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A Yield Stress Model for a Solution-Treated Ni-Based Superalloy during Plastic Deformation

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Abstract: Up to now, there are few reports on the yield behavior of Ni-based superalloy during plastic deformation. However, an accurate yield stress model is significant for simulating the plastic forming process by cellular automaton or finite element methods. Therefore, the yield behavior of a solution-treated Ni-based superalloy is studied by hot compression tests. In order to evaluate yield stresses from the measured flow stress curves, the yield process is analyzed in terms of dislocation theory. Then, yield stresses at different deformation temperatures and strain rates are clearly determined. The experimental results show that the yield stresses are highly sensitive to deformation temperature and strain rate. The determined yield stress almost linearly increases with the increase of the logarithm of strain rate or the reciprocal of deformation temperature. A yield stress model is developed to correlate the yield behavior of the studied solution-treated Ni-based superalloy with deformation temperature, strain rate, and strengthening effect of alloying elements. The developed model can well describe the yield behavior of the studied solution-treated Ni-based superalloy.

Keywords: alloy, yield behavior, dislocation, strengthening effect, alloying elements

Introduction

Generally, metals and alloys undergo a yield process before plastic deformation [1, 2]. In the yield process,

the movement and multiplication of dislocations greatly affect the plastic deformation of alloys [3–5]. Thus, a comprehensive understanding of yield behavior is useful for developing accurate constitutive models to describe the plastic flow behavior [6–10]. Meanwhile, the physical mechanisms controlling the subsequent plastic deformation can be well interpreted [11–15].

Due to their high strength, excellent corrosion resistance, thermal fatigue properties, and thermal stability [16, 17], Ni-based superalloys are widely used in different industrial equipment, especially in the critical parts of nuclear fission reactors, fusion reactors, and aircraft engines [18–22]. In recent years, some efforts have been made to study the yield behavior of some typical Ni-based superalloys. Zhao et al. [23] investigated the discontinuous yield behavior in GH4049 superalloy during high-speed compression. Roth et al. [24] proposed a yield stress model for Ni-based alloys, considering dependences of yield stress on the chemical composition and deformation temperature. Momeni et al. [25] carried out the comparative study on the discontinuous yield behaviors in IN718 and ALLVAC 718 Plus superalloys. Weiss and Grushko [26] developed a yield stress model for IN718 superalloy at low-subroom temperatures. Galindo-Nava et al. [27] proposed a yield stress model for Ni-based superalloys, considering the unimodal and multimodal precipitation distribution. Crudden et al. [28] developed a yield stress model for Ni-based superalloys, accounting for the effects of chemical composition.

Although, some investigations have been carried out on the yield behavior of Ni-based superalloys at in-service temperatures or below, few reports the yield behavior during plastic deformation. However, an accurate yield stress model is vital for simulating the plastic forming process by cellular automaton or finite element methods. In this study, the yield behavior of a solution-treated Ni-based superalloy is investigated by isothermal hot compression tests. Yield stresses are directly determined from the measured true stress–true strain curves. Based on the determined yield stress, a yield stress model is developed to describe the effects of deformation temperature, strain rate, and critical alloying elements on yield behavior.

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Material and experiments

The material used in this study is a commercial GH4169 Ni-based superalloy. Its chemical composition (wt.%) is 52.82Ni-18.96Cr-5.23Nb-3.01Mo-1.00Ti-0.59Al-0.01Co-0.03C-(bal.)Fe. Cylindrical specimens of 8 mm diameter and 12 mm height were cut from the billet. The specimens were solution treated at 1313 K for 45 min to dissolve all possible precipitates [29, 30]. Then, the solution-treated specimens were water-quenched to reserve the precipitates-free microstructure. Hot compression tests were conducted on a Gleeble 3500 thermo-mechanical simulator. The experimental conditions lie in the deformation temperature range of 1223–1313 K and strain rate range of 0.001 – 0.1 s^{-1} . Figure 1 shows the detailed procedure for the hot compression tests.

Yield process of the studied Ni-based superalloy

In this section, the yield behavior of metals and alloys will be interpreted, and an unambiguous way to determine the yield stress is introduced.

Figure 1: Procedure for the hot compression tests.

Figure 2: True stress–true strain curves: (a) typical experiment results; (b) detailed yield behavior.

Figure 2 shows the representative stress–strain curve obtained from the hot compression tests. The measured stress–strain curve can be divided into three stages: stage I (elastic stage), stage II (anelastic stage), and stage III (plastic stage) [18, 31]. In stage I, the elastic deformation is time-independent. σ_e is the stress limit of this stage. In stage II, dislocation starts to move. The movement of dislocation leads to the time-dependent deformation. However, the deformation in this stage is still reversible when the external applied stress is removed. σ_y is the yield stress, which is the stress limit of stage II. In stage III, the stress is larger than σ_y . Plastic deformation starts to occur in this stage. The turning point between the reversible and plastic deformation is called the yield point. Generally, 0.2% plastic-strain off-set yield stress is used to determine the yield point. However, the 0.2% plastic-strain off-set yield stress is a rather arbitrary notion.

In stage II, the total strain ε can be written as,

$$\varepsilon = \varepsilon_e + \varepsilon_a \quad (1)$$

where ε_e is the elastic strain, $d\sigma/d\varepsilon_e = E$, σ is the external applied stress, E is the elastic modulus. ε_e is reversible and time-independent. ε_a is the stored anelastic strain resulting from dislocation movement. ε_a is reversible and time-dependent.

The external applied stress σ in this stage can be written as,

$$\sigma = \sigma_a + \sigma_f \quad (2)$$

where σ_a is the back stress due to the anelastic deformation of dislocation, $\sigma_a = \theta_a \varepsilon_a$, θ_a is the anelastic modulus. σ_f is the friction stress against the anelastic deformation of dislocation. σ_f is induced by surrounding atoms, and is only a function of strain rate [32].

In stage II, the first derivative (θ_{II}) of σ with respect to ε can be expressed as,

$$\begin{aligned}\theta_{II} &= \frac{d\sigma}{d\varepsilon} = \frac{d\sigma}{d\varepsilon_e + d\varepsilon_a} = \frac{1}{\frac{d\varepsilon_e}{d\sigma} + \frac{d\varepsilon_a}{d\sigma_a + d\sigma_f}} \\ &= \frac{1}{\frac{d\varepsilon_e}{d\sigma} + \frac{1}{\frac{d\sigma_a}{d\varepsilon_a} + \frac{d\sigma_f}{d\varepsilon_a}}} = \frac{1}{\frac{1}{E} + \frac{1}{\theta_a}} = \frac{E\theta_a}{E + \theta_a}\end{aligned}\quad (3)$$

Plastic deformation of alloys derives from the movement and multiplication of linear defects known as dislocations. Figure 3(a) illustrates a dislocation with segment length L . Both ends of the illustrated dislocation (A and B) are pinned and cannot move. It is generally called the double-pinned Frank-Read dislocation source, and is the main source for dislocation multiplication. Figure 3(b) shows the process of dislocation multiplication. The dislocation segment bows out under shear stress τ . With the increase of τ , the radius of the bowed segment R decreases. When R falls to $L/2$, the dislocation segment reaches a critical state. If τ is removed before this state, the dislocation segment will go back to its initial position, showing an anelastic property. If the dislocation segment continues to bow out, a large dislocation outer loop and a new dislocation segment AB will be generated. This generation cannot be reversed and is a complete plastic behavior. The outer loop will spread outward, while the new dislocation segment will repeat the multiplication process.

τ can be written as,

$$\tau = \tau_{FR} + \tau_f \quad (4)$$

where τ_{FR} is the athermal part, purely geometrical. τ_f is the frictional part, contributed by the surrounding atoms. τ_f is only dependent on strain rate and deformation temperature.

The relationship between τ_{FR} and R can be written as,

$$\tau_{FR} = \frac{0.5\mu b}{R} \quad (5)$$

where μ is shear modulus, $\mu = E/(2(1+\nu))$, ν is Poisson ratio, b is Burger's vector. Thus, the critical value for τ_{FR} is,

Figure 3: Schematic of dislocation movement and multiplication: (a) Frank-Read dislocation source; (b) dislocation multiplication.

$$\tau_{FR}^c = \frac{\mu b}{L} \quad (6)$$

The shear strain induced by the curving deformation is,

$$\gamma = NbS \quad (7)$$

Where N is the number of dislocation segment per unit volume. S is the area swept by the curving deformation of dislocation segment AB.

$$S = \frac{1}{2}R^2(\beta - \sin\beta) \quad (8)$$

where β is the subtended angle of bowed segment. Based on the geometric relationship between R , β and L , the first derivative (θ_s) of shear stress with respect to shear strain can be written as,

$$\theta_s = \frac{d\tau}{d\gamma} = \frac{d\tau_{FR}}{d\gamma} = \frac{2\tau_{FR}^3 \sqrt{1 - \left(\frac{\tau_{FR}L}{\mu b}\right)^2}}{\mu b^2 \frac{\rho}{L} \left(\tau_{FR}L - \mu b \arcsin\left(\frac{\tau_{FR}L}{\mu b}\right) \sqrt{1 - \left(\frac{\tau_{FR}L}{\mu b}\right)^2} \right)} \quad (9)$$

where ρ is dislocation density. It is defined as the total length of dislocation per unit volume, $\rho = NL$. According to eq. (6), eq. (9) can be rewritten as,

$$\theta_s = \frac{2\mu p^3 \sqrt{1-p^2}}{\rho L^2 (p - \arcsin(p) \sqrt{1-p^2})} \quad (10)$$

where $p = \tau_{FR}/\tau_{FR}^c$.

According to the Taylor model [33], the relationship between the applied stress σ and shear stress τ can be written as,

$$\sigma = M\tau \quad (11)$$

where M is the Taylor factor, and is generally taken as 3.06 for face center cubic materials. Thus,

$$p = \frac{\tau_{FR}}{\tau_{FR}^c} = \frac{\sigma_a}{\sigma_c} = \frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \quad (12)$$

where σ_c is the back stress at which the dislocation segment AB yields and starts generating dislocation outer loops.

Based on the principle of virtual work ($\tau d\tau = \sigma d\varepsilon$) [34],

$$\theta_a = M^2 \theta_s \quad (13)$$

By combining eqs. (10), (12), and (13), θ_a can be written as,

$$\theta_a = \frac{2M^2 \mu \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right)^3 \sqrt{1 - \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right)^2}}{\rho L^2 \left(\left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right) - \arcsin \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right) \sqrt{1 - \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right)^2} \right)} \quad (14)$$

Thus, θ_{II} can be written as,

$$\theta_{II} = \frac{EM^2 \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right)^3 \sqrt{1 - \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right)^2}}{M^2 \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right)^3 \sqrt{1 - \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right)^2} + \rho L^2 (1 + \nu)} \quad (15)$$

$$\left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} - \arcsin \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right) \sqrt{1 - \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right)^2} \right)$$

In stage III, the external applied stress σ can be written as,

$$\sigma = \sigma_y + \sigma_w \quad (16)$$

where σ_w is the component of flow stress contributed by dislocation/dislocation interaction, and can be expressed as [35, 36],

$$\begin{cases} \sigma_w = \alpha \mu b \sqrt{\rho} \\ \frac{d\rho}{d\varepsilon} = k_1 \sqrt{\rho} - k_2 \rho \end{cases} \quad (17)$$

where α is a constant depending on the interactional strengthen of dislocations, b is Burger's vector, k_1 and k_2

are parameters of work-hardening and dynamic recovery, respectively. Thus, the first derivative (θ_{III}) of σ with respect to ε in stage III can be expressed as,

$$\theta_{III} = \frac{d\sigma}{d\varepsilon} = \frac{d\sigma_w}{d\varepsilon} = \frac{1}{2} (\alpha \mu b k_1 - k_2 \sigma_w) = \frac{1}{2} (\alpha \mu b k_1 + k_2 \sigma_y - k_2 \sigma) \quad (18)$$

Therefore, the first derivative of σ with respect to ε in stages II and III are determined. Equation (19) shows that θ_{II} and θ_{III} are obviously different. This indicates that the $\theta - \sigma$ plot has a turning point at σ_y .

$$\begin{cases} \theta_{II} = \frac{EM^2 \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right)^3 \sqrt{1 - \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right)^2}}{M^2 \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right)^3 \sqrt{1 - \left(\frac{\sigma - \sigma_f}{\sigma_y - \sigma_f} \right)^2} + \rho L^2 (1 + \nu)} \\ \theta_{III} = \frac{1}{2} (\alpha \mu b k_1 + k_2 \sigma_y - k_2 \sigma) \end{cases} \quad (19)$$

Results and discussion

Yield behavior and the determination of yield stresses

Based on the analysis of yield process in section "Yield process of the studied Ni-based superalloy", it is well known that a turning point should be found on the $\theta - \sigma$ plot at which the yield occurs. Figure 4 shows the representative $\theta - \sigma$ plots of the studied Ni-based superalloy under the tested conditions. Turning points can be clearly found on the plots. Thus, the yield stresses under the tested conditions are clearly determined.

Figure 4: $\theta - \sigma$ plots at different: (a) deformation temperatures; (b) strain rates.

Figure 5: Relationship between the yield stress and deformation temperature.

Figure 4 shows that the yield stresses are very sensitive to the deformation temperature and strain rate. Thus, the yield stresses determined from $\theta - \sigma$ plots can be plotted in terms of deformation temperature and strain rate, respectively. Figure 5 shows the relationship between yield stress and deformation temperature. The decrease of yield stress with increasing deformation temperature mainly results from the assistance of thermal fluctuations. High thermal fluctuation renders small external force to tear a dislocation from its core atmosphere.

Figure 6 shows the relationship between the yield stress and strain rate. An approximately linear relationship can be found between the logarithm of strain rate and yield stress. This linear relationship is related to the

Figure 6: Relationship between the yield stress and strain rate.

barrier at pinning points of dislocations, which were also observed by some other researchers [20, 37].

Yield stress model

The studied Ni-based superalloy is a kind of materials enriched with alloying elements. So, the strengthening effect of alloying elements should be carefully considered when a yield stress model is developed.

The contribution of alloying elements to σ_y can be superimposed as [38],

$$\sigma_y = \left(\sum_i (k_{s,i} C_i^q)^{1/q} \right)^q \quad (20)$$

where C_i and $k_{s,i}$ are the concentration and the strengthening constant of alloying element i , respectively. q is the exponential constant of alloying element concentration, and can be taken as $1/2$.

Table 1 shows the concentration and strengthening constants of different alloying elements of the studied Ni-based superalloy [39]. Here, Niobium (Nb), molybdenum (Mo), and titanium (Ti) are the main strengthening elements, while chromium (Cr) and aluminum (Al) primarily provide the resistance to oxidation and corrosion. The strengthening constants shown in Table 1 were measured at the deformation temperature of 77 K. Thus, the superimposed strength based on the data shown in Table 1 corresponds to the yield stress ($\sigma_{y,77}$) of the studied superalloy at 77 K.

The strengthening effect of alloying elements can be interpreted in view of the short range interaction between dislocations and alloying atoms. Based on this interpretation, Feltham [40] established the following relationship between yield stress, deformation temperature, and strengthening effects of alloying elements.

$$T \left(\frac{\sigma_y}{M\mu} \right)^{0.5} = K_1 - K_2 \frac{\sigma_y}{M} \quad (21)$$

where K_1 depends on strain rate, and the concentration/strengthening constant of alloying elements. K_2 is a material constant, which is only a function of strain rate.

Based on eq. (21), K_1 can be evaluated by appropriate values for yield stress and K_2 . Since the strengthening constants shown in Table 1 were measured at 77 K, a new parameter (K_2^m) is proposed,

$$K_1 = 77 \left(\frac{\sigma_{y,77}}{M\mu_{77}} \right)^{0.5} + K_2^m \frac{\sigma_{y,77}}{M} \quad (22)$$

Table 1: Concentration and strengthening constants of alloying elements in the studied Ni-based superalloy.

Alloying element	Cr	Fe	Nb	Mo	Ti	Al	Co	C
$C_i(\text{wt.}\%)$	18.96	18.33	5.23	3.01	1.00	0.59	0.03	0.03
$k_{s,i}(\text{MPa at Fraction}^{-1/2})$	337	153	1183	1015	775	225	39.4	1061

where μ_{77} and $\sigma_{y,77}$ are the shear modulus and the yield stress at 77 K, respectively.

Therefore, the yield stress model that takes account of the strengthening effects of alloying elements, deformation temperature, and strain rate can be summarized as,

$$\begin{cases} T\left(\frac{\sigma_y}{M\mu}\right)^{0.5} = K_1 - K_2 \frac{\sigma_y}{M} \\ K_1 = 77 \left(\frac{\sigma_{y,77}}{M\mu_{77}}\right)^{0.5} + K_2^m \frac{\sigma_{y,77}}{M} \\ \sigma_{y,77} = \left(\sum_i (k_{s,i} C_i^q)^{1/q}\right)^q \end{cases} \quad (23)$$

where K_2^m is used to correlate the strengthening constants at 77 K with K_1 .

The shear modulus for the studied Ni-based superalloy can be calculated as [41],

$$\mu = 7.89 \times 10^4 \left(1 - 0.64 \frac{T - 27}{1726}\right) \quad (24)$$

Based on the measured yield stresses, K_2^m and K_2 in eq. (23) can be determined. Figure 7 shows the strain rate dependence of the determined K_2^m and K_2 . It is clear that there are linear relationships between $\ln(K_2^m)$ and $\ln(\dot{\epsilon})$, as well as K_2 and $\lg(\dot{\epsilon})$. Thus, K_2^m and K_2 can be expressed as,

$$\begin{cases} K_2^m = 0.4343\dot{\epsilon}^{0.1618} \\ K_2 = 0.11747\lg(\dot{\epsilon}) + 0.00118 \end{cases} \quad (25)$$

Summarily, the yield stress model for the studied Ni-based superalloy that takes account of the effects of deformation temperature, strain rate and alloying elements can be expressed as,

$$\begin{cases} T\left(\frac{\sigma_y}{M\mu}\right)^{0.5} = K_1 - K_2 \frac{\sigma_y}{M} \\ \mu = 7.89 \times 10^4 \left(1 - 0.64 \frac{T - 27}{1726}\right) \\ K_1 = 77 \left(\frac{\sigma_{y,77}}{M\mu_{77}}\right)^{0.5} + K_2^m \frac{\sigma_{y,77}}{M} \\ \sigma_{y,77} = \left(\sum_i (k_{s,i} C_i^q)^{1/q}\right)^q \\ K_2^m = 0.4343\dot{\epsilon}^{0.1618} \\ K_2 = 0.11747\lg(\dot{\epsilon}) + 0.00118 \end{cases} \quad (26)$$

where K_2^m and K_2 can be determined by fitting to the measured yield stresses. The relationships between K_2^m , K_2 and strain rate can be further determined.

Figure 8 shows the comparisons between the measured (σ_y^m) and predicted (σ_y^p) yield stresses. The Pearson correlation coefficient (PCC) is employed to evaluate the linear dependence between σ_y^m and σ_y^p ,

Figure 8: Comparisons between the measured and predicted yield stresses. **Figure 7:** Strain rate dependence of: (a) K_2^m ; (b) K_2 (Symbols for the experimental results, the dashed line for the fitting line).

Figure 9: Effect of Nb content at different: (a) deformation temperatures; (b) strain rates.

$$\text{PCC} = \frac{\sum_{i=1}^{N_t} (\sigma_y^m - \bar{\sigma}_y^m) (\sigma_y^p - \bar{\sigma}_y^p)}{\sqrt{\sum_{i=1}^N (\sigma_y^m - \bar{\sigma}_y^m)^2 \sum_{i=1}^N (\sigma_y^p - \bar{\sigma}_y^p)^2}} \quad (27)$$

where $\bar{\sigma}_y^m$ and $\bar{\sigma}_y^p$ are the average value of σ_y^m and σ_y^p , respectively. N_t is the number of the tested conditions, $N_t = 12$. The calculated PCC is 0.979, which indicates that σ_y^p well correlates with σ_y^m .

Nb is an important precipitate-formatting element for the studied Ni-based superalloy. Thus, the strengthening effect of alloying element Nb is studied by the developed yield model. By changing the concentration constant of alloying element Nb (C_{Nb}) shown in eq. (26), yield stresses under different Nb content can be calculated. Figure 9 shows the strengthening effect of alloying element Nb at different deformation temperatures and strain rates. The yield stress decreases with the decrease in Nb content. The deformation temperature has negative effects on the strengthening effect of alloying element Nb (Figure 9a), while the strain rate has positive effects (Figure 9b). This can be attributed to the thermal assistance. High temperature weakens the bonding energy between alloying atoms and dislocations.

Conclusion

The yield behaviors of a solution-treated Ni-based superalloy are studied. The main conclusions can be drawn as follows.

- (1) The yield process is analyzed in terms of dislocation theory. A turning point can be found from the $\theta - \sigma$ plot at which the yield occurs. Thus, the yield stresses for the studied Ni-based superalloy are clearly determined.

- (2) A yield stress model that takes account of the strengthening effect of alloying elements, deformation temperature, and strain rate is developed. The predicted yield stresses for the studied Ni-based superalloy well agree with the measured ones.
- (3) The developed yield stress model shows that yield stress decreases with the decrease in Nb content. The deformation temperature has negative effects on the strengthening effect of alloying element Nb, while the strain rate has positive effects.
- (4) However, it should be emphasized that the developed yield stress model is only valid for the solution-treated Ni-based superalloy. The further research will incorporate the effects of precipitate strengthening into the yield stress.

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References

- [1] Y.C. Lin and M.S. Chen, *Mater. Des.*, 32 (2011) 1733–1759.
- [2] D.T. Nguyen, J.G. Park and Y.S. Kim, *Trans. Indian Inst. Met.*, 69 (2016) 1343–1350.
- [3] M.S. Chen, W.Q. Yuan, H.B. Li and Z.H. Zou, *Comput. Mater. Sci.*, 136 (2017) 163–172.

- [4] Y.C. Lin, F.Q. Nong, X.M. Chen, D.D. Chen and M.S. Chen, *Vacuum*, 137 (2017) 104–114.
- [5] P.J. Geng, W.G. Li, X.H. Zhang, J.X. Shao, Y. Deng, J.Z. Ma and H.B. Kou, *J. Alloys Compd.*, 684 (2016) 116–119.
- [6] R. Bobbili, V. Madhu and A.K. Gogia, *J. Mater. Res. Technol.*, 5 (2016) 190–197.
- [7] R. Bobbili, B.V. Ramudu and V. Madhu, *J. Alloys Compd.*, 696 (2017) 295–303.
- [8] G. Chen, X.C. Zhao and H. Wu, *Adv. Mech. Eng.*, 9 (2017) 1–21.
- [9] G. Chen, Z.S. Zhang, Y.H. Mei, X. Li, D.J. Yu, L. Wang and X. Chen, *Mech. Materials*, 72 (2014) 61–71.
- [10] D.D. Chen, Y.C. Lin, Y. Zhou, M.S. Chen and D.X. Wen, *J. Alloys Compd.*, 708 (2017) 938–946.
- [11] J. Cai, X.L. Zhang, K.S. Wang and C.P. Miao, *High Temp. Mater. Process.*, 35 (2016) 1037–1045.
- [12] X.L. Han, S.D. Zhao, C.Y. Zhang, S.Q. Fan and F. Xu, *High Temp. Mater. Process.*, 34 (2015) 529–538.
- [13] M.C. Cui, S.D. Zhao, C. Chen, D.W. Zhang and Y.Y. Li, *Adv. Mech. Eng.*, 9 (2017) 1–11.
- [14] M. Azarbarmas, M. Aghaie-Khafri, J.M. Cabrera and J. Calvo, *Mater. Sci. Eng. A*, 678 (2016) 137–152.
- [15] Y.Q. Ning, B.C. Xie, H. Li and M.W. Fu, *Adv. Eng. Mater.*, 17 (2015) 648–655.
- [16] D.Y. Hu, Q. Yang, H.W. Liu, J.X. Mao, F.C. Meng, Y.F. Wang, M.X. Ren and R.Q. Wang, *Int. J. Fatigue*, 95 (2017) 90–103.
- [17] Y.C. Lin, H. Yang, L. Li, *Vacuum*, 144 (2017) 86–93.
- [18] Y.C. Lin, X.M. Chen, D.X. Wen and M.S. Chen, *Comput. Mater. Sci.*, 83 (2014) 282–289.
- [19] H.B. Li and Y.L. Feng, *High Temp. Mater. Process.*, 35 (2016) 55–64.
- [20] S.A. Sajjadi, A. Chaichi, H.R. Ezatpour, A. Maghsoudlou and M. A. Kalaie, *J. Mater. Eng. Perform.*, 25 (2016) 1269–1275.
- [21] Y.H. Liu, Y.Q. Ning, Z.K. Yao, H. Li, X.P. Miao, Y.Z. Li and Z.L. Zhao, *J. Alloys Compd.*, 675 (2016) 73–80.
- [22] A. Momeni, *J. Mater. Res.*, 31 (2016) 1077–1084.
- [23] Z.L. Zhao, Y.Q. Ning, H.Z. Guo, Z.K. Yao and M.W. Fu, *Mater. Sci. Eng. A*, 620 (2014) 383–389.
- [24] H. Roth, C. Davis and R. Thomson, *Metall. Mater. Trans. A*, 28 (1997) 1329–1335.
- [25] A. Momeni, S.M. Abbasi, M. Morakabati and H. Badri, *Metall. Mater. Trans. A*, 48 (2017) 1216–1229.
- [26] B. Weiss and B. Grushko, *Weld. J.*, 62 (1983) 282–289.
- [27] E. Galindo-Nava, L. Connor and C. Rae, *Acta Mater.*, 98 (2015) 377–390.
- [28] D. Crudden, A. Mottura, N. Warnken, B. Raeisinia and R. Reed, *Acta Mater.*, 75 (2014) 356–370.
- [29] Y.X. Liu, Y.C. Lin and Y. Zhou, *Mater. Sci. Eng. A*, 691 (2017) 88–99.
- [30] X.M. Chen, Y.C. Lin, D.X. Wen, J.L. Zhang and M. He, *Mater. Des.*, 57 (2014) 568–577.
- [31] Y.C. Lin, M.S. Chen and J. Zhong, *Mech. Res. Commun.*, 35 (2008) 142–150.
- [32] P. Alexopoulos, C. Cho, C. Hu and L. Che-Yu, *Acta Mater.*, 29 (1981) 569–577.
- [33] R. Stoller and S. Zinkle, *J. Nucl. Mater.*, 283 (2000) 349–352.
- [34] Y. Estrin, L. Toth, A. Molinari and Y. Bréchet, *Acta Mater.*, 46 (1998) 5509–5522.
- [35] Y.C. Lin, D.X. Wen, M.S. Chen, Y.X. Liu, X.M. Chen and X. Ma, *J. Mater. Res.*, 31 (2016) 2415–2429.
- [36] Y.C. Lin, D.X. Wen, M.S. Chen and X.M. Chen, *Appl. Phys. A*, 122 (2016) 805.
- [37] G. Stepanov, *J. Appl. Mech. Tech. Phys.*, 23 (1982) 141–146.
- [38] L. Gypen and A. Deruyttere, *J. Mater. Sci.*, 12 (1977) 1028–1033.
- [39] Y. Mishima, S. Ochiai, N. Hamao, M. Yodogawa and T. Suzuki, *Trans. Jpn. Inst. Met.*, 27 (1986) 656–664.
- [40] P. Feltham, *J. Phys. D.*, 1 (1968) 303–308.
- [41] Y.X. Liu, Y.C. Lin, H.B. Li, D.X. Wen, X.M. Chen and M.S. Chen, *Mater. Sci. Eng. A*, 626 (2015) 432–440.