Volume 54 2009 Issue 1

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EXPERIMENTAL STUDIES OF A CHANGE OF DOMINANT SLIP SYSTEM IN TENSILE CU-6at.%Al SINGLE CRYSTALS

ANALIZA DOŚWIADCZALNA ZMIANY DOMINUJĄCEGO SYSTEMU POŚLIZGU W ROZCIĄGANYCH MONOKRYSZTAŁACH Cu-6%at.Al

The paper contains quantitative analysis of a change of dominant slip system in Cu-6%at.Al single crystals of the initial tensile orientation near the <112> crystallographic direction deformed at room temperature. It was proved that the dominant slip system changes three times during tension of the single crystals and each of the change is associated by a sudden decrease of work hardening rate and plastic strain localization. Measurements show that cumulative shear strain of the dominant slip system is about half of the order of a magnitude larger than for any other active secondary system. It was also proved that the change of the dominant slip takes place when hardening rate of the dominant system becomes equal to hardening rate of the system that takes over the domination.

Praca zawiera ilościową analizę zjawiska zmiany dominującego systemu poślizgu w rozciąganych w temperaturze otoczenia monokryształach Cu-6%at.Al o orientacji początkowej bliskiej kierunkowi krystalograficznemu <112>. W pracy wykazano, że dominujący system poślizgu zmienia się trzykrotnie i każdej zmianie towarzyszy gwałtowny spadek prędkości umocnienia odkształceniowego kryształu oraz lokalizacja odkształcenia plastycznego. Przeprowadzone pomiary pokazują, że skumulowane odkształcenie ścinające w systemie dominującym jest o około pół rzędu wielkości większe niż w każdym innym aktywnym systemie wtórnym. W pracy zostało również wykazane, że zmiana dominującego systemu poślizgu ma miejsce wtedy, gdy prędkości umocnienia dotychczas dominującego systemu oraz systemu przejmującego dominacje są sobie równe.

1. Introduction

During tension of FCC single crystals with orientation from a center of the standard stereographic triangle only one slip system (primary) dominates the deformation process. Taylor and Elam conducting their studies on the Cu and Al single crystals observed one of the fundamental consequences of the phenomenon, *i.e.* rotation of crystal lattice which is usually illustrated on the stereographic projection as a movement of the crystal tensile axis towards a pole of the direction of the primary slip system [1]. They observed the following situation: when the tensile axis crosses a symmetry line and lays in the different stereographic triangle where another slip system (conjugate) is stressed the most, the crystal lattice still continues the rotation towards the same direction. This behavior was called overshoot. Since then, the overshoot have been the subject of many studies [see review paper 2] that confirmed its generalization and importance in plasticity of FCC crystals. Following that, in the paper [3] special attention was drawn to the point of the end of the overshoot (Fig.1). It was found that the end of the overshoot generates tensile instability associated with inhomogeneous plastic deformation of the Lüders type or sudden necking leading to a rupture. At this point of the tensile deformation (Fig.1b), a sudden drop of the values of work hardening rate on $d\sigma/d\varepsilon = f(\varepsilon)$ curves is also observed (Fig.1a). It should be emphasized that the beginning of the tensile instability takes place before the Considère criterion is met, where $d\sigma/d\varepsilon = \sigma$. In paper [3] it was also shown the mechanical instability that appears at the end of the overshoot is caused by a change of the dominant slip system and the critical parameters of the phenomenon (stress, strain) depend on the initial orientation and the stacking fault energy of the deformed single crystal. On the other hand very systematic studies of the amount of dominant and secondary slip in copper single crystals of different orientation deformed in tension at room temperature are reported in [4]. The data show that at the end of overshoot the cumulative primary

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slip is usually of the order of a magnitude greater than that associated with secondary slip systems and that the changeover from primary into conjugate slip dominance takes place. However, in most cases of copper crystal orientation used the amount of slip in the new dominnat slip system was either impossible to determine due to the neck formation or the obtained data could be affected by inhomogenous distribution of macroscopic crystal strain due to the Lüders type deformation. In this paper the quantitative determination of dominant and secondary slip is analysed for copper-aluminium single crystals deformed in tension at room temperature. The choice of chemical composition and the initial alloy crystal orientation allows to study more efficiently the amount of the new dominant and the new secondary slip. To perform the quantitative analysis of the change of dominant slip, the full deformation matrix will be calculated from current orientation and shape measurements at succesive strain intervals for Cu-6at.%Al single crystals of the initial tensile axis located near the <112! > crystallographic direction in the basic stereographic triangle.

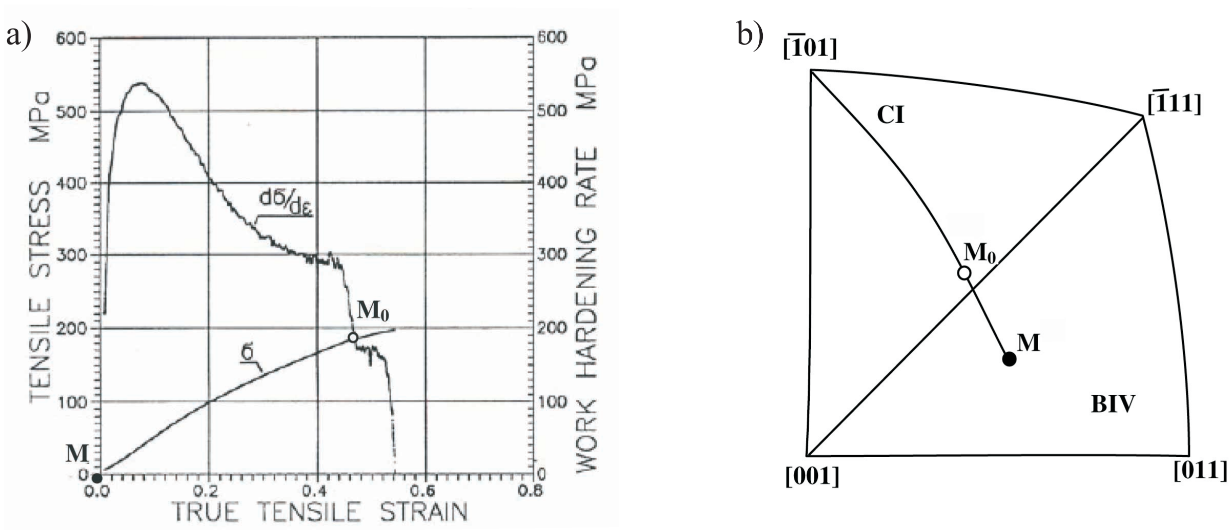


Fig. 1. The tensile stress and the work hardening rate versus true tensile strain (a), for 0.5 orientation copper single crystals. **M** and $\mathbf{M_0}$ indicate the initial and the end of overshoot position of the deformed copper crystal. **BIV** and **CI** stereographic triangles where the primary **BIV** ≡ (111)[-101] and conjugate **CI** ≡ (-1-11)[011] slip systems are stressed the most, respectively

The aim of the following paper is to perform the quantitative experimental analysis of the distribution of macroscopic crystal strain in all independent slip systems before and after the change of the dominant slip as well as to determine the mechanical condition of the transformation.

2. Geometrical basis

The quantitative analysis of the distribution of the macroscopic crystal strain in independent slip systems requires determination of a deformation gradient matrix **F** firstly, which relates the shape of a sample before and after the deformation (Fig.2).

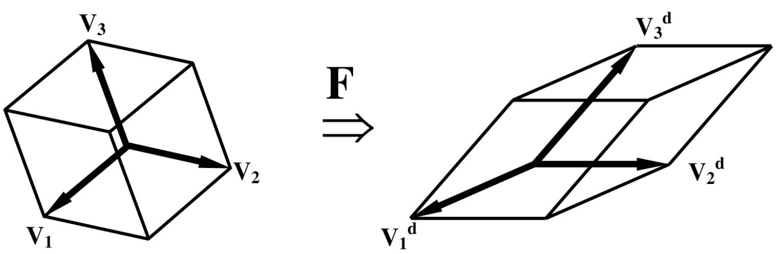


Fig. 2. The geometrical illustration of the deformation gradient matric **F** which relates the undeformed [**V**] and deformed [$\mathbf{V^d}$] configuration of the vector bases

Let vectors **V1**, **V2** and **V3** describe the sample before deformation, and vectors $\mathbf{V1^d}$, $\mathbf{V2^d}$ and $\mathbf{V3^d}$ the deformed one. The change of crystallographic geometry of the sample that underwent deformation can be described with equation (1), where components of the vectors **V** and $\mathbf{V^d}$ are written in columns of the matrix, [**V**] and [$\mathbf{V^d}$] respectively.

$$\left[\mathbf{F}\right]\left[\mathbf{V}\right] = \left[\mathbf{V^d}\right]. \tag{1}$$

The equation for the deformation gradient matrix [**F**] is obtained by sutable multiplication of the equation (1) by the inverse of matrix [**V**].

$$\left[\mathbf{F}\right] = \left[\mathbf{V^d}\right]\left[\mathbf{V}\right]^{-1}. \tag{2}$$

FCC single crystals deform simultaneously in more than one slip system. The description of the deformation gradient matrix in the case of simultaneous slip in systems A,B…H (eq. 3) was proposed by Chin, Thurston and Nesbitt [5],

$$\mathbf{F} = \lim_{\mathbf{a}\to 0} (\mathbf{F_H}...\mathbf{F_B F_A})^{\alpha/\mathbf{a}} = \mathbf{e}^{\mathbf{F_s}}, \tag{3}$$

where α is a total shear in the system F_A, **a** is the infinitesimal shear in a unit step, and $\mathbf{F_s}$ is a certain summation matrix of contributions of all independent slip systems to the total crystal strain. The $\mathbf{F_s}$ matrix remains in the following relationship with the single-column matrix **U**, which incorporates the values of simple shears in eight independent slip systems, where matrix **M** describes the crystallography of the independent slip systems of FCC crystals.

$$[\mathbf{F_S}] = [\mathbf{M}][\mathbf{U}]. \tag{4}$$

By combining equations (2)-(4), one may receive finally the equation 5 that relates the matrix **U** with the macroscopic change of the crystallographic geometry of the sample due to deformation.

$$[\mathbf{U}] = [\mathbf{M}]^{-1}\mathbf{ln}\left|\left[\mathbf{V^d}\right]\left[\mathbf{V}\right]^{-1}\right|. \tag{5}$$

3. Experimental method

The Cu-6%at.Al single crystals of the prismatic shape and following dimensions $4 \times 4 \times 100$ mm^3 were used in the studies. The initial tensile orientation of the crystals was about 3° apart from the [$\bar{1}$ 12] direction. The single crystals were grown by modified Bridgman method in natural temperature gradient furnace in a vacuum better than 10^{-4} hPa. The samples were mechanically and chemically polished with a Mitchell reagent in order to obtain surfaces of great smoothness. Then, the grid of measurement points was placed on the lateral surfaces by microhardness tester, Instron Wolpert Testor 2100, with the value of indentation load in the range between 0,1N and 0,2N. The grid was placed in the middle of the 70mm gauge length of the sample. Taking advantage of micrometric **X-Y** table coupled with an optical microscope, the measurement of the mutual placement of the points on both lateral sides P1 and P2 was conducted (Fig. 3). Next, the crystallographic orientation of P1 and P2 planes was determined by using X-ray diffractometer, Bruker D8 Advance. The samples were deformed in the tensile test at ambient temperature and initial strain rate of 10^{-3} s^{-1}. The deformed samples were placed in specially-prepared holder that was adapted to fasten it both at the X-ray diffractometer and micrometric **X-Y** table. Sets of data describing the changes of geometric crystallography of the sample after each deformation step were obtained from the performed measurements in the form of vector **V** and $\mathbf{V^d}$ (Fig. 3). Finally, taking advantage of the equation 5 the magnitude of the shear strains in eight independent slip systems was calculated with a resolution of the method of about 5×10^{-3}.

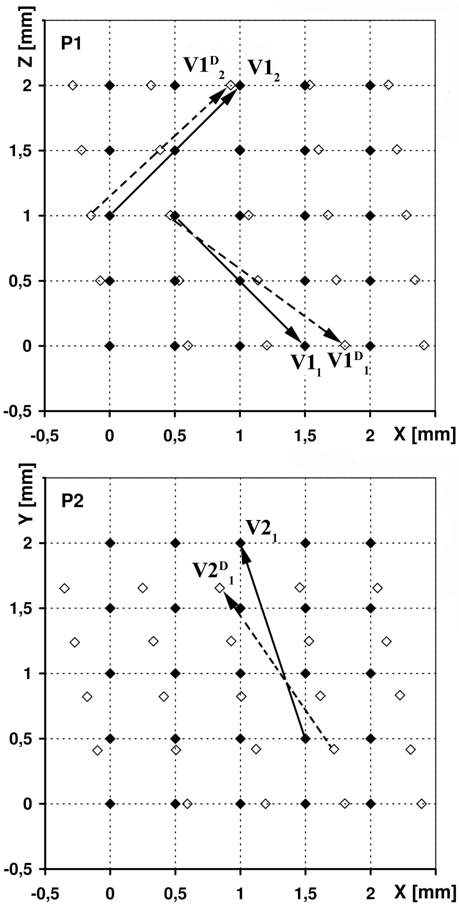


Fig. 3. The grids on the lateral surfaces P1 (a) and P2 (b) of the tested single crystal before (♦) and after (◊) the deformation. The **X** axis is parallel to the tensile direction

Also, for a comparison the characteristics of analogous tensile test without the grid of measurements points was determined as well. No substantial differences were found comparing both characteristics (with and without the grid), especially in regard the critical stresses and strains, at which the changes of the dominant deformation systems took place.

4. Results and discussion

Figure 4 shows the stress-strain ($\sigma = f(\varepsilon)$) and strain hardening rate-strain ($d\sigma/d\varepsilon = f(\varepsilon)$) curves of the studied Cu-6%at. Al single crystals. The stress-strain curve gives valuable information about critical moments of the crystal tensile deformation, which are noticed at the true strain of about 0.2 and 0.4, or alternatively, at the flow stresses of 135 MPa and 225 MPa, respectively. At the first critical deformation point (σ=135 MPa, $\varepsilon \approx 0.2$) a local deformation band appears on the sample, which propagates along the sample in form of Lüders type deformation. The change of deformation mechanism is accompanied by the rapid drop of strain hardening rate from about 950 MPa down to 400 MPa. After the period of the Lüders type deformation the macroscopically homogeneous deformation mode is restored. However, the hardening rate still maintains at the level of about 450 MPa. Similar change of the deformation mechanism of the crystal occurs at the second critical point of the process (σ=225MPa, $\varepsilon \approx 0.4$), which is accompanied by a decrease of the hardening rate from 450 MPa down to 250 MPa. It should be noticed that after the second critical point the tensile process runs close to the Considére point, and thus another change of the dominant slip system at the flow stress of about 250 MPa leads to rapid rupture of the tested sample.

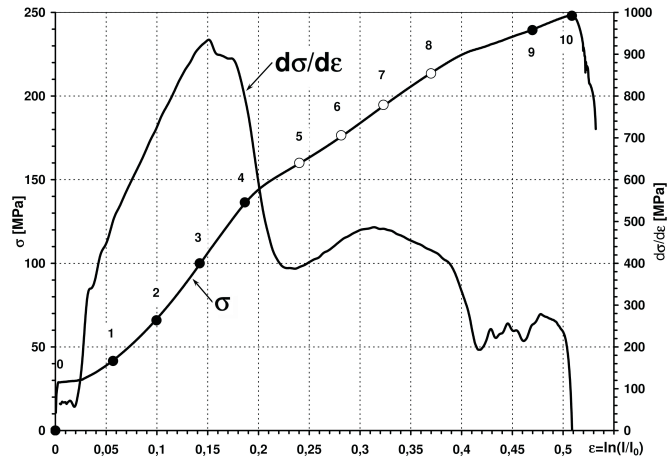


Fig. 4. The tensile stress and work hardening rate versus true elongation of the tested Cu-6at.%Al single crystals. Points marked from **1** to **10** indicate the strain intervals at which the meesurements of the crystal orientation and shape changes were taken. Points (●) and (○) indicate the strain intervals of the dominance of the BIV and CI slip system, respectively

The systematic measurements of the rotation of the crystal lattice performed at 10 different strain levels supply very important information about the changes of the deformation mechanisms. The rotation of the crystal lattice is illustrated by displacement of the tensile axis at a chosen fragment of the stereographic projection (Fig. 5). The initial orientation of the tensile direction (point **0**) is located within the standard triangle close to [$\bar{1}$12] orientation, where the most highly stressed slip system is BIV (Fig.1) or using the crystallographic description (111)[$\bar{1}$01]. Since the BIV slip system dominates the deformation process, the tensile direction moves along with the increase of the strain towards the [$\bar{1}$01] pole crossing the [001]-[$\bar{1}$11] symmetry line as soon as in the first deformation step (point **1**). At the point number **4**, *i.e.* at the critical strain of about 0.2, the overshoot reaches the maximal value of about 5 degrees, and starting from this point the reverse process occurs, the tensile axis moving backwards to the initial stereographic triangle. At the end of overshoot, the change of the dominant slip system takes place from the primary, BIV, to the conjugate one, CI, which is now the most stressed slip system. The dominance of the new slip system CI causes the tensile direction to move towards the [0 1 1] pole. After another crossing of the symmetry line (point **7-8**), the tensioned crystal is in the state of the second overshoot which ends at the strain of about 0.4 due to another change of the dominance, this time CI->BIV. The analysis of the sequence of the placement of the points **9** and **10** it can be assumed that the third change of the dominant slip system is associated with the sudden loss of the stability of the tensile test and rupture. Summarizing, the X-ray results show that the phenomenon of the change of the dominant slip system happened three times in the total range of tensile test of the studied Cu-Al single crystals.

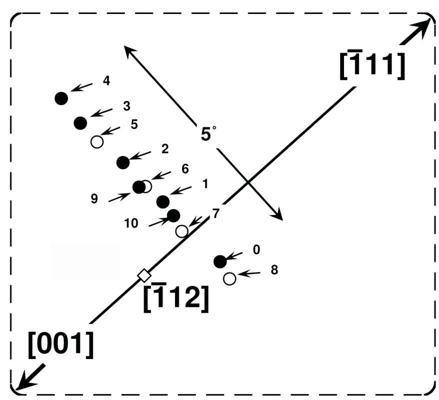


Fig. 5. The rotation of the tensile axis of the tested Cu-6at.%Al single

crystal shown on the fragment of stereographic projection containing the [001]-[-111] symmetry line

Figure 6 presents the distribution of the macroscopic deformation of the crystal in the eight independent slip systems as a function of the true tensile strain. The selected independent slip systems belong to the four {111} planes, where A, B, C and D are $(\bar{1}11)$, (111), $(\bar{1}\bar{1}1)$, $(1\bar{1}1)$ respectively. The numerical values of each slip system at each deformation step are collected in the Table. The results show unambiguously that the total deformation of the crystal is produced mainly by two slip systems, BIV and CI. The BIV slip system dominates the deformation process in the range of the true crystal elongation between 0 and 0.2 and in the range above 0.4 up to breaking of the sample. Whereas, the CI slip system dominates the deformation process in the range between 0.2 to 0.4 of the true strain. The rest of the independent slip systems operating on the A and D slip planes bring the influence in the total crystal deformation of about an order of magnitude lower before the transformation but playing much more important accommodation function afterwards.

The changeover from the BIV to CI slip system dominance is presented also in the form of resultant shear vectors operating on all four slip planes before (Fig. 7a) and after the critical point of the transformation (Fig. 7b). It should be noticed that the change of dominant shear vector from the B plane onto the plane C is accompanied by the change of a sign of the shear vectors of the accommodating slip systems operating on the A and D planes.

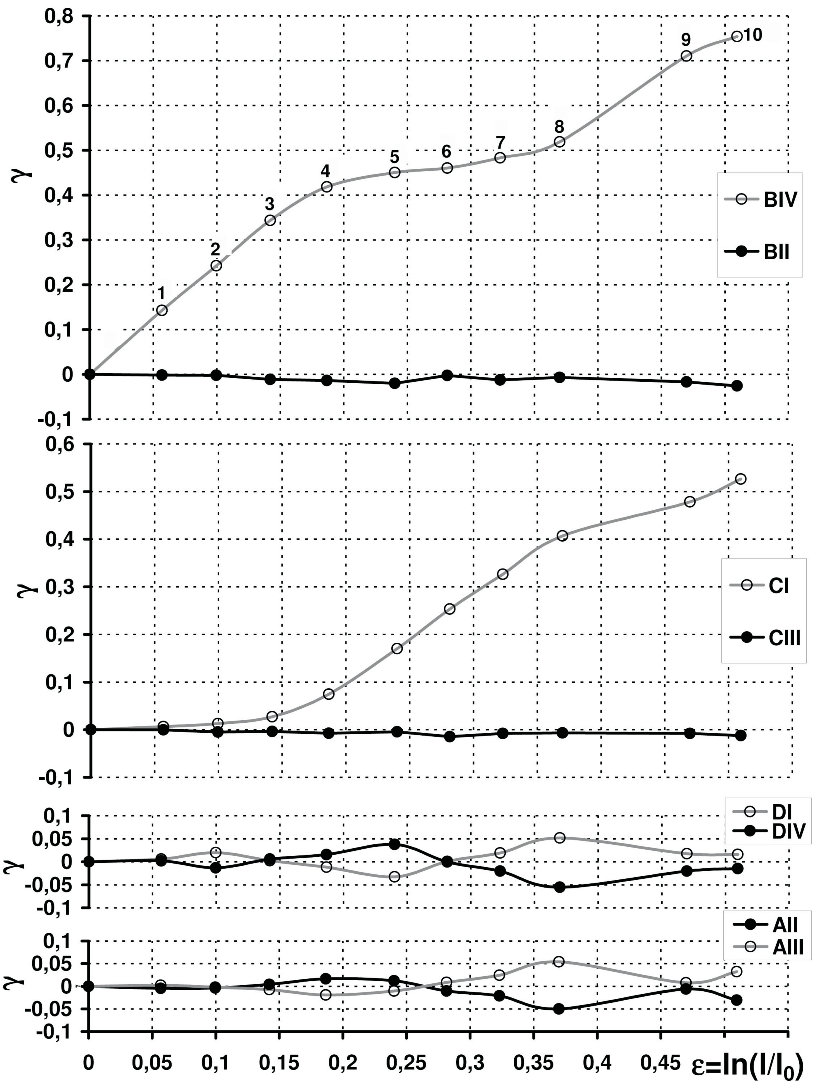
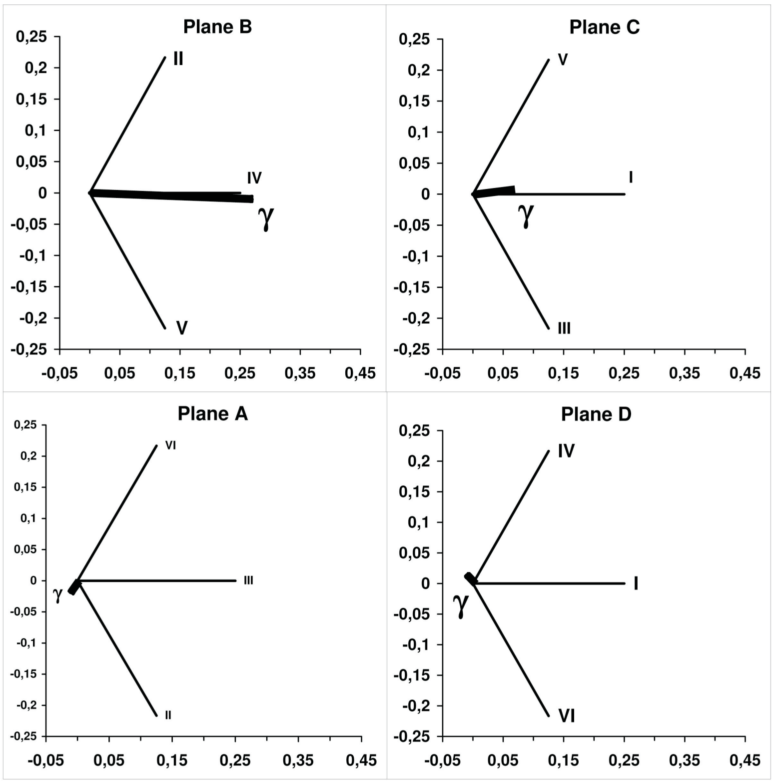


Fig. 6. The evolution of the shears in eight independent slip systems of the tested Cu-6at.%Al single crystals. The full crystallographic description of the independent slip systems see in the text and in the paper [3]

a)



b)

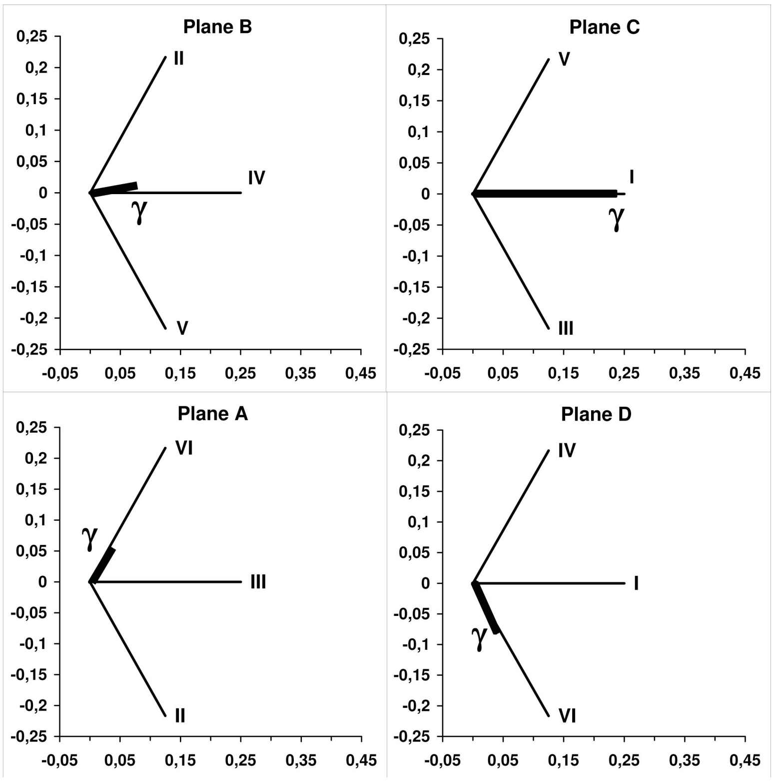


Fig. 7. The resultant shear vectors γ operating on the four **A**, **B**, **C**, **D** closed packed planes of {111} type of the tested single crystals. For the crystallographic description see comment in the text

TABLE

The numerical values of cumulative shears of the eight independent slip systems for ten strain intervals used in the experiment

step	$\varepsilon=\ln(l/l_0)$	BIV	BII	CI	CIII	AII	AIII	DI	DIV
1	**0.057**	0.143	-0.001	0.007	0.000	-0.004	0.002	0.006	0.002
2	**0.100**	0.242	-0.002	0.013	-0.005	-0.004	-0.002	0.020	-0.013
3	**0.142**	0.344	-0.011	0.027	-0.004	0.004	-0.007	0.002	0.006
4	**0.187**	0.418	-0.014	0.074	-0.007	0.017	-0.019	-0.012	0.015
5	**0.240**	0.450	-0.020	0.170	-0.005	0.012	-0.010	-0.033	0.038
6	**0.281**	0.461	-0.002	0.253	-0.015	-0.011	0.009	0.001	0.000
7	**0.323**	0.483	-0.012	0.326	-0.008	-0.021	0.024	0.019	-0.020
8	**0.370**	0.519	-0.007	0.407	-0.007	-0.050	0.054	0.052	-0.055
9	**0.470**	0.711	-0.017	0.478	-0.008	-0.006	0.008	0.017	-0.020
10	**0.510**	0.754	-0.026	0.526	-0.012	-0.031	0.032	0.016	-0.015

Taking advantage of the X-ray studies the function m(ε) of the changes of Schmid orientation factor from the true strain can be easily found for each operating slip system. On the contrary the product σ(ε) x m(ε) is equivalent to the shear stress $\tau(\varepsilon)$ operating in a given slip system. The information on the hardening rate $d\tau/d\gamma$ in the BIV and CI slip systems, which interchange their domination functions during the whole deformation process can be easily obtained connecting the relationships $\tau(\varepsilon)$ with $\gamma(\varepsilon)$ from the figure 6. Figure 8 presents a development of the shear stresses in the BIV and CI systems during the tensile process and the corresponding changes of the strain hardening rate in these systems. The obtained relationships show that the change of dominance happens when the hardening rate of the dominating system is equal to the hardening rate of the system taking over the dominance. In both observed critical points for the process, *i.e.* at true strain ε ≈0.2 and ε ≈0.4, $(d\tau/d\gamma)_{BIV} = (d\tau/d\gamma)_{CI}$. Additionally, the obtained relationships $\tau_{BIV}(\varepsilon)$ and $\tau_{CI}(\varepsilon)$ show that even in the case of the Cu-Al single crystals of the initial orientations close to the [0 0 1]-[[$\bar{1}$ 1 1] symmetry line of the stereographic triangle the end of the overshoot phenomenon is characterized by still significant value of the latent hardening τ_{CI}/τ_{BIV} ratio.

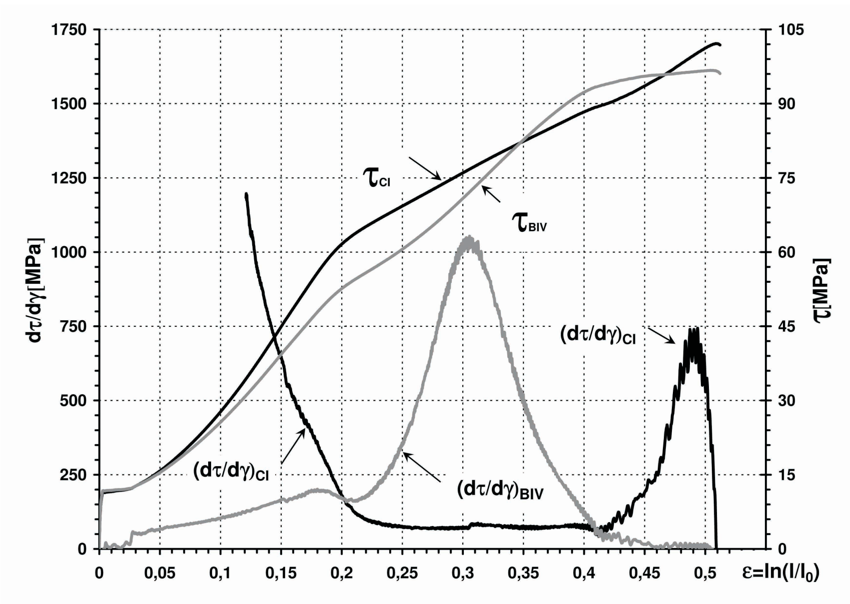


Fig. 8. The evolution of the resolved shear stresses τ operating in the BIV and CI slip systems and the coresponding rates of work hardening $\mathbf{d\tau/d\gamma}$ versus true elongation of the tested single crystals

5. Conclusions

On the basis of the results obtained on the Cu-Al single crystals in this paper and those of Cu single crystals with the various orientation of the tensile axis and presented in the paper (4) it can be concluded:

(i) The change of the dominant slip system in the tensile FCC single crystals have transformational character.

(ii) The change of the dominant slip system is accompanied by a sudden drop of the work hardening rate $d\sigma/d\varepsilon$ and occurs when the hardening rates $d\tau/d\gamma$ of the dominating system and system taking over deformation become equal.

(iii) The multiple change of the dominant slip system takes place in the FCC single crystals with the tensile axis close to the symmetry line of the stereographic triangle.

Acknowledgements

The financial support of the Ministry of Higher Education and Science of Poland, grant nr 3T08A 080 29, and of the AGH University of Science and Technology, grant nr 10.10.180.405 and 11.11.180.255 is acknowledged.

REFERENCES

[1] G. I. T a y l o r, C. F. E l a m, The Distortion of an Aluminium Crystal during a Tensile Test, Proc. Roy. Soc. (London), Ser. A, **102**, 643 (1923).

[2] S. J. B a s i n s k i, Z. S. B a s i n s k i, Plastic Deformation and Work Hardening, Dislocation in Solids, ed. F.R.N. Nabarro, North – Holland Publishing Company, 1979.

[3] Z. S. B a s i n s k i, M. S. S z c z e r b a, J. D. E m b u r y, Tensile instability in face-centered cubic materials, Philosophical Magazine A, **76**, 743 (1997).

[4] Z. S. B a s i n s k i, S. J. B a s i n s k i, Quantitative determination of secondary slip in copper single crystals deformed in tension, Philosophical Magazine **84**, 213 (2004).

[5] G. Y. C h i n, R. N. T h u r s t o n, E. A. N e s b i t t, Finite Plastic Deformation Due to Crystallographic Slip, Trans metall. Soc. AIME, **236**, 69 (1966).

Received: 20 December 2008.