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Report:

Large plastic deformations usually result in very heterogeneous dislocation microstructures, where the heterogeneity occurs at different scales. Within single grains of a polycrystalline sample a cellular dislocation structure develops with dislocation walls which encircle regions with low dislocation density. At the grain level one can observe that different grains have different dislocation densities, as a consequence of their different amount of plastic slip experienced during the deformation. In the present experiment the influence of grain orientation on the stored energy during hot deformation of a commercial Al-Mg alloy has been studied. The stored energy was estimated from the dislocation density, which was evaluated from Laue peaks measured with high resolution. The investigated texture components were Brass, Cube, Goss, Copper and S, which represent of about 65% - 70% of the total grain orientations present in a hot rolled polycrystal. For the evaluation of the dislocation density only those diffraction vectors were selected which do not conflict with reflections of the same Miller indices but belong to other texture components present in the sample. This was achieved by rotating the sample around the beam and the diffractometer axis. A beam cross section of 2x0.5 mm^2 was used, which permitted the investigation of a large sample volume, so the obtained information can be considered a good average.

Figure 1 shows 311 peaks corresponding to the above texture components of the Al-Mg alloy, hot deformed to a true strain of 1.4. There are only minimal differences between the 5 peaks, although one can see that in the whole $\Delta(2\theta)$ range the narrowest peak corresponds to the Brass component (in black) and the widest to the S component (in blue). The dislocation density of these two texture components is in good accordance to this qualitative remark and with annealing experiments done on this deformed state, which show that the S texture component recrystallizes first, i.e. it has the highest amount of stored energy [1,2]. Actually the widths of the peaks are proportional to the square root of the formal value of the dislocation density given by: $\rho^* = \frac{\pi}{2} g^2 b^2 C\rho$, where g and b are the magnitudes of the diffraction and Burgers vectors, respectively. ρ is the true value of the dislocation density and C is the average contrast factor of dislocations and depends on the relative orientation of the Burgers and line vector of dislocations with respect to the diffraction vector [3]. This geometrical dependence, related to the anisotropic strain field of the dislocations, is however the reason for the well-known strain anisotropy effect. In our case the strain anisotropy depends mainly on the Burgers vector population of the different slip systems. If this population is known (for example, in the case of a

powder sample due to averaging over all directions a uniform population can be considered) then the average contrast factor C, for a given diffraction vector can be numerically calculated. The knowledge of the average contrast factor becomes important in the case of the application of the *modified* Williamson-Hall (*m*WH) and *modified* Warren-Averbach methods [4]. Recently we have shown [5] that by applying the restricted random dislocation distribution model of Wilkens [3], the peak widths corresponding to different intensity values follow a straight line in the *m*WH plot. This result is of major importance, because the linear behaviour of the peak widths can be an indication of the correct Burgers vector populations. The knowledge of these is very important for the theories of plastic deformation. During our experiment we have measured 5 different reflections for each texture component, which permitted us to check the linear behaviour of the peak widths in the *m*WH plot. The widths at half-maximum and at 10% relative intensity are shown in Fig. 2 for the S and Brass texture components. In this case the Burgers vector populations were estimated form crystal plasticity simulations. The expected linear behaviour is obtained with 5% relative error, which is a very good result and represents an internal check of the evaluations. Similar linear behaviour of the peak widths was obtained for the Goss component and deviations form the line were obtained for the Copper and particularly the Cube component. The latter is an indication that the present crystal plasticity approximation did not adequately describe the Burgers vector population existing in this texture component. The numerical simulations of the *m*WH plot together with the data measured at ESRF are published in [5].
We mention here that by measuring at least as many Bragg peaks as the number of slip systems, based on the above equation, where the principle of superposition applies for the calculation of the average contrast factor, the population of Burgers vectors can be obtained experimentally from the X-ray measurements. We intend to do such a study in the near future on ID 31.

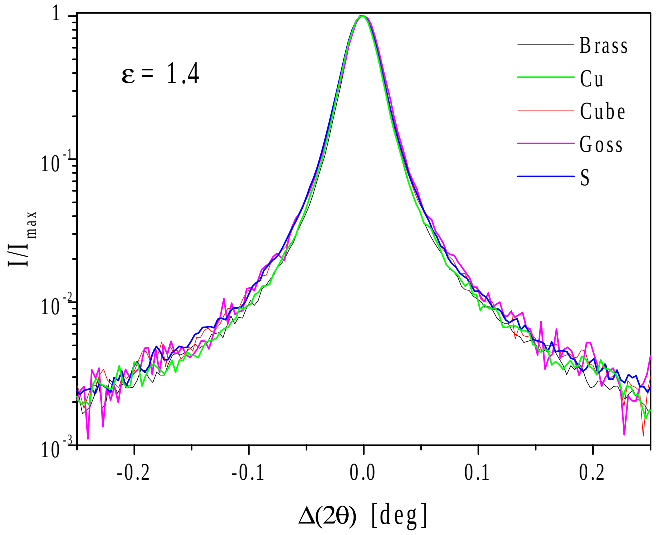

Fig. 1. 311 peaks measured on the different texture components of a hot deformed Al-

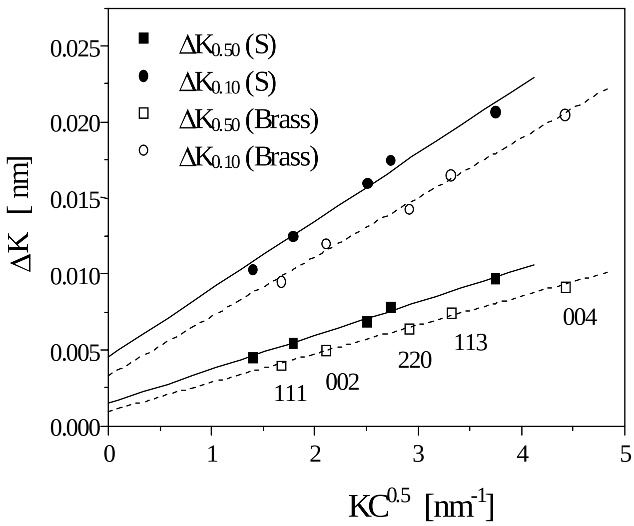

Fig. 2. *Modified* Williamson-Hall plot of the peak widths at half-maximum and 10% relative intensity, for the case of Brass and S texture components.

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