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QUENCHING DEFECTS IN SOLID SOLUTIONS AND THEIR EFFECT ON PRECIPITATION

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Résumé. — Le réseau de dislocation dans les alliages d'aluminium trempés est très différent de celui qui existe dans l'aluminium pur trempé. Ceci a été étudié par microscopie électronique (transmission) et par dilatométrie, et cet effet est probablement lié à l'absorption de paires atome dissous — lacune par les dislocations vis qui glissent, et à leur redistribution en boucles de dislocations prismatiques.

Le vieillissement entraîne la précipitation sur les dislocations et les effets peuvent être analysés à l'aide de la sous-structure de dislocations, du désaccord géométrique entre le précipité et le réseau de la matrice.

Abstract. — The dislocation structure of quenched aluminium alloys is very different from quenched pure aluminium. This effect has been studied by transmission electron microscopy and is probably associated with the absorption of vacancy-solute atom pairs on gliding screw dislocations and their redistribution as prismatic dislocation loops.

Ageing causes precipitation on dislocations and the effects can be analysed in terms of the dislocation substructure and the misfit geometry of the precipitate and matrix lattices.

The excess vacancies in a metal or alloy quenched from a high temperature may anneal out in a number of ways to produce a variety of dislocation substructures. The density of dislocations can be quite high, e.g. $10^9 - 10^{11}$ lines /sq.cm, and if the alloy is a supersaturated solid solution these dislocations can act as important sites for the segregation and subsequent precipitation of solute atoms.

The quench structures of some pure metals seem to follow the general prediction of Kuhlmann-Wilsdorf [1] that excess vacancies will collect to form small voids which subsequently collapse to give prismatic dislocation loops. However there are a number of anomalies such as the heterogeneous distribution of prismatic loops in quenched aluminium [2], the relatively few prismatic loops found in quenched copper and nickel and their apparent association with dislocation lines [3] and, most strikingly, the near absence of prismatic loops in concentrated aluminium alloys and the presence of long regular helical dislocations formed by the absorption of vacancies on screw dislocations [4]. Dilatometer experiments by Takamura [5] suggest that, in the case of gold, the size of the specimen and hence the amount of plastic deformation during quenching is an important factor affecting the behaviour of the excess vacancies. Recent experiments using transmission electron microscopy [6] have shown that this may be a general result and that the mode of influence of the plastic deformation is the absorption and redistribution of vacancies on gliding screw dislocations. We shall consider these results briefly.

Figure 1 is a typical electron micrograph of a quenched aluminium alloy. There are a large number of prismatic dislocation loops which have formed from homogeneously nucleated vacancy clusters and several regular helical dislocations. It is found that the proportion of vacancies which

Fig. 1. — Thin foil of Al-16 % Ag water quenched from 525 °C showing prismatic dislocation loops and helical dislocations ($\times 40,000$, printed $\times 0.79$).
boundaries where plastic flow would be expected during quenching due to stress concentrations existing at these points. This suggests that the formation of helices and rows of loops is associated with plastic deformation during quenching. To test this hypothesis, some Al-7 % Mg specimens were quenched in a jig so that (a) there was a minimum amount of plastic deformation during quenching and (b) the specimen was homogeneously deformed by about 2 % during quenching. The results of this experiment, shown in figure 2, are particularly striking since the presence of magnesium atoms in solid solution prevents homogeneous nucleation of dislocation loops from vacancy clusters after this treatment. Hence the loop structure arises solely from plastic deformation effects. Very few loops are observed in figure 2(a) where the amount of plastic deformation is restricted but in figure 2(b) many rows of loops and helices are observed along traces of \( \{111\} \) planes.

We now consider the behaviour of a glissile dislocation loop expanding in a matrix supersaturated with vacancies. The edge components of the dislocation can continue to glide conservatively while absorbing vacancies by climb but the movement of screw dislocations must be restricted due to their interaction with the vacancies and the consequent formation of jogs, cavities or simply loose atmospheres of vacancies. Eventually the dislocation must either cease to move or break free from its atmosphere leaving a row of defects behind. In the former case the vacancies will distribute themselves uniformly along the dislocation to give a helix while in the latter case many defects will be observed in the path of the dislocation. This suggestion enables us to give an approximate explanation of the great variety of quench structures observed in metals and alloys of high stacking fault energy. In metals like Al, Cu and Ni, vacancies are absorbed on gliding screw dislocations activated by quenching stresses and are re-distributed as prismatic dislocation loops. Thus in Al we observe patches of homogeneously nucleated dislocation loops separated by regions which have been active slip planes during quenching, showing irregular dislocations and dislocation loops formed by the re-distribution process. In Cu and Ni homogeneous nucleation appears to be difficult and hence dislocation loops are only observed in slip bands. In dilute Al alloys, typified by Al-7 % Mg in figure 2, the effect of the solute atoms is to restrict the dislocations to a single slip plane and hence the re-distributed defects are more closely aligned than in pure Al. In concentrated Al alloys the redistribution process seems to be prevented and hence each dislocation can only move a short distance before being immobilised by the absorption of vacancies. Thus more dislocations must be produced to relieve the quenching stresses and the result is a high density of helical dislocations.

Another type of quenching defect can be produced near particles in an alloy. If the particles are present at the homogenization temperature, the defects usually take the form of arrays of prismatic dislocation loops whose sign is determined by the relative thermal expansion coefficients of the particle and matrix [7]. However if the particle is
formed during quenching there may be a high local supersaturation of vacancies at the particle matrix interface which is sufficient to operate a Bardeen-Herring source [8]. Figure 3 is an example of the array of dislocation loops produced by such a mechanism.

We now consider the effect of this quench dislocation substructure on subsequent precipitation phenomena. Dislocations can affect precipitation in various ways chiefly by their action in causing segregation of solute atoms or by distortion of the matrix lattice so that it approaches the structure of the precipitate. A good example of the latter effect occurs in the Al-Ag system where the c. p. h. γ' transition precipitate nucleates on the narrow stacking faults at dislocations in the matrix [9]. A stacking fault in an f. c. c. matrix is, of course, equivalent to thin layer of c. p. h. lattice and hence...
this is an easy nucleation site for the γ precipitate. Figure 4 shows a helical dislocation which has partly transformed to a row of γ' precipitates. In this alloy system, dislocations appear to be the only nucleating sites for γ' precipitates.

In the Al-Cu system, the θ' transition precipitate is nucleated in the matrix and on dislocations so here the dislocation is only acting as a catalyst for precipitation. It is found that the dislocation only catalyses those orientations of precipitate where the Burgers vector of the dislocation can partly accommodate the misfit between the precipitate and the matrix [10]. Thus in figure 5, the Burgers vector of the dislocation is $\frac{a}{2} (110)$ and precipitates parallel to (100) and (010), with misfits in the $<100>$ and $<010>$ directions, have nucleated on the dislocations.

Figure 6 shows the formation of small rod-shaped precipitates on prismatic dislocation loops in an Al-Mg alloy. Precipitation has only occurred on one segment of the loop. Since these loops are formed by the Kuhlmann-Wilsdorf [1] reaction, the Burgers vector makes an angle of 55° to the plane of the loop. Hence only one segment of the loop is pure edge dislocation and it is this segment which is the most active nucleation site for precipitation. The dislocation loops shown in figure 3 are pure edge dislocation and hence precipitation occurs all round the loop (fig. 7).

We have shown that the dislocation substructure introduced by quenching is more complicated than was originally thought and that plastic deformation during quenching is important in determining the nature of the substructure. Alloys have more complicated substructures than pure metals owing to the effects of precipitates and solute atoms in solution. On ageing, dislocations act as preferential nucleating sites for precipitation and can also lead to the formation of some precipitate structures which would not be stable in a perfect lattice.

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REFERENCES