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Letters

Precipitation Hardening in Magnesium-Tin Alloys

In aged magnesium-tin alloys the Mg_2Sn precipitates lie mainly parallel to the (0001) plane of the matrix [1, 2] and the precipitation is preceded by formation of short-range order [2]. In the first part of this work the influence of tin on the critical resolved shear stress (crss) of magnesium was determined [3]. In the second part, undertaken in order to investigate the behaviour of precipitation-hardened magnesium-tin alloys, microhardness measurements were made on magnesium-1.65% tin (7.75 wt %) alloys and the precipitation was checked in the electron microscope on carbon replicas. This work is described here.

Small pieces of the alloys, cut from the same ingot, were homogenised, quenched and aged at 200, 250 and 300° C. The same procedure for homogenisation as employed by Hardie and Parkins [4], has been used here. For the ageing treatment the samples were enclosed in evacuated pyrex tubes. From hardness measurements on pure magnesium and on the alloy, both heat-treated in the same way, it was deduced that the polishing method did not alter the hardness appreciably. After three hours homogenisation at 550° C, the hardness did not change any more, whereas the crss remained constant after 6 h. In both cases the measurements were made on samples in the as-quenched condition. A homogenisation time of 5 h was applied to all samples, though there was still a proportion of relatively big Mg_2Sn particles.

The microhardness of magnesium-tin alloys increases with increasing tin content (fig. 1). The agreement with other published results [4] is perfect if the difference in method is taken into account. This increase in hardness is linear, but the straight line does not pass through the result for pure magnesium. The initial steep increase up to a "transitional" concentration, followed by a line of lower slope has been reported recently in other magnesium alloys tested in tension [5]. After the homogenisation, the samples were aged at the three mentioned temperatures. The precipitation was checked in the electron microscope, and the microhardness measured with a Vickers pyramid. The hardness of this alloy,

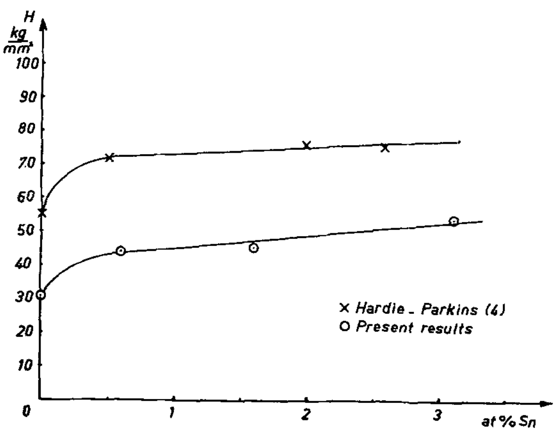

Figure 1 Microhardness of homogenised magnesium-tin alloys plotted as a function of tin content.

plotted against the log of ageing time, passes through a maximum (fig. 2). Owing to uncertainties inherent in microhardness measurements [6], especially for cph structures [7], the Rockwell-B hardness was determined for one ageing temperature. The maximum occurred after the same ageing time, but the hardness itself was lower. The latter fact was probably due to the higher load used, since an increase in load for the microhardness determinations similarly produced a decrease of hardness. The microhardness of the compound Mg_2Sn was found to be as high as

147 kg/mm². The first increase in hardness coincides with the first appearance of precipitates and the maximum is reached at complete precipitation. It is at present not possible to

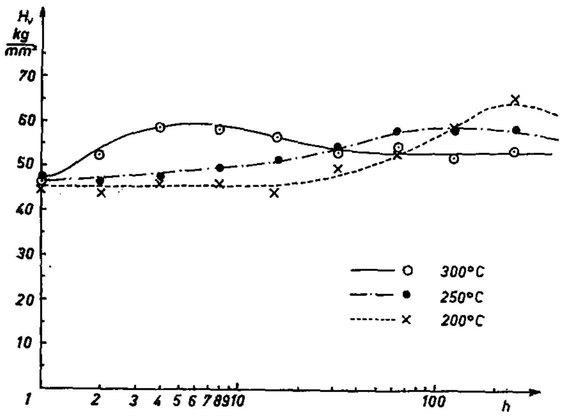

Figure 2 Microhardness of magnesium-1.65 at.% tin alloys, homogenised, quenched and aged at 200, 250 and 300° C.

explain exactly why the hardness decreases after the precipitation reaction is finished. Fig. 3 shows an example of the Mg_2Sn precipitate after completion of the reaction. The lamellae are all parallel and occur only in one direction. However, sometimes the precipitates were found to lie in different directions (fig. 4).

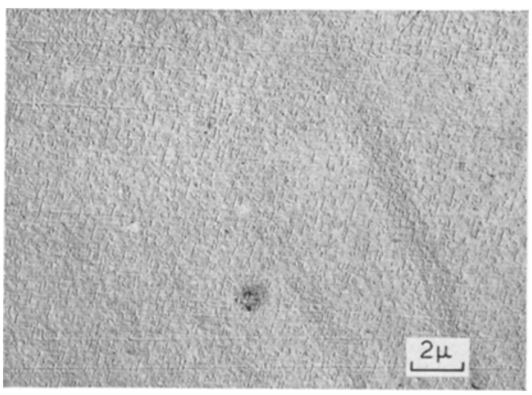

Figure 3 Mg_2Sn precipitates along (0001) plane of (Mg, Sn) matrix after an ageing period of 500 h at 250° C; carbon replica.

Gann [8] found an increase in hardness of aged magnesium-tin alloys, relative to the homogenised alloys. However, this hardening was much lower than in magnesium-zinc and magnesium-aluminium alloys. In a study by

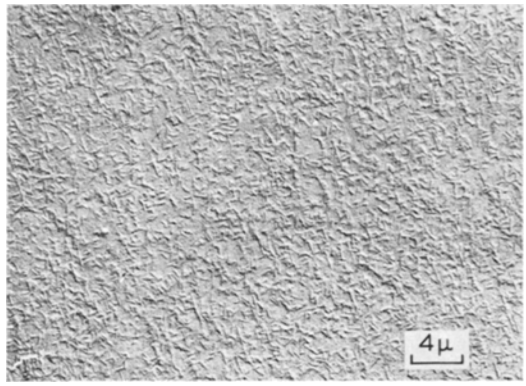

Figure 4 Mg_2Sn precipitates in several directions after an ageing period of 125 h at 200° C; carbon replica.

Clark [9] the maximum hardness of magnesium-zinc alloys containing 5 wt % zinc (1.9 at. %) in the aged condition was found to be 75 kg/mm². For the present results we obtain a maximum hardness of 65 kg/mm² for a magnesium-1.65 at. % tin alloy, which is appreciable and in contradiction with Gann's conclusion.

The influence of the precipitate on the mechanical properties of single crystals was

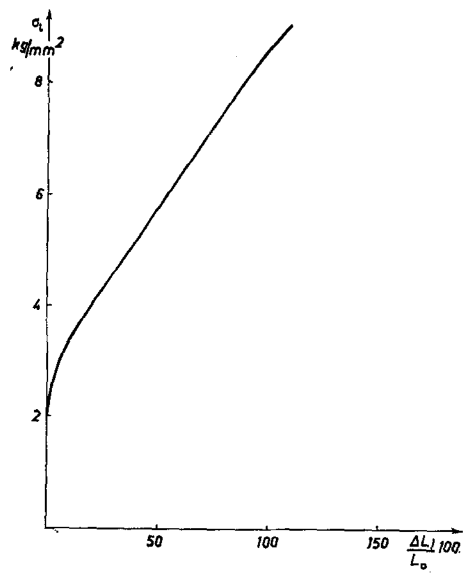

Figure 5 Stress-strain curve of magnesium-tin single crystal containing Mg_2Sn precipitates and tested at room temperature.

studied on one sample. A single crystal oriented for basal slip ($\chi_0 = \lambda_0 = 49° 30'$) and treated in the same way as the samples for hardness measurements (i.e. aged for 15 h at 300° C after being homogenised) was tested in tension at room temperature at an initial elongation rate of 2×10^{-3}/sec. The crss amounted to 1310 g/mm². In the Orowan model [10] it is assumed that dislocations bow out between particles and the crss is given to a first approximation by $\tau_0 = \tau_{om} + 2G\mathbf{b}/l$, where τ_{om} is the crss of the matrix [3], G is taken as the shear modulus of pure magnesium, **b** the Burgers' vector and l the mean precipitate spacing in the basal plane (estimated at 2.5×10^{-3} mm).

Though the (τ, γ) curve (fig. 5) conforms to alloys containing non-deforming particles for which equation 1 is valid, the above formula predicts a crss of 519 g/mm², which is much too low when compared with the experimental 1310 g/mm². At the moment it is not possible to decide whether this difference is caused by an increase in the shear modulus due to the presence of the precipitate or that the by-passing process [11], following the bowing of dislocations, is extremely difficult as can be expected for cph alloys. The rate of strain-hardening in the linear part of the (τ, γ) curve is of the same order of magnitude as with pure magnesium i.e. 1100 g/mm².

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Interlamellar Slip in Polyethylene

The purpose of this note is to comment upon the interpretation of the small-angle X-ray patterns in parts 2 and 3 of a study of orientation effects in polyethylene [1, 2]. In these papers, the authors propose that the crystalline regions of an oriented, branched polyethylene consist of stacks of lamellae. Changes in the wide-angle and small-angle X-ray patterns produced by annealing, compressing or extending an oriented specimen within certain limits are interpreted in terms of interlamellar slip. We shall define a precise model for a stack of lamellae and consider the effects of the rotation of the stack and of inter-lamellar slip within it on its diffraction pattern.

Fig. 1a shows the model we adopt. Each lamella is a single crystal and is surrounded by a layer of amorphous material of different electron density. Fig. 1b shows the Fourier transforms of the linear lattice formed by taking an identical point within each lamella, and of the lamella itself. The lattice transforms into a set of planes, intersecting the plane of the diffraction pattern in layer lines. These are spread out to a width of L^{-1} because of the finite length of the lattice, L. The lamella transforms into a spike normal to the plane of the lamella. The diffracted amplitude is given by the product of these transforms and therefore the diffraction pattern has the appearance shown by the shaded regions. The addition of the diffraction from the twin of the stack of lamellae shown in fig. 1a produces the four-point pattern shown in fig. 2a. This corresponds to the pattern obtained by annealing at 70° C after drawing and rolling in the 0y direction [1]. The material oriented in this way is therefore conceived as consisting of stacks of lamellae of the type shown in fig. 1a, and its twin as distributed in a matrix of amorphous material. The nature and extent of the amorphous material is not specified, but it must bound the stack by material of different electron density from that of a lamella.

This model is similar to the model adopted in [1] and [2], except that in [2] it is assumed that the lateral extent of the lamella (width W in fig. 1a) is large enough not to cause a spread in the diffraction pattern. The spread of the small-angle spots is, on this view, the result of variations in the orientations of the lamellae. To accord with