

Article

The Action Mechanism of Rolling Texture on the Anisotropic Behavior of a Pure Titanium Plate

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Abstract: This work combined theoretical calculation with experimental characterization to methodically study the anisotropy mechanism and evolution of the plastic behavior of pure titanium. Initially, a constant-strain uniaxial tensile test was used to measure the anisotropy of the yield behavior along the rolling direction (RD) and transverse direction (TD). Subsequently, the information of crystal orientation both before and after deformation was statistically characterized using electron backscatter diffraction (EBSD). Ultimately, the main deformation mechanism was determined by combining Schmid law with an analysis of the variation of SF values of each deformation mode with the angular relationship between the loading axis and the grain's c-axis. The findings demonstrate that, for each slip system, the variation trend and value of the SF are influenced by the angle formed by the loading axis and the grain's c- and a-axes. The primary result of dislocation slip activation is the change of the tilt angle of the grain c-axis from ND to TD, but this has little effect on the tilt angle of the grain c-axis from ND to RD. Prismatic $\langle a \rangle$ slip dominates the tensile deformation along the RD. Pyramidal $\langle a \rangle$ slip and pyramidal $\langle c+a \rangle$ slip will be activated during the subsequent hardening, whereas basal $\langle a \rangle$ slip is difficult to activate. The prismatic $\langle a \rangle$ slip in the soft-oriented grain will be preferentially activated during the tensile deformation along the TD, and the prismatic $\langle a \rangle$ slip and pyramidal $\langle a \rangle$ slip will become the dominant deformation modes during the subsequent hardening. Some soft-oriented grains could activate basal $\langle a \rangle$ slip and pyramidal $\langle c+a \rangle$ slip, but dislocation slip is restricted and coordinated by $\{10\text{--}12\}$ ET.

Keywords: texture; titanium plate; anisotropy; slip and twinning; Schmid law

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1. Introduction

Titanium metal is a highly functionalized and structured material with exceptional qualities, including low thermal expansion, biocompatibility, high specific strength, and resistance to corrosion [1–4]. However, the material's own structured characteristics and the crystal's preferential orientation during the subsequent forming process can easily lead to significant anisotropy in the macro-mesoscopic deformation behavior of the material, which in turn leads to the uncontrollable performance of titanium products and the expensive processing cost [5–7]. The complex and varied plastic deformation modes of this type of metal are caused by the poor symmetry of the hexagonal structure, and the CRSS needed to activate each deformation mode differs greatly due to the different degrees of atomic density [8–11]. The easiest to activate in titanium at room temperature is prismatic $\langle a \rangle$ slip $\{1\ 0\ -1\ 0\} \langle 1\ 1\ -2\ 0 \rangle$, which is followed by basal slip $\{0\ 0\ 0\ 1\} \langle 1\ 1\ -2\ 0 \rangle$ and pyramidal slip $\{1\ 0\ -1\ 1\} \langle 1\ 1\ -2\ 0 \rangle$, in that order. It is evident that the aforementioned three slip modes are limited to producing deformation along the grain a-axis [12–14]. Although the pyramidal $\langle c+a \rangle$ slip can coordinate the deformation in the c-axis direction, the nucleation of the slip system due to the high CRSS often requires strain accumulation to

a certain extent or significant stress concentration locally [15–18]. Twins become one of the key mechanisms for coordinating plastic deformation in low-stacking-fault-energy metals when the activation of dislocation slip is restricted and unable to satisfy the demands of plastic deformation [19–22]. Although not a thermally activated phenomenon, the critically resolved shear stress and shear strain also have an impact on the activability of the twin behavior [23–26]. Consequently, the coupling effect between the deformation modes and the differences between CRSS has been found to lead to the complex and unpredictable macro-mesoscopic deformation behavior of titanium [27,28].

Different degrees of crystal preference orientation in the material can readily result from the rotation of the grain's internal orientation caused by the activation of dislocation slip (twin) at the grain scale [29–31]. Specific crystal orientations lead to significant anisotropy in the activation characteristics of each deformation mode, and the difference in the activated deformation modes in turn leads to new texture redirection [32–35]. For example, the basal bimodal texture, in which the tilt angle of the grain c-axis from ND to TD is about 20–50°, is the most common texture type in rolled titanium plate under the traditional cold-rolling process [36–39]. Under this type of texture, the prismatic $\langle a \rangle$ slip has an absolute initial advantage when deformed in the RD-TD plane, but the activation of prismatic $\langle a \rangle$ slip may lead to the deterioration of the crystal orientation of other slip systems when they are first activated [40,41]. Moreover, the higher strain hardening phenomenon leads to an increase in the difficulty of the $\langle a \rangle$ type slip to adapt to the deformation in the direction of plate thickness, which in turn reduces the thinning ability in the forming process, such as during sheet strip rolling and stamping [42,43]. A large number of studies have shown that twins are not the cause of the initial yield of titanium, but the texture redirection and Hall–Petch effect induced by twins still have an important impact on the subsequent strain hardening behavior [44–48]. Therefore, a deeper comprehension of the connection between the crystal orientation and the activability of each deformation mode is essential for the reasonable formulation of processes.

In conclusion, the foundation of rational process formulation and material property control is to clarify the relationship between process, texture, deformation mode, and mechanical behavior. In order to systematically reveal the variation law of the activability of each deformation mode with loading mode and crystal orientation, this research studies the mechanism of texture on the anisotropic behavior of pure titanium plates. First, the textured titanium plate was tested with a constant-strain uniaxial tensile test in different directions, and the evolution of the crystal orientation information with strain during the deformation process was examined using EBSD. Then, based on Schmid law, the variation of the activation of each deformation mode with the angular relationship between the c-axis and the loading axis of the grain under different stress states was studied. Finally, considering the proportional relationship between the CRSS of each slip system, the dominant deformation mode was identified and predicted. This work can provide significant theoretical guidance for the process formulation of titanium plastic forming.

2. Materials and Methods

The experimental material was annealed TA2 pure titanium sheet with a thickness of 2 mm, and the composition is shown in Table 1. The tensile specimen was cut by WEDM, and the dimensions are shown in Figure 1. The uniaxial tensile tests were conducted along RD and TD at room temperature using the Zwickell 100 high-precision electronic universal materials testing machine. The tensile strains were set to 0.02, 0.04, 0.06, 0.08, and 0.12, and the average strain rate was 0.002 s^{-1} . Each set of specimens was repeated two times to ensure good reproducibility. For ease of description, RD (0.04) was defined to mean a tensile of strain 0.04 along RD. The microstructure and crystal orientation both before and after deformation were examined using optical microscopy and EBSD, with the observation position being the RD-TD plate plane. The surface of the deformed specimen was first mechanically ground with silicon carbide sandpaper from P1000 to P2500 and then mechanically polished with a silica suspension with a particle size of $0.05 \text{ }\mu\text{m}$. The

corrosion reagent of the metallographic sample was 1.5 mL HF + 15 mL HNO₃ + 85 mL H₂O, and the corrosion time was 60 s. The 10% perchloric acid + 90% methanol solution was used to electropolish the EBSD specimen for 5 s at 35 V and 5 °C. The microscopic texture detection instrument was JSM-7001F (JEOL Ltd., Tokyo, Japan), and the grain orientation information was constructed with a scanning step size of 0.5 µm. OIM software (EDAX Inc., Mahwah, NJ, USA) was utilized to evaluate the EBSD data, and the crystal information was processed and analyzed using the data with a confidence interval greater than 0.1.

Table 1. Chemical composition of pure titanium plate.

Elements	O	C	N	H	Fe	Ti
Composition (ppm)	820	52	40	3	220	Balance

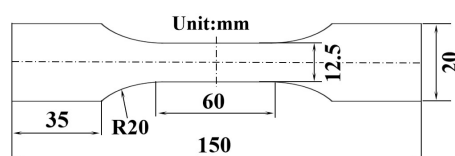


Figure 1. Tensile specimen size.

Figure 2 shows the microstructure of the initial plate. The colors of the grains in the orientation map (OM) correspond to their crystallographic orientation with respect to ND: Red grains indicate $\langle 0\ 0\ 0\ 1 \rangle // \text{ND}$, green grains indicate $\langle 2\ -1\ -1\ 0 \rangle // \text{ND}$, and blue grains represent $\langle 1\ 0\ -1\ 0 \rangle // \text{ND}$. The OM (Figure 2a) shows that the initial structure is basically equiaxed, and the average intercept grain size is about 20 µm. To display the type and quantity of twin boundaries, the grain boundaries of the $\{1\ 0\ -1\ 2\}$ extension twin (ET) and $\{1\ 1\ -2\ 2\}$ contraction twin (CT) are superimposed on a grayscale map. Combined with the grayscale map (Figure 2b), it can be found that there are no twins inside the grain, and the color distribution is relatively uniform, indicating that the orientation difference inside the initial grain is small, and the dislocation density is low. Based on the IPF (Figure 2c), it is possible to determine that the initial texture type is the basal bimodal texture, in which the c-axis of the grain is tilted about $35 \pm 5^\circ$ from ND to TD. The angles between the $\langle 0\ 0\ 0\ 1 \rangle$ axis of the grain and RD, TD, and ND are concentrated at $75\sim 90^\circ$, $30\sim 65^\circ$, and $20\sim 50^\circ$. The crystal orientation data of the textured titanium plate obtained by EBSD can be used to calculate the Schmid factor (SF) of each deformation mechanism under different loading conditions since the activability of plastic deformation mechanisms like slip or twin will be affected by the angle between the loading direction and the grain orientation. In order to facilitate the conversion of the coordinate system and the prediction of the plastic deformation mode in combination with the SF, the crystal coordinate system $O\text{-}X_2Y_2Z_2$ ($X_2 = [1\ 0\ -1\ 0]$, $Y_2 = [-1\ 2\ -1\ 0]$, and $Z_2 = [0\ 0\ 0\ 1]$) and the sample coordinate system $O\text{-}X_1Y_1Z_1$ (RD = $[1\ 0\ 0]$, TD = $[0\ 1\ 0]$, and ND = $[0\ 0\ 1]$) were established.

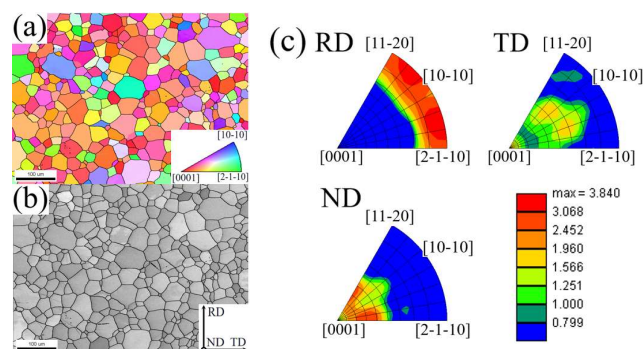


Figure 2. Initial crystal orientation information of pure titanium: (a) orientation map (OM) indexed with inverse pole figure (IPF); (b) grayscale map; (c) IPF.

3. Results

3.1. Macroscopic Stress–Strain Curves

The metallographic structure and stress–strain curves during tensile along RD and TD are displayed in Figure 3. It is reasonable to believe that the macro- and mesoscopic characteristics under a certain strain can qualitatively reflect the deformation characteristics of the material during continuous deformation under this strain, as can be seen from Figure 3, where there is good continuity between the stress–strain curves under different strains. The higher initial yield strength when tensile along TD and the higher strain hardening rate during tensile along RD indicate that the basal bimodal texture clearly causes anisotropy in both the initial yield and subsequent strain hardening of the pure titanium plate. According to the metallographic structure, twins seldom occur during RD tensile; however, with TD tensile, twins start to form when the strain exceeds 0.04. Thus, dislocation slip, not twin, is the predominant deformation mechanism of the initial yield for a pure titanium plate with a basal bimodal structure, as may be inferred from the aforementioned phenomena.

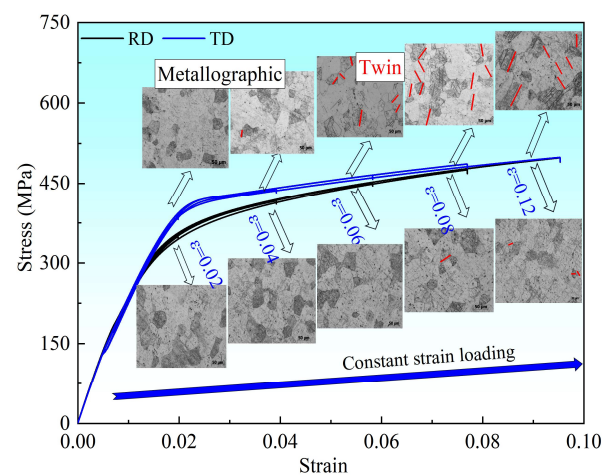


Figure 3. Stress–strain curves and metallographic structure during tensile along RD and TD.

3.2. Microstructure Evolution

Figure 4 shows the evolution of the mesoscopic deformation characteristics of the RD tensile specimen with strain. The $85^\circ \{1\ 0\ -1\ 2\}$ ET is represented by the red line in the grayscale map, while the $64^\circ \{1\ 1\ -2\ 2\}$ CT is shown by the blue line. There is no twin and a rather consistent color distribution inside the grain at RD (0.04), as shown by the OM and grayscale map in Figure 4a. As the strain approaches 0.08, Figure 4b,c demonstrate the presence of a notable color gradient inside the grain, and in certain grains, $64^\circ \{1\ 1\ -2\ 2\}$ CT is visible. Figure 5 calculates the twin area fractions and average KAM values under different loading conditions. The uneven plastic deformation and dislocation accumulation between and within the grains are often linked to changes in local orientation. Figure 5a illustrates this phenomenon, with the average KAM value of the RD tensile increasing rapidly from 0.317 at RD (0.04) to 0.612 at RD (0.12). The c-axis orientation of certain grains is displayed in the upper right corner of the grayscale map to facilitate the interpretation of the twin activation law. When combined with the IPF color card, it becomes even clearer that the majority of grains' $\langle 0\ 0\ 0\ 1 \rangle$ axis is primarily tilted from ND to TD, while the $\langle 0\ 0\ 0\ 1 \rangle$ axis is only slightly tilted from ND to RD. As a result, the tensile stress component perpendicular to the grains' $\langle 0\ 0\ 0\ 1 \rangle$ axis when tensile along RD is favorable for the nucleation of $\{1\ 1\ -2\ 2\}$ CT. Figure 5b illustrates that during RD tensile, the twin area fraction of $\{1\ 1\ -2\ 2\}$ CT is less than 4% and that the rate of increase in twin area fraction drastically drops after the strain approaches 0.08. The aforementioned phenomena suggest that dislocation slip during tensile along RD is the

predominant mechanism of plastic deformation, while twins have essentially little effect on the total amount of plastic deformation.

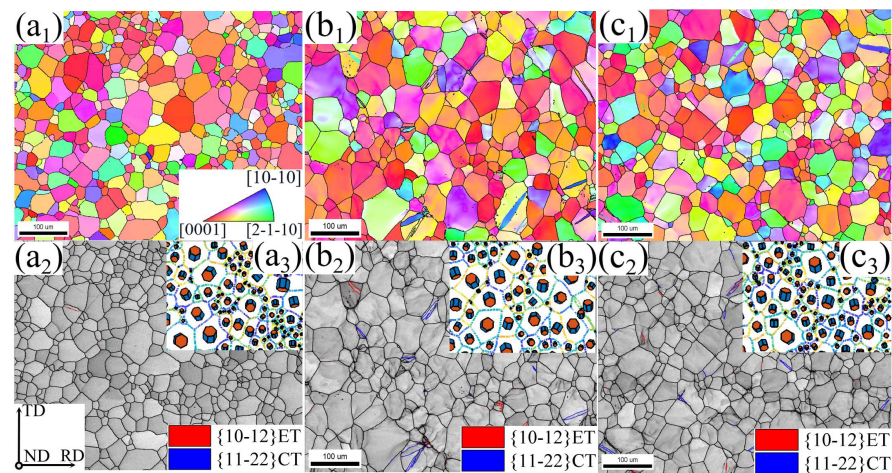


Figure 4. Evolution of the mesoscopic deformation characteristics of the RD tensile specimen with strain: (a₁) RD (0.04)-OM; (a₂) RD (0.04)-grayscale map; (a₃) RD (0.04)-grain orientation; (b₁) RD (0.08)-OM; (b₂) RD (0.08)-grayscale map; (b₃) RD (0.08)-grain orientation; (c₁) RD (0.12)-OM; (c₂) RD (0.12)-grayscale map; (c₃) RD (0.12)-grain orientation.

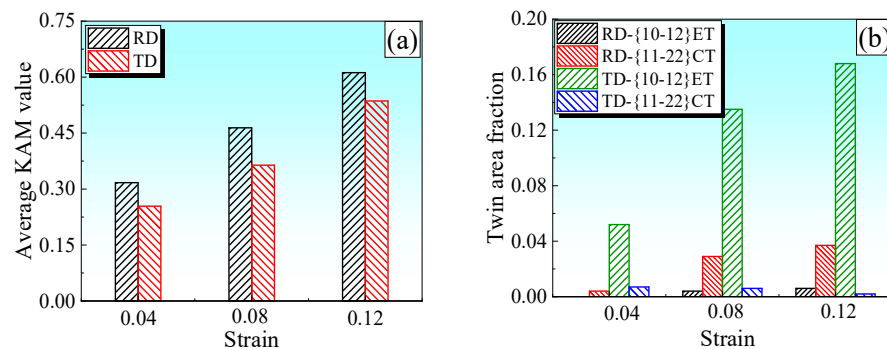


Figure 5. The average KAM value and twin area fraction under different loading conditions: (a) average KAM value; (b) twin area fraction.

Figure 6 illustrates the evolution of the mesoscopic deformation characteristics of the TD tensile specimen with strain. This figure shows that the TD tensile specimen's color gradient inside its grain is significantly weaker than that of the RD tensile specimen. When viewed together with Figure 5a, it is also evident that the TD tensile specimen's KAM value is significantly smaller than that of the RD, indicating a relatively weak degree of dislocation accumulation during TD tensile. In combination with Figure 5b, it becomes evident that certain grains exhibit 85° $\{10-12\}$ ET at TD (0.04), and the twin area fraction experiences a rapid increase, rising from around 5.2% at TD (0.04) to approximately 16.8% at TD (0.12). Although the twin undoubtedly contributes significantly to the coordination of deformation during TD tensile, it should be highlighted that, at least within a specific strain range, the amount of plastic deformation caused by the twin is still much smaller than that of the dislocation. The twins are mostly nucleated in certain grains, with crystal orientation tilted from $[0001]/ND$ to $[10-10]/ND$ and $[2-1-10]/ND$, according to the IPF color card. The $\langle 0001 \rangle$ axis of most grains during TD tensile is primarily tilted by ND to TD at a specific angle, as can be seen from the crystal orientation in the upper right corner of the grayscale map in Figure 6. The TD tensile stress will produce tensile stress along the grain $\langle 0001 \rangle$ axis, which is favorable to the nucleation of $\{10-12\}$ ET. The rules may be summed up as follows when comparing the twin characteristics under different strains: (1) The primary twin layer has a very thin thickness that progressively thickens as

strain increases, forming a convex lens-like shape. (2) The twin wafer layer nucleates at the grain boundary and gradually grows into the grain, and most of the twins run through the entire grain. (3) The twin density increases significantly with the aggravation of the deformation degree. Twin density is defined as the proportion of twin grains and the number of twins in a given grain. (4) The occurrence of one or more twin layers within the grain and the entanglement between them can achieve grain refinement to a large extent. (5) It can be seen from OM that the emergence of twins realizes a rapid transformation of crystal orientation, and the redirected grains may make it easier for later dislocation slip to activate.

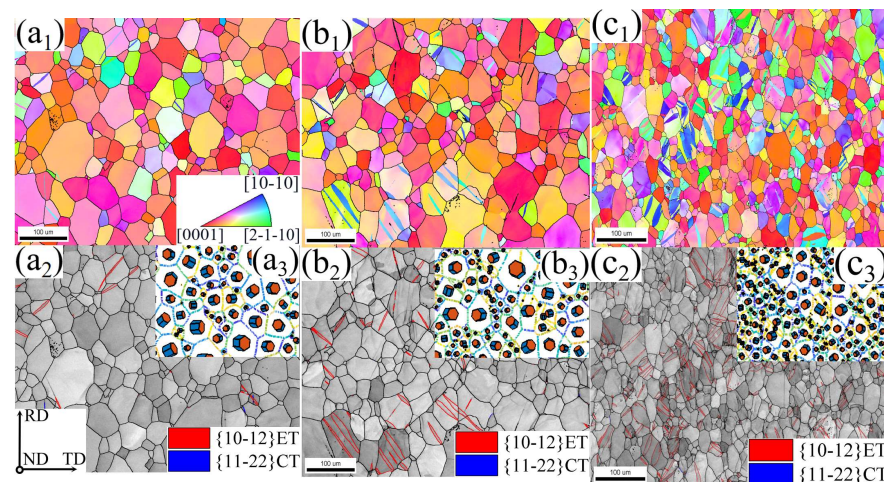


Figure 6. Evolution of the mesoscopic deformation characteristics of the TD tensile specimen with strain: (a₁) RD (0.04)-OM; (a₂) RD (0.04)-grayscale map; (a₃) RD (0.04)-grain orientation; (b₁) RD (0.08)-OM; (b₂) RD (0.08)-grayscale map; (b₃) RD (0.08)-grain orientation; (c₁) RD (0.12)-OM; (c₂) RD (0.12)-grayscale map; (c₃) RD (0.12)-grain orientation.

3.3. Texture Evolution

Figure 7 illustrates the evolution of the texture components with strain during tensile along RD. The RD-IPF shows that as strain increases, the angle between the grain $\langle 0\ 0\ 0\ 1 \rangle$ axis and RD is basically distributed between 75° – 90° . The $\langle 2\ -1\ -1\ 0 \rangle$ axis of the crystal orientation tilts gradually to the $\langle 1\ 0\ -1\ 0 \rangle$ axis, and the $\langle 1\ 0\ -1\ 0 \rangle$ //RD texture components generate after RD (0.08). The angle between the grain $\langle 0\ 0\ 0\ 1 \rangle$ axis and TD is discretely distributed at 20° – 70° , according to the TD-IPF. The most typical texture component is the $\langle 1\ 0\ -1\ 2 \rangle$ //TD, which tilts by about 45° from the $\langle 0\ 0\ 0\ 1 \rangle$ axis to the $\langle 1\ 0\ -1\ 0 \rangle$ axis. The $\langle 0\ 0\ 0\ 1 \rangle$ //TD texture component forms gradually as strain increases. The angle between the $\langle 0\ 0\ 0\ 1 \rangle$ axis and ND at RD (0.04) is mostly distributed in the range of 30° – 50° , according to the ND-IPF. As strain increases, however, the polarization phenomenon is seen, meaning that certain grains are tilted toward the $\langle 0\ 0\ 0\ 1 \rangle$ axis, while the remaining grains are tilted toward the $\langle 2\ -1\ -1\ 0 \rangle$ and $\langle 1\ 0\ -1\ 0 \rangle$ axes and even generate the $\langle 2\ -1\ -1\ 0 \rangle$ //ND texture components. In conclusion, it is evident that the activation of dislocation slip during tensile along RD primarily causes the grain c-axis' tilt angle to change from ND to TD. However, the approximately vertical angular relationship between the grain c-axis and RD is unaffected, which allows the basal bimodal texture to remain unchanged in its entirety.

Figure 8 shows the evolution of the texture components with strain during tensile along TD. From the TD-IPF, it can be found that the angle between the grain $\langle 0\ 0\ 0\ 1 \rangle$ axis and TD is discretely distributed at 40° – 70° . The $\langle 1\ 0\ -1\ 2 \rangle$ //TD texture component at about 45° position (from the $\langle 0\ 0\ 0\ 1 \rangle$ axis tilt to the $\langle 10\ -10 \rangle$ axis) gradually changes to the $\langle 1\ 0\ -1\ 1 \rangle$ //TD texture component at 60° with the increase in strain. The RD-IPF demonstrates how, as strain increases, the grain progressively deflects from the $\langle 2\ -1\ -1\ 0 \rangle$ axis to the $\langle 1\ 0\ -1\ 0 \rangle$ axis, which causes the $\langle 1\ 0\ -1\ 0 \rangle$ //RD texture components to

disappear and the $\langle 2 - 1 - 1 0 \rangle // \text{RD}$ texture components to appear. The $\langle 0 0 0 1 \rangle // \text{RD}$ texture generated by the $\{1 0 - 1 2\}$ ET-induced redirection appears when the strain reaches 0.08, but its overall strength is significantly less than that of the basal bimodal texture. The angle between the $\langle 0 0 0 1 \rangle$ axis of most grains and RD is still distributed at $75^\circ\text{--}90^\circ$, which means that the approximate vertical angular relationship between the grain's c-axis and RD does not change significantly during TD tensile, according to a comparison of the results of different strains. The angle between the $\langle 0 0 0 1 \rangle$ axis and the ND of the grain gradually decreases as strain increases, according to the ND-IPF. For example, the tilt angle from the $\langle 0 0 0 1 \rangle$ axis to $\langle 2 - 1 - 1 0 \rangle$ axis is distributed at $30^\circ\text{--}45^\circ$, $25^\circ\text{--}45^\circ$, and $0^\circ\text{--}35^\circ$ at TD (0.04), TD (0.08), and TD (0.12), respectively, and the texture components of $\langle 0 0 0 1 \rangle // \text{ND}$ gradually become obvious. In summary, it can be seen that the activation of dislocation slip during tensile along TD mainly leads to the change of the tilt angle of the grain c-axis from ND to TD but does not significantly change the approximately vertical angular relationship between the grain c-axis and RD, so the basal bimodal texture is still maintained on the whole.

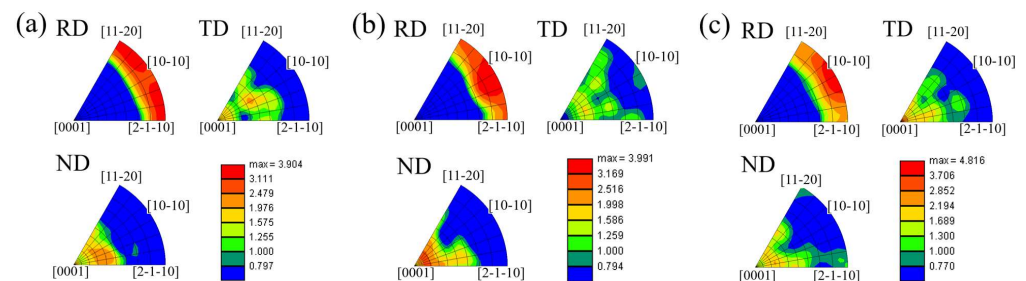


Figure 7. Evolution of texture components with strain during tensile along RD: (a) RD (0.04); (b) RD (0.08); (c) RD (0.12).

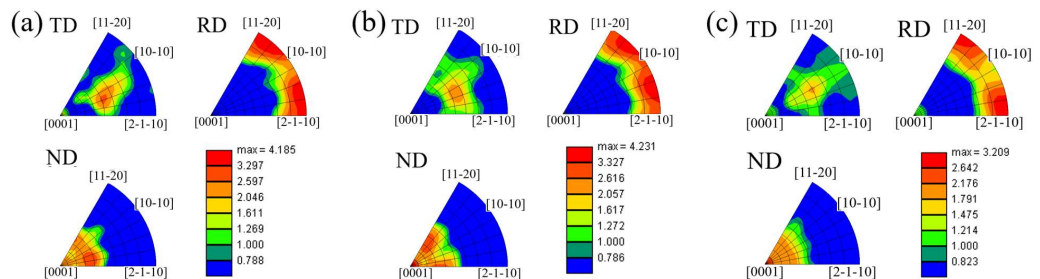


Figure 8. Evolution of texture components with strain during tensile along TD: (a) TD (0.04); (b) TD (0.08); (c) TD (0.12).

4. Discussion

4.1. Effect of Texture on the Activability of Slip Systems

Figure 9 shows the evolution of the SF distribution of each deformation mode with strain when tensile along RD. The $(0 0 0 1)$ PF is highlighted by the color card corresponding to SF, and it is easier to understand how the SF values of each deformation mode change with crystal orientation. Statistics for the grain proportion of soft orientation ψ ($\text{SF} = 0.4\text{--}0.5$), the average SF value m , and the distribution interval of SF are displayed in the histogram. The majority of the grains during RD tensile are favorable to the activation of prismatic $\langle a \rangle$ slip, pyramidal $\langle a \rangle$ slip, and pyramidal $\langle c+a \rangle$ slip but not to the activation of the basal $\langle a \rangle$ slip, as can be seen from SF distribution map (Figure 9a). To facilitate analysis, θ_a and θ_b , respectively, represent the angle of inclination of the grain c-axis from ND to RD and TD. It is evident from the $(0 0 0 1)$ PF-SF (Figure 9b) that the color gradient of the SF values for each deformation mode mostly changes with θ_a change, but θ_b change is negligible. It is evident from the $(0 0 0 1)$ PF-SF that the texture redirection at this point has no influence on the SF values of any slip system since the previous results demonstrate

that θ_a always maintains an almost vertical angular connection during the tensile along the RD. The SF distribution histogram (Figure 9c) also shows that each slip system's ψ and m values, which correspond to 90%–0.45, 5.7%–0.16, 95%–0.45, and 97%–0.47, respectively, are found to change little with strain in the prismatic $\langle a \rangle$ slip, basal $\langle a \rangle$ slip, pyramidal $\langle a \rangle$ slip, and pyramidal $\langle c+a \rangle$ slip.

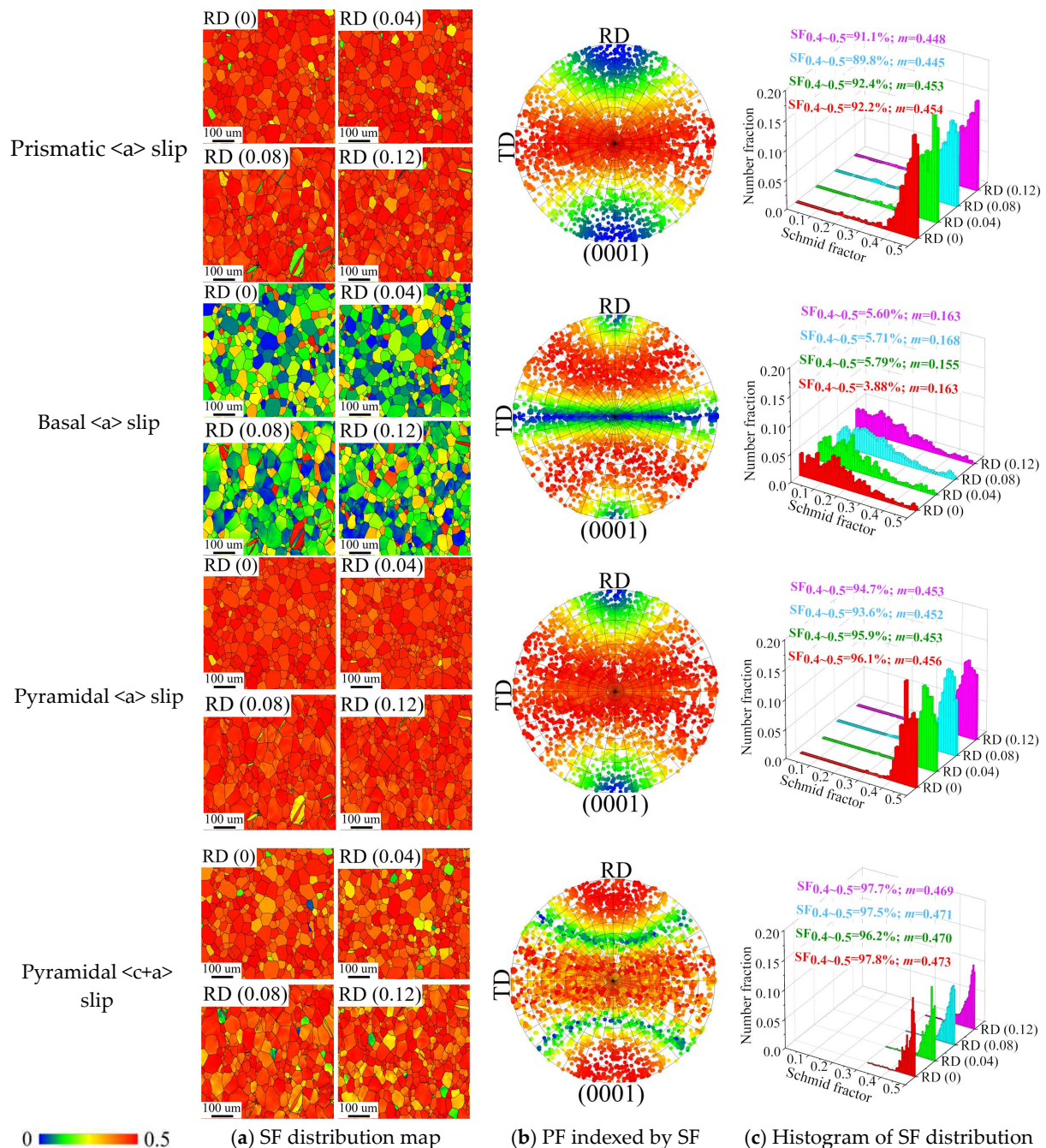


Figure 9. The evolution of the SF distribution of each deformation mode with strain when tensile along RD: (a) SF distribution map; (b) PF indexed by SF; (c) the histogram of SF distribution.

Figure 10 shows the evolution of the SF distribution of each deformation mode with strain when tensile along TD. The SF values of each slip system in the grain in the statistical region have an uneven distribution, as can be seen from SF distribution map (Figure 10a), which means that the types of slip systems activated during TD tensile are much more complex than those generated during RD tensile. This is due to the discrete distribution

of the angle between the TD tensile axis and the grain c-axis. The color gradient of the SF values of each deformation mode mainly changes with the change of the tilt angle θ_b when tensile along TD, and changes less with the tilt angle θ_a , as can be seen from the (0 0 0 1) PF-SF (Figure 10b). This suggests that the SF values of each slip system have a strong angle θ_b dependence when tensile along TD. From OM and (0 0 0 1)PF-SF, it can be seen that the matrix with {1 0 – 1 2}ET is not conducive to the activation of prismatic <a> slip, pyramidal <a> slip, and pyramidal <c+a> slip, but it is conducive to the activation of the basal <a> slip. Consequently, {1 0 – 1 2} ET becomes the primary mode of coordinated c-axis deformation in this part of the grain. Simultaneously, it is shown that the <0 0 0 1> // RD texture components created by {1 0 – 1 2} ET-induced texture redirection are favorable for prismatic <a> slip, pyramidal <a> slip, and pyramidal <c+a> slip activation but not for basal <a> slip activation. The SF distribution histogram (Figure 10c) shows that the ψ and m values of prismatic <a> slip increase slightly as strain increases, going from around 26%–0.29 at TD (0) to roughly 35%–0.32 at TD (0.12). When the strain approaches TD (0.08), for example, the ψ and m values of the basal <a> slip start to dramatically decrease, going from around 53%–0.36 at TD (0.08) to approximately 46%–0.33 at TD (0.12). When the strain is smaller than TD (0.08), the cone slip's ψ and m values increase, going from around 64%–0.38 at TD (0) to roughly 73%–0.42 at TD (0.08). It does, however, decrease to around 68%–0.4 at TD (0.12) as the strain increases. The ψ value of pyramidal <c+a> slip increases from about 71% at TD (<0.08) to 77% at TD (0.12), indicating that the crystal orientation at this point is increasingly favorable for the activation of pyramidal <c+a> slip. On the whole, the evolution of the ψ and m values of each slip system with strain is not a monotonic relationship; in addition to statistical errors, it is also influenced by the weight of the deformation modes that are active at different strain stages.

4.2. Quantitative Calculation of SF

Figure 11 is a schematic diagram of the positional relationship between the loading axis and each deformation mode. As can be seen from Figure 11, SF is affected by the external force F and the positional relationship between the grain c-axis and the grain a-axis (θ , α), where $\theta \in (0 \sim 90^\circ)$, and $\alpha \in (0 \sim 30^\circ)$. The SF values of α and $60^\circ \alpha$ are equal, as can be seen from the symmetry of the hexagonal lattice structure. As a result, $F = (\sin\theta \sin\alpha \sin\theta \cos\alpha \cos\theta)$ is the vector form of the external force F represented by (θ , α).

For a particular slip system (twin system), SF is the product of the cosine of the external force F and the angle between the normal direction n and the slip direction d of the slip system (twin system), respectively:

$$SF = [F][n] \bullet [F][d] \quad (1)$$

From the above experimental results, it can be determined that the basal <a> slip system {0 0 0 1} <1 1 – 2 0>, prismatic <a> slip {1 0 – 1 0} <1 1 – 2 0>, pyramidal <a> slip {1 0 – 1 1} <1 1 – 2 0>, and pyramidal <c+a> slip {1 1 – 2 2} <–1 – 1 2 3> as well as the {1 0 – 1 2} ET and {1 1 – 2 2} CT will be involved in the plastic deformation process of pure titanium. Equation (2) is utilized to transform the corresponding versions of each deformation mode in the hexagonal crystal system's four-axis exponent ($h k i l$) [$u v t w$] into the normalized triaxial index ($h_1 k_1 l_1$) [$u_1 v_1 w_1$] in the orthogonal coordinate system:

$$\begin{bmatrix} h_1 & k_1 & l_1 \end{bmatrix} = \begin{bmatrix} (2h + k) & \sqrt{3}k & \sqrt{3}l(a/c) \end{bmatrix} / d_{hkl} \quad (2)$$

$$\begin{bmatrix} u_1 & v_1 & w_1 \end{bmatrix} = \begin{bmatrix} (2u + v)\sqrt{3}/2 & 3v/2 & wc/a \end{bmatrix} / d_{uvw}$$

where d_{hkl} and d_{uvw} are the moduli of [$h_1 k_1 l_1$] and [$u_1 v_1 w_1$], respectively, and their expressions are as follows:

$$d_{hkl} = \left[(2h + k)^2 + 3k^2 + 3l^2(a/c)^2 \right]^{1/2}$$

$$d_{uvw} = \left[3(u + v/2)^2 + 9v^2/4 + w^2(c/a)^2 \right]^{1/2} \quad (3)$$

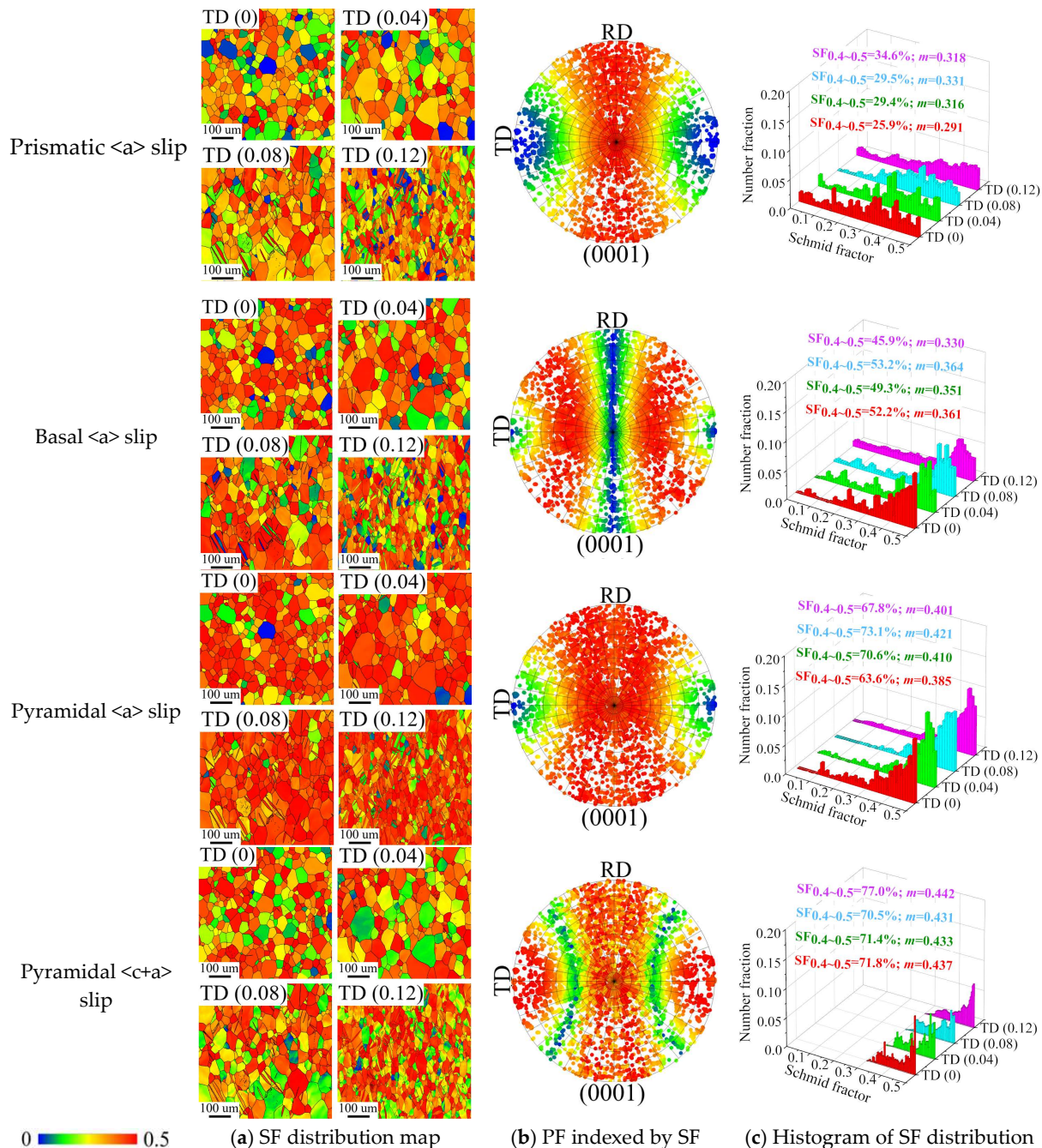


Figure 10. The evolution of the SF distribution of each deformation mode with strain when tensile along TD: (a) SF distribution map; (b) PF indexed by SF; (c) the histogram of SF distribution.

By substituting the external force F and the converted triaxial exponent into Equation (1), the relationship curves between SF and θ for each deformation mode can be obtained. Equivalent variants with maximum SF values will be activated preferentially. Figure 12 shows the variation law of the $SF_{\max}$ - θ curve of each deformation mode with the ψ value. To make the analysis easier, $\alpha = 0^\circ$, 15° , and 30° were used to calculate the $SF_{\max}$ - θ curves. The $SF_{\max}$ value is mostly influenced by the ψ value, whereas the θ value primarily influences

the trend of the $SF_{\max}-\theta$ curve for each deformation mode, as seen in Figure 12. The comparison of the various deformation modes reveals that when crystal preference orientation and loading mode change, the $SF_{\max}$ value of each deformation mode varies dramatically. This is one of the primary causes of the anisotropy of the material's plastic behavior.

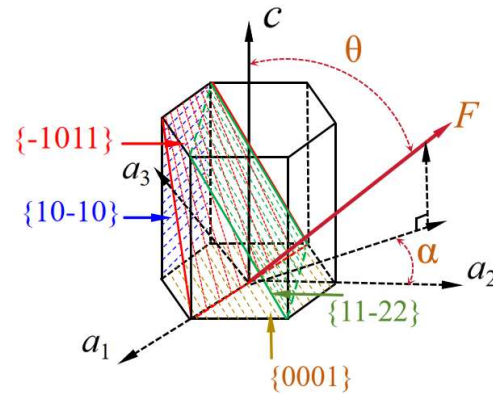


Figure 11. Schematic diagram of the positional relationship between the loading axis and each slip system.

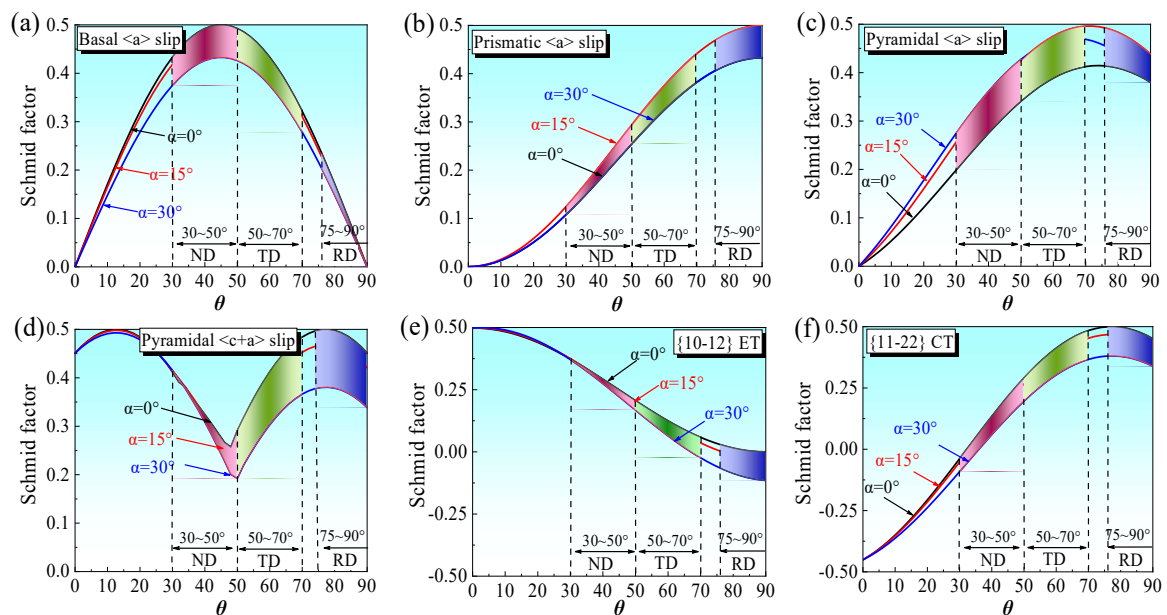


Figure 12. The variation law of the $SF_{\max}-\theta$ curve of each deformation mode with the ψ value: (a) basal $\langle a \rangle$ slip; (b) prismatic $\langle a \rangle$ slip; (c) pyramidal $\langle a \rangle$ slip; (d) pyramidal $\langle c+a \rangle$ slip; (e) $\{10-12\}$ ET; (f) $\{11-22\}$ CT.

Figure 13 displays the distribution range of the angle between the loading axis and the grain $\langle 0001 \rangle$ axis for the rolled textured titanium plate under different loading situations. The IPF clearly shows the distribution of the angle θ between each loading axis and the grain $\langle 0001 \rangle$ axis. Combined with the statistical results of histograms, it can be seen that the angle θ between the $\langle 0001 \rangle$ axis of grain and the $\langle 100 \rangle$ axis is mainly distributed at $75\sim 90^\circ$, and the number of grains accounts for 80%. Similarly, the angles θ between the $\langle 0001 \rangle$ axis and the $\langle 010 \rangle$ and $\langle 001 \rangle$ axes are concentrated at $50\sim 70^\circ$ and $30\sim 50^\circ$, respectively, and the corresponding grain number accounts for around 75% and 62% of the grain, respectively. It seems reasonable to assume that the majority of the material's plastic behavior is dominated by grains in this specific range. The distribution range of the SF values of each deformation mode under different loading conditions can be obtained by labeling the obtained θ values in Figure 12, and the specific SF values are shown in Table 2.

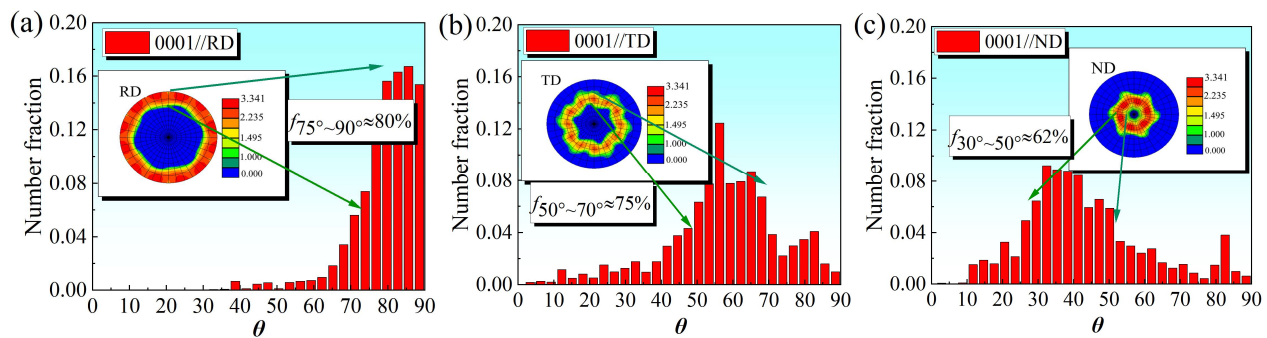


Figure 13. Distribution range of the angle between the grain $\langle 0001 \rangle$ axis and each loading axis under different loading conditions: (a) $\langle 100 \rangle$ axis; (b) $\langle 010 \rangle$ axis; (c) $\langle 001 \rangle$ axis.

Table 2. Theoretical prediction values of SF for each deformation mode under different loading conditions.

Deformation Mode	RD (75~90°, 90°)			TD (50~70°, 60°)			ND (30~50°, 40°)		
	$m_{\min}$	$m_{\max}$	m	$m_{\min}$	$m_{\max}$	m	$m_{\min}$	$m_{\max}$	m
prismatic $\langle a \rangle$ slip	0.41	0.50	0.46	0.25	0.44	0.33	0.11	0.29	0.17
basal $\langle a \rangle$ slip	0.00	0.24	0.12	0.28	0.49	0.38	0.37	0.50	0.42
pyramidal $\langle a \rangle$ slip	0.38	0.49	0.43	0.34	0.50	0.41	0.20	0.43	0.31
pyramidal $\langle c+a \rangle$ slip	0.34	0.50	0.45	0.19	0.48	0.42	0.19	0.41	0.40
$\{10\text{--}12\}$ ET	0.00	0.03	0.00	0.00	0.20	0.18	0.17	0.38	0.30
$\{11\text{--}22\}$ CT	0.34	0.50	0.40	0.20	0.48	0.30	0.00	0.29	0.09

The prismatic $\langle a \rangle$ slip, pyramidal $\langle a \rangle$ slip, pyramidal $\langle c+a \rangle$ slip, and $\{1\ 1\ -\ 2\}$ ET have m values of around 0.46, 0.43, 0.45, and 0.40, respectively, as shown in Table 2, suggesting that the crystal orientation is favorable for the activation of the aforementioned slip (twin) system. The m -values of $\{1\ 0\ -\ 1\}$ ET and basal $\langle a \rangle$ slip, on the other hand, are nearly zero, indicating that they are poorly activated. The m values of prismatic $\langle a \rangle$ slip, basal $\langle a \rangle$ slip, pyramidal $\langle a \rangle$ slip, and pyramidal $\langle c+a \rangle$ slip are around 0.33, 0.38, 0.41, and 0.42 when tensile along TD, respectively. Therefore, the activation of the above-mentioned slip system needs to take the influence of CRSS into account. In addition, it is observed that when tensile along ND, the basal $\langle a \rangle$ slip and pyramidal $\langle a \rangle$ slip are more likely to be activated, but the prismatic $\langle a \rangle$ slip is more difficult to activate. For twins, the $m_{\max}$ values of $\{1\ 0\ -\ 1\}$ ET and $\{1\ 1\ -\ 2\}$ CT during RD tensile are about 0.03 and 0.50, respectively, i.e., $\{1\ 1\ -\ 2\}$ CT is more likely to be activated. During TD tensile, the $m_{\max}$ values of $\{1\ 0\ -\ 1\}$ ET and $\{1\ 1\ -\ 2\}$ CT are around 0.20 and 0.48, respectively; nevertheless, it is really more probable that the former will be activated. The following are the causes of the aforementioned phenomenon: 1) Compared to $\{1\ 1\ -\ 2\}$ CT, the critical strain and CRSS needed for $\{1\ 0\ -\ 1\}$ ET are significantly lower. 2) The SF values are dispersed outside of the designated regions of Figure 12e,f, and $\{1\ 0\ -\ 1\}$ ET tends to nucleate in a small portion of the grains with a considerable tilt angle from ND to TD. Simultaneously, $\{1\ 0\ -\ 1\}$ ET is observed to be more likely to activate when tensile along ND. In conclusion, the theoretical SF calculation results in this section may accurately estimate the SF value of each slip system; nevertheless, but the local nucleation characteristics of twins need to be considered.

4.3. Identification of the Dominant Deformation Mechanism

The beginning of plastic deformation at the grain scale is associated with the activation of deformation modes such as dislocation slip (twin), and the combined effect of SF and CRSS is frequently utilized for predicting its activability [49–51]. The Schmid law indicates that the following requirements are satisfied by the equivalent critical shear stress $\tau_{eff,s}$

needed to activate the slip system s , the external stress σ_s needed to activate the slip system s , and the average SF value m_s of the slip system s :

$$\tau_{eff,s} = \sigma_s \cdot m_s \quad (4)$$

The prismatic $\langle a \rangle$ slip with low CRSS when stretched along RD and TD has absolute preferred activation based on the soft-oriented grain ratio ψ or SF values. This means that the activability of other deformation modes may be measured using prismatic slip as a reference. The values of σ_s , $\tau_{eff,s}$, and m_s of each slip system satisfy the following proportional relationship, which may be derived from Equation (4):

$$\frac{\sigma_{basal\langle a \rangle}}{\sigma_{prism\langle a \rangle}} = \frac{m_{prism\langle a \rangle}}{m_{basal\langle a \rangle}} \cdot \frac{\tau_{eff,basal\langle a \rangle}}{\tau_{eff,prism\langle a \rangle}} \quad (5)$$

$$\frac{\sigma_{pyram\langle a \rangle}}{\sigma_{prism\langle a \rangle}} = \frac{m_{prism\langle a \rangle}}{m_{pyram\langle a \rangle}} \cdot \frac{\tau_{eff,pyram\langle a \rangle}}{\tau_{eff,prism\langle a \rangle}} \quad (6)$$

$$\frac{\sigma_{pyram\langle c+a \rangle}}{\sigma_{prism\langle a \rangle}} = \frac{m_{prism\langle a \rangle}}{m_{pyram\langle c+a \rangle}} \cdot \frac{\tau_{eff,pyram\langle c+a \rangle}}{\tau_{eff,prism\langle a \rangle}} \quad (7)$$

Obviously, the discrimination results of the dominant deformation mode will be directly affected by the proportionate relationship between the CRSS of each deformation mode. Combined with the results of the previous investigation, it is found that the proportional relationship between the CRSS of each slip system is concentrated in prismatic $\langle a \rangle$ slip: pyramidal $\langle a \rangle$ slip: basal $\langle a \rangle$ slip: pyramidal $\langle c+a \rangle$ slip = 1:1.2~1.4:1.7~2.1:2.5~4.0 [34,51]. The variation law of the needed activation stress ratio of each slip system with the $\tau_{eff,s}$ ratio may be determined by substituting the aforementioned CRSS proportional relationship and the m value of SF obtained in Section 4.2 into Equations (5)–(7), as shown in Figure 14. As shown in Figure 14a, when $\tau_{eff,basal\langle a \rangle} / \tau_{eff,prism\langle a \rangle} \in [1.7, 2.1]$, $\sigma_{basal\langle a \rangle} / \sigma_{prism\langle a \rangle}$ are 6.51~8.05 and 1.48~1.82, respectively, when tensile along RD and TD. This indicates that the activation of the basal $\langle a \rangle$ slip requires at least 6.51 times (RD tensile yield strength ~285 MPa) and 1.48 times (TD tensile yield strength ~400 MPa) the current tensile yield strength. For the same reason, Figure 14b demonstrates that when $\tau_{eff,pyram\langle c+a \rangle} / \tau_{eff,prism\langle a \rangle} \in [2.5, 4.0]$, the distribution of $\sigma_{pyram\langle c+a \rangle} / \sigma_{prism\langle a \rangle}$ is 1.96~3.14 and 2.56~4.09, respectively, when tensile along RD and TD. This indicates that 2.56 and 1.96 times the current yield strength are necessary to activate the pyramidal $\langle c+a \rangle$ slip. In the case of the pyramidal $\langle a \rangle$ slip (Figure 14c), the distribution of $\sigma_{RD,pyram\langle a \rangle} / \sigma_{RD,prism\langle a \rangle}$ is 1.28~1.50 at $\tau_{eff,pyram\langle a \rangle} / \tau_{eff,prism\langle a \rangle} \in [1.2, 1.4]$. This means that in order to activate the pyramidal $\langle a \rangle$ slip, at least 1.28 times the current yield strength (RD tensile initial yield) is needed. However, for $\tau_{eff,pyram\langle a \rangle} / \tau_{eff,prism\langle a \rangle} \in [1.2, 1.4]$, $\sigma_{TD,pyram\langle a \rangle} / \sigma_{TD,prism\langle a \rangle}$ are 0.96~1.13 at $\tau_{eff,pyram\langle a \rangle} / \tau_{eff,prism\langle a \rangle}$, suggesting a high probability that the pyramidal $\langle a \rangle$ slip is activated when tensile along TD.

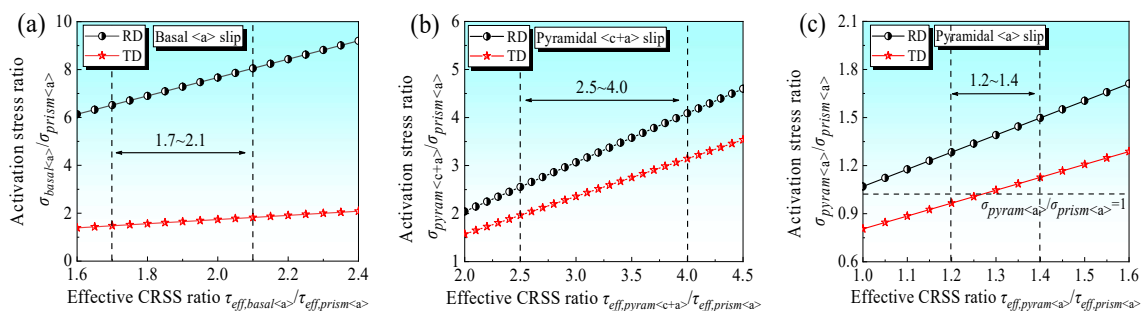


Figure 14. The variation of the needed activation stress ratio of each slip system with the $\tau_{eff,s}$ ratio under different loading conditions: (a) basal $\langle a \rangle$ slip; (b) pyramidal $\langle c+a \rangle$ slip; (c) pyramidal $\langle a \rangle$ slip.

It is possible to deduce that prismatic $\langle a \rangle$ slip dominates the initial yield of RD tensile, whereas the other three slip systems are difficult to activate, based on the evolution features of texture components and Schmid law analysis. Texture redirection has little effect on the angle formed by the loading axis and the grain c -axis during RD tensile; therefore, each slip system's SF value changes very little (Figure 7). It can be predicted that, considering applied stress increases, the pyramidal $\langle a \rangle$ slip will activate, and the basal $\langle a \rangle$ slip will be difficult to activate when combined with the following strain hardening degree (the hardening value is around 200 MPa at $\varepsilon = 0.08$). Moreover, the pyramidal $\langle c+a \rangle$ slip is also activated in the soft-oriented portion of the grain to fulfill the requirements of c -axis deformation. During TD tensile deformation, the prismatic $\langle a \rangle$ slip is preferentially activated in the soft-oriented part of the grain; however, the other two slip systems are more difficult to activate, even if the activation of the pyramidal $\langle a \rangle$ slip during TD tensile is greater than that of RD. However, the prismatic $\langle a \rangle$ slip activation in the poor crystal orientation leads to a higher initial yield strength when tensile along TD. Prismatic $\langle a \rangle$ slip and pyramidal $\langle a \rangle$ slip are predicted to become the predominant deformation modes with an increase in applied stress when combined with the following strain hardening degree (the hardening value is around 80 MPa at $\varepsilon = 0.08$). The smaller activation stress ratio of $\sigma_{\text{basal } \langle a \rangle} / \sigma_{\text{prism } \langle a \rangle}$ when tensile along TD means that basal $\langle a \rangle$ slip is possible within the part of the grain with a smaller θ (with a greater SF value than $m_{\text{basal } \langle a \rangle}$). A larger activation stress than $\sigma_{\text{pyram } \langle c+a \rangle} / \sigma_{\text{prism } \langle a \rangle}$ means that the pyramidal $\langle c+a \rangle$ slip is less likely to be activated, or it may only be activated in grains with large theta angles or in areas where there is a significant local stress concentration. The c -axis deformation of the grain in the grain with hard orientation for prismatic $\langle a \rangle$ slip, pyramidal $\langle a \rangle$ slip, and pyramidal $\langle c+a \rangle$ slip can be coordinated by {10–12} ET after a certain cumulative strain, as can be observed when combined with Figure 5. Furthermore, the twin's activation is correlated with the crystal orientation and the local stress concentration.

To summarize, the macro-mesoscopic deformation characteristics of materials are significantly influenced by the activability of each deformation mode. The reasonable prediction of the dominant deformation modes of material yield is directly impacted by the CRSS values of each slip system and the proportional relationship between them. Additionally, research on the deformation mechanism's activation can help us better understand the constraints and conditional characteristics of different approaches to producing CRSS values in subsequent research. Consequently, in order to guide the rational formulation of the process and further realize the regulation of material characteristics, it is imperative that the analysis and identification process of the dominant mechanism of anisotropic deformation be thoroughly studied.

5. Conclusions

- In this work, the influence mechanism of rolling texture on the anisotropy of the plastic behavior of a pure titanium plate was studied by Schmid law in combination with a macroscopic tensile test and mesoscopic EBSD characterization. The conclusions are as follows:
- The special crystal orientation corresponding to the basal bimodal texture affects the activability of each slip system by changing the SF value, and the changes in the type and difficulty of the activated deformation mode lead to the obvious anisotropy of the macro- and micro- deformation characteristics of the rolled titanium plate. The angle θ between the loading axis and the c -axis mostly influences the variation trend of the SF of each slip system, whereas the angle α between the loading axis projection and the a -axis impacts the value of SF. The theoretical derivation results can well predict the SF value of each slip system under different loading conditions;
- The tilt angle θ_b of the grain c -axis from ND to TD becomes polarized due to dislocation slip during tensile along RD and TD, but the tilt angle θ_a of the grain c -axis from ND to RD stays approximately vertical. The SF value of each deformation mode changes with the change of θ_a during RD tensile, but the change of θ_a is minor, resulting in the

ψ - m values of each deformation mode being almost constant. Since the SF values of each deformation mode change mainly with the change of θ_b during TD tensile, the discrete character of the SF values of each deformation mode results from the broad range of θ_b ;

- The prismatic $\langle a \rangle$ slip essentially dominates the RD tensile deformation. The pyramidal $\langle a \rangle$ slip and pyramidal $\langle c+a \rangle$ slip will be activated during the subsequent hardening, but the basal $\langle a \rangle$ slip cannot. During TD tensile, the prismatic $\langle a \rangle$ slip is preferentially activated in the soft-oriented part of the grain. Prismatic $\langle a \rangle$ slip and pyramidal $\langle a \rangle$ slip become the dominant deformation modes during the subsequent hardening. The basal $\langle a \rangle$ slip will be activated in the soft-oriented grains, while the pyramidal $\langle c+a \rangle$ slip is more difficult to activate. {10-12} ET coordinates the c -axis deformation in the grains where the dislocation slip is limited.

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