Effect of Copper Modification on Impact Strength of Zinc-Aluminum Alloys

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The effects of the microstructure and copper content on the notched impact strength of the eutectoid Zn-Al alloy modified with 0.58 mass% Cu and 4.95 mass% Cu were studied as a function of temperature from room temperature to 350°C. The fracture mode changes from transgranular to completely intergranular above 350°C. It is suggested that the change in fracture behavior is associated with the formation at the grain boundaries, of a low melting point compound with the eutectic composition.

(Received July 19, 1993)

Keywords: zinc-aluminum alloys, impact properties, brittle-ductile transition

I. Introduction

Zinc-based alloys with 20 - 30 mass% Al and more than 1.5 mass% Cu are potentially very attractive because of their low forming and casting temperatures, (under certain conditions the material behaves as superplastic material) and high strength at room temperature. The Zn-Al-Cu alloys have shown to be competitive and alternative material to other well-developed alloys due to their applications in the field of extrusion(1) and foundry(2). Their use is expected to grow in functional applications involving high stresses (~400 MPa) and good atmospheric corrosion resistance.

The present study was initiated in order to investigate in more detail on the fracture behavior of the eutectoid Zn-Al alloy, with fine grained (superplastic) and pearlitic structure with up to 4.95 mass% Cu. It is important to realize that the properties of these alloys are similar to those of the steel and vary with heat treatment. Alloys with up to 2 mass% of copper are put in the superplastic condition by heating above the eutectoid temperature (276°C), then water quenched. The microstructure of the superplastic material consists of extremely fine grains, which are almost unobservable except by electron microscopy. The pearlite structure appears after slow cooling from the temperature above the eutectoid reaction to room temperature.

Above 276°C and less than 1 mass% Cu, the alloy shows the f.c.c structure of Al (β phase). Upon cooling, the β phase decomposes into two-phases designated α + η, where α is the Al-rich phase and η is the hexagonal Zn-rich phase.

Alloys with copper content between 2 and 5 mass%, shows the presence of the intermetallic compound CuZn4 (ε), which is stable above 268°C, and decomposes(3) during prolonged aging (~9 h at 170°C) in the ternary ordered, Al-rich T' phase. Below 350°C, there are three phase transformations(4), associated with this ε phase as follows:

\[ \beta + T' \leftrightarrow \alpha + \varepsilon \text{ at } 285°C \]
\[ \beta + \varepsilon \leftrightarrow \alpha + \eta \text{ at } 276°C \]
\[ \alpha + \varepsilon \leftrightarrow T' + \eta \text{ at } 268°C. \]

After these decompositions, the stable phases of the alloy at room temperature are composed of an Al-rich solid solution α, a Zn-rich solid solution η, and the T' phase.

The impact strength of a material gives a measure of ability of the material to withstand impact loading and is important since many commercial articles are frequently subjected to such loading conditions during service. The Zn-Al alloys normally break in a brittle manner at room temperature with very small amounts of absorbing energy. Improvements in the notched impact strength (NIS) of these alloys are highly desirable.

II. Experimental, Specimens and Procedure

The ingots were melted in an electric furnace under atmospheric conditions, using Zn (99.5 mass%), Al (99.5 mass%) and Cu (99.99 mass%). The melting was carried out in graphite crucible and the molten metal was poured into a copper mold. The resultant compositions are shown in Table 1.

After homogenization treatment of the ingots for 48 h at 370°C, the samples for Charpy notched impact testing were cut in sizes of 10 × 10 × 55 mm and notched with a cutter of tip radius 0.25 mm. In order to get the fine grained structure, a number of samples were heated to 350°C for 30 min and quenched into water at room temperature. No quench cracks were observed in the specimens. Another set of samples was furnace cooled
Table 1  Chemical compositions of the specimen*.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Zn</th>
<th>Al</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>77.68</td>
<td>21.74</td>
<td>0.58</td>
</tr>
<tr>
<td>B</td>
<td>77.26</td>
<td>20.4</td>
<td>2.34</td>
</tr>
<tr>
<td>C</td>
<td>74</td>
<td>20.95</td>
<td>4.95</td>
</tr>
</tbody>
</table>

*Composition mass% Si and Fe are below 0.07 mass%.

from 350°C to room temperature in order to obtain the pearlite structure.

Microprobe and X-ray analyses were carried out on polished samples after thermal treatments. Scanning electron micrographs were taken for showing the resultant microstructures. The samples were fractured by impact at several temperatures, 24 h after the thermal treatment, from room temperature to 350°C, using 50°C steps heating. At least three samples were tested from each tempering treatment at each temperature. Before testing, each sample was housed in a small electric furnace previously stabilized at the test temperature. When an homogeneous temperature had been reached in the sample (after about 15 min), it was removed from the furnace and placed between the test piece supporting holders of a Shimadzu charpy impact testing apparatus and tested.

Testing was carried out within 5 s of removal from the furnace. According to a preliminary testing, errors in temperature due to change after removal from the furnace were small and do not alter significantly the conclusion drawn from these results.

III. Results and Discussion

1. Microstructure and phase stability

The microstructure and phase transformations of the eutectoid Zn–Al alloys with less than 1 mass% Cu are well characterized and discussed in detail elsewhere. The metallurgical behavior of the representative alloys “B” and “C” was investigated in the present work.

(1) Alloy B

In addition to the Al-rich phase, α, and the Zn-rich η, which appears in alloy A with 2.34 mass% Cu, induces the appearance of a new Zn-rich ε intermetallic phase, with hexagonal structure and composition CuZn6. Figure 1(a) shows the resultant microstructure of the alloy “B” after furnace cooling. The intermetallic phase ε (CuZn6) identified by micro-analysis appears as small (≈ 2 μm) rounded grains distributed along the lamellar structure. The cooling rate used in the present experiment was not enough to transform completely the ε phase into the T′ phase and four phases α, ε, η, and small amounts of the T′ phase were observed by X-rays analysis of the alloy at room temperature. In contrast, the resulting structure after water cooling of the alloy from 350°C, was consisted of fine grains (≈ 0.5 μm) of the α and η phases with needles of ε phase, whose composition was around Cu14Zn44 closer to CuZn5 than to CuZn4 shown in Fig. 1(b). These needles formed with the Zn-rich ε phase, convert to round grains with composition CuZn5 after furnace cooling. The microstructure of the low Cu alloy A, furnace cooled, is pearlite without the presence of the ε phase. After water quench, the alloy A shows the typical fine grained microstructure of the superplastic metals.

(2) Alloy C

Higher amounts of Cu (4.95 mass%) favors the increment of the ε phase. In the furnace-cooled alloy “C”, the brittle T′ phase was clearly observed around the grains of the ε phase as shown in Fig. 2(a) (black rim around the grains). This results from the decomposition of the ε phase in the T′ phase according to the four phase transformation:

$$\alpha + \varepsilon \leftrightarrow T' + \eta$$

at 268°C.

The resultant microstructure of the alloy “C” after quenching is shown in Fig. 2(b). It is composed of fine grains of α and η, lamellar structure regions plus a coarse needle shape Zn-rich ε phase, with composition around CuZn5.
2. Effect of temperature, microstructure and copper content on (NIS)

The (NIS) of furnace cooled alloys A and B increased steadily with temperature up to 250°C as shown in Figs. 3(a) and (b). Below 200°C in alloys A and B, the NIS values of the water quenched (WQ) alloy are lower than the specimens of the furnace cooled (FC) alloy. After 200°C the NIS values in the alloy “A”, converge and are almost the same for the specimens with pearlitic (FC) and fine grained structures (WQ) (see Fig. 3(a)). When the copper content is increased to 2.34 mass%, (alloy B, Fig. 3(b)) the maximum for the fine grained structure (WQ) is lower than the maximum for the pearlite structure (FC). It is possible that the incoherent needle shaped ε phase (CuZn3), formed in the water-cooled alloy will produce stress concentrations at the tip of the needle, facilitating the propagation of the fracture and making the water-quenched alloy more brittle than the furnace cooled one.

There is a sharp ductile to brittle transition between 250 and 300°C for alloys A and B. This transition was not observed, in the alloy C as shown in Fig. 3(c) with 4.95 mass% Cu, where NIS values increased continuously up to 350°C. The NIS values obtained by alloy “C”, are below the values obtained by alloys A and B, because of the high content of the brittle phases ε and T’.

In general the fracture surfaces of the samples tested at
room temperature showed facets of “intergranular” rupture and scattered areas of dimples, as shown in Fig. 4(a). When increasing the testing temperature up to 200°C, the fracture surfaces exhibited features of higher ductility as shown in Fig. 4(b).

After reaching a maximum in the NIS values around 250°C, alloys A and B show a full intergranular fracture at temperatures higher than 250°C as shown in Fig. 4(c). This change from transgranular to intergranular fracture occurs between 250 and 300°C. In this temperature range the grain boundary material is weaker than the phase exist within the grain. The brittle behavior observed above the eutectoid temperature could be originated by preferential softening or melting of a thin region with a eutectic composition formed at or near the grain boundaries when the temperature is increased. This ternary or binary eutectic phase has a melting point close to 377°C for the ternary and 382°C for the binary, so it is possible that the softening of this phase after 300°C, be enough to produce the observed intergranular fracture. In alloy C this transition from transgranular to intergranular fracture is not observed in the temperature range studied in this work. This is because at this alloy composition, according to the ternary phase diagram, there are not formation of any phases of low melting temperature.

IV. Conclusion

The effect of the copper content on the Notched Impact Strength of the eutectoid Zn–Al alloy, was studied as a function of temperature.

Increasing the amount of Cu from 0.58 to 2.34 mass%, increases the Notched Impact Strength of the material below the eutectoid temperature. After the eutectoid temperature there is a sharp reduction of the impact resistance, possibly due to the presence of a thin liquid intergranular film with eutectic composition. Increasing the Cu content (= 5%) in the alloy produces a continuous increment of the NIS values with temperature.

Acknowledgment

The authors would like to greatly thank Messrs. L. Baños, J. Guzman and E. Caballero for their cooperation in the experimental work. J. Negrete is enrolled in the Ph.D. program of CICESE.

REFERENCES