



***In-situ* observation of Ni-Co based wrought superalloy high-temperature deformation: lattice rotation and grain boundary response**

Yingbo Bai, Rui Zhang, Chuanyong Cui, Yizhou Zhou & Xiaofeng Sun

To cite this article: Yingbo Bai, Rui Zhang, Chuanyong Cui, Yizhou Zhou & Xiaofeng Sun (2024) *In-situ* observation of Ni-Co based wrought superalloy high-temperature deformation: lattice rotation and grain boundary response, Materials Research Letters, 12:11, 869-876, DOI: [10.1080/21663831.2024.2396033](https://doi.org/10.1080/21663831.2024.2396033)

To link to this article: <https://doi.org/10.1080/21663831.2024.2396033>



© 2024 The Author(s). Published by Informa UK Limited, trading as Taylor & Francis Group.



[View supplementary material](#)



Published online: 30 Aug 2024.



[Submit your article to this journal](#)



Article views: 1882



[View related articles](#)



[View Crossmark data](#)



Citing articles: 15 [View citing articles](#)



REPORT



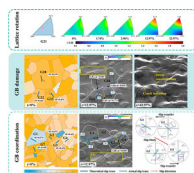
In-situ observation of Ni-Co based wrought superalloy high-temperature deformation: lattice rotation and grain boundary response

Yingbo Bai^{a,b}, Rui Zhang^a, Chuanyong Cui^a, Yizhou Zhou^a and Xiaofeng Sun^a

^aShi-changxu Innovation Center for Advanced Materials, Institute of Metal Research, Chinese Academy of Sciences, Shenyang, People's Republic of China; ^bSchool of Materials Science and Engineering, University of Science and Technology of China, Shenyang, People's Republic of China

ABSTRACT

The sensitivity of grain boundary (GB) cracks often limits the high-temperature plasticity of superalloys. The 750 °C tensile deformation of fine-grained Ni-Co based wrought superalloys was observed in situ using a scanning electron microscope (SEM) equipped with an electron backscatter diffraction (EBSD) probe to clarify the texture formation and GB response. The GB stress concentration depends on the grain orientation and slip system alignment on both sides. This reflects the ability of GBs to cope with deformation incompatibility. This study provides valuable insights for predicting the failure of polycrystalline superalloys and offers new ideas for microstructure evaluation and GB design.



IMPACT STATEMENT

The lattice rotation and GB response of Ni-Co based fine-grained wrought superalloys under 750 °C deformation were explained through in-situ observation, which helps predict the superalloys' service behavior.

ARTICLE HISTORY

Received 9 May 2024

KEYWORDS

Superalloy; *In-situ*; grain boundaries; slip; high-temperature deformation

1. Introduction

Ni-based wrought superalloys exhibit exceptional properties within the range of 700 ~ 850 °C, rendering them an ideal material for manufacturing turbine discs in aircraft engines [1–4]. In view of the reliability of components, the deformation failure mechanism of alloys within the operating temperature range has gained increasing attention. Numerous studies have reported the initiation of GB cracks in Ni-based polycrystalline superalloys at high temperatures, both in tensile and fatigue testing [5–7]. The intensification of atomic motion weakens GBs at high temperatures, thus nullifying the Hall-Petch effect [8,9]. Further research is required to enhance the coordination of GBs to reduce the probability of intergranular cracking, despite the

GBs weakening at high temperatures being confirmed [10–12].

During the plastic deformation process of polycrystalline materials, lattice rotation and GB coordination commonly occur [13,14]. In order for slip deformation to persist, each grain must conform to the constraint of external loading and the surrounding grains. Hence, lattice rotation is necessary to accommodate this deformation incompatibility [13,15]. As a result of changes in lattice orientation, deformation texture is formed following plastic deformation, accompanied by deformation mismatch at GBs. This can result in the accumulation of geometrically necessary dislocations (GNDs) at GBs, ultimately leading to GB failure [16,17]. While fine-grained structures benefit the room temperature strength and hot

CONTACT Rui Zhang ✉ rzhang@imr.ac.cn Shi-changxu Innovation Center for Advanced Materials, Institute of Metal Research, Chinese Academy of Sciences, Shenyang, 110016, People's Republic of China; Chuanyong Cui ✉ chycui@imr.ac.cn Shi-changxu Innovation Center for Advanced Materials, Institute of Metal Research, Chinese Academy of Sciences, Shenyang, 110016, People's Republic of China

Supplemental data for this article can be accessed at <https://doi.org/10.1080/21663831.2024.2396033>.

© 2024 The Author(s). Published by Informa UK Limited, trading as Taylor & Francis Group.

This is an Open Access article distributed under the terms of the Creative Commons Attribution-NonCommercial License (<http://creativecommons.org/licenses/by-nc/4.0/>), which permits unrestricted non-commercial use, distribution, and reproduction in any medium, provided the original work is properly cited. The terms on which this article has been published allow the posting of the Accepted Manuscript in a repository by the author(s) or with their consent.

working performance of Ni-based wrought superalloys used for turbine discs, an increase in the number of GBs has led to a potential sharp drop in high-temperature plasticity. Numerous studies have reported low plasticity at service temperatures, with the initiation and propagation of GB cracks as the primary forms of failure [1,18–22]. The incompatibility of GB deformation continues to hinder the service performance of these alloys. Researchers have described the slip interaction based on lattice geometric matching [23,24]. Still, the further development of GB deformation incompatibility and the response mechanism leading to failure have yet to receive attention. Especially for fine-grained superalloys, in-situ observation of their deformation process remains challenging. This study aims to address this issue by revealing the deformation response of GBs at service temperature.

Utilizing *in-situ* SEM/EBSD technology, this research delved into the tensile deformation of Ni-Co based wrought superalloys at 750 °C. This investigation examined the lattice rotation and GB response of fine-grained structures under high-temperature deformation. Ultimately, the causes of GB stress concentration and the GB-mediated slip law were revealed. This study offers valuable insights into predicting the GB deformation response of Ni-based wrought superalloys, emphasizing the importance of considering GB-mediated slip in microstructure evaluation.

2. Materials and methods

This study used a novel Ni-Co based wrought superalloy (13Cr, 25Co, 6Ti, 3Al, 1.2Mo, 3W, 0.04C by mass %) for aerospace turbine discs. The *in-situ* tensile sample was bone-shaped with double chamfers, cut from forged billets, and had a thickness of 1.2 mm, as shown in Figure 1 (a). After being mechanically polished and then electrolytically polished in a solution of 20 vol% H₂SO₄+80 vol% CH₃OH at 25 V for 25 s, the sample was tensile deformed in a vacuum environment at 750 °C until it failed, with a rate of 2 μm/s. The *in-situ* tensile curve recorded by displacement sensors was corrected based on the elastic modulus of the reference tensile curve [25]. The final engineering strain of the *in-situ* sample was 43%. The SEM (Apreo 2s) with an EBSD probe was used to observe the deformation process of the sample in situ shown in Fig. S1 in the supplementary materials. EBSD data and SEM images were collected at strains of 1.74%, 2.96%, 1.97%, and 22.97%, and SEM images were taken separately at strains of 32.97% and 42.97%. The EBSD data was analyzed using OIM software and Matlab with MTEX code, and the theoretical slip trace of a single grain was plotted. The face-centered cubic (FCC) crystal has twelve equivalent slip systems, numbered as shown in Table 1. The numbering of grains used for

Table 1. Twelve possible slip systems of the experimental alloy.

No.	Slip plane	Slip direction	No.	Slip plane	Slip direction
γ_1	(1 1 1)	[0 1 $\bar{1}$]	γ_7	($\bar{1}$ 1 1)	[0 1 $\bar{1}$]
γ_2	(1 1 1)	[$\bar{1}$ 0 1]	γ_8	($\bar{1}$ 1 1)	[1 0 1]
γ_3	(1 1 1)	[1 $\bar{1}$ 0]	γ_9	($\bar{1}$ 1 1)	[1 1 0]
γ_4	(1 1 $\bar{1}$)	[1 $\bar{1}$ 0]	γ_{10}	(1 $\bar{1}$ 1)	[$\bar{1}$ 0 1]
γ_5	(1 1 $\bar{1}$)	[1 0 1]	γ_{11}	(1 $\bar{1}$ 1)	[1 1 0]
γ_6	(1 1 $\bar{1}$)	[0 1 1]	γ_{12}	(1 $\bar{1}$ 1)	[0 1 1]

research can be found in supplementary material Fig. S1 (a). The as-received alloy had a fine-grained structure (average grain size $\approx 5.4 \mu\text{m}$) with some large-sized L1₂ structure primary γ' phases on the GBs, which can be distinguished during data analysis.

3. Results and discussion

Figure 1 (b) and (c) depict the color mapping of EBSD inverse pole figures (IPF) in the initial stage and at 22.97% strain, respectively. The deformed grains were elongated along the loading direction (LD) and displayed signs of lattice rotation. The normalized intensity pole figures (Figure 1 (d)) and LD inverse pole figures (Figure 1 (e)) more clearly demonstrate the overall orientation distribution changes under varying strains. The initial orientation distribution was relatively discrete, and this shape persisted during the elastic deformation stage. Following the onset of plastic deformation, most crystal orientations concentrated towards $\langle 111 \rangle$, and the pole intensity gradually increased as deformation progressed, ultimately forming a $\langle 111 \rangle$ tensile deformation texture.

Parallel to LD, many metals with an FCC structure demonstrate a rotation of the most deformed grains towards $\langle 111 \rangle$, which is likely connected to the crystals' slip characteristics [26]. Due to the orientation relationship, the slip plane of FCC grains at the $\langle 111 \rangle$ pole is perpendicular to the direction of external forces, making it challenging for the slip system to initiate according to Schmidt's law. Hence, $\langle 111 \rangle$ is in a hard orientation position. Following deformation, a $\langle 111 \rangle$ texture forms, and the polycrystalline structure's average Schmid factors (SF) usually decreases. This suggests that the alloy's deformation potential is gradually consumed as the slip progresses.

To further delve into the laws of texture formation and lattice rotation, we analyzed the slip and deformation of specific grains. Three grains with different initial orientations and slip system activity are displayed in Figure 2 showing their slip and lattice rotation. The grains have original SF ranging from low to high. The original orientation of G1 was located near $\langle 111 \rangle$, and its SF was 0.374, making it difficult for its slip system to activate under external stress. Even during a large plastic

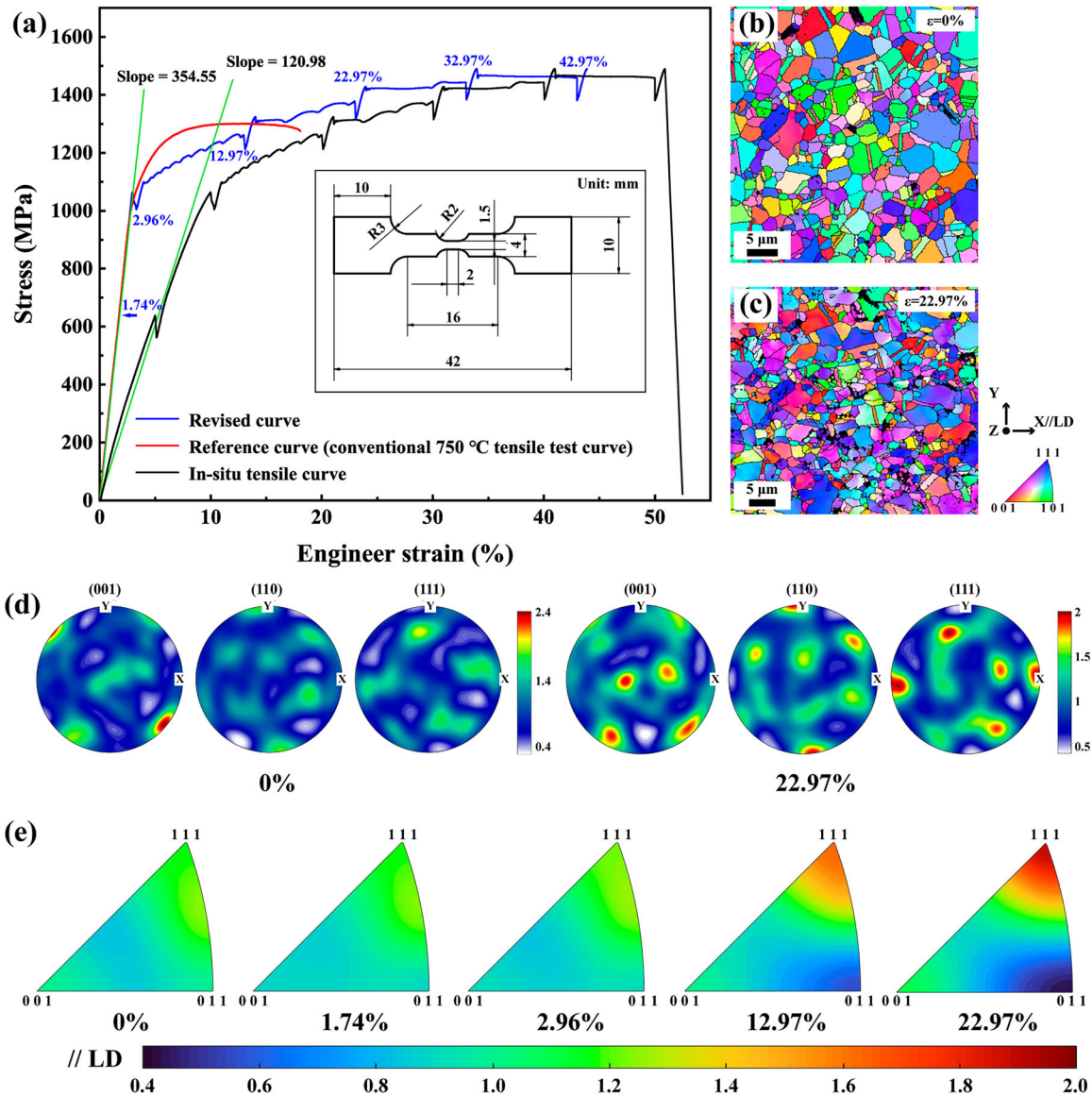


Figure 1. (a) *In-situ* tensile curves and sample dimensional drawing; IPF maps at in the initial stage (b) and at strain of 22.97% (c); Normalized intensity pole figures (d) and LD inverse pole figures (e) of different strains.

deformation stage, the degree of slip and deformation of G1 remained low, with its lattice rotation being more influenced and constrained by the surrounding deformed grains rather than dominated by its own slip. The inverse pole figures indicate that the pole intensity of G1 mainly swept around its original position without significant displacement (Figure 2 (d)). However, under large deformation, the slip system γ 10 with the maximum SF in G1 was activated (Figure 2 (c)). G20 and G21, on the other hand, had higher SF (0.420 and 0.496), enabling them to participate in plastic deformation earlier. The larger the SF, the earlier the slip system was activated (Figure 2(g, k)), and their lattice orientation continuously rotated with the slip. After the sample deformation reached 22.97%, the crystal cells of these two grains rotated away from their

initial positions. The inverse pole figures of G20 and G21 display the movement path of their pole intensity (Figure 2(h, l)).

The activation of polycrystalline slip cannot be determined solely by the SF condition. The deformation and rotation of grains are subject to complex constraints from their surroundings, making it impossible to predict the degree and direction of grain rotation directly. Interestingly, both G20 and G21 showed pole intensity movement close to the $\langle 111 \rangle$ direction on LD. The supplementary materials (Fig. S2) analyzed the lattice rotation angle and orientation distribution of 30 grains, but no specific trend was observed. Approximately 53% of the grains were oriented closer to the $\langle 111 \rangle$ pole after deformation (Fig. S2(e)). There is currently no

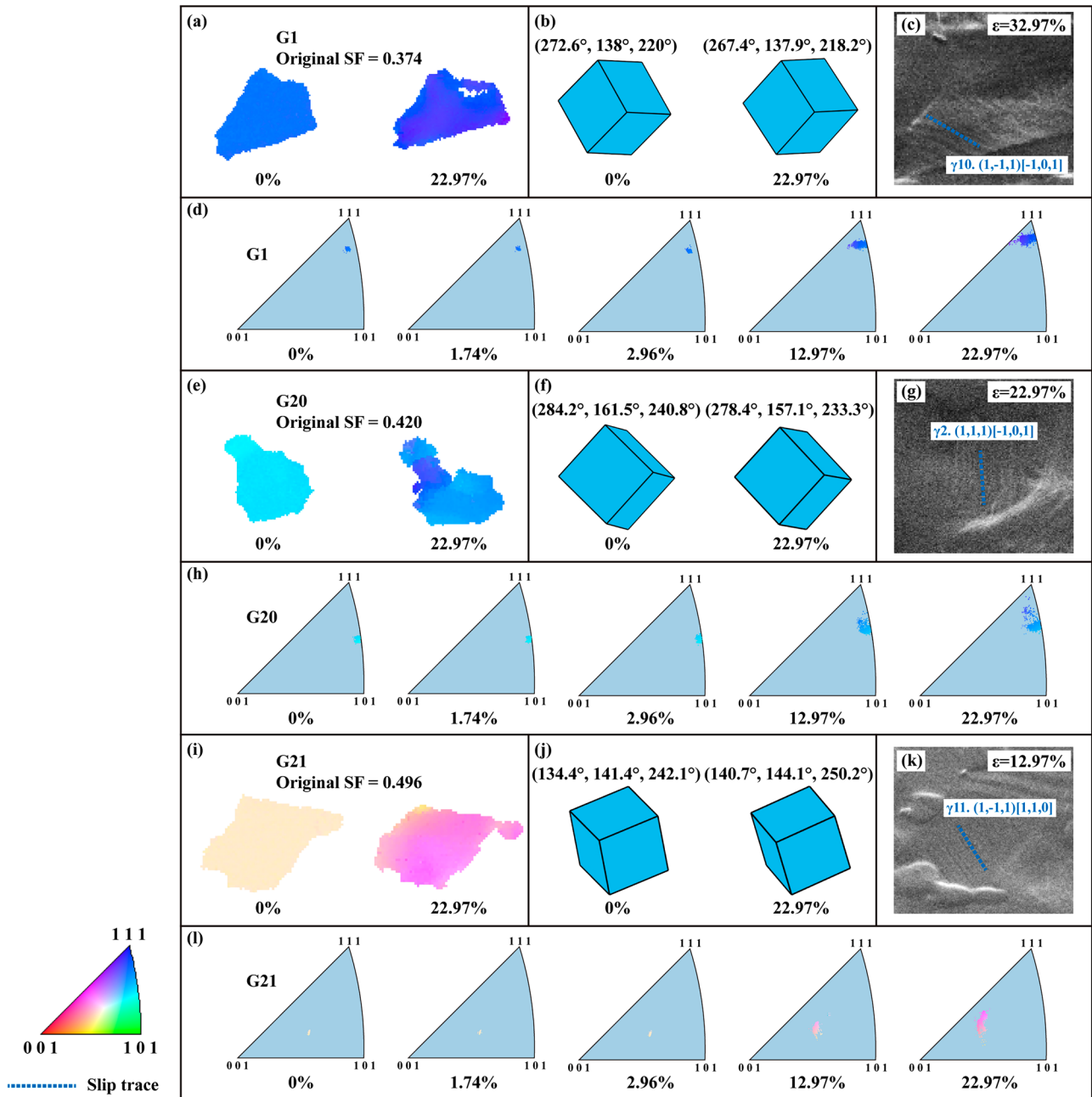


Figure 2. Orientation changes and slip of grains with different SF. (a-d) G1; (e-h) G20; (i-l) G21; (a, e, i) IPF maps of grains; (b, f, j) Cell models before and after deformation; (c, g, k) Traces of activated slip systems; (d, h, l) LD inverse pole figures of different strains.

evidence to suggest that the $\langle 111 \rangle$ pole (hard orientation position) was a specific ‘destination’ for lattice rotation. Moreover, the deviation path of the orientation does not align with the route model provided by existing research [27,28]. However, as a restricted position, grains located near the $\langle 111 \rangle$ direction were challenging to maintain slip activity and lattice rotation, and this orientation captured movable grains during plastic deformation. As a result, the pole intensity in the $\langle 111 \rangle$ direction increased continuously with deformation, and more and more grains entered a slip-restricted

state, forming a near-failure state of the $\langle 111 \rangle$ texture (Figs. S2 (a-c)).

When examining the behavior of grains, it is crucial to not only consider their individual slip and deformation but also their interactions with one another. The deformation response of GBs, as the primary carrier of lattice slip deformation incompatibility, reflects the plastic deformation ability of polycrystalline materials. However, in high-temperature environments, the strength of GBs may decrease, leading to potential GB failure [6,17]. It is generally believed that the difference in slip activation

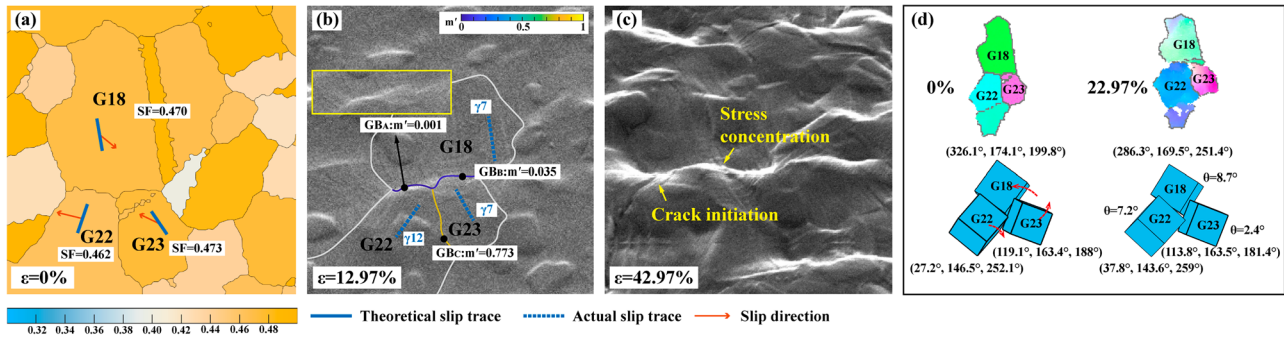


Figure 3. (a) SF maps and theoretical slip traces of undeformed grains G18, G22, and G23; (b, c) SEM images and slip traces of G18, G22, and G23 with different strains; (d) IPF maps and cell models of G18, G22, and G23 before and after deformation.

degree between grains is the primary cause of accumulating GB stress concentration [16,29]. To clarify the formation mechanism of high-stress concentration interfaces, we initially observed the GBs exhibiting apparent deformation incompatibility.

In situations where GB-mediated slip was not an option, as illustrated in Figure 3, a visibly raised and wavy interface appears in the near failure stage, ultimately becoming the favored location for both crack nucleation and propagation. Despite expectations of a deformation mismatch between soft and hard-oriented grains, a significant concentration of GB stress occurred between high SF grains G18, G22, and G23. All three of the maximum SF slip systems were activated during the early stages of plastic deformation, with slip traces intersecting at the trigonal GB. Calibration of the slip system confirmed that the conflict of activated slip system orientations caused stress accumulation in GBs. The degree of alignment of slip systems on both sides determines whether dislocations slip transfer through GB, which serves as an effective mechanism for releasing stress concentration through GB-mediated slip. The m' factor is used to evaluate the alignment degree of grain slip systems [23,30], which is defined as [31]:

$$m' = \cos\alpha \cdot \cos\beta \quad (1)$$

where α and β represent the angles between the normal direction of slip planes and the slip directions in adjacent grains, respectively. Nieto Valeira et al. [32] calculated that the conditions for slip transfer in FCC-structured pure Ni was $m' \geq 0.8$.

After analyzing the m' factor between the grains depicted in Figure 3, it was discovered that the slip system displays a completely incoherent orientation relationship, particularly for GB_A and GB_B , with m' values less than 0.05. Consequently, the slip on both sides cannot be coordinated, and the lattice maintains its rotational independence to adapt to slip and plastic deformation. GBs absorb lattice dislocations released by

slip, exacerbating intergranular deformation damage. All three grains underwent lattice rotation after deformation, and GB_A - GB_B was pulled in different directions by cell movement, resulting in a height difference (as shown in Figure 3 (d)). In contrast, GB_C with a higher m' value showed no significant stress concentration after deformation. The m' values of the high-stress concentration protrusion interface in the observation area and the SF of the grains on both sides were statistically analyzed and listed in the supplementary materials (Fig. S3 and Table. S1) throughout the deformation-failure process. The formation of all stress concentrations was similar to that in Figure 3. This indicates that both the SF and m' parameters affect the sensitivity of GB cracks, and the misalignment of slip systems isolated the deformation between grains, especially for high SF grain groups. The low m' value GB between soft-oriented grains was the easiest to accumulate GNDs, and the incompatible highly local deformation further compresses the dislocation tolerance of GB, leading to weak interfaces with high-stress concentration.

It has been observed that some GBs are better suited to plastic deformation due to their ability to avoid incompatibility issues through mediated slip. These interfaces tend to maintain a relatively flat appearance after deformation, in contrast to high-stress concentration GBs. In Figure 4(b and c), good slip transfer occurred between G1 and G2, as well as between G3 and G4, with slip traces aligned on both sides passing through GB (locally enlarged images (1) and (2)). The m' values of GB_D and GB_F are 0.899 and 0.969, respectively. The lattice and slip plane models of grains G1-G4 are shown in Figure 4(d). The slip planes on both sides of the high m' GBs are almost parallel, creating conditions for slip transfer between grains. The slip transfer achieved good coordination between adjacent grains, and GB_D and GB_F maintained a flat morphology after plastic deformation without prominent stress concentration characteristics. On the contrary, the GB_E ($m' = 0.023$) cannot

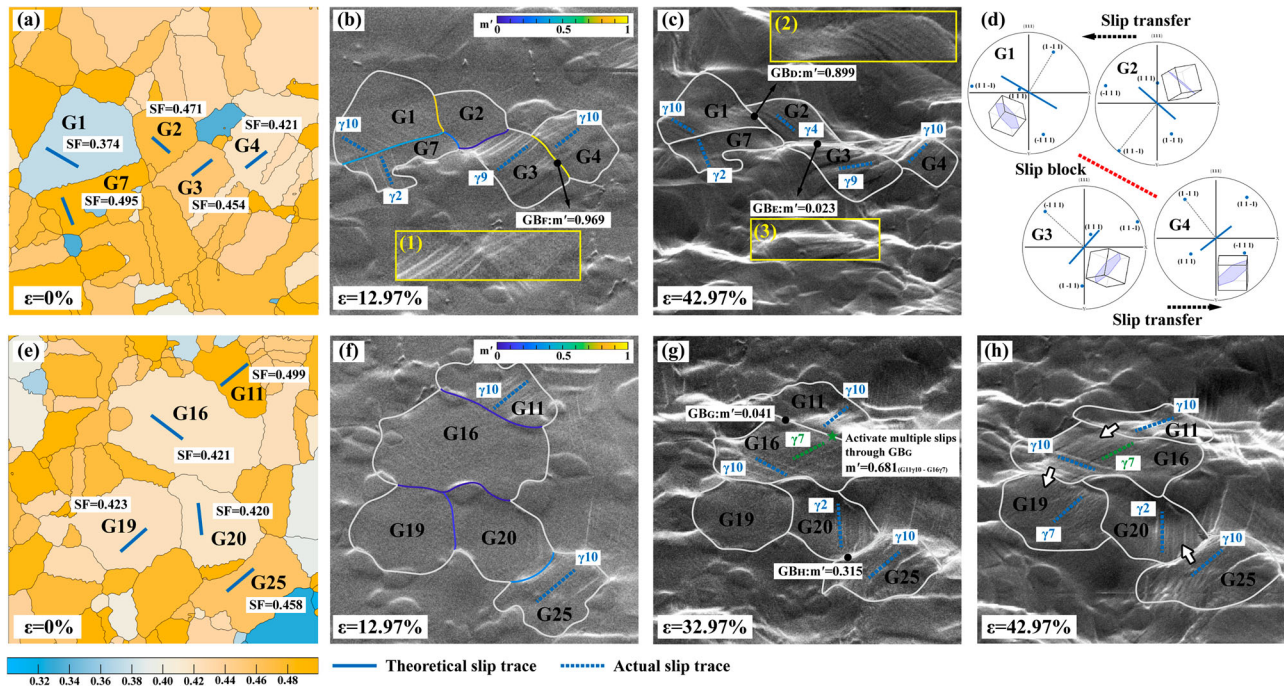


Figure 4. (a) SF maps and theoretical slip traces of undeformed grains G1-G4, and G7; (b, c) SEM images and slip traces of G1-G4, and G7 with different strains; (d) Slip plane positions of G1-G4 marked on the pole figures and lattices; (e) SF maps and theoretical slip traces of undeformed grains G11, G16, G19, G20, and G25; (f-h) SEM images and slip traces of G11, G16, G19, G20, and G25 with different strains.

achieve deformation coordination of G2 and G3, and the discontinuous deformation on both sides leads to GB arching, which is consistent with the previously described failure-prone GBs.

The slip transfer phenomenon is rare due to the limitations of lattice arrangement. According to statistics, the proportion of GBs with $m' \geq 0.8$ in the experimental alloy is less than 14%, which implies that most GBs require other mechanisms to coordinate grain deformation on both sides. Figure 4 (e-h) shows the continuous activation of slip between grains with misaligned slip systems. In the early stage of plastic deformation, the slip systems of high SF grains G11 and G25 initiated first, and slip block was formed at the low m' GB_F and GB_H. The two GBs exhibited stress concentration and height difference due to the slip and rotation of grains on one side. The stress concentration of GB provided the impetus for the movement of dislocations in grains with deformation the other side, essentially the absorption and re-emission of lattice dislocations by GB.

In G20, the maximum SF slip system $\gamma 2$ was activated by stress concentration at GB_H, and the dislocation slip started from GB. The dislocations released by GB slipped inward into the grain (Figure 3 (g)), indicating that G20 and G25 completed deformation coordination at the cost of a certain degree of intergranular stress concentration. Similarly, the stress concentration at the trident GB between G16, G19, and G20 activated the maximum SF

slip system $\gamma 7$ of G19 in the later stage of deformation (Figure 4 (h)). Specifically, the activation of stress blocking was also affected by the alignment degree of the slip systems. When the deformation of the sample reached 32.97%, the stress concentration at GB_G activated the $\gamma 7$ of G16, with an SF of only 0.288. Compared to the maximum SF slip system $\gamma 10$, the orientation relationship of $\gamma 7$ in G16 was more aligned with the slip system produced by G11 (with a $m' = 0.681$ between the introduced slip system). Hence, when examining GB-mediated slip, the m' factor has a more significant impact than SF. The shear stress inside the lattice ($\gamma 10$) and the dislocation release mediated by GB_G ($\gamma 7$) lead to the activation of different slip systems in G16. It is crucial to note that such multiple slips are rare in alloys, with the vast majority of observed grains having only one maximum SF slip system activated. Unlike other studies on Ni-based superalloys' slip characteristics [23,27], the lower slip motion tolerance of the fine-grained structure leads to the deformation activity mainly being single slip, ensuring the feasibility and reliability of alloy structure prediction.

4. Conclusions

In summary, we conducted *in-situ* SEM/EBSD observations of the tensile deformation process at 750 °C in a Ni-Co based wrought superalloy. The $\langle 111 \rangle$ orientation restricts the slip and rotation of grains, and most

grains were captured by $\langle 111 \rangle$, which was the cause of the deformation texture. Furthermore, stress concentration was primarily observed at the GBs between soft-oriented grains with slip system misalignment, where low m' values GBs were incapable of coordinating the slip impact on both sides. Conversely, when slip systems were well-aligned, the GBs could avoid deformation incompatibility by mediating slip, even if there was a slip activity difference between the grains on both sides.

Disclosure statement

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

Funding

This work was supported by the National Key R&D Program of China (Nos. 2019YFA0705300), the National Science and Technology Major Project of China (No. 2019-VI-0006-0120), the Youth Innovation Promotion Association, CAS (No. 2023202), and the Natural Science Foundation Project of Liaoning Province (No. 2023-MS-024).

Data availability statement

Data available within the article or its supplementary materials.

References

- [1] Yuan Y, Gu YF, Osada T, et al. Deformation mechanisms in a new disc superalloy at low and intermediate temperatures. *Scripta Mater.* 2012;67(2):137–140. doi:10.1016/j.scriptamat.2012.03.042
- [2] Antonov S, Zheng Y, Sosa JM, et al. Plasticity assisted redistribution of solutes leading to topological inversion during creep of superalloys. *Scripta Mater.* 2020;186:287–292. doi:10.1016/j.scriptamat.2020.05.004
- [3] Katnagallu S, Vernier S, Charpagne M-A, et al. Nucleation mechanism of hetero-epitaxial recrystallization in wrought nickel-based superalloys. *Scripta Mater.* 2021;191:7–11. doi:10.1016/j.scriptamat.2020.09.012
- [4] Liu P, Zhang R, Yuan Y, et al. Microstructural evolution of a Ni-Co based superalloy during hot compression at γ' sub-/super-solvus temperatures. *J Mater Sci Technol.* 2021;77:66–81. doi:10.1016/j.jmst.2020.10.042
- [5] Roy AK, Pal J, Mukhopadhyay C. Dynamic strain ageing of an austenitic superalloy-Temperature and strain rate effects. *Mater Sci Eng A.* 2008;474(1–2):363–370. doi:10.1016/j.msea.2007.05.056
- [6] Jiang L, Ye X, Cui C, et al. Intermediate temperature embrittlement of one new Ni-26W-6Cr based superalloy for molten salt reactors. *Mater Sci Eng A.* 2016;668:137–145. doi:10.1016/j.msea.2016.04.032
- [7] Wu Y, Qin X, Wang C, et al. Microstructural evolution and its influence on the impact toughness of GH984G alloy during long-term thermal exposure. *J Mater Sci Technol.* 2021;60:61–69. doi:10.1016/j.jmst.2020.06.005
- [8] Du B, Yang J, Cui C, et al. Effects of grain refinement on the microstructure and tensile behavior of K417G superalloy. *Mater Sci Eng A.* 2015;623:59–67. doi:10.1016/j.msea.2014.11.041
- [9] Bai Y, Zhang R, Cui C, et al. Thermal Stability of GH4065A superalloy based on microstructural and performance evolution following long-term aging. *Met Mater -Int.* 2024;30(2):333–347. doi:10.1007/s12540-023-01510-x
- [10] Zheng L, Chellali R, Schlesiger R, et al. Intermediate temperature embrittlement in high-purity Ni and binary Ni(Bi) alloy. *Scripta Mater.* 2011;65(5):428–431. doi:10.1016/j.scriptamat.2011.05.024
- [11] Yamaguchi M, Shiga M, Kaburaki H. Grain boundary decohesion by impurity segregation in a nickel-sulfur system. *Science.* 2005;307(5708):393–397. doi:10.1126/science.1104624
- [12] Chellali MR, Zheng L, Schlesiger R, et al. Grain boundary segregation in binary nickel–bismuth alloy. *Acta Mater.* 2016;103:754–760. doi:10.1016/j.actamat.2015.11.003
- [13] Yan Z, Srikakulapu K, Gao Y, et al. γ' -size effect on the deformation microstructure and texture in a Ni-based superalloy using in-situ neutron diffraction. *Scripta Mater.* 2023;236:115665. doi:10.1016/j.scriptamat.2023.115665
- [14] Perrin C, Berbenni S, Vehoff H, et al. Role of discrete intragranular slip on lattice rotations in polycrystalline Ni: experimental and micromechanical studies. *Acta Mater.* 2010;58(14):4639–4649. doi:10.1016/j.actamat.2010.04.033
- [15] Qu P, Yang W, Liu C, et al. Tensile deformation dominated by matrix dislocations at intermediate temperatures revealed using in-situ EBSD in superalloys. *Mater Res Lett.* 2024;12(2):116–124. doi:10.1080/21663831.2024.2304255
- [16] Ojima M, Adachi Y, Suzuki S, et al. Stress partitioning behavior in an fcc alloy evaluated by the in situ/ex situ EBSD-Wilkinson method. *Acta Mater.* 2011;59(10):4177–4185. doi:10.1016/j.actamat.2011.03.042
- [17] Alabort E, Barba D, Sulzer S, et al. Grain boundary properties of a nickel-based superalloy: Characterisation and modelling. *Acta Mater.* 2018;151:377–394. doi:10.1016/j.actamat.2018.03.059
- [18] He LZ, Zheng Q, Sun XF, et al. Low ductility at intermediate temperature of Ni-base superalloy M963. *Mater Sci Eng A.* 2004;380(1):340–348. doi:10.1016/j.msea.2004.03.057
- [19] Sharghi-Moshtaghin R, Asgari S. The influence of thermal exposure on the γ' precipitates characteristics and tensile behavior of superalloy IN-738LC. *J Mater Process Technol.* 2004;147(3):343–350. doi:10.1016/j.jmatprotec.2004.01.006
- [20] Osada T, Gu Y, Nagashima N, et al. Optimum microstructure combination for maximizing tensile strength in a polycrystalline superalloy with a two-phase structure. *Acta Mater.* 2013;61(5):1820–1829. doi:10.1016/j.actamat.2012.12.004
- [21] Dong JX, Zhang MC, Xie XS, et al. Interfacial segregation and cosegregation behaviour in a nickel-base alloy 718. *Mater Sci Eng A.* 2002;328(1–2):8–13. doi:10.1016/s0921-5093(01)01491-5
- [22] Razumovskiy VI, Lozovoi AY, Razumovskii IM. First-principles-aided design of a new Ni-base superalloy: Influence of transition metal alloying elements on grain boundary and bulk cohesion. *Acta Mater.* 2015;82:369–377. doi:10.1016/j.actamat.2014.08.047

- [23] Chen J, Lu J, Cai W, et al. In-situ study of adjacent grains slip transfer of Inconel 718 during tensile process at high temperature. *Int J Plast.* **2023**;163:103554. doi:[10.1016/j.ijplas.2023.103554](https://doi.org/10.1016/j.ijplas.2023.103554)
- [24] Kacher JP, Liu GS, Robertson IM. Visualization of grain boundary/dislocation interactions using tomographic reconstructions. *Scripta Mater.* **2011**;64(7):677–680. doi:[10.1016/j.scriptamat.2010.12.020](https://doi.org/10.1016/j.scriptamat.2010.12.020)
- [25] Ma J, Lu J, Tang L, et al. A novel instrument for investigating the dynamic microstructure evolution of high temperature service materials up to 1150 °C in scanning electron microscope. *Rev Sci Instrum.* **2020**;91(4):043704. doi:[10.1063/1.5142807](https://doi.org/10.1063/1.5142807)
- [26] Clausen B, Lorentzen T, Leffers T. Self-consistent modelling of the plastic deformation of f.c.c. polycrystals and its implications for diffraction measurements of internal stresses. *Acta Mater.* **1998**;46(9):3087–3098. doi:[10.1016/S1359-6454\(98\)00014-7](https://doi.org/10.1016/S1359-6454(98)00014-7)
- [27] Zhang W, Lu J, Wang J, et al. In-situ EBSD study of deformation behavior of Inconel 740H alloy at high-temperature tensile loading. *J Alloys Compd.* **2020**; 820:153424. doi:[10.1016/j.jallcom.2019.153424](https://doi.org/10.1016/j.jallcom.2019.153424)
- [28] Nagarajan S, Jain R, Gurao NP. Microstructural characteristics governing the lattice rotation in Al-Mg alloy using in-situ EBSD. *Mater Charact.* **2021**;180:111405. doi:[10.1016/j.matchar.2021.111405](https://doi.org/10.1016/j.matchar.2021.111405)
- [29] Polonsky AT, Lenthe WC, Echlin MP, et al. Solidification-driven orientation gradients in additively manufactured stainless steel. *Acta Mater.* **2020**;183:249–260. doi:[10.1016/j.actamat.2019.10.047](https://doi.org/10.1016/j.actamat.2019.10.047)
- [30] Guo Y, Britton TB, Wilkinson AJ. Slip band–grain boundary interactions in commercial-purity titanium. *Acta Mater.* **2014**;76:1–12. doi:[10.1016/j.actamat.2014.05.015](https://doi.org/10.1016/j.actamat.2014.05.015)
- [31] Luster J, Morris MA. Compatibility of deformation in two-phase Ti-Al alloys: dependence on microstructure and orientation relationships. *Metall Mater Trans A.* **1995**;26(7):1745–1756. doi:[10.1007/BF02670762](https://doi.org/10.1007/BF02670762)
- [32] Nieto-Valeiras E, Llorca J. Criteria for slip transfer across grain and twin boundaries in pure Ni. *Materialia.* **2022**;21:101303. doi:[10.1016/j.mtla.2021.101303](https://doi.org/10.1016/j.mtla.2021.101303)