Communication

Influence of State of Stress on Dynamic Recrystallization in a Titanium-Modified Austenitic Stainless Steel

SUMANTRA MANDAL, A.K. BHADURI, and V. SUBRAMANYA SARMA

The influence of the state of stress on the microstructure and dynamic recrystallization (DRX) in a titanium-modified austenitic stainless steel is assessed by performing plane-strain and uniaxial hot compression studies. Although the state of stress does not alter the mechanisms of DRX nucleation, the kinetics of DRX is hindered during plane-strain deformation vis-à-vis uniaxial deformation.

DOI: 10.1007/s11661-011-1015-2

Dynamic recrystallization (DRX) is the most important restoration mechanism during the hot deformation of low-stacking fault energy, face-centered cubic alloys. The DRX process influences mainly the final microstructure and mechanical properties of the deformed material and, thus, the formability of materials.[1] Microstructure control through DRX requires a detailed knowledge of microstructural evolution as a function of processing parameters (*i.e.,* strain, strain rate, and temperature). A special emphasis needs to be given in understanding the nucleation mechanisms during DRX. In our recent study, microstructural evolution and DRX behavior in a Ti-modified austenitic stainless steel (referred to as alloy D9) during hot working was investigated over a range of temperature (1173 K to 1373 K [900 °C to 1100 °C]) and strain (0.2 to 0.8) at a strain rate 10 seconds^{-1}.[2] The nucleation of new DRX grains has been found to occur by bulging of the parent grain boundary. The processing parameters have been found to influence strongly the grain boundary character distribution and DRX in alloy D9.

In addition to the processing parameters, the mode of deformation (*i.e.,* state of stress) is expected also to influence the microstructure evolution and DRX behavior of materials. Although most of the studies on DRX reported in the literature are based on uniaxial hot-compression tests,[1,3–7] a few studies are also based on plane-strain hot compression.[8,9] However, to the best of our knowledge, the influence of the state of stress (*i.e.,* uniaxial *vs* plain-strain compression) on DRX kinetics and mechanisms has not been reported in the open literature. The objective of the current study is to investigate the role of state of stress on microstructural evolution and DRX behavior of alloy D9.

The alloy D9 (in solution-annealed condition) used in this study is an austenitic stainless steel containing Fe-0.05C-1.51Mn-0.51Si-0.002S-0.01P-15.05Cr-15.07Ni-2.25Mo-0.21Ti-0.01Co-0.006N (in wt pct). Uniaxial hot-compression tests were conducted on cylindrical specimen of 10 mm diameter and 15 mm height, whereas the plane-strain hot compression tests were conducted on a $20 \times 15 \times 10$ mm^3 solid bar using a specially designed anvil in a Gleeble thermomechanical simulator. The compression tests were performed in the temperature range of 1173 K to 1373 K [900 °C to 1100 °C] (in steps of 100 K) at a strain rate of 10 seconds^{-1}. Various degrees of deformation (to true strains of 0.2, 0.4, 0.6, and 0.8) were imparted at each temperature to study the influence of strain on microstructural evolution. The hot-deformed specimens were water quenched with a spray nozzle as soon as the specified amount of strain was imparted (within 1 to 2 seconds) to freeze the hot-deformed microstructure. Electron back scatter diffraction (EBSD) scans were performed on all processed samples using a TSL–OIM system attached to scanning electron microscope operating at 30 kV. Misorientations above 2 deg were considered grain boundaries, whereas Brandon's criterion[10] was used for identifying coincident site lattice boundaries.

The study of the microstructural evolution at 1173 K (900 °C) during plane-strain compression revealed that DRX is insignificant at strains of 0.2, 0.4, and 0.6 (Figure 1(a)). It could be observed that the grain and twin boundaries at 0.6 strain are mainly straight. The preexisting twin boundaries have started to lose their orientation [as the misorientation deviates from the original Σ3 misorientation of 60 deg with an allowed deviation up to 8.7 deg as per Brandon criterion (Figure 1(a))], and gradually, these are converted to random high-angle boundaries. It is important to note that the grain boundaries become serrated in nature, which signifies the initiation of DRX in similar deformation condition (*i.e.,* 0.6 strain at 1173 K [900 °C]/10 seconds^{-1}) during uniaxial compression of alloy D9 (Figure 1(b)). The DRX grains nucleate predominantly at parent grain boundaries (shown by arrow in Figure 1(b)). During plane-strain deformation at 1173 K [900 °C]/10 seconds^{-1}, DRX is found to initiate only at 0.8 strain. It is observed that DRX grains nucleate along prior grain boundaries by bulging mechanism, and this is similar to the nucleation mechanism observed in uniaxial compression of the same alloy.[2]

The microstructural evolution at 1273 K (1000 °C) reveals that DRX is marginal at 0.4 strain. However, a rapid progress of DRX at higher strains is observed at this deformation temperature because higher deformation

SUMANTRA MANDAL, Scientific Officer 'E', is with the Materials Technology Division, Indira Gandhi Centre for Atomic Research, Kalpakkam 603102, India. Contact e-mail: sumantra@igcar.gov.in A.K. BHADURI, Associate Director, is with the Materials Development and Technology Group, Indira Gandhi Centre for Atomic Research. V. SUBRAMANYA SARMA, Associate Professor, is with the Department of Metallurgical and Materials Engineering, Indian Institute of Technology Madras, Chennai 600036, India.
Manuscript submitted August 5, 2011.
Article published online December 10, 2011

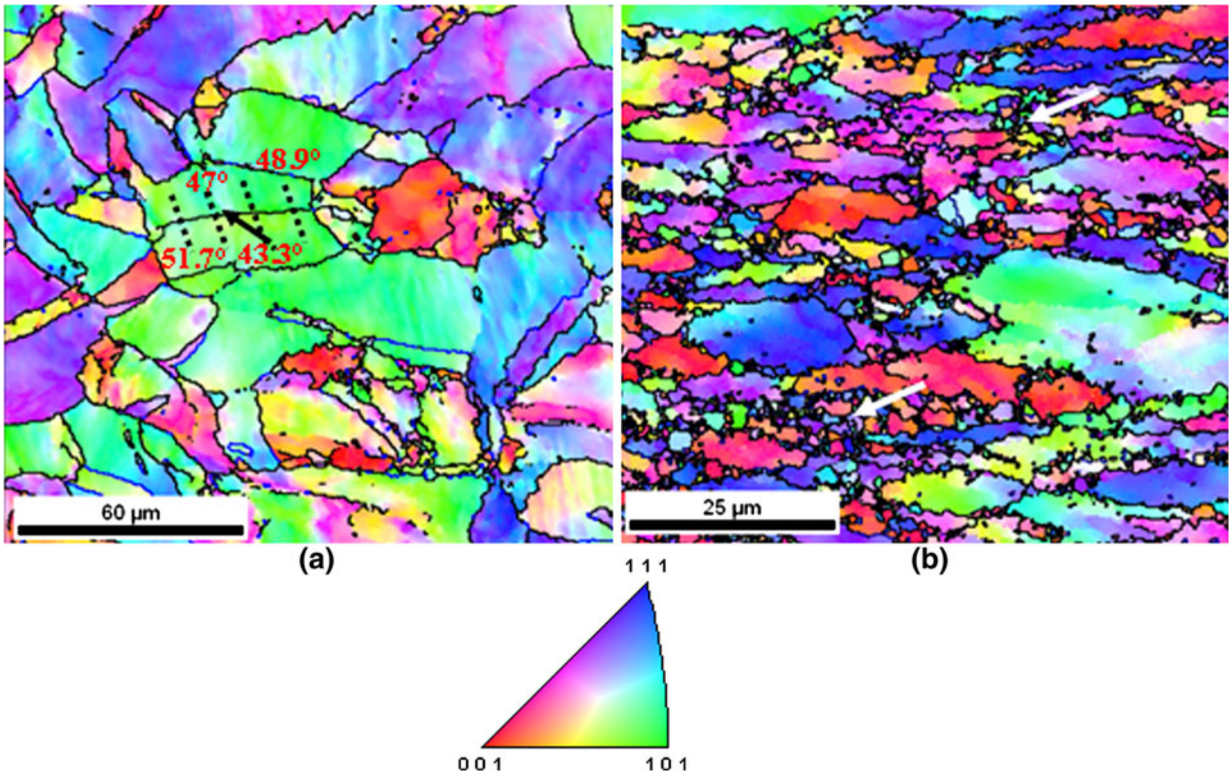

Fig. 1—Inverse pole figure (IPF) map of the specimen deformed to 0.6 strain by (*a*) plane-strain and (*b*) uniaxial compression at 1173 K [900 °C] (color code: Σ3—blue, other high-angle boundaries—black). (Color figure online).

increases the defect density resulting in increased mechanical instability (*i.e.,* higher driving force) in the microstructure. A partially recrystallized structure is obtained at 0.8 strain following both the plane-strain (Figure 2(a)) and uniaxial (Figure 2(b)) deformation. The proportion of Σ1 boundaries in the specimen following plane-strain deformation is higher (~43 pct) (Figure 2(a)) compared with the specimen subjected to uniaxial compression (~35 pct) (Figure 2(b)). The size of DRX grains was found to be virtually unchanged with increase in strain at a given temperature. This indicates that the grains grew rapidly to their final size after nucleation at a critical strain because work hardening occurs in the DRX grains during subsequent deformation and driving force for growth is reduced.[11] The DRX grains contain a high fraction of Σ3 boundaries. The straight and parallel morphology of these Σ3 boundaries suggest these are mainly coherent annealing twins.

Microstructural investigation on samples deformed at 1373 K (1100 °C) reveals that the fraction of DRX grains increases rapidly with strain when compared with those deformed at 1273 K (1000 °C). An increase in temperature provides more thermal energy for grain boundary migration and thereby accelerates the DRX process. The microstructure following plane-strain deformation to 0.8 strain level at 1373 K (1100 °C) is shown in Figure 2(c). The microstructure reveals equiaxed grain structure as a result of rapid development of DRX. Although most grains in this microstructure are new DRX grains, a closer look at Figure 2(c) reveals the presence of fraction of Σ1 boundaries (yellow lines). This indicates that the microstructure in this condition is not fully recrystallized. In contrast, a fully DRX structure is achieved in similar deformation condition during uniaxial compression of alloy D9 (Figure 2(d)).

The trends in the evolution of Σ1 and Σ3 boundaries following plane-strain compression at 10 seconds^{-1} are found to be similar to that of uniaxial compression. A representative plot of the fractions of Σ3 and Σ1 boundaries after plane-strain and uniaxial deformation at 1373 K (1100 °C) is shown in Figure 3. The average deviation of Σ3 boundaries following plane-strain and uniaxial compression to different strains at 1373 K (1100 °C)/10 seconds^{-1} is also shown in this figure. It may be noted here that the fraction of Σ3, Σ1, and average deviation of Σ3 in each sample is analyzed from at least two maps (area of each map ~500 × 500 μm^2) to ensure statistical significance. The data reported in this study is the average value obtained from these maps. A continuous increase in the fraction of Σ3 boundaries and a decrease in the fraction of Σ1 boundaries with increasing strain is observed during both plane-strain and uniaxial compression. However, the fraction of Σ3 boundaries after plane-strain deformation is relatively lower and the fraction of Σ1 boundaries is higher when compared with those subjected to uniaxial compression. Figure 3 indicates also that the average deviation of Σ3 boundaries decreases with increase in deformation level during both plane-strain and uniaxial compression.

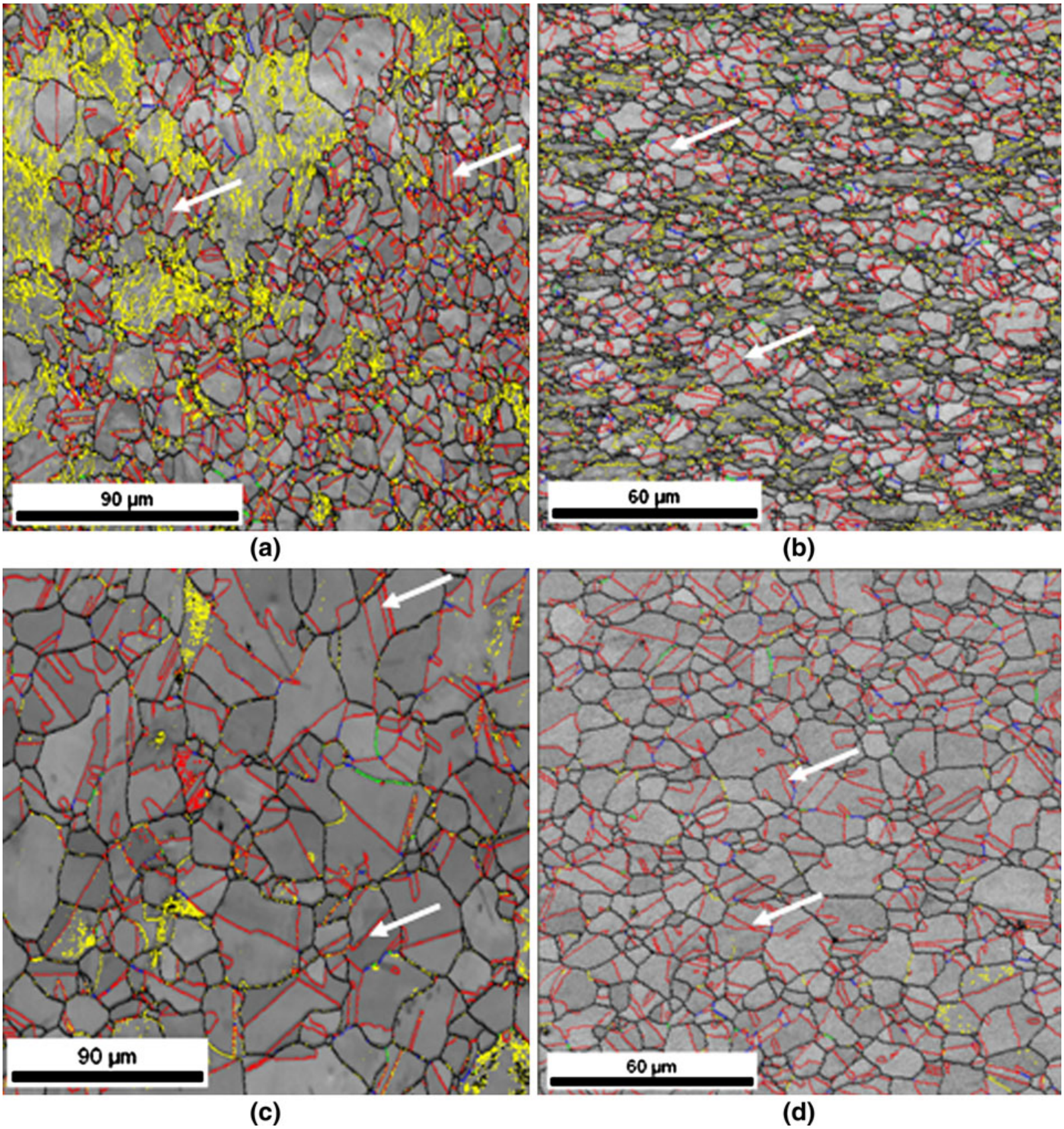

Fig. 2—Grain boundary image quality maps of the specimen compressed to 0.8 strain at (*a*) 1273 K [1000 °C] by plane-strain, (*b*) 1273 K [1000 °C] by uniaxial compression, (*c*) 1373 K [1100 °C] by plane-strain, and (*d*) 1373 K [1100 °C] by uniaxial compression (color code: Σ3—red; Σ9—blue; Σ27—green; Σ1—yellow; other boundaries—black). (The arrows indicate the presence of Σ3 boundaries within DRX grains.). (Color figure online).

Based on the observation of decrease in average deviation of Σ3 boundaries with an increase in deformation level, it is inferred most Σ3 boundaries are freshly nucleated. Otherwise, an increase in imposed deformation level would have increased the average deviation of these Σ3 boundaries. It is believed that formation of Σ3 twins during DRX is caused by grain boundary migration[12–14] as in the case of annealing twins[15] and not by the slip activity within the grains.[14] As suggested by Gleiter,[16] annealing twins may form during recrystallization by "growth accidents." During the growth of nucleated DRX grains, a "growth accident" may take place at the moving grain boundary and generate twin boundaries. This is in line with our observations that most annealing twins are found within the DRX grains (see arrows in Figure 2), suggesting thereby that these are formed during the growth of DRX grains. Twinning caused by "growth accidents" will change the growth rate and can add to the nucleation rate and thereby influence the kinetics of DRX.[12] The twin boundaries were found to play an important role during nucleation and subsequent expansion of the DRX process during uniaxial compression in alloy D9.[2] In line with the observation reported on

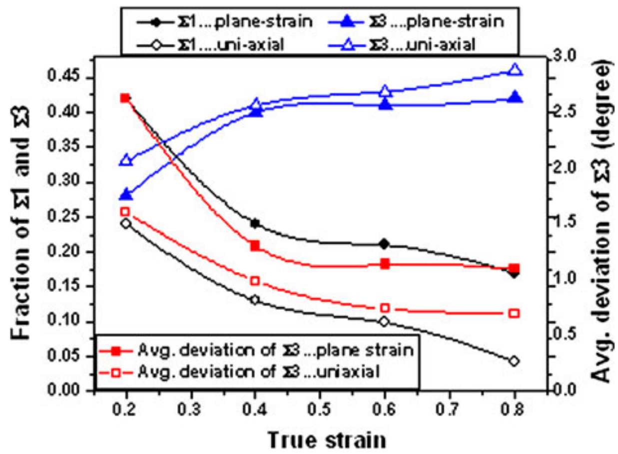

Fig. 3—Fraction of Σ1, Σ3 boundaries and average deviation of Σ3 boundaries following plane-strain and uniaxial compression to different strains at 1373 K [1100 °C]/10 s^{-1}.

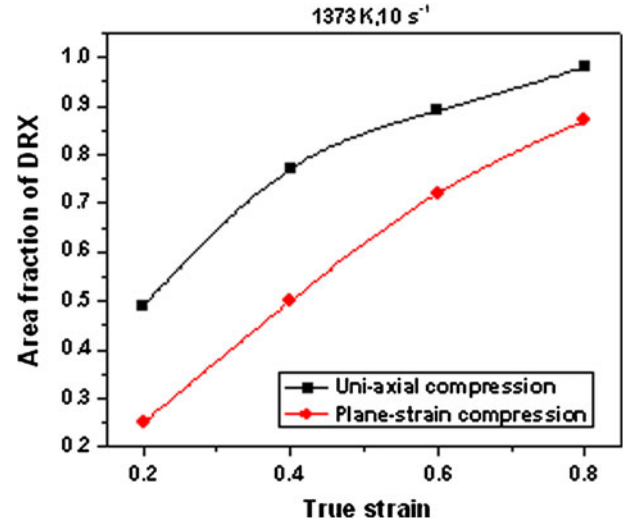

Fig. 4—DRX fraction at various strain levels after uniaxial and plane-strain compression at 1373 K [1100 °C]/10 s^{-1}.

uniaxial compression studies,[2] the presence of Σ3 boundaries within DRX grains underlines also the role of annealing twins on DRX process during plane-strain hot compression. This emphasizes that nucleation mechanisms of DRX during plane-strain hot compression of alloy D9 are similar (*i.e.,* by bulging and annealing twinning) to that of uniaxial compression. In other words, the state of stress does not affect/alter the nucleation mechanisms of DRX.

It could be observed that a considerable proportion (~17 pct) of Σ1 boundaries exist in the microstructure of the specimen when deformed to 0.8 strain by plane-strain compression at 1373 K (1100 °C)/10 seconds^{-1} (Figure 3). However, a low proportion (~4 pct) of Σ1 boundaries could be observed following uniaxial compression (Figure 3). These results indicate that DRX is hindered during plane-strain deformation vis-à-vis uniaxial deformation. It has been observed that there is absence of DRX at 0.6 strain after plane-strain compression at 1173 K/10 seconds^{-1} (Figure 1(a)), whereas DRX is observed in similar deformation condition after uniaxial compression (Figure 1(b)). With a view to understand the effect of state of stress on the kinetics of DRX, the area fraction of DRX at different strain levels after plane-strain and uniaxial compressions at 1373 K (1100 °C) are evaluated and compared (Figure 4). The DRX grains have been partitioned out employing the grain orientation spread (GOS) approach. This parameter has been explained in detail elsewhere.[2,17] It was shown that GOS with a cut-off of 1 deg is a suitable criterion to partition DRX grains from the deformed matrix in hot deformed alloy D9.[2] It is observed from Figure 4 that the area fraction of DRX is consistently higher at all strains after uniaxial compression. At the maximum deformation temperature (1373 K [1100 °C]), and strain (0.8) examined in this study, the area fraction of DRX is ~90 pct during plane-strain compression, whereas a fully DRX structure evolves in similar deformation condition during uniaxial compression (Figure 4). These results indicate that DRX is hindered during plane-strain deformation vis-à-vis uniaxial deformation.

To the best of our knowledge, the influence of the state of stress on DRX kinetics has not been reported in the open literature. However, some studies were carried out to understand the effect of state of stress on the static recrystallization (SRX) kinetics. Kaiser and Taylor[18] did not observe any difference in the SRX kinetics of Fe ingot deformed by tension, torsion, and rolling. However, this observation seems to be inconclusive as there were significant variations in prior thermomechanical history, grain sizes, and compositions among the various samples used in their experiments. In contrast, it was observed by Barto and Ebert[19] that SRX was accelerated in Mo deformed in tension and wire drawing when compared with the rolled samples. This difference in SRX kinetics was attributed to the difference in dislocation densities produced by the various modes of deformation. The influence of the state of stress on DRX kinetics in the current study could be explained by the difference in stored energy accumulated during hot deformation in plane-strain and uniaxial deformations. It is apparent that the mode of deformation, which causes more dislocations to be generated, has a higher nucleation frequency during DRX. It seems that more dislocations are generated during uniaxial compression, which eventually results in higher DRX fraction. The influence of the state of stress on the grain boundary mobility is another important aspect that can affect/alter the kinetics of DRX. The mobility of the grain boundaries is influenced by their orientation, and how the orientation is influenced by the state of stress needs to be established. It may be argued here that a smaller initial grain size employed for uniaxial compression (~7 μm compared with ~10 μm for plane-strain compression, with both the grain size being measured considering twin as grain boundary) would have accelerated DRX. This view would stem from the fact that finer grain size material has more triple junctions, which are highly favorable sites for the nucleation of DRX grains.[20] However, the grain size has been observed to affect the DRX kinetics when there is a significant size difference (~4 to 5 times) in the initial microstructure.[21] In view of the marginal difference in initial grain size in the current study, its role on DRX kinetics is considered to be not significant.

To summarize, the current study reveals that nucleation mechanisms of DRX during plane-strain hot compression of alloy D9 are similar (*i.e.,* by bulging and annealing twinning) to that of uniaxial compression. This indicates that the state of stress (*i.e.,* uniaxial *vs* plane-strain compression) does not affect/alter the nucleation mechanisms of DRX. In contrast, the state of stress is found to influence the kinetics of DRX. The area fraction of DRX is found to be consistently lower in plane-strain compression compared with similar deformation condition in uniaxial compression. At the maximum deformation temperature (1373 K [1100 °C]), and strain (0.8) examined in this study, the area fraction of DRX during plane-strain compression is ~90 pct, whereas a fully DRX structure evolves in similar deformation condition during uniaxial compression. This confirms that DRX is hindered during plane-strain deformation vis-à-vis uniaxial deformation. This is attributed to the possible differences in the dislocation density evolution during uniaxial and plane-strain deformation and due to the variation in grain boundary mobility with the applied state-of-stress.

REFERENCES

1. P. Poelt, C. Sommitsch, S. Mitsche, and M. Walter: *Mater. Sci. Eng. A*, 2006, vol. 420, pp. 306–14.
2. S. Mandal, A.K. Bhaduri, and V. Subramanya Sarma: *Metall. Mater. Trans. A*, 2011, vol. 42A, pp. 1062–72.
3. M. Frommert and G. Gottstein: *Mater. Sci. Eng. A*, 2009, vol. 506, pp. 101–10.
4. M. Jafari and A. Najafizadeh: *Mater. Sci. Eng. A*, 2009, vol. 501, pp. 16–25.
5. D. Ponge and G. Gottstein: *Acta Mater.*, 1998, vol. 46, pp. 69–80.
6. Z. Yu and H. Choo: *Scripta Mater.*, 2011, vol. 64, pp. 434–37.
7. S. Mitsche, C. Sommitsch, D. Huber, M. Stockinger, and P. Poelt: *Mater. Sci. Eng. A*, 2011, vol. 528, pp. 3754–60.
8. I. Salvatori, T. Inoue, and K. Nagai: *ISIJ Int.*, 2002, vol. 42, pp. 744–50.
9. J.H. Kang and S. Torizuka: *Scripta Mater.*, 2007, vol. 57, pp. 1048–51.
10. D.G. Brandon: *Acta Metall.*, 1966, vol. 14, pp. 1479–84.
11. A.G. Beer and M.R. Barnett: *Metall. Mater. Trans. A*, 2007, vol. 38A, pp. 1856–67.
12. E. Brünger, X. Wang, and G. Gottstein: *Scripta Mater.*, 1998, vol. 38, pp. 1843–49.
13. X. Wang, E. Brunger, and G. Gottstein: *Scripta Mater.*, 2002, vol. 46, pp. 875–80.
14. H. Miura, T. Sakai, R. Mogawa, and G. Gottstein: *Scripta Mater.*, 2004, vol. 51, pp. 671–75.
15. S. Mahajan, C.S. Pande, M.A. Imam, and B.B. Rath: *Acta Mater.*, 1997, vol. 45, pp. 2633–38.
16. H. Gleiter: *Acta Metall.*, 1969, vol. 17, pp. 1421–28.
17. S. Mandal, S.K. Mishra, A. Kumar, I. Samajdar, P.V. Sivaprasad, T. Jayakumar, and B. Raj: *Phil. Mag.*, 2008, vol. 88, pp. 883–97.
18. H.F. Kaiser and H.F. Taylor: *Trans. ASM.*, 1939, vol. 27, p. 227.
19. R.L. Barto and L.J. Ebert: *Metall. Trans.*, 1971, vol. 2, pp. 1643–49.
20. S. Andiarwanto, H. Miura, and T. Sakai: *Mater. Sci. Forum*, 2002, vols. 408–412, pp. 761–66.
21. A. Dehghan-Manshadi and P.D. Hodgson: *Metall. Mater. Trans. A*, 2008, vol. 39A, pp. 2830–40.