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Dislocation arrangement and density in deformed Al–Pd–Mn single-quasicrystals

P. Schall [a,*], M. Feuerbacher [a], M. Bartsch [b], U. Messerschmidt [b], K. Urban [a]

[a] *Institut für Festkörperforschung, Forschungszentrum Jülich GmbH, D-52425 Jülich, Germany*
[b] *Max-Planck-Institut für Mikrostrukturphysik, D-06120 Halle/Saale, Germany*

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Abstract

We present results of microstructural investigations on plastically deformed Al–Pd–Mn single-quasicrystals. Deformations were carried out in uniaxial geometry at different temperatures and to different values of plastic strain and in bending geometry. In the uniaxial geometry, various glide systems are activated. The dependence of the total dislocation density on temperature and on plastic strain is studied. Additional heat treatments on deformed samples show that recovery processes occurring at the high deformation temperatures significantly influence the deformation process. The experimental data are interpreted using a kinetic equation describing the evolution of the dislocation density during deformation. In the bending geometry a defined arrangement of geometrically necessary dislocations of edge type is observed. We find straight dislocation lines with line directions in the plane spanned by the direction of the applied force and the bending axis. The Burgers vectors have twofold direction corresponding to glide planes with the highest shear stress. © 2000 Elsevier Science B.V. All rights reserved.

Keywords: Quasicrystals; Plastic deformation; Dislocations

1. Introduction

In many respects, quasicrystals show plastic properties which are unusual in comparison with crystalline materials, especially metals. They are brittle at room temperature but at elevated temperatures of about 80% of the absolute melting temperature they show extensive ductile behaviour. The most prominent feature of the plastic behaviour is the absence of a work hardening regime at higher strains. The plastic properties of quasicrystals have mostly been investigated on icosahedral Al–Pd–Mn, which can be grown in the form of large single-quasicrystals of high structural quality. The thermodynamic activation parameters have been determined [1–3] and the microstructure of deformed samples was studied [4,5].

In this paper, we present a detailed study of the evolution of the dislocation density in uniaxially deformed Al–Pd–Mn quasicrystals and of the dislocation structure in samples deformed in bending geometry.

2. Deformation tests in uniaxial geometry

Deformation tests at a constant strain rate of $10^{-5}\,s^{-1}$ were performed on icosahedral Al–Pd–Mn single-quasicrystals at temperatures between $T = 695$ and 820°C. At 730°C deformations to different values of plastic strain were carried out. The deformed samples were studied in a transmission electron microscope. From the electron micrographs, the total dislocation density ρ was determined as described in [6] as a function of temperature and plastic strain.

Fig. 1(a) shows true stress–true strain curves at $T = 695$, 730, 790, and 820°C. The flow stress at the upper yield point decreases from 560 MPa at 695°C to 97 MPa at 820°C. Fig. 1(b) shows true stress–true strain curves of the deformations at 730°C. The end of each deformation test is marked by a full circle. The curves show good coincidence, and we observe a pronounced yield drop followed by a continuous decrease of the flow stress. Fig. 2(a)–(c) shows typical examples of the dislocation structure at different stages of deformation up to the lower yield point. A significant increase of the dislocation density can be seen.

The dislocation density ρ as observed in the samples subjected to the deformations shown in Fig. 1 is represented in Fig. 3(a) and (b) as a function of temperature and plas-

* Corresponding author. Tel.: +49-2461-612409; fax: +49-2461-616444.
E-mail address: p.schall@fz-jeulich.de (P. Schall).

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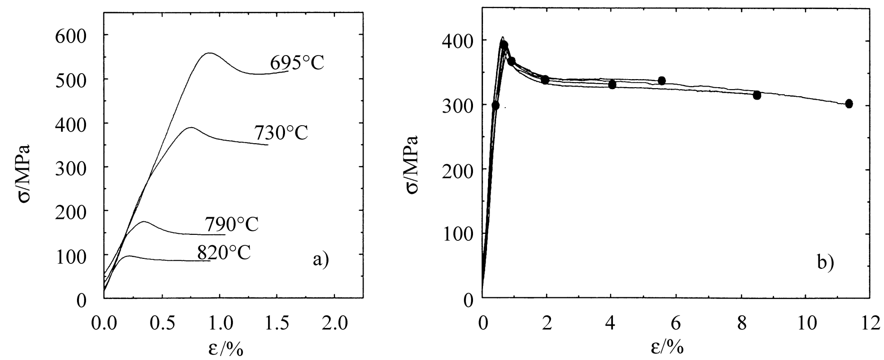

Fig. 1. True stress–true strain curves of icosahedral Al–Pd–Mn at a strain rate of $10^{-5}\,s^{-1}$ at different temperatures (a) and at 730°C by different amounts of plastic strain (b).

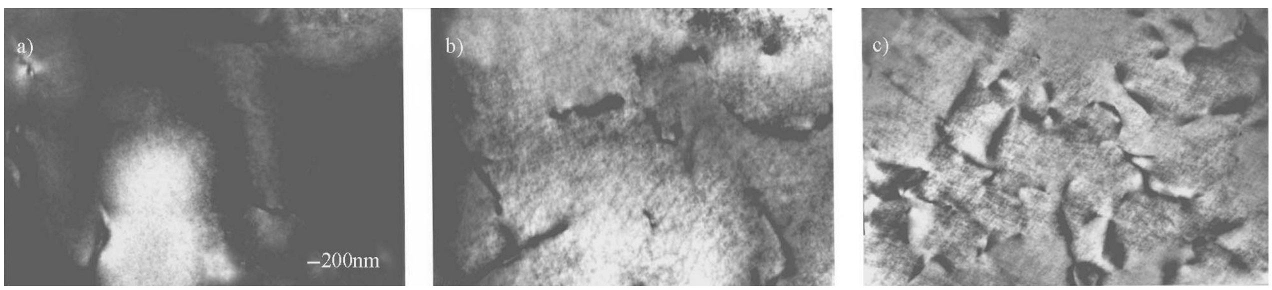

Fig. 2. Dislocation structures in icosahedral Al–Pd–Mn deformed by 0.1 (a), 0.3 (b), and 1.5% plastic strain (c).

tic strain $\varepsilon_{\mathrm{plast}}$, respectively. It is found to decrease strongly with increasing deformation temperature by about two orders of magnitude between 695 and 820°C. The development of ρ with plastic strain shows a linear increase at the onset of plastic deformation, a further nonlinear growth up to $12 \times 10^9\,\mathrm{cm}^{-2}$, and a subsequent decrease down to about 50% of the maximum value. A linear regression of the first five data points yields a multiplication constant of

$$M = 6.5 \times 10^{11}\,\mathrm{cm}^{-2}, \tag{1}$$

where M is assumed to be constant during deformation.

Heat treatments of different duration were performed on deformed material with the highest deformation induced dislocation density ($\varepsilon_{\mathrm{plast}} = 5.2\%$) at the same temperature as the deformation ($T = 730$°C). Fig. 4 depicts the dislocation density as a function of annealing time. During heat treatment of about 45 minutes the dislocation density decreases to about one third of the as-deformed value. The initial decrease of ρ can be well described by the phenomenological equation [7]

$$\left(\frac{\mathrm{d}\rho}{\mathrm{d}t}\right) = -A\rho^2 \quad \text{or} \quad \frac{1}{\rho} = At + B, \tag{2}$$

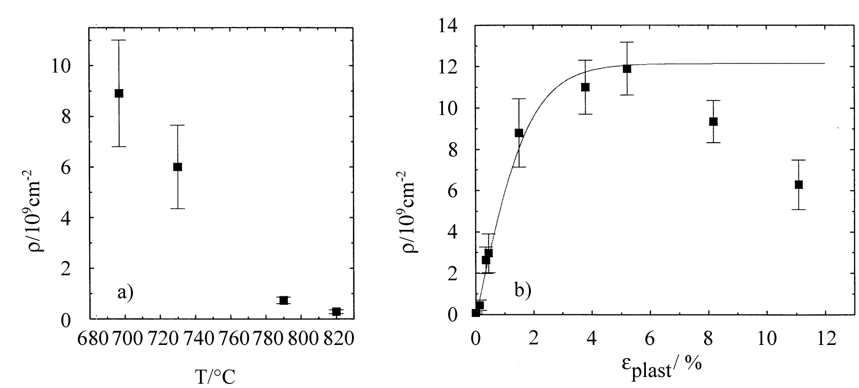

Fig. 3. Dislocation density ρ in deformed icosahedral Al–Pd–Mn as a function of temperature (a) and of plastic strain (b), experimental results (■) and calculation according to (3) (—).

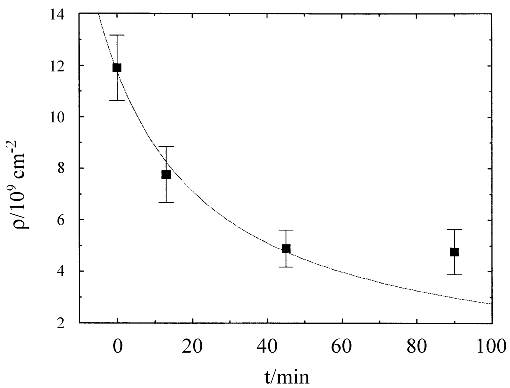

Fig. 4. Dislocation density ρ in deformed icosahedral Al–Pd–Mn ($\varepsilon_{\mathrm{plast}} = 5.2\%$, 730°C) as a function of annealing time t (■) and fit according to (2) (—).

where A and B are constants (Fig. 4). The rapid decrease in dislocation density upon annealing shows that recovery plays an important role in quasicrystal deformation behaviour. The annihilation rate and the multiplication constant can be entered into a kinetic rate equation balancing the dislocation density during deformation. The net rate of the variation of the dislocation density then results from the difference between the creation rate due to multiplication and the loss rate due to annihilation. Assuming a mechanism of static recovery as described by (2) yields.

$$\frac{\mathrm{d}\rho}{\mathrm{d}\varepsilon} = M - \frac{A}{\dot{\varepsilon}}\rho^2, \tag{3}$$

Eq. (3) describes a saturation of the dislocation density at a steady state value ρ_{ss} defined by $(\mathrm{d}\rho/\mathrm{d}\varepsilon)|_{\rho_{\mathrm{ss}}} = 0$. The development of the dislocation density calculated according to (3) with the experimentally determined values for M and A of (1) and (3) is shown in Fig. 3(b) in comparison with the experimental values for ρ. Up to a plastic strain of about 5% the calculated dislocation density shows good agreement with the experimental values, whereas the saturation behaviour in the second part deviates significantly from the measured dislocation density. The saturating behaviour as described by (3) is observed in many crystalline materials for the total dislocation density at high temperatures [8]. The decrease of the dislocation density in quasicrystal deformation can be understood as a result of structural changes of the material due to moving dislocations introducing disorder into the structure. The weakening of the structure due to dislocation motion can be qualitatively described by the cluster friction model [5]. In this model, the Mackay-type clusters, which are the elementary building blocks of the structure of icosahedral Al–Pd–Mn, are assumed to act as rate-controlling obstacles to dislocation motion. The net density of clusters is reduced by moving dislocations introducing disorder into the structure. This structure change leads to an increased dislocation mobility. Therefore, at high strains less dislocations are needed to maintain the constant strain rate and the dislocation density decreases with increasing strain. In terms of the steady-state dislocation density ρ_{ss}, the dislocation densities measured at high strain are values of a dynamic equilibrium in a succession of steady states belonging to different structural states.

3. Deformation tests in bending geometry

Samples of about 1.2 mm × 1.8 mm × 7 mm in size, the long axis oriented parallel to a fivefold direction, were deformed in bending geometry. A load of about 160 N was applied at a temperature of 710°C. Immediately after unloading, the samples were quenched in water. Fig. 5 shows bright field electron micrographs of the dislocation structure as observed in a specimen cut perpendicular to the long axis of the bent sample. We observe an arrangement of long straight dislocation lines with five different line directions. From the length of the segments the inclination of the dislocation lines with respect to the foil plane of the microscope sample is estimated to be less than 2° assuming a specimen thickness of 150 nm. This inclination can be explained by the inaccuracy of the orientation of the foil plane

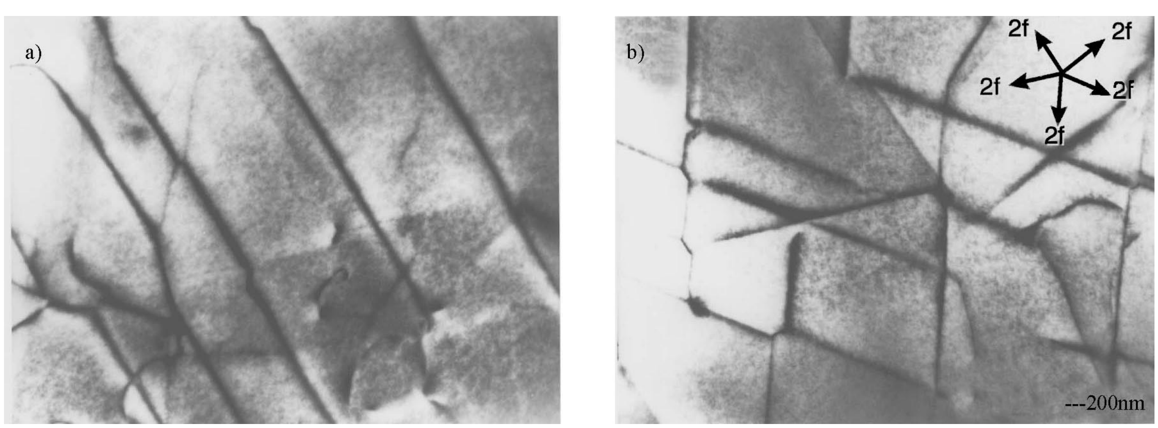

Fig. 5. (a) and (b). Dislocation structures in bent icosahedral Al–Pd–Mn samples.

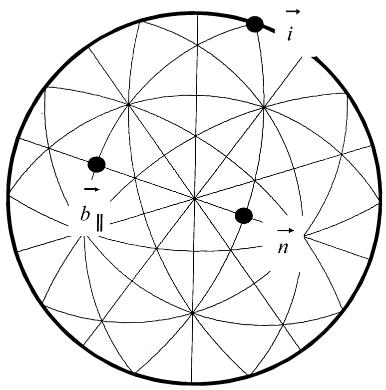

Fig. 6. Stereographic projection showing the Burgers vector direction $\vec{b}_{||}$, the line direction $\vec{l}$ and the glide plane normal $\vec{n}$ of one dislocation.

with respect to the long axis of the bent sample. We conclude that the line directions of the dislocations are perpendicular to the long axis of the deformed sample. A comparison with diffraction images shows that within this plane perpendicular to the long axis the dislocations are aligned parallel to the five twofold directions perpendicular to the fivefold axis.

Contrast extinction studies were performed to determine the direction of the Burgers vector in physical space, $\vec{b}_{||}$. We find a strong extinction condition (SEC) [9] if we choose a reciprocal lattice vector $\vec{g}_{||}$ parallel to the line direction of the dislocation for imaging under two-beam conditions. It follows that the Burgers vectors $\vec{b}_{||}$ of the dislocations are perpendicular to their line directions, that is the dislocations are of pure edge type. A second SEC was used to determine the direction of $\vec{b}_{||}$ completely. The Burgers vectors were found to be parallel to twofold directions inclined to the fivefold long axis by 58°. Fig. 6 shows an example of the direction of the dislocation line $\vec{l}$, the Burgers vector $\vec{b}_{||}$ and the normal of the corresponding glide plane, $\vec{n} = \vec{l} \times \vec{b}_{||}$, for one dislocation in a stereographic projection seen along the fivefold long axis of the sample. Rotating the stereographic projection by $n(2\pi/5)$, where $n = 0 \dots 4$, yields all five glide systems activated during the bending process of the fivefold oriented sample.

A finite element calculation of the stress field in the sample under the present bending conditions shows that, as is well known, the bending deformation can be regarded in a good approximation as pure tension and compression parallel to the long axis in the upper and lower parts of the sample, respectively. Then the driving force for dislocation glide is proportional to the Schmid factor $m_s = \cos\varphi\cos\lambda$, where φ and λ are the angles between the long sample axis and the normal of the glide plane, and the long sample axis and the Burgers vector of the dislocation, respectively. The Schmid factor takes on values in the range $0 < m_s < 0.5$. In the bending geometry where the dislocations have been shown to be of pure edge type with line directions perpendicular to the long axis (Fig. 6) the Schmid factor can be expressed as $m_s = \cos\lambda\sin\lambda$. The observed dislocations with a twofold Burgers vector inclined by $\lambda = 58^\circ$ to the long axis have a high Schmid factor of 0.447. However, dislocations with a Burgers vector along the other twofold direction under $\lambda = 32^\circ$ to the fivefold long axis have the same Schmid factor. Dislocations with these Burgers vectors were not observed in the present investigation, indicating that the corresponding slip system is not activated in this deformation geometry.

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