phys. stat. sol. (a) **171**, 35 (1999)

Subject classification: 61.72.Lk; 61.72.Ff; 62.20.–x; S5.12; S6

Recombination-Enhanced Dislocation Motion in SiGe and Ge

I. YONENAGA[1]) (a), M. WERNER (b), M. BARTSCH (b), U. MESSERSCHMIDT (b), and E. R. WEBER (c)

(a) Institute for Materials Research, Tohoku University, Sendai 980-8577, Japan

(b) Max-Planck-Institut für Mikrostrukturphysik, D-06120 Halle/Saale, Germany

(c) Department of Materials Science, University of California, Berkeley, CA 94720, USA

(Received October 5, 1998)

In-situ straining experiments on dislocation motion in Ge and Si–5 at% Ge alloy single crystals are performed in a high voltage transmission electron microscope. In comparison with previous results by other methods, the dislocation velocities are found to be enhanced due to a recombination enhancement owing to the excess carrier injection by the electron beam. The reduction in the activation energy of dislocation motion is ascribed to the recombination-assisted kink formation. The kink migration energy is estimated to be 0.7 eV in Ge and 1.5 eV in SiGe.

1. Introduction

The phenomenon that dislocation glide is enhanced during carrier injection into a crystal by forward biasing or by laser irradiation is well known as recombination-enhanced dislocation motion (REDM). Quantitative analyses have been reported in Si, GaAs and several other semiconducting materials [1]. Werner et al. [2] revealed a higher efficiency of dislocation motion enhancement by direct carrier injection in Si using a high voltage transmission electron microscope (HVTEM), observed at temperatures higher than those by laser illumination [3] or by injection of low voltage electrons [4]. In Ge, only a negligible effect is known caused by low-voltage electron injection [4]. Recently, Inoue et al. [5] reported the enhanced motion of kinks on 30° partial dislocations in Ge by the excitation with a concentrated electron beam during high resolution electron microscopic (HRTEM) observation [5]. A rather macroscopic feature of REDM in Ge can be expected using direct carrier injection during HVTEM observation. It is also interesting whether REDM exists in semiconductor alloys, or not.

This paper reports on dislocation glide in Ge and Si-rich SiGe alloy single crystals by *in-situ* observation in an HVTEM.

2. Experimental

The specimens prepared from an intrinsic Ge crystal and an undoped Si-5 at% Ge alloy crystal were pre-compressed to introduce dislocations. Thin plates parallel to the (111)

[1]) Corresponding author: Tel: +81222152042; Fax: +81222152041;
e-mail: yonenaga@imr.tohoku.ac.jp

plane with a [123] tensile axis were cut from them and then thinned to electron transparency by mechanical polishing followed by chemical polishing. *In-situ* experiments were performed inside a high voltage electron microscope JEOL 1000 (HVTEM) operating at 1000 kV by using a double tilting heating tensile stage [2, 6, 7]. The dynamic motion of dislocations was *in-situ* observed and simultaneously recorded on video tapes. The stress acting on the dislocations was determined by analysing the curvature of dislocation segments fixed at pinning points [8].

3. Results

In both materials dislocations moved easily during the *in-situ* HVTEM observation. Some dislocation sources ceased to operate upon decreasing the electron beam current and again started to operate when the electron beam current was recovered. 60° dislocations, emerging through both surfaces, on the primary slip system were analyzed for velocity measurements. The length of such dislocations was longer than about 1 μm, which means that the dislocation motion is expected to be in *so-called* length independent region [9]. We used rather thick samples for HVTEM observation. Dislocations in motion showed a small bowing that might be due to the influence of surface pinning, but this should result in a back stress, i.e. small change of the effective stress.

Fig. 1 shows the velocities of 60° dislocations in the Ge crystal at temperatures of 428, 474, 566 and 657 °C under an *in-situ* observation with the electron beam current of 10 μA plotted against the resolved shear stress. For the sake of comparison, the velocity versus stress relation of 60° dislocations in Ge [10] in the dark measured by the etch pit technique is included. The logarithm of the velocity of dislocations depends linearly on the logarithm of the stress at a fixed temperature with approximately the same slope described by a stress exponent of $m = 1.5$. The dislocation velocity increases with beam

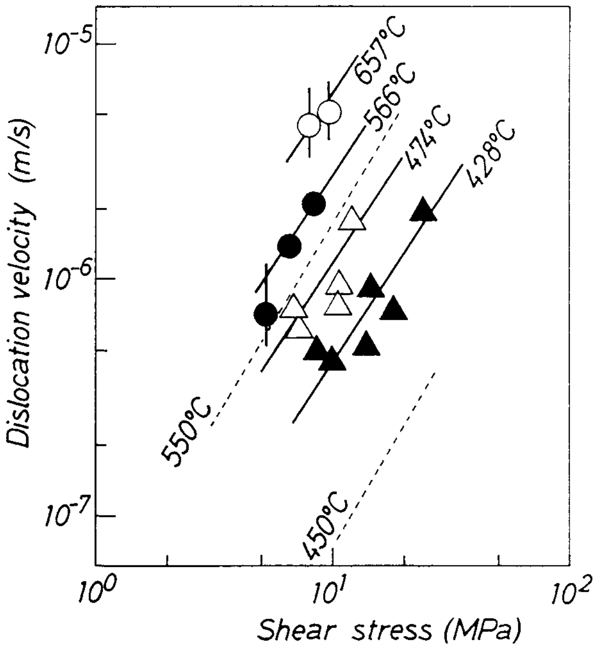

Fig. 1. Velocities of 60° dislocations in the Ge crystal at temperatures of 428, 474, 566 and 657 °C under *in-situ* observation at an electron beam current of 10 μA plotted against the resolved shear stress. Dashed lines show the velocity vs. stress relation of 60° dislocations in Ge in the dark obtained by the etch pit technique [10]

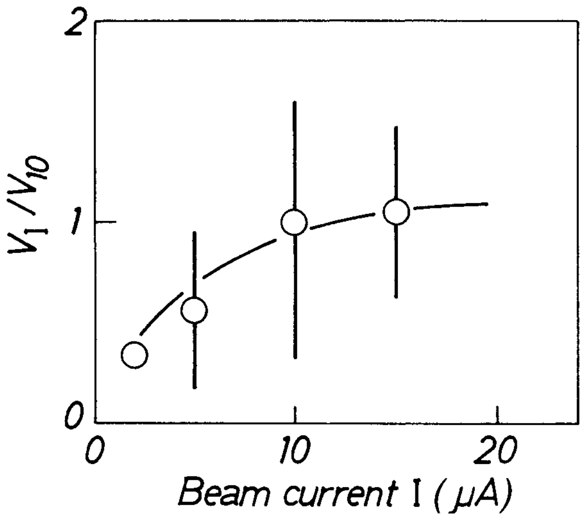

Fig. 2. The beam current variation of the ratio of V_I, the velocity under irradiation with the beam current of I, and V_{10}, that with 10 μA, measured during *in-situ* HVTEM at 566 °C under a stress of 8 MPa in Ge

current as shown in Fig. 2, with a saturation above a beam current of 10 μA. The dislocation velocities in the *in-situ* observation on the SiGe alloy were higher than those measured by the etch pit technique without carrier injection [11].

An Arrhenius plot of the velocities of 60° dislocations in Ge and SiGe during the *in-situ* observation with the electron beam current of 10 μA estimated for the stress $\tau = 20$ MPa

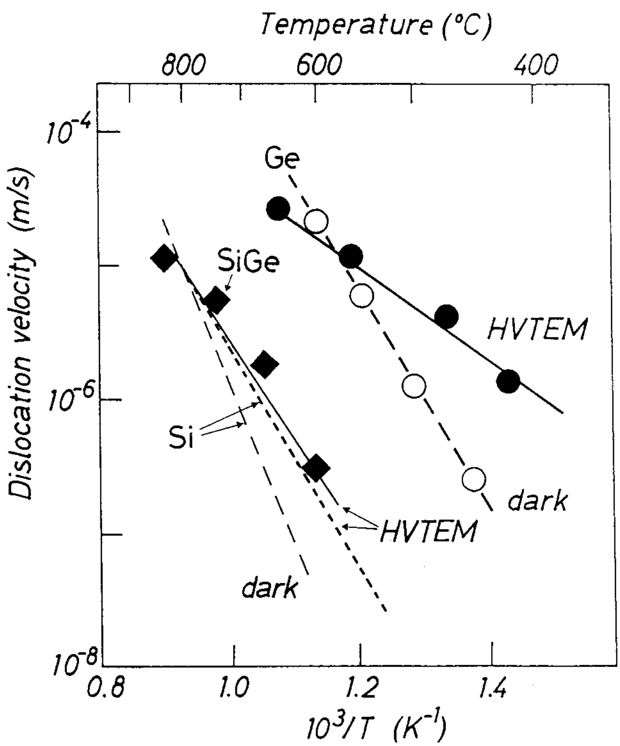

Fig. 3. Arrhenius plot of the velocities of 60° dislocations in Ge and SiGe during the *in-situ* observation with the electron beam current of 10 μA estimated for the shear stress of $\tau = 20$ MPa. The velocities of 60° dislocations in Ge [10] and Si [12] in the dark obtained by the etch pit technique and by *in-situ* X-ray topography and those in Si [2] during *in-situ* observation by HVTEM are included for the sake of comparison

is given in Fig. 3. For the sake of comparison, the velocities of 60° dislocations are included for the same stress in Ge [10] and Si [12] measured in the dark by the etch pit technique and by *in-situ* X-ray topography, respectively, together with those obtained in Si under carrier injection by using the HVTEM [2]. Based on previous experiments using the melting point of metals deposited on TEM samples used in the HVTEM, the temperature increase for the conditions used here can be estimated to be not more than 20 K. Moreover, a change in sample temperature results in a shift, but not a change of the slope of the Arrhenius plot. The dislocation velocities in Ge are clearly found to be enhanced especially at low temperatures by carrier injection during the *in-situ* observation. It is known that the dislocation velocities in SiGe are slightly higher than those in Si under electron irradiation. These results can be interpreted by REDM caused by carrier injection during *in-situ* HVTEM observation.

The velocity v of 60° dislocations in Ge and SiGe is well expressed by an empirical equation as a function of the stress τ and the temperature T:

$$v = v_0(\tau/\tau_0)^m \exp\left(-Q/k_B T\right), \quad (1)$$

where k_B is the Boltzmann constant and τ_0 is 1 MPa. The values of m and Q in Ge and SiGe, determined experimentally during the *in-situ* HVTEM observation, are given in Table 1 together with those in pure Ge and Si in the dark.

The activation energy Q of glide of 60° dislocations in Ge during the *in-situ* observation is obtained to be only 0.7 eV, much lower than that in the dark [10]. On the other hand, the value of Q in Si–5 at% Ge is 1.5 eV, similar to that in Si reported under electron irradiation [2, 4] or illumination [3]. This may be attributed to the rather small Ge content in the alloy investigated.

4. Discussion

The elementary process of dislocation glide in a semiconductor consists of the thermally activated nucleation of a double kink on a straight dislocation line lying along the Peierls valley and the subsequent expansion of the generated kink pair along the dislocation line across the Peierls barrier of second kind [9]. Thus, the activation energy Q for dislocation glide is the sum of single or double kink formation energy, depending on the length regime, and kink migration energy. Kolar et al. [13] have determined the magnitudes of the single kink formation and migration energies to be 0.73 and 1.24 eV, respectively, by means of *in-situ* high resolution electron microscopy of moving partial dislocations at

Table 1
Values of m and Q for 60° dislocations in Ge and Si–5 at% Ge crystals experimentally determined during the *in-situ* HVTEM observation together with those in pure Ge [10] and Si [12] in the dark

crystal	*in-situ* HVTEM obs.		in the dark		estimated W_m (eV)
	m	Q (eV)	m	Q (eV)	
Ge	≈ 1.5	0.7	1.7	1.7	0.7
Si–5 at% Ge	≈ 1.6	1.5		2.3	1.5
Si	≈ 1.6	1.6	1.0	2.2	1.6

600 °C in high purity FZ-Si and suggested that the elementary process of dislocation motion is controlled by the process of kink migration. However, it remains uncertain whether the kink nucleation, kink migration, or kink obstacles control the dislocation motion in semiconductors.

REDM is thought to be caused by nonradiative carrier recombination at dislocations. The energy released by the recombination is converted to a lattice vibration and can help either double kink formation or kink migration in the dislocation glide process. The injected high density of carriers as in the HVTEM seems to attribute mainly to recombination-assisted kink formation rather than to recombination-enhanced kink migration: One new double kink moves the whole dislocation segment by one step, whereas one kink migration step moves only the kink by one step [2]. Indeed, wavy partial dislocations with a high density of geometrical kinks and double kink pairs have recently been reported to exist in Ge induced by the excitation with a concentrated electron beam during HRTEM observation [5].

Using the model discussed above, a large number of kinks seems to be generated continuously and leads to the enhancement of dislocation motion by beam injection during the *in-situ* HVTEM observation. Thus, the reduction of the activation energy of dislocation motion during the *in-situ* HVTEM observation can be understood as a reduction of the kink formation energy E_k and the observed activation energy Q for glide may be comparable with the kink migration energy W_m in the Ge crystal and the SiGe alloy. As seen in Table 1, the estimated kink migration energy of 0.7 eV in Ge is considerably smaller than that of 1.6 eV in Si [2] and Jendrich-Haasen's deduction [14] of 1.11 eV from internal friction measurements. However, the magnitude is well comparable to the kink migration energy of 0.75 eV by Inoue et al. [5] on wavy partial dislocations with a high density of geometrical kinks [5] and to the kink migration energy of 0.8 to 0.9 eV by Louchet et al. [15] on *in-situ* HVTEM observation. On the other hand, the kink migration energy in SiGe is estimated to be 1.5 eV, comparable to that in Si, possibly due to the rather small Ge content in the alloy investigated.

In comparison with the results obtained by measurements of bulk samples without carrier injection we find a change of slope of 1 eV in Ge which is larger than the bandgap of Ge. This cannot be explained by the recombination of thermalized carriers which cannot release more energy than the bandgap. A possible explanation might be the fact that the high-energy beam produces very highly energetic "hot" carriers. If such hot carriers interact with dislocations before thermalization, they can well release energies above the bandgap of the semiconductor sample and thus e.g. facilitate kink formation. Hot carrier trapping at deep level defects has been proposed already for the EL2 defect in GaAs [16]. However, it is obvious that this interesting question requires further studies.

5. Conclusion

In-situ straining experiments on Ge and Si-rich SiGe alloy crystals in a high voltage transmission electron microscope proved the dislocation mobility to be enhanced during the observation. The effect is attributed to the nonradiative recombination of a high density of electron–hole pairs excited by the electron beam at dislocations, which seems to result mainly in the recombination-assisted kink formation. From the reduction in the activation energy of the dislocation motion, the kink migration energy is estimated to be 0.7 eV in Ge, smaller than that in Si, whereas the kink migration energy is 1.5 eV in SiGe, comparable to that in Si.

Acknowledgement I. Yonenaga thanks the Murata Science Foundation for the financial support of his visiting work in Halle. The stay of E. R. Weber in Halle was supported by a US senior scientist award of the Alexander von Humboldt Foundation.

References

[1] K. MAEDA and S. TAKEUCHI, in: Dislocation in Solids, Vol. 10, Eds. F. R. N. NABARRO and M. S. DUESBERY, North-Holland Publ. Co., Amsterdam 1996 (p. 443).
[2] M. WERNER, E. R. WEBER, M. BARTSCH, and U. MESSERSCHMIDT, phys. stat. sol. (a) **150**, 337 (1995).
[3] K. H. KÜSTERS and H. ALEXANDER, Physica **116B**, 594 (1983).
[4] N. MAEDA, K. KIMURA, and S. TAKEUCHI, Bull. Acad. Sci. USSR, Ser. Phys. **51**, 93 (1987).
[5] M. INOUE, K. SUZUKI, H. AMASUGA, Y. MERA, and K. MAEDA, J. Appl. Phys. **83**, 1953 (1998).
[6] U. MESSERSCHMIDT and M. BARTSCH, Ultramicroscopy **56**, 163 (1994).
[7] M. WERNER, M. BARTSCH, U. MESSERSCHMIDT, and D. BAITHER, phys. stat. sol. (a) **146**, 133 (1994).
[8] U. MESSERSCHMIDT and F. APPEL, Kristall und Technik **14**, 1331 (1979).
[9] J. P. HIRTH and J. LOTHE, Theory of Dislocations, Wiley-Interscience, New York 1982.
[10] I. YONENAGA and K. SUMINO, Appl. Phys. Lett. **69**, 1264 (1996).
[11] I. YONENAGA, unpublished result.
[12] M. IMAI and K. SUMINO, Phil. Mag. A **47**, 599 (1983).
[13] H. R. KOLAR, J. C. H. SPENCE, and H. ALEXANDER, Phys. Rev. Lett. **77**, 4031 (1996).
[14] U. JENDRICH and P. HAASEN, phys. stat. sol. (a) **108**, 553 (1988).
[15] F. LOUCHET, D. COCHET MUCHY, Y. BRECHET, and J. PELISSIER, Phil. Mag. A **57**, 327 (1988).
[16] V. YA. PRINZ and S. N. RECHKUNOV, phys. stat. sol. (b) **118**, 159 (1983).