Review

Research Progress on Slip Behavior of \(\alpha\)-Ti under Quasi-Static Loading: A Review

Runqi Zhang \(^1\), Qinyang Zhao \(^2\,*\), Yongqing Zhao \(^1,3\,*\), Dizi Guo \(^3\) and Yu Du \(^3\)

\(^1\) School of Materials Science and Engineering, Northeastern University, Shenyang 110819, China
\(^2\) School of Material Science and Engineering, Chang'an University, Xi'an 710064, China
\(^3\) Northwest Institute for Nonferrous Metal Research, Xi'an 710016, China
* Correspondence: zqy@chd.edu.cn (Q.Z.); trc@c-nin.com (Y.Z.)

Abstract: This paper reviews the dislocation slip behavior of \(\alpha\) phase in \(\alpha\), near \(\alpha\) and \(\alpha + \beta\) titanium alloys dominated by \(\alpha\)-Ti deformation under quasi-static loading. The relation of slip activity, slip transfer, slip blocking, twinning and crack initiation is discussed, mainly combined with in situ tensile technology. The slip behavior in Ti-alloys is analyzed in detail from the aspects of critical resolved shear stress (CRSS), grain orientation distribution and geometric compatibility factor \(m'\). In addition, slip blocking is an important factor of the formation of twins and micro-cracks. The interaction of slip behavior and interfaces is clarified systematically. Finally, the insufficiency of current research, future research directions and key difficulties of study are also discussed.

Keywords: slip activity; slip transfer; slip blocking; crack initiation

1. Introduction

Nowadays, titanium (Ti) alloys are widely used in aerospace, biomaterials and other fields \([1–3]\). There are benefits from their high specific strength, excellent biocompatibility and controllable microstructures produced by different fabrication technologies \([4–6]\). For instance, the proportion of Ti adopted in the Boeing 787 has risen to 15% at present \([7]\). These reasons above attract extensive interest in terms of investigating its deformation mechanisms \([8,9]\). However, most research is not particularly in-depth, especially on the quantitative analysis of the influencing factors and the degree of coordinated deformation in polycrystalline materials. As a result, it is difficult to improve the poor strength–plasticity matching of some Ti-alloys. With the development of industry, especially in the aerospace field, the demand and application of Ti-alloys with excellent comprehensive mechanical properties have become extremely urgent \([10]\). Studies on improving the comprehensive mechanical properties of Ti-alloy materials based on the deformation and fracture mechanisms are still extremely important.

The mechanical properties, deformation and fracture of Ti-alloys are affected by alloying elements, microstructure, deformation temperature, deformation mode, texture, etc. \([11–13]\). Essentially, the influences of these factors are attributed to the blocking and transfer capacity of the dislocation slip \([14–17]\). Alloying elements Al, Zr, Sn, Mo, V, Cr, Nb, Fe, B, Si and interstitial elements O, N and H are usually added into the Ti matrix to obtain better mechanical properties through solid solution strengthening \([18–21]\). Further, \(\alpha\)-stable elements and neutral elements preserve Ti-alloys as a hexagonal close-packed (HCP) structure and improve the strength of the alloys. \(\beta\)-stable elements can stabilize the body-centered cubic (BCC) structure to room temperature, forming \(\alpha + \beta\) and \(\beta\) Ti-alloys \([17]\). Different morphologies (equiaxed \(\alpha\), lamella \(\alpha\) and \(\alpha\) colony, etc.) of the \(\alpha\)-phase and \(\beta\)-phase (\(\beta\) lath, etc.) are formed by various processing methods, and their interaction also affects the deformation mechanism and mechanical properties \([22]\). Temperature is another factor that affects the deformation mechanism of Ti-alloys, such as grain boundary sliding (GBS), which also plays a crucial role during deformation at 600 °C \([23,24]\). In cryogenic
environments, such as liquid nitrogen (77 K), liquid hydrogen (20 K) and liquid helium (4.2 K), the TWIP (twinning induced plasticity) effect of Ti-alloys is enhanced significantly. The deformation twins further affect the deformation behavior due to the dynamic Hall–Petch effect and change the orientation of grains, etc. [25]. In addition, the deformation and fracture mechanisms are affected not only by stress, such as tensile stress, compressive stress, shear stress, etc., but also by strain modes, such as fatigue load, dwell fatigue, torsional fatigue, triangular wave, sine wave, etc. Furthermore, grain orientation greatly affects plastic deformation due to the inherent strong anisotropy of the HCP structure (α-Ti) in Ti-alloys [26–28]. It has been reported that texture has an effect on twinning, slip transfer and slip blocking during deformation of polycrystalline materials [29].

Much work has been carried out during the last 20 years to identify and analyze deformation mechanisms of α-Ti in Ti-alloys under quasi-static mechanical loading. This paper reviews the slip behavior of α-Ti in α, near α and α + β alloys with equiaxed, bimodal and lamellar structures under quasi-static loading, mainly combined with in situ tensile technology. First, the mechanical properties of alloys with different compositions under continuous load and slip types known currently at room temperature are listed. Second, the activity requirements of slip systems, the influence factors of slip activity and the effects of slip transfer and blocking on deformation and fracture are discussed. Finally, the future focus and directions of study regarding the deformation mechanism of Ti-alloys are considered.

2. The Influence of Alloying Elements on Mechanical Properties
2.1. Mechanical Properties of α, near α and α + β Ti-Alloys

Similar to other metals, the mechanical properties of pure titanium are not ideal. The methods of alloying used widely in engineering applications significantly improve the mechanical properties of Ti-alloys. Even commercial pure titanium (CP-Ti) also needs to control the content of interstitial elements to improve its mechanical properties. The minimum yield strength of CP-Ti is about 170 MPa, while a yield strength of over 700 MPa and an elongation of 10% to 46% can be available after various processing [30–32]. Generally, in order to further improve the mechanical properties of α-type Ti-alloys, α-stable elements and neutral elements are often added to strengthen the α matrix. Among them, Al is the element with the most significant strengthening effect [19,33,34]. Specifically, new research shows that Zr is categorized as a weak β-stable element rather than a neutral element [34]. The mechanical properties of near-α alloys and α + β type Ti-alloys are improved significantly due to the addition of β-stable elements. Some alloys show good cryogenic strength–plasticity matching and high temperature strength with the addition of Sn and Si elements [35,36]. Table 1 shows the mechanical properties of some typical α, near α and α + β Ti-alloys.

It can be seen from Table 1 that the strengths of annealed Ti-alloys at cryogenic, room temperature and high temperature show an upward trend generally as the degree of alloying increases. Especially, the addition of interstitial elements and Al element has the greatest impact without considering the influence of processing technology. For α, near α and α + β Ti-alloys, alloying elements in α matrix create the effect of solid solution strengthening. In addition, more β-stable elements in Ti-alloy remain more β phase when cooling from high temperature to room temperature. The β phase dispersed in α matrix can influence the strength of α, near α and α + β Ti-alloys.
Table 1. Mechanical properties of typical α, near α and α + β Ti-alloys.

| Alloy                  | Processing Technology            | T/K     | YS/MPa  | UTS/MPa  | EL/%     | Ref  |
|------------------------|----------------------------------|---------|---------|----------|----------|------|
| CP-Ti (Grade 1)        | Hot Rolled + Annealed            | 293     | 356     | 454      | 50.3     | [16] |
|                        | Warm Rolled                      | 77      | 681     | 899      | 65.1     |      |
| CP-Ti (Grade 2)        | 950 °C ooled to 600 °C + AC      | 293     | 370     | 480      | 25       | [37] |
|                        | 80% Cold Rolled                  | 293     | 980     | 1057     | 12       |      |
| CP-Ti (Grade 3)        | Rod + 750 °C Annealed            | 293     | 485     | 625      | 25.6     | [39] |
|                        | Equal Channel Angular Extrusion  | 293     | 820     | 853      | 15.1     |      |
|                        | 40% Cold Rolled                  | 293     | 833     | 907      | 14.1     | [40] |
| CP-Ti (Grade 4)        | 40% Cryogenic Rolled             | 293     | 960     | 1060     | 10.1     |      |
| Ti-2Al-2.5Zr           | Rod + Annealed                   | 298 K   | 475     | 594      | 27.5     | [41] |
| Ti-2Al-1.5Mn           | Mill-annealed                    | 293 K   | 523     | 642      | 24       | [42] |
| Ti-3Al-2.5V            | Hot-extruded + Annealed          | 293 K   | 590     | 660      | 18       |      |
| Ti-4Al-1Mn             | Hot Forged + Annealed            | 293 K   | 530     | 665      | 19       |      |
|                        | Hot Rolled                       | 77      | 1266    | 1331     | 13.8     | [16] |
|                        | 40% Cold Rolled                  | 293     | 906     | 958      | 16       |      |
| Ti-5Al-2.5Sn           | Rod+ Annealed                    | 523 K   | 215     | 305      | 36.0     | [45] |
| Ti-5Al-3Mo-2V          | Hot Forged                       | 673 K   | 200     | 270      | 26.5     |      |
| Ti-6Al                 | Hot Forged + Annealed            | 300 K   | 1035    | 1105     | 14.2     | [46] |
| Ti-6Al ELI             | Hot Forged + Annealed            | 293 K   | 780     | 840      | 15       | [14] |
| Ti-6Al-2Mo             | Hot Forged + Annealed            | 293 K   | 645     | 695      | 16       | [47] |
| Ti-6Al-3Nb             | Hot Forged + Annealed            | 293 K   | 757     | 835      | 14       | [14] |
| Ti-6Al-4V              | Hot Forged + Annealed            | 293 K   | 844     | 791      | 14       | [14] |
| Ti-6Al-4V ELI          | Hot Forged + 700 °C/2 h          | 77 K    | 1482    | 1573     | 12.1     | [46] |
|                        | Hot Forged                       | 293 K   | 929     | 998      | 15       |      |
|                        | Hot Rolled                       | 293 K   | 861     | 941      | 17       | [46] |

2.2. Solid Solution Strengthening of α Phase

The common character of α, near α and α + β Ti-alloys is that composed of α phase mostly [10,48]. The strengthening of the α phase has a critical influence on the deformation mechanisms and mechanical properties of these of types Ti-alloys [30]. The change of Ti lattice due to the solution of alloy elements can be expressed by the solid solution coefficient. The solid solution coefficient $\gamma$ can be used to evaluate the dislocation nucleation of the α phase in the alloy, and the dislocation mobility $\eta$ can be used to evaluate the degree of influence on the dislocation slip. A new mathematical model has been proposed by Huang [14] to investigate the effects of alloying elements on dislocation nucleation and slip based on Yu Rui-huang electron theory [49–51]. The effect of adding alloying elements on improving dislocation mobility and mechanical properties of alloys was evaluated quantitatively. The solid solution coefficient $\gamma$ can be listed as [14]:

$$\gamma = \left| \frac{E_{A}^{\alpha\beta-Ti-M}}{E_{A}^{\alpha\beta-Ti}} \right| - 1$$

where $E_{A}^{\alpha\beta-Ti}$ represents cohesive energy of the strongest covalent bond of α-Ti without alloying atom M. $E_{A}^{\alpha\beta-Ti-M}$ represents cohesive energy of the strongest covalent bond of α-Ti-M, respectively.

$\eta$ represents the resistance to dislocation gliding on the slip planes in the same crystal lattices, which is related to the density of cohesive bond energy on the slip plane. The method of calculating the dislocation mobility $\eta$ can be listed as [14]:

$$\eta = \sum \frac{E_{A}}{s} I_{A}$$

where $E_{A}$ represents cohesive energy of the valence bond distributing on the slip plane. $I_{A}$ represents the equivalent bond number of A covalent bond on slip plane. $s$ represents the area of the slip plane [52].
Huang’s research [14] explains the solid solution strengthening effect of alloying elements in α phase. It shows that, with the increase in alloying element content in α phase, γ value reduces and η value increases. The smaller the solid solution coefficient γ, the easier the dislocations in the alloy to nucleate and proliferate. The larger the value of η, the stronger the ability of the alloy solute atoms to pin dislocations, thereby generating greater resistance to make dislocations more difficult to slip. Further effects of η on strength can be found in the literature [14]. It is worth mentioning that the accuracy and validity of using this method to study the effect of other elements on strength remain to be verified.

3. Slip Behavior in α-Ti
3.1. Determination of Slip System in α-Ti

In the initial stage of plastic deformation, crystal plane slipping by means of dislocations is usually the dominant mechanism [53,54]. For Ti-alloys, the type of slip in α-Ti can be divided into basal slip, prismatic slip, ⟨a⟩ type Π1 pyramidal slip and ⟨c + a⟩ type Π1 and Π2 pyramidal slip [55–58], as shown in Figure 1.

![Slip systems in α-Ti](image)

Figure 1. Slip systems in α-Ti: basal (B), prismatic (P) and first-order pyramidal (Π1) planes containing Burgers vector a; first and second-order (Π2) pyramidal slip planes containing Burgers vector c + a [55].

In order to study the activation of slip systems in Ti-alloys, slip systems activated must be determined first. For in situ tensile experiment, the observation of slip traces by scanning electron microscope (SEM) can analyze the activation of the slip system intuitively and conveniently. With the development of in situ technology and electron back-scattered diffraction (EBSD) technology, determining the activation of the slip system through the grain surface slip trace and grain orientation (Euler angle φ1, φ, φ2) has become an important method for the research of slip. The Schmid factor and the angle (θ) between the slip trace and tensile direction can be summarized into the following formula [47,55,59]:

\[ SF = \frac{n.R_C}{|n| \cdot |R_C|} \cdot \frac{s.R_C}{|s| \cdot |R_C|} \]

\[ \theta = \arccos \left( \frac{S_T.R_C}{|S_T| \cdot |R_C|} \right) \]

\[ S_T = n \times N_c; R_C = g.R; N_c = g.N \]

\[ g = \begin{pmatrix} \cos \phi_1 \cos \phi_2 - \sin \phi_1 \sin \phi_2 \cos \phi & \sin \phi_1 \cos \phi_2 + \cos \phi_1 \sin \phi_2 \\ -\cos \phi_1 \sin \phi_2 - \sin \phi_1 \cos \phi_2 \cos \phi & \sin \phi_1 \cos \phi_2 - \cos \phi_1 \sin \phi_2 \\ \sin \phi_1 \sin \phi & \cos \phi_1 \sin \phi \end{pmatrix} \]

where \( n \) represents the normal direction of slip plane, where \( s \) represents the slip direction, where \( S_T, R_C \) and \( N_c \) represent crystal coordinates (define the tensile direction as the RD axis and define the normal direction as the ND axis [60,61]) and where \( g \) represents the coordinate transformation matrix. Generally, \( \theta \pm 5^\circ \) deviation tolerance is used in the comparison between theoretical values and experimental observations.
The activation of slip systems in α grains can be listed by geometrical method. Some reports have shown the Schmid factors of each type slip into the grain in the form of contour lines in the triangle of inverse pole figure (IPF) [62].

As shown in Figure 2, McLean P. Echlin et al. [62] took grain A in Ti-6Al-4V as an example to calculate the angle between slip traces and tension direction. Then, the Schmid factor and theoretical slip trace of each slip system in each grain can be determined. The difficulty of slip activation in grains with different orientations can be evaluated by the distribution of the Schmid factor in the triangle of IPF.

![Figure 2](image-url)

**Figure 2.** The pyramidal slip system activity in α-grain A: (a) position of slip trace in IPF; (b) theoretical slip trace; (c) distribution of Schmid factor in the triangle and position of grain A [62].

3.2. Slip Activity

3.2.1. CRSS

The significant difference in CRSS values of different slip modes makes the deformation behavior on a microstructure scale more complicated [63–65]. Some authors have worked to identify the incipient slip activity in α-Ti and the CRSS values of slip systems. In these studies, only α phase is considered separately. The measurement of CRSS includes simulation calculation and single crystal experiment method. These studies have differences on the methods of calculating CRSS, and there are also different values in the literature. One of the important factors is the physical model used to measure CRSS. Most authors used a Taylor (uniform strain assumption) or a Sachs (uniform stress assumption) model and the self-consistent models [66]. To improve accuracy, T. Dick [67] established a more advanced creative method, finite element (FE) computations, by using crystal plasticity material models to re-evaluate the CRSS of α-Ti in Ti-6Al-4V precisely. In the single crystal experiments, the difference between the measured values may be caused by the difference in the purity of the alloy and the uniformity of the alloying elements. The slight difference between the contents of interstitial elements oxygen, nitrogen and replacement elements can influence activation of the slip systems [64,67]. The CRSS measurements and activity of each slip type of typical alloys are listed in Tables 2 and 3.
Table 2. CRSS of $\alpha$-Ti in some typical Ti-alloys.

| Materials         | CRSS (MPa) | CRSS Normalized to Prismatic | Ref |
|-------------------|------------|-------------------------------|-----|
|                   | Basal      | Prism                        | Pyr. $<$c + a$>$ | Pyr. $<$c + a$>$ | Pyr. $<$c + a$>$ |
| CP-Ti             | 98         | 224                          | 1.2       | 2.3       | [68]   |
| CP-Ti             | 160        | 770                          | 0.95      | 3.9       | [70]   |
| Ti-6Al            | 322        | 846                          | 1.0       | 2.6       | [58]   |
| Ti-6Al            | 365        | 620                          | 1.1       | 1.8       | [71]   |
| Ti-3Al            | 160        | 120                          | 1.3       | [70]       |
| Ti-6Al            | 190        | 200                          | 0.9       | [70]       |
| Ti-6Al-2Sn-4Zr-2Mo| 362        | 410                          | 0.9       | 2.6       | [72]   |
| Ti-6Al-4V         | 444        | 392                          | 404       | 1.1       | 1.6     | [73]   |
| Ti-6Al-4V         | 420        | 370                          | 490       | 1.1       | 1.3     | 1.6     | [74]   |
| Ti-6Al-4V         | 373        | 388                          | 1.0       | [55]       |
| Ti-6Al-4V         | 325        | 355                          | [72]      |
| Ti-6Al-4V         | 338        | 352                          | [64]      |
| Ti-6Al-4V         | 400        | 380                          | 640       | 1.0       | 1.7     | [67]   |
| Ti-8Al-1Mo-1V     | 276        | 207                          | 248       | 1.3       | 1.2     | [76]   |

Table 3. Statistics on the proportion of slip-activated grains.

| Material         | Temperature | $\epsilon$ (%) | Pd (%) | Pds (%) | Ref |
|------------------|-------------|----------------|--------|---------|-----|
|                  |             |                |        | Basal   | Prism | Pyram $<$c + a$>$ | $<$c + a$>$ |
| Ti-6Al           | RT          | 10             | –      | 26      | 68   | 14           | –   | [47]   |
| Ti-6Al-4V        | 27K         | 0.5            | 33     | 18.5    | 12.5 | 0            | 1.0 | [57]   |
| Ti-6Al-4V        | 225 K       | 0.5            | 26.4   | 14      | 10.7 | 1.6          | 0   | [57]   |
| Ti-6Al-4V        | 550 K       | 0.5            | 51     | 11.4    | 12.3 | 23.1        | 4.2 | [57]   |
| Ti-6Al-2Sn-4Zr-2Mo| RT         | 18.3           | 80     | 28      | 48   | 4            | –   | [72]   |
| Ti-6Al-2Sn-4Zr-6Mo| RT         | 18.4           | 83.9   | 61.0    | 19.5 | 3.4          | –   | [72]   |
| Ti-5Al-2.5Sn     | RT          | 3.5            | –      | 57.8    | 68.7 | 4.8          | 6.1 | [77]   |
| Ti-6Al-4V        | RT          | 1.2            | –      | 16.7    | 66.7 | 9.8          | 0   | [24]   |
| Ti-6Al-4V        | RT          | 1.2            | –      | 87      | 52   | 33           | 3   | [64]   |

Note: Pd: percentage of deformed grains; Pds: percentage of activated slip planes in the deformed grains.

It can be seen from the measurement data in the above reports that, as the content of alloying elements in the titanium alloy increases, the solid solution strengthening effect is stronger, leading to a higher CRSS of the same type of slip system in $\alpha$-Ti, which means that the slip activity is more difficult. In fact, alloying elements can not only increase the CRSS value but also have different effects on the CRSS value of the three slip modes. For example, with the addition of Al element, the CRSS ratio of basal slip and prismatic slip gradually decreases [70,78]. In $\alpha$ phase, the CRSS values of prismatic slip and basal slip are smaller [79,80]. There are slight differences in the comparison of the CRSS values between prismatic slip and basal slip in the reports; the stress ratio is between 0.8 and 1.2, which reveals that the CRSS of basal slip is similar to prismatic slip. According to the previous reports, the CRSS of the $<$c + a$>$ type pyramidal slip is much higher than $<$c + a$>$ type slip when the deforming temperature is lower than 700 K, which means the kinds of slip systems are more difficult to be initiated [57].

Table 3 shows some studies on the relative activity frequency of each slip system. For example, in the tensile test of Ti-6Al-4V alloy at 77 K, 225 K and 550 K, the relative activity frequency of each slip system shows obvious differences due to the CRSS influenced by temperature. However, in addition to the alloy composition, texture is the most important factor affecting the activity of each slip system at room temperature, such as the study of Ti-6Al-2Sn-4Zr-2Mo and Ti-6Al-2Sn-4Zr-6Mo alloys with different textures.

3.2.2. The Effect of Grain Orientation on Slip Activity

A slip system with a lower CRSS does not mean it is preferentially activated during deformation since grain orientation is also a key factor for the activity of each slip system by determining the Schmid factor [81]. As shown in Table 3, in slipped grains, there was a higher degree difference in the percentage of slip system in $\alpha$-Ti. The source of these differences is attributed to the orientation of the grains or the texture in polycrystalline...
materials. With the influence of the Schmid factor, assuming that the polycrystalline material with a stronger [0001] texture on ND is subjected to tensile stress on the TD or RD, pyramidal slips will be activated in a greater proportion of grains compared with other textures. However, affected by a higher CRSS, pyramidal slip is not dominant in the plastic deformation yet compared to other types of slip. For example, in the study of S. Hémery, P. Villechaise [72], the basal and prismatic slip of Ti-6Al-2Sn-4Zr-2Mo with [10-10] texture and Ti-6Al-2Sn-4Zr-6Mo with [0001] texture are shown in Figure 3. In the incipient plastic deformation, the basal slip and prismatic slip are first activated due to smaller CRSS. With the influence of texture, the basal slip in Ti-6242 grains activated at a lower stress. As the macroscopic stress increases, the prismatic slip gradually replaces the dominant role of the basal slip, which always plays a predominant role in Ti-6246 alloy. If the texture effect was not considered, two slip systems with a Schmid factor greater than 0.36 would be activated. The slip frequency of Ti-6242 is higher than that of basal slip, while Ti-6246 is the opposite. In addition, the c/a ratio of CP-Ti is 1.58; the increase in c/a ratio is one of the reasons for the activation of the basal slip in Ti-6Al-4V [57]. It can be seen that, even if the two alloys are similar, the type and activity frequency of slip systems are not static. The characteristics of some alloys, especially texture, can dominate basal or prismatic slip at incipient plastic deformation. The reasons for texture formation and the influence of texture on slip activity are still being studied.

Figure 3. Slip activation of the (a) Ti-6242 and (b) Ti-6246 alloys under different stresses and different Schmid factors at 800 MPa [72].

3.3. Slip Transfer
3.3.1. Slip Transfer across α-Grain Boundary

As we all know, grain boundaries are strong barriers for dislocation slip, which is the reason why the strengthening effect can be achieved by grain refinement. The yield strength as the function of the square root of grain size obeys the Hall–Petch relation. However, the obstacle effect is not absolute: slip transfer occurs when the stress reaches a certain
level at the grain boundary [82]. In the coordinated deformation process of polycrystalline Ti-alloys, slip transfer is one of the crucial factors affecting the plastic and toughness. In recent years, studies have shown that the small-angle grain boundary determined by grain orientation is beneficial to slip transfer [83]. Earliest, Livingston and Chalmers [84] conducted a study on the cross-grain boundary slip activity in grain pairs and proposed a quantitative rationale for slip transfer to occur. A similar effort was made to parametrize slip transfer between grain pairs in Ti-Al alloys by Luster and Morris [85], defining slip transfer factor \( m' = \cos \phi \cos \lambda \). As shown in Figure 4, horizontal (orange and blue) planes signify slip or twinning planes on either side of the boundary, where \( \kappa \) is the angle between slip directions, where \( \psi \) is the angle between plane normals and where \( \theta \) is the angle between plane traces on the boundary. The value of slip transfer factor \( m' \) is between 0 and 1. The larger the the value of \( m' \), the higher the align degree of the slip systems in the grain pair and the easier the slip transfer, resulting in observations of slip transfer in TEM specimens. T.R. Bieler’s research [86] on pure tantalum, TiAl, commercial purity titanium, titanium alloys and other materials found that slip transfer not only promotes plastic deformation but also has a great impact on the fracture process. Therefore, in the studies of the deformation mechanism of titanium alloys, the slip transfer behavior is a crucial factor; it is necessary to study it.

![Figure 4. Geometry of slip transfer across a grain boundary [86].](image)

The finding of Huang et al. [47] regarding the in situ tensile deformation process of Ti-6Al alloy is that slip transfer can not only coordinate the increased macroscopic strain effectively but also release the stress concentration at the grain boundary. As shown in Figure 5, regarding the slip transfer factor \( m' = 0.963 \) of slip system (10–10) [1–201] between the A and B \( \alpha \)-grains, slips are transferred to the B grain easily when the larger Schmid factor slip systems activated in A grain. However, the slip transfer between the C and D grains is difficult with the slip transfer factor \( m' = 0.560 \), which further leads to higher internal stress at the grains. The higher internal stress at the grain boundary shown in the kernel average misorientation (KAM) map may induce nucleation of micro-cracks at the grain boundary and promote fracture of the alloy.

![Figure 5. Slip transfer and blocking in Ti-6Al alloy: (a) IPF map; (b) SEM map; (c) KAM map; (d) linear distributions of KAM values perpendicular to grain boundaries A/B and C/D [47].](image)
S. Hémyer et al. [87, 88] used in situ tensile experiments to make a series of statistics on the slip transfer between α grains of Ti-6Al-4V under different tensile stresses. The study summarized the relationship between slip transfer factor m′ and the type of incoming–outgoing slip systems. The report shows that the probability of slip blocking decreases with increasing the value of m′. When the value is close to 1, the slip transfer will not pile up. It can be seen from Figure 6, which is not mentioned in the literature, that, for Ti-6Al-4V, the slip transfer between the α grains hardly occurs if the value of m′ is less than 0.6. Otherwise, the grain-pairs proportion allowing slip transfers increases greatly with m′. The study also generated statistics on the slip types of incoming slip systems and outgoing slip systems during slip transfer. It is interesting that most of the basal slip in outgoing grain comes from the basal slip of incoming grains, and most of the prismatic slip in outgoing grain comes from the prismatic slip of incoming grains. In addition, the basal and prismatic slip come from the pyramidal slip of some grains. Due to the high CRSS of pyramidal slip, these grains are few in number. However, in the statistics of slip transfer with a small misorientation (κ ∈ [0–15°] U ψ ∈ [0–15°]), the pyramidal slip that came from prismatic and basal slip transfer occupied a relatively large part of all the pyramidal slip grains. It clearly indicates that the slip transfer at the small angle α grain boundary prefers to induce the slip with higher CRSS. However, currently, the research remains lacking regarding slip transfer of small-angle grain boundaries to induce slip activity, and there are almost no principles proposed from an atomic level. Furthermore, pyramidal slip, as one of the three major types of α-Ti slip, requires more profound investigation.

![Figure 6](image-url)

**Figure 6.** (a) The distribution of the frequency of slip transfer and blocking of different tensile stress relative to m′; (b) the distribution of outgoing slip systems with respect to incoming slip systems; (c) distribution of outgoing slip system excluding κ ∈ [0–15°] U ψ ∈ [0–15°] domain with respect to the incoming slip system [87].

The orientation relationship determines the geometric compatibility between neighbor grains. During the plasticity deformation of polycrystalline materials, the existence of texture always influences slip transfer. McLean P. Echlin et al. [62] found that, during the tensile test of the annealed Ti-6Al-4V alloy after cold rolling, there is a long slip trace that penetrates 21 equiaxed grains. As shown in the IPF map (Figure 7a), the alloy has obvious texture in the annealed microstructure after cold rolling. With the strain continuously increasing, the slip traces characterized by scanning electron microscope digital image correlation (SEM-DIC) became more uniform and denser [89]. The activated slip systems transferred through multiple grains with similar orientations in these gathering areas, which made the slip traces longer. However, slip traces are blocked at the regional boundaries with other grain orientation gathering areas [88]. If the gathering area of grains with similar orientation is larger, the slip traces of the activated slip systems will extend longer by transfer and improve the ductility of the alloy. It provides feasibility for improving the plasticity by controlling texture, and, meanwhile, solves the problem of strength–ductility mismatch in some Ti-alloys.
Figure 6. (a) The distribution of the frequency of slip transfer and blocking of different tensile stress levels. (b) The fraction of total slip bands having localized slip band lengths measured using SEM-DIC.

3.3.2. Slip-Transfer-Caused Twinning

Twins result in more slip systems activation than parent grains, interfacial mismatches and localized stress–strain fields with the matrix, which leads to higher strength and ductility [90,91]. Twinning has an extremely important influence on the entire deformation process. Slip transfer can induce the nucleation of twins at the grain boundary and propagate with further deformation. In the study of slip transfer by Wang et al., as shown in Figure 8a,b, the deformation twins in grain 2 are thicker at the grain boundary with grain 1 and gradually become thinner close to the grain boundary with grain 3. This indicates that the deformed twins nucleated from the 1–2 grain boundary and grew to the 2–3 grain boundary. According the report, the twins in grain 2 are T1 twins, and the twinning system is (10–12) [−1011]. The twinning system has a smaller Schmid factor compared with twinning system (0–112) [01–11]. However, the transfer factor \( m' \) between the twinning system in grain 2 with the prismatic slip system in grain 1 is as high as 0.93. The prismatic slip in grain 1 hit the grain boundary, transferred to grain 2 on the other side and boosted the formation of twin dislocations at the grain boundary to promote nucleation of twins [92]. In Figure 8d, with the increase in strain, the dynamic process of slip-induced twinning nucleation and growth is clearly recorded by in situ observation [93]. Similarly, in Figure 8c, when the orientation of the grains allows, not only slip transfer but also twin transfer can promote twinning nucleation in neighbor grains.

C. Lavogiez et al. [94] analyzed the process of slip-transfer-caused twinning in Ti-6Al-4V with tensile fatigue load by SEM and EBSD. The boundary area of the neighboring grains being studied was marked in Figure 9a,b. According to reports, when the loading direction is along the X axis, the SF value of the incoming prismatic slip system (−1100) [11–02] corresponding to the slip bands in grain A is 0.49; the SF value of \( c + a \) pyramidal slip system (0–112) [−1–122] corresponding to the slip bands in grain B is 0.47. However, the geometric compatibility factor \( m' \) of the slip traces in the pair of grains is only 0.007. The low \( m' \) value rules out \( c + a \) pyramidal slip resulting from slip transfer of incoming prismatic slip bands. In addition, in grain B, new grains with larger misorientation from parent grain appear near the grain boundary of grain A. From measurement of misorientation angle (Figure 9c) and pole figures (PFs) (Figure 9d), the common pole observed on the pole figures further confirms the [10–12] <10–11> deformation twinning. Interestingly, in the above studies, the incoming slips that induce twinning nucleation by slip transfer are all prismatic slip. However, there is no report on whether the twinning nucleation can be caused by incoming basal or pyramidal slip. The research regarding what types of twinning nucleation are induced by slip transfer on α-grain boundary is lacking.
in Figure 8a,b, the deformation twins in grain 2 are thicker at the grain boundary with grain 1; (b) IPF of a part of grains in (a); (c) slip and twin transfer induced twinning [86]; (d) in situ observation of the twinning process induced by slip transfer [93].

Figure 9. (a) SEM micrograph of twinning caused by slip transfer in a pair of grains; (b) the corresponding IPF map along the loading direction; (c) the misorientation profile along path I; (d) [10–12] and [10–11] pole figures at positions 1 and 2 [94].

3.3.3. Slip Transfer of α/β Phase Boundary

A complex interplay of morphological and orientation relationship occurs due to the interaction between α and β crystal [95]. Easy slip transfer without the production of
interface defects and debris will result in homogenous plastic deformation [84]. Shockley partials are found to play an important role in the underlying dislocation transfer across \(\alpha/\beta\) interfaces, which offer viable dislocation reactions [96]. As shown in Figure 10, Xiaodong Zheng et al. [97] used a transmission electron microscope (TEM) to characterize the atomic arrangement of the \(\alpha/\beta\) interface in three dimensions. A model was used to systematically show the approximately parallel relationship between certain crystal planes and crystal orientations in the \(\alpha\) phase and \(\beta\) phase. According to reports, there were (01–10)\(\alpha\) || (–121)\(\beta\), (2–1–10)\(\alpha\) || (11–1)\(\beta\), (0002)\(\alpha\) || (101)\(\beta\) and two pairs of nearly parallel planes (–1100)\(\alpha\) || (–101)\(\beta\) (–5°) and (–1–120)\(\alpha\) || (0–10)\(\beta\) (–5°), which also provides a basis for slip transfer between \(\alpha\) and \(\beta\) phases. The selected area electron diffraction (SAED) pattern in Figure 10b revealed the relation between slip directions in \(\alpha\) and \(\beta\) phases. The three \(a\)-type slips on the (0001) basal plane of \(\alpha\) phase are marked with \(a_1 = [2–1–10]\), \(a_2 = [–12–10]\) and \(a_3 = [–1–120]\), and three slip directions in \(\beta\) phase are labeled by \(b_1 = [11–1]\), \(b_2 = [–111]\) and \(b_3 = [0–10]\), where \(a_1\) is parallel to \(b_1\), while \(a_2\) and \(a_3\) deviate a few degrees from \(b_2\) and \(b_3\). These crystallographic relations are consistent with the Burgers orientation relation (OR) [97,98].

Dislocation emission at the interface often occurs, especially in the early stages of deformation. In situ observations by TEM (JEOL 2010, equipped with a SIS CCD camera)
were studied. P. Castany et al. [99] conducted research in detail on the slip of dislocations at the interface between α and β phases with tensile loading. According to the report, the α_p/β interface promoted the nucleation and multiplication of dislocations. As shown in Figure 11a, two dislocations are emitted from the interface due to local stress. In the next image, the two dislocations slightly slip, and a new dislocation is emitted from the interface. Determined by the $g \cdot b = 0$ invisible criterion, these dislocations have a-type Burgers vector and slip on the prismatic planes. The reason is that there is no near-Burgers orientation relationship between $\alpha_p$ and $\beta$.

![Image](image-url)

**Figure 11.** (a) In situ TEM observation of the dislocation emission from $\alpha_p/\beta$ interface; (b) in situ sequence showing simultaneous slip of dislocations in two $\alpha$ laths and one $\beta$ lath [99].

As shown in Figure 11b, the dislocations emitted at the interface $\alpha_s/\beta$ can also slip in the layered colony when the phase interface has a near-Burgers orientation relationship [98]. Two $\alpha_s$ colonies, labeled $\alpha_{s1}$ and $\alpha_{s2}$, are separated by a thin $\beta$ lath. The screw dislocation Burgers vector is $b_1 = a/3[11\overline{2}0]$ and moves from right to left along the basal plane. Despite the existence of $\beta$ lath, the orientation relationship between $\alpha_{s1}$ and $\alpha_{s2}$ is still aligned well, and $b_1 = a/3[11\overline{2}0]$ is equal to or close to $b_2 = a/2[11\overline{1}]$. The dislocation line passed through the thin $\beta$ lath and slipped both sides of it. Occasionally, although the dislocation on one side was pinned, that on the other side continued to slip.

### 3.4. Slip Blocking

#### 3.4.1. Slip Blocking Caused Twinning

In Ti-alloys with a fully lamellar microstructure, deformation twins tend to nucleate at the $\alpha/\beta$ interface [61,100] due to the localized stress/strain concentration at the interface caused by intrinsic misfit dislocations and dislocation accumulation [100]. Deformed twins tend to nucleate at the $\alpha/\beta$ interface in fully lamellar structure because the crystal lattice intrinsic misfit dislocations and dislocation plugging will cause local stress/strain concentration at the interface. Xiaodong Zheng et al. [95] studied the nucleation process of twins at the phase interface and conducted a modified displacement gradient accommodation (m-DGA) analysis. The dissociation process of a basal $<a\bar{c}>$ dislocation into $[1\overline{1}02]$ twins was determined. Burgers vectors for basal slip were $a_1 = \pm 1/3[2\overline{1}0\overline{1}0]$, $a_2 = \pm 1/3[12\overline{1}0]$ and $a_3 = \pm 1/3[\overline{1}2\overline{1}0]$. As shown in Figure 12, the incoming dislocation with Burgers vector $a_1 = \pm 1/3[2\overline{1}0\overline{1}0]$ is screw type and has zero shear along the twinning shear associated with the $[1\overline{1}02]$ twins. Therefore, $a_1$ cannot be used as a twin dislocation to cause twin nucleation. However, when dislocations with Burgers vectors $a_2 = \pm 1/3[12\overline{1}0]$ and $a_3 = \pm 1/3[\overline{1}2\overline{1}0]$ were blocked and piled up at interfaces, $a_2$ and $a_3$ decomposed at the $\alpha/\beta$ interface into twinning dislocations of $(01\overline{1}2)$ and $(\overline{1}1\overline{0}2)$ twinning variants along the
crystal direction of [0–111] and [1–101], respectively. It is demonstrated by the two beam diffraction TEM technique (see the literature [95] for details).

![Twinning nucleation mechanism](image1)

**Figure 12.** Schematic of the twinning nucleation mechanism from <a> type dislocations dissociation (a) schematic of reaction of basal <a> type dislocations at different planes into different twin dislocations and residual dislocations; (b) dissociation of a₂ and a₃ dislocations at terrace plane; (c) dissociation of a₂ dislocation at ledge plane. [97].

### 3.4.2. Slip Blocking Induced Crack Initiation

The fracture process is rather complex, even in single crystals. It depends on crystal defects, chemical composition that may influence surface energies and environment that may have an essential effect on crack tip behavior, etc. [101]. In this paper, we focus on the influence of slip blocking on crack nucleation. As we all know, the formation of microcracks in polycrystalline materials mostly comes from the stress concentration due to the poor coordinated deformability. As shown in Figure 13, during the in situ tensile study of Ti-6Al alloy and CP-Ti, Wang et al. found that the grain boundaries, especially with larger misorientation, are harmful to slip transfer. Cracks always nucleate at the intersection of grain boundaries where the dislocations are piled up due to difficult slip transfer. Similar to some grain pairs in Figure 13b, many slip traces can be seen in grains 1 and 2 in Figure 13a, and there is almost no slip trace between grains 1 and 2 and grains 3 and 4, which proves that slip blocking happened. A great number of dislocations inevitably piled up at the intersections of grain boundaries. In this case, nano-void formed, then developed into micro-cracks.

![Crack initiation](image2)

**Figure 13.** Crack initiation in intersection of α grain boundaries: (a) Ti-6Al [47]; (b) CP-Ti; (c) schematic of the nano-void nucleation; (d) schematic of the dimple nucleation [102].

The micro-cracks in lamellar structure tend to nucleate at the interface of intergranular α, and the deformability of intergranular α sheet is poorer than that of α lath colony [103]. Motomichi Koyama et al. [103] conducted in situ tensile observations on the lamellar structure of Ti-6Al-4V and found that micro-crack was nucleated along the interface of intergranular α-sheet, and this phenomenon was closely related to blocking of the dislocation slip. As discussed above, slip transfer between α lamellae within α colony is convenient due to the same orientation in the grains, and the α/β interface was not an important
obstacle to changing the dislocation slip path. For example, the yellow line in Figure 14 clearly showed that the slip traces in α can pass through β lath easily, and the α colonies showed excellent slip transfer ability. However, the slip is hindered near the intergranular α sheet with high misorientation, resulting in strain localization and formation of interface damage [104]. In addition, for the Ti-alloy with lamellar structure, the effective grain size for deformation is equivalent to the size of α colonies beam, and, for the intergranular α sheet, the thickness can be considered as the effective grain size. The parallel distribution of α colonies can be approximated to the larger effective grain size due to the same orientation. As we all know, smaller grain size has higher dislocation blocking ability and smaller dislocation capacity, so the strain localization is easier to form near the intergranular α sheet. Moreover, the height of the surface fluctuation caused by the slip is determined by the number of dislocations emitted from the surface [105]. The above reasons lead to the accumulation of dislocations at the interface of the intergranular α sheet, and the shear stress concentration near the alloy surface can aggravate the formation of micro-cracks, as shown in the evolution model in Figure 14c–e.

![Figure 14.](image)

Figure 14. (a) Slip traces in magnified SE image of lamellar structure in Ti-6Al-4V; (b) KAM map near intergranular α sheet; (c) schematic of original microstructure; (d) schematic of dislocation pile up at intergranular; (e) schematic of the crack nucleation [103].

In near α and α + β alloys, a bimodal or tri-modal structure will provide more sites for micro-cracks. As shown in Figure 15, Tan et al. [106] studied the tensile deformation process of a damage tolerance TC21 alloy. Research shows that it has more advantages to form a crack in the primary α lath than equiaxed α phase and β-trans matrix. During deformation, the plastic deformation in α phase was dominant in the early stage of deformation, and the separated β grains were compelled to deform plastically [107]. Compared with α lath colony and the α in β-trans matrix, after a large amount of deformation, equiaxed α grains exhibited better deformability and more slip systems activated. The mode of deformation is changed from single slip to multiple slip. Although dislocations pile up at the interface, multiple slip bands can partially release the stress concentration of the α/β interface. As for the primary α lath colony, its effective grain size is close to α lath thickness, and the deformation ability of α lath is greatly restricted during the deformation process, so dislocations more easily pile up at the α/β interface and develop into a crack source. The
size of α in β-trans matrix is smallest and the deformability of it is the worst. Due to the small size and high dispersion, slip blocking, dislocation pile up and stress concentration occur during deformation. As a result, under the same load, β-trans matrix was highly strengthened and it is more difficult to deform than other positions. Therefore, the α/β interface does not need to accommodate high deformation, which is also the reason why the probability of micro-cracks is between the other two cases.

![Image](image-url)

**Figure 15.** (a) Crack nucleation at interface of equiaxed α phase; (b) crack nucleation at α lath; (c) crack nucleation at interface of β-trans matrix; (d) probability of crack nucleation sites [106].

In a word, the lamellar structure has the worst plastic deformation capacity. For example, when the lamellar Ti-6Al-4V alloy is subjected to a tensile test, only a fracture extension of less than 7% is achieved, while the fracture elongation of Ti-6Al-4V alloy with equiaxed α structure can reach more than 10% [24,62]. A crack in Ti-alloy is closely related to slip transfer and blocking at the interface. In future research, we can continue to study how to control the structure to adjust slip transfer and blocking at the interface to improve the mechanical properties.

4. Summary

In this paper, a review of the slip behavior of α-Ti in Ti-alloys under quasi-static loading was performed according to the recent research. These studies were mainly focused on the mechanical properties and deformation mechanisms of α-Ti in different microstructures. The in situ tensile technology was effectively utilized for the investigation of the slip behavior of α-Ti in Ti-alloys under quasi-static loading. The research progress is summarized as follows:

1. Through statistics on the mechanical properties of a variety of typical widely used α, near α and α + β Ti-alloys, it was found that alloying elements have the most profound impact on the mechanical properties. Numerous scholars have quantified the effects of alloy atoms on the dislocation slip by combining electron theory and experimental results, which provides important references for the design of new Ti-alloys in future research.

2. The slip behavior in α-Ti is dominant in the deformation process of the Ti-alloy, especially at the initial deformation stage. Studies have shown that, for slip behavior in α-Ti affected by CRSS and Schmid factor together, prismatic slip occurs most
frequently among the three slip modes. The CRSS values measured by various methods show that the CRSS of prismatic slip is the lowest in most alloys, the CRSS of basal slip is relatively close to it and the research on a few alloys shows the CRSS of basal slip is slightly smaller than that of prismatic slip. The CRSS of pyramidal slip is much higher than that of basal slip and prismatic slip, and it appears at the lowest frequency during deformation.

(3) In the process of slip transfer, the geometric compatibility factor $m'$ can be used as a criterion for slip transfer between neighbor grains. The larger the value, the slip transfer more easily leads to better coordination in polycrystalline Ti-alloys during deformation; meanwhile, when the Schmid factor of the twinning system is large enough and the value of $m'$ is small, slip blocking is prone to induce twins. Lack of twins results in increasing stress concentration, which becomes the reason for the formation of crack sources.

However, at present, there are still limitations in the main research direction and research depth of the deformation mechanism of Ti-alloys. The research deficiencies and suggestions are summarized as follows:

(1) For the electronic theory, the effects of elements in Ti-alloy on mechanical properties are still under verification and need to be supplemented. In addition to alloying elements, the properties of Ti-alloys are affected by many factors, such as the grain size and second phase. In research on the effects of alloying elements on properties through electronic theory, most studies try to exclude the influence of other factors in polycrystalline Ti-alloy materials due to the difficulties of preparing single crystals. However, the influence of these factors is difficult to eliminate completely, so effects exist on the reliability of experimental results. In future studies, single crystals should be used as much as possible. More accurate and systematic models should be established to exclude other factors.

(2) As relatively traditional metal material, Ti-alloys have been studied in depth on their structure and mechanical properties. However, even if in quasi-static deformation, the deformation mechanism is reflected in the constantly changing deformation process. The process is difficult to characterize consistently, which leads to the lack of systematic research on the deformation mechanism and low degree of microscopcity, especially at the atomic scale. The in situ observation of deformation can record the deformation behavior at different deformation rates and elongations in real time. In future research, in situ methods can be further used to study the deformation mechanism of Ti-alloys.

(3) Whether intragranular deformation or intergranular coordinated deformation, slip behavior in $\alpha$-Ti will be affected by the polycrystalline texture. By modifying slip behavior via the specific texture, the mechanical properties could be improved. Moreover, stronger texture may lead to anisotropy of mechanical properties. In order to adapt to different application conditions, controlling the texture provides a new method for processing design.

(4) Whether slip transfer or blocking, the influence of the interfaces is important. For the $\alpha$, near $\alpha$ and $\alpha + \beta$ alloys, the interfaces mainly comprise the grain boundary, the $\alpha/\beta$ lamellar interface and the interface between equiaxed $\alpha$ and $\beta$-trans matrix, etc. The dynamic evolution of their interaction is complex in the case of slip transfer across the interface. However, so far, little research has systematically clarified it on the atomic scale. In future work, further systematic studies in this direction could be carried out to optimize the mechanical properties of Ti-alloys by adjusting the interface to control slip transfer and blocking, which caused twinning and crack initiation.
Author Contributions: Literature search, R.Z., Q.Z., Y.Z., D.G. and Y.D.; Figures, R.Z.; Data collection, R.Z. and Q.Z.; Data interpretation, R.Z., Q.Z., Y.Z., D.G. and Y.D.; Writing, R.Z., Q.Z., Y.Z., D.G. and Y.D.; Study design, Y.Z. All authors have read and agreed to the published version of the manuscript.

Funding: This research was funded by the Science and Technology Major Project of Shaanxi Province of China [No. 2020dzx04-01-02] And National Natural Science Foundation of China [No.52101122].

Acknowledgments: The authors are grateful for the above financial support.

Conflicts of Interest: The authors declare no conflict of interest.

Abbreviations

Critical resolved shear stress (CRSS); hexagonal close-packed (HCP); body-centered cubic (BCC); grain boundary sliding (GBS); twinning induced plasticity (TWIP); commercial pure titanium (CP-Ti); yield strength (YS); ultimate tensile strength; elongation (EL); scanning electron microscope (SEM); electron backscattered diffraction (EBSD); inverse pole figure (IPF); Schmid factor (SF); finite element (FE); kernel average misorientation (KAM); digital image correlation (DIC); pole figures (PFs); transmission electron microscope (TEM); orientation relation (OR); modified displacement gradient accommodation (m-DGA).

References

1. Zhao, G.; Xu, X.; Dye, D.; Rivera-Diaz-del-Castillo, P.E. Microstructural evolution and strain-hardening in TWIP Ti alloys. Acta Mater. 2020, 183, 155–164. [CrossRef]

2. Bahl, S.; Suwas, S.; Chatterjee, K. Comprehensive review on alloy design, processing, and performance of titanium alloys as biomedical materials. Int. Mater. Rev. 2021, 66, 114–139. [CrossRef]

3. Chong, Y.; Bhattacharjee, T.; Tian, Y.; Shibata, A.; Tsuji, N. Deformation mechanism of bimodal microstructure in Ti-6Al-4V alloy: The effects of intercritical annealing temperature and constituent hardness. J. Mater. Sci. Technol. 2021, 71, 138–151. [CrossRef]

4. Teixeira, O.; Silva, F.J.G.; Ferreira, L.P.; Atzeni, E. A Review of Heat Treatments on Improving the Quality and Residual Stresses of the Ti-6Al-4V Parts Produced by Additive Manufacturing. Metals 2020, 10, 1006. [CrossRef]

5. Sen, M.; Suman, S.; Mukherjee, S.; Banerjee, T.; Sivaprasad, S.; Tarafder, S.; Bhattacharjee, A.; Kar, S.K. Low cycle fatigue behavior and deformation mechanism of different microstructures in Ti-5Al-5Mo-5V-3Cr alloy. Int. J. Fatigue 2021, 148, 106238. [CrossRef]

6. Xu, Z.; Huang, C.; Tan, C.; Wan, M.; Zhao, Y.; Ye, J.; Zeng, W. Influence of microstructure on cyclic deformation response and micromechanics of Ti55531 alloy. Mater. Sci. Eng. A 2021, 803, 140505. [CrossRef]

7. Lu, K. The future of metals. Science 2010, 328, 319–320. [CrossRef]

8. Mosleh, A.O.; Kotow, A.D.; Vidal, V.; Mochugovskiy, A.G.; Velay, V.; Mikhailovskaya, A.V. Initial microstructure influence on TiAlMoV alloy’s superplastic deformation behavior and deformation mechanisms. Mater. Sci. Eng. A 2021, 802, 140626. [CrossRef]

9. Elshaer, R.N.; Ibrahim, K.M. Effect of cold deformation and heat treatment on microstructure and mechanical properties of TC21 Ti alloy. Trans. Nonferrous Met. Soc. China 2020, 30, 1290–1299. [CrossRef]

10. Wu, X.; Zhang, B.; Zhang, Y.; Niu, H.; Zhang, D. Correlation between microstructures and tensile deformation behavior of a PM near Ti6Al2Sn4Zr2Mo0.1Si alloy. Mater. Sci. Eng. A 2021, 825, 141909. [CrossRef]

11. Kun, Z.; Zai, Z.; Yumaa, Z.; Xinhua, W.; Williams, J.; Enquan, L.; Jisheng, M.; Ren, Z.; Lim, C.V.S.; Aijun, H. Effect of deformation reduction on microstructure, texture, and mechanical properties of forged Ti-6Al-4V. J. Mater. Eng. Perform. 2021, 30, 1147–1156.

12. Zhao, Z.; Wang, G.; Zhang, Y.; Gao, J.; Hou, H. Microstructure evolution and mechanical properties of Ti-6Al-4V alloy prepared by multiaxis equal channel angular pressing. J. Mater. Eng. Perform. 2020, 29, 905–913. [CrossRef]

13. Huang, Z.W.; Yong, P.L.; Liang, N.N.; Li, Y.S. Slip, twinning and twin-twin interaction in a gradient structured titanium. Mater. Charact. 2019, 149, 52–62. [CrossRef]

14. Huang, S.; Zhao, Q.; Zhao, Y.; Lin, C.; Wu, C.; Jia, W.; Mao, C.; Ji, V. Toughening effects of Mo and Nb addition on impact toughness and crack resistance of titanium alloys. J. Mater. Sci. Technol. 2021, 79, 147–164. [CrossRef]

15. Lei, L.; Zhao, Y.; Zhao, Q.; Wu, C.; Huang, S.; Jia, W.; Zeng, W. Impact toughness and deformation modes of Ti-6Al-4V alloy with different microstructures. Mater. Sci. Eng. A 2021, 801, 140411. [CrossRef]

16. Sun, Q.Y.; Gu, H.C. Tensile and low-cycle fatigue behavior of commercially pure titanium and Ti-5Al-2.5Sn alloy at 293 and 77 K. Mater. Sci. Eng. A 2001, A316, 80–86. [CrossRef]

17. Im, Y.-D.; Lee, Y.-K. Effects of Mo concentration on recrystallization texture, deformation mechanism and mechanical properties of TiMo binary alloys. J. Alloys Compd. 2020, 821, 155808.

18. Chou, K.; Marquis, E.A. Oxygen effects on and phase transformations in a metastable TiNb alloy. Acta Mater. 2019, 181, 367–376. [CrossRef]
19. Pang, E.L.; Pickering, E.J.; Baik, S.I.; Seidman, D.N.; Jones, N.G. The effect of zirconium on the omega phase in Ti-24Nb-[08]Zr (at.%) alloys. *Acta Mater.* 2018, 153, 62–70. [CrossRef]

20. Huang, S.; Zhao, Q.; Lin, C.; Wu, C.; Zhao, Y.; Jia, W.; Mao, C. Effects of oxygen content on Charpy impact properties and crack resistance of titanium alloys. *Mater. Sci. Eng. A* 2021, 818, 141394. [CrossRef]

21. Chen, F.; Gu, Y.; Xu, G.; Cui, Y.; Chang, H.; Zhou, L. Improved fracture toughness by microalloying of Fe in Ti-6Al-4V. *Mater. Des.* 2020, 185, 108251. [CrossRef]

22. Zhu, C.; Peng, G.; Lin, Y.C.; Zhang, X.-Y.; Liu, C.; Zhou, K. Effects of Mo and Cr contents on microstructures and mechanical properties of near-Ti alloy. *Mater. Des.* 2021, 205, 110882. [CrossRef]

23. Hao, F.; Xiao, J.; Feng, Y.; Wang, Y.; Ju, J.; Du, Y.; Wang, K.; Xue, L.; Nie, Z.; Tan, C. Tensile deformation behavior of a near-titanium alloy Ti-6Al-2Zr-1Mo-1V under a wide temperature range. *J. Mater. Res. Technol.* 2020, 9, 2818–2831. [CrossRef]

24. Li, W.; Yamasaki, S.; Mutsuhara, M.; Nakashima, H. In situ EBSD study of deformation behavior of primary phase in a bimodal Ti-6Al-4V alloy during uniaxial tensile tests. *Mater. Charact.* 2020, 163, 110282. [CrossRef]

25. Wang, Q.; Ren, J.Q.; Zhang, B.B.; Xin, C.; Wu, Y.K.; Zhang, L. Simultaneously improved strength and elongation at cryogenic temperature in Ti5Al1V1Sc1Zr0.8Mo alloy with a bimodal structure. *Mater. Sci. Eng. A* 2021, 824, 141792. [CrossRef]

26. Sofinowski, K.; Smid, M.; van Petegem, S.; Rahimi, S.; Connollely, T.; van Swygenhoven, H. In situ characterization of work hardening and springback in grade 2–titanium under tensile load. *Acta Mater.* 2019, 181, 87–98. [CrossRef]

27. Bosh, N.; Mumumliler, C.; Mozaffari-Jovein, H. Deformation twinning in cp-Ti and its effect on fatigue cracking. *Mater. Charact.* 2019, 155, 15–25. [CrossRef]

28. Bennassoud, F.; Cheikh, M.; Velay, V.; Vidal, V.; Matsumoto, H. Role of grain size and crystallographic texture on tensile behavior induced by sliding mechanism in Ti-6Al-4V alloy. *Mater. Sci. Eng. A* 2020, 774, 138835. [CrossRef]

29. Li, H.; Boehler, C.J.; Bieler, T.R.; Crimp, M.A. Analysis of slip activity and heterogeneous deformation in tension and tension-creep of Ti-5Al-2.5Sn (wt %) Using in-situ SEM experiments. *Philos. Mag.* 2012, 92, 2923–2946. [CrossRef]

30. Boyer, R.R. An overview on the use of titanium in the aerospace industry. In *Proceedings of the Fifth IUMRS International Conference on Advanced Materials (ICAM ’99)*, 9–13 October 2001; Elsevier: Zurich, Switzerland, 1996; pp. 103–114.

31. Li, X.; Duan, Y.L.; Xu, G.F.; Peng, X.Y.; Dai, C.; Zhang, L.G.; Li, Z. EBSD characterization of twinning in cold-rolled CP-Ti. *Mater. Charact.* 2013, 84, 41–47. [CrossRef]

32. Shi, X.; Cao, Z.; Fan, Z.; Qiao, J. Texture Evolution Behavior and Its Triggered Mechanical Anisotropy of CP Ti During Severe Cold Rolling and Subsequent Annealing. *Acta Metall. Sin.* 2020, 33, 1271–1282. [CrossRef]

33. Min, K.K.; Young, K.H.; Miyazaki, S. Effect of Zr Content on Phase Stability, Deformation Behavior, and Young’s Modulus in Ti-Nb-Zr Alloys. *Mater. Sci. Eng. A* 2020, 751–758. [CrossRef]

34. Han, D.; Zhao, Y.; Zeng, W. Effect of Zr addition on the mechanical properties and superplasticity of a forged SP700 titanium alloy. *Mater. Sci. Eng. A* 2021, 84, 41–47. [CrossRef] [PubMed]

35. Lu, Z.; Zhang, X.; Ji, W.; Wei, S.; Yao, C.; Han, D. Investigation on the deformation mechanism of Ti5Al-2.5Sn ELI titanium alloy at cryogenic and room temperatures. *Mater. Sci. Eng. A* 2021, 818, 141380. [CrossRef]

36. Narayana, P.L.; Seong-Woong, K.; Jae-Keun, H.; Reddy, N.S.; Jong-Taek, Y. Tensile properties of a newly developed high-temperature titanium alloy at room temperature and 650 C. *Mater. Sci. Eng. A* 2018, 718, 287–291. [CrossRef]

37. An, X.; Xu, C.; Guo, Y. Effect of electropulsing rolling on the microstructure and properties of industrial pure titanium TA1. *Adv. Mater. Res.* 2015, 1095, 99–102. [CrossRef]

38. Palan, J.; Kubina, T.; Motyka, P. The effect of annealing on mechanical and structural properties of UFG titanium grade 2. In *Proceedings of the 4th International Conference on Recent Trends in Structural Materials*, Pilsen, Czech Republic, 9–11 November 2016; IOP Publishing: Bristol, UK, 2017; p. 012055.

39. Zhiguo, F.; Hong, J.; Xiaoqiang, S.; Jie, S.; Xiaoning, Z.; Chaoying, X. Microstructures and mechanical deformation behaviors of ultrafine-grained commercial pure (grade 3) Ti processed by two-step severe plastic deformation. *Mater. Sci. Eng. A* 2009, 527, 45–51.

40. Choi, S.-W.; Jeong, J.S.; Won, J.W.; Hong, J.K.; Choi, Y.S. Grade 4 commercially pure titanium with ultrahigh strength achieved by twinning-induced grain refinement through cryogenic deformation. *J. Mater. Sci. Technol.* 2021, 66, 193–201. [CrossRef]

41. Zhentao, Y.; Lian, Z.; Du, J.; Haicheng, G. Investigation on monotonic and cyclic stress–strain characteristics of Ti-2Al-2.5Zr alloy, Symposium CC—Light Metals. In *Proceedings of the Fifth IUMRS International Conference on Advanced Materials (ICAM ‘99)*, Beijing, China, 13–18 June 1999; Elsevier: Zurich, Switzerland, 2000; pp. 192–198.

42. Xiu-Yang, F.; Jian-Xun, Z. Microstructural evolution and mechanical properties in laser beam welds of Ti-2Al-1.5Mn titanium alloy with transversal pre-extrusion load. *Int. J. Adv. Manuf. Technol.* 2016, 85, 337–343.

43. Pyshmintsev, I.Y.; Kosmatskii, Y.I.; Gornostaeva, E.A.; Illarionov, A.G.; Vodolazskii, F.V.; Radaev, P.S.; Karabanalov, M.S. Structure, Phase Composition and Mechanical Properties of Hot-Extruded Ti3Al2.5V Pipe After Vacuum Annealing. *Metallurgist* 2019, 63, 751–758. [CrossRef]

44. Gaur, R.; Gupta, R.K.; AnilKumar, V.; Banwait, S.S. Effect of cold rolling and heat treatment on microstructure and mechanical properties of Ti-4Al-1Mn titanium alloy. *J. Mater. Eng. Perform.* 2018, 27, 3217–3233. [CrossRef]

45. Wang, H.; Sun, Q.Y.; Xiao, L.; Sun, J.; Ge, P. Low-cycle fatigue behavior and deformation substructure of Ti-2Al-2.5Zr alloy at 298 and 673 K. *Mater. Sci. Eng. A* 2010, 527, 3493–3500. [CrossRef]
46. Nayan, N.; Singh, G.; Prabhu, T.A.; Murty, S.V.S.N.; Ramamurty, U. Cryogenic Mechanical Properties of Warm Multi-Pass Caliber-Rolled Fine-Grained Titanium Alloys: Ti-6Al-4V (Normal and ELI Grades) and VT14. *Metall. Mater. Trans. A* 2018, 49, 128–146. [CrossRef]

47. Huang, S.; Zhao, Q.; Lin, C.; Wu, C.; Zhao, Y.; Jia, W.; Mao, C. In-situ investigation of tensile behaviors of Ti6Al alloy with extra low interstitial. *Mater. Sci. Eng. A* 2021, 809, 140958. [CrossRef]

48. Liu, C.; Wang, X.; Zhou, G.; Li, F.; Zhang, S.; Zhang, H.; Chen, L.; Liu, H. Dislocation-controlled low-temperature superplastic deformation of Ti-6Al-4V alloy. *Front. Mater.* 2020, 7, 60692. [CrossRef]

49. Lin, C.; Yin, G.; Zhao, Y.; Wang, J. Analysis of the effect of alloy elements on allotropic transformation in titanium alloys with the use of cohesive energy. *Comput. Mater. Sci.* 2016, 111, 41–46. [CrossRef]

50. Lin, C.; Huang, S.; Yin, G.; Zhang, A.; Zhao, Z.; Zhao, Y. A simple model to ascertain the initial formation concentration of athermal phase in titanium alloys. *Comput. Mater. Sci.* 2016, 123, 263–267. [CrossRef]

51. Lin, C.; Yin, G.; Zhao, Y. Calculation of the cohesive energy of solids with the use of valence electron structure parameters. *Comput. Mater. Sci.* 2015, 101, 168–174. [CrossRef]

52. Lin, C.; Yin, G.; Zhang, A.; Zhao, Y.; Li, Q. Simple models to account for the formation and decomposition of athermal phase in titanium alloys. *Scr. Mater.* 2016, 117, 28–31. [CrossRef]

53. Ostapovets, A.; Serra, A. Slip dislocation and twin nucleation mechanisms in hcp metals. *J. Mater. Sci.* 2017, 52, 533–540. [CrossRef]

54. Yumeng, L. Study on multi-steps dynamic compression deformation behavior of Ti-5.5Al alloy. In Proceedings of the 5th Annual International Workshop on Materials Science and Engineering, Changsha, China, 17–18 May 2019; IOP Publishing: Bristol, UK, 2019; p. 012017.

55. Bridier, F.; Villechaise, P.; Mendez, J. Analysis of the different slip systems activated by tension in a α/β titanium alloy in relation with local crystallographic orientation. *Acta Mater.* 2005, 53, 555–567. [CrossRef]

56. Shi, J.; Guo, Z.; Sui, M. Slip system determination of dislocations in α-Ti during in situ tensile deformation. *Acta Metall. Sin.* 2016, 52, 71–77.

57. Anne, B.R.; Okuyama, Y.; Morikawa, T.; Tanaka, M. Activated slip systems in bimodal Ti6Al4V plastically deformed at low and moderately high temperatures. *Mater. Sci. Eng. A* 2020, 798, 140211. [CrossRef]

58. Hasija, V.; Ghosh, S.; Mills, M.J.; Joseph, D.S. Deformation and creep modeling in polycrystalline Ti-6Al alloys. *Adv. Eng. Mater.* 2003, 51, 4533–4549. [CrossRef]

59. Li, Y.; Chen, J.; Wen, H.; Li, Z. Modelling of texture evolution for materials of hexagonal symmetry-II. *Acta Mater.* 2006, 54, 63, 737–740. [CrossRef]

60. Cao, F.; Zhang, T.; Ryder, M.A.; Lados, D.A. A Review of the Fatigue Properties of Additively Manufactured Ti-6Al-4V. *JOM* 2018, 70, 349–357. [CrossRef]

61. Beyerlein, I.J.; Demkowicz, M.J.; Misra, A.; Uberuaga, B.P. Defect-interface interactions. *Prog. Mater. Sci.* 2015, 74, 125–210. [CrossRef]

62. Echlin, M.P.; Stinville, J.C.; Miller, V.M.; Lenthe, W.C.; Pollock, T.M. Incipient slip and long range plastic strain localization in microtextured Ti-6Al-4V titanium. *Acta Mater.* 2016, 114, 164–175. [CrossRef]

63. Hemery, S.; Thomas, C.; Villechaise, P. Combination of in-situ SEM tensile test and FFT-based crystal elasticity simulations of Ti-6Al-4V alloy for an improved description of the onset of plastic slip. *Mech. Mater.* 2017, 109, 1–10. [CrossRef]

64. Hemery, S.; Nait-Ali, A.; Villechaise, P. Combination of in-situ SEM tensile test and FFT-based crystal elasticity simulations of Ti-6Al-4V alloy for an improved description of the onset of plastic slip. *Mech. Mater.* 2017, 109, 1–10. [CrossRef]

65. Hutchinson, W.B.; Barnett, M.R. Effective values of critical resolved shear stress for slip in polycrystalline magnesium and other hcp metals. *Scr. Mater.* 2010, 63, 737–740. [CrossRef]

66. Phillipe, M.J.; Serghat, M.; van Houtte, P.; Esling, C. Modelling of texture evolution for materials of hexagonal symmetry-II. Application to zirconium and titanium or near alloys. *Acta Metall. Mater.* 1995, 43, 1619–1630. [CrossRef]

67. Dick, T.; Cailletaud, G. Fretting modelling with a crystal plasticity model of Ti6Al4V. *Comput. Mater. Sci.* 2006, 38, 113–125. [CrossRef]

68. Knezovic, M.; Lebensohn, R.A.; Casacu, O.; Revil-Baudard, B.; Proust, G.; Vogel, S.C.; Nixon, M.E. Modeling bending of α-titanium with embedded polycrystal plasticity in implicit finite elements. *Mater. Sci. Eng. A* 2013, 564, 116–126. [CrossRef]

69. Jicheng, G.; Wilkinson, A.J. Anisotropy in the plastic flow properties of single-crystal titanium determined from micro-cantilever beams. *Acta Mater.* 2009, 57, 5693–5705.

70. Williams, J.C.; Baggerly, R.G.; Paton, N.E. Deformation behavior of HCP Ti-Al alloy single crystals. *Metall. Mater. Trans. A* 2002, 33, 837–850. [CrossRef]

71. Venkataramani, G.; Deka, D.; Ghosh, S.; Nordholt, J.B. Crystal plasticity based Fe model for understanding microstructural effects on creep and dwell fatigue in Ti-6242, Journal of Engineering Materials and Technology. *Trans. ASME* 2006, 128, 356–365.

72. Hemery, S.; Villechaise, P. Comparison of slip system activation in Ti-6Al-2Sn-4Zr-2Mo and Ti-6Al-2Sn-4Zr-6Mo under tensile, fatigue and dwell-fatigue loadings. *Mater. Sci. Eng. A* 2017, 697, 177–183. [CrossRef]

73. Jones, I.P.; Hutchinson, W.B. Stress-state dependence of slip in titanium-6Al-4V and other H.C.P. metals. *Acta Metall.* 1981, 29, 951–968. [CrossRef]
74. Bridier, F.; McDowell, D.L.; Villechaise, P.; Mendez, J. Crystal plasticity modeling of slip activity in Ti-6Al-4V under high cycle fatigue loading. *Int. J. Plast.* 2009, 25, 1066–1082. [CrossRef]

75. Hemery, S.; Villechaise, P. On the influence of ageing on the onset of plastic slip in Ti-6Al-4V at room temperature: Insight on dwell fatigue behavior. *Scr. Mater.* 2017, 130, 157–160. [CrossRef]

76. Chan, K.S. A micromechanical analysis of the yielding behavior of individual Widmanstatten colonies of an α+β titanium alloy. *Metall. Mater. Trans. A* 2004, 35 A, 3409–3422. [CrossRef]

77. Li, H.; Mason, D.E.; Bieler, T.R.; Boehlert, C.J.; Crimp, M.A. Methodology for estimating the critical resolved shear stress ratios of α-phase Ti using EBSD-based trace analysis. *Acta Mater.* 2013, 61, 7555–7567. [CrossRef]

78. Kwasniak, P.; Cloutet, E. Influence of simple metals on the stability of a basal screw dislocations in hexagonal titanium alloys. *Acta Mater.* 2019, 180, 42–50. [CrossRef]

79. Kasemer, M.; Echlin, M.P.; Steinville, J.C.; Pollock, T.M.; Dawson, P. On slip initiation in equiaxed Ti-6Al-4V. *Acta Mater.* 2017, 136, 288–302. [CrossRef]

80. Kawano, Y.; Mayama, T.; Kondou, R.; Ohashi, T. Crystal plasticity analysis of change in active slip systems of α-phase of Ti-6Al-4V alloy under cyclic loading. In Proceedings of the 13th Asia-Pacific Symposium on Engineering Plasticity and its Applications, AEPa 2016, Hiroshima, Japan, 4–8 December 2016; Trans Tech Publications Ltd.: Hiroshima, Japan, 2017; pp. 183–188.

81. Hemery, S.; van Truong, D.; Signor, L.; Villechaise, P. Influence of microtexture on early plastic slip activity in Ti-6Al-4V polycrystals. *Metall. Mater. Trans. A* 2018, 49, 2048–2056. [CrossRef]

82. Basu, I.; Fidder, H.; Ocelik, V.; de Hosson, J.T.M. Local stress states and microstructural damage response associated with deformation twins in hexagonal close packed metals. *Crystals* 2018, 8, 1. [CrossRef]

83. Joseph, S.; Bantouanas, I.; Lindley, T.C.; Dye, D. Slip transfer and deformation structures resulting from the low cycle fatigue of near-alpha titanium alloy Ti-6242Si. *Int. J. Plast.* 2018, 100, 90–103. [CrossRef]

84. Livingston, J.D.; Chalmers, B. Multiple slip in bicrystal deformation. *Acta Metall.* 1957, 5, 322–327. [CrossRef]

85. Luster, J.; Morris, M.A. Compatibility of deformation in two-phase Ti-Al alloys: Dependence on microstructure and orientation relationships. *Metall. Mater. Trans. A* 1995, 26 A, 1745–1756. [CrossRef]

86. Bieler, T.R.; Eisenlohr, P.; Zhang, C.; Phukan, H.J.; Crimp, M.A. Grain boundaries and interfaces in slip transfer. *Curr. Opin. Solid State Mater. Sci.* 2014, 18, 212–226. [CrossRef]

87. Bieler, T.R.; Eisenlohr, P.; Zhang, C.; Phukan, H.J.; Crimp, M.A. Grain boundaries and interfaces in slip transfer. *Curr. Opin. Solid State Mater. Sci.* 2018, 120, 435–442. [CrossRef]

88. Hemery, S.; Nizou, P.; Villechaise, P. In situ SEM investigation of slip transfer in Ti-6Al-4V: Effect of applied stress. *Mater. Sci. Eng. A* 2018, 709, 277–284. [CrossRef]

89. Hemery, S.; Nait-Ali, A.; Gueguen, M.; Wendorf, J.; Polonsky, A.T.; Echlin, M.P.; Steinville, J.C.; Pollock, T.M.; Villechaise, P. A 3D analysis of the onset of slip activity in relation to the degree of micro-texture in Ti6Al4V. *Acta Mater.* 2019, 181, 36–48. [CrossRef]

90. Jiao, B.; Zhao, Q.; Zhao, Y.; Li, L.; Hu, Z.; Gao, X.; Zhang, W.; Li, J. The relationship between slip behavior and dislocation arrangement for large-size Mo-3Nb single crystal at room temperature. *J. Mater. Sci. Technol.* 2021, 92, 208–213. [CrossRef]

91. Jin, G.; Yang, D.; Wang, Q.; Ren, J.; Wang, Y.; Xin, C.; Xiao, L.; Yang, D. Deformation and fracture mechanisms of gradient nanograined pure Ti produced by a surface rolling treatment. *Mater. Sci. Eng. A* 2019, 754, 121–128. [CrossRef]
103. Koyama, M.; Yamanouchi, K.; Qinghua, W.; Shien, R.; Tanaka, Y.; Hamano, Y.; Yamasaki, S.; Mitsuhara, M.; Ohkubo, M.; Noguchi, H.; et al. Multiscale in situ deformation experiments: A sequential process from strain localization to failure in a laminated Ti-6Al-4V alloy. *Mater. Charact.* 2017, 128, 217–225. [CrossRef]

104. Bieler, T.R.; Eisenlohr, P.; Roters, F.; Kumar, D.; Mason, D.E.; Crimp, M.A.; Raabe, D. The role of heterogeneous deformation on damage nucleation at grain boundaries in single phase metals. *Int. J. Plast.* 2009, 25, 1655–1683.

105. Koyama, M.; Sawaguchi, T.; Ogawa, K.; Kikuchi, T.; Murakami, M. Continuous transition of deformation modes in Fe-30Mn-5Si-1Al alloy. *Mater. Trans.* 2010, 51, 1194–1199. [CrossRef]

106. Tan, C.; Sun, Q.; Xiao, L.; Zhao, Y.; Sun, J. Characterization of deformation in primary α phase and crack initiation and propagation of TC21 alloy using in-situ SEM experiments. *Mater. Sci. Eng. A* 2018, 725, 33–42. [CrossRef]

107. Liu, X.; Qian, Y.; Fan, Q.; Zhou, Y.; Zhu, X.; Wang, D. Plastic deformation mode and α/β slip transfer of Ti5Al-2.5Cr-0.5Fe-4.5Mo-1Sn-2Zr-3Zn titanium alloy at room temperature. *J. Alloys Compd.* 2020, 826, 154209. [CrossRef]