Magnetization curves of plastically deformed Fe metals and alloys

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The hysteresis loop of plastically deformed Fe metals and alloys was measured for single crystals, polycrystalline iron, and A533B steel samples, and the dislocation density and structure of these samples were observed by electron microscopy. The relation between structure-sensitive properties and applied stress σ was analyzed in connection with grain boundaries and dislocations. The coercive force H_c increases with σ and is represented by one curve in all the samples. The magnetic susceptibility χ_c above the coercive field can be described by a simple relation to the magnetic field H . $\chi_c = c/H^3$ in a limited region of H . The parameter c depends only on lattice defects such as dislocations and the grain size and has a simple relation to them, but is independent of the kinds of samples or of the process of plastic deformation. The susceptibility χ_c is explained in comparison with the susceptibility χ_r in the range of approach to saturation. The magnetic properties c and H_c are discussed considering the dislocation density and its distribution including the effect of grain boundaries. These magnetic properties are useful for nondestructive tests of metal fatigue. © 2000 American Institute of Physics. [S0021-8979(00)01102-6]

I. INTRODUCTION

The currently employed nondestructive tests for metal fatigue are useful in the investigation of cracks and fractures in the materials, such cracks being produced by the pile of dislocations, where the dislocation density ρ is maximum locally. The work-hardening process has long been studied from the viewpoint of dislocation theory and we can obtain detailed information on dislocations as related to metal fatigue. However, few investigators have conducted nondestructive tests before the appearance of cracks. Nondestructive tests could be improved remarkably if tests for dislocations were established. Dislocations are generally observed by electron microscopy, but this method is not nondestructive. We thus propose obtaining information on dislocations nondestructively before the nucleation of cracks takes place. With an indirect method, it would be possible to use the physical property which is most sensitive to dislocations.

The most sensitive properties are the magnetic ones. The relation between magnetism and dislocations has been long investigated. The most famous magnetic phenomenon due to dislocations is the roll-induced magnetic anisotropy in permalloy.¹ Two different origins of induced anisotropy have been postulated; in the first, such anisotropy is thought to be caused by magnetoelastic coupling energy due to the strain field around the dislocations,^{2,3} and in the second, the cause is conjectured to be the dipole-dipole interaction due to the atomic arrangement in the APB, the so-called atom pair model.⁴⁻⁷ These two models have progressed independently; the former model has been developed and systematized by the Stuttgart group of the Max-Planck Institute,^{3,8-11} and the latter has been extended to the magnetic transition due to plastic deformation by one of the present authors and his

group. It has been concluded that both models are simultaneously realized in the Ni₃Fe-ordered alloy.¹²

Depending on the materials, there are three mechanisms by which the magnetic properties change due to plastic deformation. The first mechanism is found in ferromagnetic materials such as Fe, Ni, and Co metals.⁸⁻¹⁰ The second is applied to the magnetic transition due to plastic deformation in the intermetallic compounds and ordered alloys.¹³ The Ni₃Al intermetallic compound is ferromagnetic, which changes to paramagnetic by plastic deformation.^{14,15} Another example is the Fe-Al intermetallic compound which is paramagnetic, but becomes ferromagnetic by plastic deformations.¹⁶ This mechanism has been investigated theoretically and experimentally.¹⁷ The third mechanism is phase transformation due to plastic deformation as shown in SUS 304 stainless steel; the structure of stainless steel changes from fcc to bcc by martensitic transformation. This mechanism has been explained by Olson and Cohen and many other investigators.¹⁸⁻²¹ In the second and third types, magnetic properties such as spontaneous magnetization M_s and Curie temperature change drastically due to plastic deformation. Only the first mechanism is considered in the present study.

The mechanism caused by the strain field due to dislocations has an influence on the magnetization curve through magnetoelastic coupling, i.e., the strain field of dislocations exerts an influence on the spin distribution through the magnetostriction. The structure-sensitive properties change due to plastic deformation in ferromagnetic materials: the initial susceptibility decreases and the coercive force H_c and the reversible susceptibility χ_r increase with plastic deformation. H_c increases by about 100 times due to plastic deformation. These phenomena were studied theoretically and experimentally considering the dislocation structure elucidated by Seeger and Kronmüller and their co-workers 3 decades ago.^{3,9,10} These structure-sensitive properties are highly

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TABLE I. Composition of A533B steel.

A533B	C	Si	Mn	P	S	Cu	Ni	Mo	Al
wt %	0.18	0.15	1.50	0.004	0.001	0.03	0.66	0.56	0.01

related to the dislocation density and its distribution. The above-mentioned researchers were mainly interested in the academic aspects of plastic deformation and concentrated on the investigation of ideal cases using single crystals. However, investigation of the relationship of these properties with dislocation density and its distribution in the case of practical materials is necessary in nondestructive tests for dislocations. One of the purposes in the present study is to apply the previous results to practical ferromagnetic materials.

One of the important and urgent areas for the application of nondestructive tests is the nuclear power station, especially the pressure vessel. The deterioration of the pressure vessel determines the lifetime of the nuclear reactor. Since most pressure vessels are made of A533B steel, in the present study the structure-sensitive properties are investigated in connection with the plastic deformation in A533B steel as well as that in standardized samples such as Fe single crystals and polycrystalline specimens.

The coercive force is a good parameter to estimate the value of ρ ; H_c increases with the square root of ρ .^{8,9} This being confirmed for single crystals with tensile tests. Fatigue tests were conducted for Fe single crystals by measurement of the structure-sensitive properties, in which the value of H_c becomes saturated in the advanced stage.¹¹ The high-field susceptibility χ_r is more sensitive to the dislocation density and its distribution than H_c , but the value is difficult to obtain experimentally. There are two difficulties in the application of nondestructive tests: demagnetization and the necessity of a high external field. The main purpose of the present study is to find a parameter which is more sensitive to ρ than H_c and easier to measure than the susceptibility χ_r . This parameter, as well as H_c and χ_r , should be universal and be applicable to other materials such as SA508 steel, which is the other component material of the pressure vessel.

Experimental study on the relation between the structure-sensitive properties and dislocations has been conducted in single crystals. The investigation of grain boundaries is essential for the application of previous findings to the practical materials. A purpose of the present study is to investigate the grain-boundary effect on the structure-sensitive properties.

II. EXPERIMENTAL PROCEDURE

We prepared three sheets of Fe metals and alloys: single crystals and polycrystalline samples of pure iron (99.99%), and A533B steel. The compositions of A533B steel are shown in Table I. A533B steel is a manganese-molybdenum steel modified by increased nickel and reduced chromium content. During the manufacturing process it is austenitized between 860 and 890 °C. This is followed by water quenching and tempering processes between 650 and 665 °C for

FIG. 1. Shape and size of the samples. (a) Sample size of A533B steel using the tensile test and the fatigue test. (b) Sample size of polycrystalline and A533B steel for the hysteresis loop measurement. (c) Sample size of single crystals.

139 min. The grain sizes of A533B steel and polycrystalline samples as determined by metallurgical microscopic observation are about 10 and 40 μm , respectively. Each sample was tensile deformed at room temperature by an Instron-type testing machine. The tensile direction of Fe single crystals is [100]. The single crystals were cut into picture frame samples with [100] and [010] frames by spark cutting for the magnetic measurement. The other samples were shaped into a doughnut form. The shapes and the sizes are shown in Fig. 1. The magnetization curves are measured using a fluxmeter at room temperature, the magnetizing and detecting coils on these samples being wound with 80 and 100 turns, respectively. It was ascertained that the measured results were independent of the number of coil turns between 50 and 150 turns. The dislocations in each sample are observed by electron microscope.

III. STRUCTURE-SENSITIVE PROPERTIES AND THE LATTICE DEFECTS

Structure-sensitive magnetic properties are influenced by dislocations and grain boundaries. The effect of dislocations has been studied theoretically and experimentally^{3,9} and the results are briefly introduced above. The effect of grain boundaries has been studied experimentally and H_c was found to increase with the decrease of the grain size.²²

A. Coercive force and dislocations

The magnetization process proceeds by domain-wall movement in a single crystal. The coercive force is given by

$$H_c = \frac{1}{2M_s F \cos \phi} \left(\frac{\partial E_w}{\partial x} \right)_{\max} \quad (1)$$

where F is the total area of the domain wall in the unit volume, ϕ is the angle between the normal of the Bloch wall and H , and E_w is the domain-wall energy depending on the position and results from the strain field of lattice defects. In a previous study, H_c is obtained in a single magnetic do-

main; the dislocations exert a force on the domain wall and play the role of an obstacle to the 180° Bloch wall movement. This model can be applied to single crystals and polycrystalline samples with large grain sizes. $(\partial E_w / \partial x)_{\max}$ is calculated and found to be proportional to $\sqrt{\rho}$.^{9,10} H_c depends on the distribution of dislocations. The value of H_c in making a group consisting of n dislocations becomes $\sqrt{n}$ times larger than the homogeneous distribution.¹⁰

The initial susceptibility decreases according to the inverse of the square root of ρ . The validity of these relations was confirmed for single crystals of Ni, Fe, and Co metals.

B. Magnetic susceptibility and dislocations

The magnetic moments are deflected in the vicinity of a dislocation line from the direction of easy magnetization of the magnetocrystalline anisotropy; dislocations contain the strain field in themselves and this influences the direction of magnetic moments through magnetoelastic coupling. In other words, a magnetic anisotropy exists in the vicinity of the dislocations. The magnetic anisotropy energy is a function of ρ , the type of dislocations, and their structure in general, i.e., $E_\sigma(\rho, \xi, \beta_i, \theta)$; ξ is a parameter representing the dislocation distribution, β_i is the angle representing the dislocation type, and θ is the angle between H and M_s .

The other property related to dislocations is magnetic susceptibility. When the deviation of the magnetic moment from the external field is small or the external field is high, the magnetization is given by

$$M = M_s \cos \theta \approx M_s \left(1 - \frac{\theta^2}{2} \right), \quad (2)$$

and θ is given by

$$\theta \approx \frac{1}{HM_s} \frac{\partial E_\sigma}{\partial \theta}. \quad (3)$$

Then, Eq. (2) becomes

$$M = M_s \left(1 - \frac{c}{2H^2} \right), \quad (4)$$

where c is given by

$$c = \frac{1}{M_s^2} \left(\frac{\partial E_\sigma}{\partial \theta} \right)^2. \quad (5)$$

The magnetic susceptibility can be written as

$$\chi_c = \frac{\partial M}{\partial H} = \frac{c}{H^3}. \quad (6a)$$

The magnetic susceptibility is measured in the low-field range just above the coercive field H_c . The relation between χ_c and H is confirmed and the parameter c is obtained as a function of the applied stress. The physical meaning of c will be explained in detail later. This experimental finding shows that c has an intimate relation with the strain field around the dislocations. The relation of Eq. (6a) was also ascertained in the range of ferromagnetic saturation field by the Stuttgart group.^{3,9} We introduce the susceptibility χ_r in the range of

saturation field to distinguish it from χ_c . The introduction of χ_r helps us to understand the meaning of c . Equation (6a) is rewritten as

$$\chi_r = \frac{\partial M}{\partial H} = \frac{c'}{H^3}. \quad (6b)$$

The parameter c' has been calculated in simple cases.⁹ The magnetic moments in the vicinity of the dislocations are deflected within the exchange length. The anisotropy due to dislocations increases proportional to ρ . The parameter c' in Eq. (6b) is given by⁹

$$c' = \frac{9}{8} F(\beta_i) \frac{G^2 b^2 \lambda_s^2 n \rho}{4 \pi M_s} \left(\ln \frac{R'_o}{l_H} - 2.00 \right), \quad (7)$$

where G is the shear modulus, $\mathbf{b}$ is the Burgers vector, λ_s is the magnetostriction constant, $R'_o = 1/(2\sqrt{\rho})$ is the distance within which the long-range stress field is shielded, and l_H is the exchange length of the magnetic field. $F(\beta_i)$ is the anisotropy function representing the dislocation character.⁹ The value of χ_r depends on the distribution of the dislocations. The value of χ_r in making a group consisting of n dislocations becomes n times larger in comparison to the homogeneous distribution.

C. Grain boundary and coercive force

To date, experimental investigations between dislocations and magnetic properties have considered the ideal case of single crystals. However, the effect of grain boundaries on magnetic properties should be studied so that the above results can be applied to nondestructive tests.

The easy direction of magnetization is discontinuous in the grain boundary. The magnetic moments are deflected near the grain boundary against the exchange interaction. The grain boundary acts as an obstacle to the Bloch wall movement. Domain-wall energy E_w is influenced by the different atomic structure of the grain boundary. H_c and the initial susceptibility depend on the grain size. We can separate the effect of grain boundaries into two cases: (i) The grain size is sufficiently large enough for the Bloch wall to move in the grain. (ii) The grain size is equivalent to or smaller than the thickness of the Bloch wall and Bloch walls do not exist in the grain. The mechanism of the effect is different in the two cases. The value of H_c becomes very large in the latter case and cannot be explained by the present experiments. We deal with the former case in the present study.

The deflection of moments exists along the grain boundary at the triple point where three grains contact and at the other multiple point where more than four grains contact each other. The Bloch wall receives the forces of f_a , f_t , and f_m in the grain boundary, the triple points, and the multiple point, respectively. These forces depend on the relative crystal orientation of the grains. H_c is given by Eq. (1). The maximum force of the domain wall by the grain boundaries is given by

$$\left(\frac{\partial E_w}{\partial x} \right)_{\max} = \frac{k_1 \langle f_a \rangle}{a^2 d} + \frac{k_2 \langle f_t \rangle}{a d^2} + \frac{k_3 \langle f_m \rangle}{d^3}, \quad (8)$$

FIG. 2. Stress-strain curves of single-crystal samples. Samples were deformed by the stress indicated by the symbol ▲.

FIG. 3. Stress-strain curves of polycrystalline samples. Samples were deformed by the stress indicated by the symbol ●.

FIG. 4. Stress-strain curves of A533B steel. Samples were deformed by the stress indicated by the symbol ◆.

FIG. 5. Electron micrograph of the single-crystal sample with an applied stress of 110 MPa.

where d is the mean diameter of the grain, k_i are constants depending on the shape of the grain boundary, and a is the lattice constant. The second and third term are negligible compared with the first one and the coercive force is given by

$$H_c = \frac{k_1 \langle f_a \rangle}{2 M_s a^2 F \cos \phi} \frac{1}{d}. \quad (9)$$

H_c is inversely proportional to d . This relation agrees well with the experimental result in α -Fe metal.²²

In the polycrystalline samples, dislocations and grain boundaries would be considered in the nondestructive tests. The problem of which effect is more dominant, d or ρ , becomes important in the polycrystalline samples.

IV. EXPERIMENTAL RESULTS

The value of ρ as shown by electron microscopy is the local value and contains microscopic information on the crystal. This value, however, contains large measurement errors compared with the value of σ . The internal stress and the magnetic properties give macroscopic information on dislocations. It is reasonable to first obtain the relationship between magnetic properties and σ and then to compare it with that shown by electron microscopy.

A. Stress-strain curves

The stress-strain curves of Fe single crystals and polycrystalline samples are shown in Figs. 2 and 3, respectively. The strain rate is 1.5%/min in a single crystal and 1.2%/min in polycrystalline samples. The solid triangles and circles represent the maximum stress point of each sample; five polycrystalline samples with different applied stresses were prepared. The critical shear stress in single crystals is smaller than that of polycrystalline samples. The stress-strain curves are the same as those of many previous studies.

The stress-strain curve of A533B steel is quite different from those of single and polycrystalline samples, as shown in Fig. 4. The strain rate is 1.2%/min. The critical shear stress is about 550 MPa, three times greater than that of the polycrystalline samples. The applied stress of 462 MPa in A533B steel is within the elastic region. The large difference of the stress-strain curves between A533B steel and poly-

FIG. 6. Electron micrograph of the polycrystalline sample with an applied stress of 275 MPa.

FIG. 7. Electron micrograph of an A533B steel sample before plastic deformation.

FIG. 8. Electron micrograph of an A533B steel sample with the applied stress of 663 MPa.

crystalline samples is caused by the dislocation density in their initial state, as shown in the next section.

B. Electron microscopy observation

Dislocations in Fe metals have long been investigated by many investigators. The present observation is similar to the previous ones. A few typical micrographs of dislocations are given here. Figures 5 and 6 are the electron micrographs of a single crystal sample and a polycrystalline sample with applied stress of 110 and 275 MPa, respectively. Dislocations are distributed homogeneously in both samples.

Figures 7 and 8 are electron micrographs of A533B steel before plastic deformation and with applied stress of 663 MPa, respectively. The A533B steel contains a high value of ρ initially, in which dislocations are homogeneously distributed and form a group with the same sign of the Burgers vector. In the sample with 663 MPa stress, above which the sample was broken, dislocations form a cell structure or a band structure and their density increases remarkably.

The value of ρ was measured by counting the intersection of the dislocation lines with straight lines drawn at random on photographic films. The observed value of ρ may include an error of about 30% due to the uncertainty of the foil thickness and the diffraction conditions. The dependence of ρ on σ is represented in Fig. 9. A533B steel contains a high value of ρ even before plastic deformation. The value of ρ is $3 \times 10^{10} \text{ cm}^{-2}$. The relation between ρ and σ is not smoothly connected between single crystals and the polycrystalline group, due to the grain boundary. The grain boundary contributes to the increase of σ in plastic deformation. The differences between the polycrystalline samples and A533B steel can be attributed to the grain size.

C. Hysteresis loops

Figures 10 and 11 are the hysteresis loops of the single crystals plastically deformed by 0, 55, and 115 MPa, and A533B samples with plastic deformation by 0, 550, and 663 MPa, respectively. The coercive forces in these A533B steel samples are 7.92, 11.41, and 13.10 Oe, respectively. The magnetic susceptibility χ_c of the single crystal with 115 MPa applied stress and the A533B steel sample with 663 MPa of applied stress is plotted against H above H_c in Figs.

FIG. 9. Dependence of dislocation density ρ on the stress σ in single crystal ($\blacktriangle$), polycrystalline ($\bullet$), and A533B ($\blacklozenge$) steel samples.

FIG. 10. Hysteresis loops of single-crystal samples with 55 and 115 MPa of applied stress and without plastic deformation.

FIG. 11. Hysteresis loops of the A533B steel samples with 550 and 663 MPa of applied stress and without plastic deformation.

FIG. 12. Dependence of the magnetic susceptibility χ_c on the magnetic field above the coercive field H_c in a single-crystal sample with an applied stress by 115 MPa.

FIG. 13. Dependence of the magnetic susceptibility χ_c on the magnetic field above the coercive field H_c in the A533B steel sample with an applied stress by 663 MPa.

FIG. 14. Dependence of the magnetic susceptibility χ_c on the magnetic field above the coercive field H_c in a single-crystal sample with plastic deformation by 115 MPa. Both are plotted by the logarithmic values. The slope of the straight line -3 indicates the validity of Eq. (6a).

12 and 13, respectively. The same property of the single crystal with 115 MPa applied stress is plotted in logarithm values to test the relation of Eq. (6a) in Fig. 14 and that of A533B steel with 663 MPa in Fig. 15. The relation of Eq. (6a) is valid in the limited region, $p_1 H_c < H < p_2 H_c$, where p_1 and p_2 are parameters representing the limited magnetic field. The values of p_1 , p_2 , and H_c are shown in Table II. In the field smaller than $p_1 H_c$, the relation of Eq. (3) is not valid because the approximation, $\sin \theta \approx \theta$, is not obeyed. Above $p_2 H_c$, χ_c decreases rapidly with the increase of H and the relation cannot be verified in the present measurement. The relation of Eq. (6a) is found in all the samples. The parameter c is obtained for each sample and is shown in Table II. The smallest value of c is $5 \times 10^3 \text{ G Oe}^2$ in the undeformed sample of single crystals and its value is $1.5 \times 10^7 \text{ G Oe}^2$ in the A533B sample with an applied stress of 680 MPa.

The coercive force increases with the increase of σ as shown in Fig. 16. H_c values are arranged in one curve without a clear discontinuity between the different kinds of samples. A discontinuity, however, is observed in the slope of the curve between the single crystal and the polycrystalline samples, but no discontinuity in the slope is observed between the polycrystalline and A533B steel samples.

The dependence of c on σ is represented in Fig. 17. The parameter c is plotted on a logarithmic scale. The relation between c and σ can be expressed by one straight line with small scattering, which means that c depends on only σ and is independent of the kind of sample. Internal stress is caused by the grain boundary as well as by the dislocations. Both the dislocations and the grain boundary determine the value of c .

V. DISCUSSION

Shear stress σ has a simple relation with the dislocation density and its distribution in a single crystal: i.e., $\rho \propto \sigma - \sigma_0$ in stage I where σ_0 is the critical shear stress, and $\rho \propto \sigma^2$ in stage II. In the polycrystalline samples, however, the yield shear stress τ depends on the grain size as well as on the dislocation density, i.e., the Hall-Petch relation, $\tau = \tau_0 + k_y / \sqrt{d}$, where τ_0 is the yield shear stress of a single crystal

FIG. 15. Dependence of the magnetic susceptibility χ_c on the magnetic field above the coercive field H_c in an A533B steel sample with plastic deformation by 663 MPa. Both are plotted by the logarithmic values. The slope of the straight line -3 indicates the validity of Eq. (6a).

TABLE II. Coercive force H_c and the parameter c in Fe single crystals, Fe polycrystalline samples, and A533B steel. The magnetic susceptibility χ_c is measured in the applied field H , between p_1H_c and p_2H_c . The parameter c has a relation with H : $\chi_c = c/H^3$. Fatigue tests were also performed in A533B steel.

	σ (MPa)	H_c (Oe)	p_1	p_2	c (G Oe ²)
Single crystals	0	0.22	4.98	5.57	5.01×10^3
	55	0.33	4.51	5.35	1.00×10^4
	90	0.46	3.47	4.80	1.58×10^4
	100	0.75	2.92	4.10	3.16×10^4
	110	0.96	2.16	3.91	3.98×10^4
	115	0.97	2.21	3.89	4.47×10^4
Poly-crystalline samples	0	0.68	3.94	5.54	1.26×10^4
	168	1.06	2.78	4.63	3.98×10^4
	242	2.58	1.50	2.35	1.58×10^5
	272	2.99	1.54	2.28	1.58×10^5
	292	3.30	1.62	2.31	2.00×10^5
	293	3.38	1.78	2.40	2.51×10^5
A533B steel	297	3.67	1.74	2.34	2.82×10^5
	0	7.92	1.60	4.04	1.60×10^6
	462	8.27	1.57	3.87	1.70×10^6
	550	11.41	2.80	4.38	7.10×10^6
	580	13.10	3.05	4.81	1.40×10^7
	663	13.10	3.05	4.81	1.50×10^7
	500 MPa, 50 cycles	7.57	1.48	3.73	1.78×10^6
	510 MPa, 500 cycles	9.78	1.29	6.45	7.94×10^6

and k_y is a constant. The shear stress in this case yields information on the lattice structure, including the lattice defects.

The magnetic properties change with the lattice defects, and the relation between H_c and σ is investigated in the three kinds of samples. Figure 16 reveals that a discontinuity appears in the slope of the curve between the single crystals and polycrystalline samples. The value of H_c in the polycrystalline samples, even before plastic deformation, is much larger than that of the single crystals. The initial value of ρ in the polycrystalline samples is so small that the dislocations do not contribute to the value of H_c , which is reflected mainly by the grain boundary. The grain size of the polycrystalline samples is about 50 μm , which is much smaller than the critical size of 500 μm .²² The mechanism of coercive force in the polycrystalline samples is explained by another model, namely, the rotation of magnetic moments within the grain dominates the magnetization process. The value of H_c in the single-domain model is much larger than the present

FIG. 16. Dependence of the coercive force H_c on the applied stress σ in single crystal ($\blacktriangle$), polycrystalline ($\bullet$), and A533B ($\blacklozenge$) steel samples.

results. The mechanism below the critical grain size is under consideration. The following conclusions can be drawn from the experimental results. The relation can be represented by one curve though more than two mechanisms exist depending on the grain size. The coercive force is available as a parameter to estimate the process of deterioration due to dislocations.

The physical origin of χ_r is clear according to this model; the deflection of magnetic moments around the dislocations within the exchange length is important in χ_r . The exchange lengths of stray field, the magnetocrystalline energy, and the magnetic field are given, respectively, by⁹

$$l_s = \sqrt{\frac{2A}{2\pi M_s^2}}, \quad (10)$$

FIG. 17. Dependence of the coefficient of susceptibility c on the applied stress σ in single crystal ($\blacktriangle$), polycrystalline ($\bullet$), and A533B ($\blacklozenge$) steel samples. c is plotted in the logarithmic scale.

FIG. 18. Comparison of χ_c and χ_r . (a) In the range of approach to the saturation field, the exchange length of the field l_H is comparable to the effective exchange length l_{eff} and the susceptibility χ_r of magnetic moments within l_H is measured. (b) When the applied field is small, l_H is much larger than l_{eff} and the moments near l_H contribute to the susceptibility χ_c .

$$l_K = \sqrt{\frac{A}{K_1}}, \quad (11)$$

and

$$l_H = \sqrt{\frac{2A}{HM_s}}, \quad (12)$$

where A is the exchange constant and K_1 is the first magnetocrystalline anisotropy constant. The effective exchange length l_{eff} is defined by⁹

$$\frac{1}{l_{eff}} = \frac{1}{l_s} + \frac{1}{l_K} + \frac{1}{l_H}. \quad (13)$$

The values of l_s , l_K , and l_H for α -Fe at 1000 Oe are 3.3, 20.9, and 15.3 nm, respectively. The exchange length of the field plays a role only within the law of approach to saturation at room temperature. For small fields, the magnetic moments around the dislocations are determined by l_s and l_K . The value of χ_r can be experimentally obtained in the range of approach to magnetization above 1000 Oe.

There is a difference in the applied field between χ_c and χ_r . The value of χ_c is obtained in the weak field between 2 and 50 Oe. The mechanism of χ_c is different from that of χ_r , but both are reflective of the deflection of magnetic moments around the dislocations. The strain field of a dislocation decreases inversely with the distance r from the dislocation core. The magnetic moments within the strain field of a dislocation, being far from the core, turns to the external field though the field is weak. The moments which are located at $r = l_H$ ($l_{eff} \ll l_H$) contribute to the susceptibility χ_c , as shown in Fig. 18(b). Though the susceptibility due to the moments at far distances is weak, the observed value is large, since the number of turned moments increases proportional to r^2 . The number of moments which contribute to χ_c is in proportion to $2\pi\rho r dr$ when the dislocations separate sufficiently.

The magnetization at the field where χ_c is obtained decreases with the increases of σ as shown in Figs. 10 and 11. The magnetic moments within l_{eff} do not turn to the applied field because the moments are firmly fixed by the magnetoelastic coupling energy. The applied field is too low to affect

the deflection of moments within l_{eff} . These firmly fixed moments do not contribute to the value of magnetization in low fields. The number of fixed moments is proportional to ρ . Higher applied fields are required to rotate the firmly fixed moments within l_{eff} parallel to the external field. The magnetic susceptibility χ_r can be measured in the high field, where $l_{eff} \sim l_H$ [see Fig. 18(a)]. The susceptibility χ_c has a different behavior; the moments outside and far from the region of l_{eff} give the value of χ_c . The detailed relation between χ_r and χ_c should become clear in the following.

The grain boundary may exert two influences on χ_c . The value of χ_c depends on the grain size, since the magnetic moments are deflected within the grain boundary. The other effect is an indirect influence: dislocations pile up at the grain boundary. The former effect is independent of plastic deformation but the latter depends on deformation. The latter effect is related to the distribution of dislocations. These effects of dislocations and the grain size also affect the internal stresses. So the relation between c and σ is significant in comparison with that of c and ρ . The separation of the two effects, due to dislocations and grain boundaries, would be interesting from the academic viewpoint, but is not necessary for the nondestructive tests, since both are the cause of cracks.

The mechanism of χ_c is due to the rotation of the magnetic moments and the value of c is independent of the kinds of samples; the magnetostriction constant is the same and the modulus of rigidity is nearly the same in the present samples. The relation between c and σ is sufficiently simple to be represented by a single line as shown in Fig. 17. This means that the parameter c depends only on the internal structures, such as crystal structures and lattice defects. It is independent of the external conditions such as the methods of plastic deformation, tensile deformation, compression, rolling, or fatigue tests. This is easily understood from the fact that the dislocations in A533B steel are first induced by rolling. The independence of c from the external condition is an important factor in the use of nondestructive tests.

For the purpose of confirmation, an A533B steel sample was fatigue tested under conditions of 500 MPa in 50 cycles and of 510 MPa in 500 cycles, and the values of c and ρ were obtained by the same method as that used with the previous samples. The obtained values are consistent as to the relations $\rho \sim \sigma$ in Fig. 9 and $c \sim \sigma$ in Fig. 17; the value of c is $7.94 \times 10^6 \text{ G Oe}^2$, which is between those of 550 and 580 MPa, and the value of H_c is 9.78 Oe, which is between those of 462 and 550 MPa. The value of ρ is $7 \times 10^{10} \text{ cm}^{-2}$ between the samples of 550 and 580 MPa. The relation between ρ and c is more consistent than that between ρ and H_c . One of the reasons for this is the fact that the value of c is more accurate than H_c . The inconsistency of H_c and c values is attributed to the dislocation distribution. The dislocations form the cell structure in the sample with fatigue tests of 510 MPa in 500 cycles, which is different from that of the tensile samples. The value of c would indicate a larger value in the distribution making a group than in the homogeneous distribution. In the samples with 500 MPa and 50 cycles, no difference is detected in the electron microscopy observation compared with the undeformed sample. The

value of H_c is smaller than that of the sample without deformation. The small value of H_c depends on the initial condition of the samples, since the samples of the fatigue tests were purchased from a different source. The value of c in the samples with 500 MPa and 50 cycles is larger than that without deformation. Dislocations tend to form a cell structure to a slight extent due to the fatigue tests. The small change of the dislocation distribution increases the value of c . The parameter c is very sensitive to dislocations, especially to their distribution.

The parameter c is more sensitive to lattice defects than H_c . The smallest value of c is $5 \times 10^3 \text{ G Oe}^2$ in a single crystal without plastic deformation and the largest value is $1.5 \times 10^7 \text{ G Oe}^2$ in A533B steel with shear stress of 663 MPa. The ratio of the largest value to the smallest is 3×10^3 in the parameter c . The extreme ratio in H_c is 60. The high sensitivity of c is good for application to nondestructive tests. The susceptibility χ_r is also sensitive to lattice defects such as the value of ρ . The sensitivity of χ_r is nearly the same as that of χ_c . The susceptibility χ_c has an advantage the application to nondestructive tests compared with χ_r . The applied field for the measurement is much lower in χ_c than in χ_r ; the former is 50 Oe at most and the latter is more than 1000 Oe. Both, however, have a disadvantage in that the demagnetization factor should be considered, though it is unnecessary for the coercive force. This demagnetization problem can possibly be overcome in the process of equipment development and design.

The value of ρ increases about 10 times due to the applied stress of 663 MPa in A533B steel, but the value of H_c increases only 60%, though an increase by a factor of 3 times is expected theoretically. The value of H_c reflects the largest value of the local strain field of dislocations. Accordingly, within the regions of high dislocation density, it increases only modestly, whereas in the low-density region, the density increases significantly. The same tendency was observed in a previous study,¹¹ where the value of H_c became saturated for high dislocation density. The value of c reflects the increase of ρ directly and increases as much as that of ρ . The

parameter c would be effective in the investigation of high dislocation density such as the nucleation of cracks.

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