

Solid solution-hardening by hydrogen in Fe-Cr-Ni-based austenitic steel
studied by strain rate sensitivity measurement:
contributions of effective stress and solute drag

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Abstract

Dissolution of atomic hydrogen (H) into Fe-Cr-Ni austenitic steels causes a significant increase in flow stress (*i.e.*, solid solution-hardening, SSH) during their plastic deformation. In this study, the characteristics and kinetics of H-induced SSH in AISI Type 310S steel with 7600 at. ppm H were studied through the measurements of strain rate sensitivity, S , by stress relaxation and strain rate jump experiments at 296 K. Two factors were found to contribute to the SSH: (i) the role of H as *thermally activatable obstacles* that increase the effective stress; (ii) the resistance acting on moving dislocations due to their dragging of H atmosphere (*solute drag*). These factors operated cooperatively or competitively in determining the S value and its dependencies on stress, strain, and strain rate. The peak SSH can be achieved where the sum of dislocation glide resistances from (i) and (ii) is maximized. The anticipated strain rate range for establishing such a situation coincided accurately with the author's previous experiments.

Keywords:

Austenitic steel; Hydrogen; Solid solution-hardening; Plasticity; Thermal activation

1. Introduction

The dissolution of hydrogen (H) atoms into Fe-Cr-Ni-based austenitic stainless steels and other face-centered cubic (FCC) metals/alloys triggers a significant strengthening effect *via* solid solution-hardening (SSH) [1–10]. The phenomenon manifests most straightforwardly as increases in tensile flow stress and indentation hardness, to an extent equivalent to other conventional interstitial atoms, including carbon and nitrogen [2,3]. Considering the general understanding that one of the root causes of SSH is lattice strain around solute atoms [11–16], the remarkable hardenability of H is somewhat abnormal

since the lattice dilatation around an H atom ($\approx 2 \times 10^{-30} \text{ m}^3$ [17,18]) is much smaller than those around carbon and nitrogen ($\approx 8.6 \times 10^{-30} \text{ m}^3$ [19]). This apparent deviation from the standard theory urges us to rationalize the H-induced SSH by envisaging specific interactions between H and dislocations [2,3].

An elaboration of the H-induced SSH in Fe-24Cr-19Ni-based steel (AISI Type310S) was recently performed by the authors through a series of tensile tests in the temperature range of 173–423 K [3]. In analogy with the general behavior of SSH, solute H increased the effective (*i.e.*, thermal) stress component of the total flow stress [5,20], which can be relaxed with the aid of temperature and time. However, even though such a standard temperature dependence was identified above room temperature, the SSH was maximized around 300 K in terms of yield stress, decaying as the temperature decreased to 173 K [3]. At $\approx 300 \text{ K}$, H atoms are highly diffusible and energetically favorable to segregate around moving dislocations [21–23]. Thus, in addition to H atoms dispersed in the FCC lattice, some dynamic interactions between H and dislocations might be the rationales behind such a maximized SSH. Some potential mechanisms that explain the phenomenology of H-induced SSH have been discussed by the authors' group [3,20].

The effective stress component of SSH stems from short-range (*i.e.*, the extent of a few interatomic distances) interactions between dislocation and solute atoms, where thermal vibration aids its progress [24]. Under a given temperature, a conventionally used physical parameter for measuring the significance of those short-range interactions and resultant effective stress is strain rate sensitivity, S [25–27]: increase or decrease in the flow stress after a tenfold increase in strain rate. The parameter S normally exhibits positive values and increases as the predominant rate-controlling obstacles for deformation are weaker and more thermally activatable, a typical example of which are solute atoms [28,29]. Indeed, strain rate sensitivity (or the amount of stress relaxation, which has the same physical sense with an increasing S , cf. eq. (2)) in Type310S steel becomes higher after the introduction of H [4,20,30], substantiating our statement on the role of solute H in increasing the effective stress. On the other hand, there may be another contribution to S and SSH from a longer-range remote interaction when solute atoms are diffusible enough and able to follow the movement of dislocations. That is the force exerted by dragging the solute atmosphere, called *solute drag* [23,31–33]. Although H-induced drag resistance has been invoked in H-charged FCC metals and alloys both theoretically and experimentally [1,3,7,23,34], its effectiveness and impact on the strain rate- and temperature-dependent plastic flow remain elusive. Since the resistance to dislocation motion by atmosphere dragging is also sensitive to the dislocation velocity and strain rate, the S value (*i.e.*, change in flow stress in response to the strain rate variation) is anticipated to work as an effective probe of the significance of solute drag.

The effective stress stemming from short-range interactions exhibits its logarithmic dependence on the dislocation velocity and strain rate [35]. For solute atoms, the effective stress can be approximated as background friction and almost invariable during a tensile test under a constant strain rate when the solutes are sufficiently dense compared with the interspacing between dislocations [29,36,37]. Meanwhile, the linear dependence of drag force on dislocation velocity [16,33,38] potentially makes it more rate-sensitive than short-range interactions. This possibly renders the flow stress a function of mobile dislocation density (*i.e.*, strain) even when the macroscopic strain rate is constant. As such, investigating the simultaneous dependencies of S on strain rate and strain should provide some key insights into the relative contributions of effective stress and solute drag on the total flow stress under a given condition. In this study, S values in Type310S steel with 7600 at. ppm H were measured by stress relaxation and strain rate jump tests throughout the tensile deformation at room temperature with various base strain rates (BSR) in order to figure out the components (*i.e.*, effective stress and solute drag) responsible for the H-induced SSH. Comparison of the experimental results with theoretical formulae demonstrated the competitive or cooperative contributions of H-induced effective stress and solute drag to the S value in plausible strain rate ranges and mobile dislocation density.

2. Methodology

2.1 Material, specimen, and test equipment

A commercial rod of Type310S steel with a diameter of 20 mm and a chemical composition of Fe-0.02C-0.37Si-1.10Mn-0.023P-0.001S-19.18Ni-24.18Cr (mass %) was prepared. The rod was supplied after solution annealing at 1353 K and subsequent water-quenching. Fig. 1 (a) shows the initial microstructure on the plane perpendicular to the rod axis, captured by electron backscattered diffraction (EBSD) in a scanning electron microscope (SEM). The heat treatment resulted in an average grain size with an equivalent diameter of 40 μm , excluding annealing twins. Because of its high Cr and Ni contents, the Type310S steel is immune to martensitic transformation, and the austenite phase remains totally stable during the deformation around room temperature [39]. Cylindrical tensile specimens with a diameter of 6 mm and a gage length of 30 mm (Fig. 1 (b)) were machined from the rod center, and the surfaces of their gage parts were mirror-polished *via* buffing by a cloth with 1 μm diamond suspension.

All the mechanical tests were performed by a screw-driven electromechanical test frame (Shimadzu AGX-plus) with a 100 kN load capacity. The temperature during the tests was kept at 296 ± 1 K. For the measurement of tensile strain, a strain-gage extensometer was clipped on the specimen's gage part. Prior to the experiments, H-charging was done by exposing the specimens to a gaseous hydrogen environment at 100

MPa and 543 K (the maximum pressure and temperature feasible in our equipment) for 200 h. This condition can realize a uniform H concentration in the specimen's gage part [4]. The H concentrations in the gage part were measured after mechanical tests by thermal desorption analysis (TDA) equipped with a thermal conductivity detector (STF-20A Series, J-Science Lab, Japan). In the H-charged specimen, the H concentration was 140 mass ppm (7600 at. ppm), while it was 5 mass ppm (300 at. ppm) in the non-charged one, as shown in the H desorption profiles in Fig. 1 (c).

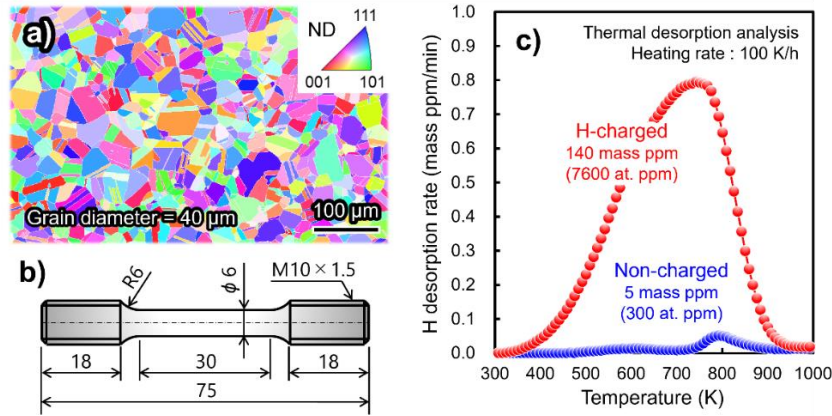


Fig. 1 (a) Microstructure of the Type310S steel on the plane normal to the rod axis, captured by EBSD. (b) Configuration and dimensions (mm) of the tensile specimen. (c) H desorption curves from the deformed gage parts of the specimens measured by TDA.

2.2 Transient mechanical tests

Two types of mechanical transient, *i.e.*, stress relaxation and strain rate jump, were employed in the course of deformation under the base crosshead speeds of 0.0003, 0.003, and 0.03 mm/s. These correspond to the initial tensile strain rates (base strain rates, BSR), $\dot{\epsilon}$, of 10^{-5} , 10^{-4} , and 10^{-3} /s, respectively. Note that the base crosshead speed and BSR here mean the standard tensile (not shear) deformation rate before and after the stress relaxation or strain rate jump was implemented. As references, monotonic tensile tests with these three BSRs were also carried out.

2.3 Theoretical background and the details of experiments

2.3.1 Stress relaxation test

The repeated stress relaxation procedure invented by Spätig et al. [40] was implemented at multiple stress/strain levels during the tensile deformation (Fig. 2 (a)). This methodology was established for a precise measurement of activation parameters controlling the plastic flow by taking the changes in the internal material state during the relaxation into account. Indeed, exhaustion of mobile dislocations and work-hardening during stress relaxation possibly affect experimental results obtained by a single

relaxation [41–43]. Although another rigorous way was recently proposed by Prasad et al. [44], we stayed with a more conventional method for the purpose of comparatively studying the differences between non-charged and H-charged samples. For each relaxation, crosshead holding for 30 s and subsequent stress reversion were repeated 5 times. A shorter holding time is recommended in [40] to minimize the internal state changes in the material with the progress of relaxation. Thus, we selected 30 s as the possibly shortest time, wherein the magnitude of stress drop was sufficient enough as compared with the fluctuation of load cell signal.

During the relaxation period, the elastic strain in the specimen's gage part is gradually relaxed through sluggish plasticity with a plastic strain rate of $\dot{\epsilon}_p = -\dot{\sigma}/M$, where $\dot{\sigma}$ is the rate of stress decay and M is the elastic modulus of the specimen-machine assembly. The test enables us to measure the activation volume, V , which is the area swept by a dislocation segment during overcoming an obstacle, multiplied by the Burgers vector, b (later explained in Section 4.1.2 and Fig. 9). V is recognized as a physical parameter for thermally activated deformation that characterizes the distribution and strength of the rate-controlling obstacles [45,46]. Given that the kinetics of plastic deformation obeys the Arrhenius type rate equation (later described in Section 4.1.2), V is inversely correlated with the strain rate sensitivity, S , as follows [24,47]:

$$S = \frac{\partial \tau}{\partial \ln \dot{\gamma}} = \frac{1}{M_T} \frac{\partial \sigma}{\partial \ln \dot{\epsilon}} = \frac{kT}{V} \quad (1)$$

where τ and $\dot{\gamma}$ are applied shear stress and shear strain rate, respectively. k is Boltzmann's constant, and T is temperature. τ and $\dot{\gamma}$ were calculated from the tensile stress, σ , and tensile strain rate, $\dot{\epsilon}$, assuming an average Taylor factor, M_T , of 3.06 in FCC polycrystals (*i.e.*, $\sigma = M_T \tau$ and $\dot{\gamma} = M_T \dot{\epsilon}$). Note that the present definition of strain rate sensitivity is slightly different from the generally defined one, in which a logarithm of stress is used instead in the numerator [44]. Given that the stress reversion process (Fig. 2 (a)) is nearly elastic, S can also be measured as a parameter purely reflecting the stress-dependence of dislocation velocity with a minimized effect of mobile dislocation exhaustion [24,40]:

$$S = \frac{1}{M_T} \frac{\Delta \sigma_1}{\ln(\dot{\epsilon}_{2b}/\dot{\epsilon}_{1e})} = \frac{\Delta \tau_1}{\ln(\dot{\gamma}_{2b}/\dot{\gamma}_{1e})} \quad (2)$$

where $\Delta \sigma_1$ ($= M_T \Delta \tau_1$) is the magnitude of the stress drop in the first relaxation. $\dot{\epsilon}_{1e}$ ($= \dot{\gamma}_{1e}/M_T$) and $\dot{\epsilon}_{2b}$ ($= \dot{\gamma}_{2b}/M_T$) are the strain rates at the end of the first relaxation and the beginning of the second relaxation, respectively (Fig. 2 (a)). Repeating this procedure over all five relaxations can derive an average and reliable value for S . For more details on the test method and the derivation of V and S , the readers should refer to [40]. It should be noted that some strain aging mechanisms related to carbon diffusion or carbon-vacancy interactions can affect the amount of stress transients as well [48,49]. Nonetheless,

Hannula et al. evaluated room temperature aging in Type316L steel containing 0.07% carbon, reporting that the aging-induced stress change is merely around 1 MPa under the aging time of an order of 10 s [49]. Considering the much lower carbon content in our present material (0.02%), the influence of strain aging should not be significant compared with the amount of total stress relaxation.

2.3.2 Strain rate jump test

The S can also be evaluated by a strain rate jump test (Fig. 2 (b)), where the flow stress increase, $\Delta\sigma$ ($= M_T\Delta\tau$), after a rapid escalation of strain rate from $\dot{\epsilon}_1$ ($= \dot{\gamma}_1/M_T$) to $\dot{\epsilon}_2$ ($= \dot{\gamma}_2/M_T$) is measured [24,47]. Eq. (2) is slightly modified for this test type:

$$S = \frac{1}{M_T} \frac{\Delta\sigma}{\ln(\dot{\epsilon}_2/\dot{\epsilon}_1)} = \frac{\Delta\tau}{\ln(\dot{\gamma}_2/\dot{\gamma}_1)} \quad (3)$$

In this study, a tenfold increase from each BSR was implemented at various stress and strain levels. The determination of S is straightforward when the stress-strain curve after the strain rate jump is smooth. However, a small yield point (Fig. 2 (b)) was sometimes observed, making the definition of S not explicit. In such a case, a tangent line having the same slope as the stress-strain curve was drawn through the yield point (Fig. 2 (b)). Since the temporal stress drop after the yield point is often attributed to dislocation multiplication [24,50], this latter definition of S may be more appropriate to measure the change in dislocations glide resistance by excluding the multiplication effect.

One has to be aware that the present S is all calculated based on the shear stress-strain, which was converted from the measured tensile stress-strain through M_T . Since the numerator of S (eq. (1)-(3)) is defined linearly rather than logarithmically, the obtained results should be different by a factor of M_T if tensile stress-strain is used as has been done in some previous works [28,29].

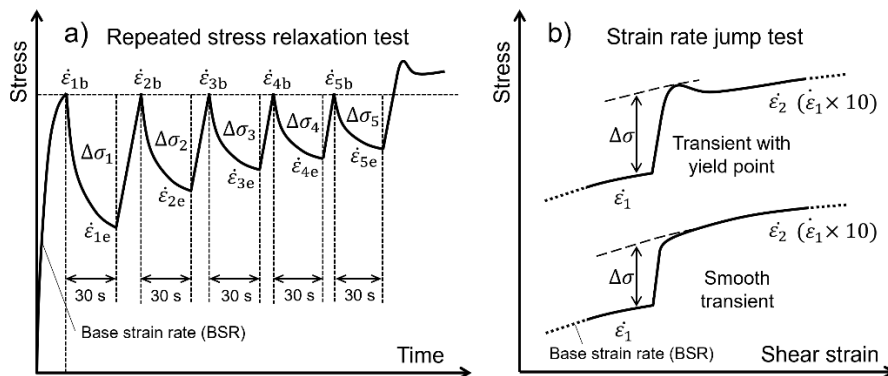


Fig. 2 Schematic drawings of (a) repeated stress relaxation and (b) strain rate jump tests. Refer to Sections 2.2.1 and 2.2.2 for the corresponding descriptions of methodologies.

3. Results

3.1 Rate-dependence of stress-strain behavior

The stress-strain curves in the monotonic tensile tests at different strain rates (10^{-5} , 10^{-4} , and 10^{-3} /s) are shown in Fig. 3. In what follows, true stress and strain are used for all descriptions rather than nominal values (*i.e.*, the calculation of shear stress through M_T (Section 2.3.1) is also based on σ_t). In non-charged specimens, flow stress was monotonically increased with the tenfold increase in the strain rate. The flow stress in H-charged specimens consistently exhibited higher values than that in non-charged specimens under all the strain rates due to the H-induced SSH effect [2–4]. A positive correlation between flow stress and strain rate was also observed even in H-charged specimens, although the flow stress gap between 10^{-4} /s and 10^{-3} /s was somewhat smaller than that between 10^{-5} /s and 10^{-4} /s. In the inset of Fig. 3, the yield (0.2% proof) stresses, which were measured by an average of 3 specimens for each, are plotted as a function of strain rate. The gap of yield stress between non-charged and H-charged specimens at 10^{-5} /s and 10^{-4} /s was somewhat greater than that at 10^{-3} /s by an extent of almost 10 MPa.

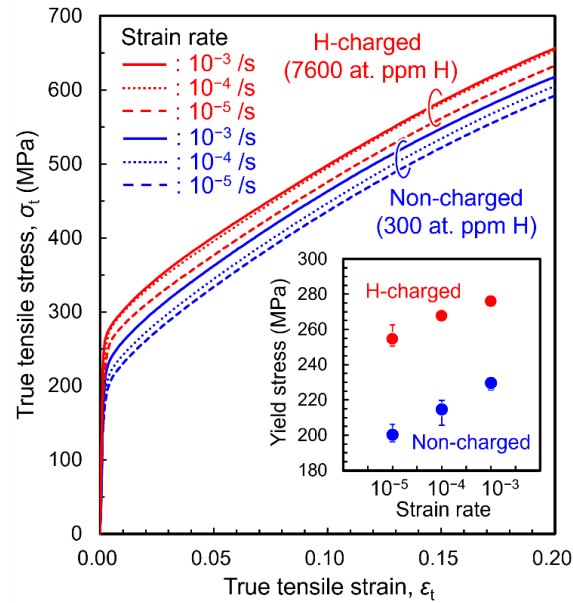


Fig. 3 True stress-true strain curves of non-charged (blue) and H-charged (red) specimens in the monotonic tensile tests at room temperature with three different strain rates. The inset magnifies around the yield point.

Fig. 4 compares the stress-strain behaviors during stress relaxation and strain rate jump tests under three different BSRs with those in monotonic tensile tests. In stress relaxation tests, a gradual drop of flow stress after crosshead holding was observed in all the experimental conditions (Fig. 4 (a)~(c)). With an increase in BSR, the magnitude of the stress drop was amplified monotonically in both non-charged and H-charged

specimens. Meanwhile, a prompt increase in the flow stress was evident after the rapid strain rate jump (Fig. 4 (d)~(f)).

Disregarding some exceptions (see Section 3.2.2), the extent of these stress transients in H-charged specimens was more remarkable than those in non-charged specimens when the comparison was done under the same BSR (cf. magnifications in the insets of Fig. 4). This result indicates an increase in S due to solute H, reproducing the previous results reported by the authors and other researchers [20,30]. Note that stress-strain curves in stress relaxation and strain rate jump tests coincided well with the monotonic deformation curves, except for the relaxation and strain rate jump phases. Thus, it can be deduced that the overall flow behavior is not affected by the mediation of these procedures. For the information about stress and strain values at which stress relaxation and strain rate jump were implemented, as well as the magnitude of stress relaxation in all the examined stress/strain levels, the readers shall refer to Table A1 and Fig. A1 in Supplementary Material.

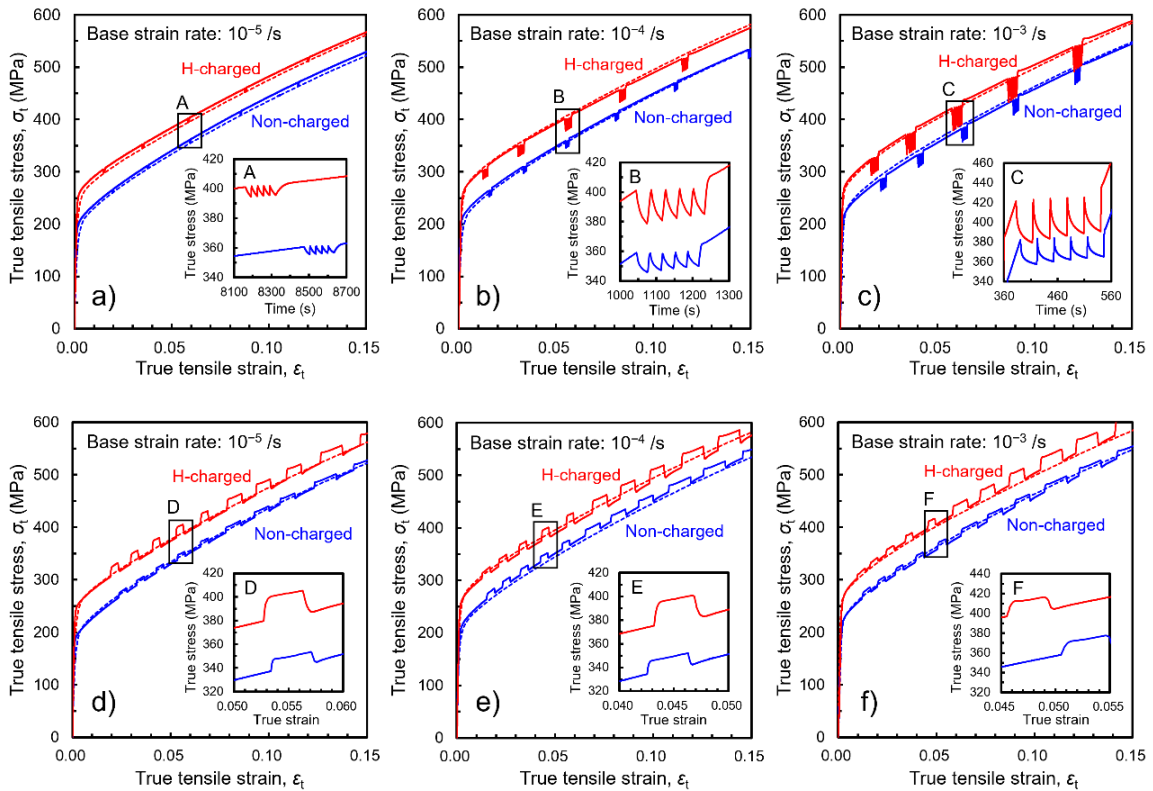


Fig. 4 True stress-true strain curves in monotonic tensile (dashed lines) and transient mechanical (solid lines) tests with different base strain rates (BSR). The results of stress relaxation tests are shown in (a)~(c), while those of strain rate jump tests are displayed in (d)~(f). Magnifications of the rectangle A~F are shown in the insets.

3.2 Stress-, strain-, and strain rate-dependences of S

3.2.1 Stress relaxation test

The S values measured by stress relaxation tests under three different BSRs are summarized in Fig. 5 as the functions of shear flow stress, τ - τ_y (τ_y was approximated as $\sigma_{0.2}/M_T$, where $\sigma_{0.2}$ is 0.2% proof stress), and tensile strain, wherein the distinctions between non-charged and H-charged specimens are evident. Note that each curve in Fig. 5 was constructed from the results of two individual specimens to check the reproducibility.

In non-charged specimens, the S values increased gradually and linearly with the increases in applied stress and strain. The magnitude of S was barely dependent on the BSR, converging all three curves on each other. On the other hand, the S was increased approximately twofold by solute H. Almost horizontal (*i.e.*, stress- and strain-insensitive) plots were obtained in H-charged specimens, although they accompanied slight concave downs in the small stress/strain domain at BSR = $10^{-3}/s$ and $10^{-4}/s$. Moreover, they exhibited slight and non-monotonic dependence on BSR. While the S once increased with an increase in BSR from $10^{-5}/s$ to $10^{-4}/s$, it decreased again when BSR was further raised to $10^{-3}/s$: the result at BSR = $10^{-3}/s$ lay in-between those at 10^{-5} and $10^{-4}/s$.

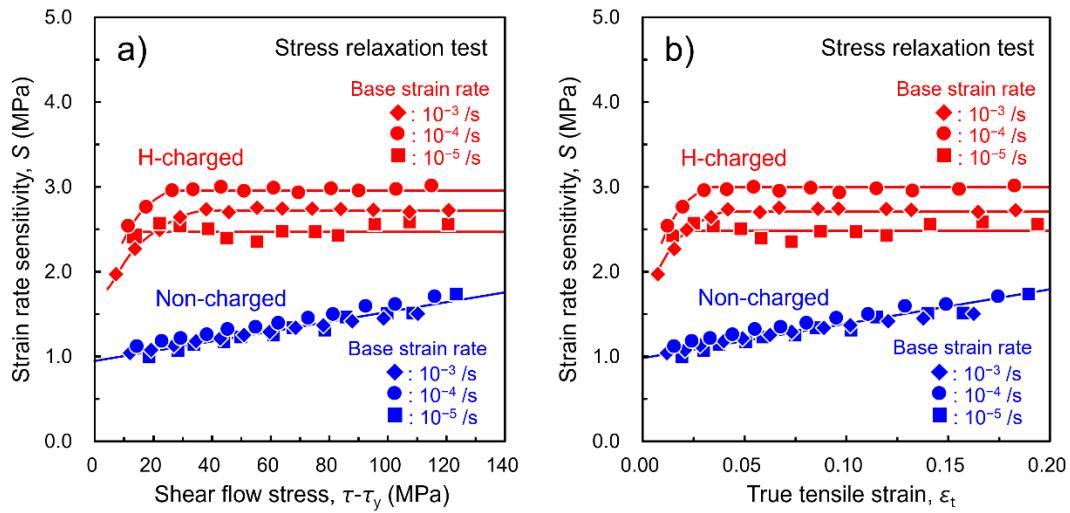


Fig. 5 Strain rate sensitivity, S , in non-charged (blue) and H-charged (red) specimens measured by stress relaxation tests with different base strain rates (BSR). The results are plotted as a function of (a) flow stress and (b) true tensile strain.

3.2.2 Strain rate jump test

The shapes and BSR-dependencies of the S -stress/strain plots drastically changed when measured by strain rate jump tests. As shown in Fig. 6, the S in non-charged specimens also became slightly BSR-dependent (*i.e.*, it monotonically increased with an increase in BSR), although the plots maintained their linearity. The S values closest to

those in stress relaxation tests (Fig. 5) were obtained in the strain rate jump test at BSR = 10^{-5} /s.

The S in the H-charged specimen exhibited a behavior almost identical to stress relaxation tests at BSR = 10^{-5} /s. However, unique S -stress/strain plots manifested when BSR was increased to 10^{-4} and 10^{-3} /s. That is, the initial concave part at low stress/strain was more pronounced as the BSR increased, decreasing the S to the level equivalent to that of the non-charged case. Notably, the initial S at low stress/strain was even smaller than those in non-charged specimens at BSR = 10^{-3} /s, the exceptional case described in Section 3.1. The S in H-charged specimens then gradually escalated with increases in stress and strain. Eventually, it far exceeded the S in non-charged specimens, and its magnitude finally followed the rank order of BSR at high stress ($\tau - \tau_y > 100$ MPa) and large strain ($\epsilon_t > 0.15$). One has to append that, at BSR = 10^{-4} and 10^{-3} /s, the S at this large stress/strain regime (3.5~4.0) was greater than the values measured in stress relaxation tests (2.5~3.0): compare Fig. 6 with Fig. 5.

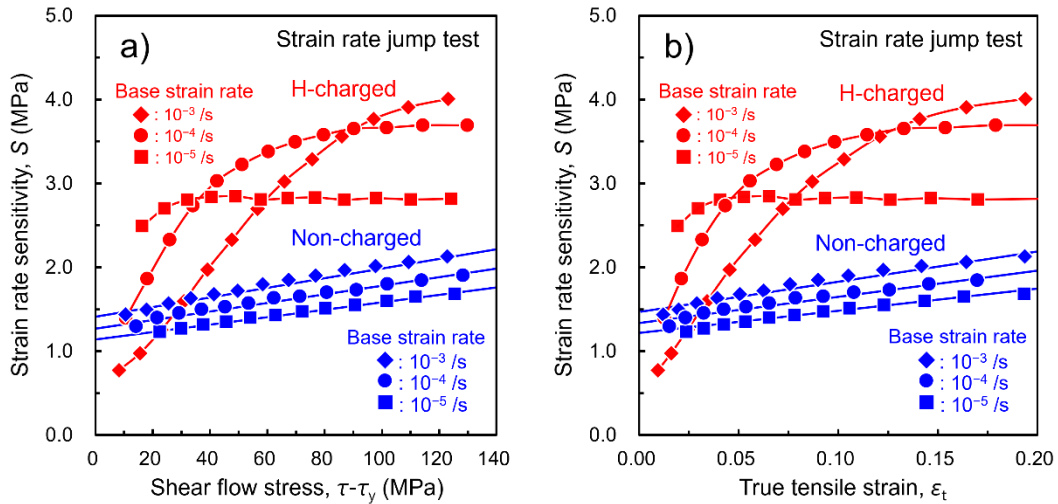


Fig. 6 Strain rate sensitivity, S , in non-charged (blue) and H-charged (red) specimens measured by strain rate jump tests with different base strain rates (BSR). The results are plotted as a function of (a) flow stress and (b) true tensile strain.

3.3 Transient behavior after strain rate jump

Another impact of solute H and BSR was discovered on the stress transient behavior after the strain rate jump. In Fig. 7, magnifications of the stress-strain curves around the points of strain rate jump at $\epsilon_t \approx 0.03$ and 0.11 are depicted for BSR = 10^{-5} and 10^{-3} /s. At a small strain regime (≈ 0.03) with a lower BSR (10^{-5} /s), the stress-strain curves after the strain rate jump were relatively smooth in both non-charged and H-charged specimens, having the slopes almost the same as those before the jump (Fig. 7 (a)). Meanwhile, a slight yield-drop appeared after the strain rate jump in the H-charged specimen at BSR =

10⁻³/s (cf. red arrow in Fig. 7 (b)). Interestingly, such a yield-drop was indistinct in the non-charged specimen (Fig. 7 (a) and (b)), as well as for the H-charged specimen at BSR = 10⁻⁵/s (Fig. 7 (a)), under this small strain domain. With an increase in ϵ_t above 0.1, the yield-drop after the strain rate jump began to manifest in all the testing conditions (Fig. 7 (c) and (d)), possibly ascribable to some multiplication event of dislocations [24,50]. Although the results are omitted from Fig. 7, the tendency of stress-strain behavior under BSR = 10⁻⁴/s was almost similar to the case of BSR = 10⁻⁵/s. The appearance of yield-drop after the mechanical transient has often been attributed to some strain aging phenomena [48,49]. However, its presence under the faster BSR and its absence under slower BSR contradict any basic theories of strain aging. Note that a similar yield-drop phenomenon was identified at the reloading process after stress relaxation as the applied strain became relatively large (Fig. 4 (a)-(c)). This may also be due to the dislocation multiplication event, trying to continue the deformation by compensating for the density of mobile dislocations that were exhausted during the relaxation process [51].

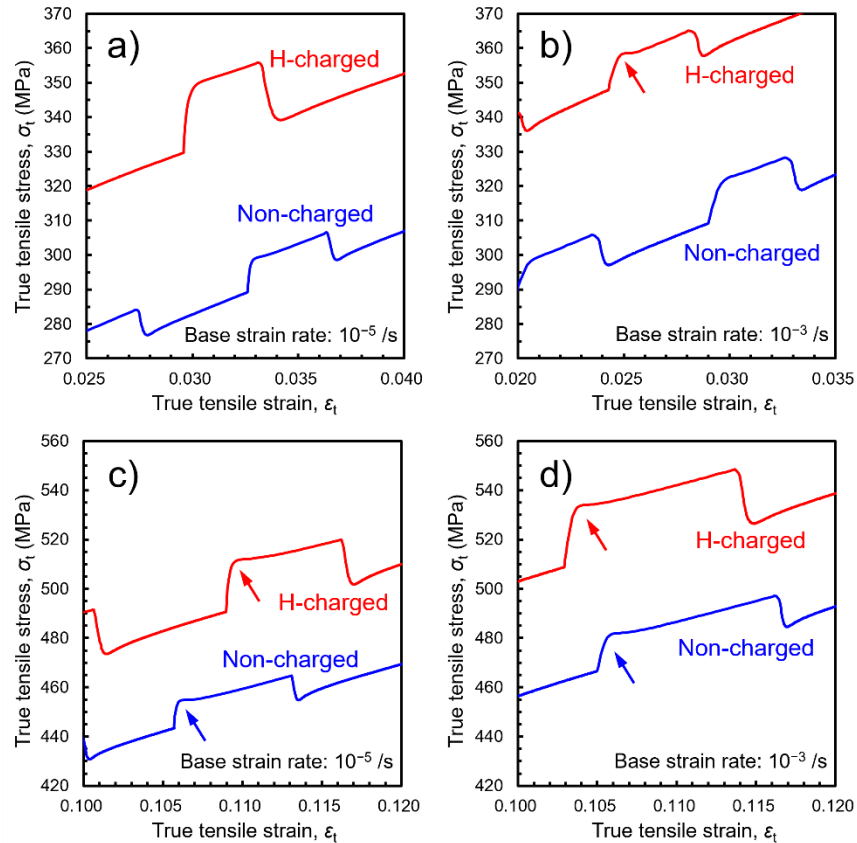


Fig. 7 True stress-true strain curves during the strain rate jump tests with the BSR of (a)(c) 10⁻⁵/s and (b)(d) 10⁻³/s. (a) and (b) correspond to the small strain regime below 0.05, while (c) and (d) show a larger strain beyond 0.1. The arrows denote the minute yield-drop manifesting after the strain rate jump.

4. Discussion

4.1 Strain rate sensitivity in non-charged specimens

4.1.1 Linearity of the SRS vs. flow stress plot: *Cottrell-Stokes law*

When plotted versus flow stress, the S values in non-charged specimens exhibited almost linear behavior accompanying a gradual increase in S with a positive slope. Such linearity between S and flow stress has been reported in a variety of FCC metals and alloys, a feature of the so-called *Cottrell-Stokes law* [52–54]: effective stress is proportional to the total flow stress.

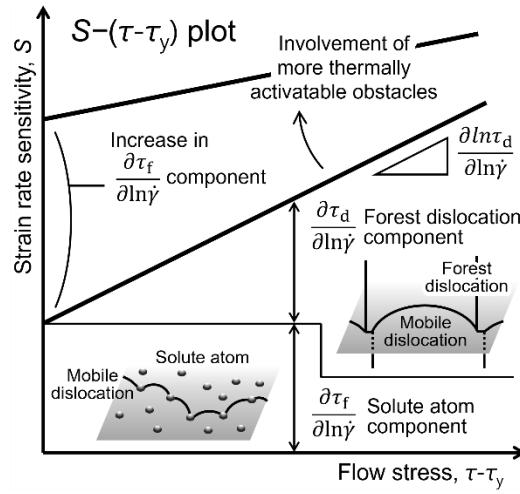


Fig. 8 Schematic drawing of the strain rate sensitivity (S) vs. flow stress ($\tau - \tau_y$) plot on the basis of eq. (4). The linearity of the plot appears when it obeys Cottrell-Stokes law.

The S value in engineering alloys is determined as a sum of the contributions from multiple thermal obstacles [26,29,36,37]. The primary obstacles in the present material are substitutional solutes (*i.e.*, Cr and Ni) and forest dislocations introduced by plastic deformation. Given that their contributions are additive, an accepted case for a combination of solutes and forest dislocations [36,37,46], eq. (1) can be decomposed into two separate terms.

$$S = \frac{\partial \tau}{\partial \ln \dot{\gamma}} = \frac{\partial \tau_y}{\partial \ln \dot{\gamma}} + \frac{\partial \tau_d}{\partial \ln \dot{\gamma}} = \frac{\partial \tau_y}{\partial \ln \dot{\gamma}} + \frac{\partial \ln \tau_d}{\partial \ln \dot{\gamma}} (\tau - \tau_y) \quad (4)$$

where τ_d is the amount of strain-hardening, equals to $\tau - \tau_y$. A major part of S at the yield point, τ_y , stems from the contribution of solutes. Thus, the first term in eq. (4) (*i.e.*, the intercept of $S-(\tau - \tau_y)$ plot) reflects the S component of solutes only, while the second term expresses the forest dislocation component. The linearity of $S-(\tau - \tau_y)$ plot indicates a constant coefficient of the second term, a parameter reflecting the strength of the intersection between mobile and forest dislocations [29,36]. In Fig. 8, the general

behavior of $S-(\tau-\tau_y)$ plot is schematically drawn, together with how it is affected by the involvement of a new type of thermally activatable obstacles (discussed in [Section 4.2.1](#)). One can also recognize how this general behavior breaks down in the case of H-charged specimens from the comparison of [Fig. 8](#) with [Fig. 5](#) and [Fig. 6](#). This is due to the contribution of solute drag as will be discussed later in [Section 4.2.2](#).

The activation parameters measured by mechanical transient tests can reflect not only the mobility of individual dislocations but also their annihilation process: dynamic recovery. Nonetheless, the typical consequence when dynamic recovery is interposed is the deviation of $S-(\tau-\tau_y)$ plot from linearity and its shape change into a concave upward [29,54]. Such behavior was not observed in non-charged specimens ([Fig. 5](#) and [Fig. 6](#)), indicating that the influence of dynamic recovery was minor in the presently examined stress/strain range. Also, we have preliminarily confirmed in Fe-Cr-Ni alloy system that the impact of solute H on the propensity for dynamic recovery is insignificant [55].

4.1.2 BSR vs. S in strain rate jump test

Although $S-(\tau-\tau_y)$ plots in non-charged specimens satisfied linearity in both stress relaxation and strain rate jump tests, a difference was identified as their BSR-dependence. BSR barely affected S in the former tests, whereas an increase in BSR caused an upward shift of $S-(\tau-\tau_y)$ plot in the latter tests. Since the shift occurred parallelly with an almost constant slope (*i.e.*, a constant second term in eq. (4)), it is due to changes in the interaction of mobile dislocations with solutes rather than their interactions with forests.

The dislocation configuration and its force-distance profile during a thermal activation event for overcoming a short-range obstacle is usually drawn in the form of [Fig. 9](#) [46]. Several shapes of obstacle profiles have been established, amongst which parabolic, triangular, or Cottrell-Bilby potentials with finite slopes are most widely employed for the case of solute-dislocation interactions [13,24,56]. In [Fig. 9](#) (c) and (d), ΔG (shaded with black) expresses the free energy change during the thermally activated overcome, *i.e.*, energy supplied by thermal fluctuation. This enters into the numerator of the exponent in the Arrhenius type rate equation for plasticity [24,50].

$$\dot{\gamma} = \dot{\gamma}_0 \exp\left(-\frac{\Delta G}{kT}\right) \quad (5)$$

where $\dot{\gamma}_0$ is a pre-exponential factor involving the density and velocity of mobile dislocations. The area below ΔG (shaded with gray) in [Fig. 9](#) (c) and (d) is the externally applied mechanical work, which is the product of applied stress and activation volume, τV . V is a purely geometric parameter multiplying the effective obstacle width, d , their mean interspacing, L , and the Burgers vector, b : $V = bdL$ ([Fig. 9](#) (a) and (b)). It is also defined by the stress derivative of ΔG [24] as

$$V = - \left. \frac{\partial \Delta G}{\partial \tau} \right|_T \quad (6)$$

By combining eq. (6) with eq. (5), one can obtain eq. (1). From eq. (5), it can be noticed that ΔG is a linearly decreasing function with a logarithm of strain rate:

$$\Delta G = kT \ln \left(\dot{\gamma}_0 / \dot{\gamma} \right) = kT (\ln \dot{\gamma}_0 - \ln \dot{\gamma}) \quad (7)$$

ΔG probabilistically decreases as the strain rate increases, making the d value statistically narrower. In other words, a higher stress is required for dislocations to overcome these obstacles due to a lesser aid from thermal fluctuation. Given that the L is constant during the deformation (e.g., for solutes), a decrease in d leads to a decrease in V . This, in turn, causes an increase in S (cf. eq. (1)), a plausible reason for a larger S under a higher BSR in non-charged specimens (Fig. 6). According to eq. (7), an increase in strain rate has the same physical sense as a decrease in deformation temperature. In fact, decreasing V and increasing S with a lowering temperature is a common trend [27,57].

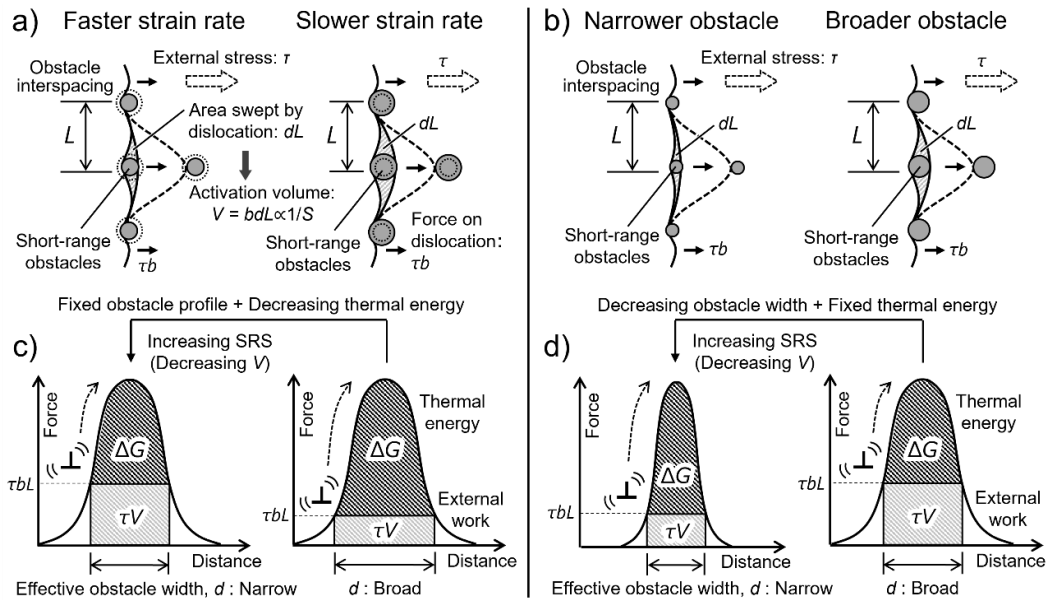


Fig. 9 Illustration of the movement of a dislocation segment *via* overcoming short-range obstacles. (a) and (c) are the top view of the slip plane, while (b) and (d) are its side view. (a) and (b) denotes the effect of strain rate on the free energy and activation volume, whereas the influence of obstacle extent on these parameters is shown in (c) and (d).

4.1.3 BSR vs. S in stress relaxation test

To understand the BSR-insensitivity of S in the stress relaxation test, not only BSR but also the *strain rate during stress relaxation* should be considered. This is because the S in the stress relaxation test is derived from the difference in plastic strain rates between the end of one relaxation and the beginning of the next relaxation (Section 2.2.1). In Fig.

(a), the plastic strain rates, $\dot{\epsilon}_p$, in non-charged specimens during 1st~5th relaxations at $\epsilon_t \approx 0.1$ are shown for three different BSRs. Although the experimentally obtained stress-time data seemed smooth in Fig. 4 (a)-(c), they necessarily involve some noises due to tiny fluctuation of the load cell signal. Thus, its time derivative also involves some scatters especially when $\dot{\epsilon}_p$ was below $10^{-5}/s$.

Due to a necessity for deformation continuity, $\dot{\epsilon}_p$ at the beginning of all the 1st relaxations coincided well with their corresponding BSR. Then, $\dot{\epsilon}_p$ gradually decayed as the relaxation progressed, which is typical behavior of the logarithmic deformation transient [24], eventually settling at $4 \times 10^{-5} \sim 7 \times 10^{-5}/s$. After the reversion of stress to the original level, the 2nd relaxation started with $\dot{\epsilon}_p$ slightly smaller than that at the beginning of 1st relaxation. These $\dot{\epsilon}_p$ ranges were scarcely affected by ϵ_t . The variation of $\dot{\epsilon}_p$ during the relaxation was greatest at BSR = $10^{-3}/s$, while it was smallest at BSR = $10^{-5}/s$. Because of the limited crosshead-holding time of 30 s, the $\dot{\epsilon}_p$ at the end of each relaxation exhibited slightly higher values at BSR = $10^{-3}/s$ and $10^{-4}/s$ than at BSR = $10^{-5}/s$. With the prolongment of the holding time, these strain rates would converge with each other and eventually settle into zero upon the effective stress component is fully relaxed.

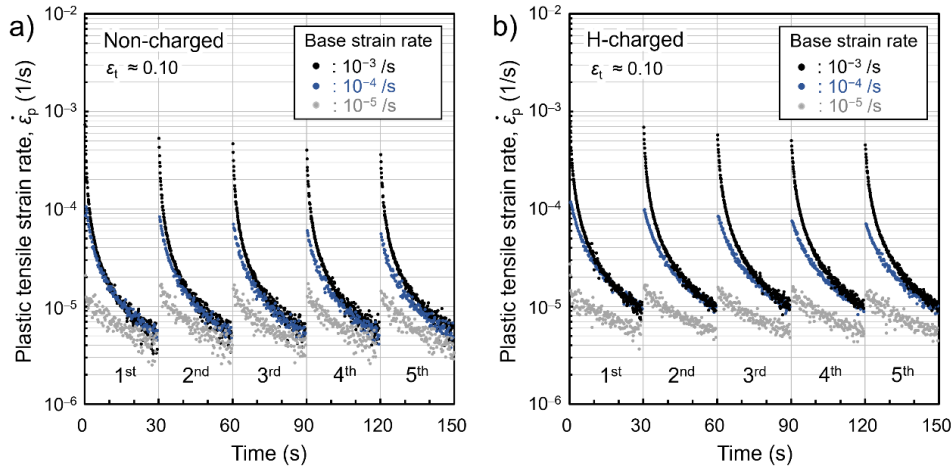


Fig. 10 Plastic strain rate, $\dot{\epsilon}_p$, during the stress relaxation series in (a) non-charged and (b) H-charged specimens at a true tensile strain, ϵ_t , of around 0.10. These plastic strain rate ranges during the relaxations were almost similar for each BSR condition, even when the strain level was larger and smaller.

An important indication from Fig. 10 is that, in the stress relaxation test, the denominator of eq. (2) or eq. (3) depends on BSR (*i.e.*, it should increase with an increase in BSR), while it does not in strain rate jump test as long as the $\dot{\gamma}_2/\dot{\gamma}_1$ ratio is constant. Therefore, an increase in S with BSR, which is due to the reason explained in Section 4.1.2, can be compensated by this increase in the denominator term. In other words, the S value inherently includes the influences of continuously decreasing $\dot{\epsilon}_p$ (*i.e.*, sluggish

decrease in dislocation velocity during the relaxation), an inevitable consequence owing to the nature of this experimental methodology. From a physical perspective, the strain rate jump test may reflect an instantaneous change in the glide resistance to mobile dislocations more adequately, whereas some transient effects in a finite timeframe are added to the result of stress relaxation tests. A similar viewpoint of the strain rate during relaxation is also vital to interpreting the complex S -BSR correlations in H-charged specimens and their dependences on the test types (Fig. 5 and Fig. 6).

4.2 Strain rate sensitivity in H-charged specimens

4.2.1 Solute H as *thermal obstacles*

In the stress relaxation tests of H-charged specimens, Fig. 5 (a) shows that the back extrapolations of the horizontal parts of $S-(\tau-\tau_y)$ plots yield positive intercepts, the values larger than those in non-charged specimens. According to eq. (4) and Fig. 8, a larger intercept implicitly indicates a greater $\partial\tau_y/\partial\ln\dot{\gamma}$ term and an additional contribution of another type of solute atoms to S . In the present case, it is precisely H. They act as additional short-range obstacles, amplifying the effective stress and causing SSH [3,20].

The role of solute H as a short-range obstacle and its interaction form with dislocations have been discussed in the authors' companion papers [3,20], thereby not described here in depth. In short, H atoms segregating into dislocation core, as well as those dispersed in austenite lattice, act as weak and more thermally activatable obstacles than other intrinsic obstacles contained in the material. The bowing-out of short dislocation segments from the row of segregated solute H, accompanying an atomic jump of those H toward the direction of dislocation movement, maybe a particular rate-controlling process of deformation. Such a model was deduced from the interrelationship between H concentration, yield stress, and activation volume [20]. Additionally, it has also been inferred from previous work that the effective stress component stemming from this type of H-dislocation interaction is proportional to the total amount of solute H contained in the material [20].

Considering its more thermally activatable nature, H potentially has a narrower force-distance profile in Fig. 9 (d), whereas the height of its profile is not exactly known. Since a weaker and more localized obstacle generally yields a smaller V and, in turn, a larger S [29], it has a positive contribution to $\partial\tau_y/\partial\ln\dot{\gamma}$ term in eq. (4) (Fig. 5 (a) and Fig. 8). At the same time, it may be time for dislocations to overcome these newly involved obstacles that control the deformation kinetics [26,28,46] in a relaxation timeframe of 30 s. A high density of these additional rate-controlling obstacles potentially weakens the effectiveness of $\partial\tau_d/\partial\ln\dot{\gamma}$ term in eq. (4) and decreases the slope of the $S-(\tau-\tau_y)$ plot. Indeed, our stress relaxation tests in [20] demonstrated a gradual slope decrease with an

increase in H concentration, an extreme case of which is the horizontal plot in Fig. 5 (a).

Given that H acts as a short-range obstacle, one would attempt to understand the BSR-dependences of S in H-charged specimens in a context similar to Section 4.1. Nonetheless, the concave curvature of $S-(\tau-\tau_y)$ plots (Fig. 5 (a)), as well as their non-linearity in strain rate jump tests (Fig. 6 (a)), are beyond the spectrum that can be covered by the description employed in Section 4.1. Thus, another hypothesis, *solute drag*, should be involved to establish a more comprehensive interpretation.

4.2.2 Contribution of solute drag

A resistance to dislocation motion by dragging solute atmosphere is envisaged in a situation where the diffusivity of solute is comparable to an average velocity of dislocations [16,31–33]. This can apply to the present experiments because of the high mobility of H in austenite lattice even at room temperature [3,58,59]. Cottrell first attempted to roughly estimate the critical strain rate for solute drag under a given solute diffusivity and mobile dislocation density [16]. Later on, the theory was numerically refined, quantifying the drag stress and its variation concerning the dislocation velocity [31–33]. Here, the latest formulation by Sills and co-workers [23,38] is employed to assess the contribution of solute drag in Fig. 5 and Fig. 6. These formulae were successful in explaining some anomalies in the flow behavior of H-charged austenitic steel in our previous publication [3].

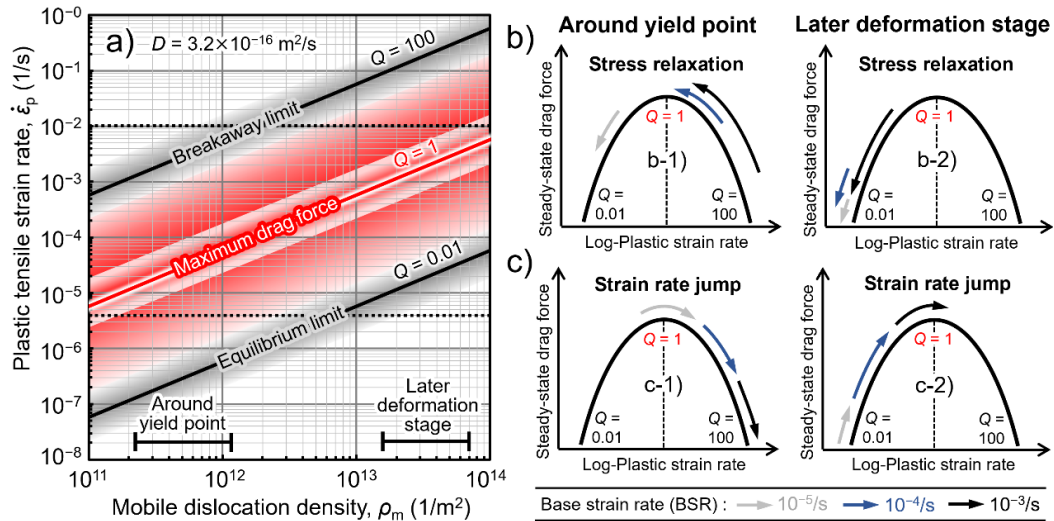


Fig. 11 (a) strain rate range, where drag of H atmosphere is feasible to exert a resistance force (drag force) against steady-state (*i.e.*, constant velocity) dislocation motion, as a function of mobile dislocation density. (b) and (c) indicate the situations in (b) stress relaxation and (c) strain rate jump tests under different base strain rates, which were estimated from (a). The calculations were performed based on eq. (8) ~ eq. (10).

Based on Orowan's formula, the critical strain rate for the dynamic interaction between mobile dislocation and diffusible solutes, $\dot{\epsilon}_c$, is given by

$$\dot{\epsilon}_c = \frac{4QDkT}{M_T\beta} \rho_m b \quad (8)$$

where D is the bulk diffusion coefficient of solute, and ρ_m is the density of mobile dislocations [23]. The parameter Q characterizes the velocity of mobile dislocations, which is non-dimensionalized as:

$$Q = \frac{v_d\beta}{4DkT} \quad (9)$$

Here, β is an elastic interaction parameter for an edge dislocation that is defined *via* shear modulus, μ , Poisson's ratio, ν , and the volume expansion per solute atom, ΔV ($\approx 2 \times 10^{-30}$ m³ for H in Fe-Cr-Ni austenitic steels [2,18]):

$$\beta = \Delta V \frac{\mu b}{3\pi} \left(\frac{1+\nu}{1-\nu} \right) \quad (10)$$

The dislocation is pulled away from the atmosphere when $Q > 100$: breakaway limit. Meanwhile, under $Q < 0.01$, the atmosphere can follow the dislocation by maintaining its near-equilibrium distribution: equilibrium limit. At $0.01 < Q < 100$, the atmosphere lags behind the dislocation, and a resultant non-equilibrium distribution exerts a drag resistance with its maximum at $Q \approx 1$: maximum drag force (Fig. 11 (b) and (c)) [23,38].

The H diffusivity, D , in Type310S steel was measured by Perng and Altstetter [60]. Extrapolation of their data yields $D = 3.2 \times 10^{-16}$ m²/s at 296 K. In Fig. 11 (a), these breakaway limit ($Q = 100$), maximum drag ($Q = 1$), and equilibrium limit ($Q = 0.01$) are drawn for arbitral values of $\dot{\epsilon}_p$ and ρ_m . Note that a band with finite width decorates each limit line to take the possible errors of ΔV and D into account. Our theoretical calculation estimated that a dense H atmosphere around a stationary edge dislocation is formed in a Type310S steel with 7600 at. ppm H at room temperature [3]. The whole or a part of the atmosphere is feasible to migrate with dislocation when $\dot{\epsilon}_p$ and ρ_m lay in-between the range of $0.01 < Q < 100$ (area shaded with red in Fig. 11 (a)). Note that these Q -based criteria can also be applied even to the dislocations extended into two Shockley partials, although the magnitude of drag resistance is somewhat affected by dislocation's character (*i.e.*, fractions of edge and screw components) [38]. For slightly extended dislocations like in Type310S steel with medium stacking fault energy [61], the impact of solute drag on the overall flow stress appears as a statistical average of such character-dependent variation.

4.2.3 Effect of BSR in stress relaxation tests

A critical but experimentally unmeasurable parameter that enters into eq. (8) is ρ_m . Nonetheless, a theoretical fitting of the stress-strain curve was implemented by Alden [62], giving $\rho_m = 10^{11} \sim 10^{12}/\text{m}^2$ around the yield point and its sudden increase over $10^{13}/\text{m}^2$ at a later deformation stage for austenitic steel (cf. horizontal bars in Fig. 11 (a)). In Fig. 10 (b), the $\dot{\epsilon}_p$ during the stress relaxation in H-charged specimens at $\epsilon_t \approx 0.1$ are presented. These ranges of $\dot{\epsilon}_p$ roughly yields the situations during the relaxation and their BSR-dependence as the black, blue, and gray arrows in Fig. 11 (b).

At BSR = $10^{-5}/\text{s}$, the drag force spontaneously decreases with the slowing down of $\dot{\epsilon}_p$ at a small strain around the yield point (gray arrow in Fig. 11 (b-1)). Therefore, the dislocation velocity is primarily dominated by their short-range (thermal) interactions with obstacle H (Section 4.2.1) rather than the interaction with an atmosphere. Moreover, due to the narrow range of $\dot{\epsilon}_p$ (Fig. 10 (b)), the variation of drag force during relaxation should also be small. The drag force becomes further absent at a later deformation stage (gray arrow in Fig. 11 (b-2)). These limited contributions of drag force led to a totally horizontal $S-(\tau-\tau_y)$ plot at BSR = $10^{-5}/\text{s}$ (Fig. 5 (a)).

The situation around the yield point is somewhat different for BSR = 10^{-4} and $10^{-3}/\text{s}$ (blue and black arrows in Fig. 11 (b-1)). The drag force gradually increases with the decay of $\dot{\epsilon}_p$. Thus, the dislocations, which would be mobile if there was no drag resistance, could eventually lose their mobility as the relaxation progressed. This should have a negative contribution to S by limiting the relaxation stress drop, $\Delta\tau_1$, in eq. (2), as well as by increasing $\dot{\gamma}_{2b}$ term due to the release of dislocations from drag resistance upon reloading. Such a negative contribution can be ascribed to the initial concave down of $S-(\tau-\tau_y)$ plots at small stress/strain under these two BSRs (Fig. 5). Explicitly, as expected from Fig. 11 (a) and (b-1), the initial concave down was more substantial for BSR = 10^{-3} than for $10^{-4}/\text{s}$. As ϵ_t becomes larger, the situation same as BSR = $10^{-5}/\text{s}$ (*i.e.*, the limited contribution of drag force) can be reached even for BSR = 10^{-4} and $10^{-3}/\text{s}$ (Fig. 11 (b-2)). Namely, $S-(\tau-\tau_y)$ plots become horizontal after a certain extent of deformation (Fig. 5).

Even though the initial concave down of $S-(\tau-\tau_y)$ plot at low stress/strain can be interpreted by solute drag, the BSR-dependence of S in the horizontal part of $S-(\tau-\tau_y)$ plots (*e.g.*, $\tau-\tau_y > 40$ MPa in Fig. 5 (a)) is not straightforward. Possibly, in the case of H-charged specimens, the same explanation with Section 4.1.2 can be applied even for stress relaxation tests due to the greater magnitude of $\Delta\tau_1$ (Fig. 4 (a)~(c)) that cannot be compensated by an increase in $\dot{\gamma}_{2b}/\dot{\gamma}_{1e}$ in eq. (2). This can be the reason for the constantly larger S at BSR = $10^{-4}/\text{s}$ than at $10^{-5}/\text{s}$ (Fig. 5). On the other hand, the density of segregated H atoms in dislocation core, which work as thermal obstacles [20], could be reduced at a higher BSR, rendering the S at BSR = $10^{-3}/\text{s}$ smaller than that at $10^{-4}/\text{s}$.

Measurement conditions (*e.g.*, relaxation time) must be optimized to clarify these points, some dedicated experiments for which are ongoing by the authors.

4.2.4 Effect of BSR in strain rate jump tests

The shapes of $S-(\tau-\tau_y)$ plots were more complex in strain rate jump tests, an interpretation for which is now attempted based on Fig. 11 (c). Around the yield point at $\text{BSR} = 10^{-4}/\text{s}$ and $10^{-3}/\text{s}$ (Fig. 11 (c-1)), the drag force sharply decreases upon the strain rate jump. This should have a significant negative contribution to S , while this effect could be limited at $\text{BSR} = 10^{-5}/\text{s}$. One thus envisages that the initial concave down, discussed in Section 4.2.3, appears more explicitly at $\text{BSR} = 10^{-4}/\text{s}$ and $10^{-3}/\text{s}$. Notably, the $S-(\tau-\tau_y)$ plots in Fig. 6 indeed exhibited such a tendency.

Fig. 11 (c-1) expects that (I) $\dot{\epsilon}_p$ reached the breakaway limit ($Q = 100$), and the drag force disappeared upon the strain rate jump from 10^{-3} to $10^{-2}/\text{s}$, while the drag force might more or less exist after the jump from 10^{-4} to $10^{-3}/\text{s}$. Furthermore, (II) the decreasing rate of drag force potentially becomes steeper as $\log-\dot{\epsilon}_p$ gets away from $Q = 1$ (cf. blue and black arrows in Fig. 11 (c-1)) [38]. Both these (I) and (II) amplify the negative contribution of drag force to S at a higher BSR condition, one of the reasons for consistently smaller S at $\text{BSR} = 10^{-3}/\text{s}$ than at $10^{-4}/\text{s}$ in the low stress/strain domain in Fig. 6. Additionally, since a major part of H atmosphere is left behind the moving dislocation under such a circumstance, (III) the concentration of H along the dislocation core (thermal obstacles to dislocation motion) could also be reduced: increase in the intercept of $S-(\tau-\tau_y)$ plots by H (Fig. 8) could be minor. A combination of (I)~(III) works to reduce the S in H-charged specimens, making it equivalent to or even smaller than the values in non-charged specimens at the same stress/strain level (cf. $\tau_y = \sim 40$ MPa and $\epsilon_t = \sim 0.05$ in Fig. 6). Note also that the relief of dislocations from their drag resistance seemingly appeared on the characteristics of stress-strain behavior after the strain rate jump. Namely, the density of dislocations with higher mobility suddenly increases, causing the yield-drop phenomenon captured in Fig. 7 (b).

According to Fig. 11 (a), the above factors (I)~(III) are gradually weakened with an increase in ρ_m : $\dot{\epsilon}_p$ -range of $10^{-4} \sim 10^{-2}/\text{s}$ gets away from $Q = 100$ and lays close to or below $Q = 1$. The S thus eventually recovers to the same level with stress relaxation tests, wherein a larger strain (ρ_m) is required for the recovery at $\text{BSR} = 10^{-3}/\text{s}$ than at $10^{-4}/\text{s}$ (Fig. 6). Afterwards, the situation turns into Fig. 11 (c-2) when ρ_m exceeds $10^{13}/\text{m}^2$. In this regime, the strain rate jump oppositely increases the drag force and can positively contribute to S , the impact of which is most significant at $\text{BSR} = 10^{-4}/\text{s}$ (cf. blue arrow in Fig. 11 (c-2)). This is a rationale behind the larger S at $\text{BSR} = 10^{-4}/\text{s}$ as compared with at $10^{-5}/\text{s}$ (Fig. 6). Ultimately, the additive contribution by drag force could escalate S in

strain rate jump tests to a greater level than those in stress relaxation tests as the deformation goes into a later stage with $\tau_y > 80$ MPa and $\varepsilon_t > 0.1$ (Fig. 6).

Meanwhile, the S value at $\text{BSR} = 10^{-3}/\text{s}$ exceeded that at $10^{-4}/\text{s}$ in a further large stress/strain domain with $\tau_y > 100$ MPa and $\varepsilon_t > 0.15$, although the reason for it is yet to be known. Probably, multiple factors (*e.g.*, inherent dependence of S on BSR (Section 4.1.2), role of H as thermal obstacles (Section 4.2.1), and solute drag) might be complexly mixed here, and individual contribution of these influencing factors cannot be distinguished unfortunately in the present test program. Note also that the present analysis and interpretations are all based on the assumption that the density and arrangement of dislocations are not affected significantly by the presence of H. Fortunately, it was confirmed by our past evaluations of the deformation microstructures and strain-hardening properties in the same material [3,4,20,55].

5. Summary: roles of effective stress and solute drag on SRS and SSH

The experimental assessment of the strain rate sensitivity, S , and its match-up with theoretical formulae substantiated the competitive or cooperative contributions of the H-induced effective stress and atmosphere drag resistance to the flow stress in Type310S austenitic steel. Even though the confirmation was carried out indirectly and some uncertain issues were left, it can reinforce our previous statement [3,20] estimating the presence of solute drag, particularly around the yield point. Those experiments were performed only under the condition of 7600 at. ppm H, leaving the H concentration-dependence of the magnitude of drag stress being unraveled. Nonetheless, considering the linear concentration-dependence in the conventional theories of solute drag [16,33], the H-induced drag resistance should also be proportional to H concentration in a similar manner to the effective stress component caused by H [20].

Fig. 12 schematically summarizes the relationship between effective stress and glide resistance by solute drag as a function of a logarithm of the mobile dislocation velocity. The effective stress should increase linearly in terms of the Arrhenius type rate equation (eq. (5)), whereas the decreasing density of H atoms along the dislocation core in the regime above $Q = 1$ could make it deviate from the linearity toward the lower side. When the velocity is in the regime below $Q = 1$, the effective stress and drag force additively contribute to enhancing both the flow stress and S . A typical example of such a situation was the strain rate jump test with $\text{BSR} = 10^{-4}/\text{s}$ at a later stage of deformation (Fig. 11 (c-2)) where the S value was significantly large (Fig. 6). On the other hand, although these two factors are additive under a constant velocity of dislocations, their prompt acceleration diminishes the drag resistance and negatively contribute to S when the velocity is greater than $Q = 1$. A representative case was the strain rate jump test with

BSR = $10^{-3}/s$ around the yield point (Fig. 11 (c-1)), where S in the H-charged specimen sank below that in the non-charged one. The last fact indicates that the net H-related S value can notably be negative due to a sudden loss of solute drag. Nonetheless, a sufficient positive contribution from other intrinsic thermal obstacles (*e.g.*, solute atoms and forest dislocations) compensates for such a negative impact of H (Fig. 8). This maintains the total S value still positive, preventing the emergence of plastic flow instability [24,63].

Fig. 12 anticipates that, in terms of the yield stress, the SSH by solute H at room temperature should be maximized around $Q = 1$ where effective stress and solute drag most effectively cooperate. The corresponding strain rate is $10^{-5} \sim 10^{-4}/s$ if one refers to the small ρ_m regime in Fig. 11 again. Indeed, our measurement for the same Type310S steel demonstrated such a trend when the yield stress in the H-charged specimen was evaluated in the strain rate range of $5 \times 10^{-7} \sim 5 \times 10^{-3}/s$ [3,4]. Also in the present study, the gap of yield stress between non-charged and H-charged specimens was greater at the strain rates of 10^{-5} and $10^{-4}/s$ than that at $10^{-3}/s$ (Fig. 3).

In the present study, we only dealt with the condition of 7600 at. ppm H, leaving the influence of H concentration on the mechanical transient and S parameter both unraveled.

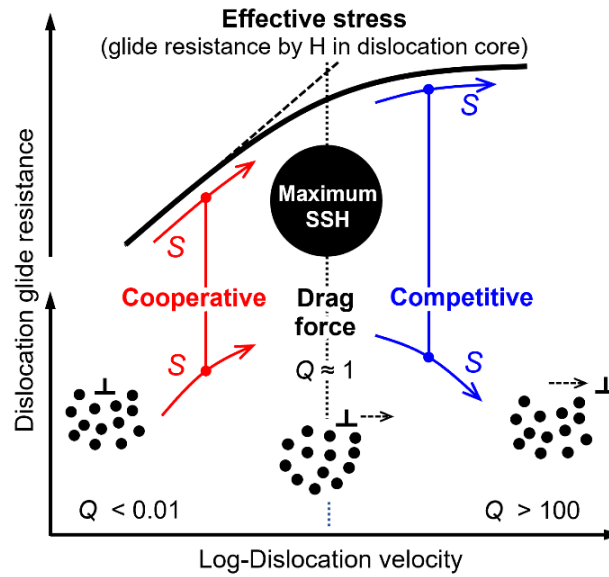


Fig. 12 Schematic illustration of the cooperative or competitive contribution of effective stress and solute drag force by H to the strain rate sensitivity at room temperature.

6. Conclusions

The H-induced SSH at 296 K was studied in Type310S austenitic steel containing 7600 at. ppm H. Two types of transient mechanical tests: stress relaxation; and strain rate jump, were performed under various BSRs, determining the SRS as a measure of stress components responsible for SSH under a given deformation condition.

The SRS clarified two possible roles of H that contribute to the flow stress components: (i) H atoms as short-range, thermally activatable obstacles; and (ii) a resistance to dislocation motion by dragging the diffusible H atmosphere (*i.e.*, solute drag). The significance of both these components was strain rate dependent. Factor (i) consistently exerts a positive contribution to SRS, a typical characteristic when an obstacle type that is more thermally activatable than others is newly involved. Meanwhile, factor (ii) either increases or decreases SRS depending on whether the dislocation velocity range of interest, which is a function of mobile dislocation density and strain rate, lays below or above the limit of maximum drag force. The most effective H-induced SSH can be derived when the sum of the contributions from (i) and (ii) is optimally maximized. In terms of yield stress, for instance, such a situation is achieved in the strain rate range of $10^{-5}\sim 10^{-4}$ /s.

Acknowledgments

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Data availability

The raw/processed data for reproducing the findings in this study are available from the corresponding author upon request by readers. However, the availability will be determined on a case-by-case basis in consultation with the organizations who funded to this research.

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