

Article

High-Temperature Deformation Behavior of Ti-55531 Alloy with a Lamellar Microstructure

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Abstract

The macroscopic mechanical properties of near- β titanium alloys depend inherently on their complex microstructural morphologies and phase transformation kinetics. This study investigates the deformation behavior of a lamellar Ti-55531 alloy during isothermal compression and in situ tensile testing. Pronounced strain rate sensitivity dictates the deformation of the lamellar microstructure. Low strain rate deformation (0.001 s^{-1}) induces dynamic recovery and recrystallization, which in turn drive α -lamellae fragmentation and the nucleation of new α phase during compression. The α precipitation is governed by a strict Burgers orientation relationship (BOR), but extensive plastic deformation may lead to the breakdown of the BOR. During tension, continuous slip transfer between adjacent phases is critically restricted by α/β interfacial thickness. As plastic strain accumulates, lath-like, V-shaped, and acicular α phases precipitate concurrently within β grains, creating a complex α microstructure.

Keywords: Ti-55531 alloy; in situ tension; orientation relationship; dislocation slip

1. Introduction

Near- β titanium alloys serve as crucial load-bearing components in the aerospace sector due to their high specific strength, structural hardenability and corrosion resistance [1–4]. The mechanical properties of Ti-55531, a typical near- β titanium alloy formulated as Ti-5Al-5Mo-5V-3Cr-1Zr, are fundamentally governed by its microstructural morphology [5–7]. This structural diversity inherently arises from complex phase transformations during processing.

Among titanium microstructures, the lamellar morphology attracts sustained research interest for providing a superior combination of strength and ductility [8–10]. The overall mechanical performance depends strongly on the morphology and scale, spatial distribution, and crystallographic orientation of the α phase relative to the β matrix [11–13]. Therefore, elucidating the microstructural evolution and α precipitation mechanisms during hot deformation becomes vital for optimizing thermomechanical processing and tailoring the resultant structural properties.

During hot deformation, the mutual interaction between the α and β phases dictates the structural evolution and continuous mechanical response. The grain size, texture, and strain state of the β phase determine the precipitation morphology, variant selection, and spatial arrangement of the α phase [5,14,15]. Conversely, the geometric characteristics and volume fraction of the α precipitates influence the dynamic recovery, recrystallization, and strain compatibility of the β matrix [16,17]. Previous investigations demonstrated the



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dynamic recrystallization behavior of α and β phases for Ti-5553 alloy during the sub-transus thermomechanical processing [18]. Even during hot compression, the combined effects of deformation heat, crystal defects, and dynamic recrystallization of the β matrix can trigger the precipitation of the secondary α phase in TC8 titanium alloy [19]. However, a comprehensive understanding of α precipitation within an initially lamellar microstructure remains limited. The mechanisms by which strain affects different types of α phase, particularly the competitive formation between grain boundary α and intragranular α , require further clarification.

Furthermore, in situ mechanical testing enables the real-time observation of microstructural progression [12,20,21]. The in situ tensile tests on Ti-6Al-4V elucidated the activation sequences of slip systems during early-stage deformation [11]. Nevertheless, detailed observations concerning the initiation and propagation of slip bands, along with their dynamic interactions with discrete microstructural features such as grain boundaries (GBs) and α lamellae in near- β titanium alloys at elevated temperatures, are still scarce.

Motivated by these considerations, this study investigates a Ti-55531 alloy featuring a fine lamellar initial microstructure. Isothermal compression and in situ high-temperature tensile tests are utilized to evaluate the influence of deformation parameters on microstructural evolution, including the precipitation behavior of the α phase and its crystallographic orientation relationship with the β matrix. To be specific, (i) we qualitatively reveal the precipitation behavior of the α phase during hot compression and demonstrate the dependence of nucleation sites on deformation temperature. (ii) We provide direct evidence showing that severe plastic deformation can locally disrupt the BOR. (iii) Through elevated-temperature in situ tensile testing, we correlate slip transmission with interfacial thickness and GBs characteristics, offering real-time insights that cannot be obtained from static observations. This study will establish a solid basis for microstructural tailoring and process optimization of advanced titanium alloys.

2. Materials and Methods

The forged Ti-55531 alloy with a diameter of 110 mm was supplied by the Beijing Institute of Aeronautical Materials. Its β -transus temperature was determined to be 840 ± 5 °C. Cylindrical specimens with $\Phi 8$ mm $\times$ 15 mm for compression testing and dog-bone tensile samples with a gauge length of 17 mm were machined. To generate a fine lamellar microstructure, these samples were initially held at 900 °C for 10 min to produce a single β -phase matrix, furnace-cooled at a rate of 8 °C/min to 770 °C (dual-phase region) for 1.5 h, and subsequently air-cooled to room temperature to induce the precipitation of α phase, as shown in Figure 1.

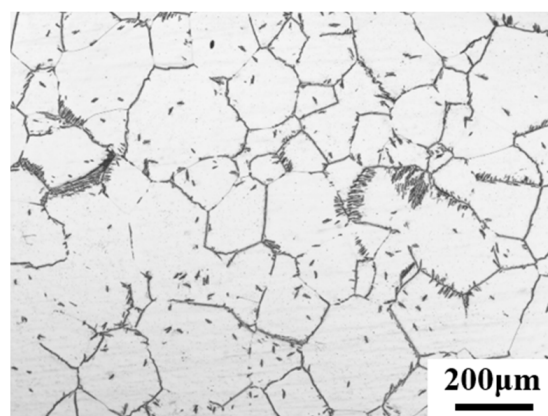


Figure 1. The microstructure of the undeformed Ti-55531 alloy after heat treatment.

Conventional forging of titanium alloys is generally carried out across the $\beta \rightarrow \alpha$ transformation temperature. To investigate the complex interactions between coexisting α and β phases, isothermal compression tests were executed on a Gleeble 1500 thermomechanical simulator (Duffers Scientific, Albany, NY, USA) at 650 °C and 700 °C (dual-phase region) with graphite lubricant. The applied strain rate was 0.001 s^{-1} with a strain of 50%. Specimens were heated at 20 °C/s to the target temperatures and deformed immediately without any holding time to minimize unexpected thermal phase transformations. After compression, the samples were promptly water-quenched to preserve the high-temperature deformed microstructures. To observe real-time slip traces, the in situ high-temperature tension was conducted employing a Lasertec VL2000DX-SVF17SP ultra-high temperature confocal laser scanning microscope equipped with a deformation stage (onekura Manufacturing Co., Ltd., Osaka, Japan).

Microstructural characterization was performed utilizing a Carl Zeiss Axio Imager.A2m optical microscope and a Zeiss SUPRA 55 field emission scanning electron microscope (SEM, Carl Zeiss AG, Oberkochen, Baden-Württemberg, Germany). Crystallographic orientation and phase analyses were acquired via an electron backscatter diffraction (EBSD) system attached to the SEM. All samples were electropolished in a mixture comprising 10 vol.% perchloric acid, 60 vol.% methanol, and 30 vol.% n-butanol at an applied voltage of 30 V and a temperature of -25 °C . For optical observation, the specimens were chemically etched in a reagent consisting of 1 vol.% hydrofluoric acid, 3 vol.% nitric acid, and the balance water.

3. Results

3.1. Flow Behavior and Microstructural Evolution During Hot Compression

The macroscopic flow stress behavior of the alloy inherently reflects its intrinsic resistance to deformation at elevated temperatures. However, the flow stress cannot be used directly because it is inevitably affected by both friction and the adiabatic temperature rise during deformation. A correction of the flow stress is therefore necessary. In this study, the strain rate is relatively low, so the decrease in flow stress caused by temperature deviation is negligible. The detailed procedure for the friction correction is described in References [22–24]. Figure 2 illustrates the corrected flow stress versus true strain curves of the Ti-55531 alloy with fine lamellar microstructure at a low strain rate of 0.001 s^{-1} . The flow stress curves exhibit analogous evolutionary patterns across different temperatures. The stress initially rises abruptly to a peak value and subsequently undergoes a rapid decline, which is then followed by a marginal rebound before continuously decreasing with further accumulated strain. Such flow behavior stems from the dynamic competition between work hardening and softening mechanisms. Meanwhile, the deformation temperature exerts a profound impact on the magnitude of the flow stress. The peak stress reaches approximately 140 MPa at 650 °C, whereas it notably decreases to around 100 MPa upon elevating the temperature to 700 °C. This temperature-dependent reduction in flow stress is attributed to the enhanced thermal activation. Elevated temperatures supply sufficient thermal energy to overcome the barriers against dislocation mobility, thereby predominantly facilitating the operation of dynamic recovery and dynamic recrystallization.

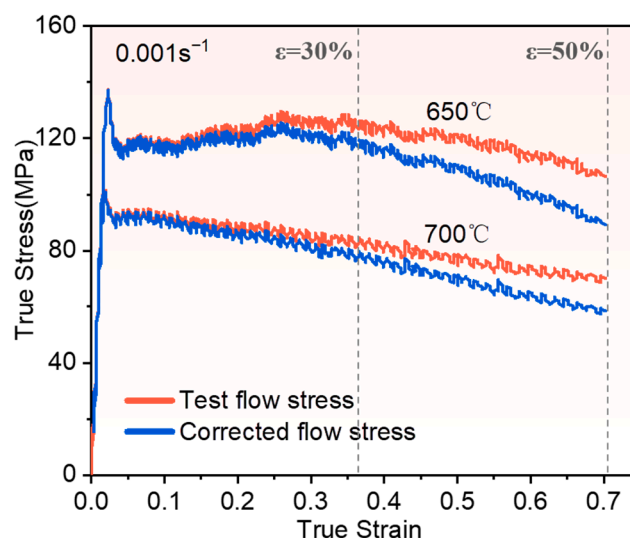


Figure 2. Corrected flow stress curves for Ti-55531 alloy at different deformation temperatures under a strain rate of 0.001 s^{-1} .

Figure 3 shows the microstructural evolution of the Ti-55531 alloy. At a strain of 30%, the α lamellae break down as the deformation temperature rises from $650 \text{ }^{\circ}\text{C}$ to $700 \text{ }^{\circ}\text{C}$. The α phases still nucleate preferentially at β GBs, and only a small amount of α phase precipitates within grains. Certain β GBs become thicker due to α precipitation along boundaries during deformation. With strain rising to 50%, the coupled thermal and mechanical effects lead to severe deformation of the β grains under compressive stress. Original straight β GBs bend and fragment, and α lamellae undergo more extensive fragmentation. Fine α phases form at β GBs, and additional α precipitates evolve within β grains. This indicates that high-density dislocations and subgrain boundaries introduced by large deformation act as alternative nucleation sites for intragranular α phase. Also, increasing the deformation temperature weakens the thermodynamic driving force for α formation, thereby reducing its volume fraction. At a strain of 50% and deformation temperatures of $700 \text{ }^{\circ}\text{C}$ and $650 \text{ }^{\circ}\text{C}$, the volume fractions of α phase were $21.8 \pm 3.6\%$ and $33.6 \pm 5.2\%$, respectively. The Vickers hardness values of the compressed alloy reached $294 \pm 7 \text{ HV}$ and $309 \pm 11 \text{ HV}$, respectively. Such hardness discrepancy exhibits a strong correlation with the volume fraction of α phase. The specimen deformed at $650 \text{ }^{\circ}\text{C}$ possesses a substantially higher α phase content, which accounts for its elevated hardness.

3.2. The Orientation Relationship Between the α and β Phases

To further clarify the interplay between deformation and phase transformation, EBSD characterization was performed. The deformed microstructure of the alloy at $650 \text{ }^{\circ}\text{C}$ with 50% strain is shown in Figure 4. The phase distribution map (Figure 4a) reveals that lamellar α precipitates at GBs. Both phases experience pronounced dynamic recrystallization under current deformation conditions, with respective recrystallized fractions of 33% for the β phase and 42.6% for the α phase (Figure 4b). Orientation characteristics of different α variants are further explored (Figure 4c,d). For α growing along GBs (α_{GB}), the $\{0002\}\alpha$ pole figure coincides with the $\{111\}\beta$, verifying the orientation relationship of $\{111\}\beta // \{0002\}\alpha_{\text{GB}}$. In contrast, other α at GBs and intragranular α exhibit no coincident poles with neighboring grains. This finding indicates that both possess distinct crystallographic orientations, likely arising from independent nucleation.

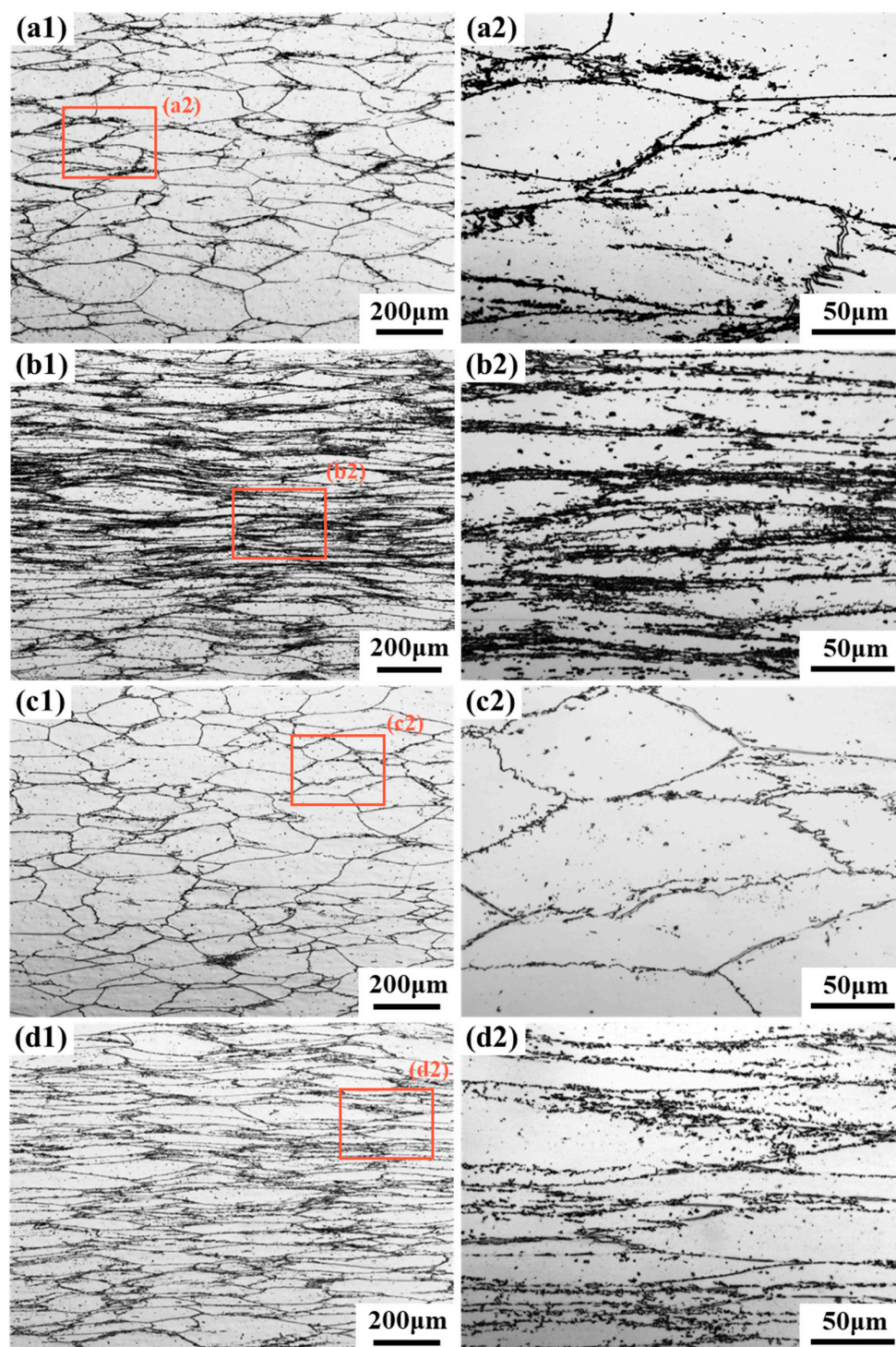


Figure 3. Deformation microstructure of Ti-55531 alloy during deformation at different strains and temperatures at a strain rate of 0.001 s^{-1} : (a1,a2) 30–650 °C; (b1,b2) 50–650 °C; (c1,c2) 30–700 °C; (d1,d2) 50–700 °C.

Figure 5 presents EBSD results obtained at a deformation temperature of 700 °C. As the deformation temperature increases, α phase nucleates at β GBs and also grows inside grains (Figure 5a). Compared with the alloy deformed at 650 °C, the dynamic recrystallization fractions of both β and α phases decrease (Figure 5b). Meanwhile, the lamellar α_1 and α_2 observed in orientation maps are identified as deformed α phases located at GBs. Only α_1 shares identical pole positions at $\{10\bar{1}0\}$ with β grains at $\{100\}$, yielding an orientation relationship of $\{100\}\beta // \{10\bar{1}0\}\alpha$. No overlapping poles exist between α_2 and adjacent grains, proving

that the two lamellar structures derive from separate nucleation with distinct crystallographic orientations. Under the applied deformation conditions, the nucleation orientation between deformed α and parent β phases deviates from the strict BOR.

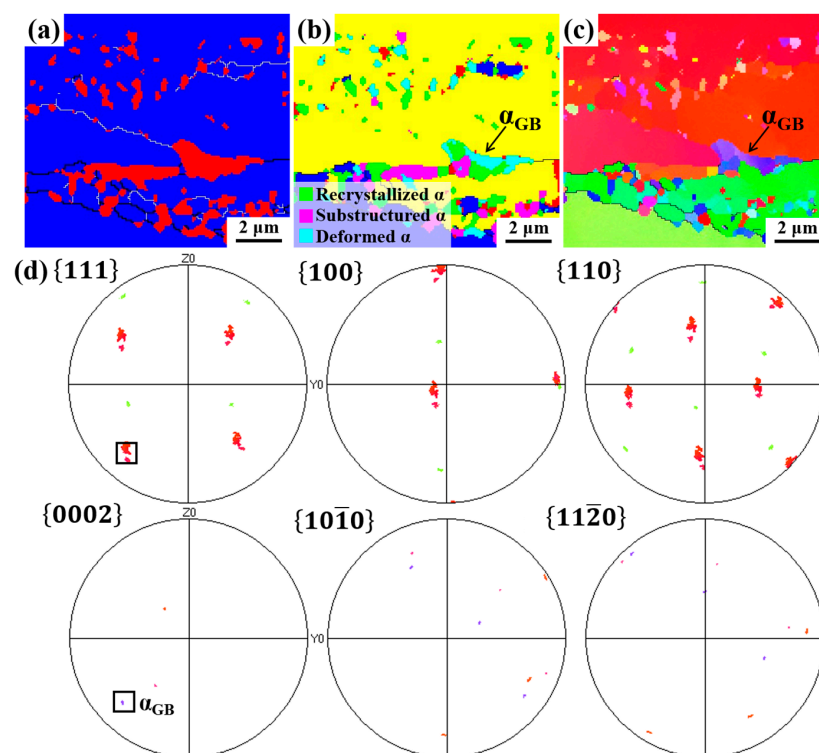


Figure 4. EBSD analysis for Ti-55531 alloy at a deformation temperature of 650 °C and a strain of 50%: (a) Phase distribution map; (b) recrystallization distribution map; (c) orientation distribution map; (d) pole figure maps for both β and α phases.

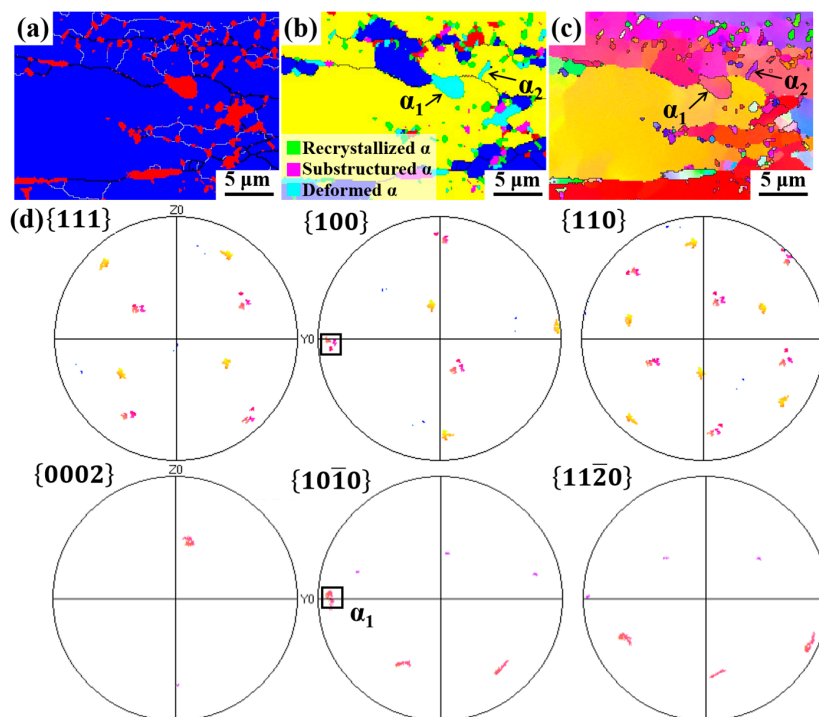


Figure 5. EBSD analysis for Ti-55531 alloy at a deformation temperature of 700 °C and a strain of 50%: (a) Phase distribution map; (b) recrystallization distribution map; (c) orientation distribution map; (d) pole figure maps for both β and α phases.

3.3. Microstructural Evolution During In Situ Tensile Testing

To characterize phase transformation under varied stress conditions, the in situ tensile test at 600 °C was performed on this alloy. The corresponding stress–displacement curve is presented in Figure 6. The initial tensile rate was set at 0.1 mm/min, and the rate was changed to 1 mm/min once the displacement reached approximately 3 mm.

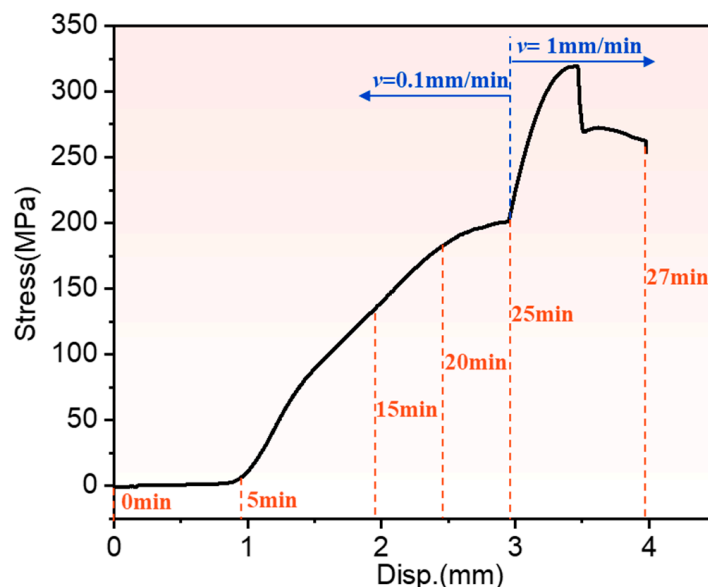


Figure 6. The in situ stress–displacement curve for Ti-55531 alloy at 600 °C.

Figure 7 shows the microstructure of the Ti-55531 alloy at different stages during in situ tension. The initial microstructure of the alloy consisted of β grains with an average size of about 180 μm . The α lamellae precipitate along β GBs or nucleate at boundaries and grow into the grain interior. At a deformation time of about 5 min, β GBs coarsening occurs, indicating the onset of α phase precipitation at the boundaries. At about 10 min, initial slip lines appear within the grains (marked in yellow, Figure 7c). As deformation progresses to 15 min (14.4% strain), slip lines become localized within a specific subset of grains. After 25 min, slip activity intensifies significantly, with parallel slip lines extending across the entire grains. At deformation termination, some slip lines continue to propagate, whereas others stop upon encountering β GBs or intragranular α lamellae. Notably, fewer slip lines appear in grains A and B, a feature possibly linked to grain orientation.

Before tensile deformation, α lamellae nucleating at β GBs extend into the grain interior, as evidenced by the phase distribution map (Figure 8a). The recrystallization distribution map (Figure 8b) reveals that partial β phase undergoes static recrystallization during heat treatment, as indicated by the blue regions. The α lamellae also recrystallize due to the undercooling induced by furnace cooling. The substructured α phase predominantly precipitates at GBs with limited inward growth. The nucleation sites of α phase exhibit selectivity, and not all GBs serve as preferential nucleation sites. Certain GBs show negligible α precipitation, indicating a strong crystallographic dependency of α nucleation on the matrix orientation. A combined analysis of α and β pole figures reveals that the $\{11\bar{2}0\}$ poles corresponding to α_{GB} , α_2 and α_3 coincide with the $\{111\}\beta$ pole figure (as marked), thereby confirming the $\{111\}\beta // \{11\bar{2}0\}\alpha$ orientation relationship. The α_1 fulfills the BOR of $\{0002\}\alpha // \{110\}\beta$, $\{11\bar{2}0\}\alpha // \{111\}\beta$.

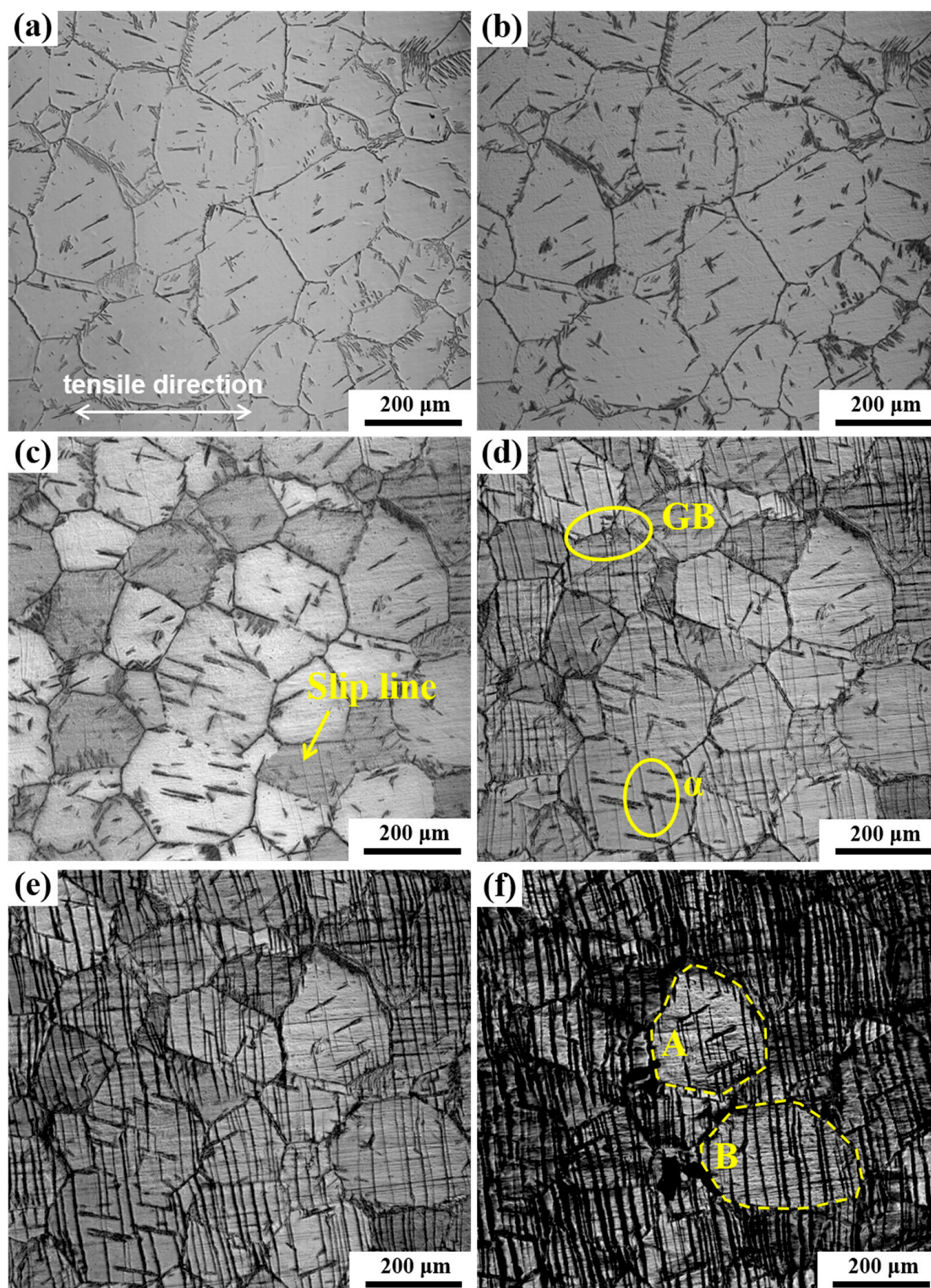


Figure 7. Microstructural evolution of Ti-55531 alloy during in situ tensile process at 600°C: (a) 0 min (undeformed microstructure); (b) 5 min; (c) 10 min; (d) 15 min; (e) 25 min; (f) 27 min.

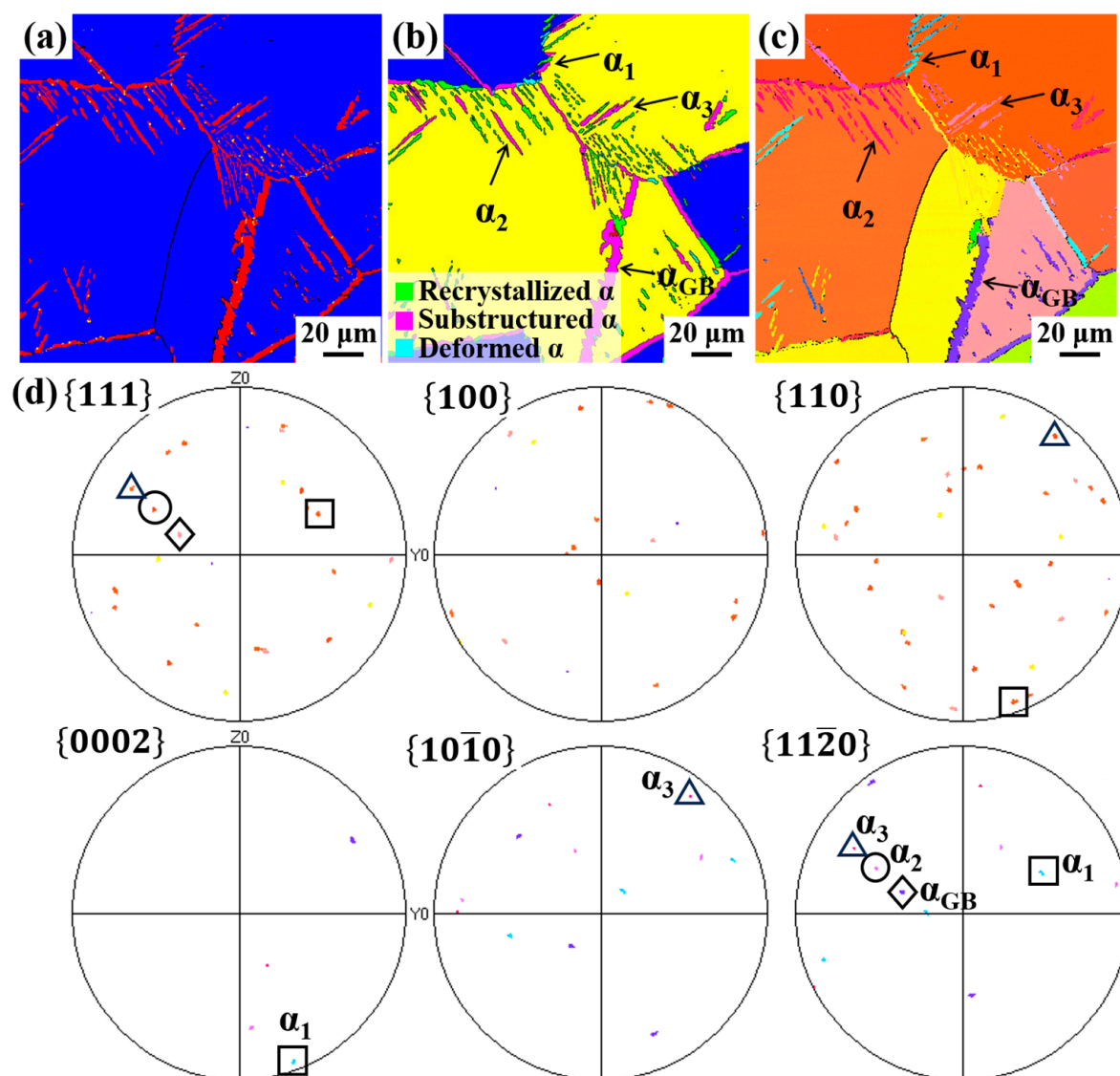


Figure 8. EBSD analysis for undeformed Ti-55531 alloy: (a) Phase distribution map; (b) recrystallization distribution map; (c) orientation distribution map; (d) pole figure maps for both β and α phases.

Figure 9 presents microstructural features of Ti-55531 alloy after tensile deformation. During thermomechanical deformation, parent β grains elongate mildly parallel to the tensile direction, and boundary-located α lamellae aggregate and interconnect along GBs (Figure 9a). Numerous lath-shaped, V-shaped and acicular α phases precipitate inside grains, occupying nearly the entire grain with a thickness of about 100 nm (Figure 9b). The resultant microstructure consists of coarse primary α lamellae embedded with refined α phase. Phase distribution mapping fails to resolve the detailed morphology of ultra-fine α , yet confirms complete α coverage over individual β grains (Figure 9c). In Figure 9d, regions of identical color merge into continuous domains, confirming texture development during tensile deformation. Cross-comparison between β and α pole figures (Figure 9e) identifies three coincident poles marked by rhombus, triangle and square symbols, corresponding to the $\{110\}\beta // \{0002\}\alpha$ orientation relationship.

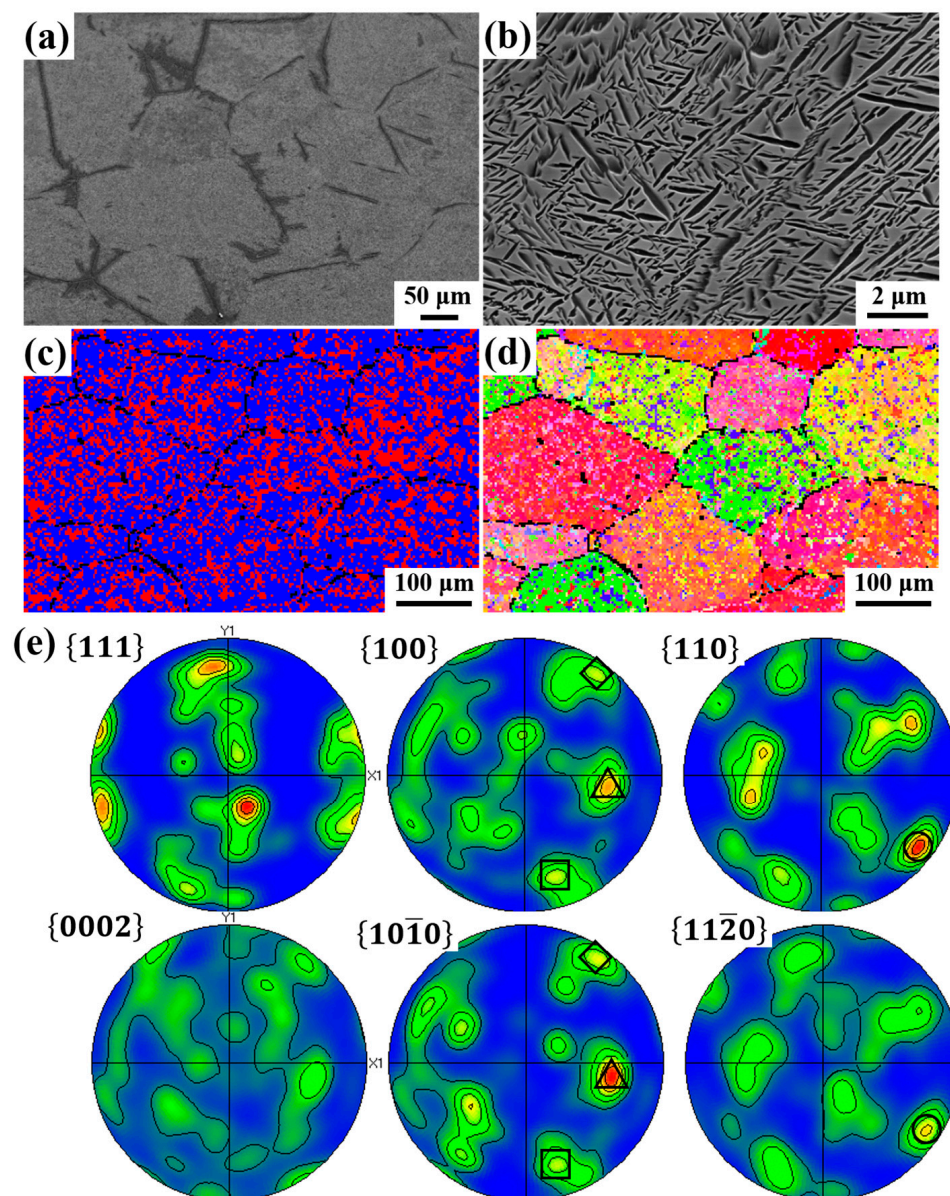


Figure 9. Microstructure analysis for Ti-55531 alloy after tensile deformation: (a,b) SEM image of the microstructure; (c) phase distribution map; (d) orientation distribution map; (e) pole figure maps for both β and α phases.

4. Discussion

4.1. Mechanisms of Lamellar Microstructure Evolution During Hot Deformation

Microstructural evolution of lamellar Ti-55531 during hot compression originates from competitive interplay between work hardening and dynamic softening. First, continuous mutual competition between dislocation hardening and dynamic recovery creates steady minor stress oscillations. Dislocation multiplication dominates early deformation under the strain rate of 0.001 s^{-1} and lifts flow stress toward peak values, while simultaneous dislocation rearrangement and annihilation soften the β matrix. Subsequent dynamic recovery and dynamic recrystallization reduce flow stress [25,26]. Second, thermomechanical loading induces kinking, bending and fragmentation of α lamellae, with such structural modifications closely linked to localized stress concentration and shear band initiation [12,16]. These actions intermittently release local stress concentrations and thus produce repeated minor stress drops across the whole strain range. Third, dynamic precipitation of fine intragranular α precipitates continuously hinders dislocation motion. Temporary dislo-

cation pile-up at newly formed α particles slightly increases flow stress, and subsequent dislocation bypass induces local stress relaxation, which sustains curve fluctuations.

Severe deformation up to 50% introduces irregular serrated curvature to initially straight β GBs from nonuniform strain distribution. Modified boundaries supply extra nucleation sites for α precipitation. Abundant intragranular GBs simultaneously act as preferential nucleation substrates for interior α phase. Elevated temperature imposes dual influences on microstructural development. Accelerated atomic mobility at high thermal levels facilitates dynamic recrystallization and supports dislocation rearrangement alongside sub-boundary generation [18,27]. Improved thermal conditions reduce the thermodynamic driving force for α precipitation, with a consequent decrease in the α volume fraction.

4.2. Precipitation Behavior and Nucleation Site Preference of the α Phase

The precipitation of the α phase exhibits preferential nucleation behavior. At low strain levels, nucleation predominantly occurs at β high-angle grain boundaries (HAGBs), where the elevated interfacial energy effectively reduces the thermodynamic barrier. With increasing macroscopic strain, deformation-induced substructures, including subgrain boundaries and dislocations, provide additional heterogeneous sites, thereby facilitating intragranular α precipitation [28,29]. To clarify the respective roles of deformation-induced defects in promoting α -phase nucleation, EBSD was employed for further analysis, as shown in Figure 10. Dislocation density is the primary driving force for α nucleation. The precipitation density of α phase in high-strain regions is substantially higher than that in low-strain regions, confirming that stored strain energy governs the α nucleation. Dynamic recrystallization indirectly affects α precipitation. Although dynamic recrystallization grains contain low dislocation densities, HAGBs possess high interfacial energy and therefore remain favorable nucleation sites for α phase. Subgrain boundaries and LAGBs are also preferred nucleation sites, as they collectively account for over 60% of the total boundaries. The low-energy interfaces provided by such boundaries significantly facilitate nucleation of α phase. However, the nucleation behavior exhibits a clear temperature dependence. At 650 °C, about half of the α phase nucleates and grows in regions with high dislocation density within the grains, rather than at GBs. At 700 °C, α precipitation occurs almost exclusively at GBs. This transition arises from the substantial decrease in intragranular stored energy at elevated temperatures due to enhanced dynamic recovery, which forces nucleation to rely on the interfacial energy supplied by GBs to overcome the critical nucleation barrier.

In addition, there is a strict crystallographic correspondence between the precipitating phase α and the matrix β . The nucleation and growth of the α phase proceed only when the crystallographic orientation of the β phase satisfies the BOR. Nevertheless, substantial lattice rotation during continuous plastic flow can disrupt the initial crystallographic alignment, causing precipitates in high-strain regions to deviate from the strict BOR [30]. The deviation from the ideal BOR is shown in Figure 11. Due to the severe fragmentation of the α phase, the analysis was conducted on regions of similar α orientations. The results show that all compressed alloys exhibit deviation from the ideal BOR, with the deviation being more severe at 700 °C than at 650 °C. At 650 °C, the overall lattice rotation remains restricted, and the α lamellae maintain orientations with a minor deviation from the ideal BOR. At 700 °C, however, the α phase breaks down into much finer and isolated globular particles. Without the orientational constraint of continuous lamellae, these particles can easily rotate under deformation stress, which further widens their misorientation with the surrounding β matrix.

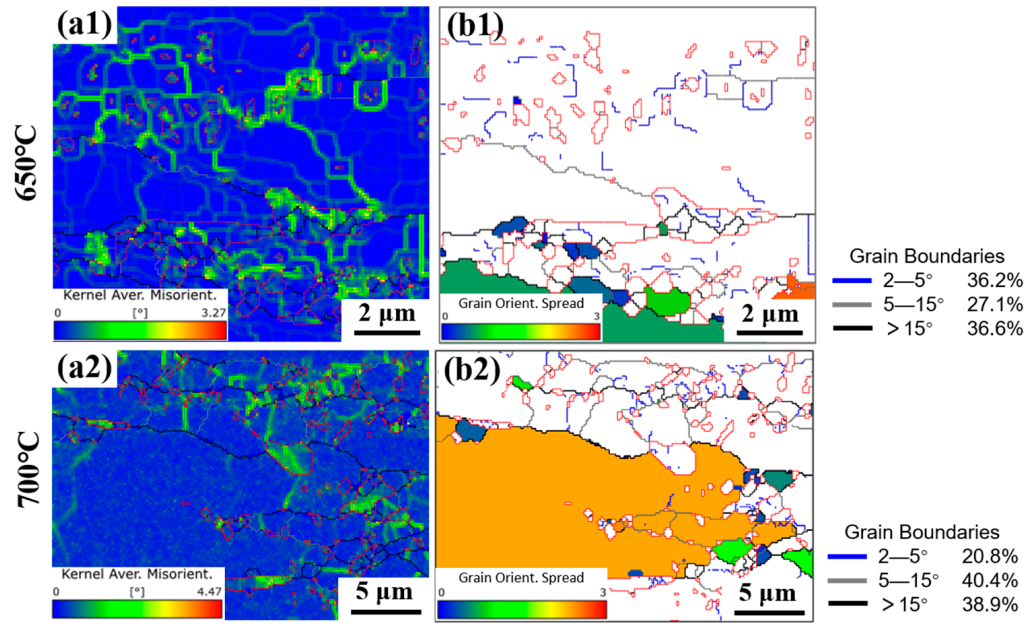


Figure 10. EBSD analysis for Ti-55531 alloy at deformation temperatures of 650 °C and 700 °C with a strain of 50%: (a1,a2) Kernel average misorientation (KAM)map; (b1,b2) grain orientation spread (GOS) map. The α/β phase boundaries, representing the nucleation sites of α phase, are marked in red. Subgrain boundaries (2–5°), low-angle grain boundaries (LAGBs, 5–15°), and HAGBs (>15°) are delineated in blue, gray, and black, respectively. Dynamic recrystallization β grains are identified by a GOS value of $\leq 3^\circ$.

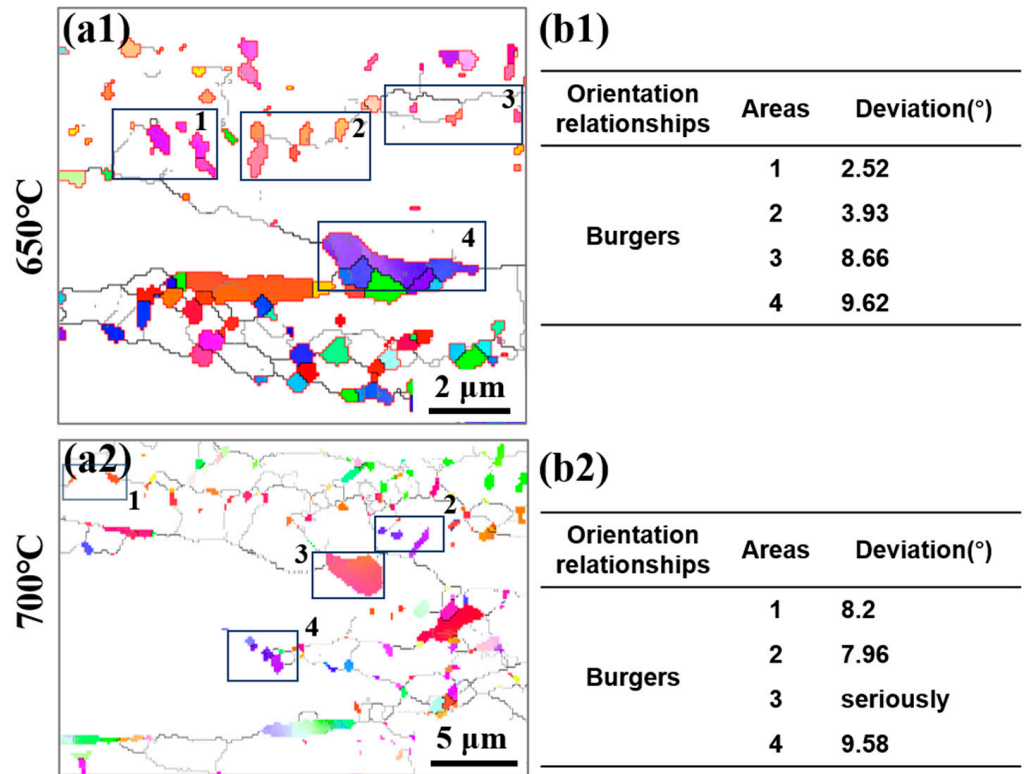


Figure 11. EBSD analysis for Ti-55531 alloy at deformation temperatures of 650 °C and 700 °C with a strain of 50%: (a1,a2) Orientation distribution map of α phase; (b1,b2) BOR deviation.

During in situ tensile loading, thermomechanical coupling effects become pronounced. Deformation-induced substructures act as potent nucleation sites while simultaneously

accelerating precipitate growth via the pipe diffusion of solute atoms. As a result, abundant plate-like, V-shaped, and acicular α precipitates form within the grains.

4.3. Slip Mechanism at Elevated Temperatures

In situ tensile testing elucidates the slip behavior of the lamellar microstructure during high-temperature tension. At the initial stage, slip systems are activated solely within soft oriented grains, confining slip traces to individual interiors. With increasing macroscopic strain, the critical resolved shear stress requirements are met in additional grains, promoting the participation of nearly all crystals in the plastic strain.

The capability of slip transmission across boundaries correlates with interface features [31–33]. The misorientation between adjacent grains dictates the deflection of the slip path, where LAGBs facilitate continuous transmission. The thicknesses of β GBs and α lamellae determine the barrier strength against dislocation motion. For thin α lamellae maintaining the BOR with the matrix, dislocations glide across the phase interfaces. Conversely, thicker α plates or unfavorable crystallographic orientations induce dislocation pile-ups, causing localized stress concentrations and terminating the slip traces.

5. Conclusions

The hot compression and in situ tensile behaviors of Ti55531 alloy with a lamellar microstructure were investigated. The primary findings regarding microstructural evolution and slip mechanism are summarized below.

(1) During hot compression at a strain rate of 0.001 s^{-1} , the flow stress curves display obvious work hardening and strain softening behavior. Peak stress values decrease with increasing deformation temperature, reaching approximately 140 MPa at 650 °C and 100 MPa at 700 °C. Elevated deformation temperature and increased strain accelerate the fragmentation of α lamellae alongside the bending and breaking of β GBs.

(2) During the hot deformation, the nucleation and growth sites of α precipitates are dependent on the deformation temperature. Nucleation can occur both in high-strain regions and at GBs. Both GB and intragranular α precipitates predominantly maintain the standard BOR characterized by $\{110\}\beta // \{0002\}\alpha$. The crystallographic orientation relationship between certain α precipitates and adjacent β grains is altered by severe plastic deformation.

(3) The in situ high-temperature tension elucidates the active slip mechanisms governing the lamellar structure. Slip traces are initially confined within individual grains at the early deformation stage, but with further straining they spread across the majority of grains. The transmission of these slip lines across boundaries depends on the thickness of the β GBs and the crystallographic orientation relationship between adjacent grains.

Author Contributions: Conceptualization, Y.W. and X.H.; methodology, C.L.; validation, W.Z.; formal analysis, Y.W. and X.H.; investigation, C.L.; resources, W.Z.; data curation, C.L. and Y.W.; writing—original draft preparation, C.L.; writing—review and editing, W.Z. and X.H. All authors have read and agreed to the published version of the manuscript.

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Data Availability Statement: The original contributions presented in this study are included in the article. Further inquiries can be directed to the corresponding authors.

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

Abbreviations

The following abbreviations are used in this manuscript:

BOR	Burgers orientation relationship
GBs	Grain boundaries
SEM	Scanning electron microscope
EBSD	Electron backscatter diffraction
HAGBs	High-angle grain boundaries
LAGBs	Low-angle grain boundaries
KAM	Kernel average misorientation
GOS	Grain orientation spread

References

1. Boyer, R.R. An overview on the use of titanium in the aerospace industry. *Mater. Sci. Eng. A* **1996**, *213*, 103–114. [\[CrossRef\]](#)
2. Banerjee, D.; Williams, J.C. Perspectives on titanium science and technology. *Acta Mater.* **2013**, *61*, 844–879. [\[CrossRef\]](#)
3. Zhao, Q.; Sun, Q.; Xin, S.; Chen, Y.; Wu, C.; Wang, H.; Xu, J.; Wan, M.; Zeng, W.; Zhao, Y. High-strength titanium alloys for aerospace engineering applications: A review on melting–forging process. *Mater. Sci. Eng. A* **2022**, *845*, 143260. [\[CrossRef\]](#)
4. Zhuo, L.; Zhan, M.; Xie, Y.; Chen, B.; Ji, K.; Wang, H. Recent advances in near- β titanium alloys: Microstructure control, deformation mechanisms, and oxidation behavior. *Adv. Eng. Mater.* **2024**, *26*, 2401837. [\[CrossRef\]](#)
5. Tan, C.; Yang, T.; Huang, C.; Zhang, J.; Wang, X.; Lu, H.; Wen, L.; Zhang, G. Enhanced strength–ductility synergy in Ti55531 titanium alloys through gradient microstructural design strategy. *Mater. Sci. Eng. A* **2024**, *909*, 146823. [\[CrossRef\]](#)
6. Li, Z.Y.; Wu, G.Q.; Huang, Z. Relationships between Microstructure and Mechanical Properties of Ti-5Al-5Mo-5V-3Cr-1Zr Alloy. *Mater. Res. Express* **2018**, *5*, 036518. [\[CrossRef\]](#)
7. Hao, M.; Wang, Y.; Li, P.; Zhang, T.; Zhu, J.; Wang, D. Heterogeneous microstructure enhanced comprehensive mechanical properties in titanium alloys. *JOM* **2021**, *73*, 3082–3091. [\[CrossRef\]](#)
8. Shen, J.; Zhang, C.; Pan, Z.; Jensen, D.J.; Yu, T. Hetero-lamellar structure design for enhanced strength and ductility of Ti enabled via additive manufacturing. *Mater. Res. Lett.* **2026**, *14*, 29–37. [\[CrossRef\]](#)
9. Qin, D.; Li, Y. The role of microstructure and stress state in dynamic mechanical behavior of Ti-5Al-5V-5Mo-3Cr alloy. *Mater. Charact.* **2019**, *147*, 421–433. [\[CrossRef\]](#)
10. Jianchong, L.; Hai, N.; Jiaqi, Z.; Dong, H.; Shukui, L.; Qunbo, F.; Dongmei, H. Effects of microstructural factors on adiabatic shear behaviors of Ti6441 alloys. *Rare Met. Mater. Eng.* **2013**, *42*, 1105–1107. [\[CrossRef\]](#)
11. Liu, Y.X.; Chen, W.; Li, Z.Q.; Liu, L.L.; Liu, D. In Situ observation on the deformation behavior of primary α -Ti in a textured Ti-6Al-4V. *MSF* **2020**, *993*, 365–373. [\[CrossRef\]](#)
12. Long, W.; Ou, M.; Mao, X.; Liang, Y. In situ deformation behavior of TC21 titanium alloy with different α morphologies (equiaxed/lamellar). *Rare Met.* **2021**, *40*, 1173–1181. [\[CrossRef\]](#)
13. Chen, W.; Yang, S.; Lin, Y.C.; Shi, S.; Chen, C.; Zhang, X.; Zhou, K. Cyclic deformation responses in α -phase of a lamellar near β -Ti alloy. *Mater. Sci. Eng. A* **2020**, *796*, 139994. [\[CrossRef\]](#)
14. Chuan, W.; Liang, H. Hot Deformation and dynamic recrystallization of a near-beta titanium alloy in the β single phase region. *Vacuum* **2018**, *156*, 384–401. [\[CrossRef\]](#)
15. Tan, C.; Yang, T.; He, J.; Dang, Q.; Lu, H.; Wen, L.; Zhang, G. Influence of electropulsing treatment and step-quench process on the gradient structure evolution of Ti55531 alloy. *J. Mater. Eng. Perform.* **2025**, *34*, 12651–12662. [\[CrossRef\]](#)
16. Zhou, Y.; Zhou, J.; Shu, Q.; Li, S.; Gongye, F.; Long, S.; Deng, H. Deformation mechanism and constitutive consideration for Ti-5Al-5Mo-5V-3Cr-1Zr alloy compressed at elevated temperatures. *J. Mater. Eng. Perform.* **2020**, *29*, 5104–5113. [\[CrossRef\]](#)
17. Wu, C.; Huang, L.; Li, C.M. Experimental investigation on dynamic phase transformation and texture evolution of Ti55531 high strength titanium alloy during hot compression in the $\alpha + \beta$ region. *Mater. Sci. Eng. A* **2020**, *773*, 138851. [\[CrossRef\]](#)
18. Qin, D.; Guo, D.; Zheng, L.; Li, Y. Dynamic Recrystallization of Ti-5553 alloy during sub-transus thermomechanical processing: Mechanisms and its role in formation of a bi-modal structure. *J. Alloys Compd.* **2018**, *769*, 725–731. [\[CrossRef\]](#)
19. Wang, K.; Li, M.Q. Effects of heat treatment and hot deformation on the secondary α phase evolution of TC8 titanium alloy. *Mater. Sci. Eng. A* **2014**, *613*, 209–216. [\[CrossRef\]](#)
20. Wang, Y.; Chu, S.; Zhang, X.; Sun, W.; Zhang, Q.; Liu, Q.; Dar, S.M.; Zhou, B.; Sun, J.; Liang, G.; et al. In-situ investigation of the slip activities, deformation twinning, and cracking behavior of a bimodal Ti-6Al-3Nb-2Zr-1Mo alloy. *Mater. Sci. Eng. A* **2026**, *953*, 149705. [\[CrossRef\]](#)
21. Zhang, X.; Zhang, S.; Zhao, Q.; Zhao, Y.; Li, R.; Zeng, W. In-situ observations of the tensile deformation and fracture behavior of a fine-grained titanium alloy sheet. *J. Alloys Compd.* **2018**, *740*, 660–668. [\[CrossRef\]](#)

22. Hu, M.; Dong, L.; Zhang, Z.; Lei, X.; Yang, R.; Sha, Y. Correction of flow curves and constitutive modelling of a Ti-6Al-4V alloy. *Metals* **2018**, *8*, 256. [[CrossRef](#)]
23. Ebrahimi, R.; Najafizadeh, A. A new method for evaluation of friction in bulk metal forming. *J. Mater. Process. Tech.* **2004**, *152*, 136–143. [[CrossRef](#)]
24. Haihao, T.; Yufeng, X.; Tao, S.; Deyu, Z.; Lei, C. Flow stress prediction of near- β Ti-55511 alloy during iso-thermal compression based on corrected Arrhenius model with material parameter evolution and BP-ANN model. *Rare Met. Mater. Eng.* **2023**, *52*, 823–833.
25. Luo, J.; Gao, J.; Li, L.; Li, M.Q. The flow behavior and the deformation mechanisms of Ti-6Al-2Zr-2Sn-2Mo-1.5Cr-2Nb alloy during isothermal compression. *J. Alloys Compd.* **2016**, *667*, 44–52. [[CrossRef](#)]
26. Nan, Y.; Ning, Y.; Liang, H.; Guo, H.; Yao, Z.; Fu, M.W. Work-hardening effect and strain-rate sensitivity behavior during hot deformation of Ti-5Al-5Mo-5V-1Cr-1Fe alloy. *Mater. Des.* **2015**, *82*, 84–90. [[CrossRef](#)]
27. Liang, H.; Guo, H.; Ning, Y.; Peng, X.; Qin, C.; Shi, Z.; Nan, Y. Dynamic recrystallization behavior of Ti-5Al-5Mo-5V-1Cr-1Fe alloy. *Mater. Des.* **2014**, *63*, 798–804. [[CrossRef](#)]
28. Li, C.; Zhang, X.; Li, Z.; Zhou, K.-C. Hot deformation of Ti-5Al-5Mo-5V-1Cr-1Fe near β titanium alloys containing thin and thick lamellar α phase. *Mater. Sci. Eng. A* **2013**, *573*, 75–83. [[CrossRef](#)]
29. Li, M.; Xu, J.; Zhu, L.; Wang, X.; Li, X.; Shang, G.; Tao, C.; Yang, J.; Zhu, Z.; Zhao, X. Hot deformation behavior and globularization mechanism of the lamellar α phase in a new low-cost titanium alloy. *Mater. Sci. Eng. A* **2026**, *960*, 150108. [[CrossRef](#)]
30. Wang, D.; Fan, Q.; Yang, L.; Gong, H.; Yuan, J.; Chen, K.; Zhu, X.; Cheng, X.; Zhou, Z. Effects of rolling reduction on Burgers orientation relationship and slip behavior of a Ti-5.5Mo-7.2Al-4.5Zr-2.6Sn-2.1Cr alloy. *J. Mater. Res. Technol.* **2021**, *15*, 3099–3109. [[CrossRef](#)]
31. Seal, J.R.; Crimp, M.A.; Bieler, T.R.; Boehlert, C.J. Analysis of slip transfer and deformation behavior across the α/β interface in Ti-5Al-2.5Sn (wt.%) with an equiaxed microstructure. *Mater. Sci. Eng. A* **2012**, *552*, 61–68. [[CrossRef](#)]
32. Zhao, P.; Shen, C.; Savage, M.F.; Li, J.; Niezgoda, S.R.; Mills, M.J.; Wang, Y. Slip transmission assisted by shockley partials across α/β interfaces in Ti-alloys. *Acta Mater.* **2019**, *171*, 291–305. [[CrossRef](#)]
33. Guo, Y.; Britton, T.B.; Wilkinson, A.J. Slip band–grain boundary interactions in commercial-purity titanium. *Acta Mater.* **2014**, *76*, 1–12. [[CrossRef](#)]

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