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ORIGINAL REPORT



Unraveling the origin of ductility in multilayered Ti/Nb composites: role of dislocation evolution

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ABSTRACT

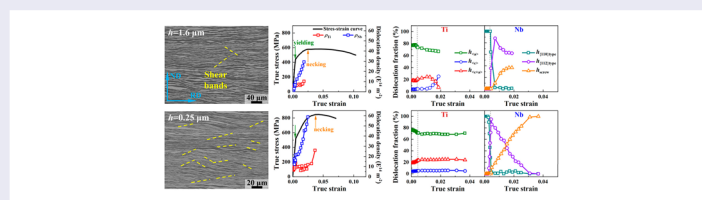
Dislocation evolution in multilayered Ti/Nb composites processed by accumulative roll bonding under uniaxial tension was quantitatively evaluated by high-energy X-ray diffraction. The composites with different heterogeneous microstructures show distinct dislocation activities in Ti, which majorly contributes to their different ductility. For the composite in which extensive shear bands have formed before loading, the stable fraction of $\langle c+a \rangle$ dislocations and the suppressed activation of $\langle c \rangle$ dislocations contribute to its uniform deformation. We emphasize that tracking the dynamic evolution of dislocations is fundamental for understanding ductility of the multilayers, whereas the increased $\langle c+a \rangle$ dislocations found by post-mortem analysis do not necessarily dominate.

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IMPACT STATEMENT

Dynamic evolution of dislocations, rather than only dislocation densities obtained by post-mortem analysis, is found to explain ductility of multilayered composites, which can be tailored by adjusting heterogeneous microstructures.

Introduction

Heterostructured materials have recently attracted extensive interests in the materials community owing to their great potential in evading strength-ductility dilemma. The typical heterostructures include gradient, laminate, multi-modal, harmonic structure, etc. [1,2]. Among them, the laminated materials show exceptional properties in extreme environments, such as thermal stability [3], creep resistance [4], radiation damage tolerance [5,6], as well as shock [7] and wear resistances [8], which make them could be potentially applied in aerospace, nuclear and automobile industries. As a vital microstructural characteristic, the heterophase interface between the constituent metals within one laminate plays a crucial role in the mechanical behavior. On the one hand, the

interface may have a significant influence on the macroscopic stress-strain responses due to the deformation incompatibility [9,10]. Then, interface affected zone [11] and hetero-deformation induced (HDI) hardening [12] were proposed, which are closely related to the accumulation of geometrically necessary dislocations (GNDs) at the interfaces. Therefore, the mechanical properties of the material are closely dependent on the varying interface spacing [9,11]. On the other hand, the heterophase interface can affect the local stress/strain states by transferring or accommodating the load during deformation. As a consequence, extraordinary deformation mechanisms, i.e. unusual dislocations and twinning, can be activated. For example, non-Schmid slip systems and unconventional $\{10\bar{1}2\}$ $\langle 10\bar{1}\bar{1} \rangle$ extension de-twinning are

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activated in a laminated aluminum with alternative stacking of fine-grained Al layers and coarse-grained Al layers [13] and Ti/Al composites [14], respectively, due to the local multiaxial stress state. This complex stress state, differing from global uniaxial tension, maybe caused by the deformation incompatibility between constituent layers [15] as well as the promoted layer-to-layer interactions by the formation of shear bands [16]. Especially, the enhanced interface constraint effect on the deformation mechanisms has been reported in hexagonal close-packed (HCP) laminated materials [17]. Concomitantly, the variation of deformation mechanisms is pivotal to determine the exceptional mechanical properties, particularly in HCP metals, owing to their complicated deformation modes. Mechanical properties of the HCP-based laminates have been extensively investigated, especially in the HCP/BCC (body-centered-cubic) multilayers [8,18,19]. The dependence of mechanical properties on microstructure, such as layer thickness [18,20], interface structure [19] and interface orientation [8], have been fully elucidated. However, the underlying mechanism for the exceptional properties of bulk materials has not been clarified. Some researchers have tried to overcome these obstacles from the perspective of local strain/stress fields inside the multilayers [6,17,21–23]. Furthermore, the deformation mechanisms in the HCP-based laminated materials are investigated using transmission electron microscopy (TEM) [17] and crystal plasticity modeling [24], which reveals the important role of $\langle c+a \rangle$ dislocations in plastic deformation.

In spite of these, a quantitative assessment of dislocation activity in heterostructured laminated materials under deformation has not been reported. For the direct identification of various dislocations, TEM analysis allows for very local measurements, and provides reliable results only when the dislocations are not dense, i.e. primarily at small loading strains. Moreover, the on-site deformation to high strains within electron microscope are difficult [25]. Fortunately, the synchrotron-based X-ray diffraction can easily measure dislocation density in a large diffraction volume during loading, due to its high brilliance and temporal–spatial resolution [26]. Further, dislocation activity can be tracked by performing a line-profile analysis on the measured physical diffraction profiles [27]. In this work, we applied *in situ* synchrotron-based high-energy X-ray diffraction (HE-XRD) during tensile loading to determine the dislocation density and activity for the prototypic Ti/Nb composites fabricated by accumulative roll bonding (ARB). The correlation between dislocation slip and macroscopic mechanical behavior is then elaborated. This work quantitatively elucidates the activation of different slip systems in the HCP-based composites under deformation, which sheds light

on the understanding of deformation behavior as well as the strengthening and toughening mechanisms of this group of heterostructured materials.

Experimental

The investigated multilayered Ti/Nb composites were fabricated by the ARB process as described elsewhere [16]. In this study, three composites with the nominal layer thicknesses (h) of 48, 1.6 and 0.25 μm were selected for investigation, corresponding to the equivalent strains of 3.5, 7.4 and 9.6 after three, six and eight ARB cycles, respectively. The microstructures were examined by a JEOL 7001F scanning electron microscope (SEM), and a Zeiss Crossbeam550 electron microscope equipped with a Symmetry electron backscattered diffraction (EBSD) detector. The *in situ* HE-XRD measurements during uniaxial tensile tests were carried out at P21.2 beamline of PETRA III synchrotron source (Germany), using the transmission geometry. A monochromatic X-ray beam with an energy of 67 keV was used. The high angular resolution required for line-profile analysis was achieved in the far-detector position of 1 m. The tensile tests were performed on a compact load frame with a nominal strain rate of 10^{-3} s^{-1} . Flat dog bone-shaped tensile specimens were machined from the as-ARBed plates along the rolling direction (RD).

Results and discussion

The SEM images (Figure 1a₁–c₁) show the lamellar morphologies of the investigated Ti/Nb composites, in which the dark layers are Ti and the bright are Nb. No visible interfacial diffusion is identified. Both the Ti and Nb layers are flat in the $h = 48 \mu\text{m}$ composite. In $h = 1.6 \mu\text{m}$ composite, the metallic layers become wavy due to the necking and fracture of Nb layers, and the sporadic shear bands appear (yellow dashed lines in Figure 1b₁). The angles between the shear bands and RD are $30 \sim 35^\circ$. With the further ARB processing, abundant shear bands are distributed in the bulk $h = 0.25 \mu\text{m}$ composite. The Nb layers are completely fragmented and distributed dispersedly. Moreover, the tilt angles of shear bands decrease to $15 \sim 20^\circ$. This is because that the Ti layers are work hardened in the later ARB process, and the deformation incompatibility between these two metals is relieved. Thus, under the compressive stress during rolling, the Nb fragments become re-oriented, and the inclination angle of shear bands is weakened. Figure 1a₂–c₂ are all Euler maps showing grain morphologies in each constituent metal obtained by EBSD mapping. The composites consist of alternating coarse-grained Ti layers and fine-grained Nb layers. Ti has

elliptical grains that were formed by dynamic recrystallization during warm-roll bonding. The grain size distributions can be found in Figure S1 (Supplementary material). The average size (along RD) in Ti is ~ 550 , 484 and 376 nm for the $h = 48 \mu\text{m}$, $h = 1.6 \mu\text{m}$ and $h = 0.25 \mu\text{m}$ composites, respectively. The grains in Nb are dramatically elongated along RD, in appearance of a fiber shape. The average grain size along the normal direction (ND, i.e. through thickness direction) in Nb is ~ 178 , 110 and 24 nm for the $h = 48 \mu\text{m}$, $h = 1.6 \mu\text{m}$ and $h = 0.25 \mu\text{m}$ composites, respectively (Figure S1a₃-c₃). Besides, crystallographic orientations (texture) of the composites have been determined elsewhere [16]. Three composites exhibit similar textures in each constituent metal. Namely, Ti exhibits a strong $\{11\bar{2}5\}\langle 10\bar{1}0\rangle$ texture component, and Nb shows the R-cubic ($\{001\}\langle 110\rangle$) orientation and some weak γ -fiber ($\langle 111\rangle$ //ND) textures. In a word, across the heterophase interfaces, there are apparent microstructure heterogeneities including grain morphologies, textures and constituent metals, which is expected to enable a strong interface effect on deformation behavior of the bulk composites. Tensile stress-strain curves of the Ti/Nb composites are shown in Figure 1a₃-c₃. The $h = 48 \mu\text{m}$ and $h = 1.6 \mu\text{m}$ composites start to neck at comparable applied strains (indicated by the orange arrows in Figure 1a₃-b₃). The $h = 0.25 \mu\text{m}$ composite shows a slightly higher uniform elongation, despite of the fact that it has experienced severe plastic deformation during ARB with a processing strain up to 9.6.

The HE-XRD patterns obtained during the uniaxial tensile tests are shown in Figure S2 (Supplementary material). To avoid the effects of macroscopic stress on the dislocation determination, the patterns at the azimuth angle of 50° are used. Evidently, the diffraction peaks broaden with the increasing load, especially after yielding of the composites (inset in Figure S2). To quantify the contributions of various dislocation types to the macroscopic deformation, the broadening of peak widths is further examined with the line-profile analysis that combines the modified Williamson-Hall (MWH) and the modified Warren-Averbach (MWA) methods. The MWH method determines dislocation contrast factors solely from the peak broadening, and then the MWA method provides dislocation densities based on the peak shape. Consequently, the combined MWH-MWA method can account for the individual broadening contributions of various dislocation types ($\langle a \rangle$, $\langle c \rangle$ and $\langle c+a \rangle$ -type dislocations for Ti; $\{110\}\langle 111 \rangle$, $\{112\}\langle 111 \rangle$ and screw-type dislocations for Nb). It is worth mentioning that this method has been proved to be the most efficient way to determine the dislocation types in HCP metals [28,29]. The detailed analysis procedure for the $h = 0.25 \mu\text{m}$ composite before

uniaxial tension is served as an example and illustrated as follows.

The diffraction peak broadening can be described as the MWH equation [30]:

$$\Delta K = 0.9/D + (\pi A^2 b^2 \rho / 2)^{1/2} (K \bar{C}^{1/2}) + O(K^2 \bar{C}) \quad (1)$$

where K , D , b , ρ are the Bragg position, the average crystallite size, the magnitude of Burgers vector, and the dislocation density, respectively. A is a parameter depends on the effective outer cut-off radius of dislocations and O is the negligible high-order terms. The average contrast factor $\bar{C}$ corresponding to each $\{hk.l\}$ reflection in HCP crystals is given as [31]:

$$\overline{C_{hk.l}} = \overline{C_{hk.0}}(1 + q_1 x + q_2 x^2) \quad (2)$$

where $x = (2/3)[l/(ga)]^2$, $\overline{C_{hk.0}}$ is the average contrast factor of $\{hk.0\}$ reflections, q_1 and q_2 are parameters associated with the anisotropic elastic constants, g is the magnitude of diffraction vector, and a is the lattice constant. If inserting Equation (2) into Equation (1), the following equation is obtained:

$$(\Delta K - \alpha)^2 / K^2 = \beta(1 + q_1 x + q_2 x^2) \quad (3)$$

where $\alpha = 0.9/D$ and $\beta = (\pi A^2 b^2 / 2) \overline{C_{hk.0}} b^2$. Thus, the values of α , β , q_1 and q_2 can be determined by the optimal quadratic fitting between $(\Delta K - \alpha)^2 / K^2$ and x , as shown for the $h = 0.25 \mu\text{m}$ composite before tension in Figure 2a, using eight peaks including $\{10.0\}$, $\{10.1\}$, $\{11.0\}$, $\{11.2\}$, $\{20.1\}$, $\{20.3\}$, $\{21.1\}$ and $\{11.4\}$ reflections. The R^2 value in Figure 2a is the coefficient of determination indicating the fitting quality. Consequently, the measured values q_1 and q_2 are determined as: $q_1 = -0.5849$, $q_2 = 0.1789$, and D is 68 nm. These parameters (q_1 and q_2) can be employed to determine the relative fractions of dislocation populations with different Burgers vector types (i.e. $\langle a \rangle$, $\langle c \rangle$ and $\langle c+a \rangle$ for HCP). The details of the analysis procedure are given elaborately in Ref. [32,33]. Figure 2b illustrates the frequency distribution of the $\langle a \rangle$, $\langle c \rangle$ and $\langle c+a \rangle$ -type dislocations. Then, the average fractions of these three dislocation populations (denoted as h_a , h_c , h_{c+a}) are determined as 76.7%, 4.5% and 18.8%, respectively, which are the gravity centers of the histograms. Subsequently, the $\overline{C_{hk.0}} b^2$ value can be calculated according to Ref. [29], which is then used in the MWA analysis to estimate the dislocation density. The MWA equation is proposed as [34]:

$$\ln A(L) = \ln A^s(L) + Y(L)(K^2 \overline{C_{hk.l}} b^2) + O'(K^2 \overline{C_{hk.l}}^2 b^4) \quad (4)$$

$$Y(L) = -(\rho \pi L^2 / 2) \ln(R_e / L) \quad (5)$$

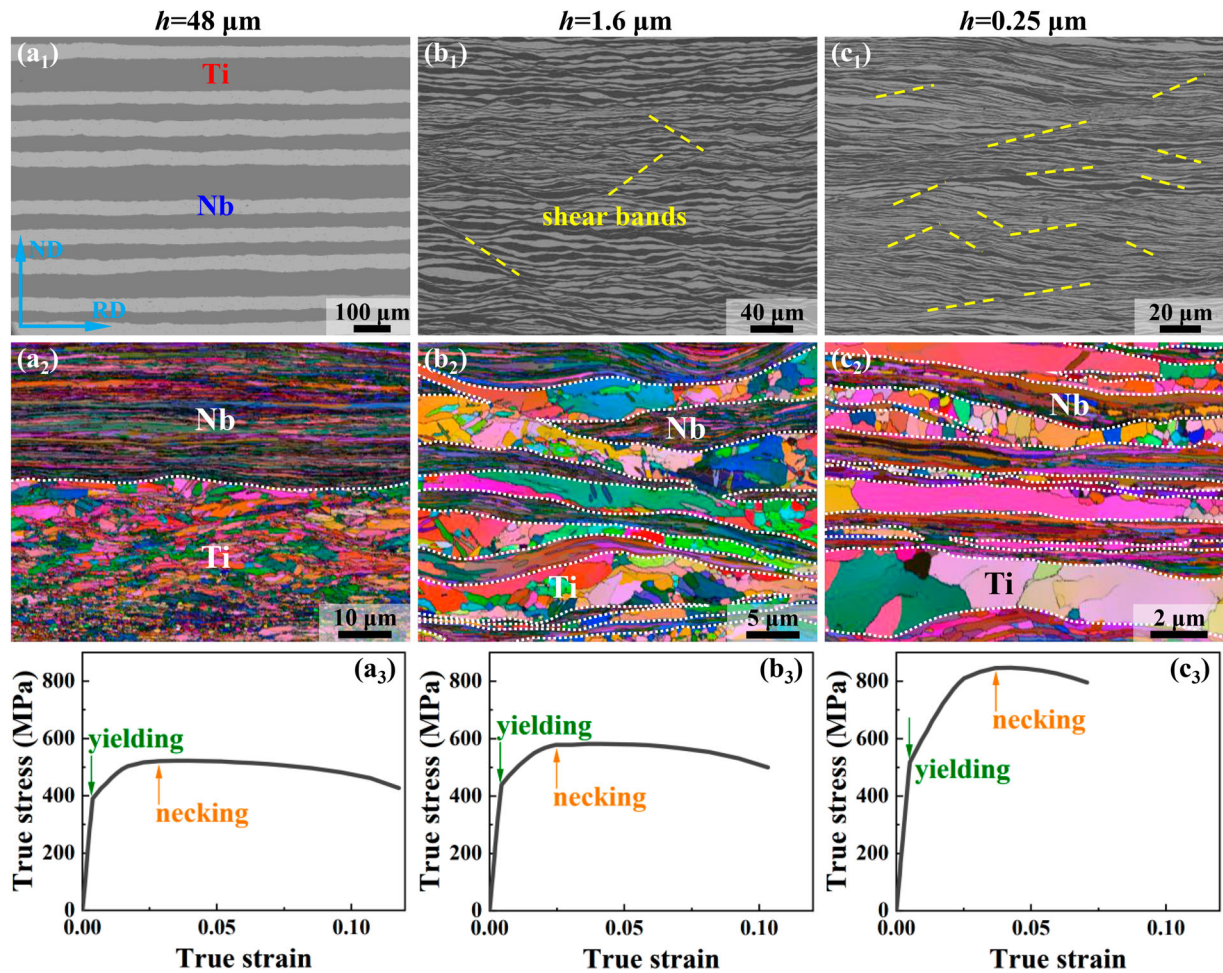


Figure 1. SEM images, all Euler maps and tensile stress-strain curves of the Ti/Nb composites with different layer thicknesses: (a) $h = 48 \mu\text{m}$, (b) $h = 1.6 \mu\text{m}$ and (c) $h = 0.25 \mu\text{m}$. The white dotted lines in (a₂–c₂) indicate the interfaces between Ti and Nb. The green and orange arrows in (a₃–c₃) indicate yielding and necking of the composites, respectively.

where $A(L)$ and $A^s(L)$ represent the Fourier coefficients of the diffraction profile and the size contribution, respectively; L , O' and R_e represent the Fourier length, high order terms and the effective outer cut-off radius of dislocations, respectively. By fitting $\ln A(L)$ against $K^2 \overline{C_{hk,l}^2} b^2$, the coefficient $Y(L)$ can be determined, as shown by the MWA plots in Figure 3a. Furthermore, Equation (5) can be rewritten as follows:

$$Y(L)/L^2 = (\pi\rho/2) \ln L - (\pi\rho/2) \ln R_e \quad (6)$$

Thus, the dislocation density (ρ) can be obtained by fitting the linear relationship between $Y(L)/L^2$ and $\ln L$ (Figure 3b). The principles for the combined MWH-MWA analysis in Nb is similar (details in Supplementary Note), and the dislocation density analysis for Nb in the $h = 0.25 \mu\text{m}$ composite before tension is shown in Figure S3 (Supplementary material).

Figure 4a₁–c₁ shows dislocation density evolution in each constituent metal (i.e. ρ_{Ti} and ρ_{Nb}) during uniform plastic deformation, along with the enlarged

stress–strain curves of the Ti/Nb composites. After yielding of the bulk Ti/Nb composites, dislocation densities in both Ti and Nb start to increase until necking, except for Nb in the $h = 48 \mu\text{m}$ composite. In this composite (Figure 4a₁), dislocation density first rises dramatically and then drops before necking. This is attributed to the premature fracture of the Nb layers during tension (indicated by the gray chain line), which releases the strain gradient/accumulation between the constituent metals. It should be noted that the dislocation density of Nb is always higher than that of Ti during the uniform deformation. Similar behavior was reported in a Zr/Nb (HCP/BCC) multilayer under nano-scratching [8]. The maximum dislocation densities of Nb can reach $3.3 \times 10^{15} \text{ m}^{-2}$ and $2.9 \times 10^{15} \text{ m}^{-2}$ for the $h = 48 \mu\text{m}$ and $h = 1.6 \mu\text{m}$ composite, respectively. It is worth mentioning that the continuous rise of dislocation density can reach $\sim 6 \times 10^{15} \text{ m}^{-2}$ for Nb in the $h = 0.25 \mu\text{m}$ composite (Figure 4c₁). Due to repulsive interactions among dislocations, crystalline materials have a maximum limit of dislocation density. The

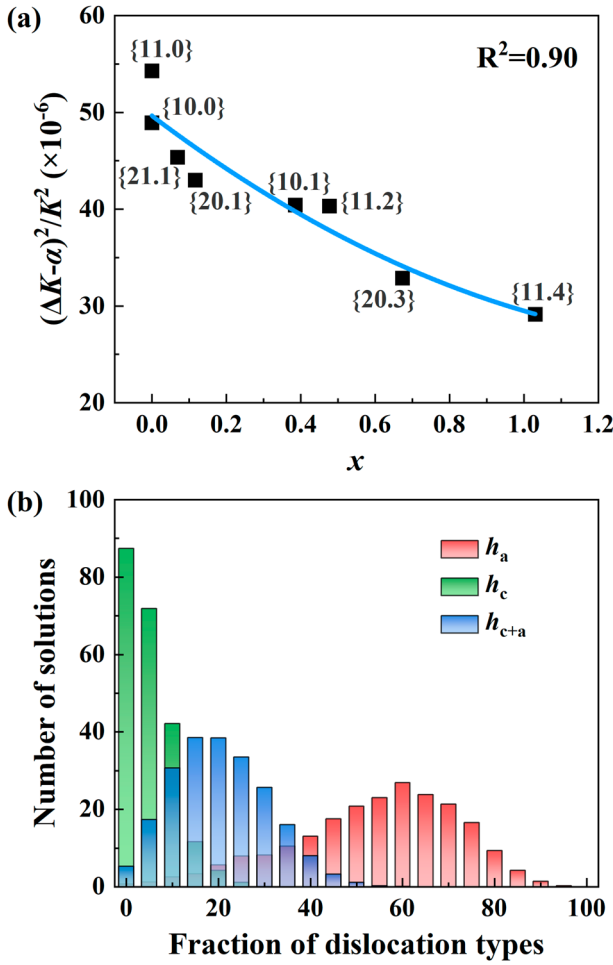


Figure 2. (a) Modified Williamson-Hall plot and (b) histogram of the frequency distribution of the $\langle a \rangle$, $\langle c \rangle$ and $\langle c+a \rangle$ -type dislocations (i.e. h_a , h_c and h_{c+a}) in Ti of the $h = 0.25 \mu\text{m}$ composite before tensile loading. The R^2 value in (a) is the coefficient of determination indicating the fitting quality.

maximum limit of dislocation density was demonstrated to be around 10^{16} m^{-2} in a severely deformed iron [35]. In a D&P steel, an extremely high dislocation density of $\sim 1.3 \times 10^{16} \text{ m}^{-2}$ was reported to contribute to its outstanding strength and ductility [36]. Recently, based on molecular dynamics (MD) simulation of a Zr/Nb multilayer [8], the dislocation density of the similar order as that in the Ti/Nb composites has been obtained. However, a much higher dislocation density even on the order of 10^{17} m^{-2} was reported in the Zr and Nb single crystals, which could be ascribed to the absent hindering effect of grain/phase boundaries on dislocation multiplication. Considering the small grain size in the $h = 0.25 \mu\text{m}$ composite, the grain boundary strengthening (σ_g) is approximately 309 and 265 MPa for Ti and Nb, respectively, which are calculated based on the Hall-Petch relationship (i.e. $\sigma_g = k \cdot d^{-1/2}$, where k is the Hall-Petch coefficient and d is the average grain size/thickness). It

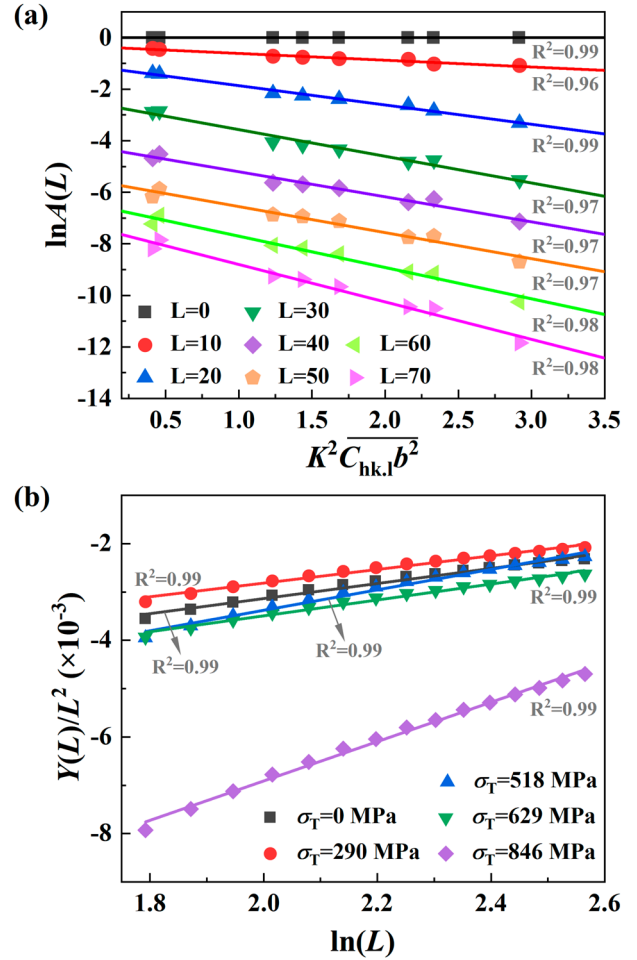


Figure 3. Modified Warren-Averbach analysis for Ti in the $h = 0.25 \mu\text{m}$ composite: (a) modified Warren-Averbach plot obtained from the diffraction profile before tensile loading, (b) plots of $Y(L)/L^2$ as a function of $\ln L$ at various applied true stress (σ_T). The R^2 values are the coefficients of determination indicating the fitting quality.

should be noted that the grain size of Ti is the average dimension along RD and ND of the ellipsoidal grains, and that of Nb is the thickness along ND due to its severely elongated morphology. Under an assumption of mixture rule, the contribution of grain size to this composite is 287 MPa. Based on the method in Ref. [16], the heterophase interface induced strengthening is evaluated as ~ 25 and 228 MPa for Ti and Nb in the $h = 0.25 \mu\text{m}$ composite, respectively. Evidently, the total strengthening from grain boundaries and heterophase interfaces ($\approx 287 + (25 + 228)/2 = 413.5$ MPa) is far below the ultimate tensile strength of the bulk composite (815 MPa). Therefore, the considerable dislocation density in Nb predominantly contributes to the high strength of the bulk composite with a layer thickness of $0.25 \mu\text{m}$.

Furthermore, the evolution of various dislocation activities in Ti and Nb is shown in Figure 4a₂-c₂ and

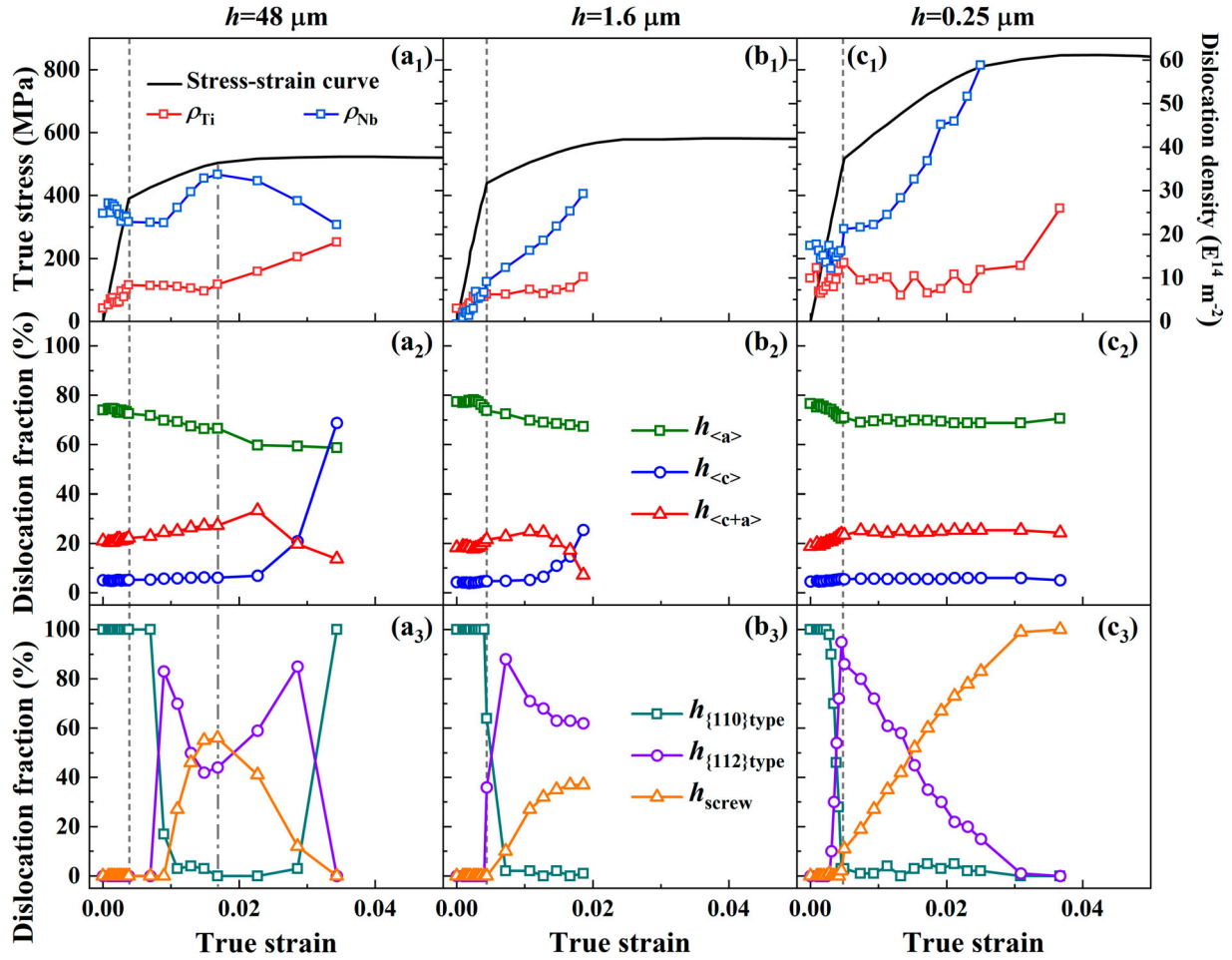


Figure 4. Evolution of dislocation densities/types versus true strain in the Ti/Nb composites with different layer thicknesses: (a) $h = 48 \mu\text{m}$, (b) $h = 1.6 \mu\text{m}$ and (c) $h = 0.25 \mu\text{m}$. (a₁–c₁) Dislocation density evolution in each constituent metal along with enlarged true stress–strain curves of the Ti/Nb composites, (a₂–c₂) evolution of dislocation activities in Ti, and (a₃–c₃) evolution of dislocation activities in Nb. The gray dashed lines indicate yielding of the Ti/Nb composites, and the gray chain lines indicate the premature fracture of the Nb layers.

Figure 4a₃–c₃, respectively. It is clear that the $\langle a \rangle$ dislocations (with fraction higher than 60%) prevail in Ti for all the studied composites, which is in agreement with their easier activation compared to other dislocations. Thus, $\langle a \rangle$ dislocations have a considerable contribution to the plastic deformation of Ti. The dislocation activities in Ti are similar in the $h = 48 \mu\text{m}$ and $h = 1.6 \mu\text{m}$ composites. Specifically, the fraction of $\langle a \rangle$ dislocations decreases gradually during plastic deformation. The activation of $\langle c+a \rangle$ dislocations first increase slightly and then drops significantly, while the fraction of $\langle c \rangle$ dislocations first keeps constant and then increases dramatically at higher applied strains. As $\langle c \rangle$ dislocations are sessile [37] and do not provide any independent slip modes, the significant activation of $\langle c \rangle$ dislocations lead to the premature failure of the $h = 48 \mu\text{m}$ and $h = 1.6 \mu\text{m}$ composites. In the $h = 0.25 \mu\text{m}$ composite, the dislocation activities in Ti generally keep stable, with the fractions of $\langle a \rangle$, $\langle c \rangle$ and $\langle c+a \rangle$ dislocations of $\sim 70\%$, 5% and 25% , respectively.

As for the dislocation activities in Nb, the variation tendency during uniform plastic deformation is similar in the three composites. After yielding, the fraction of $\{110\}\langle 111 \rangle$ dislocations drops quickly and becomes marginal in the following deformation, while the fraction of $\{112\}\langle 111 \rangle$ dislocations jumps immediately and then decrease gradually. During the whole period of plastic deformation, the fraction of screw dislocations increases continuously. This result in Nb can be expected because plastic flows in BCC metals are generally governed by screw dislocations [38]. The abnormal dislocation activities of Nb in the late stage of deformation (after the gray chain lines) for the $h = 48 \mu\text{m}$ composite is attributed to the premature fracture of Nb layers, which could potentially change the local stress state. Therefore, the different ductility of the studied Ti/Nb composites with various layer thicknesses can be attributed to their distinct dislocation activities in Ti, especially the different evolutions of $\langle c+a \rangle$ dislocations (as shown by the

$h = 0.25 \mu\text{m}$ composite). This coincides well with the finding in our previous work that the satisfactory ductility of the ARB processed Ti/Nb composites mainly comes from the strain hardening plateau of the various Ti grain families, which extends strain hardening of the bulk composites to the wider deformation ranges [16]. Generally, the activation of dislocations is closely related to the orientation of grains. However, these three composites exhibit the similar textures. Thus, there should be another factor leading to the abnormal dislocation activity in Ti of the $h = 0.25 \mu\text{m}$ composite. It has been demonstrated that shear bands that are related to the intensive wavy heterophase interfaces are favorable for $\langle c+a \rangle$ dislocation multiplication and motion in the HCP metals via promoting deformation accommodation between the constituent metals [39]. Therefore, the abundant and disperse shear bands in the $h = 0.25 \mu\text{m}$ composite are considered to strongly contribute to the stable activation of $\langle c+a \rangle$ dislocations and the resultant better ductility. This may provide an innovative design concept under the consideration of dislocation activities to obtain high-performance materials, via regulating the heterogeneous microstructural characteristics (such as shear bands and heterophase interfaces). Besides, the dislocation activities with plastic deformation are similar in both Ti and Nb constituents in the $h = 48 \mu\text{m}$ and the $h = 1.6 \mu\text{m}$ composites, indicating that the variation of layer thickness plays an insignificant role in dislocation type evolution for the composites with layer thicknesses in the micron-scale. Moreover, we emphasize that the dislocation activity in the Ti/Nb multilayers with submicron- and nano-scaled layer thicknesses depends on the size of individual layers significantly, as the confined layer slip (CLS) and/or dislocation transmission mechanisms may be activated and become to be dominant [19,40]. However, this is beyond the scope of the present work and needs further investigation. What's more, the dislocation densities for various slip modes in the Ti/Nb composites are shown in Figure S4 (Supplementary material). Although the fractions of $\langle c+a \rangle$ dislocations decrease significantly after deformation in the $h = 48 \mu\text{m}$ and $h = 1.6 \mu\text{m}$ composites, their dislocation densities are still higher than those before loading. Therefore, the increased $\langle c+a \rangle$ density obtained by means of post-mortem analysis can not accurately explain the ductility of the bulk HCP-based composites. By contrast, the dynamic evolution of dislocation activities as revealed by *in situ* measurements is the intrinsic origin.

Conclusion

In summary, by applying *in situ* synchrotron-based HEXRD technique, dislocation activities in the multilayered

Ti/Nb composites with various layer thicknesses are determined quantitatively. At the end of uniform deformation of the bulk materials, dislocation densities in Nb reach the order of 10^{15} m^{-2} . Specifically for the composite with layer thickness of $0.25 \mu\text{m}$, dislocation density in Nb layers reaches $\sim 6 \times 10^{15} \text{ m}^{-2}$, which is responsible for its outstanding strength. On the other hand, the difference of ductility for the composites mainly originates from their distinct dislocation activities in Ti. The predominant $\langle a \rangle$ dislocations majorly contribute to the plastic deformation of Ti. During plastic deformation, the decreasing fraction of $\langle c+a \rangle$ dislocations along with the increasing fraction of $\langle c \rangle$ dislocations in Ti leads to the early necking of the composites with layer thickness of 48 and $1.6 \mu\text{m}$. Contrastively, the stable fraction of $\langle c+a \rangle$ dislocations and the suppressed activation of $\langle c \rangle$ dislocations contribute to the better ductility of the composite with layer thickness of $0.25 \mu\text{m}$, even though it has experienced a high processing strain up to 9.6. The changed dislocation activities are attributed to the abundant and disperse shear bands in the composites fabricated by the highest ARB cycles. Furthermore, it is pointed out that dynamic evolution of dislocation activities is the intrinsic origin for the ductility of bulk HCP-based materials, rather than the increased $\langle c+a \rangle$ dislocations. These findings provide new insights into the fundamental deformation mechanism in multilayered composites. The research paradigm we have established for *in situ* tracking dislocation activities in the Ti/Nb composites can be extended to the design of more advanced materials with heterogeneous microstructures.

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Disclosure statement

No potential conflict of interest was reported by the author(s).

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