A R C H I V E S O F M E T A L L U R G Y A N D M A T E R I A L S
Volume 50 2005 Issue 3

M.S. SZCZERBA*, A. PIĄTKOWSKI**, P. PAŁKA*, M. PEREK* T. TOKARSKI*

TWIN-TWIN INTERACTIONS IN Cu-8AT.%Al SINGLE CRYSTALS DEFORMED AT 4.2K

ODDZIAŁYWANIE BLIŹNIAK-BLIŹNIAK W MONOKRYSZTAŁACH Cu-8AT.%Al. ODKSZTAŁCANYCH W 4,2 K

The 4.2 K plastic deformation mechanism of Cu-8at.% Al single crystals oriented for double glide in tension is analyzed. Special attention is paid to crystallographic and structural effects of the intersection of first order twins. On the basis of mechanical tests performed, x-ray measurements and electron microscopy observations it was found that the intersection of the first order twins leads to a formation of second order twins, and what is important to the appearance of some crystal regions of crystallographic orientation, which do not relate according to twin orientation neither to the matrix nor to the first order twins. The experimental evidence of formation of the additional crystallographic orientation regions and the second order twins confirm the theoretical predictions based on disclination modeling of twin-twin interactions [1], as well as, other experimental observations of mechanical twins intersections in polycrystalline austenitic steels [2]. The obtained results should contribute to better understanding of the mechanisms of deformation texture formation both in single crystals and polycrystalline fcc metals and alloys with low stacking fault energy. A suggestion is also put forward that crystallographic description of dislocations inherited by the second order twins should take into account the product of correspondence matrices $[C_I][C_{II}]$ connected with the first and second order twins, respectively.

Keywords: twin interactions, fcc single crystals

W pracy poddano analizie mechanizm odkształcenia plastycznego monokryształów Cu-8% at. Al. rozciąganych w temperaturze 4.2 K wzdłuż orientacji faworyzującej poślizg pdwójny. Szczególną uwagę zwrócono na krstalograficzne i strukturalne skutki procesu przecinania się bliźniaków pierwszego rzędu. Na podstawie przeprowadzonych badań mechanicznych, pomiarów dyfrakcji rentgenowskiej oraz obserwacji elektrono-mikroskopowych stwierdzono, iż przecinanie się bliźniaków pierwszego rzędu prowadzi do powstania w odkształcanym krysztale bliźniaków rzędu drugiego oraz obszarów o orientacjach krystalograficznych,

* DEPARTMENT OF STRUCTURE AND MECHANICS OF SOLIDS, AGH UNIVERSITY OF SCIENCE AND TECHNOLOGY, 30-059 KRAKÓW, AL. MICKIEWICZA 30, POLAND

** INSTITUTE OF METALLURGY AND MATERIALS SCIENCE, POLISH ACADEMY OF SCIENCES, 30-059 KRAKÓW, UL. REYMONTA 25, POLAND

które nie pozostają w relacji bliźniaczej zarówno względem sieci osnowy jak i bliźniaków pierwszego rzędu. Uzyskane w pracy doświadczalne dowody powstawanie w miejscach przecięć bliźniaków pierwszego rzędu obszarów o "nowych" orientacjach krystalograficznych, oraz bliźniaków drugiego rzędu potwierdzają przewidywania teoretyczne oparte o modele dysklinacyjne [1], jak również inne obserwacje doświadczalne przecinania się bliźniaków odkształcenia w stalach austenitycznych [2]. Uzyskane w pracy wyniki badań powinny przyczynić się do głębszego poznania mechanizmów powstawania tekstur odkształcenia w monokryształach i polikryształach metali i stopów RSC o niskiej energii błędu ułożenia. W pracy wysunięto również sugestię iż krystalograficzny opis dyslokacji dziedziczonych przez bliźniaki drugiego rzędu winien postępować zgodnie z iloczynem macierzy korespondencji $[C_I][C_{II}]$ związanych odpowiednio z bliźniakami pierwszego i drugiego rzędu.

1. Introduction

Tensile deformation at room temperature of the Cu-8at.% Al single crystals along original orientation axis close to [112] crystallographic direction results in formation of the first order twins (it means twins inside the crystal matrix regions), when the tensile axis oscillating around the symmetry line [001]-[111] of the basic stereographic triangle reaches during the deformation the second overshoot position [3]. This situation changes the decrease of the deformation temperature down to liquid helium temperature where the first twins form as early as the tensile axis reaches already the position of the first overshoot [4], which is at much lower tensile strain. Thus, one may expect that continuation of low-temperature tensile deformation up to the second overshoot leads to activation of another family of mechanical twins, and consequently to twin-twin intersection process, inevitably. The aim of this paper is presenting a way to generate twin-twin interactions in the structure of the deformed Cu-8at.% Al single crystals in a controlled manner, and to bring a description of the most important crystallographic and structural effects of these interactions.

2. Experimental method

The Cu-8at.% Al single crystals in the shape of rectangular prism with dimensions $3\times3\times70mm^3$ and main axis orientation [$\bar{1}$ 1 2] (tensile direction TD), and side walls P1 and P2 parallel to crystallographic planes (1 1 0) and (1 $\bar{1}$ 1) were deformed in tension at liquid helium temperature. The 4.2 K tensile test was conducted with initial strain rate $10^{-4}s^{-1}$ at testing machine of special construction adapted to cryogenic measurements of mechanical properties of metals and alloys single crystals. The x-ray measurements and electron microscopy observations were carried out on D8 Advance Bruker diffractometry system and Philips CM-200 electron microscope, respectively. Special attention was paid to the crystallographic and structural effects of the twin-twin interactions occurring in the tensile Cu-8at.% Al single crystals right after the conditions of the second overshoot had been reached. More details of the experimental procedure see further in the text.

3. Results and discussion

It is useful to divide the obtained low-temperature mechanical characteristics (true tensile stress-true tensile strain, $\sigma - ln(l/lo)$) of studied Cu-8at.% Al single crystals into three stages (Fig. 1) The stage I concerns the initial deformation region ($ln(l/lo) \leqslant 0.2$), where the flow stress σ changes quickly from the initial yield point of about 60 MPa up to 140 MPa. At the end of the stage I the deformed crystals reach the state of the first overshoot illustrated by the crystallographic position of tensile axis deviated about 3° from the [0 0 1]-[$\bar{1}$ 1 1] edge of the basic stereographic triangle towards the [$\bar{1}$ 0 1] direction which is the slip direction of the dominant deformation system BIV (Fig. 2, see also the terminology of slip and twin systems in fcc lattice introduced in [3]). In such situation the C3 twin system becomes stressed enough to be activated. Just at the flow stress of 140 MPa, or alternatively at the tensile strain $ln(l/lo) = 0.2$, the dominant slip system BIV is being replaced by the first order twinning of the C3 twin system (T_{C3}), and the stage II of the tensile characteristics begins. Taking into account current orientation of the tensile direction the critical stress of the T_{C3} twinning resolved on the twin plane and twin direction is about 60 MPa. The stage II takes the deformation interval $0.2 \leqslant (ln(l/lo) \leqslant 0.95$, and is connected with further substantial increase of flow stress σ up to 360 MPa. During this stage, the deformation

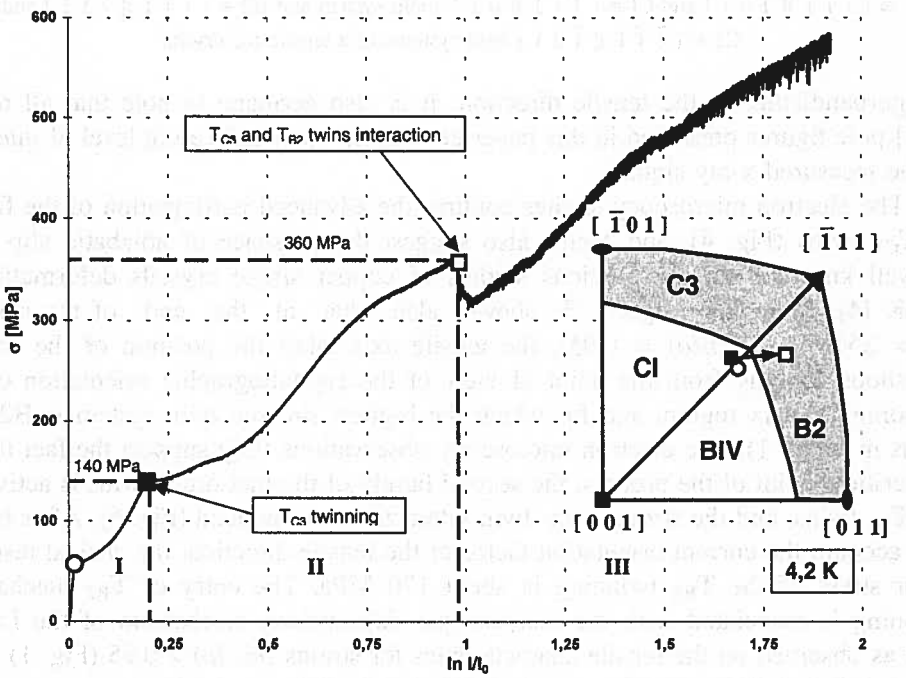

Fig. 1. The 4.2 K tensile curve of the Cu-8at.% Al single crystals oriented for double glide. The insert shows the first (■) and the second (□) overshoot position of the tensile axis

of the single crystals proceeds due to T_{C3} twinning supported mainly by slip operating in BIV and CI systems, the highest stressed slip systems (Fig. 2).

The experimental evidence for significant participation of the T_{C3} twins at the end of the II stage of crystal deformation is shown in figure 3 which shows the data of the {111} pole figures measurements registered in three different planes parallel to P1, P2 and P3 faces of the deformed crystals, respectively. It is to note that P3 crystal face is

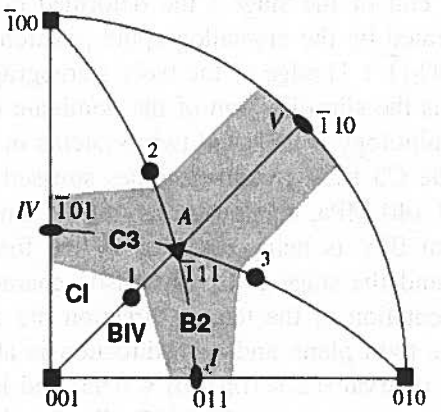

Fig. 2. [0 0 1] stereographic projection representation of Schmid's law for the choice of BIV ≡ (1 1 1)[$\bar{1}$ 0 1] and CI ≡ ($\bar{1}$ $\bar{1}$ 1)[0 1 1] slip system and B2 ≡ (1 1 1)[$\bar{2}$ 1 1] and C3 ≡ ($\bar{1}$ $\bar{1}$ 1)[$\bar{1}$ 2 1] twin systems of a tensile fcc crystal

perpendicular to the tensile direction. It is also germane to note that all of the {111} pole figures presented in this paper contain numbers indicating level of intensity of the measured x-ray signals.

The electron microscopy studies confirm the advanced participation of the family of T_{C3} twins (Fig. 4), and seems also suggest the presence of adiabatic slip (Fig. 5) well known from the previous studies of copper single crystals deformation at 4.2 K [4]. Moreover, figure 3 shows also that at the end of the stage II (σ = 350MPa, $ln(l/lo)$ = 0.95), the tensile axis takes the position of the second overshoot, looking from the point of view of the crystallographic orientation of the remaining matrix region, and for which the highest stressed twin system is B2 (see insert in figure 1). The electron microscopy observations fully support the fact that at this critical point of the process, the second family of the first order twins is activated, the T_{B2} twins, and the strong twin- twin interactions are noticed (Fig. 6). After taking into account the current orientation factor of the tensile direction, the critical resolved shear stress of the T_{B2} twinning is about 170 MPa. The entry of T_{B2} mechanical twinning is associated with the macroscopic deformation mechanism of the Lüders type as observed on the tensile characteristics for strains $ln(l/lo) > 0.95$ (Fig. 1).

Another analysis of the measured {111} pole figures, similar to that of figure 3 but carried out on the deformation stage III samples, disclose important fact of the appearance of crystal regions, which do not remain according to twin orientation

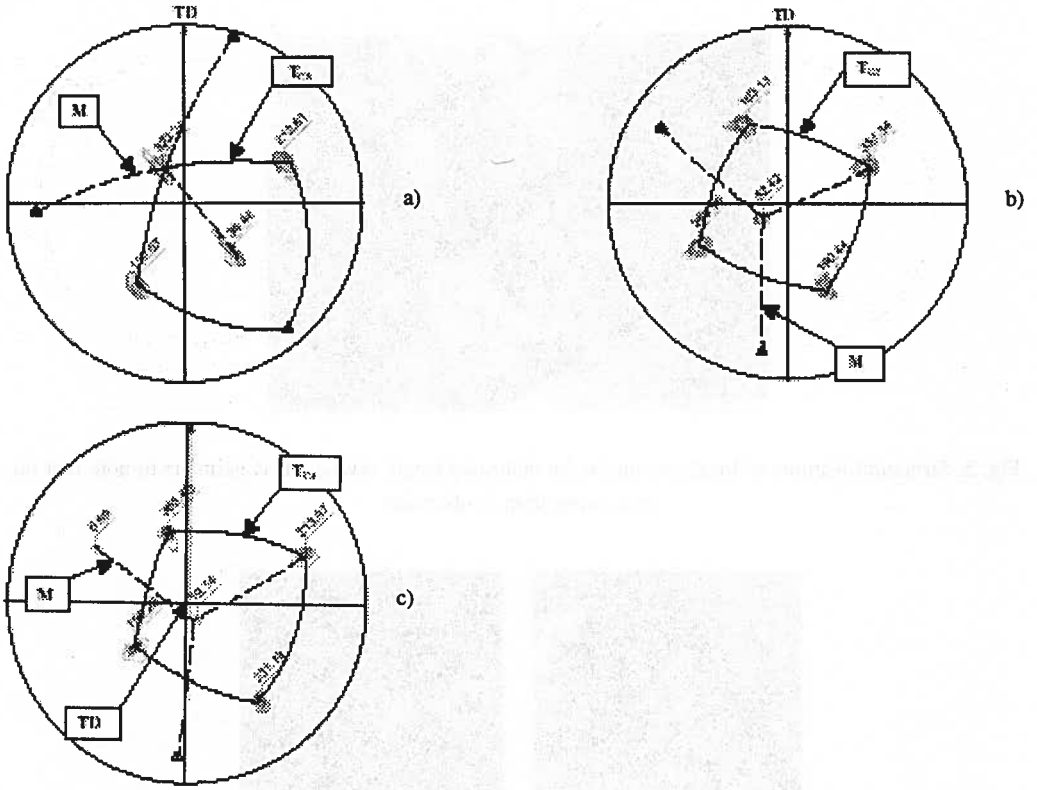

Fig. 3. {111} pole figures recorded in three planes of the deformed single crystals at the end of the stage II of deformation; a) plane P1, b) plane P2, c) plane P3

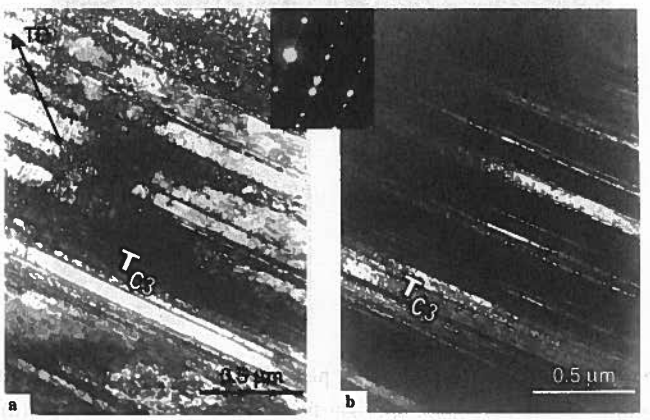

Fig. 4. TEM pictures of the thin foil parallel to the plane P2 of the deformed crystals (a) bright field, (b) dark field of T_C twinning operating at the end of the stage II of 4.2 K tensile deformation

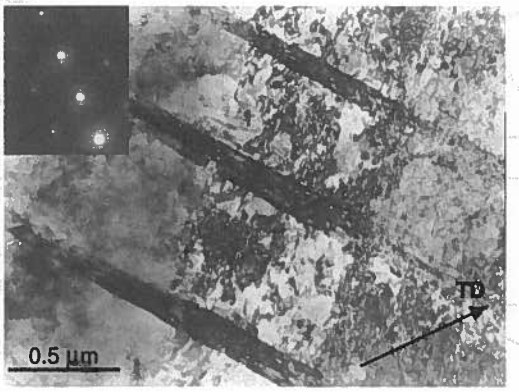

Fig. 5. Structural features of localized slip in the deformed single crystals. It is germane to note that no twin orientation is observed

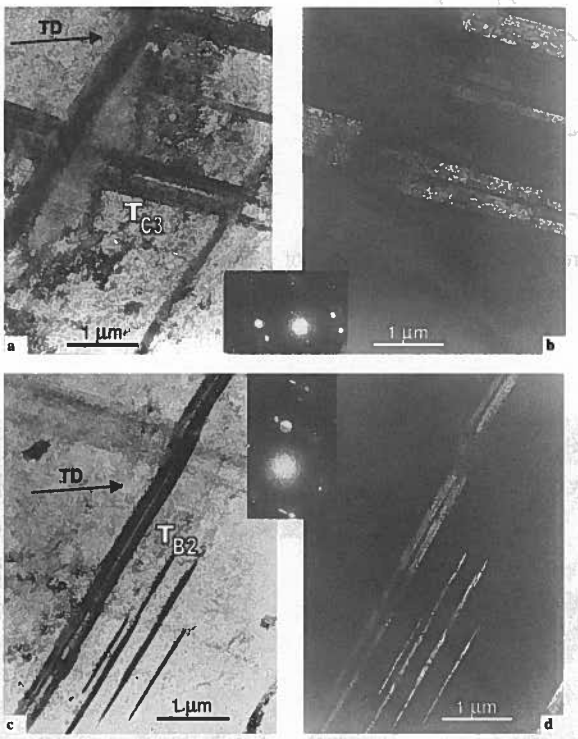

Fig. 6. TEM pictures of the thin foil parallel to the plane P2 showing twin-twin interactions (T_{C3} – T_{B2}) at the stage III of 4.2 K tensile deformation; bright field (a),(c) and dark field (b),(d) image of the T_{C3} twins and the T_{B2} twins, respectively

relationship neither to the lattice of the T_{C3} twins nor to the lattices of the vanishing matrix and practically vanished new generated T_{B2} twins (Fig. 7). The necessity of the appearance of crystal regions of the new orientation ("N") is well-suggested in the disclination modeling of twin-twin interactions [1]. According to the theory, the successful penetration of an incident twin across another barrier twin must result in the appearance in crystal lattice the "N" orientation belonging exactly to the first order twins intersected regions which are accommodated by quadrupole disclination arrangements (Fig. 8). Moreover, the practical lack of x-ray signals from the lattices of the T_{B2} twins and the vanishing matrix regions prove operation of the prevailing process of reactivation of the T_{C3} twinning, as suggested by the theory [1]. Figure 9 shows schematically the growth of the "N" orientation regions at the expense of the vanishing matrix and T_{B2} twins.

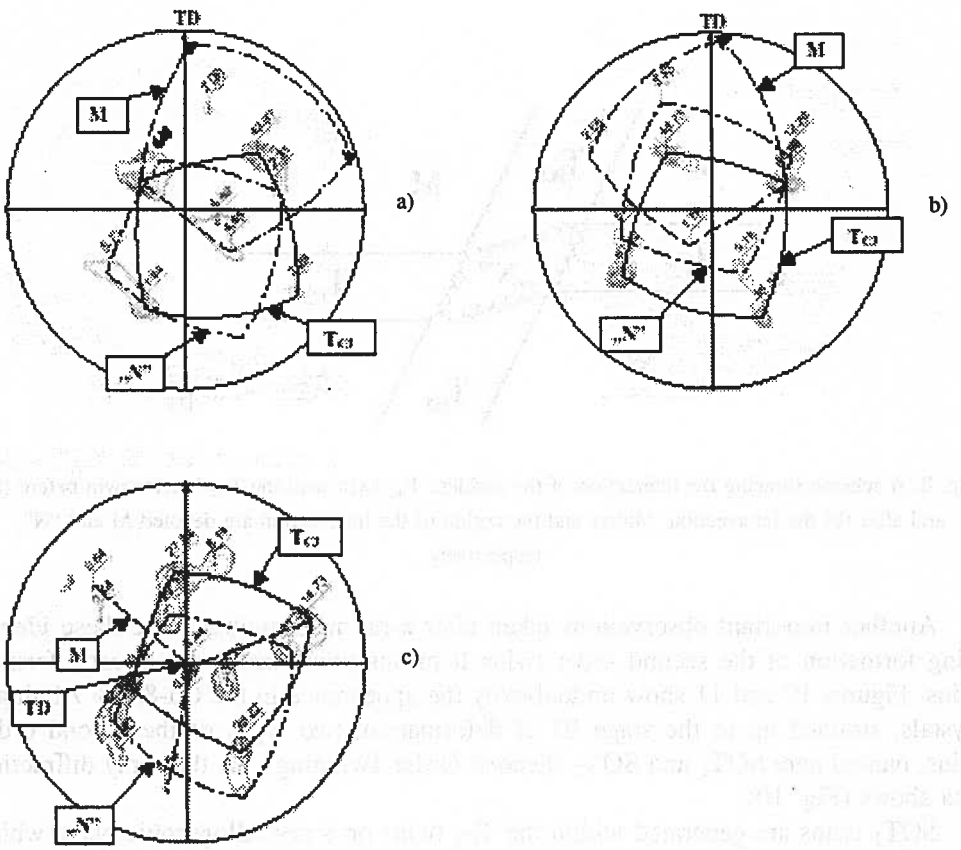

Fig. 7. {111} pole figures recorded in three planes of the deformed single crystals showing formation of the "N" orientation crystal regions; a) plane P1, b) plane P2, c) plane P3

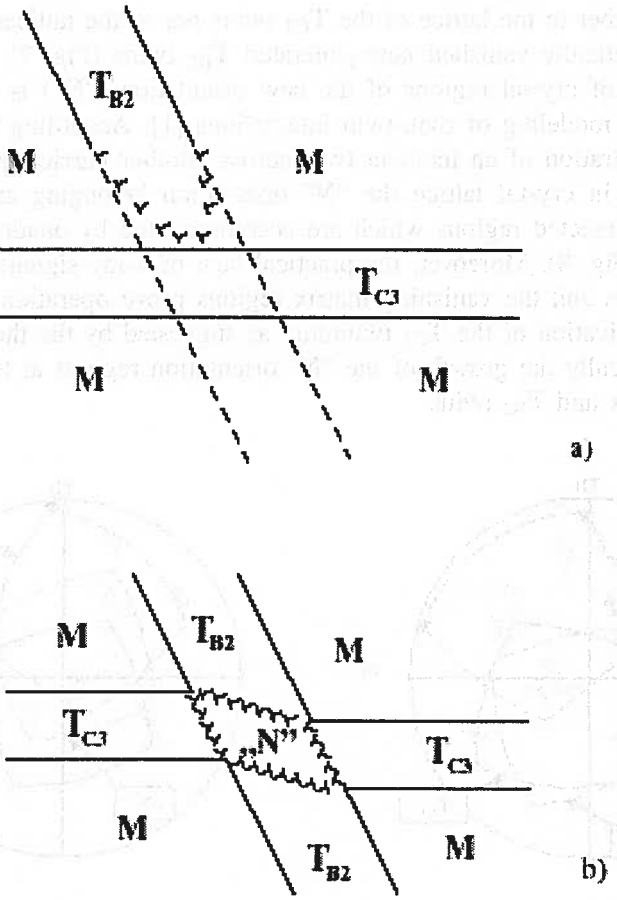

Fig. 8. A scheme showing the interaction of the incident T_{B2} twin with the T_{C3} barrier twin before (a) and after (b) the intersection. Matrix and the region of the intersection are denoted M and "N", respectively

Another important observations taken after x-ray measurements are these identifying formation of the second order twins It means twins inside the already formed twins. Figures 10 and 11 show undoubtedly the appearance in the Cu-8at.% Al single crystals, strained up to the stage III of deformation, two types of the second order twins, named here SOT_1 and SOT_2 (Second Order Twinning). As the x-ray diffraction data shows (Fig. 10),

SOT_1 twins are generated within the T_{C3} twins on a crystallographic plane which was inherited from the non-K1 and non-K2 {111} type planes of the matrix (see terminology of the twinning geometry established in [5] and the geometrical predictions of the correspondence matrix method drawn in [6]). Whereas, figure 11 proves that second order twins (SOT_2) may also form within the crystal regions connected with the "N" orientation. The results of electron microscopy studies confirm fully the presence

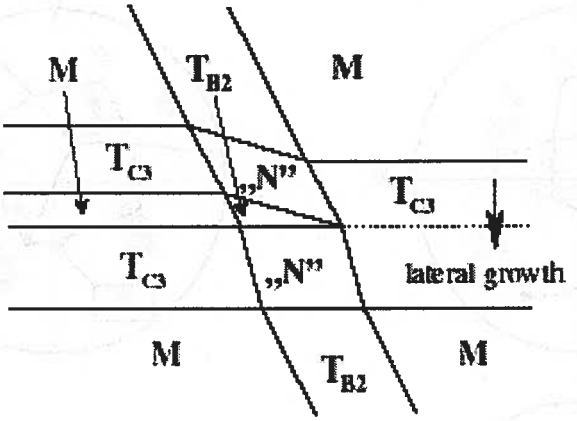

Fig. 9. A scheme showing crystallographic consequences of the reactivation of the T_{C3} twins. Note the lateral growth of the T_{C3} twin and the "N" orientation at the expense of matrix and T_{B2} twin crystal regions

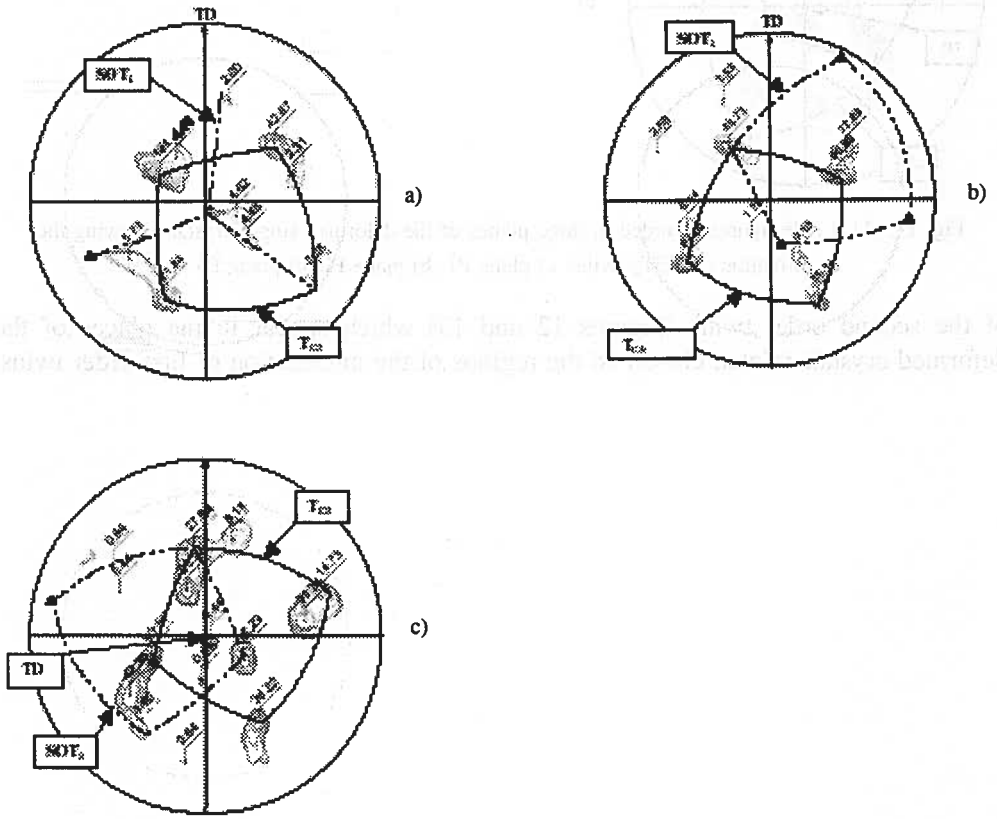

Fig. 10. {111} pole figures recorded in three planes of the deformed single crystals showing the formation of the SOT_1 twins; a) plane P1, b) plane P2, c) plane P3

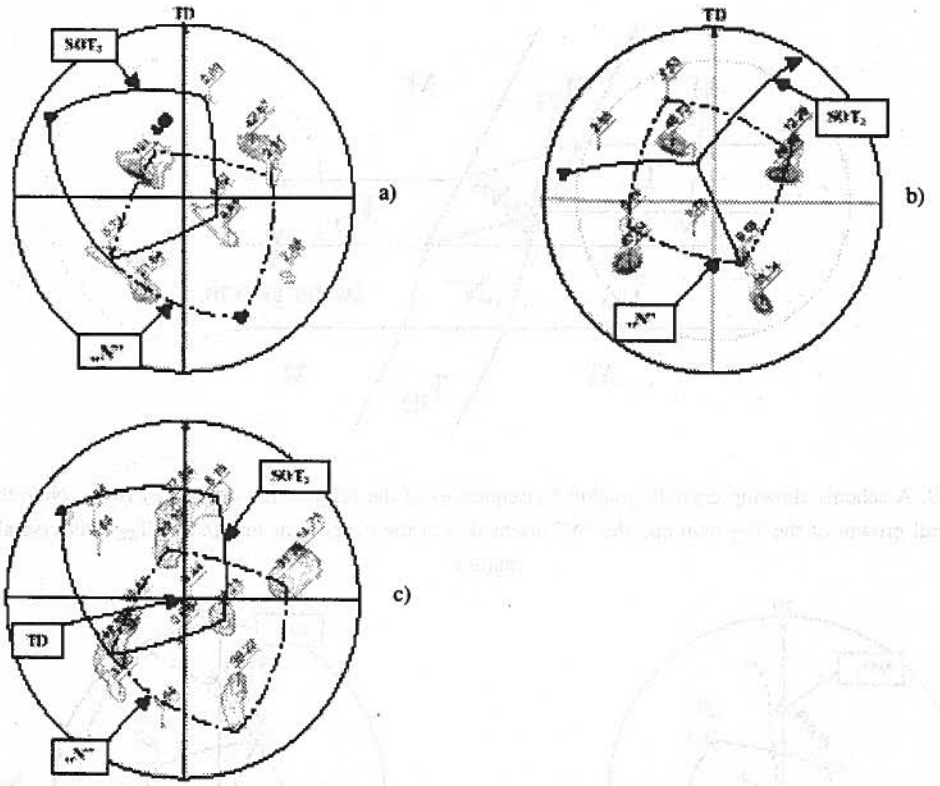

Fig. 11. {111} pole figures recorded in three planes of the deformed single crystals showing the formation of SOT_2 twins; a) plane P1, b) plane P2, c) plane P3

of the second order twins (Figures 12 and 13) which appear in the places of the deformed crystals related closely to the regions of the intersection of first order twins.

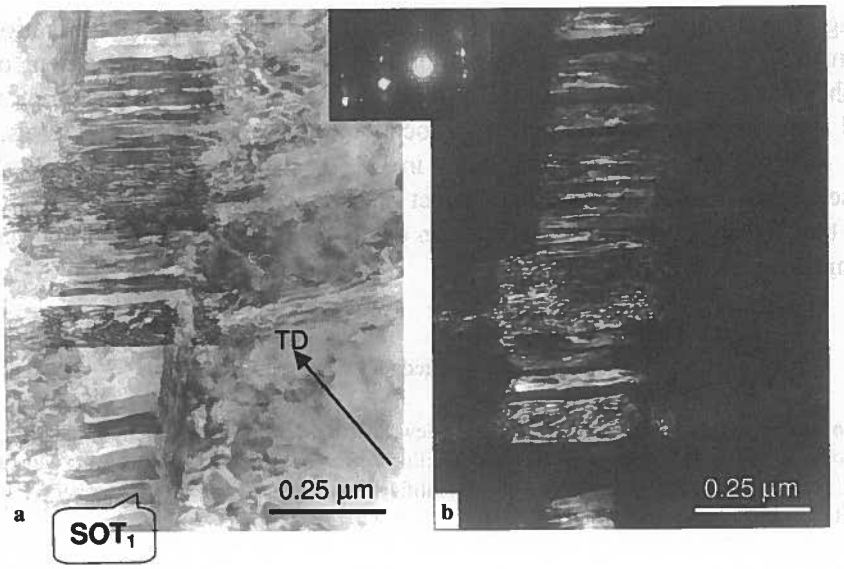

Fig. 12. TEM images (a) bright field, (b) dark field of SOT_1 twins

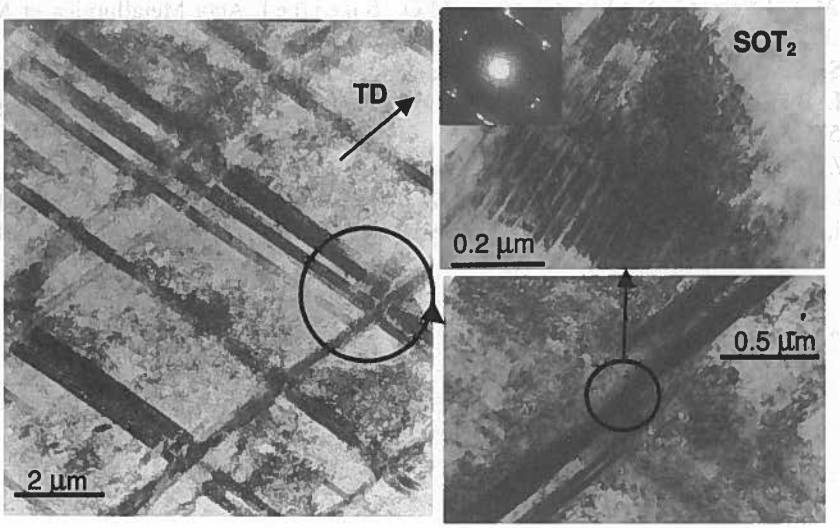

Fig. 13. TEM images of SOT_2 twins

4. Summary

Summarizing, it is worth underlining that the obtained results on the twin-twin interactions should contribute to better understanding of the mechanisms of deformation texture formation in single crystals and polycrystalline fcc metals and alloys with low

stacking fault energy. The work clearly suggests the necessity of incorporation of the mechanism of the "N" orientation formation into the deformation phenomena of highly strengthened fcc metals and alloys. The need of further electron microscopy work on second order twin transformation of dislocation substructure of fcc crystals, similar to that initiated for the first order twins in [6], is also suggested. The development of these studies should include the product of correspondence matrices $[C_I][C_{II}]$ as a proper tool for crystallographic description of dislocations inherited during the second order mechanical twinning.

Acknowledgements

One author (MSS) is indebted to Dr M. Niewczas of McMaster University, Hamilton, Ontario, Canada of permission to use the Z.S. Basinski facilities of cryogenic mechanical testing laboratory. The financial support of the State Committee for Scientific Research under the grant number 11.11.180.255 and 10.10.180.281 is also acknowledged.

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Received: 10 May 2005.