Microstructure transformation and twinning mechanism of 304 stainless steel tube during hydraulic bulging

G S Song, K S Ji, H W Song and S H Zhang

1 School of Materials Science and Engineering, Shenyang Aerospace University, Shenyang 110136, People’s Republic of China
2 Institute of Metal Research, Chinese Academy of Sciences, Shenyang 110016, People’s Republic of China
E-mail: hwsong@imr.ac.cn

Keywords: hydraulic bulging, the hollow shaft, martensite transformation, K-S relationship, deformation twin

Abstract
At room temperature, the hollow shaft of AISI 304 stainless steel tubes was produced by a hydraulic bulging process. The behavior of strain-induced austenite to martensite transformation and the twin crystallographic nature of AISI 304 stainless steel tubes at different positions after hydraulic bulging were discussed. The results have demonstrated that strain-induced austenite to martensite transformation occurred in AISI 304 stainless steel tubes during hydraulic bulging, resulting in the formation of the \( \alpha' \)-martensite phase, and the volume fraction of martensite gradually increased with an increase in strain. The austenite and \( \alpha' \)-martensite phases maintained lattice coherency throughout and followed the Kurdjumov–Sachs (K-S) relationship in terms of lattice coherency. During the deformation process, de-twinning occurred in the austenite and the deformation twins were formed in \( \alpha' \)-martensite. With the increase in strain, the volume fraction of the annealing twins gradually reduced until complete disappearance in the austenite. The volume fraction of the deformation twins increased in the martensite with an increase in strain, and finally reached saturation.

1. Introduction

Austenitic stainless steels have many excellent properties, such as high strength, corrosion resistance, good toughness, and pliability [1–3]. Therefore, austenitic stainless steels have been applied to the manufacturing of household products to the aviation, aerospace, and nuclear industries [4, 5]. The deformation microstructure of austenitic stainless steel is associated with the stacking fault energy (SFE). Related results show that austenitic stainless steels have low SFE, which inhibits slip, which is the primary deformation mechanism. During the cold deformation process, with the increase of dislocation density, strain-induced austenite to martensite transformation and deformation twin forming may occur in austenitic stainless steels [1, 6–8].

At room temperature, the microstructures of austenitic stainless steels are at a metastable state, so austenite is easy to be transformed into martensite. The strain-induced austenite to martensite transformation is a key factor in the high strength and good pliability of austenitic stainless steels [9, 10]. Many scholars have studied the strain-induced austenite to martensite transformation [11–15]. Odmobokova et al [16] studied the microstructure, texture evolution, and strengthening of 316L-type and 304L-type austenitic stainless steels during cold rolling. It was found that the microstructures were composed of flat austenite and martensite grains formed due to the formation of deformation twins and micro-shear bands and strain-induced austenite to martensite transformation during cold rolling. Das et al [17] used the tensile deformation of the metastable austenitic stainless steel with varying strain rates at room temperature to study the spatial distribution pattern of martensite in austenite grains at different strain scales. The results have shown that the distribution pattern of the martensite was closely related to the local stress and strain states of the stretched specimen, and the distribution pattern of the martensite were also strongly dependent on the crystallographic nature of martensite variation under applied stress. Meanwhile, the strain rate also affected the distribution pattern of the martensite. Garin et al [18] studied the influence of plastic deformation on the transformation of the austenite when the sample was thinned into the range of 0–25%. This study revealed that the retained austenite decreased with the
increase of the thinning scale, and, accordingly, the martensite fractions increased gradually. Nakada et al [19] investigated the behavior of strain-induced austenite to martensite transformation in cold-rolled and cold-drawn specimens of 316 type stainless steel. It was found that the strain-induced austenite to martensite first nucleated at the boundaries of the austenite matrix and the deformation twins. In cold-rolled samples, the martensite formed at the twin boundary had a Kurdjumov-Sachs (K-S) relationship with both the austenite matrix and the deformation twin (‘double K-S relationship’). In cold-drawn samples, two kinds of deformation twins with different twin planes were formed, and, therefore, strain-induced austenite to martensite transformation was formed from regions where the deformation twin boundaries intersected. Therefore, the formed martensite had an imperfect ‘triple K-S relationship’ with the austenite matrix and deformation twins.

As another micro deformation mode to slip, deformation twinning plays a key role in some metal deformations [20–22]. Sang et al [23] studied the interactions between twin and dislocation at different stages of the dynamic microstructure evolution of the Fe–38Mn austenitic steel during hot shear-compression deformation. It was proven that the step-like structure located at the boundaries of pre-existing annealing twins was induced by the interactions of twins and dislocations, in the case of a low strain level. These steps provided a favorable nucleation position for new grains. Benefiting from the large strain and high strain rate during the hot shear-compression deformation, deformation twins of different sizes appeared with further increased strain. Lee et al [24] studied the orientation characteristics and the formation mechanism of deformation twins in 18Cr18Mn–2Mo–0.9 N austenitic stainless steel. The deformation microstructures exhibited the character that plane dislocations were formed in the low strain region, and stacking fault and well-developed deformation twins were formed in the high strain region. The deformed twin had a twin system of {1 1 1} {1 1 2} and displayed strong orientation dependence along the direction of the tensile axis.

However, most of the previous studies on the microstructure formation and mechanism of deformation of stainless steels were performed under unidirectional loading. Related studies under multidirectional loading are lacking. In the stainless steel parts production process, plastic deformation of austenite stainless steel usually occurs under multidirectional loading, such as parts produced by hydraulic bulging technology of the tube. Hydraulic bulging is a newly developed forming technology in the field of plastic processing. It is an advanced manufacturing technology to realize a lightweight structure. Until now, it has been applied in the automotive industry [25]. Hydraulic bulging of the tube was accomplished by the interaction of pressure imposed by the fluid medium on the tube and axial thrust by external loading to obtain flexible forming technology of tubular parts with different cross-sections [26–28].

In the current work, microstructures and grain orientations of an AISI 304 stainless steel hollow shaft produced by hydraulic bulging were measured. Furthermore, the behavior of strain-induced austenite to martensite transformation and the twinning mechanism of AISI 304 stainless steel tubes after hydraulic bulging were investigated. It is helpful to understand the micromechanism of AISI 304 stainless steel tubes during complex deformation, and it has great significance in guiding the practical applications of austenite stainless steel.

2. Experimental materials and methods

The experimental material was a commercial AISI 304 stainless steel welded tube with a diameter (ø) of 22 mm and wall thickness (t) of 1.5 mm. The chemical composition of the steel tube is listed in Table 1. After solid solution heat treatment at 1110 °C and 0.5 h, the AISI 304 stainless steel welded tube was then used to produce a hollow shaft via hydraulic bulging performed on the THF-1500T CNC internal high-pressure forming machine. As shown in Figure 1, the hydraulic bulging process of the tube primarily includes six stages: feeding, die closing, liquid filling, pressure imposing, feed filling, part forming and part unloading. First, after the die and tube assembly was completed, the high pressure liquid was then filled into the tube and the internal pressure was applied. When the liquid pressure on the tube reached 70 MPa, the axial feeding started, and the internal pressure imposed on the tube by the liquid continued to increase and the axial feeding distance was 11 mm. Finally, when the axial feeding was completed, the internal pressure imposed by the liquid on the tube reached 180 MPa and the hydraulic bulging process was carried out, and the hollow shaft was removed. The hollow shaft is schematically presented in Figure 2.

| Table 1. Chemical composition of AISI 304 stainless steel tube (wt%). |
|-----------------|---|---|---|---|---|---|---|
| C | Si | Mn | P | S | Cr | Ni | N |
| 0.036 | 0.35 | 1.26 | 0.039 | 0.008 | 18.21 | 7.91 | 0.063 |
Figure 2 shows that the hollow shaft has an egg shaped section, which means the initial tube experienced an irregular radial deformation. A 3 mm sample was cut along the radial section of the shaft in the axial thickness. The sectional shape of the sample is shown in figure 3(a), where five different positions (numbered as positions 1, 2, 3, 4, and 5) were selected on the corresponding radial sections of the sample to measure their microstructures and grain orientations. Similarly, as shown in figure 3(b), a ring sample with an axial thickness of 3 mm was cut (numbered as position 0) from the original tube. In figure 3, RD represents the radial direction, ED represents the extrusion direction, C represents the center of the sample, and the green colored part represents the welding seam of the original tube. The radial wall thickness at positions 0–5 in figure 3 and the distance between those positions and the center, C, of the tube were measured and the micro-region x-ray diffraction (Micro-XRD) experiments were carried out to determine the volume fraction of each phase in the microstructure. The electron backscattering diffraction (EBSD) experiments were carried out at three positions along the sample (positions 1, 2, and 4) and the original tube (position 0) in figure 3, to determine the grain orientation characteristics.

The volume fraction of austenite and martensite of the selected positions in figure 3 were measured by a Bruker D8 Discover two-dimensional x-ray micro-region diffractometer. In these experiments, the source of x-ray radiation was Co, and the parameter, 2θ angle, started from 40° and ended at 100°, where a two-dimensional spectra was scanned every 1.5°, accordingly, and every position was scanned 5 times. The scanning time of each two-dimensional spectrum was 400 s. The diameter of the collimated tube of the beam device was 0.3 mm. For every position, greater than five measured two-dimensional spectra were integrated into a whole, and then the two-dimensional spectra were converted into a diffraction line, where the angle range of the spectral line was 40–105°, and the step size was 0.005°. In EBSD experiments, the grain orientation was measured by a Zeiss Gemini SEM500/300 field emission scanning electron microscope with an EBSD system.
The magnification was 500×, the step size was 1 μm, and the observation direction was the extrusion direction (ED) as shown in figure 3. The samples were ground, mechanically polished, and electrolytically polished. The electrolyte was HClO₄-C₂H₅OH mixed solution with a volume ratio of 1:9, the electrolytic voltage was 30 V, and the electrolytic time was 25 s. The measured grain orientation data was processed by HKL Channel5 software.

3. Experimental results and analysis

3.1. Strain-induced austenite to martensite transformation

During the hydraulic bulging process of the hollow shaft, the deformation of each measured position of the shaft section was different because of the irregular section shape, which leads to different radial wall thicknesses of the shaft sections after deformation. The radial wall thickness and strain of each measured position on the shaft, the distance between those positions and the original center C are listed in table 2. The strain of each position during the hydraulic bulging of the hollow shaft was simulated by the Dynaform software, and other data in table 2 was obtained from the experimentally measured results. As shown in table 2, the radial wall thickness of the section decreased after the tube had been bulged, and the thinning amount was significantly different at different positions, and the thinning amount increased as distance increased between those positions and the original center C of the section.

![Figure 3. Schematic of the section shapes and measured positions on the shaft and the tube; (a) the shaft; (b) the original tube.](image)

| Positions | 0  | 1   | 2   | 3   | 4   | 5   |
|-----------|----|-----|-----|-----|-----|-----|
| Distance from C/mm | 10.25 | 14.25 | 14.69 | 17.14 | 18.59 | 19.17 |
| Wall thickness /mm | 1.5 | 1.404 | 1.324 | 1.286 | 1.157 | 1.058 |
| Equivalent strain | 0 | 0.427 | 0.446 | 0.480 | 0.585 | 0.618 |

The Micro-XRD diffraction patterns at different positions on the shaft section are presented in figure 4. The strains in the figure correspond to those in table 2. It can be observed from figure 4 that the diffraction pattern of the original tube showed a single diffraction peak of austenite, which indicated that the original microstructure of the tube was only composed of the austenite. Moreover, the grain orientation distributions in figure 4 showed that the {111} \( \perp \) ED and {200} \( \perp \) ED textures had been formed. Figure 4 also shows that after hydraulic bulging, the diffraction peaks of the martensite and the austenite simultaneously appear in the diffraction pattern of the shaft section, which means that, during the hydraulic bulging process, parts of the austenite in the tube were transformed into martensite. Meanwhile, as shown in figure 4, the Micro-XRD diffraction patterns of the martensite show that the {110} \( \perp \) ED texture was formed during deformation. With the increase in strain, the texture strength of {211} \( \perp \) ED in martensite grains significantly increased, and the texture strength of {220} \( \perp \) ED in austenite grains also improved. During the plastic deformation process, the metal of the cubic crystal structure can produce different kinds of textures because of the different deformation modes. For the rolling of face centered cubic (FCC) metals, {112} \{111\}, {011} \{100\}, and {123} \{634\} textures usually formed [29]. The {011} \{100\} and {011} \{211\} textures in the austenite and {001} \{110\}, {112} \{110\} and {332} \{113\} textures in the martensite usually formed for austenitic stainless steel rolling [10].

According to the crystallographic natures of different crystals, the martensite can be divided into \( \alpha' \)-martensite of body centered cubic (BCC) type and \( \varepsilon \)-martensite of hexagonal close packed (HCP) type. For
deformations at room temperature, results of the XRD test showed that the diffraction peaks of the austenite and α′-martensite could be clearly observed, but the diffraction peak of the ε-martensite could be hardly observed \[30\], which was due to the ε-martensite, as a transition phase, quickly transforming into α′-martensites during the deformation process \[31\]. As shown in figure 4, in the range of 2θ = 40° ~ 105°, the main diffraction peaks were γ(111), γ(200), γ(220), α′(110), α′(200), and α′(211). No corresponding diffraction peaks of ε-martensites were observed, which indicated that the strain-induced austenite to martensite transformation occurred in AISI 304 stainless steel tube during the hydraulic bulging process.

To further quantitatively analyze the relationship between the volume fraction of α′-martensite and the strain, the diffraction peaks in figure 4 were quantitatively calculated by the direct comparison method \[5, 32\], and the calculated results are shown in figure 5. The results showed that the content of α′-martensite in AISI 304 stainless steel increased with the increase in strain during deformation.

Figure 6 presents the distributions of different phases at different positions on the shaft. In figure 6, the green regions represent the austenite, the yellow regions represent the α′-martensite, the black lines represent grain boundaries, and TD represents the tangential direction. As shown in figure 6(a), the microstructure of the original tube was only composed of the austenite, while figures 6(b)–(d) show that α′-martensite appeared in the austenitic matrix after deformation, and the volume fraction of the α′-martensite increased with the increase in strain. As shown in figure 6, the strip or block shaped α′-martensite appeared and were mainly distributed within the austenitic grains, and a small amount of them were distributed at the grain boundaries.

To analyze the crystallographic nature of the martensite transformation during former deformation, six austenite and six martensite grains were separately extracted from the phase distribution figures as shown in figures 6(b)–(d), and numbered as γ1~γ6 and α1~α6, respectively. According to the orientation values of each grain measured by EBSD, the pole figures of the austenite and α′-martensite grains were separately drawn, as shown in figure 7. For the orientation characteristics of austenitic grains in the figure, the yellow triangles represent lattice planes of \{1\}11, and the purple circles represent lattice planes of \{1\}10; for the orientation characteristics of martensite grains, green triangles represent lattice planes of \{1\}11, and red circles represent lattice planes of \{1\}10. Black circles in the figure represent parallel lattice planes between the austenite and martensite. Black rectangles and lines of L1 ~ L3 and L1′~L3′ in the figure represent parallel lattice directions between the austenite and martensite. Black rectangles and lines of L1 ~ L3 and L1′~L3′ in the figure represent parallel lattice directions between the austenite and martensite. It can be seen from the pole figures that the following orientation relationships exist between the austenite and α′-martensite: (111)·1 // (101)·γ1, (111)·2 // (101)·γ2, L1 // L1′, (111)·3 // (110)·α3, (111)·4 // (011)·α4, L2 // L2′, (111)·5 // (101)·α5, (111)·6 // (101)·α6 and L3 // L3′, which indicate that the austenite and the martensite follow a K-S relationship during the process of strain-induced
3.2. Twinning

During the plastic deformation process of metals, in addition to slip, twinning is an important micro deformation mode [20, 21], especially for HCP metals with low atomic arrangement symmetry. Twinning is relatively easier to initiate during the deformation process [34]. The twin formed during the deformation process is called the deformation twin. For the metals with low SFE, an annealing twin is easy to be formed.
during the annealing process after deformation [35]. Annealing twin forming is essentially the release of internal stress causing lattice shearing during the annealing process.

For the original tube and hollow shaft, orientation micrographs are presented in figure 8, which shows, during the hydraulic bulging process of the hollow shaft, how the deformation affects the microstructure of the AISI 304 stainless steel. As shown in figure 8(a), the microstructure of the original tube is composed of grains of non-uniform size, and there are some lath-shaped twin bands formed within the grains. Most twin bands intersected with grain boundaries, and, for some twin bands with different orientations, they were
simultaneously present within a single grain and corresponded to different twin variants. Twin bands in figure 8(a) were formed at the recovery stage of the tube solid solution heat treatment, and these annealing twins were actually formed by the large-angled grain boundary migration [21, 35, 36]. As shown in figures 8(b)–(d), the microstructure of the tube has significantly changed after hydraulic bulging, and the annealing twins in the original tube gradually disappeared with strain increase, which indicated de-twinning had occurred during the deformation process. With the increase in strain, more fine grains were formed within original grains, and the grain size was much smaller than that of parent grains. The volume fraction of newly formed lath-shaped martensite within grains also increased with the strain increase.

Figure 9 shows distribution patterns of the austenite and martensite in the original tube and shaft, and twin boundaries within each phase are also shown in this figure, where the green region represents the austenite, the yellow region represents the $\alpha'$-martensite, the red lines represent twin boundaries in the austenite, and the black lines represent twin boundaries in the $\alpha'$-martensite. According to the comparisons among figures 9(a), (b), (d), and (f), it can be seen that the volume fraction of twins in austenite of the original tube was greatly reduced after hydraulic bulging, which indicated that de-twinning had occurred during the deformation process. As shown in figures 9(b), (d), and (f), fine twins were distributed within austenite grains after deformation. As shown in figures 9(c), (e), and (g), the volume fraction of martensite gradually increased with the increase of strain, showing that more austenite was transformed into martensite, and, simultaneously, fine deformation twins were formed within martensite grains. The deformation twins were distributed within parent grains and the volume fraction of deformation twins increased with the increase in strain.

This analysis reveals that during the solid solution heat treatment of the tube, annealing twins were formed in the austenite, and the de-twinning occurred in the subsequent deformation process. Additionally, a large number of deformation twins were formed in the $\alpha'$-martensite.

Due to symmetric differences in atomic arrangements of different crystal structures, twin systems formed by different crystal structures usually display different crystallographic characteristics. The twin system characteristic is usually described by the crystallographic nature of twin plane and twin direction. The predominant twin system of FCC metals is $\{111\}11 \bar{2}$ twin [37, 38], and for BCC metals is $\{112\}11 \bar{1}$ twin [39]. It is known that obvious misorientation exists between the twin and its corresponding parent grain. Combining the rotation axis with the rotation angle is a good way to describe the misorientation between the twin and the parent grain [34]. For the twin $\{111\}11 \bar{2}$, the corresponding rotation axis and rotation angle are separately $60^\circ$ and $(111)$ [34], and the corresponding measurements for the twin $\{112\}111$ are also $60^\circ$ and $(111)$ [40].
The distributions of misorientation angle and rotation axis in austenitic grains within the samples are shown in figure 10. It is clearly observed that the 60° misorientation angle occupies the highest percentage among the whole range of misorientation angle values. In addition, the inverse pole figures in figure 10 reveal that the rotation axis is concentrated on the \(\langle 111\rangle\) crystal orientation. Above results revealed that a large number of twins with \(\langle 111\rangle\) 60° misorientation existed in the austenite.

As shown in figure 10, it was found that the number of twins in the original tube was significantly higher than that in the shaft, which means that detwinning results in the decrease of twins. Furthermore, by comparing the misorientation angle distributions in figure 10(b), (c), and (d), it can be found that the number of twins within...
grains decreased from position 1 to position 4. According to the equivalent strains corresponding to positions 1, 2, and 4, as listed in table 1, it was concluded that the \{111\} 112 twin volume fraction decreased with increasing strain scale. It is known that the twin content decreased with increasing the strain scale during the detwinning process \cite{41, 42}. Additionally, in some cases, the original twin completely disappears when the strain was higher than some values \cite{43, 44}.

The distributions of rotation axis and misorientation angle in martensitic grains within the shaft are shown in figure 11. It can be seen that the frequency of the 60° misorientation angle was significantly higher than others, and inverse pole figures in this figure display that the rotation axis is concentrated on the \{111\} crystal orientation, which means that many \{112\} 111 twins were formed in martensite during the hydraulic bulging process.

With the comparison between figure 11(a) and (b), it was found that the frequency of the 60° misorientation angle at positions 2 and 4 was higher than that in position 1, meaning the twin content was increased with increasing strain scale. As shown in figure 11(b) and (c), the frequency of the 60° misorientation angle was almost the same at positions 2 and 4, revealing the twin volume percentage was nearly equivalent at both positions, implying that the deformation twin reached saturation as the strain scale reached a certain degree.

### 4. Conclusions

(1) During the hydraulic bulging process of AISI304 stainless steel tube, the strain-induced austenite to martensite transformation occurred in austenite to form \(\alpha'\)-martensite. With an increase in strain, the volume fraction of \(\alpha'\)-martensite gradually increased.

(2) The analysis on the orientation relationship between the austenite and martensite showed that, during the process of strain-induced austenite to martensite transformation, the lattice coherency was always maintained between the austenite and \(\alpha'\)-martensite, and the K-S relationship was followed.
(3) During the hydraulic bulging process, the original annealing twin in austenitic grains has been de-twined, causing the volume fraction of the annealing twin in austenite to become greatly reduced.

(4) A large number of the deformation twins were formed in the newly formed $\alpha'$-martensite grains, but with an increase in strain, the deformation twin content will reach saturation.

Acknowledgments

This research was funded by National Natural Science Foundation of China (51875547).

ORCID iDs

K S Ji @ https://orcid.org/0000-0003-0415-9345
H W Song @ https://orcid.org/0000-0003-4872-6807

References

[1] Souza Filho I R, Zilnyk K D, Sandim M J R, Bolmaro R E and Sandim H R Z 2017 Materials Science and Engineering: A 702 161
[2] Xu Y, Zhang S H, Cheng M and Song H W 2012 Scripta Materialia 67 771
[3] Padilha A F, Plaut R I and Rios P R 2003 ISIJ International 43 135
[4] Lo K H, Shek C H and Lai J K L 2009 Materials Science and Engineering: R: Reports 65 39
[5] Zhang X S, Xu Y, Zhang S H, Cheng M, Zhao Y H, Tang Q S and Ding Y X 2017 Acta Metallurgica Sinica 53 335
[6] Olson G B and Cohen M 1975 Metallurgical Transactions A 6 791
[7] Talonen J and Hanninen H 2007 Acta Materialia 55 6108
[8] Tavares S S M, Pardal J M, Silva M J G D, Abreu H F G and da Silva M R 2009 Materials Characterization 60 907
[9] Hilkhuijsen P, Geijelselaers H J M and Bor T C 2013 Journal of Alloys and Compounds 577 5609
[10] Kumar B R, Singh A K, Das S and Bhattacharya D 2004 Materials Science and Engineering: A 364 132
[11] Tsakiris V and Edmonds D V 1999 Materials Science and Engineering: A 273–275 430

Figure 11. The distribution of rotation axis and misorientation angle in martensite within the shaft; (a) position 1; (b) position 2; (c) position 4.
[12] Souza Filho I R, Sandim M J R, Cohen R, Nagamine L C C M, Hoffmann I, Bolmaro R E and Sandim H R Z 2016 Journal of Magnetism and Magnetic Materials 419 156
[13] Gauss C, Filho I R S, Sandim M J R, Suzuki P A, Ramirez A J and Sandim H R Z 2015 Materials Science and Engineering: A 651 507
[14] Ryoo D Y, Kang N and Kang C Y 2011 Materials Science and Engineering A 528 2277
[15] Ahn T H, Lee S B, Park K T, Oh K H and Han H N 2014 Materials Science and Engineering: A 598 56
[16] Odnobokova M and Belyakov A 2018 and Kaibyshev R Philosophical Magazine 99
[17] Das A 2016 Materials Science and Engineering: A 658 484
[18] Garin J J and Mannheim R L 2003 Journal of Materials Processing Technology 143–144 347
[19] Nakada N, Ito H, Matsuoka Y, Tsuichiya T and Takaki S 2010 Acta Materialia 58 895
[20] Murr L E, Meyers M A, Nious C S, Chen Y J, Pappu S and Kennedy C 1997 Acta Materialia 45 157
[21] Molodov K D, Al-Samman T and Molodov D A 2017 Acta Materialia 124 397
[22] Abd El-Aty Ali, Xu Yong, Guo Xunzhong, Zhang Shi-Hong, Ma Yan and Chen Dayong 2018 Strengthening mechanisms, deformation behavior, and anisotropic mechanical properties of Al-Li alloys: A review Journal of Advanced Research 10 49–67
[23] Sang D L, Fu R D, Li Y J, Wang Y P and Kang J 2018 Journal of Alloys and Compounds 735 2395
[24] Lee T H, Oh C S, Kim S J and Takaki S 2007 Acta Materialia 55 3649
[25] Fuchizawa S 2007 Journal of Plasticity Engineering 14 171
[26] Lang I H, Yuan S J, Wang Z R, Wang X S, Danckert J and Nielsen K B 2004 Proceedings of the Institution of Mechanical Engineers, Part B: Journal of Engineering Manufacture 218 43
[27] Abrantes J P, Szabo-Ponce A and Batalha G F 2005 Journal of Materials Processing Technology 164–165 1140
[28] Mummer K and Altan T 2001 Journal of Materials Processing Technology 108 384
[29] Humphreys F J and Hatherly M 2004 Recrystallization and Related Annealing Phenomena 67
[30] De A K, Murdock D C, Mataya M C, Speer J G and Matlock D K 2004 Scripta Materialia 50 1445
[31] Petit B, Gray N, Cherkasov M, Bolle B and Humbert M 2007 International Journal of Plasticity 23 323
[32] Moser N H, Gross T S and Korkolis Y 2014 Metallurgical and Materials Transactions A 45 4491
[33] Liu Z C, Li Y P and Ren H P 2018 Heat Treatment Technology and Equipment 39 1
[34] Song G S, Zhang S H, Cheng M and Wang B 2012 Advanced Materials Research 472–475 700
[35] Yang G, Sun J I, Zhang L N, Wang L M and Wang C 2009 Journal of Iron and Steel Research 21 39
[36] McCormack Scott J, Wen Wei, Pereloma Elena V, Tomé Carlos N, Gazder Azdair A and Saleh Ahmed A 2018 On the first direct observation of de-twinning in a twinning-induced plasticity steel Acta Materialia 156 172–82
[37] Mahajan S 2013 Scripta Materialia 68 95
[38] Christian J W and Mahajan S 1993 Progress in Materials Science 39 1
[39] Visser W and Ghonem H 2017 Materials Science and Engineering: A 687 28
[40] Bouyne E, Flower H M, Lindley T C and Pineau A 1998 Scripta Materialia 39 295
[41] Wu B L, Duan G S, Du X H, Song L H, Zhang Y D, Philippe M J and Eslin C 2017 Materials and Design 135 57
[42] Şarl N, Yıldız G D, Yıldız Y G and Yağcı N K 2019 Physica B: Condensed Matter 553 161
[43] Wu L, Jain A, Brown D W, Stoica G M, Agnew S R, Clausen B, Fieled D E and Liaw P K 2008 Acta Materialia 56 688
[44] Murphy-Leonard A D, Pagan D C, Beaudoin A, Miller M and Pand Allison J E 2019 International Journal of Fatigue 125 314