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Correlation the $\langle 112 \rangle$ 111 slip with high-temperature tension/compression asymmetry in the single-crystal nickel-based superalloy PWA1483

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ABSTRACT

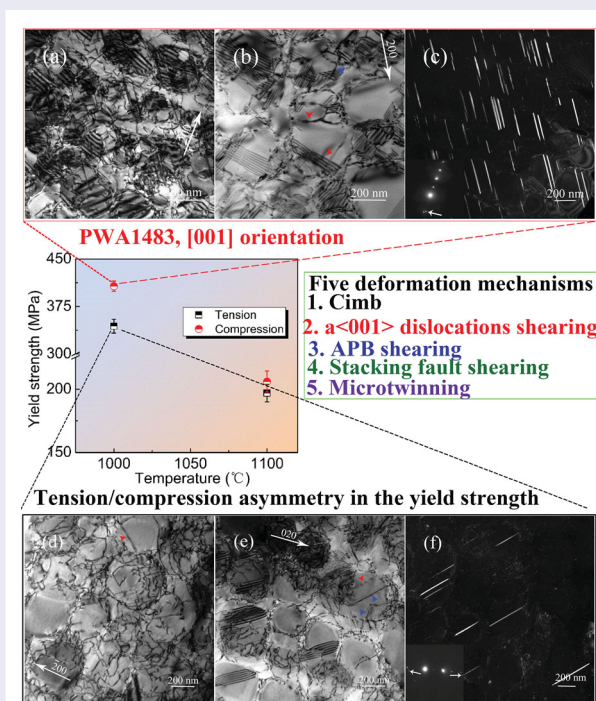
The yield strength of the [001]-orientated single-crystal nickel-based superalloy PWA1483 is investigated under tension/compression at 1000 and 1100 °C. We found that the compressive yield strength is larger than the tensile strength. Microstructural observations disclose that the dominant deformation modes are stacking fault shearing and microtwinning in compression at 1000 °C, whereas dislocation climb controls the plastic deformation under the other three deformation conditions, although the above two processes also occur. Based on these results, it is deemed that the frequent occurrence of $\langle 112 \rangle$ 111 slip accounts for the tension/compression asymmetry in the yield strength of PWA1483 at high temperatures.

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KEYWORDS

Nickel-based superalloy; high temperature; tension/compression asymmetry; stacking faults; microtwins



IMPACT STATEMENT

The tension/compression asymmetry concerning the yield strength, stacking fault shearing and microtwinning at 1000 and 1100 °C in the [001]-oriented single-crystal nickel-based superalloy is uncovered for the first time.

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1. Introduction

Single-crystal (SC) nickel-based superalloys have a wide application in the vanes/blades of advanced aircraft engines due to their excellent resistance to plastic deformation as well as oxidation and hot corrosion [1,2]. As the inlet temperature of these engines increases, hollow vanes/blades are usually fabricated to reduce the exposed temperatures during service. Generally, the complex shape of the cored vanes/blades usually induces the stress and temperature distribution, and thus it is imperative for researchers to understand the strength and deformation mechanisms under various loading conditions to exploit the maximum potential of these alloys [3–5].

The tension/compression asymmetry and corresponding mechanisms of nickel-based superalloys have been widely investigated [3,6–19]. Initially, Shah et al. [6] ascribed the tension/compression asymmetry in the yield strength of the [001]-orientated SC nickel-based superalloy Mar-M200 below the peak-stress temperature (around 760 °C) to the ‘cross-slip’ and ‘core width’ effects, which were firstly proposed by Lall et al. [20] for the $\text{Ni}_3(\text{Al}, \text{Nb})$ single crystals and also responsible for the non-Schmid effects in superalloys [7–10,20]. For the [001] direction, the tensile stress tended to constrict the two Shockley partials which constituted the $a/2 < 110 >$ dislocation, thus facilitating cross-slip of the latter dislocation from 111 planes onto 100 planes [20]. As the dislocation segments on octahedral planes were locally pinned by the segments on cube planes, the tensile yield strength was larger as compared to the compressive yield strength [7,8,10]. Subsequently, Jiao et al. [11] pointed out that the tension/compression asymmetry of the [001]-orientated SC nickel-based superalloy SC16 during fatigue testing at 850 °C was due to stacking fault shearing, and the easier occurrence of the process in tension accounted for the lower tensile stress at fatigue saturation. In contrast to these results, Yamashita and Kakehi [15] as well as Tsuno et al. [16] proposed that the tension/compression asymmetry concerning the yield strength of the SC nickel-based superalloy PWA1480 oriented near the [001] direction was attributed to the directional-formation feature of microtwins (MTs) at intermediate temperatures. As the operation of microtwinning, which was prone to occur in compression, induced weakness, the compressive yield strength was inferior to the tensile strength [15,16].

Although many experimental and theoretical investigations are available, there is still a lack of detailed information about the influence of the sense of the applied stress on the mechanical properties and elementary mechanisms of SC nickel-based superalloys at high temperatures. This study has been focused on the mechanical response and deformation mechanisms of

the [001]-oriented SC nickel-based superalloy PWA1483 under quasi-static compressive and tensile loading conditions at 1000 and 1100 °C. Emphasis is made on clarifying whether the tension/compression asymmetry appears in the alloy at high temperatures.

2. Materials and methods

The nominal composition of the SC nickel-based superalloy PWA1483 is 9.0Co, 12.2Cr, 3.8W, 5.0Ta, 3.6Al, 4.1Ti, 1.9Mo and 0.07 C with the balance Ni (all in the weight percent). Bars of SCs were grown along the $< 001 >$ direction at a constant withdrawal rate of 3 mm min^{-1} using an ALD Vacuum Technologies furnace. The diameter and length of the bars are 14 and 210 mm, respectively. Using the method developed by Guo et al. [21], the misorientations of all the bars were determined by a Rigaku D/MAX-2400 diffractometer with a Cu K α source. The orientation of one of these bars is about 5.56° deviating from the [001] direction (see Electronic Supplementary Figure S-1), and all the specimens investigated in this study stem from this bar. The specimens were given to a standard heat treatment (SHT), including a two-step solution treatment, 1204 °C/1 h + 1265 °C/1 h/air cooling (AC), and a stabilizing heat treatment, 1080 °C/4 h/AC.

Threaded tensile specimens and cylindrical samples were machined from the heat-treated bar. The gauge length and diameter of the tensile specimens are 15.0 and 3.0 mm, respectively. Whereas, the gauge height and diameter of compressive specimens are 13.5 and 5.0 mm, respectively. Both tension and compression tests were conducted on an MTS 793 testing machine using a strain rate of $2.5 \times 10^{-4} \text{ s}^{-1}$ at 1000 and 1100 °C. A three-region clamshell type resistance furnace was used for the high-temperature tests, and variations in the temperature were maintained within ± 2 °C. Besides the highly strained and rupture tests, some interrupted tests were also performed to study the deformation mechanisms using transmission electron microscope (TEM). Once the tests were finished, the resistance furnace was moved away immediately, and then specimens were cooled to room temperature (RT) through water quenching/forced-air cooling within 60 s.

Microstructures were investigated by a JEM-F200 TEM equipped with a high-angle annular dark-field (HAADF) detector and a double-tilt goniometer at 200 kV. Foils were sectioned from the heated samples perpendicular to the growth direction by a wire-cut electric discharge machine. Meanwhile, discs were also cut from the gauge sections of the deformed specimens at around 25° with respect to the loading axis. After mechanically thinning to approximate 50 μm , thin lamellae were

electropolished in a solution of ethanol with 9.0 vol.% perchloric acid and 0.5 vol.% ethylene glycol at around $-30\text{ }^{\circ}\text{C}$ and 30–50 V.

3. Results and discussion

The initial γ/γ' microstructure in PWA1483 after SHT is shown in Figure 1(a)–(b). Few dislocations are visible in the alloy, and the γ' precipitates exhibit a bimodal size distribution, as demonstrated in Figure 1(b). The coarse γ' precipitates have a cubic morphology. Whereas, the fine γ' precipitates display a spherical morphology. Quantitative measurements on the precipitate sizes were made on the HAADF images using a line intercept method, and at least 650 particles were inspected to obtain an adequate distribution. It is found that the average edge length of coarse γ' precipitates is $292 \pm 70\text{ nm}$, and the mean diameter of fine γ' precipitates is $29 \pm 12\text{ nm}$. Meanwhile, using the methods reported in Ref. [22], the γ' volume fraction is measured to be $54.2 \pm 0.8\%$, which is in accordance with the value reported for PWA1483 in Ref. [23].

Figure 1(c) shows the typical tensile and compressive engineering stress–strain curves of the experimental alloy at 1000 and 1100 $^{\circ}\text{C}$. Clearly, variations in the flow stress with strain show the similar changing trends at the beginning of plastic deformation. The yield strength was determined on the stress–strain curve by the offset method reported in Refs. [24,25], and plotted in Figure 1(d). Evidently, the compressive yield strength is larger, especially at 1000 $^{\circ}\text{C}$. That is, the alloy exhibits the tension/compression asymmetry concerning the yield strength at the two temperatures. The phenomenon is in stark contrast to the results found in other [001]-orientated SC nickel-base superalloys [8,10–13,16,26], where showed that the compressive and tensile strength are comparable at high temperatures, and the yield strength is larger in tension at low and intermediate temperatures. Obviously, the theories stated above, especially the ‘cross-slip’ and ‘core width’ effects, cannot explain the present results, and therefore the deformation mechanisms in PWA1483 at the two temperatures were investigated in the following section.

Figure 2 shows the typical microstructures in PWA1483 after approximate 2.0% compressive plastic strain at 1000 and 1100 $^{\circ}\text{C}$. Similar to previous results found in other SC nickel-based superalloys, some zig-zag and curve dislocations, which are almost oriented parallel to the cube edges of the γ' precipitates, are visible within the matrix channels when the beam direction (BD) is close to the [011] direction (see Figure 2(a)), demonstrating that dislocation climb operates during plastic deformation [27–30]. Whereas, within the γ' precipitates, a few pairs of $a/2 < 110 >$ dislocations

and $a < 001 >$ superdislocations, which are separately indicated by the blue triangle and red arrowheads in Figure 2(b), are visible. It indicates that anti-phase boundary (APB) shearing and two dissimilar $a/2 < 110 >$ dislocations collectively penetrating γ' precipitates occur occasionally during plastic deformation [12,28,31–33]. Aside from these deformation patterns, the most notable features are numerous isolated superlattice stacking faults (SSFs) and MTs within γ' precipitates, suggesting that stacking fault shearing and microtwinning dominate the plastic deformation at 1000 $^{\circ}\text{C}$ [3,13,16,28,34–37]. As the temperature increases, the five deformation patterns mentioned above are still visible in the deformed specimens, as illustrated in Figure 2(d)–(f). Whereas, the densities of SSFs and MTs decrease dramatically as compared with those at 1000 $^{\circ}\text{C}$ (see Figure 2(d) and (f)). Generally, the prime deformation mode in SC nickel-based superalloys is the viscous slip of $a/2 < 011 >$ dislocations within matrix channels and then the climb process along the γ/γ' interfaces during tensile deformation at high temperatures above 950 $^{\circ}\text{C}$ [31,38–43]. So, someone might argue that the compressive stress contributes to the creation of these planar faults during plastic deformation, since stacking fault shearing and microtwinning are inclined to occur during compressive deformation [4,5,13,16,19].

Then, after around 2.0% tensile plastic strain at 1000 and 1100 $^{\circ}\text{C}$, the microstructures in PWA1483 were also investigated. Indeed, as reported in Refs. [13,15,16], SSFs and MTs are more frequently visible in compression. However, as shown in Figure 3, the five deformation modes can still be visible in PWA1483, although the densities of SSFs and MTs are much lower as compared to that of zig-zag and curve dislocations which are almost oriented parallel to the cube edges of the γ' precipitates. It suggests that dislocation climb controls the plastic deformation in tension at the two temperatures, rather than stacking fault shearing and microtwinning. Meanwhile, in order to further demonstrate that MTs are created in PWA 1483, a high-resolution TEM (HRTEM) image of one MT is obtained and manifested in Figure 4(a). Generally, for nickel-based superalloys, stacking fault shearing and/or microtwinning play a significant role in the quasi-static compressive/tensile plastic deformation only at temperatures between RT and 950 $^{\circ}\text{C}$ [3,31,34,37,44]. However, our study discloses that the two processes occur frequently in PWA1483 at 1000 and 1100 $^{\circ}\text{C}$. To our best knowledge, these findings have never been reported so far, hence expanding the formation temperature range of SSFs and MTs and adding new datapoints to the deformation-mechanism maps of nickel-based superalloys.

Following the procures reported in Refs. [46,47], the formation process of isolated SSFs is analyzed in

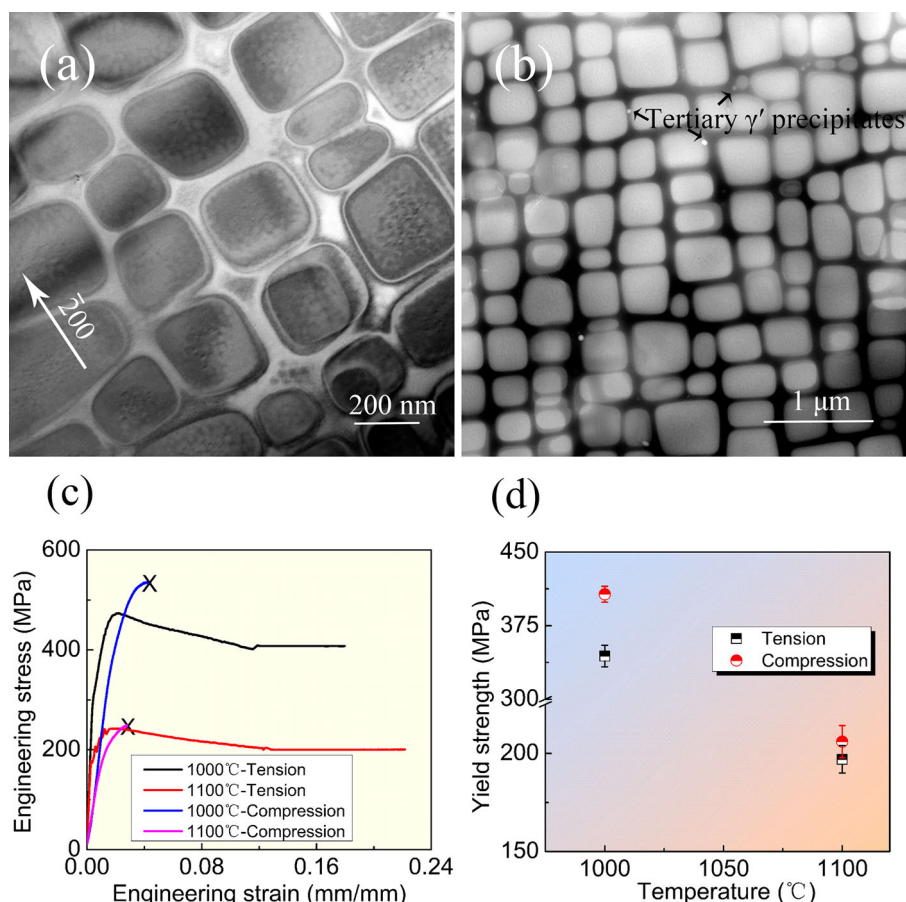


Figure 1. (a) Bright-field (BF) TEM image illustrating the γ/γ' microstructure in PWA1483 after SHT; (b) HAADF image displaying the γ' precipitates in PWA1483 after SHT; (c) typical engineering stress-strain curves of the [001]-orientated SC nickel-based superalloy PWA1483 in compression and tension at high temperatures; (d) tensile and compressive yield strength at 1000 and 1100 °C. With respect to the compressive stress-strain curves, the extensometer was not used during the tests. The crosses manifest that compression tests are broken off.

detail. As illustrated in Figure 4(b) (a centered dark-field (CDF) image), the g (020) points towards the light outer fringe. Thus, the nature of the SSF is intrinsic [46]. Meanwhile, it is found that the Burgers vector of the partial dislocation which results in formation of the SSF is the $a/3[11\bar{2}]$, and the fault is connected to the partial dislocation $a/6[211]$ (see Electronic Supplementary Figure S-2 and Table S-1). Thus, it is reasonable to conclude that the formation of SSFs in PWA1483 follows the procedures reported in other nickel-based superalloys [28,31,48–50]. As with the formation mechanism of MTs in PWA1483, it is unclear at present. Whereas, Guimier and Strudel [51] pointed out that the successive glide of the $a/3 < 121 >$ partial dislocations on neighboring 111 planes led to the creation of MTs in the nickel-based superalloy Waspaloy during tensile deformation at intermediate temperatures. Subsequently, the process was also found in the nickel-based superalloy CMSX-4 during compressive deformation at high temperatures [28]. Moreover, given the fact that the partial dislocations which penetrate γ' precipitates are also the

$a/3 < 121 >$ type, it is believed that the formation of MTs in PWA1483 follows the scenario as well. On the other hand, as the dislocation $a/3[11\bar{2}]$ acts as the leading partial in compression, compressive stress tends to extend the two partials, but the reverse is true in tension for the [001] direction [19,20]. This might explain why microtwinning is prone to occur in compression. Then, a question has arisen as to why stacking fault shearing and microtwinning are easier to occur in PWA1483, rather than in other [001]-oriented SC nickel-based superalloys. Generally, the operative deformation mechanisms in nickel-based superalloy are intimately related to the compositions of the γ and γ' phases, which are in turn dependent on the alloy composition [16,36,52]. Then, using the software Thermo-Calc together with TCNi11 database, the contents of various alloying elements in the two phases of PWA1483 at 1000 and 1100 °C were calculated due to the experimental difficulties in acquiring these values at such high temperatures (see Electronic Supplementary Figure S-3). It is found that a high content of metallic elements Cr and Co has dissolved in the

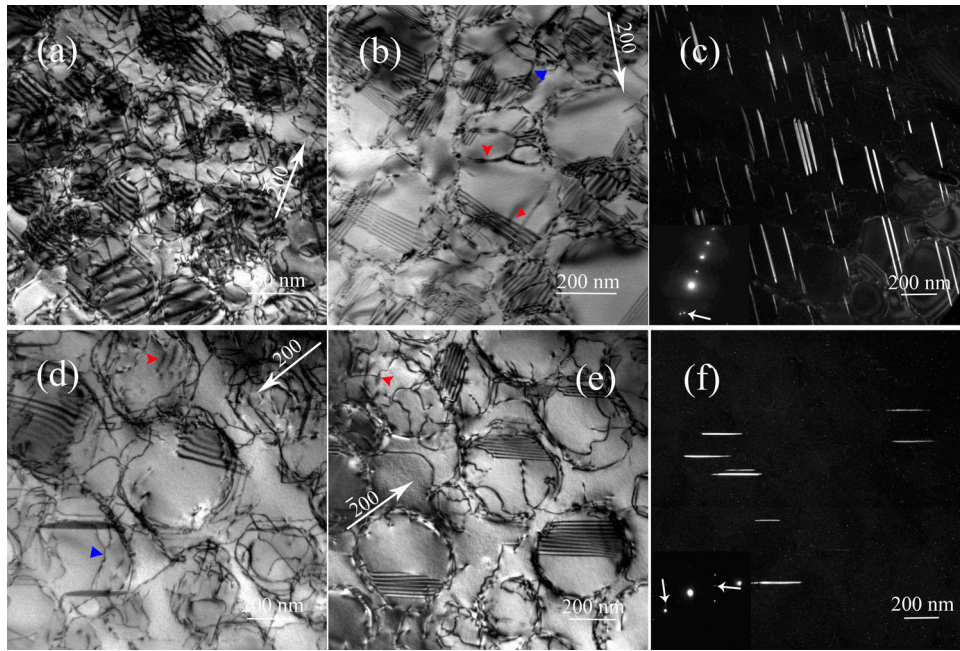


Figure 2. TEM images showing the typical microstructures in PWA1483 after around 2.0% compressive plastic strain at 1000 °C ((a)-(c)) and 1100 °C ((d)-(f)): (a) zig-zag and curve dislocations as well as isolated SSFs, $\mathbf{g} = (\bar{2}00)_{\text{BF}}$; (b) isolated SSFs, pairs of $a/2 < 110 >$ dislocations and $a < 001 >$ superdislocations, $\mathbf{g} = (200)_{\text{BF}}$; (c) MTs, DF image; (d) zig-zag and curve dislocations as well as pairs of $a/2 < 110 >$ dislocations, $\mathbf{g} = (200)_{\text{BF}}$; (e) isolated SSFs and $a < 001 >$ superdislocations, $\mathbf{g} = (\bar{2}00)_{\text{BF}}$; (f) MTs, DF image. The red arrowheads and blue triangles indicate the $a < 001 >$ superdislocations and pairs of $a/2 < 110 >$ dislocations, respectively. The white arrow indicates the spots of MTs within the diffraction pattern of the [011] zone axis.

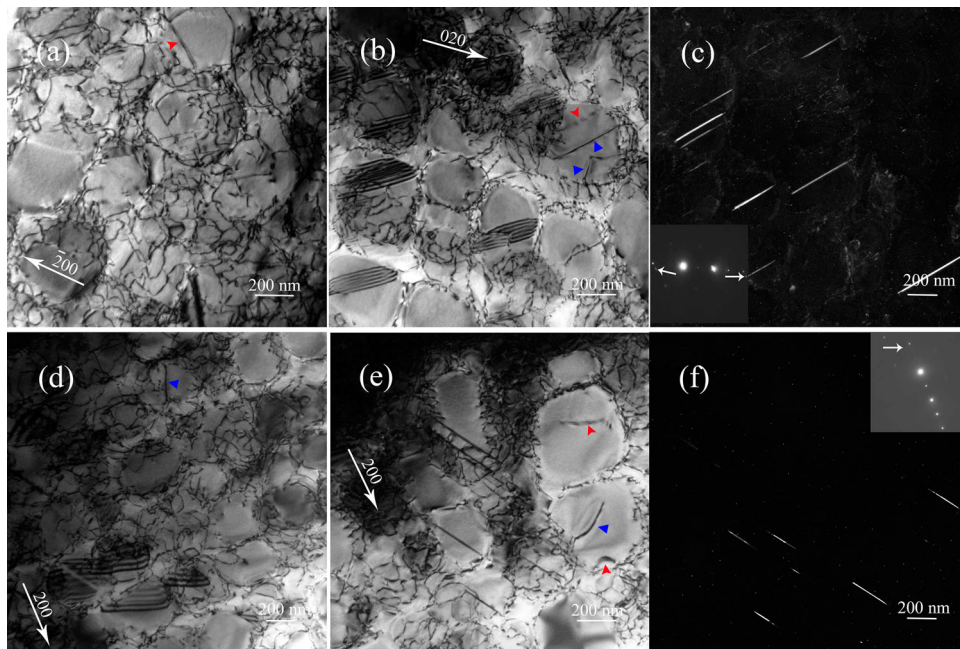


Figure 3. TEM images showing the typical microstructures in PWA1483 after around 2.0% tensile plastic strain at 1000 °C ((a)-(c)) and 1100 °C ((d)-(f)): (a) zig-zag and curve dislocations as well as $a < 001 >$ superdislocations, $\mathbf{g} = (\bar{2}00)_{\text{BF}}$; (b) isolated SSFs, pairs of $a/2 < 110 >$ dislocations and $a < 001 >$ superdislocations, $\mathbf{g} = (020)_{\text{BF}}$; (c) MTs, DF image; (d) zig-zag and curve dislocations, isolated SSFs and pairs of $a/2 < 110 >$ dislocations, $\mathbf{g} = (200)_{\text{BF}}$; (e) pairs of $a/2 < 110 >$ dislocations and $a < 001 >$ superdislocations, $\mathbf{g} = (200)_{\text{BF}}$; (f) MTs, DF image. The red arrowheads and blue triangles indicate the $a < 001 >$ superdislocations and pairs of $a/2 < 110 >$ dislocations, respectively. The white arrow indicates the spots of MTs within the diffraction pattern of the [011] zone axis.

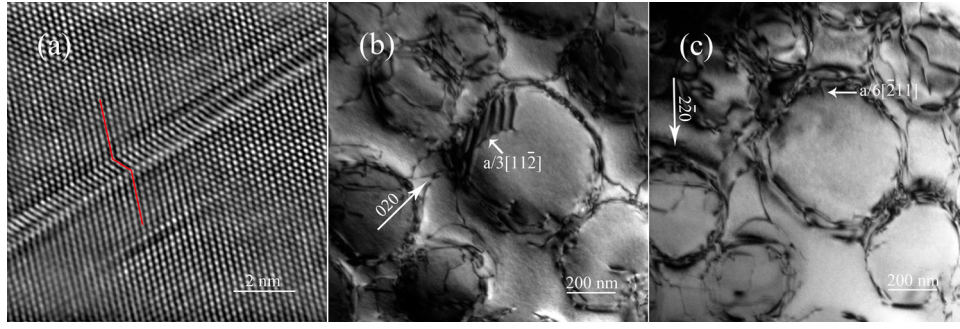


Figure 4. (a) HRTEM micrograph showing the formation of MTs in PWA1483 after around 2.0% compressive plastic strain at 1000 °C (Using the HRTEM filter script developed by Mitchell to filter the HRTEM image based on the work of Kilaas [45]); (b) CDF image showing one SSF, $g = (020)$; (c) BF image illustrating the invisibility of the SSF with the reflection ($2\bar{2}0$).

matrix, which can decrease the stacking fault energy of the γ phase dramatically and then induce the decomposition of matrix dislocations [13,34,36,44,53]. With respect to the γ' precipitate, there are large amounts of metallic elements Co, Ti, and Ta. It is reported that the former element can decrease the stacking fault energy of the γ' precipitate, whereas the latter two elements would give rise to the increased APB energy [34,53–56]. Thus, it is believed that all these factors facilitate the formation of these planar faults [57,58].

Another question needed to be answered is why the compressive yield strength is larger than the tensile yield strength, especially at 1000 °C, for the experimental alloy. In fact, the phenomenon might be understood in term of the contribution to the yield strength due to the operation of various deformation modes. As stated above, the compressive plastic deformation of PWA1483 is accomplished mainly by stacking fault shearing and microtwinning at 1000 °C, and dislocation climb dominates the plastic deformation under the other three deformation conditions, which can also be drawn based on the microstructures in the slightly deformed specimens (see Electronic Supplementary Figure S-4 and 5). Then, following the procedure reported in Refs. [59–62], increments in the critical resolved shear stresses (CRSSs) due to local climb, stacking fault shearing and microtwinning were calculated using the simply modified models:

$$\Delta\tau_{Lc} = \{[\sqrt{2}Ga(6f/\pi)^{1/2}]/2d\}[\lambda^{1.5}/(2\sqrt{2} + \lambda^{1.5})], \quad (1)$$

$$\Delta\tau_{SF}/\Delta\tau_{Twin} = \frac{\sqrt{6}}{2} \left(\frac{\pi(2-\nu)}{4(1-\nu)} \frac{nGa}{12d} + \frac{\gamma_{SF}}{na} \right), \quad (2)$$

where $\Delta\tau_{Lc}$ is the CRSS increment due to single matrix dislocations climbing over γ' precipitates, a , the lattice parameter of the matrix, G , the shear modulus, f , the volume fraction of γ' precipitates, d , the diameter of γ' precipitates, λ , a parameter that describe

the climb resistance, $\Delta\tau_{SF}$ and $\Delta\tau_{Twin}$, the CRSS increments due to stacking fault shearing and microtwinning, ν , the Poisson's ratio, γ_{SF} , the stacking fault energy of the γ' phase, n , the layer numbers of the planar fault. When $n = 1$, $\Delta\tau = \Delta\tau_{SF}$. Whereas, if $n > 2$, $\Delta\tau = \Delta\tau_{Twin}$ [59,61]. Using the following material parameters: $a = 0.365$ nm, $\nu = 0.342$, $G = 55.4$ GPa at 1000 °C (All the parameters are acquired using the software JMatPro Version 11.1 due to lack of reliable experimental data. See the Electronic Supplemental Material.), $\lambda = \sqrt{2}$ [62], $\gamma_{SF} = 0.010$ – 0.035 J m $^{-2}$ [49,61] and $n = 1$ or 3, together with Eqs. (1) and (2), $\Delta\tau_{Lc}$, $\Delta\tau_{SF}$ and $\Delta\tau_{Twin}$ are calculated to 18.6 MPa, 47.5–131.4 MPa, and 53.1–81.1 MPa, respectively. In the view of these estimates, one can get the minimum contribution of stacking fault shearing and microtwinning to the CRSS are two times larger than the CRSS increment originating from dislocation climb, which explains why the compressive yield strength is larger at 1000 °C. As the deformation temperature increases, although the content of SSFs and MTs is higher in PWA1483 after compressive deformation, plastic deformation is accomplished mainly by dislocation climb, rather than stacking fault shearing and microtwinning (see Figures 2 and 3). Thus, the yield strength decreases dramatically with temperature, and the tension/compression asymmetry is reduced.

4. Conclusion

In summary, we found that: (i) there exists the tension/compression asymmetry concerning the yield strength of the [001]-orientated SC nickel-based superalloy PWA1483 at 1000 and 1100 °C; (ii) dislocation climb, APB shearing, two different matrix dislocations jointly penetrating γ' precipitates and stacking fault shearing as well as microtwinning occur in PWA1483 simultaneously during compressive and tensile deformation at the two temperatures, and the latter two deformation modes are prone to occur in compression; (iii) the frequent

occurrence of stacking fault shearing and microtwinning accounts for the higher compressive yield strength for the experimental alloy.

Disclosure statement

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

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