The Influence of Superplasticity on the Fracture Behaviour of Zinc Based Alloys

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INTRODUCTION:- Superplasticity is the property of a metal that allows it to undergo very large plastic deformation under low applied stress. This normally takes place at a low strain rate in a particular temperature range and involves very little change in the microstructure of the metal. Many stable two-phase alloys can be rendered superplastic by inducing a fine-grain eqiaxed microstructure The alloy under study is based on the zinc aluminium eutectoid with small additions of copper and magnesium to improve the strength, creep resistance and corrosion resistance. This alloy was produced in a number of microstructural forms and the present work seeks to relate the fatigue crack growth and fracture characteristics with the various microstructures and mechanical properties.

MATERIALS:- The alloy of nominal composition 77% Zn 21.9% Al 1% Cu 0.1% Mg was produced in the following four conditions

SP1 and SP2 Superplastic sheet with flow stress of 14MN/m^2 and 33 MN/m^2 respectively at strain rate of 2.4/min and temperature 260°C.

HT1 and HT2 Rendered single phase and then isothermally transformed just below and well below the eutectoid temperature respectively.

FATIGUE AND FRACTURE TESTING:- The specimens used for these tests were centre-cracked sheets, the nominal dimensions of which were: length 400mm, width 100mm and thickness 1.5 mm. The fatigue tests were carried out under constant amplitude loading at a gross stress of 45 MN/m^2 and a minimum/maximum stress ratio of 0.1, at a frequency of 16 Hz. Measurements of crack length and number of load cycles was recorded using foil gauges and an automatic recording technique. All fatigue tests were terminated at $2a/W \simeq 0.5$. A static fracture test was then conducted on each specimen. The results for the fatigue tests for the four material conditions are given in Figure (1), where crack growth rates (mm/cycle) are plotted against stress intensity range ($MN.m^{-3/2}$). The fracture toughnesses were calculated using maximum load at failure and final fatigue crack length and are presented in Table (1).

Material	U.T.S σ_u MN/m^2	.2% P.S σ_y MN/m^2	% Elong	K_c $MN.m^{-3/2}$	$\left(\frac{K_c}{\sigma_y}\right)^2$ M
	Strain rate 0.2/min				
SP1	383	342	10	60.0	0.030
SP2	357	325	9	50.0	0.023
HT1	425	310	12	66.0	0.044
HT2	520	460	15	60.0	0.016

Table 1 Mechanical Properties and Fracture Toughness

<u>FRACTURE SURFACES</u>:- A detailed examination was made in a scanning electron microscope of the surface of the fatigue crack and the static fracture. The surface of the fatigue cracks in SP1 and SP2 were similar. Striation markings parallel with the crack front are not evident, but surface contours, possibly evident of the structure of the rolled sheet are clearly defined normal to the crack front, Figures (2) and (3). The fracture surfaces were both different from that of the fatigue crack, and were ductile in apparance, the dimple size corresponding closely with the grain size of the alloys,

Figure (4). The microstructure of SP1 is very fine grained equiaxed two phase typical of that of a superplastic alloy. SP2 had large particles of one of the phases or of coarse lamellae set within a fine grained equiaxed matrix. The fracture surface showed evidence of the large particles.

HT1 and HT2 in contrast to SP1 and SP2 had heavily structured fatigue crack surfaces again with no evidence of striations. The surface appears to follow the grain boundaries of the high temperature α' phase which existed prior to isothermal transformation, Figure (5). The fracture surface of HT1 appeared ductile with fine dimples super-imposed on a coarser structure, Figure (6). The coarse structure may again be a remnant of the α' grain structure. The microstructure of HT1 was fairly fine two-phase, the transformation having been initia-ted at the α' grain boundaries. Some very fine lamellae were present. In HT2 the microstructure was very find indeed and the fracture surface did not have the ductile appearance of any of the other fractures but shows some cleavage surfaces, Figure (7). These two factors account for the higher strength and reduction in $\left(\frac{K_c}{\sigma_y}\right)^2$ value, Table 1.

As shown in Figure (1), there is a marked difference in crack growth rate between the SP and HT material. At this stage no positive conclusion can be drawn, however, the difference may be due to the residual effect of the α' boundaries which existed in the HT material, but which have been broken down by subsequent hot working during production of the SP material.

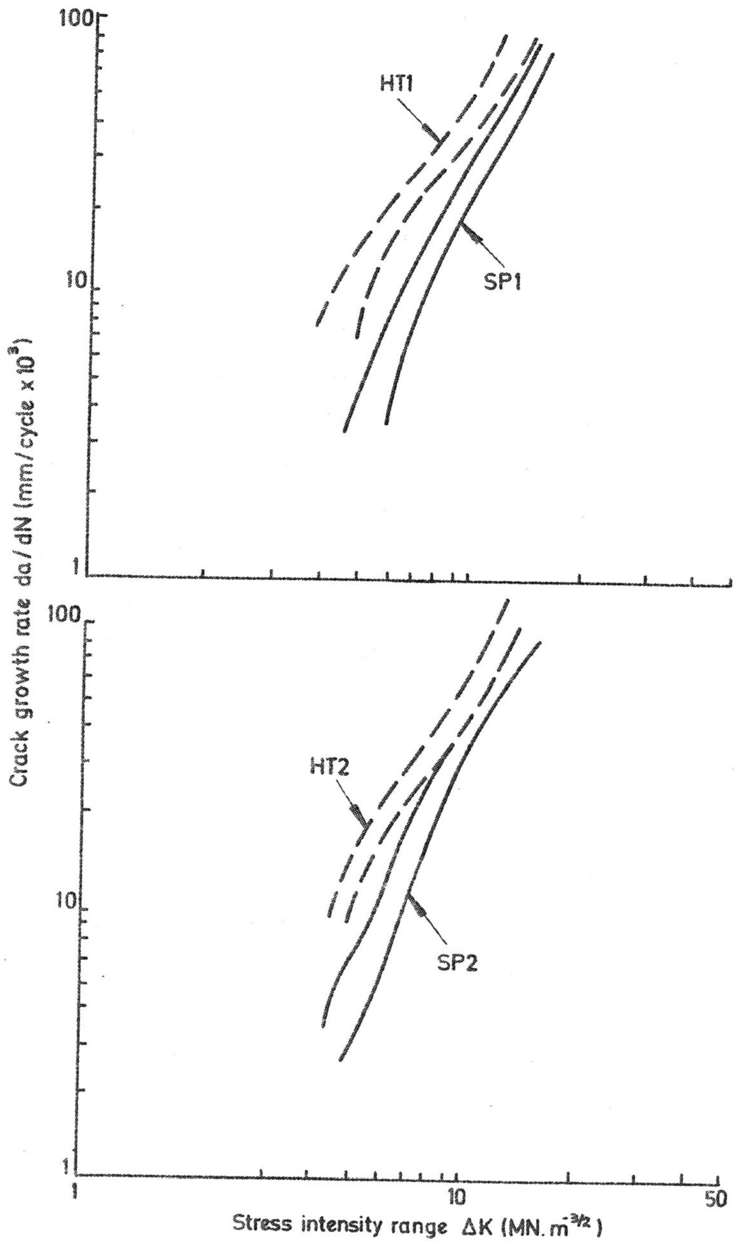

Figure 1 Scatter bands for fatigue crack growth rates of four different microstructures tested.

Figure 2. SP1 Fatigue/ fracture surface. x650

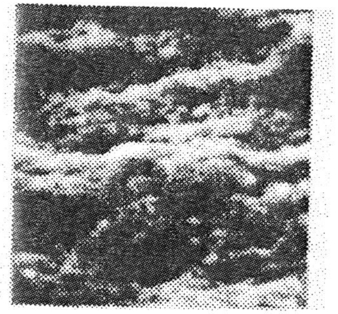

Figure 3. SP1 Fatigue surface. x1900

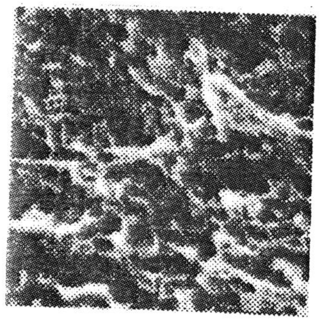

Figure 4. SP2 Static fracture surface. x1260

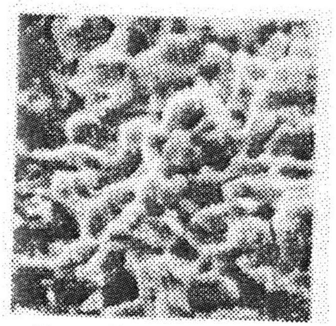

Figure 5. HT2 Fatigue surface. x800

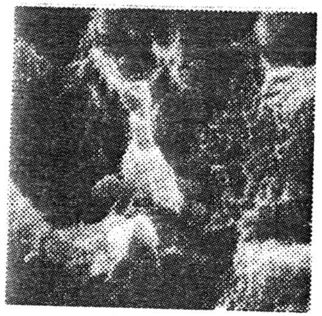

Figure 6. HT1 Static fracture surface. x3250

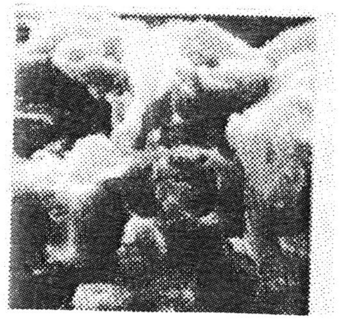

Figure 7. HT2 Static fracture surface. x3600