JOURNAL OF MATERIALS SCIENCE 18 (1983) 2613–2617

Cross-slip on the first order pyramidal plane $(10\bar{1}1)$ of a-type dislocations $[1\bar{2}10]$ in the plastic deformation of α-titanium single crystals

S. NAKA, A. LASALMONIE
Office National d'Etudes et de Recherches Aérospatiales, 92320 Chatillon-sous-Bagneux, France

The nature of the cross-slip plane was determined by electron microscopy observations of α-titanium single crystal specimens, oriented for single prismatic slip $(10\bar{1}0)$ $[1\bar{2}10]$. The occurrence of cross slip on the first order pyramidal plane $(10\bar{1}1)$ was proved for a-type dislocations $[1\bar{2}10]$. Furthermore, two types of dislocation configuration due to the double cross-slip were observed: edge dislocation dipoles and loops elongated along the Burger's vector.

1. Introduction
It has been shown in a recent paper [1] that the critical resolved shear stress (CRSS) for prismatic slip in α-titanium single crystals exhibits an anomaly between 300 and 500 K. The examination of slip lines revealed the appearance of cross-slip events at the very beginning of this anomaly (Fig. 1). This was confirmed by electron microscopy observations. Furthermore the existence of numerous edge dipoles was thought to be a consequence of double cross-slip.

The aim of this paper is to determine the nature of the cross-slip plane and to show the presence of the double cross-slip events.

2. Experimental procedure
Single crystal specimens, oriented for single prismatic slip (Schmid factor $\simeq 0.5$), were deformed in compression between 77 and 700 K. Three thin foil planes were chosen for observation in transmission electron microscopy: parallel to the slip plane, normal to the slip plane but containing the Burger's vector and normal to the Burger's vector.

In situ deformation experiments were also performed using a heating tensile stage [2]. But in this study, only polycrystalline specimens were deformed in the microscope.

3. Results
3.1. Determination of the cross-slip plane
The cross-slip became generalized when the deformation temperature rose beyond 300 K. In a specimen deformed at 684 K, cross-slip was frequently observed and one such example is shown in Fig. 2. The foil plane in this case is nearly normal to the prismatic slip plane and it contains the Burger's

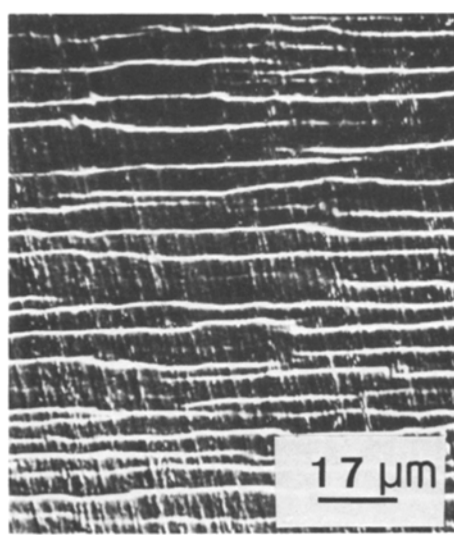


Figure 1 Cross-slip evidenced by wavy slip lines in the specimen deformed at 478 K.

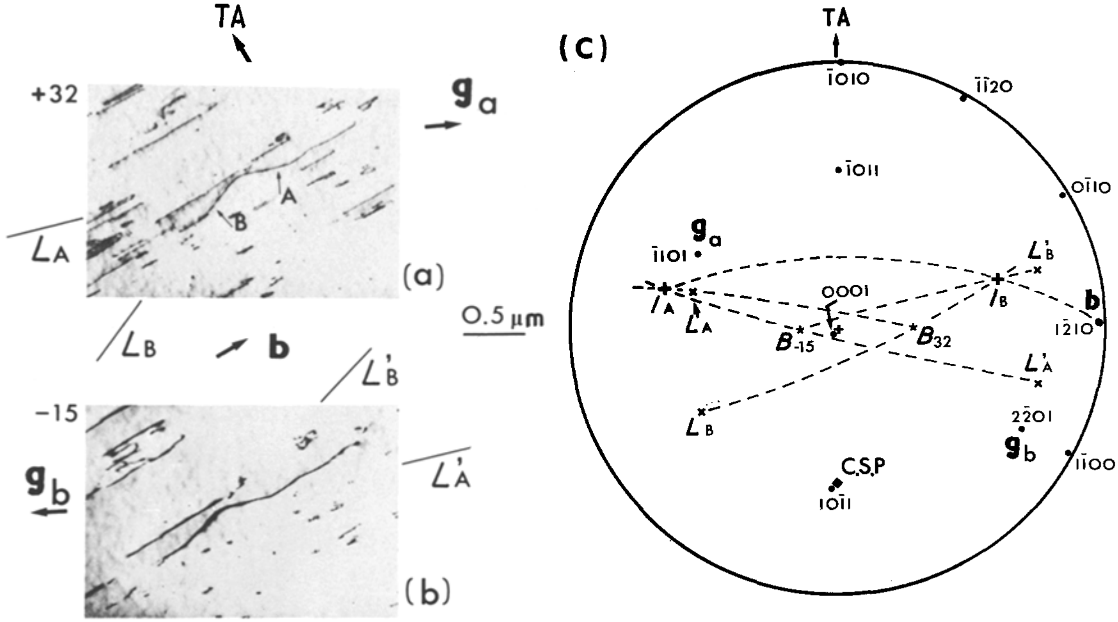
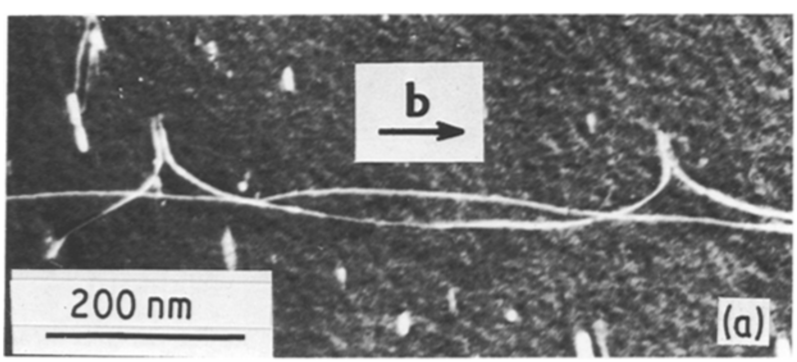


Figure 2 Determination of the cross-slip plane. The primary slip plane is normal to the picture. A and B are two deviated segments. (a) the image at a tilt angle of + 32°, (b) the image at a tilt angle of — 15° and (c) stereoscopic projection giving the crystallographic orientations I_A and I_B of Segments A and B.

vector, **b**; the tilt axis (TA) is normal to **b**. Irrespective of the tilt angle, the prismatic plane is always perpendicular to the screen. If there is no cross-slip, all the dislocations should then appear as straight lines perpendicular to TA. The dislocation segments, A and B, (Fig. 2) have thus necessarily deviated from the primary prismatic slip plane. The crystallographic orientations, I_A and I_B of segments A and B were found by using the following technique. The projections of I_A and I_B on the screen were determined for different tilting conditions.

In Fig. 2a, the tilt angle is + 32° giving the projections L_A and L_B, in Fig. 2b, the tilt angle is



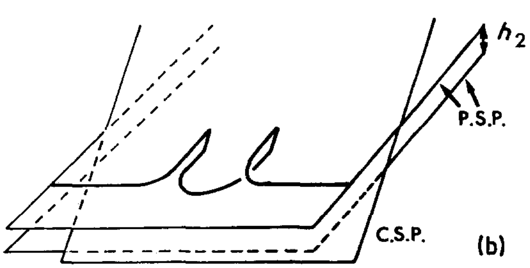


Figure 3 Formation of edge dipoles due to a double cross-slip.

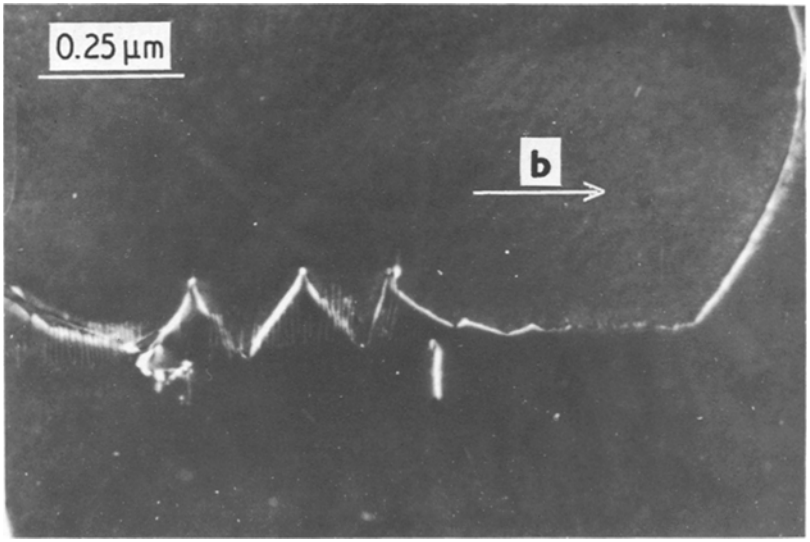


Figure 4 Serrated image of a screw dislocation due to a high density of edge dipoles.

$-15°$ and the projections are L'_A and L'_B. The beam orientations are, respectively, B_{32} and B_{-15}. I_A and I_B can be graphically determined as shown in Fig. 2c. It can be seen in Fig. 2c that I_A and I_B are in the $(10\bar{1}1)$ plane which is actually a secondary slip plane for the dislocation ($\mathbf{b} = 1/3[1\bar{2}10]$).

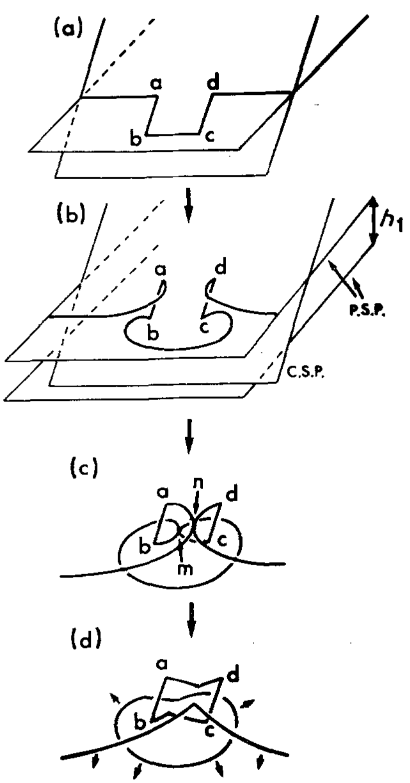


Figure 5 Formation of loops elongated along the Burger's vector by the Orowan mechanism following a double cross-slip.

3.2. Double cross-slip and formation of edge dipoles

Fig. 3 gives an example of a screw dislocation segment which has undergone a double cross-slip. The thin foil is here parallel to the primary $(10\bar{1}0)$ slip plane. The details of the dislocation shape are schematically shown in Fig. 3b: two edge dipoles are formed which lower the line tension energy if the cross-slip distance, h, is small. The presence of such dipoles was particularly evident in the specimens deformed in the temperature range 300 to 500 K, where the CRSS against temperature curve exhibits an anomaly [1]. In some cases, the edge dipoles were so close together that the image of the screw dislocation segments became very serrated (Fig. 4).

When the cross-slip distance, h, is large, the dislocation can escape by the Orowan mechanism as depicted in Fig. 5. A loop, abcd, is created which has the two segments ab and cd in the cross-slip plane and the other two segments ad and bc in the primary slip plane. This loop is not necessarily planar. Such a loop was indeed observed (see Fig. 6) in the previously observed foil of Fig. 2. The prismatic slip plane here is nearly normal to the screen; the Burger's vector is parallel to the trace of the prismatic plane; the TA is approximately parallel to the Burger's vector. When the foil is horizontal (Fig. 6a), it can be seen clearly that the loop does not lie in the prismatic plane.

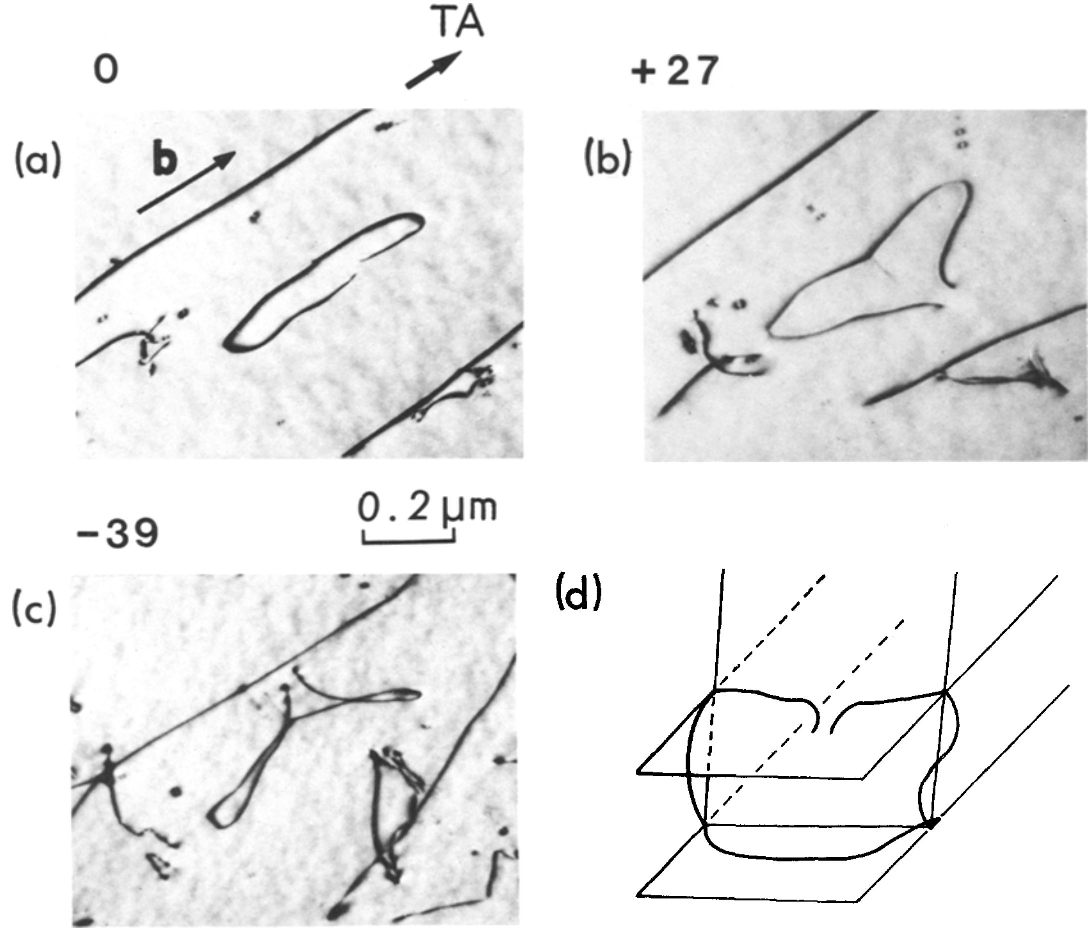


Figure 6 Identification of a loop similar to abcd of Fig. 5.

Images at two different tilt angles (Figs. 6b and c) evidence the complicated shape of the loop. However, by observing the change in the image of the loop at different tilt angles, it could be proved that the loop lay nearly in the first order pyramidal plane although it was not completely planar. Indeed, the sense of inclination of the plane containing the loop changed around a tilt angle of $-28°$ and the angle formed between the prismatic plane and the first order pyramidal plane which share the same Burger's vector is of $28.62°$. Stereo pairs enabled us to work out the configuration of the loop (Fig. 6d) which can be perfectly explained by the mechanism shown on Fig. 5.

3.3. *In situ* deformation tests

In this study, some preliminary tests were performed at 570 K on polycrystalline specimens; these tests were intended only to detect the occurrence of cross-slip. More detailed experiments will be performed later on single crystals.

During the *in situ* deformation, the main slip plane was always a prismatic one. In some areas, the traces of the moving dislocations at the foil surfaces were very wavy (see Fig. 7), thereby confirming the occurrence of cross-slip. In the bulk specimens, slip and cross-slip may occur in only three types of planes for **a**-type dislocations: the prismatic plane $(10\bar{1}0)$, the basal plane (0001) and the first order pyramidal plane $(10\bar{1}1)$. However in the thin foil, the dislocations were able to glide in a larger number of planes (pencil glide) presumably due to a thin foil effect.

4. Discussion and conclusion

This study clearly proves the occurrence of cross-slip on the first order pyramidal plane $(10\bar{1}1)$. It is not surprising, since this plane is recognized as one of the possible secondary slip planes for **a**-type dislocations in α-titanium [3]. However, the prismatic slip being the easiest slip mode, it is difficult to understand why cross-slip phenomena occur in

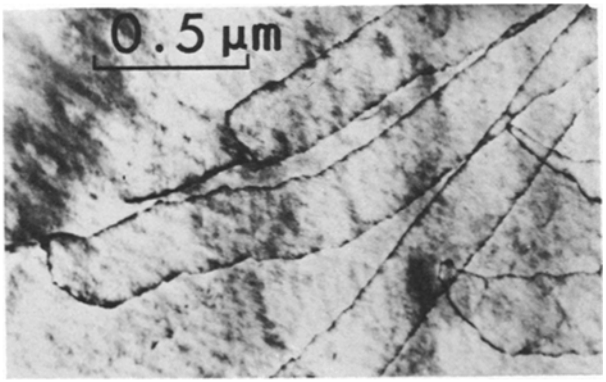


Figure 7 Wavy traces in a foil deformed at 570 K in the microscope.

the first order pyramidal plane in the single crystals oriented for single prismatic slip. Until now, few investigations have been made for the first order pyramidal slip mode, in particular in single crystals [4]. Undoubtedly, the core structure plays a prominent role in the occurrence of cross-slip phenomena. Šob *et al.* [5] proposed a core structure model in which the **a**-type dislocations split on two planes simultaneously (prismatic and first order pyramidal planes) in order to explain the strengthening of α-titanium by interstitial solutes. More recently, Bacon and Martin [6, 7] performed an atomic core simulation of the **a**-type dislocations in hexagonal metals, and de Crecy *et al.* [8] attempted an observation of core structure for **a**-type dislocations in α-titanium by high resolution electron microscopy.

These theoretical and experimental attempts have not provided any definitive answer until now for the understanding of α-titanium and all other hexagonal metals. Moreover, in the case of cross-slip in this study, it remains to be understood if this phenomenon is characteristic for pure α-titanium or is promoted by the presence of interstitial solutes.

Acknowledgements
The authors are very grateful to Dr L. P. Kubin for helping us to perform the *in situ* deformation tests in his laboratory and for useful discussions with him.

References
1. S. NAKA and A. LASALMONIE, *Mater. Sci. Eng.* **56** (1982) 19.
2. L. P. KUBIN and P. VEYSSIERE, in "Proceedings of the 10th International Congress on Electron Microscopy", Hambourg, August 1982, Vol. 2 (Deutsche Gesellschaft für Elektronenmikroskopie, 1982) p. 529.
3. J. P. POIRIER and R. Le HAZIF, *J. Microsc. Spectrosc. Electr.* **1** (1976) 595.
4. H. CONRAD, *Prog. Mater. Sci.* **26** (1981) 123.
5. M. ŠOB, J. KRATOCHVÍL and F. KROUPA, *Czech. J. Phys.* **B25** (1975) 872.
6. D. J. BACON and J. W. MARTIN, *Phil. Mag.* **A43** (1981) 883.
7. *Idem, ibid.*, **A43** (1981) 901.
8. A. de CRECY, A. BOURRET, S. NAKA and A. LASALMONIE, *Phil. Mag.* **A47** (1983) 245.

Received 26 November 1982
and accepted 6 January 1983