Materials Science & Engineering A 586 (2013) 231–235

Contents lists available at ScienceDirect

Materials Science & Engineering A

journal homepage: www.elsevier.com/locate/msea

Effects of interactions between dislocations and/or vacancies and He atoms on mechanical property changes in Ni

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ARTICLE INFO

Article history:
Received 2 April 2013
Received in revised form
31 July 2013
Accepted 3 August 2013
Available online 16 August 2013

Keywords:
Helium
Mechanical property
Dislocations

ABSTRACT

It is well known that helium (He) generated by (n, α) nuclear reactions in metals and alloys irradiated with high-energy particles prompts not only the nucleation of interstitial-type dislocation loops, but also the nucleation of voids, which degrades mechanical properties. In the present study, He atoms were implanted at 150 eV without causing displacement damage in Ni containing either dislocations, vacancies or a combination of both. The results showed that upon He implantation, the ultimate tensile strength increased and total elongation decreased in samples containing both dislocations and vacancies. However, it was also found that in samples with He trapped by dislocations, where the He concentration decreased with increasing sample depth, nickel increased both the ultimate tensile strength and total elongation at 300 K, but that the effect of He disappeared at 573 K. In addition, He trapped by vacancies did not strongly affect the ultimate tensile strength or total elongation in samples containing only vacancies.

1. Introduction

Helium (He) is generated in materials by (n, α) nuclear reactions under neutron irradiation. The rate of He generation increases with increasing neutron energy. He atoms are insoluble in most metals and alloys, and the presence of He atoms in metals greatly impacts their microstructure. He atoms promote the growth of cavities [1,2] and dislocation loops [3,4], and can induce void swelling. In addition, the interaction between He atoms and defects increases hardness and decreases ductility.

Point defects, such as interstitials and vacancies, are produced simultaneously in metals and alloys under irradiation with high-energy particles. Interstitials and vacancies aggregate to form clusters, namely, interstitial-type dislocation loops, dislocations grown from loops, and voids, all of which can strongly interact with He atoms [5–8]. Thus, it is difficult to distinguish the interaction of any one type of defect with He atoms in metals and alloys under normal irradiation conditions. However, it is possible to induce dislocations or vacancies separately in samples using a special method. The present study takes a fundamental look at the interaction between He atoms and dislocations and vacancies in materials and investigates the effects of He atoms on the mechanical properties of nickel (Ni) containing only dislocations or vacancies.

2. Experimental procedure

Pure Ni (99.999%) was used. The experimental procedures employed for specimen preparation are shown in Fig. 1. Specimens that were 0.11 mm thick were annealed at 1173 K for 1 h under high vacuum conditions (10^{-4} Pa) (procedure I). Defects were introduced by cold rolling using a rolling machine. Well-annealed specimens were cold-rolled to 10% of their original thickness (procedure II). Along with defects, which were mainly dislocations, vacancies and vacancy clusters were introduced into the specimens. To annihilate these vacancies and vacancy clusters, cold-rolled samples were annealed at 673 K for 1 h in a vacuum (procedure IV). On the other hand, vacancies were induced by electrons using an electron linear accelerator at Kyoto University at room temperature (procedure VI). The electron energy was 8 MeV, and the irradiation dose was 4.4×10^{21} e/m^2 (10^{-4} dpa). As shown in Fig. 2, tensile samples were punched from the rolled sheet before the last heat treatment in the experimental procedure shown in Fig. 1. He implantation was carried out in some samples (procedures III, V, and VII) using a mono-energetic He^+ ion beam at room temperature under a vacuum of 8.0×10^{-6} Pa. The He^+ ions were collimated and mass-analyzed, and the beam flux was monitored by a Faraday cup. To avoid displacement damage, implantation was performed at 150 eV, under the assumption that the threshold displacement energy in Ni is 40 eV [9]. The nominal He^+ dose was 1×10^{20} He^+/m^2 at a flux of 5.0×10^{15} He^+/m^{2}s. The He distribution in Ni was calculated using the SRIM code [10], the results of which are shown in Fig. 3. These results indicate that the implanted He ions are located 0–5 nm from the incident

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http://dx.doi.org/10.1016/j.msea.2013.08.011

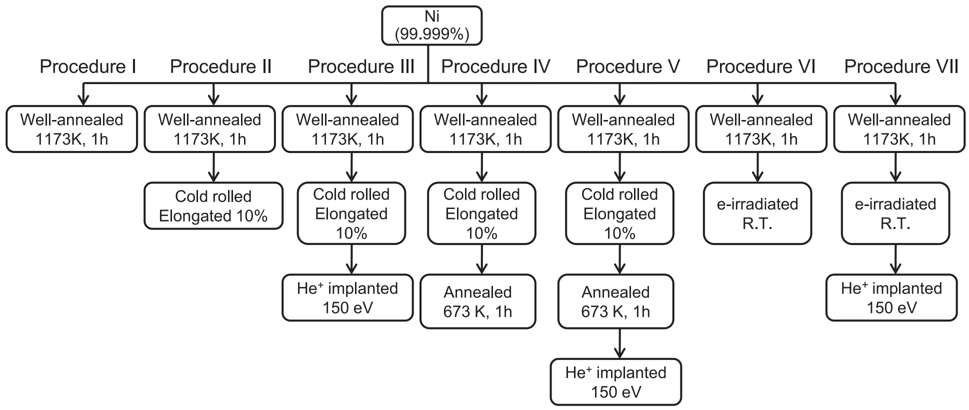

Fig. 1. Preparation of specimens.

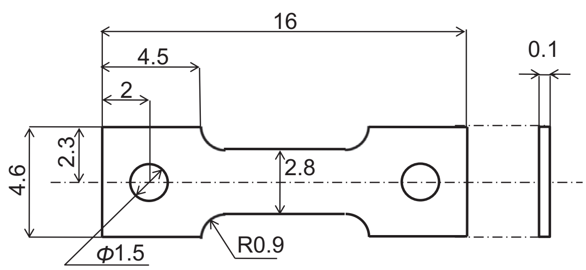

Fig. 2. Configuration and dimensions of test specimens.

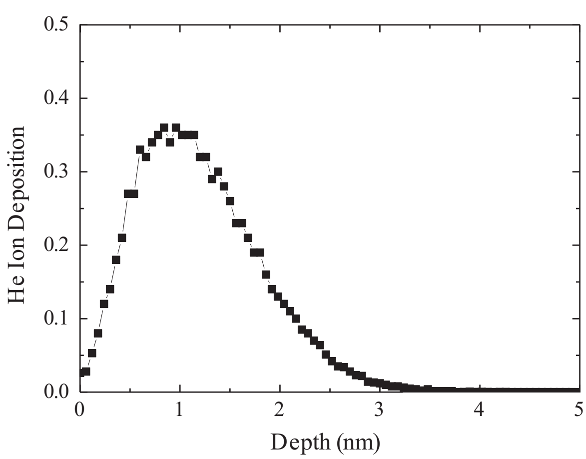

Fig. 3. Depth distribution of ion deposition in Ni implanted with He ions at 150 eV up to 1×10^{20} ions/m^2. Displacement damage could not be produced by 150-eV He ions.

surface and the He distribution peak is ~1 nm from the surface. However, the migration energy of He in Ni is 0.49 eV [11], so at room temperature, He atoms can diffuse freely in bulk Ni deeper into the sample, until they are trapped by defects, such as dislocations, vacancies and vacancy clusters. The tensile specimens were tested at a crosshead speed of 1 mm/min, which corresponds to a strain rate of 2×10^{-3} s.

Positron annihilation spectroscopy (PAS) was carried out to identify the defects in the samples after each experimental procedure (i.e., annealing, cold rolling, and He^+ implantation) by measuring the positron lifetime. The detection depth of PAS is 0.1 mm in Ni. The positron lifetime spectrometer had a time resolution of 190 ps (full width at half maximum, FWHM), and each spectrum was the accumulation of 1×10^6 counts or more. To distinguish the defect component from the bulk component, the lifetime spectrum, $L(t)$, was decomposed into two components based on a two-state trapping model, a short lifetime τ_1, and a long lifetime τ_2, after subtracting the radioactive source and background components [12]:

$$L(t) = (I_1/\tau_1)\exp(-t/\tau_1) + (I_2/\tau_2)\exp(-t/\tau_2) \quad (1)$$

where I_i are the intensities ($I_1 + I_2 = 1$).

Within the framework of the two-state trapping model in which positrons are assumed to annihilate, despite the bulk or trapped states at vacancies and vacancy clusters, τ_1 and τ_2 have the following physical meanings:

$$\tau_1 = 1/(\tau_B^{-1} + \kappa) \quad (2)$$

here, τ_B is the positron lifetime in bulk Ni, and κ, the net positron trapping rate due to the defects:

$$\kappa = I_2/I_1(\tau_B^{-1} - \tau_2^{-1}) \quad (3)$$

τ_1 results from bulk material properties, including free electrons and other defects, such as dislocations. The vacancy concentration, Cv, is given by

$$\kappa = \mu \mathrm{Cv} \quad (4)$$

where μ, the trapping coefficient of vacancy clusters, is calculated from the diffusion model via

$$\mu = 4\pi RD/\Omega \quad (5)$$

where R is the trap radius and Ω is the atomic volume. D is the diffusion coefficient of positrons, which is taken to be 2×10^{-5} m^2/s in Ni [13].

3. Results

Defects in the samples after each experimental procedure were characterized using PAS. The main results are shown in Table 1. Several samples were measured after tensile testing using PAS; their results are shown in Table 1 also. In contrast to the well-annealed sample lifetime of 106.5 ps (procedure I), the long-lifetime component, τ_2, was 164.0 ps after cold rolling (procedure II). This is close to the lifetime of single vacancies (175 ps [14]) in Ni. The long lifetime intensity, I_2, of 82.5% indicates that positrons were mainly annihilated at vacancies. τ_2 decreased to 128.8 ps after annealing rolled samples at 673 K for 1 h (procedure IV). Although this was slightly higher than the 119 ps for positron annihilation with dislocations in Ni [14], it was assigned to a dislocation network in Ni. Dislocations have previously been observed in cold-rolled samples annealed at 673 K for 1 h [15]. The dislocation density was estimated to be 2.4×10^{14} /m^2 on the basis of microstructural observation. In contrast, after irradiation by 8 MeV electrons at room temperature (procedure VI), τ_2 was 174.2 ps with an intensity of 50.9%. This indicates that only single