DISTORTION OF THE MATRIX CAUSED BY COHERENT PRECIPITATES IN THE Al-Ni SYSTEM

D.Kunstelj and K.Kranjc
Institute of Physics of the University, Zagreb

An Al-3.6 at.%Ni alloy was quenched from the melt in order to obtain supersaturated solid solution of Ni in Al. The development of clusters of Ni atoms during annealing and their influence on microhardness were examined. It was found[1] that in the second stage of annealing, coherent disc-like precipitates of the metastable η- phase[2] develop. In this stage the microhardness of the alloy is about 280 kp/mm^2. The plate-like precipitates were found to lie in the $\{100\}$ planes of the matrix[1], producing a characteristic "coffee-bean" contrast in the electron micrographs which is caused by the distortion of the matrix planes. From the diffraction patterns it was established that $(111)_\eta \ // \ (100)_{Al}$ and that e^c, the "constrained strain"[3] is about 0.07. The transformation strain ("stress-free strain"[3]) $e^t = 0.074$ was calculated from $d^{\eta}_{222} = 2.175$ Å and $d^{Al}_{200} = 2.025$ Å.

The matrix method[4,5] in the two-beam approximation of the dynamical theory of electron diffraction[4] was used in order to match the theory to the results of the experiments. This method is now widely applied when obtaining "computed micrographs" of various lattice defects[6], e.g. dislocations and stacking faults. In computation of the contrast at the precipitates we applied the linear elasticity theory which was developed for inclusions of arbitrary shape[3] and has already been applied to the spherical particles which strain the host matrix[7]. We rearranged and adapted the computer programm worked out by Thölén[8], who applied the matrix method to obtain simulated micrographs of various configurations of dislocations, so that it could be used to calculate the contrast maps of the distortion field caused by the coherent nonspherical precipitates, i.e. thin cylinders (discs).

One example of matching the experimental and the theoretical micrographs is presented in fig. 1a and 1b.

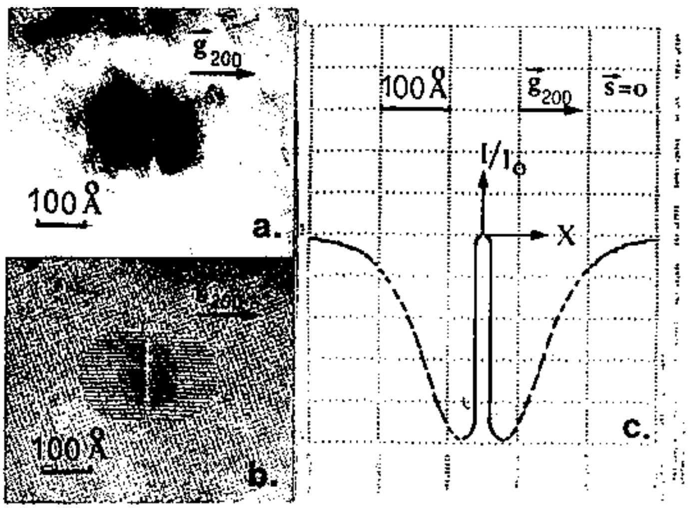

Fig. 1a) "Coffee-bean" contrast at the plate-like precipitate of the η-phase. b) contrast computed in the "dislocation loop" approximation. c) the contrast profile across the centre of the precipitate.

The intensity profile, or more precisely the contrast profile I/I_0 (I_0 being the background transmitted intensity far from the inclusion), is plotted against the line across the centre of the precipitate in the direction of $\vec{g}$ and presented in fig.1c. This profile and the simulated micrograph in fig. 1b were obtained with the use of the "dislocation loop" approximation[9], with the "equivalent" Burgers vector $b_{eq} = t.e^{c} = 1.2$ Å, t being the precipitate thickness. Approximately the same results were obtained using the theory of Eshelby[3] when aplied to thin precipitates. However, the latter[3] approach has the advantage of being applicable to thicker precipitates, when the "dislocation loop" approximation fails. The scales of grey recommended for printing simulated micrographs[6,8] were not quite suitable for the printers we used, although the qualitative matching with the experiment was good (compare fig. 1a and fig. 1b). The asymmetry of the experimental micrograph in relation to the "line of no contrast" (fig.1a) was probably due mainly to two reasons: 1. the precipitate is not exactly in the center of the foil as was supposed in the calculations, and 2. the crystal is out of the Bragg condition ($\vec{s} \neq 0$) in order to obtain a good contrast, while in calculations we took $\vec{s} = 0$.

The calculated intensity profiles and computed micrographs appeared to be very sensitive to various parameters, e.g. e^c, $\vec{s}$, depth of inclusion, foil thickness, etc. (See for example Ref.7). We also find it useful to note that the matrix method as developed by Thölén[8] is four to twenty times faster (depending on the parameters used in the programm) then the Runge-Kutta method of solving coupled linear differential equations appearing in the theory[4].

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