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## Plastic deformation of a Ti - 6% Al - 4% V alloy with a strong transverse-type crystallographic $\alpha$ -texture at elevated temperatures

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† Professor Peter Klimanek died on 22 Dec. 2010 in Dresde, Germany. He was 75.

JSL : “he was a good man, a good Professor, someone who was very human. He will live on in my memory forever”.

### Abstract

The deformation behaviour of  $(\alpha+\beta)$ -Ti with a very sharp texture is influenced by temperature, strain rate and the sense of stress. In this paper the plastic deformation of  $(\alpha+\beta)$ -Ti is investigated in a temperature range between 295 K and 1073 K in compression and tension along different directions of the sheet plane ( $0^\circ$ ,  $45^\circ$  and  $90^\circ$  from the Rolling Direction). For this temperature range two phenomena are observed : a dynamic strain ageing and a asymmetry of the yield stress associated with the movement of  $\langle \mathbf{c+a} \rangle$  dislocations. Electron microscopy of specimens strained in uniaxial tension or compression along [0001] at 570 K shows  $\langle \mathbf{c+a} \rangle$  dislocations moving on  $\{1\bar{1}01\}$  and  $\{11\bar{2}2\}$  planes and an anomalous activity of  $\langle \mathbf{c+a} \rangle$  dislocations is observed on the whole range of orientation. We discuss the effect of interstitial solutes on the activity of the  $\langle \mathbf{c+a} \rangle$  dislocations between 500 K and 750 K and the effect of the non-centrosymetry for the  $\{1\bar{1}01\}$  planes on the directional asymmetry of the yield stress associated with the  $\langle \mathbf{c+a} \rangle$  movement.

*Keywords* : Ti-6%Al-4%V alloy, Texture, deformation systems, high temperature.

## **1. Introduction**

The Ti - 6%Al - 4%V (  $\alpha + \beta$  ) alloy is the most popular and the best known of all the Ti alloys. Depending on different thermomechanical treatments, it is possible to obtain this alloy with a large variety of microstructures and crystallographic orientations [1 - 6]. In particular, flat products made by hot rolling have always relatively strong textures [7]. The mechanical properties for commercial Ti (  $\alpha + \beta$  ) alloys are strongly anisotropic because of the sharp texture associated with the anisotropy of  $\alpha$  phase monocrystal of.

Now, the need is felt for a clear understanding of the mechanisms of texture and microstructure evolution during the processing of h.c.p. metals and their alloys. A good knowledge of the deformation mechanisms (slip and twinning) and their respective critical resolved shear stresses (CRSS) is particularly important to set coherently the parameters used in the plasticity models in order to calculate the texture evolution during rolling and the plastic anisotropy for subsequent forming [8-11]. In the present work, a Ti - 6%Al - 4%V (  $\alpha + \beta$  ) sheet with a strong « edge-type » texture was tested under different stress-strain states. Deformation mechanisms and the local texture were investigated in order to understand, explain and predict the mechanical properties.

## **2. Experimental procedure**

### *2.1. Material*

The material used is a Ti - 6% Al - 4% V 5 mm thick sheets of made by Cezus (France).

**Table 1** : *Chemical analysis of the Ti-6%Al-4%V alloy (wt.%)*

Element	Ti	Al	V	Fe	Si	O	C	N	H
Specification	Balance	5.5-6.5	3.5-4.5	<0.4	<0.04	<0.2	<0.01	<0.05	<0.015

The material was in the annealed state (1000 K for 1 h) after cold rolling. Previously it had been hot rolled in the  $\alpha+\beta$  field. The chemical composition is given in Table 1. The alloy has a nominal O<sub>2</sub> content of 1300 ppm (weight) and an equivalent O<sub>2</sub> content of 0.35 at. % calculated by Conrad's equation  $O_{eq} = O + 2N + 0.75 C$  [12]. The texture analysis showed that the sheet presents a strong transverse texture with a majority of grains having their  $\langle c \rangle$  axes in transverse direction. Tests were conducted on the differently shaped tensile and compressive test pieces shown in Fig. 1.

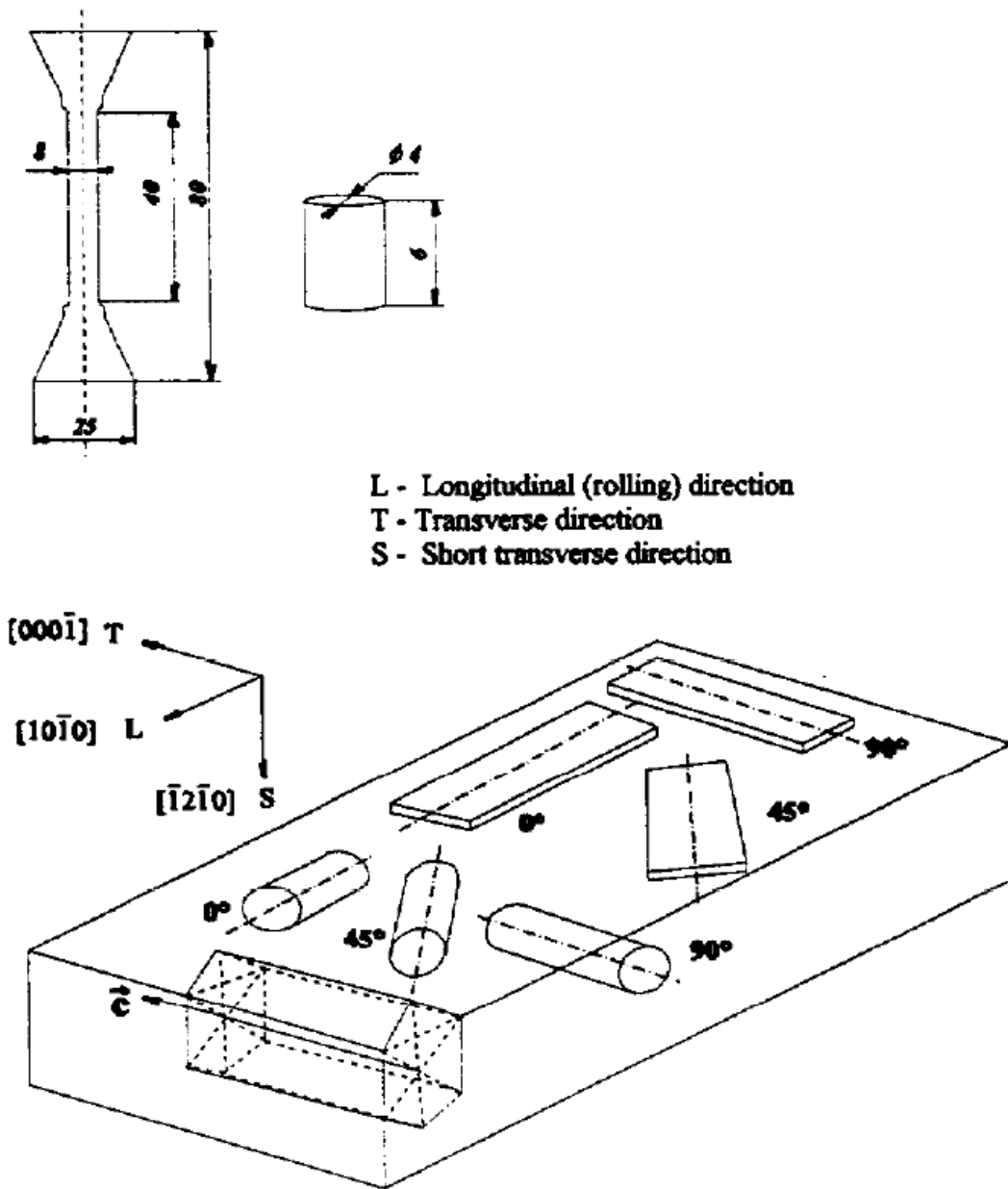


Figure 1 The orientations of the tensile and compressive test pieces in relation to crystallographic orientations in the Ti-6% Al-4% V.

## 2.2 Microstructural investigations

Metallographic sections corresponding to the rolling, longitudinal and transversal planes of the sheet (RP, LP and TP) were examined by optical microscopy. A conventional metallographic preparation according to the methods of Boyer [13] for Ti alloys was carried out.

Transmission electron microscopy (TEM) techniques were also used to analyse the starting material substructure and to determine the deformation systems. Foil preparation followed the

recommendations of Zaefferer and Schwarzer[14]. The specimens were examined at 300 KV with a Philips EM-430 electron microscope.

### *2.3. Mechanical testing*

Tensile tests were conducted between 295 K and 1073 K using a Zwick machine at a cross-head rate of  $1 \times 10^{-3} \text{ s}^{-1}$ . Load-elongation curves (or load-cross-head displacement curves for plane strain test pieces) were obtained on an XY plotting table which was linked to the load cell of the machine. Compression tests on small cylinders (4 mm diameter and 6 mm high) were performed on a MTS machine at strain rates  $\dot{\epsilon} = 1 \times 10^{-3} \text{ s}^{-1}$ . Tensile and compressive tests were systematically repeated at least twice for every direction.

## **3. Results**

### *3.1. Metallographic and crystallographic characterization*

Metallographic observations revealed an inhomogeneous two-phase structure. The equiaxed  $\alpha$ -grains have a mean size of about 15  $\mu\text{m}$ . They are decorated by a dispersion of small  $\beta$ -phase grains (Fig. 2). TEM observations showed a heterogeneous  $\alpha$  structure only partially recovered and recrystallized. The mean size of the mostly equiaxed  $\beta$  particles is below 1  $\mu\text{m}$  and the volume fraction is of the order of 10% at 295 K but increases during the test with temperature (~30% at 1073 K). No twinned areas were observed in the whole temperature range.

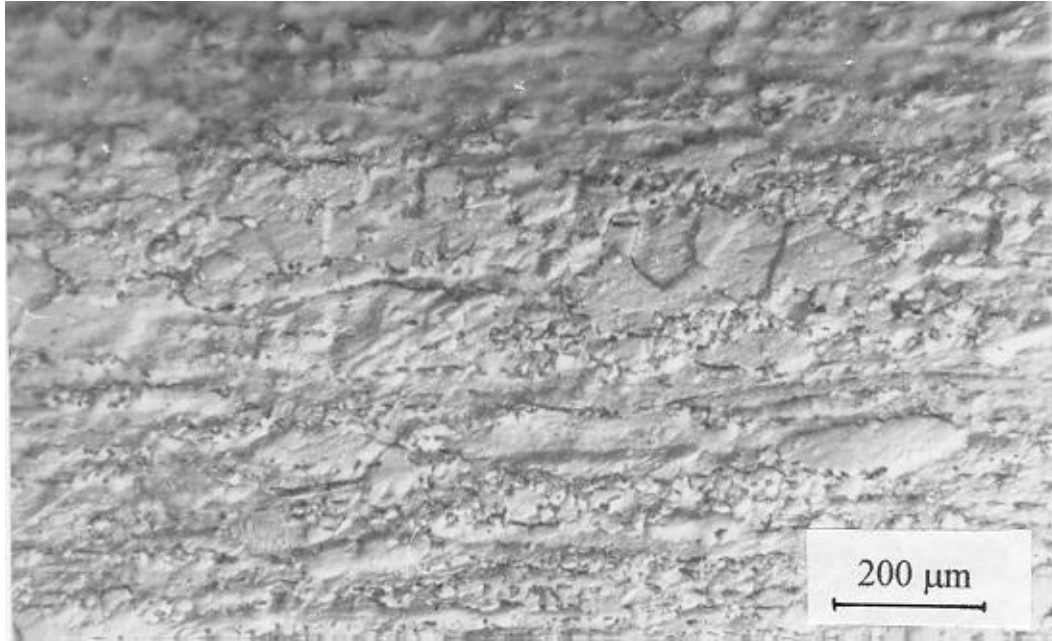


Figure 2 Optical micrographs of the 5 mm thick Ti-6% Al-4% V sheet (as received).

### 3.2. Deformation systems

The investigation of deformation systems was performed with a computer program which makes the on-line determination of Burgers vectors, dislocation line directions, glide planes and twinning systems easier [19]. The determination of Burgers vectors and line directions, however, is limited to a deformation degree less than 10 %, since individual dislocations must be recognized after successive tilts of the sample about large angles.

#### 3.2.1. Specimens strained at 295 K.

The change in the main slip system is normal and due to the strong  $\langle c \rangle$  transverse texture. Zaefferer [19] showed a correlation between the grain orientation and the degree of local deformation, evaluated through the dislocation density: The closer the crystallites are oriented to  $(11\bar{2}0)[10\bar{1}0]$  the more they are deformed, i.e. the higher their dislocation density is. Results are listed in Table 2. The dislocations remaining in the foil belong to the  $\{10\bar{1}0\}$  slip systems and were shown to be of  $\langle a \rangle$  type. They are mainly of screw type, suggesting that here the edge dislocations are more mobile.

**Table 2 :** Deformation systems on the Ti-6%Al-4%V deformed at  $\varepsilon = 0.05$

		300 K			500 K		
		0°	45°*	90°	0°	45°	90°
Main slip system		P<a>	B<a>	$\Pi_1$ <c+a>	$\Pi_1$ <c+a>	$\Pi_1$ <c+a>	$\Pi_1$ <c+a>
Secondary slip system (in order of % of occurrence)	1	$\Pi_1$ <c+a>	$\Pi_1$ <c+a>	B<a>	B<a>	B<a>	$\Pi_2$ <c+a>
	2	$\Pi_1$ <a>	$\Pi_1$ <a>	$\Pi_1$ <a>	$\Pi_2$ <c+a>		
	3	B<a>		P<a>			

$P<a> = (10\bar{1}0)\langle 11\bar{2}0 \rangle$  ;  $B<a> = (0001)\langle 11\bar{2}0 \rangle$  ;  $\Pi_1<a> = (10\bar{1}1)\langle 11\bar{2}0 \rangle$  ;  $\Pi_1<c+a> = (10\bar{1}1)\langle 11\bar{2}3 \rangle$

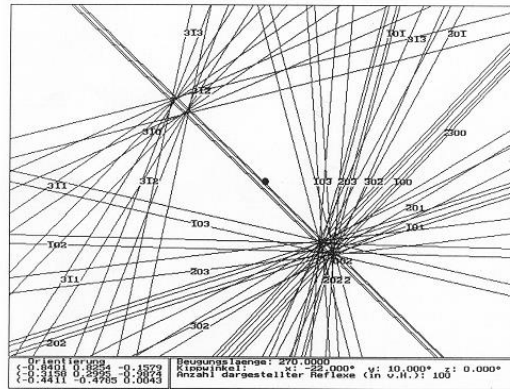
;  $\Pi_2<c+a> = (11\bar{2}2)\langle 11\bar{2}3 \rangle$

\* caution : P<a> has been observed, although it is generally mentioned in the literature.

### 3.2.2. Specimens strained at 500 K

For all orientations a high activity of <c+a> dislocations was found. Figure 3 shows the dislocation substructure in an  $\alpha$ -grain on the  $\{11\bar{2}2\}$  planes. The latter plane is unusual in Ti alloys because twinning is commonly activated in this plane.



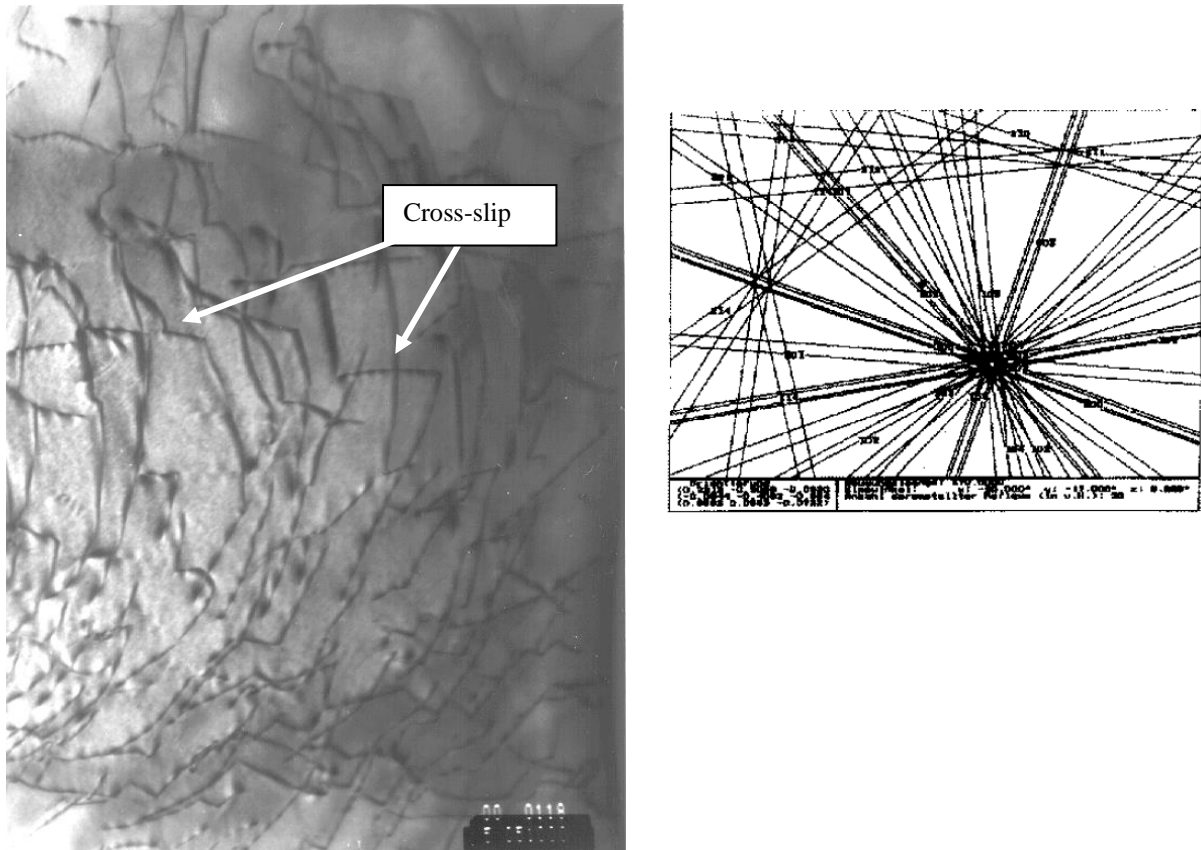


**Figure 3**  $\langle c+a \rangle$  dislocations on the  $(2\bar{1}\bar{1}2)$  and  $(10\bar{1}\bar{1})$  planes in a specimen compressed along  $[0001]$  (foil normal). Beam direction  $g = \langle 0002 \rangle$ .

### 3.2.3. Specimens strained at 1073 K

Fig. 4 shows (a) dislocations gliding on prismatic planes. There is a lot of cross-slip. It is quite difficult to find all slip systems because the CRSSs are narrow.

Dislocation pile-ups are occasionally observed. Large grains have low dislocation densities, whereas small grains present higher dislocation densities so that their Kikuchi patterns are not sharp.



**Figure 4** <a> dislocations on prismatic planes using [0001] zone,  $g = (1011)$ .

### 3.3. Mechanical properties

Fig. 5 shows flow stresses of the Ti -6%Al-4%V, which decrease normally with temperature for the specimen oriented at  $0^\circ$ ,  $45^\circ$ , and  $90^\circ$ . For the two first orientations an anomalous behaviour is noticed between 500 K and 750 K. In this range, there is a temperature dependence of the flow stress. Although it has been established that interstitial solutes are responsible for this behaviour, the exact details of the process are not clear yet. For the specimens tested at  $90^\circ$  to Rolling Direction (RD), the decrease of the flow stresses with the temperature is more continuous and smooth. A possible explanation is given in discussions (§ 4). In the same way the differences between the yield stress in tension and compression are larger for the specimens tested at  $90^\circ$  to RD.

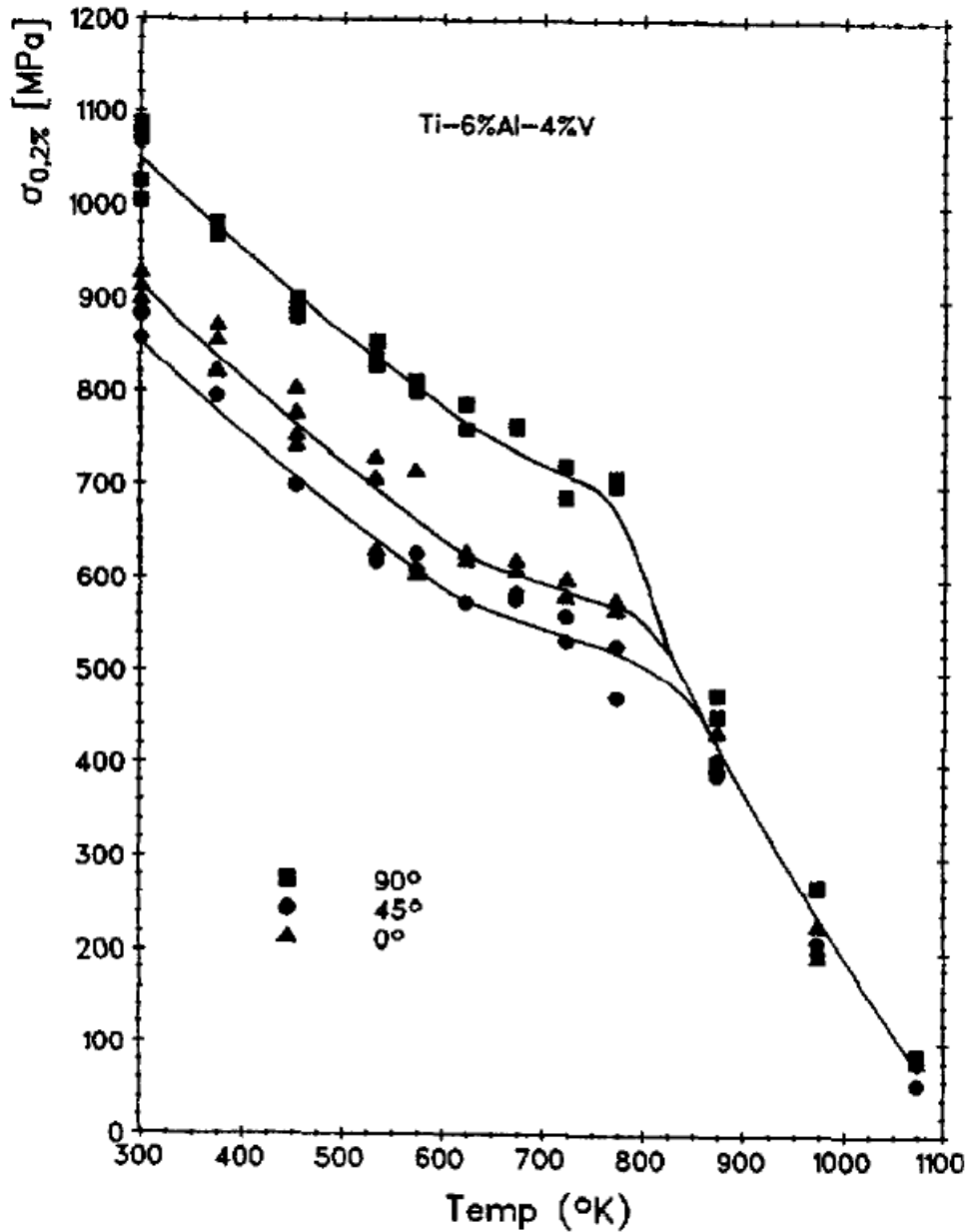


Figure 5 The variation of the yield stresses with temperature for tensile specimens.

#### 4. Discussion

At room temperature, most dislocations were of  $\langle a \rangle$  type on the prismatic planes. Secondary slip system were basal or on pyramidal  $\{1 \bar{1} 01\}$  planes.

In the range 500 K to 750 K, some of these so called « secondary slip systems » were found to be the most active for dislocation motions :  $\langle c+a \rangle$  dislocation gliding on the pyramidal planes ( $\{1 \bar{1} 01\}$  and  $\{11\bar{2}2\}$ ) was prominent.

At higher temperatures the prismatic slip is the common glide but cross-slip is very frequent. All types of slip are found.

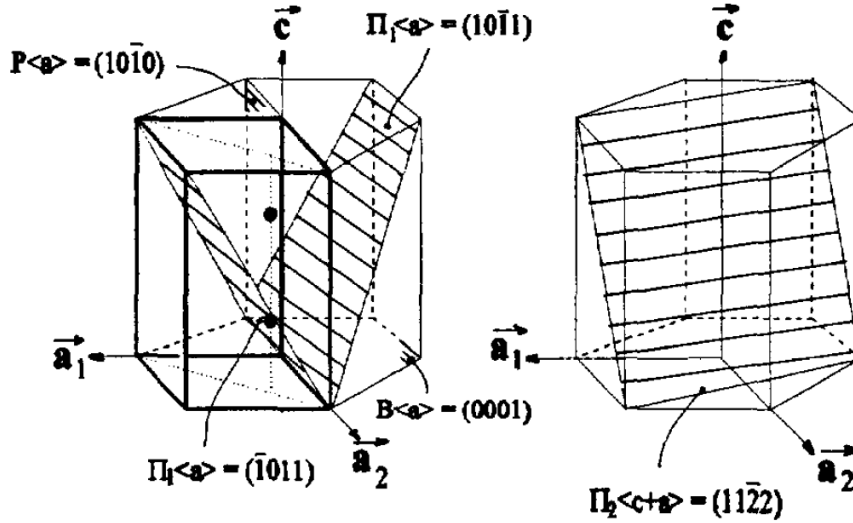


Figure 6 Possible slip planes of  $\langle a \rangle$  and  $\langle c + a \rangle$  dislocations in  $\alpha$ -Ti. The possible interstitial sites within the unit cell formed by  $a_1$ ,  $a_2$ , and  $c$  are indicated by  $\bullet$ .

Welsch and Bunk [20] investigated the influence of the oxygen concentration on the mechanical properties. They found that the room temperature deformation of a Ti-6%Al-4%V alloy containing 0.22 wt% oxygen produced a predominant planar pyramidal slip only after that the samples had been aged at 500 K for an extended period of time. They use a Churchman model [21] to explain this. The latter study a model on the effect of relatively small concentrations of oxygen and nitrogen ( $\sim 0.1$  wt%) on the slip systems in a  $\alpha$ -Ti. has been presented. The model predicts that oxygen (or nitrogen) atoms occupying octahedral positions interfere with  $\langle \bar{1}2\bar{1}0 \rangle$  type slip more severely on the basal (0001) and the prismatic  $\{10\bar{1}0\}$  slip planes than on the  $\{10\bar{1}1\}$  pyramidal slip planes. Figure 6 illustrates that the interstitial sites are never coplanar neither with the basal plane nor with the prismatic slip planes. However, they can be coplanar with one of the two possible pyramidal slip planes. At these temperatures, the mean atomic jump time  $t_j$  for interstitial solutes in titanium is given by

$$t_j \approx \frac{b^2}{D} = \frac{b^2}{D_0} \exp\left(-\frac{Q_D}{kT_s}\right)$$

where  $D$  is the diffusion coefficient,  $Q_D$  the activation energy (cal/mole) and  $T_s$  is the deformation temperature. Taking (from Conrad [12]):  $D_0 = 10^{-1} \text{cm}^2 \text{s}^{-1}$ ,  $Q_D = 26 T_m$  (cal/mole) for oxygen in Ti, (where  $T_m$  is the melting temperature in K), a temperature  $T_s = 500 \text{ K}$  is obtained for the mean jump time being of one second (1s.). This is the temperature at which the migration of these solutes first occurs at a significant rate. In the vicinity of this temperature, the respective interstitial solutes will be sufficiently mobile so that they can no longer be considered as stationary obstacles, and plastic deformation will reflect the interaction between mobile dislocations and mobile interstitial solute atoms, resulting in a behaviour known as dynamic strain ageing. Our results in microscopy confirm this possible explanation because in this range of temperature the most common system is the pyramidal  $\langle \mathbf{c+a} \rangle$  indicating that the other  $\langle \mathbf{a} \rangle$  glides have higher CRSSs. Differences between the yield stress in compressive and in tensile tests for the  $90^\circ$  oriented specimens are due to the non-centrosymmetry of the  $\{10\bar{1}1\}$  planes (see Figure 6) as explained by [15,22]. This effect is always found when the  $\{10\bar{1}1\}\langle 11\bar{2}3 \rangle$  is the most active slip.

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