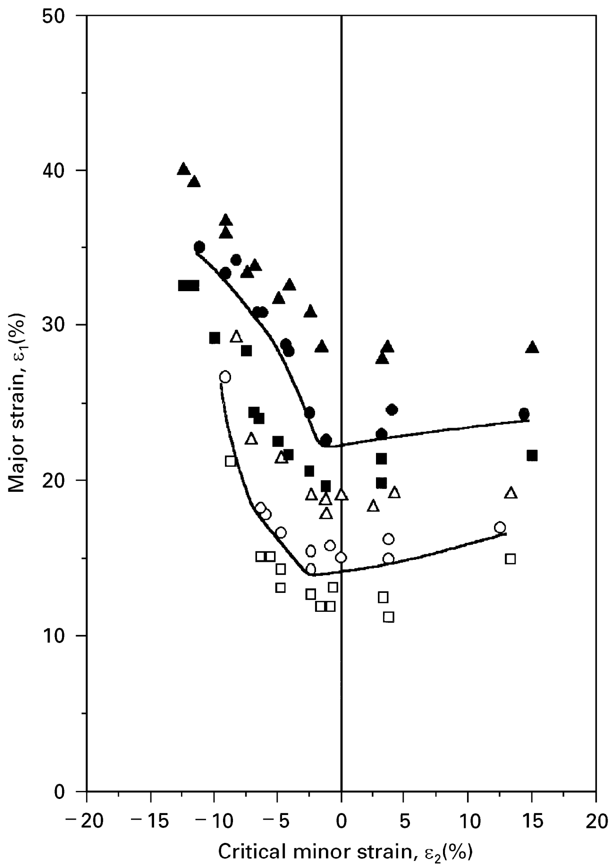

Figure 3 Forming limit curves for 0.9 mm thick A and B sheet samples. For sample A: (■) safe, (●) necking and (▲) fracture and for sample B: (□) safe, (○) necking and (△) fracture.

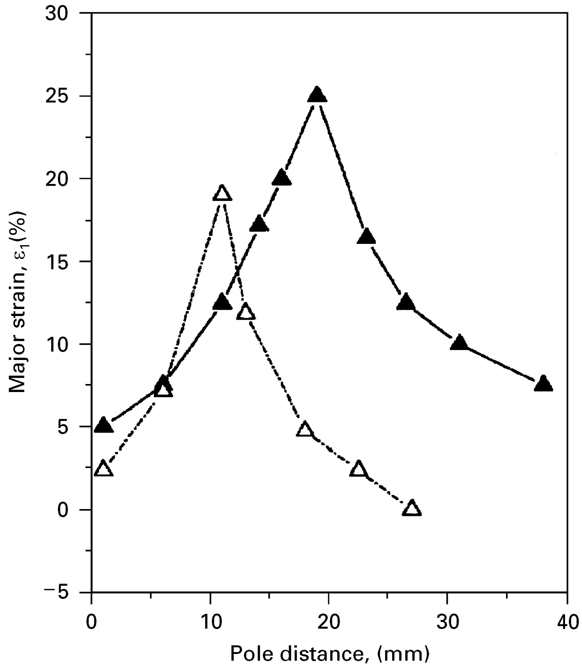

Figure 4 Major strain (ε_1) distribution for plane-strain state ($\varepsilon_2 = 0$). (▲) Sample A and (△) sample B.

The major strain (ε_1) distribution, after stretching A and B sheet specimens, (blank sizes were 150×95 mm), thereby simulating the plane-strain state is shown in Fig. 4. The failure site is displaced away from the pole to a greater extent in the case of sample A thereby attaining higher critical peak strains. The strain distribution is also more uniform for sample A and the area under the distribution curve is larger. A comparison is made with some formable high-strength aluminium alloys, taking the available limiting dome heights (Fig. 5) and FLC's (Fig. 6) from the literature [11, 13]. The mechanical properties of these alloys are listed in Table III. It should be noted that the tensile strength value of Mg-based non-heat-treatable alloys (5085-O, 5182-O and AlMg6) are lower than for the tested AlMg6.8 alloy, even for the A condition with a recrystallized structure. In the case of the partially annealed material (series B) this difference is further increased (the YS is twice that of the Mg alloys in Table III).

The LDH/R values for the alloys chosen from the literature fall between the two conditions of the tested AlMg6.8 alloy (Fig. 5), while their FLC's are roughly level with that of the A condition (Fig. 6). It must also be noted that the LDH/R and FLC values, taken from the literature, were obtained using a hemispherical punch with a 102 mm diameter.

4. Discussion

The two tested alloys show great differences in their achieved strength level and also their levels of formability. Using partial annealing, a remarkably higher strength (the YS rises from 175.5 to 283.0 MPa) is accompanied by a serious loss in the formability which, assessed by the LDH change, seems to reach its maximum value around the plane-strain state (Fig. 2).

This effect is mirrored in the FLC values (although the two FLC's are rather parallel, Fig. 3). The applied partial annealing condition allowed the retention of a partly work-hardened strength. Thus, the reduced overall level of the n' values (Fig. 1), and hence the tensile elongation (Table II), during straining of the B samples, as compared to the recrystallized A samples, can be rationalized by taking into account the higher dislocation density retained after the partial annealing. A steep increase in n' values at low strains and a decrease at higher strains is common to both conditions, but in the B samples the hardening ability is apparently exhausted earlier than in the case of the A samples (compare the values of ε' in Fig. 1). The influence of the strain-rate hardening is not discussed, since for the tested alloy, the total strain rate sensitivity was found to be $m = -0.01 \div -0.015$ [14], which is a typical value for Al–Mg alloys and is due to dynamic strain ageing [15]. Therefore, post-uniform straining is eliminated as a relevant value. The lowering of n' values at higher strains is assumed rather to be a result of strain rate softening because of the negative values of m, as has been discussed by Gosh [16]. In any case the higher dislocation density in the partially annealed structure is also thought to be more favourable for flow softening by shear banding [17, 18], influencing the early drop in the n' value in the B samples.

The strain states at failure in most stamped parts are usually very close to the plane strain condition ($\varepsilon_2 = 0$) and it has been estimated that 85% of all ductile failures occur in the region $0\% < \varepsilon_2 < 5\%$ [19]. So, any stretch formability analysis should primarily be concentrated on the plane strain region and thus the 42% decrease in the dome height at around $\varepsilon_2 = 0$ seems to be a relevant assessment of the stretchability loss after partial annealing of the tested AlMg6.8 alloy. Although the LDH was pointed out as being a highly suitable stretchability rating mode,

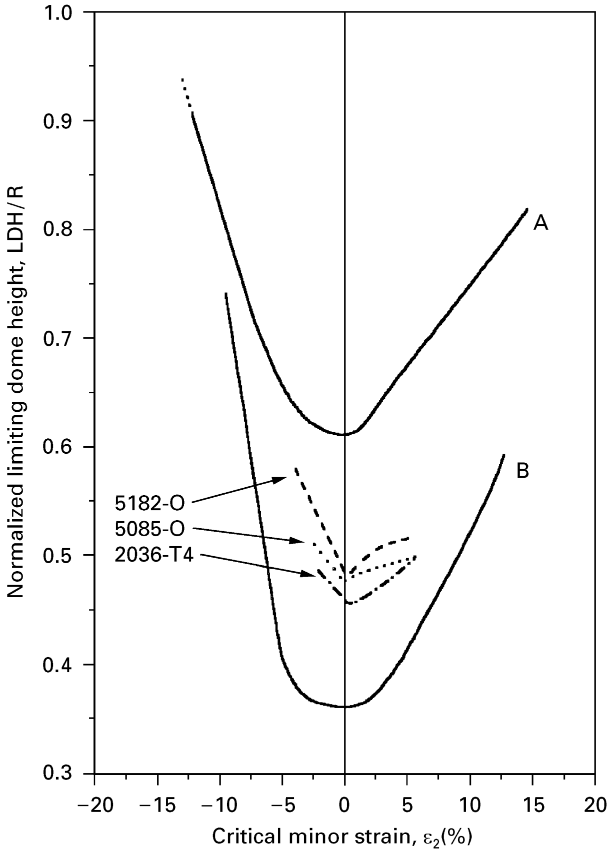

Figure 5 LDH/R values for some Al alloys from Gosh [11] compared to the results from Fig. 2. (––) 5182-O 0.89 mm thick, (- - -) 5085-O 1.09 mm thick and (– – –) 2036-T4 1.01 mm thick.

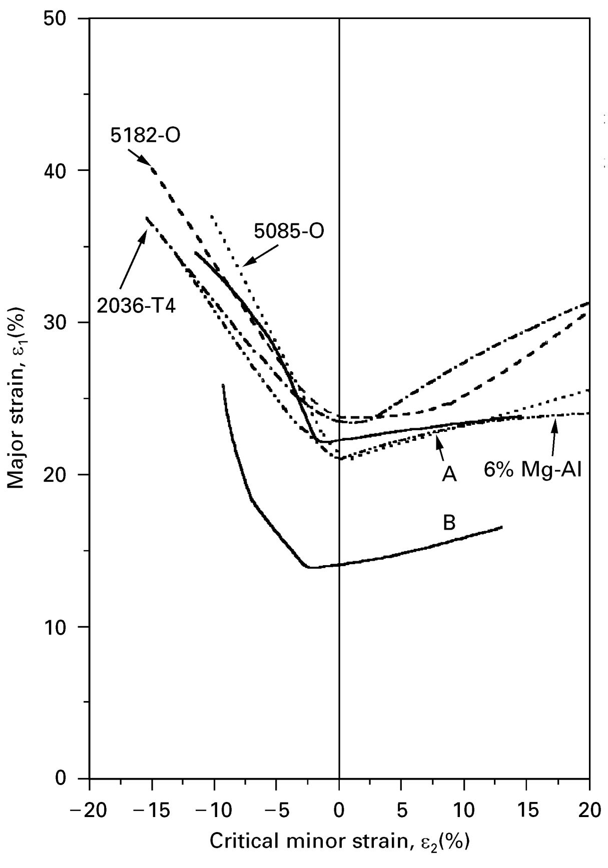

Figure 6 FLC's of some Al alloys from Gosh [11] compared to the results from Fig. 3. (– – –) 5182-O 0.89 mm thick, (- - - -) 5085-O 1.09 mm thick, (– - – -) 2036-T4 1.01 mm thick and (– - - -) 6 wt % Mg-Al.

TABLE III Mechanical properties

Alloy	Yield strength, YS (MPa)	Ultimate tensile strength UTS (MPa)	ε_t (%)	n	Reference
5085-O	139.5	320.3	27.6	0.266	11
5182-O	142.4	335.3	25.3	0.240	–
2036-T4	189.8	358.5	23.0	0.166	–
AlMg6-H111	163.0	311.9	34.2	0.285	13

n-terminal value; test specimen gauge length 50.8 mm and a 12.5 mm width.

since it includes both the limit strain and the strain distribution [11], in the FLC representation, the plane strain stretchability also appeared to be lowered by 35%. This similarity in stretch formability assessment is due to a large difference in general strain-hardening ability (Fig. 1). The 7% difference between the values produced by the two stretchability testing modes is thought to be due to a difference in the strain distribution in the two conditions. In plane-strain stretching the peak strain for the A samples, with a better hardening ability, is located closer to the edge than in case of B samples (Fig. 4). This is a result of a stronger frictional effect between the punch and sheet metal on retarding the deformation transmission toward the pole in materials with a higher strain-hardening ability, as has been previously discussed by Yoshida and Miyauchi [20]. The difference in area under the strain distribution curves for the two materials clearly reflects the higher stretching depth achieved in the A samples. This is probably due to the fact that the diffuse necking is controlled by the terminal n value [16]. The less uniform strain distribution around the peaks for the B samples is thought to reflect the noted difference in plane strain stretchability assessment using the LDH and FLC criterions.

Concerning the surface quality, the Lüder's elongation of about 1% in the A samples leads to the appearance of the well known "wedge" or "A" type sheet surface markings, leading to kinking which is often very undesirable for forming applications [1, 21]. The stretch markings were eliminated by an applied partial annealing.

Comparing the data in Tables II and III it is apparent that the tested alloy with the recrystallized structure (A) achieved a YS higher than the values listed in Table III for the Mg based formable alloys. It is difficult to compare the uniaxial ductility (ε_t) since the tensile test specimens were of different sizes. The normalized limiting dome heights of the 5182-O, 5085-O and 2036-T4 alloys are higher than that for the B condition but, at the same time, they are lower than the values attained for the A condition (Fig. 5). For the sake of valid comparison, it should be noted that the difference in the punch diameters used (75 mm in this work and 102 mm in work reported in the literature) could influence the relative position of the LDH/R curves. The normalization after using smaller diameter punches could exaggerate the curvature influence on formability to an extent that is non proportional to the appropriate stretch depth decrease [22]. It is thought that the present difference in punch diameters does not lead to any serious errors since the discussed influence becomes important at larger differences in punch diameters [22]. This can be easily justified if comparison is made using the stretch depth-LDH values without normalization. In this case, the three alloys taken from the literature approach the level for material A, but taking into account the lowering of the stretch depth due to the smaller punch diameter, the stated estimation of the formability ranking is further confirmed.

The peak strain criterion (FLC) did not reveal any such difference and the data taken from literature are rather close to the FLC obtained for the tested alloy with the recrystallized structure (Fig. 6).

In general, after testing this alloy, the observed formability changes over the wide strength range achieved by full recrystallization and partial recovery, allow us to produce different conditions between the two tested samples which seems to be very attractive in satisfying the different forming severity requirements in producing light-weight high-strength parts.

5. Summary

A stretch formability analysis was performed on 0.9 mm thick AlMg6.8 alloy sheets. One sheet had a fully recrystallized structure (18 μm average grain diameter after annealing at 320 °C for 3 h), whilst another sheet underwent partial annealing (260 °C for 3 h) and retained its basic deformed structure. The formability analysis is concentrated on the plane-strain region since it is the most rigorous condition, using both the normalized limiting dome height (LDH/R) and forming limit criterion (FLC).

The 175 MPa high yield stress value in the recrystallized AlMg6.8 alloy sheet was increased to 283 MPa by applying the partial annealing, but the plane strain stretchability measured by the LDH mode was lowered by 42%, whereas the FLC criterion showed a lowering of only 35%. The 7% difference between the two values obtained using the two testing methods is thought to be the result of a less uniform strain distribution over the punch profile in the partially annealed material.

The fully recrystallized condition (an average grain diameter of 18 μm) of the tested AlMg6.8 alloy sheet is assessed to have a rather similar or even better formability (at a similar strength level) to the considered 5182-O, 5085-O high-strength formable alloys and the heat treatable 2036-T4 alloy. The partially annealed condition could satisfy moderate forming requirements for the production of very high-strength parts but, generally, the tested alloy can be worked out to different strength-formability relations, between the two tested conditions, meeting attractive application possibilities.

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