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The role of deformation mechanisms in flow localization of 316L stainless steel

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Abstract

Type 316 SS is widely used as a structural material in a variety of current accelerator driven systems and designs as well as in a number of current and advanced fission and fusion reactor concepts. The material is found to be very sensitive to irradiation damage in the temperature range of 150–400 °C, where low levels of irradiation exposure, as little as 0.1 dpa, can substantially reduce the uniform elongation in tensile tests. This process, where the plastic flow becomes highly localized resulting in very low overall ductility, is referred to as flow localization. The process controlling this restriction of flow is related to the difference between the yield and ultimate strengths such that dramatic irradiation-induced increases in the yield strength results in very limited plastic flow until necking. In this study, the temperature dependence of this process is examined in light of the operating deformation mechanisms. It is found that twinning is an important deformation mechanism at lower temperatures but is not available in the temperature range of concern since the stress to activate twinning becomes excessively high. This limits the deformation and leads to the flow localization process.

1. Introduction

During the past several decades, there has been a long and continuing effort to understand the influence of irradiation exposure on reactor structure materials. In the spallation neutron source (SNS) and other advanced nuclear applications, type 316L (LN) and 304L (LN) stainless steel have been selected as components and structure materials, particularly because of their relatively high strength and good capacity to resist brittle fracture. However, under the condition of irradiation exposure, 316L SS will suffer a severe reduction of ductility. This loss of ductility typically is exhibited through following features: elevated yield strength accompanied by a very low uniform elongation under tensile loading conditions. The damage microstructure developed during irradiation restricts dislocation flow so that the plastic deformation is confined to very small volumes or regions of materials in the tensile load. Under this circumstance, premature necking at yield often followed by brittle fracture. This process is characterized as *flow localization*. Due to its importance in many applications, numerous studies have been performed to investigate the relationship among ductility loss, radiation-induced

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microstructure change and testing temperature. Previous studies [1–3] show that regardless of the irradiation level, the true stress for the onset of plastic instability (necking) is approximately constant, defined as the '*critical stress*'. In other words, the true stress to failure is an inherent material property rather than a function of irradiation-induced defect microstructure or strain hardening behavior. Limited studies of the tensile behavior of 300-series stainless steels following irradiation exposure have been carried out at intermediate temperatures in the past. A major conclusion by Pawel et al. [4] is that the 316L SS could suffer a minimum ductility in the range of 150–350 °C even at the irradiation levels as low as 1 dpa. This temperature range and irradiation conditions are typical of several current nuclear systems including spallation neutron source (SNS), current light water reactors (LWRs) and some Generation IV reactor concepts designs. Past efforts to understand flow localization concentrated primarily on dislocation pinning and channeling effects. Several models based on the 'barrier' mechanism have been proposed to explain the large increase in yield strength. However, current research shows that the flow localization process is more closely associated with the final stage of tensile deformation instead of the initial stage. This fact is indicated in Fig. 1, in which the true stress–strain curves of 316L SS irradiated to different dose level and tested at 50 °C are shifted to superpose at the point of plastic instability. It can be observed that the plastic flow exhibits a uniform behavior for various irradiation levels before the onset of plastic instability for all irradiation conditions. Therefore, the controlling microstructural mechanism for flow localization should be working for both unirradiated and irradiated materials and could include deformation-induced twinning and large scale planar slip in addition to the cleared channel flow due to the irradiation-induced defect microstructure. This insight provides an advantage in studying the final stages of flow localization, since the point of instability occurs at the same true stress level for both irradiated and unirradiated materials. Thus the controlling mechanism can be studied directly from unirradiated material.

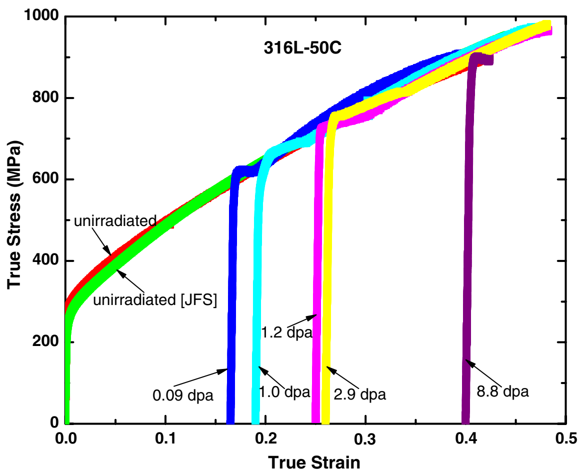

Fig. 1. True stress–strain curves for 316L SS tested at 50 °C.

2. Experiments

The material studied in this investigation was fully annealed 316L stainless steel with nominal composition given in Table 1. A series of tensile tests were performed on this material at temperatures ranging from RT to 400 °C. The tests were stopped at various strain levels to investigate the microstructural evolution during tensile loading. After testing, specimens were sliced from the center of specimen and polished. They were then examined using a JEOL 7000F SEM with electron backscattering diffraction (EBSD) capabilities. The results and details are discussed in the following sections.

3. Analysis

Recent work [1–3] shows the existence of a critical stress for the flow localization of face-centered cubic metals and alloys. This critical stress is a direct consequence of the post yield hardening properties of the materials, which seem to be independent of irradiation conditions. The observations show that,

Table 1
Composition for 316L stainless steel

Material	Composition (wt%)										
	C	Co	Cr	Cu	Mn	Mo	N	Ni	P	S	Si
316L SS	0.20	0.155	16.847	0.322	1.755	2.214	0.039	10.234	0.027	0.002	0.388

once flow begins, it follows the same strain hardening behavior, regardless of the position along the stress–strain curve. In addition, the point of departure from uniform elongation always occurs at the same true stress level, i.e. the critical stress. The critical stress value is highly dependent on the test temperature. A strong shift downward in the critical stress value can be noticed from about 900 MPa at 50 °C to 700 MPa at 164 °C. These results also offer insight into the severe problem with flow localization in 316L SS between 200 and 400 °C [4], where the critical stress levels seem to be the smallest. This loss of ductility was puzzling since the irradiation-induced hardening is nearly uniform across the temperature range from room temperature to at least 300 °C, and only dependent on doses. Test temperature has a weak impact on yield strength and no apparent evidence has been found that the elevation in yield strength controls the flow localization process. Recent work [5], which identifies the critical stress as the controlling mechanism, showed that the critical stress has strong temperature dependence: it drops rapidly between room temperature and 200 °C, and then reaches a plateau between 200 and 400 °C. Thus the strain to fracture is controlled by the difference between the yield strength and the critical stress [5]. The margin is large at room temperature even at moderately high irradiation doses, but the margin diminishes rapidly with increasing temperature. Between 200 and 400 °C, even small doses reduce the margin to nearly zero. Fig. 2 shows this temperature dependence of the critical stress. The average increase in yield strength is also presented in Fig. 2 and no plastic strain hardening prior to necking (flow localization) will occur after two curves intercept.

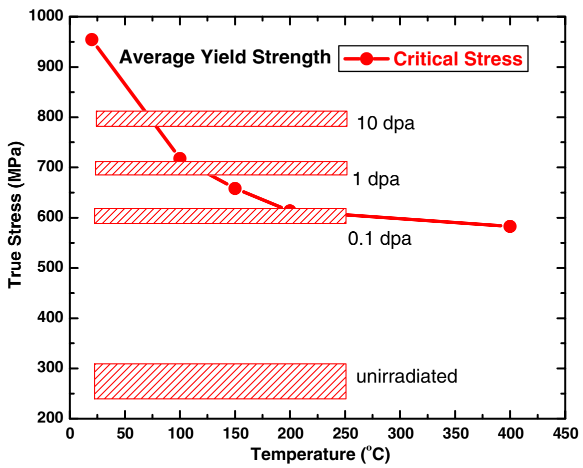

Fig. 2. The relationship between critical stress, average yield strength and test temperatures for type 316L SS; the yield strengths are idealized in a straight band to better demonstrate the difference between critical stress and yield strength.

Fig. 3 shows both the true stress (σ) vs. true strain (ε) and the strain hardening rates ($d\sigma/d\varepsilon$) vs. true strain curves for the specimens tested in the temperature range from room temperature to 400 °C. At the initiation of deformation, the strain hardening is very high and drops dramatically within the true strain range from 0 to 0.025. This rapid drop of strain hardening rates is explained by dislocation multiplication prior to large-scale plastic deformation. For the specimen tested at room temperature, the strain hardening rate reaches a relative plateau during the dislocation-based plastic strain hardening period. El-Danaf et al. [6] and Song and Gray [7] proposed that, the mechanism of mechanic twinning is activated at the onset of this stage, which prevents dislocation movement and promotes strain hardening. After the material is strained to the vicinity of plastic instability, the strain hardening rate drops quickly again. It eventually intercepts the true stress curve. This intersection point corresponds to the onset of necking; and the true stress at this point is the critical stress for plastic instability. The rapid decrease of strain hardening rate before the point of plastic instability can be explained by twin intersections as shown in Fig. 4. Compared with strain hardening curve at room temperature, the curves at 200 and 400 °C decrease much faster and no plateau can be found at the second stage until necking. The activation of twinning systems is believed to be associated with

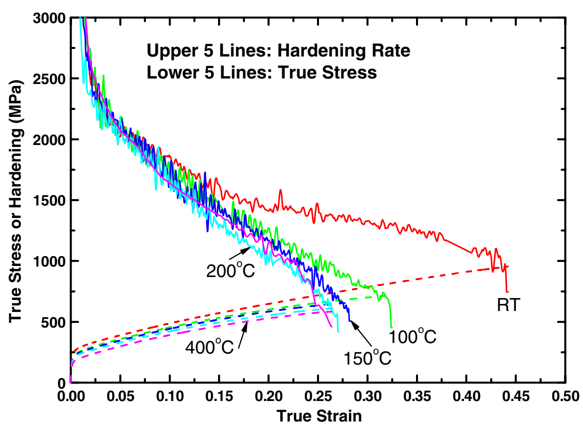

Fig. 3. Temperature dependence of the strain hardening rate (upper curves) and true stress–strain curve (lower curves) for annealed 316L SS.

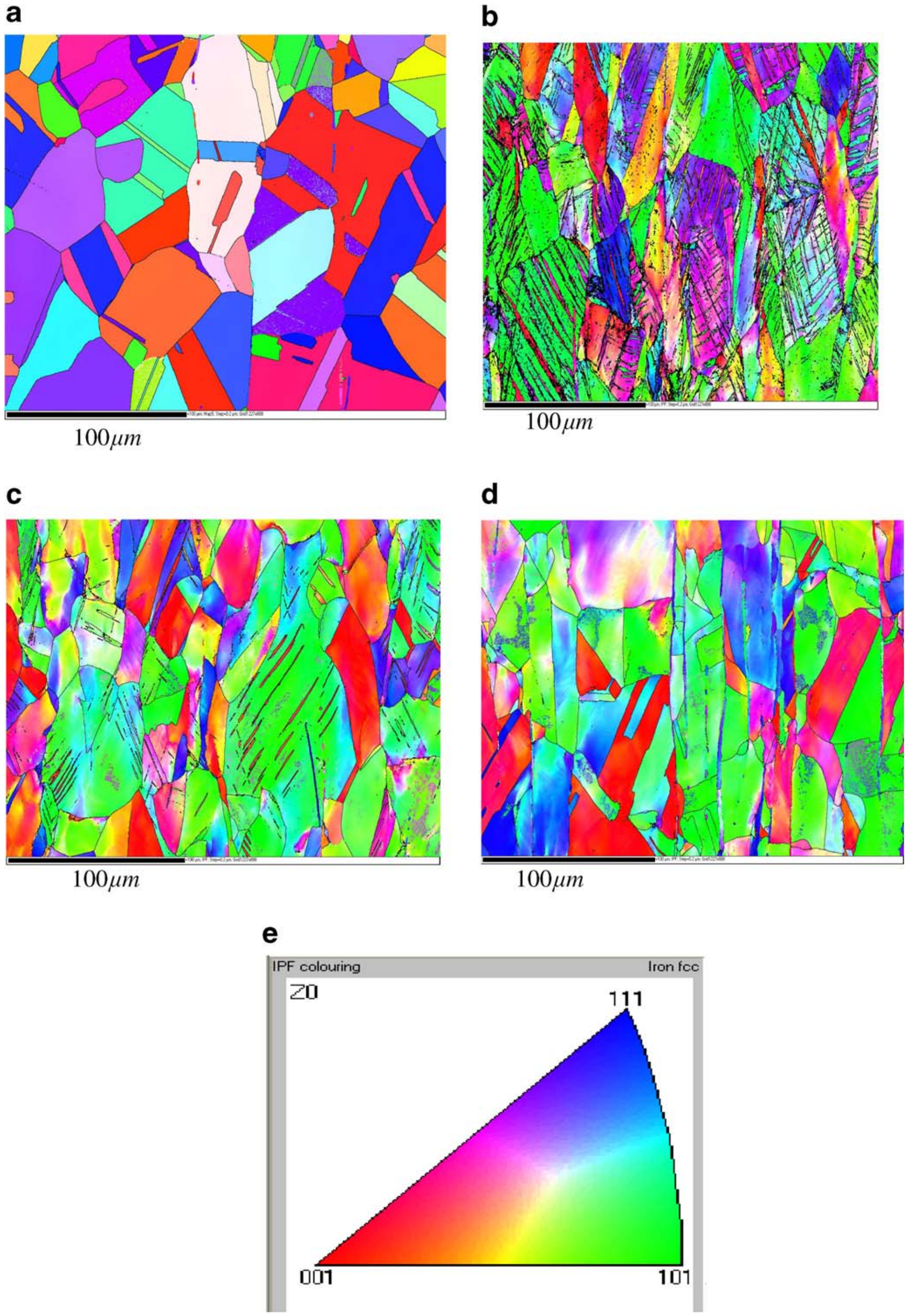

Fig. 4. SEM with EBSD pictures show the twinning structure and crystal orientation for (a) undeformed and deformed to the point of uniform elongation at (b) RT, (c) 100 °C and (d) 200 °C; (e) shows the legend for crystal orientation.

the width of Shockley partial which depends largely on the stacking fault energy (SFE). SFE is a positive function of temperature for many FCC materials. Above 200 °C, the SFE is too high to allow the twinning mechanism to activate. This effect is demonstrated in Fig. 4 and discussed later.

One aspect of the microstructural processes that controls the flow localization process can be identified by examining the relevant materials deformation modes as a function of temperature. Fig. 4 shows the deformation microstructure of 316L SS undeformed and deformed to the critical stress point (i.e. to approximately the UTS, just prior to the onset of necking), from room temperature to 200 °C. The undeformed specimen has equiaxed grain in all grain orientations. The grain structure following uniform plastic strain to the point of necking shows a grain structure that is elongated in the direction of the applied stress. The room temperature deformed specimen has extensive twinning whereas the 100 °C specimen shows much less twinning. At the higher test temperature of 200 °C, twinning is completely absent. It is noted that in the room temperature and 100 °C cases, twinning occurs in the grains with preferred crystal orientations. The specimen tested at room temperature to the point of plastic instability also shows extensive twin intersection which is believed to be the cause of rapid drop of strain hardening rate and the onset of flow localization.

4. Discussion

This study shows that the deformation mode of 316L SS demonstrates a strong temperature dependence of the critical stress, while the yield strength is relatively temperature insensitive. Fig. 2 indicates that the critical stress decreases quickly between RT and 200 °C and reaches a plateau thereafter. The average value for the yield strength irradiated at different irradiation levels are also plotted in this graph. The band between the yield strength and the critical stress indicates the amount of irradiation-induced hardening that the material can withstand before undergoing flow localization at yield. Once the radiation-induced hardening brings the yield strength to the level of the critical stress, the material will tolerate little or no plastic deformation prior to plastic instability, reducing the uniform elongation to critically low levels, below 1%. It is clear that much higher hardening levels can be tolerated at low temperatures than between 200 and 400 °C. Therefore, materials deformed at low temperature, even after high irradiation exposure levels still maintain a large difference between the critical stress and the average yield strength and are able to undergo large amounts of plastic flow and strain hardening after yield and prior to plastic instability. However, for the materials tested at intermediate temperatures from 200 to 400 °C, even at low doses, the difference between the irradiation-induced hardening and the critical stress is relatively small so that the materials suffer necking immediately after yielding. This observation is consistent with the previous investigation by Pan et al. [3]. It is also noteworthy that the difference between the yield strength and the critical stress is nearly uniform in the temperature range between 200 and 400 °C, consistent with the observed plateau in the tendency for flow localization in this temperature range.

The twinning process is believed to be a major contributing mechanism to strain hardening for FCC materials. The trigger for the activation of twinning system is determined by the critical twinning stress. In an earlier study, Byun [8] proposed a critical twinning stress expression in terms of the equivalent or uniaxial stress from the critical resolved shear stress. The expression is revised here based on the current EBSD results as

$$\sigma_{\mathrm{T}} = \frac{2\gamma_{\mathrm{SFE}}}{b_{\mathrm{p}}} \cdot \frac{1}{\mathrm{SF}} = 4.3\frac{\gamma_{\mathrm{SFE}}}{b_{\mathrm{p}}},$$

here the average Schmidt factor, SF, is about 0.465. The stacking fault energy, SFE, values are taken from literatures [9–11]. The value of b_{p} is equal to 0.145 nm [8]. The critical twinning stress is proportional to the stacking fault energy which has a strong compositional and temperature dependence in FCC materials including 316L SS. The critical

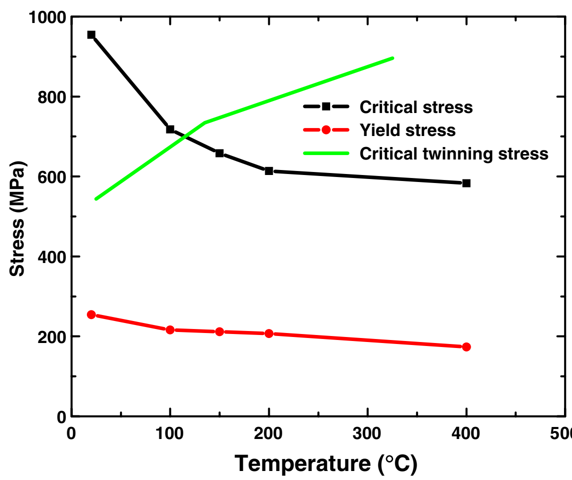

Fig. 5. Temperature dependence of critical stress, critical twinning stress and yield strength for annealed type 316L SS determined from experiment and the calculated critical stress for the initiation of twinning.

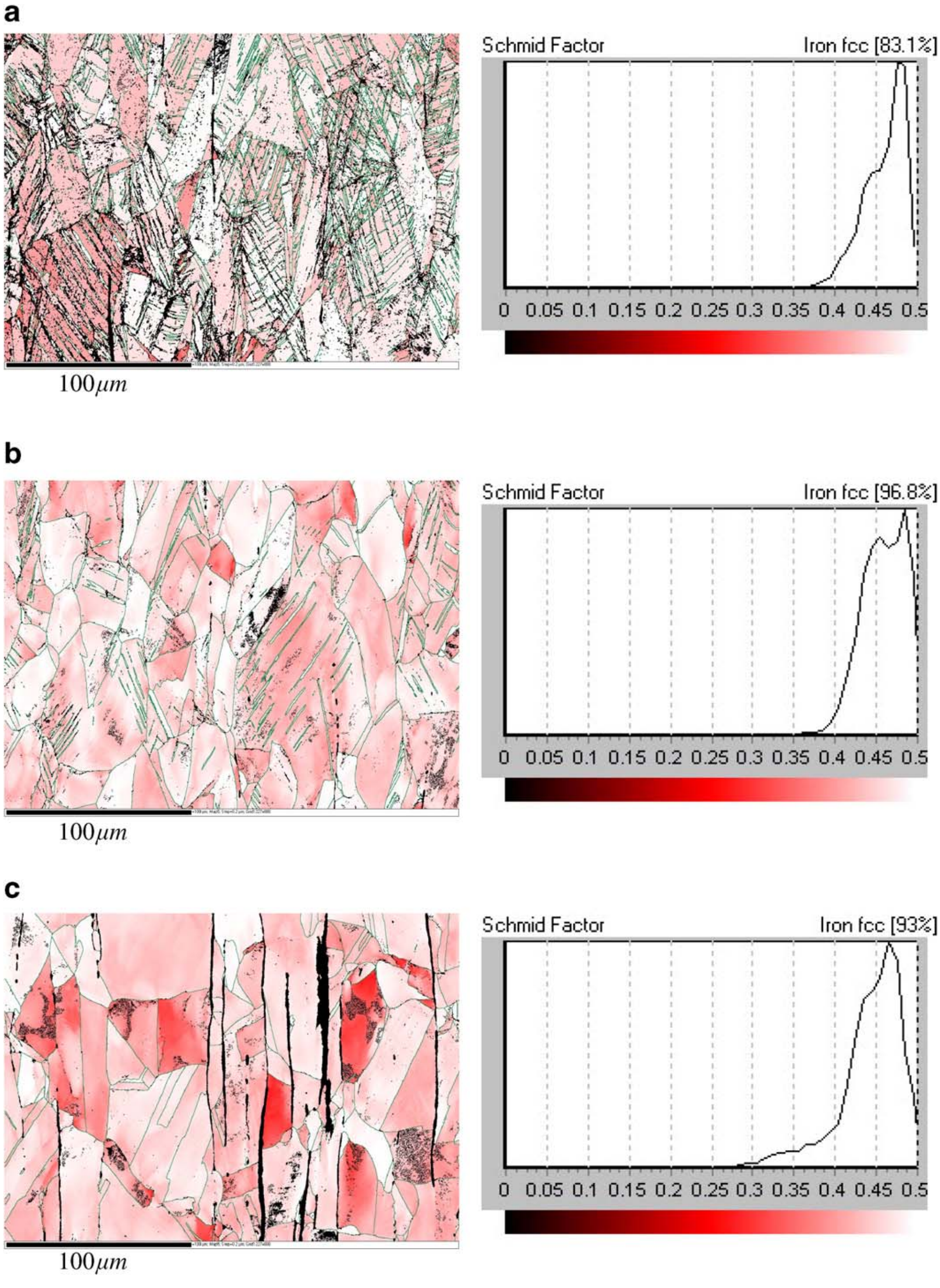

Fig. 6. Electron backscattered diffraction (EBSD) patterns from tensile specimens deformed to their UTS at (a) room temperature, (b) 100 °C and (c) 200 °C showing the major differences in the amount of twinning and the distribution of Schmidt factors.

twinning stress increases with increasing temperature and intercepts with the critical stress for necking at a temperature just above 100 °C as shown in Fig. 5. For the specimens tested below this temperature, the twinning process will be activated during the strain hardening process. It can be noted that the difference between the yield strength and critical twinning stress is smaller at low temperatures than that at higher temperatures, and consequently, the twinning process will be activated soon after yield at lower temperatures. However, for materials tested above the temperature at which the critical twinning stress is equal to critical stress, the twinning mechanism cannot be triggered during the hardening process because the true stress is smaller than the critical twinning stress. This observation is consistent with the SEM analysis shown in Fig. 4.

Schmidt factors were calculated based on the ESBD patterns and are included in Fig. 6 to indicate that twinning occurs only after some level of dislocation-based hardening at room temperature. At higher temperatures, standard dislocation-based hardening is present, but the stresses never elevate to the point where twinning can be triggered as an additional deformation mechanism. Twinning only occurs in the grains with the Schmidt factor less than about 0.485. The slip system and twinning process are competing mechanisms for plastic deformation. The grains with Schmidt factor above 0.485 easily reach the shear stress for slip before the critical twinning stress is attained. After the specimen is strained to some appropriate level where the critical twinning stress is reached, deformation in the grains with the most favorable slip orientations have been already dominated by dislocation-based hardening. Following this, twinning occurs elsewhere in the less well-oriented grains. Further indication of the initial dislocation-based hardening followed by twinning is observed in the room temperature hardening rate curves in Fig. 3. The steep initial slope in all cases is the initial dislocation-based strain hardening behavior. The pseudo-plateau for the room temperature curve is due to the onset and continued activation of twinning in that case. The inability of the material to reach the ever increasing critical stress for twinning as the temperature goes up is also reflected in the strain hardening curves where there is less indication of twinning even at 100 °C, and much steeper (downward) slopes of the strain hardening curves at all elevated temperatures. Thus the activation of the twinning process permits larger uniform elongation at room temperature. The inability of the twinning mechanism to activate at higher temperatures effectively limits the materials ability to deform, substantially reducing levels of uniform elongation. This explains the major tendency of 316SS to exhibit flow localization and very low uniform elongations following irradiation exposure between 200 and 400 °C.

5. Conclusions

Analysis of the tensile response and microstructure evolution of selected series of unirradiated and irradiated type 316L SS tested at different temperatures was carried out to establish the concept of critical stress for plastic instability and identify the controlling mechanism of flow localization. The following conclusions can be drawn:

1. Irradiated and unirradiated type 316L SS suffer flow localization, or extremely small values of uniform elongation, when the irradiation-induced hardening increases the yield strength to the level of the critical stress; the critical stress level is not a function of irradiation level.
2. The strain hardening process in 316L SS is controlled by dislocation-induced hardening mechanisms and, to some extent, by mechanic twinning; the extent of twinning is highly dependent on test temperature.
3. The ability of 316L SS to twin depends on a critical twinning stress for the activation of deformation which is directly proportional to stacking fault energy which also has a strong positive temperature dependence.
4. Schmidt factor plays an important role in the activation of twinning process and in the competing mechanism of twinning and slip. In cases here, some plastic, dislocation-based deformation occurs prior to the onset of twinning, also consistent with the need to reach a critical stress to activate twinning.
5. The strong temperature dependence of the stress to activate twinning in 316L SS means that it is not possible to induce twinning during plastic deformation at temperatures above about 200 °C. This limits plastic deformation to lower levels of uniform elongation of this material above that temperature.
6. For conditions where irradiation-induced hardening occurs, the limitation in ductility is due to the convergence of the elevated yield strength and the critical stress for plastic instability. This explains the minimum ductility in the temperature range of 200–400 °C for irradiated 316L SS.

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