An Investigation of Sc Addition on the Extrudability, Recrystallisation Resistance and Mechanical Properties of a 6082- and a 7108-Alloy

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Abstract

In the present work the effect of scandium as a dispersoid former in a 6082 and a 7108 alloy is explored, both by adding Sc alone and in combination with other dispersoid forming elements (Zr, Mn,Cr). In extrusion the breakthrough pressure is in general highest when Sc is added in combination with other dispersoid forming elements. The maximum extrusion speed before tearing decreases somewhat when Sc is added to the 6082 alloys, while taking Mn out of 6082 gives a dramatic decrease in maximum speed due to a change in the melting reaction that takes place. In 7108 with Zr the maximum speed was slightly reduced with increasing Sc content. As for the recrystallisation resistance, it is clear that Sc does not have a strong effect in 6082 under the present conditions. For the 7108 alloy, adding both Sc and Zr lead to a much higher recrystallisation resistance than by adding only Sc or only Zr. Additions of Sc had some influence on the mechanical properties which for the 6082 alloys can be related to differences in the degree of recrystallisation.

1. Introduction

The effects of adding scandium to aluminium have been investigated in great detail both in binary Al-Sc alloys and by Sc-addition to several aluminium alloy types [1]. It seems that the area where Sc has the highest benefit over other alloying elements is as a dispersoid former, and in particular if Sc is added together with zirconium. The effect of Sc or Sc+Zr as a dispersoid former in high strength 7xxx alloys is already reported in several works (*e.g.* [2]), while the lower strength 7xxx alloys does not seem to have been covered. For 6xxx alloys there does not seem to have been much work done, most likely due to the expected formation of an AlSc₂Si₂ phase on the expense of Al₃Sc for alloys with a significant Si content [3]. The aim of this paper is to add to the knowledge base of Sc as a dispersoid former in aluminium alloys by investigating extrusion and extruded profiles of a medium strength 7xxx alloy, 7108, and a 6xxx alloy which is normally intended to be non-recrystallised, 6082, on a semi-industrial scale.

2. Experimental Procedures

A total of nine alloys, five 6082-alloys (A1-A5) and four 7108 alloys (B1-B4), were DC cast as Ø95mm logs. The measured compositions of the alloys are given in Table 1. The 6xxx alloys were homogenised at 575°C/2h and cooled at a rate of 350°C/h, while the 7xxx alloys were homogenised by heating at a rate of 50°C/h to 500°C, kept at that temperature

for 4h and cooled at a rate of 200°C. The homogenised logs were cut to billets of 200mm length, which were extruded at a 800 ton instrumented laboratory press at SINTEF in Trondheim. The billets were preheated in an induction furnace at a rate of approx. 80° C/min. For the 6xxx alloy a plain Ø9mm round bar was extruded, while for the 7xxx alloy a Ø9mm round bar with two 1x1mm² ribs were extruded. Both extrusion tools were equipped with two thermocouples in order to monitor the temperature at the bearing surface. A sketch of the profile geometries is given in Figure 1.

Table 1 Measured chemical compositions of the 6082-alloys (A1-A5) and the 7108-alloys (B1-B4) The measurements were done by an optical emission spectrograph except for the Sc analysis due to the lack of Sc-standards. For Sc inductive coupled plasma spectrography was used.

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_	Alloy	Mg	Si	Zn	Fe	Cu	Mn	Cr	Zr	Sc
	A1	0.67	1.10	-	0.20	0.09	0.48	0.16	-	-
	A2	0.73	1.13	-	0.18	0.08	-	-	0.14	0.03
	A3	0.75	1.16	-	0.18	0.08	-	-	0.14	0.14
	A4	0.74	1.11	-	0.17	0.08	0.45	-	0.13	0.13
	A5	0.74	1.11	-	0.18	0.08	0.53	0.19	0.13	0.13
	B1	1.18	0.07	5.29	0.14	-	-	-	0.15	-
	B2	1.14	0.07	5.17	0.14	-	-	-	0.15	0.04
	B3	1.11	0.04	5.08	0.12	-	-	-	0.13	0.10
	B4	1.17	0.04	5.35	0.12	-	-	-	-	0.10
	 9			Thermoco	Thermocouples		 <u> </u> 9 _			



Figure 1: Sketch of the geometries for the two profiles that were used. The round bar was used for the 6082-alloys (A1-A5) while the round bar with ribs was used for the 7108-alloys (B1-B4).

The objective of the extrusion experiment was twofold. First one wanted to find the maximum extrusion speed before tearing of the profiles, and the associated critical profile exit temperature before tearing. This was done for billet temperatures of 420°C, 460°C and 500°C for the 6082 alloys and for billet temperatures of 460°C, 490°C and 520°C for the 7108 alloys. For each billet temperature several billets were extruded, each billet at a different ram speed, and the extruded profile was inspected for tearing before the next billet was extruded. If a tested ram speed did not give tearing in the profile the ram speed for the next billet was increased somewhat and, vice versa, if the tested ram speed gave tearing in the profile the ram speed for the next billet was decreased somewhat. In this manner the critical ram speed for tearing for a given billet temperature was determined with an accuracy of approx. +/- 2-3% The second objective was to provide material for determination of the recrystallisation resistance of the alloys. For this purpose three billets of each alloy were extruded with a constant billet temperature but at different ram speed as shown in Table 2. Samples from the front, middle and the end of the runout length of these profiles were collected for examination of degree of recrystallisation and testing of mechanical properties. For examination of the microstructure, transverse cross sections of the profiles were ground, polished and anodised and investigated in an optical light microscope under polarised light. The thickness of the recrystallised surface layer was used as a measure of the recrystallisation resistance of the alloys.

Table 2 Billet temperature and ram speed applied for extruding the profiles that were investigated for recrystallisation resistance.

Alloys	Billet Temperature [°C]	R	am Speed [mm/s]	
6082 (A1-A5)	460	2	5	8*
7108 (B1-B4)	490	2	4	6*

* or the highest ram speed that did not give tearing

For evaluation of the mechanical properties, two ageing procedures were used for each of the alloys. Samples of the 6082 alloys (A1-A5) were aged at either 165°C for 24h or at 185°C for 4h without a separate solution heat treatment. The 7108-samples (B1-B4) were either two-step aged at 100° C/5h + 150° C/8h without a separate solution heat treatment, or solution heat treated 480° C/30 min, water quenched and two-step aged at 100° C/5h + 150° C/6h. In the following these four ageing procedures are labelled 165° C/24h, 185° C/4h, Non-SHT and SHT, respectively.

3. Results and Discussion

Figure 2 and Figure 3 show the breakthrough pressure as a function of billet temperature and ram speed v_{ram} for the 6082 alloys (A1-A5) and the 7108 alloys (B1-B4), respectively. The breakthrough pressure of an alloy decreases with increasing billet temperature, but this effect is small within the present temperature intervals. The main part of the observed differences in breakthrough pressure at a constant ram speed is thus due to the alloy composition. Considering Figure 2, it seems that the alloys A2 and A3 have the lowest hot deformation resistance, that the alloys A1 and A4 have a slightly higher hot deformation resistance and that alloy A5 has the highest deformation resistance of these 6082-alloys. The main difference between these alloys is the content of the dispersoid-forming elements Mn, Cr, Zr and Sc. One observes that the increase in Sc content from alloy A2 to alloy A3 has a limited, if any, effect on the breakthrough pressure, and that the effect of the Sc+Zr addition in alloy A4 on the breakthrough pressure is approx. the same as the effect of the Cr addition in alloy A1. Considering Figure 3, a comparison of the alloys B1, B2 and B3 indicates that the hot deformation resistance increases somewhat with increasing Sc content of the alloys. Comparing alloy B4 and B1, it seems that substituting 0.15 wt.% Zr with 0.10 wt.%Sc leads to a slight decrease in the hot deformation resistance.



Figure 2: Breakthrough pressure of the 6082 alloys (A1-A5) as a function of the billet temperature and the ram speed $v_{\text{ram}}.$



Figure 3: Breakthrough pressure of the 7108 alloys (B1-B4) as a function of the billet temperature and the ram speed v_{ram} .

Figure 4 shows the critical ram speed before tearing occurs as a function of billet temperature. For each alloy the variation in critical ram speed with billet temperature can be approximated by a linear relationship within the investigated temperature range. Consider the alloys A1, A2 and A3, where the Mn + Cr addition in alloy A1 has been replaced by Zr + Sc in alloys A2 and A3. Mn and Cr tie up some Si in constituent particles after casting and homogenisation, whereas Zr and Sc do not. This leads to an approx. 0.2 wt.% higher effective Si content in alloys A2 and A3 vs in alloy A1 which in turn lowers the temperature of eutectic local melting of the alloy. Local melting is recognised as a reason for hot tearing of profiles during extrusion of this type of alloys [4]. The measured critical profile exit temperature of 559 – 560°C of alloys A2 and A3 matches the ternary eutectic melting reaction AI + Mg₂Si + Si \rightarrow Liq., while the critical temperature 582°C of alloy A1 lies in the temperature range for the eutectic melting reaction AI + Mg₂Si \rightarrow Liq. [5]. So, in spite of the lower hot deformation resistance of alloys A2 and A3 compared to alloy A1, the critical ram speed drops due to the lower temperature for local melting.



Figure 4: Critical ram speed before tearing as a function of billet temperature for all the alloys. The measured critical exit temperatures are indicated on the lines of critical ram speed.

In the alloys A4 and A5 which also contain Mn and Mn+Cr, respectively, the effective Si content is reduced and the critical temperature for tearing increases accordingly. Alloy A4 has a lower critical speed than alloy A1 due to a lower temperature for binary eutectic melting, while alloy A5 has a lower critical speed than A1 due to a higher amount of dispersoids and thus a higher hot deformation resistance, as found in Figure 2.

Considering the alloys B1, B2 and B3 it seems that the critical speed before tearing is reduced somewhat with increasing Sc content, even though the critical profile exit temperature is measured to be somewhat higher in the Sc-containing alloys. In alloy B4 tearing only occurs at rather high ram speeds while spalling occurs at lower ram speeds. Spalling is a surface defect characterised by local thinning of the extruded section [6]. Thus, for comparison of performance between the alloys the critical speed with respect to spalling, indicated with grey in Figure 4, should be considered as the maximum ram speed for alloy B4.

The measured thicknesses of the recrystallised layer of the as-extruded profiles are shown in Figure 5. For the 6082 alloys it is obvious from the data of alloy A2 and A3 that the additions of Sc and Zr alone do not offer adequate recrystallisation resistance under the present conditions. By comparing Alloy A1, A4 and A5 one finds that Sc and Zr improves the recrystallisation resistance somewhat when added together with Mn or Mn + Cr. However, such an improvement could also probably have been achieved by adding more Mn and Cr to the alloy. For the 7108 alloys one finds that the addition of 0.10 wt.% Sc gives a poorer recrystallisation resistance than the addition of 0.15 wt.% Zr. When Sc is added together with Zr, one finds that 0.04 wt.% Sc is sufficient to give a completely non-recrystallised structure of the profiles, which is in accordance with literature data [7].



Figure 5: Measured recrystallised layer of the Ø9mm profiles as a function of ram speed and alloy type. When the columns reach 4500 μ m this means that the profiles are completely recrystallised.

Figure 6 shows the measured yield strength and ultimate tensile strength of the alloys. For the 6082 alloys, one notices that the alloys A2 and A3 have a considerably lower strength than the other alloys. This is due to the recrystallised structure of these alloys. The strongest alloy is A5, which also is the alloy that has the highest recrystallisation resistance. For the 7108 alloys the alloy B4 that contains only Sc as a dispersoid-forming element has the lowest strength for both of the ageing procedures applied. When solution heat treatment is applied prior to age hardening, alloy B1 without Sc has the highest strength. When ageing is done without a separate solution heat treatment the ultimate tensile strength of the alloys B1, B2 and B3 is fairly equal, whereas the yield strength of alloy B3 is higher than that of alloy B1 and B2. It is not possible to give an explanation of these variations without a further investigation of the materials.



Figure 6: Measured yield strength and ultimate tensile strength of age hardened specimens of the extruded profiles. The labelling of the different age hardening procedures is explained in the experimental procedure.

4. Conclusion

In extrusion the breakthrough pressure is in general highest when Sc is added in combination with other dispersoid forming elements. The maximum extrusion speed before tearing decreases somewhat when Sc is added to the 6082 alloys, while taking Mn out of 6082 gives a dramatic decrease in maximum speed due to a change in the melting reaction that takes place. In 7108 with Zr the maximum speed was slightly reduced with increasing Sc content. As for the recrystallisation resistance, it is clear that Sc does not have a strong effect in 6082 under the present conditions. For the 7108 alloy, adding both Sc and Zr lead to a much higher recrystallisation resistance than by adding only Sc or only Zr. Additions of Sc had some influence on the mechanical properties which for the 6082 alloys can be related to differences in the degree of recrystallisation.

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