On the Thermodynamic Aspect of Interaction Between Dislocations and Highly Mobile Interstitial Solute: With Special Focus on Recent Progress in Pd–H System: A Review

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Abstract It is well known that, in kinetics, the interaction between dislocations and interstitial solute normally exerts strong solute drag effect on dislocations, leading to strong solution hardening of the metals. However, due to the low mobility of interstitial solute in many metals, thermodynamic aspect of the interaction between dislocations and interstitial solute is often unobservable and omitted. It will be shown in this article by reviewing the H-induced behaviors in metal–H systems, especially the recent progress in Pd–H system that, when the interstitial solute atoms are highly mobile and able to collect in the vicinity of mobile dislocations easily, the scenario will be remarkably different. The interaction between dislocations and these highly mobile interstitial solute atoms, in thermodynamics, will reduce the line energy of dislocations and will facilitate the generation of dislocations, leading to an increase in dislocation density and an enhanced strain hardening of metals upon plastic deformation.

KEY WORDS: Plastic deformation; Dislocation; Interstitial solute; Strengthening mechanism; Thermodynamics

1 Background

Interaction between dislocations and solute is among the most important issues in materials engineering, because the addition of solute in metals normally leads to the hardening of metals, causing the well-known solid solution hardening [1]. Depending on the types of solute elements (substitutional or interstitial), the solid solution hardening is basically originated from two mechanisms. When the sizes of solute atoms are large, they will occupy the regular sites of the crystal lattice and lead to the formation of substitutional solid solution. Due to the

difference between the solute atoms and the matrix atoms, the lattice in the vicinity of the solute atoms will be distorted. The interaction between the stress fields of dislocations and those around substitutional solute atoms will impede the motion of dislocations, causing the hardening of the metals [2]. On the other hand, if interstitial atoms are introduced into the metals and when the diffusion kinetics of these atoms is favored, the solute atoms will collect in the vicinity of dislocations, forming Cottrell cloud [3]. The interaction between the Cottrell cloud and the dislocations which is also known as solute drag effect will exert impeding force for the motion of dislocations, resulting in the hardening of metals as well [1]. Normally, since the mobility of interstitial atoms is low at low temperature, e.g., room temperature, the solute drag effect of Cottrell cloud on dislocations is strong, and the solution hardening by interstitial solute behaves more effectively in impeding the motion of dislocations and, therefore, causes more significant hardening effect on metals than that by substitutional solute [2]. Basically, the

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significant hardening effect by the solute drag effect of Cottrell cloud on metals can be attributed to two factors. (1) High concentration of solute in the vicinity of dislocation cores; for example, the maximum carbon concentration observed within Cottrell cloud along a dislocation line in bcc iron by atom probe tomography can reach (8 ± 2) at.% which is equivalent approximately to 21 ± 1 carbon atoms per atomic plane [4]. (2) Low mobility of solute at relatively low temperature; for example, the diffusion distance of carbon in bcc iron at room temperature for 1 s estimated by using the diffusion coefficient of carbon at room temperature ($D = 2 \times 10^{-20} \text{ m}^2 \text{ s}^{-1}$ [5]) is only around 3 Å which is about one jump in lattice only.

The hardening of metals by the solute drag effect of Cottrell cloud on dislocations has been well documented in the literature, whereas the other aspect regarding the interaction between interstitial solute and dislocations has not been paid enough attentions, i.e., what if interstitial solute atoms become extremely mobile and can collect in the vicinity of mobile dislocations easily? This is exactly the case of H in many metals, e.g., Fe, Nb, V, and Pd, where H atoms may perform many jumps in crystal lattice within 1 s, even at temperatures as low as 50 K [6]. Actually, for some heavier interstitial atoms, e.g., C and N in Fe, at moderate temperatures, e.g., 473–573 K, the mobility of these atoms can be rather high as well according to their diffusion coefficient [7]. In this case, the interaction between these highly mobile interstitial solute and dislocations is expected to exhibit different characteristics compared to the conventional understandings stated above. On the one hand, in kinetics, solute atoms segregating in dislocations will weaken the solute drag effect on dislocations, according to the classical solute drag theory [3, 8, 9]. On the other hand, in thermodynamics, the collection of solute in dislocations will relax the lattice stress around dislocation cores, leading to the reduction in line energy of the dislocations. However, the above thermodynamic aspect has not been addressed well in the literature. In this review, we will discuss the thermodynamic aspect of interaction between dislocations and highly mobile interstitial solute based on several metal–H systems with special focus on the recent progress in Pd–H system. The major advantage of metal–H systems is that compared to other metal–interstitial solute systems the mobility of H in lattice of many metals is high, for an instance, the diffusion coefficient of H in Pd is around $10^{-11} \text{ m}^2 \text{ s}^{-1}$ at room temperature [6]. This may ensure a fast collection of H in the vicinity of dislocations and exert strong H–dislocations interaction [10, 11]. In addition, the alloying of H into metals is simple, which can be achieved precisely by electrochemical charging [12].

2 A Novel Defactant Concept

As is well known, it is difficult to produce bubbles in pure water, because surface energy of these bubbles is high; the as-produced bubbles suffer from high instability and will annihilate immediately to reduce total Gibbs free energy of the system. However, if surfactant (surface acting agent) is introduced into the water, the formation of bubbles will be facilitated remarkably, because the surface energy of the bubbles will be reduced significantly by the interaction between surface and surfactant. The quantitative theoretical treatment was carried out by Gibbs with his famous Gibbs isotherm adsorption theorem [13], where the reduction in surface energy by the interaction between surface and surfactant can be described as

$$d\gamma_{\text{surf}} = -\Gamma_{\text{surfactant}} d\mu_{\text{surfactant}}, \quad (1)$$

where γ_{surf} , $\Gamma_{\text{surfactant}}$ and $\mu_{\text{surfactant}}$ are the surface energy, the surfactant excess at the surface, and the chemical potential of surfactant, respectively.

Recently, the concept of surfactant was extended into crystalline materials by Kirchheim [14–16]. It is shown that the segregation of solute atoms in lattice defects, e.g., vacancies, dislocations, grain boundaries/interfaces, may lead to the reduction in energy of lattice defects, which will, in turn, facilitate the formation of the lattice defects and stabilize these defects [14–16]. Similar to surfactant, the solute used to reduce the energy of lattice defects can be called as defactant (defect acting agent) [16]. The interaction between defactant and lattice defects can be described by a similar expression to the Gibbs isotherm adsorption theorem [14], i.e.,

$$d\gamma_{\text{def}} = -\Gamma_{\text{def}} d\mu_{\text{def}}, \quad (2)$$

where γ_{def} , Γ_{solute} and μ_{solute} are the energy of lattice defects, the solute excess at the defects, and the chemical potential of solute, respectively. Regarding dislocations, γ_{def} corresponds to line energy of dislocations. According to the defactant concept [14–16], solute segregation in a dislocation will cause relaxation of lattice stress around the dislocation core, leading to the reduction in line energy of dislocations.

So far, the defactant concept has been verified to be very successful in tailoring multiplication and stability of the lattice defects in metals, e.g., grain boundaries in nanocrystalline materials [14–19], non-equilibrium vacancies in interstitial alloy systems [15, 20–22], and dislocations in metals [10, 11, 23–27]. In particular, this concept is able to provide convincing thermodynamic explanations on hydrogen embrittlement in metals, where it is illuminated that the enrichment of H in local regions of crack tips in the metals can reduce the generation energy of double kinks of dislocation lines, which enhances the mobility of

dislocations [23–27]. Hydrogen embrittlement of metals can be caused by H–dislocations interaction in local region of metals. However, if the H atoms distribute homogeneously in the microstructure, the scenario can be different and the interaction of H with dislocations is expected to occur globally rather than locally. Within the framework of the defactant concept, the interaction of H with dislocations globally will reduce the line energy of dislocations globally as well. As a consequence, upon plastic deformation, the generation of dislocations will be facilitated in the whole sample rather than in a local region. This will cause an increase in dislocation density by H addition. Furthermore, since the generation of dislocations plays a crucial role in hardening and plastic deformation behaviors of metals in most cases, the facilitation of dislocation generation by H–dislocations interaction will be expected to influence these behaviors as well. The above consequences of H–dislocations interaction have been verified in the metal–H systems, especially the recent experimental investigations on Pd–H system, which will be reviewed in the following sections.

3 Increasing Dislocation Density in Cold-Deformed Pd by Alloying H

Work hardening is one of the most widely used methods to strengthen metals. By cold deformation, high density of dislocations can be produced. The interaction of the produced dislocations will impede their further slip. A higher stress will be required to activate plastic deformation, leading to work hardening of the metals. The enhanced work hardening can normally be achieved by increasing the dislocation density by increasing the degree of plastic deformation [2]. Alternatively, according to the defactant concept, if some specific solute elements are introduced as defactant into the metals, the interaction between solute and dislocations will facilitate dislocation formation as above mentioned. Then, subjected to the same degree of plastic deformation, more dislocations will be generated and will enhance the work hardening of the metals. The work in cold-rolled Pd–H system [10, 28–30] reviewed in the following testified the concept. Because H can be charged into and removed from Pd easily, this alloy element can be taken as a “temporary defactant” to enhance the dislocation generation [15].

Well-annealed Pd sheets were charged with different amount of H electrochemically in an electrolyte mixed from 1/3 phosphoric acid (85%) and 2/3 glycerin (85%) by taking Pd sheets as cathode and a platinum net as anode. The amount of H was well defined by Faraday’s law. Subsequently, the Pd sheets were cold-rolled to 50% of the initial thickness. Detailed experimental procedures are available in Refs. [10, 29]. The method used in Refs. [10,

29] to charge H into Pd enables several ppm level control in H amount and, therefore, is considered to be rather accurate and reliable [12]. Dislocation densities of the cold-rolled Pd samples were estimated by two independent methods, i.e., (1) analyzing the broadening of Bragg peaks of the measured X-ray diffraction patterns [10], and (2) measuring the diffusion of H in deformed Pd sheets [29]. It should be noted that, in Refs. [10, 29], before measuring the dislocation densities, H was removed electrochemically from the Pd samples completely.

As shown in Fig. 1a, b, the broadening of the X-ray diffraction (XRD) Bragg peaks of cold-rolled Pd–H alloys becomes more and more significant with increasing initial H/Pd ratio. Since the contribution of the crystallite size to the XRD Bragg peak broadening is negligible due to the large grain size, the observed broadening of XRD Bragg peaks is mainly attributed to the dislocations [10]. The enhanced broadening of XRD Bragg peaks shown in Fig. 1a, b with increasing initial H/Pd ratio indicates the increase in dislocation densities by increasing initial H/Pd ratio. Applying Williamson and Smallman’s simple dislocation model [31], dislocation densities of the cold-rolled Pd–H samples were estimated from slopes of the Williamson–Hall plots in Ref. [10].

Alternatively, since the dislocations in the cold-rolled Pd are the major traps for H, the trapped H in Pd upon the diffusion of H across the cold-rolled Pd sheets can be a measure of relative dislocation densities in the cold-rolled Pd. This was formulated by Deutges *et al.* [29]

$$\frac{\rho_A}{\rho_B} = \frac{\rho_A \Gamma(\mu_H) \Omega_{Pd}}{\rho_B \Gamma(\mu_H) \Omega_{Pd}} = \frac{c_{t,A}}{c_{t,B}}, \quad (3)$$

where ρ_A and ρ_B are the dislocation densities of two different samples, $\Gamma(\mu_H)$, Ω_{Pd} , $c_{t,A}$ and $c_{t,B}$ are the H excess in dislocations at a chemical potential of H (μ_H), and the concentrations of H trapped at sample A and sample B, respectively. The comparison of the trapped H at the same μ_H of two different samples allows the estimation of the relative dislocation densities of the two samples. It is shown in Fig. 2 that at the same chemical potential, with increasing t initial H/Pd ratio, the concentrations of H trapped at dislocations in cold-rolled Pd samples increase, suggesting the increase in dislocation density.

The dislocation densities of the cold-rolled Pd–H alloys estimated by the above two methods are shown and compared in Fig. 3. The results indicate that the progressions of the dislocation densities determined by the two independent methods are identical. Therefore, it can be concluded that the addition of H as defactant facilitates the generation of dislocations and, thus, increases the dislocation density in cold-rolled Pd.

As illuminated above, H can be taken as a temporary alloy element to increase dislocation density in cold-rolled

Fig. 1 **a** Measured XRD profiles of the cold-rolled Pd–H alloys with different H/Pd ratios and well-annealed Pd, **b** Williamson–Hall plots for the diffraction patterns of cold-rolled alloys and well-annealed Pd [10]

Fig. 2 Concentration of the trapped hydrogen c_t at lattice defects in cold-rolled Pd versus the chemical potential μ_H of H in Pd for samples with different H contents during deformation [29]

Fig. 3 Development of the dislocation density of cold-rolled Pd with increasing H content during cold rolling. The measurements were performed by two different methods: interpretation of the diffraction patterns of the samples using the method of Williamson–Hall plots and measurement of the relative dislocation density by evaluating the diffusion time of H through the Pd samples [29]

Pd. Since the increase in dislocation density will lead to the hardening of metals [1, 2], H can be treated as a temporary alloy element to harden metals as well in this context. Figure 4a shows Vickers hardness of the cold-rolled Pd with different initial H/Pd ratios, where it can be seen that with increasing initial H/Pd ratio, hardness of the cold-rolled Pd increases systematically. One should note that as mentioned in Ref. [10] the hardness was measured after completely removing the H in Pd. Therefore, the samples used for hardness measurements contain no H and the increase in hardness of the cold-rolled Pd can only be ascribed to the increased dislocation density. This is also verified by comparisons of the experimental data with Taylor equation, see Fig. 4b [28].

Besides cold rolling, the increase in dislocation density in Pd by alloying H is also demonstrated in a shock-loading experiment by Melikhova et al. [30]. The authors applied intensive plastic deformation on well-annealed Pd doped by various amount of H by shock-loading via a Split-Hopkinson apparatus and characterized the lattice defects, i.e., vacancy clusters and dislocations, by positron lifetime spectroscopy combined with XRD. As shown in Fig. 5a, b, both concentration of vacancy clusters and dislocation density in the shock-loaded Pd increase with increasing pre-charged H concentration. It is interesting that the changes in defects densities show strong dependence on H concentration only, and the alternation of strain rate seems to be not influencing the defects densities significantly at the same H concentration, see Fig. 5a, b. This behavior is unusual, because the increase in strain rate normally suppresses the dynamic recovery of lattice defects (at least for dislocations) upon plastic deformation and leads to the increase in defects density [32]. A possible explanation on this unusual behavior is the stabilization effect of H on dislocations. Although corresponding to the same H concentration a higher strain rate could produce more dislocations, the additional

Fig. 4 **a** Vickers hardness (Hv) as a function of the initial H/Pd ratio, **b** linear fit of the measured Hv to square root of the measured dislocation density [28]

Fig. 5 The density of dislocations **a** and the concentration of deformation-induced vacancies **b** plotted as a function of the hydrogen concentration in the pre-charged samples [30]

dislocations produced by the higher strain rate may not be stabilized sufficiently by H and will annihilate subsequently, leading to the consequence that an identical defect density observed in the samples with the same H concentration but different strain rates.

It has to be pointed out that the increase in dislocation density by solute addition is not observed extensively. As stated in the classical textbook [1], no great increase in grown-in dislocation density on addition of solute is observed. The reason can be explained by the defect concept. According to Eq. (2), a significant reduction in line energy of dislocation requires a large Γ_{def} , a large amount of solute segregating in the dislocation. In most metal-interstitial solute systems, the mobility and the solubility of solute at relatively low temperatures, e.g., room temperature, are low. The low mobility will prevent a pronounced segregation of solute at dislocations and therefore will certainly not lead to a significant reduction in line energy of dislocation. Although, in some systems, e.g., Fe–C system, a pronounced segregation of solute atoms at dislocation can be achieved at moderate temperatures, the low solubility of solute will, again, not result in a large Γ_{def} , when massive dislocations are produced upon plastic deformation. Therefore, it is believed that low solubility and relatively low mobility of interstitial solute atoms in most metals at low temperature may be the major reason for preventing the observation of increase in dislocation density by interstitial solute addition.

4 Impact of H on Deformation Behaviors of Pd

The work regarding the cold-rolled Pd–H alloys reviewed in Sect. 3 has demonstrated that thermodynamic consequence of the interaction between H and dislocations is the

reduction in line energy of dislocations, which, therefore, facilitates significantly the generation of dislocations in the cold-rolled Pd. Since plastic deformation of most metals is associated with generation and motion of dislocations, the above thermodynamic effect of such an interaction is certainly expected to influence the behaviors of plastic deformation of metals as well, which has also been observed in Pd–H system [11].

In order to avoid the loss of H, in Ref. [11], the in situ tensile tests on well-annealed Pd and Pd–H alloys were performed in an electrolyte bath used to charge H into Pd at room temperature and a constant strain rate of $6.67 \times 10^{-4} \text{ s}^{-1}$. Figure 6a shows interestingly that the presence of H does not affect the yield strength but enhances the strain hardening of Pd remarkably. As shown in Fig. 6b, the strain hardening exponent (n) by linear fitting with Holloman equation ($n = d \ln \sigma_{tr} / d \ln \varepsilon_{tr}$) [33] increases systematically with increasing H/Pd ratio, and

when H/Pd reaches 10,000 at.ppm, the value of n up to 0.75 can be achieved.

The negligible change in yield strength of Pd with the presence of H is ascribed to the extremely high mobility of H in Pd lattice [11]. As estimated in Ref. [11], the diffusion speed of H in Pd is almost 3 orders of magnitude higher than the dislocation velocity at the strain rate of $6.67 \times 10^{-4} \text{ s}^{-1}$. This means that H can move along with the mobile dislocations easily, thus exerting negligible solute drag effect on the dislocations. On the other hand, the strain-hardening behaviors are strongly associated with the dislocation multiplication upon plastic deformation. Therefore, the enhanced strain hardening of Pd by alloying H should be closely related to the influence of H on generation of dislocations. Since the interaction between H and dislocations reduces the line energy of dislocations in Pd [10, 15], the addition of H will certainly lead to the reduction in the formation energy of dislocations and, concomitantly, the driving force for their annihilation [11]. Therefore, the dislocation multiplication upon plastic deformation will be enhanced with the presence of H, leading to the enhanced strain hardening with increasing H content.

Similar experimental studies on the impacts of H on plastic deformation behaviors of metals have been investigated extensively. Both H-induced hardening and softening are observed. The former has been found in many fcc metals, e.g., Al [34], Ni [35], and austenite stainless steels [36, 37], while the latter appears in several bcc metals, e.g., high-purity Fe [38–40], ferritic steels [41], and β -Ti alloys [42, 43]. The H hardening or softening behaviors in alloys and pure metals with relatively low purity are considered to be rather complicated and related to many different factors, such as hydride formation [34], impurities [44], and surface effect [45], which are beyond the major concerns on H–dislocations interaction. Therefore, in the following, we will focus, mainly, on two simple and typical solid solution systems, i.e., Ni–H system with fcc structure and Fe–H with bcc structure.

A systematic investigation on tensile deformation behaviors of Ni–H system was carried out by Boniszewski et al. [35]. Figure 7a, b shows the measured engineering stress–strain curves of well-annealed H-free Ni and H-charged Ni tested at different strain rates. At a lower strain rate of $3.33 \times 10^{-4} \text{ s}^{-1}$, H shows negligible effect on yield strength of Ni (Fig. 7a). Further increasing the strain rate to $8.33 \times 10^{-3} \text{ s}^{-1}$, an apparent increase in yield strength of Ni by charging H is resulted and the effect becomes more significant with increasing H content (Fig. 7b). Apparently, because of a lower dislocation velocity at a lower strain rate, H exhibits a weaker/negligible solute drag effect on dislocations in Ni, whereas, at a higher strain rate, this effect becomes stronger, therefore

Fig. 6 **a** True stress–strain curves of Pd and Pd–H alloys with different H concentrations, **b** linear fitting on $\ln \sigma_{tr}$ – $\ln \varepsilon_{tr}$ curves and the resulting strain-hardening exponent n for Pd and Pd–H alloys [11]

Fig. 7 Stress–strain curves of H-free and H-charged Ni tested at different conditions: **a** room temperature, strain rate $3.33 \times 10^{-4} \text{ s}^{-1}$, **b** room temperature, strain rate $8.33 \times 10^{-3} \text{ s}^{-1}$ [35]

causing an increase in yield strength as shown in Fig. 7b. This is consistent with the above explanations on negligible effect of H on yield strength of Pd. Besides this, the addition of H appears to enhance the strain hardening of Ni as well. As shown in Fig. 7a, the yield strength of Ni charged with H is identical to that of the H-free Ni, whereas the tensile strength of Ni–H alloys is higher than that of the pure Ni at the same strain, which is similar to the behaviors observed in Pd–H system (Fig. 6a). However, since the H content in Ni shown in Fig. 7a is rather small, the effect is much less significant compared to the case of Pd–H system (Fig. 6a).

Whether H may induce hardening or softening of Fe was in a controversy in the previous literature due to the differences (mainly concerning the purity of Fe) in materials used in the studies [38]. After careful inspections on the reported experimental data, especially the work of Matsui et al. [39], Hirth [38] came to the conclusion that for ultrahigh-purity Fe with a residual electric resistivity ratio larger than 3600 at temperatures greater than 190 K where

H is mobile enough, H-induced softening will occur. The H-induced softening of Fe is ascribed to H-enhanced kink pairs formation on a screw dislocation [38–40]. Fe is bcc structure at room temperature; its yielding and plastic deformations are dominated by screw dislocations [2]. Since kink pairs on a screw dislocation move along it much more quickly than the screw dislocation itself [46], if the kink pairs formation is enhanced and their sideways motion is not hindered in the presence of H, softening of Fe will be expected to take place. Kinks on a screw dislocation have edge-type character and due to the concomitant hydrostatic stress fields are expected to attract H atoms pronouncedly, which, according to Eq. (2), will be expected to lead to a significant reduction in formation energy of the kinks and enhance their formation. This has been confirmed by atomic scale computer simulation of Wen et al. [47]. However, the H-enhanced kink pair formation is only a necessary condition for softening of Fe. If the solute drag of H on the sideways motion of the kink pair is strong at low temperatures due to a limited mobility of H, a hardening effect may still be resulted. This effect was also observed in the work of Matsui et al. [40] where hardening of the H-charged Fe is presented below 190 K. Different from the H-enhanced formation of kink pairs, the impact of H on multiplication of dislocations in Fe and other bcc metals, which is closely related to the work-hardening behaviors, is not well documented experimentally. The recent molecular dynamics simulation taken by Wagih et al. [48] indicates that multiplication of dislocations in α -Fe nanocrystals with a grain size of 80 nm upon tension can be enhanced by H, which will increase the flow stress of crystals and will favor the homogenous deformation by activating multiple slip systems of Fe. The H-enhanced multiplication of dislocations is suggested to be attributed to the reduction in line energy of dislocations by H–dislocation interaction as stressed by the defactant concept [48].

One should note that different from bcc metals, the yielding of fcc metals is dominated by the motion of edge dislocations [2]. Since the Peierls potential of an edge dislocation is much smaller than that of a screw dislocation, H may pin an edge dislocation rather than increase its mobility by reducing the Peierls stress [40]. Therefore, the negligible changes in yield strength of Pd and the slowly strained Ni in the presence of H mentioned above should be caused by the negligible solute drag effect of H on dislocations only. Despite the different mechanisms involved in the impact of H on the yielding of fcc (Pd and Ni) and bcc (Fe) metals, the mechanisms of the enhanced strain hardening by H-enhanced multiplication of dislocations upon plastic deformation are in convergence, i.e., reduction in line energy of dislocations by H–dislocation interaction.

5 Summary

This study reviews the recent progress in thermodynamics of interaction between dislocations and highly mobile interstitial solute atoms with special focus on Pd–H system. Following a brief introduction on theoretical background of the interaction between dislocations and solute atoms and a novel defactant concept, several consequences of H–dislocations interaction upon plastic deformation testified in Pd–H system and several other metal–H systems are reviewed and discussed within the framework of the defactant concept. The main conclusions and prospective future directions in this field are summarized as follows.

6 Concluding Remarks

Generally, interaction between dislocations and interstitial solute atoms exerts, in kinetics, strong solute drag effect on dislocations, leading to a strong solution hardening of metals. The thermodynamic effect of the interaction on line energy of the dislocations is normally not considered, because in many metal–interstitial solute systems, the mobility of solute in lattice of the metals is low and the solute atoms are not able to collect in the vicinity of dislocations easily, which makes the effect to be hardly observable.

The behaviors observed in Pd–H system and other metal–H systems reviewed in this article indicate that, when interstitial solute atoms become highly mobile and can collect in and move along with mobile dislocations easily, the interaction between dislocations and these highly mobile interstitial solute atoms will reduce the line energy of dislocations and will facilitate the multiplication of dislocations upon plastic deformation. This will, on the one hand, increase the dislocation density and, on the other hand, enhance the strain hardening of metals.

7 Future Directions

1. The work conducted in metal–H systems highlights the possibilities of tailoring dislocation formation and plastic deformation behaviors of metals by taking thermodynamic aspect of the interaction between dislocations and highly mobile interstitial solute into account. Although the behaviors observed in metal–H systems are hardly observed in other metal–interstitial solute systems at room temperatures, they are expected to show up at high temperatures where the mobility and solubility of the interstitial solute are high. One well-known system is Fe–C system which is important in steel industry. In the high-temperature austenite

region, the C atoms are highly mobile in Fe lattice and the maximum solubility of C reaches up to around 9 at.%. In this case, upon hot deformation, the consequences of C–dislocations interaction may probably be similar to those observed in metal–H systems, which may possibly influence the hot working and microstructure evolution of many steels significantly and should certainly deserve further investigations.

2. Due to the high affinity of Pd for H, the H-induced behaviors may possibly appear in some engineering Pd-based alloys, e.g., Pd–Ag, Pd–Au, and Pd–Ir, as well. If so, H can be taken as a temporary element to assist plastic working of these alloys, since H can be charged into and removed from the metals easily. Therefore, whether the H-induced behaviors demonstrated in the binary Pd–H system are available in these Pd-based alloys will require further testifications.
3. Although it has been testified by computer simulation that the multiplication of dislocations in bcc Fe can be facilitated by H–dislocations interaction, there is a lack of direct experimental evidence. Further experimental investigations are highly suggested, especially concerning its relevance to hydrogen embrittlement in ferrite steels. This work is currently ongoing in the authors' group.

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