Mechanical Properties and Microstructure of Rapidly Solidified Al-Fe Alloys

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Abstract

The mechanical properties of three P/M processed aluminium alloys, Al-Fe-Zr and Al-Fe-Zr-X (X=V, Si), have been investigated. These alloys contain high concentrations of transition elements, which form thermal stable precipitates. Tensile test were performed of temperature up to 400°C. The yield strength of the rapid solidified alloys decreases with increasing temperature. Above 250°C the deformation mechanism change and dispersoid volume fractions and size have not a significant effect on yield strength. Fracture mode exhibits a change about 250°C. The addition of V to the ternary Al-Fe-Zr alloys improvement the thermal stability.

1. Introduction

The rapid solidification technologies and powder metallurgy make possible to produce a new kind of dispersion–strengthened aluminium alloys. These new materials include also the Al – TM alloys with a very high strength at the room temperature and good mechanical properties at temperature up to 300°C [1]. All the binary Al – TM alloys have moderate thermal stability. The development of optimised high – temperature alloys has focused on ternary and quaternary alloys. The most prominent of these are Al – Fe alloys, containing addition of V, Cr, Mo, Zr [2]. The transition metals V, Zr have low diffusion in aluminium. The addition of these elements to the Al- Fe alloys is expected to improve the thermal stability [3]. In this article we present the mechanical properties at ambient and elevated temperatures in Al – Fe alloys containing Zr, V and Si.

2. Experimental

In the addition to the ternary Al-Fe-Zr, Al-Fe-Zr-V, Al-Fe-Zr-Si, alloys were produced by rapid solidification using a melt-spinning technique. During melt spinning liquid alloys are casted upon a fast rotating wheel. Ribbons are approximately 65µm thick and 2-3mm wide. Ribbons length is not limited in the case of a continuous process. The ribbon obtained is cut into flakes and then compacted by cold pressing to a preform. Full consolidation takes by a hot extrusion with a 20:1 ratio at the temperature of 400°C. Prior to consolidation the perform were placed in aluminium cans and vacuum degassed. The ribbons were hot consolidated to full density via the PM processing sequences given in Figure 1. The overage chemical composition and volume fraction of investigated alloys are given in Table 1. The microstructures were investigated by optical microscopy (OM), transmission electron microscopy (TEM) and scanning electron microscopy (SEM).
Table 1: Chemical composition of investigated alloys and volume fraction of dispersoids

<table>
<thead>
<tr>
<th>Alloy/Element</th>
<th>Fe (m.%)</th>
<th>Zr (m.%)</th>
<th>V (m.%)</th>
<th>Si (m.%)</th>
<th>Vol. fraction (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al-Fe-Zr</td>
<td>7,7</td>
<td>1,10</td>
<td>0,0</td>
<td>0,0</td>
<td>32,6</td>
</tr>
<tr>
<td>Al-Fe-Zr-V</td>
<td>6,2</td>
<td>0,95</td>
<td>1,2</td>
<td>0,0</td>
<td>27,4</td>
</tr>
<tr>
<td>Al-Fe-Zr-Si</td>
<td>7,7</td>
<td>0,95</td>
<td>0,0</td>
<td>2,3</td>
<td>34,8</td>
</tr>
</tbody>
</table>

Tensile test specimens were machined from rods in longitudinal directed to the extrusion axis. The cylindrical tensile specimens had cross section of 8 mm and gauge length at 20 mm. The tensile tests were performed on the GLEEBLE 1500 testing machine. The conduction heating of specimen was computer controlled in the gauge section. The tensile test were performed at temperatures between the RT and 350°C with an initial strain rate \( \varepsilon = 7 \cdot 10^{-4} \text{ s}^{-1} \).

3. Results and discussion

Microstructure

Representative SEM/TEM micrographs of the ribbons cross section are shown in Figure 2. Two zone of microstructure were observed in each of these alloys. The first zone has been designated with A and the second zone with B [4]. Typical observed microstructure in all three alloys, was a cellular matrix structure.
The investigated dispersion–strengthened aluminium alloys exhibited very small grain sizes compared to the conventional Al alloys product by casting. The microstructure of these materials depended on chemical composition and cooling rate and was influenced by the dispersoids type and the dispersoid volume fraction. The grain and the particle size influenced the deformation behaviour in tension test. Transmission and scanning electron microscope micrographs of the extruded alloys are shown in Figure 3. The microstructures are rather heterogeneous and strongly directional. As shown in Figure 3a the duplex microstructure (zone A and B) was still evident.

The dispersoids were essentially equiaxed and there was no evidence of the original cellular region, which was presented in the ribbons. The majority of the dispersoid particles was located at grain boundaries and there was a fairly homogeneous distribution (in each zone). Al-Fe-Zr, Al-Fe-Zr-Si and Al-Fe-Zr-V alloys had approximately similar microstructure with the dispersoid size 50-300nm and grain size 0.5-3µm as whiles coarsen precipitation and grain was expected. The main difference was a smaller overage size of particles and grains in the Al-Fe-Zr-Si and Al-Fe-Zr alloys.

Mechanical properties

Tensile test were carried out in the temperature range from the RT to 450°C. The tensile curves always show a very low strain hardening (Figure 4). Figure 5 shows yield strength as function at the test temperature of the investigated alloys. In the Al-Fe-Zr-V alloy, the yield strength (YS) is reached at a lower strain (318MPa, 0.2 pct) than in the Al-Fe-Zr alloy (370MPa, 0.2 pct) and in the Al-Fe-Zr-Si alloy (394MPa, 0.2 pct). In all these alloys the YS decreased with the increasing temperature. At the high temperature, the difference between the alloys reduced and the overage strength is higher than the strength of conventional
aluminium alloys. Above about 250°C the alloys exhibited a similar strength. The addition of V improved the elevated-temperature strength of the Al-Fe-Zr alloy. Figure 6 shows the changes in the Vickers hardness (at ambient temperature) as a function of isothermal annealing for all investigated alloys. The hardness decreased gradually at the first stage and then decreased continuously by annealing at 400°C, except of the Al-Fe-Zr-V alloy, where it increased at the first stage and than decreased. The addition of V improved the thermal stability of the Al-Fe-Zr alloy.

The ductility measured by the fracture area reduction, as a function of temperature is shown in Figure 7. At the room (ambient) temperature the Al-Fe-Zr-V alloy showed the highest ductility. The lowest ductility was found with the Al-Fe-Zr-Si alloy. The temperature dependence of the ductility showed pronounced differences among the three alloys. The Al-Fe-Zr-Si alloy exhibited slightly increased ductility between the room temperature and 300°C. The ductility of the Al-Fe-Zr-V alloy remained nearly constant from the room temperature up to 200°C and then decreased quite sharply and exhibited ductility dip at above 300°C. In contrast, the ductility of the Al-Fe-Zr alloys decreased more gradually over the whole temperature range.
The fracture surfaces of tensile testing specimen at room and elevated temperatures of the fractured testing alloys were similar. The representative micro-fractographs are shown in Figure 8.

Fig. 8: Micrographs of the hot extruded Al-Fe-Zr alloy after tensile test; a) at room temperature and b) at 350°C

At the room temperature the fracture surfaces was ductile with dimples. Voids nucleated at the dispersoids and subsequently, crack propagation occurring by coalescence of these voids. At 350°C the fracture surfaces exhibited new features and over the whole fracture surface there was clearly visible lack of dimples and dispersoids. The voids are initiated at the dispersoids along the grain boundaries and the cracks were propagated along the grain boundaries.

4. Conclusion

1. Rapid solidification resulted in highly supersaturated solid solution with corresponding strengthening and also produced dispersoids that induced Orowan strengthening. But the majority of dispersoid particles were present at grain boundaries and it was difficult to apply the Orowan model to the particles on the boundaries. In all of the examined alloys, the strengthening mechanism was mainly attributed to the very fine grains.

2. The difference in strength at the room temperature for Al-Fe-X alloys was due to dispersoids, which affected the grain size. Coarsening of dispersoids, which pinned on grain boundaries allowed grain growth. Still small change of grain size influenced on the strength.

3. After hot extrusion, the three alloys were characterized by a microstructure consisting of fine dispersoids with a size between 50 and 300nm in Al matrix. Small-scale inhomogeneties were found in all the three alloys in forms of area with coarse dispersoids and large grains.

4. The yield strength of the alloy decreased with the increasing test temperature and above 250°C different addition elements (Zr, V, Si) and dispersoid volume fraction have not significant effect on the strength due to a change of deformation mechanism.

5. The ductility dip of these alloys is restricted to strain rates and the finest microstructure.

6. The fracture mode was changing from ductile dimpled fracture at low temperature to intergranular fracture at high temperature.

7. The addition of Zr to the binary Al-Fe alloys refined the matrix grain size.
8. The addition of V to Al-Fe-Zr alloys improved the thermal stability and strength at elevated temperatures.

**Literature**