Effects of Partial Recrystallization on High-Cycle Fatigue Deformation and Crack Generation of a Nitrogen-Strengthened 32Mn-7Cr Austenitic Steel at Liquid-Nitrogen Temperature

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To achieve higher fatigue resistance against subsurface crack generation, both the refinement of grain structure and the introduction of mobile dislocations on various slip systems have been shown to be effective in the 32Mn-7Cr austenitic steel. A novel treatment which consisted of cold grooved rolling and partial recrystallization was introduced to modify the microstructure. High-cycle fatigue properties and fatigue-crack generation were investigated for both the solution-treated (ST) and the partially recrystallized (PR) materials at 77 K. The PR material displayed higher fatigue strength than the ST material, especially in the high-cycle regime. No subsurface crack generation was detected for the PR material; however, it appeared in the lower peak stress and/or in the longer-life range for the ST material. Intergranular facets formed a subsurface crack initiation site in the ST material. Since the dislocation structure that developed in the fatigued PR material assisted homogeneous and multidirectional plastic deformation, the localized deformation and/or the stress concentration at the grain boundaries by coplanar arrays were believed to be relieved. Therefore, intergranular cracking due to incompatibility at a grain boundary may disappear.

I. INTRODUCTION

SOME high-strength alloys such as titanium alloys, nitrogen-strengthened austenitic steels, and hard steels exhibit two kinds of fatigue-crack initiation clearly at and below room temperature.^[1] The first kind occurs at the specimen surface, and the other in the interior of the specimen. Subsurface fatigue-crack initiation is dominant in the high-cycle fatigue regime and at lower temperatures, while surface cracks are initiated by high-peak-stress and low-cycle fatigue tests. The low resistance of the alloys in the high-cycle-fatigue range is related to the subsurface crack generation, especially at the positive mean stresses.^[1] The subsurface fatigue-crack initiation sites and the deformation structure were characterized for solution-treated austenitic steels in a previous study^[2] to understand the fatigue-crack generation of the alloys. The subsurface crack-initiation sites are formed intergranularly. Dislocation movements of the $\{111\}$ - $\langle 110 \rangle$ system are restricted to their slip planes, and the cross slip of dislocations is strongly suppressed. The dislocation arrangement is developed mainly by pileups or coplanar arrays, which impinge on the grain boundary and also interact with the grain boundary. Since the planar slip is blocked or strictly localized, it is considered to be a potential source of grain-boundary cracking in a specific incompatibility of the grain boundary (Figure 1). In the investigation of the subsurface crack initiation for the Ti-Fe-O alloy,^[3] the author also pointed out that a specific region with microstructural inhomogeneity provides a potential site against strain incompatibility at grain boundaries and microcrack generation. The nuclei for the microcrack generation are the locations where

an incompatibility develops between the continuous isotropic elastic-strain field and the localized planar slip in a specific inhomogeneous microstructure. The author has proposed a model of subsurface crack generation based on the local strain incompatibility related to the microstructural heterogeneity, as illustrated in Figure 1.

In the authors' previous articles,^[2,4,5] the dependence of subsurface crack size on stress range was discussed by modifying the *Kitagawa diagram*^[6] and introducing the idea of the microcrack growth. The author proposed that a microcracking and its growth generates a subsurface crack-initiation site and that the " $\Delta K_{\text{Imax}} = \text{constant}$ " relationship on the subsurface crack size basically controls the transition from stage I to stage II. This scheme suggested that the main crack (stage II) starts to propagate at any stress level when the local stress-intensity factor range exceeds a constant at around the subsurface crack (stage I). Therefore, the peak cyclic stress is lower with a larger subsurface crack-initiation site.

Since the initiation site shifts from the surface to the interior at the lower stress where it is well below the macro-yield range, the S-N curve shows a plateau range, the *knee*, as shown in Figure 2.^[1] There is a rather sharp drop in the fatigue strength at higher cycles; thereafter, the curve flattens out and reaches a fatigue limit. The drop in fatigue strength is associated with a change in the mechanism of fatigue-crack generation such as surface (curve 1) and subsurface (curve 2). If curves 1 and 2 are independent of each other, curve 2 can relatively shift to a longer life or a higher fatigue-strength region (curve 2') and is close to overlapping or hiding curve 1 (Figure 2).^[7] Therefore, a microstructural modification to prohibit the subsurface crack generation is necessary to improve the high-cycle fatigue strength. From the knowledge of strain incompatibility and the critical stage I crack size mentioned previously, this mode of failure is favored by the coarse microstructure. A fine grain structure with a randomly distributed crystal orientation may decrease the strain incompatibility at grain boundaries and increase

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Fig. 1—A model proposed subsurface crack generation in the 32Mn-7Cr ST material.

Fig. 2—Schematic diagram of the S-N curves dependent on the fatigue fracture mechanisms.

the resistance forming a critical stage I crack due to miniaturizing each microcrack. Furthermore, pre-existing movable dislocations in grains will provide the source for the rearrangement and the multiplication of dislocations. If the movable dislocations are dispersed in a grain with different slip characters, many more grains may yield locally at an applied loading condition below the macro-yield range. Thus, pre-existing movable dislocations are thought to introduce plastic deformation into a lot of grains and relieve localized

deformation. On the contrary, the pre-existing dislocations damage themselves and may decrease fatigue strength in the lower cycles (curve 1') in the macro-yield range, as illustrated in Figure 2.

Partial recrystallization of cold-worked material is a promising method to control the fine duplex structure consisting of recrystallized grains and recovered grains. However, σ -phase precipitation must be taken into consideration in the Fe-Mn-Cr austenite steel, since the annealing condition of partial recrystallization is almost the same with its aging condition.^[8] The discontinuous precipitation of the σ phase in the Fe-Mn-Cr austenite occurs at recrystallizing interfaces in the specimens subjected to heavy cold working before aging.^[9] During the aging, the recrystallizing interfaces move toward the deformed and supersaturated grains, leaving the σ phase to precipitate behind the new recrystallized grains. The size of the σ phase decreases with the degree of cold-work strain, although its resultant number density increases. Therefore, the following microstructural modeling should be used to examine the relationship between strain incompatibility and subsurface fatigue-crack generation.

1. The introduction of partial recrystallization for a heavily multiaxial cold-worked sample.
2. The development of a fine duplex structure that consists of recrystallized grains and recovered ones.
3. The involvement of movable dislocations on various kinds of slip systems in the grains.
4. The precipitation of the σ phase and its dispersion in the grains.

Since the σ -phase precipitation in austenite steels results in an embrittlement, it generally lowers crack-propagation resistance. If a generated microcrack under cyclic deformation gives a stress-intensity factor higher than the threshold value, it subsequently propagates as a fatal crack. Consequently, there will be more focus on the fatigue-crack initiation stage.

Therefore, partial recrystallization on a heavily worked 32Mn-7Cr steel and low-temperature high-cycle-fatigue characteristics of the steel are examined in the present work. The improvement of high-cycle-fatigue strength and overcoming the subsurface crack generation by the partial recrystallization are demonstrated.

II. EXPERIMENTAL PROCEDURE

A. Materials and Partial Recrystallization

A 32Mn-7Cr austenitic steel^[10] in the solution-treated condition, which was developed for the cryogenic application, was examined. The chemical compositions of the steel are listed in Table I. A cast ingot was forged at 1423 K into a 30-mm-thick plate, then solution treated at 1293 K for 7.2 ks followed by a water quench.^[2] The 30-mm-square rectangular bars, which were cut from the solution-treated (ST) material along the transverse direction, were cooled below 273 K, and multistep grooved rolling was carried out. The reduction rate was about 83 pct (true strain, $\eta = 1.8$). The cold-rolled (CR) materials were heat treated at 973 K for 3.6 ks, followed by a water quench to obtain a partially recrystallized (PR) structure. The heat-treatment condition was determined by preliminary examinations. Figure 3 shows

Table I. The Chemical Compositions of the Test Steel in Mass Percent

C	Si	Mn	P	S	Ni	Cr	Mo	Al	N	Fe
0.14	0.60	31.58	0.022	0.006	0.23	7.04	0.04	0.012	0.133	bal

Fig. 3—Relationship between annealing (heating for 1.2 ks) temperature and the Vickers hardness for cold-worked 32Mn-7Cr steel. “CR” stands for the cold-rolled ST material with true strain $\eta = 1.77$. Hardness for the PR material (heating for 3.6 ks) is also indicated with a solid plot.

the Vickers hardness (H_v) values for the ST, CR, and heat-treated (heating for 1.2 ks) materials. The condition of 973 K heating reveals an offset of hardness transition, where the H_v is decreased with an increase of the heating temperature.

B. Testing

Tensile tests were done at a strain rate of $8.3 \times 10^{-4} \text{ s}^{-1}$ using a screw-driven-type tester. Cylindrical testpieces were cut parallel to the rolling direction (RD) for the PR material; the gage geometry was 3.5 mm in diameter and 20 mm in length. The test temperatures were 77 and 293 K. The 0.2 pct proof stress ($\sigma_{0.2}$), ultimate tensile strength, (σ_B), elongation, and reduction of area were determined by the duplicate test. The V-notched Charpy impact energy of the PR material was determined from duplicate tests at 4, 77, 210, and 293 K. The V-notched Charpy-impact specimens ($10 \times 10 \times 55 \text{ mm}$, 2 mm depth notch) were also machined parallel to the RD.

Fatigue testing was carried out with the specimen immersed in liquid nitrogen (77 K). Load-control tests were conducted using a cryogenic servohydraulic fatigue-test machine. The sinusoidal waveform loading was uniaxial, with a minimum-to-maximum stress ratio (R) of 0.01. A test frequency of 10 Hz at 77 K was chosen to minimize the increase in the specimen temperature as much as possible. Figure 4 shows the configuration of the fatigue-test specimens. Un-notched hourglass-type specimens with a waist diameter of either 6 mm (specimen a) or 3 mm (specimen c) were used for fatigue

Fig. 4—Configuration of fatigue specimens in the present study: (a) and (b) ST material and (c) and (d) PR material. (a) and (c) Hourglass-type specimens are for fatigue tests, and (b) and (d) smooth type ones are for cyclic tests.

fracture testing. The longitudinal direction of the ST specimens was perpendicular to the hot-rolling direction. That of the PR specimens was parallel to the RD.

Cyclic stress-strain testing was carried out at 77 K with the smooth-type specimens for up to 2000 cycles. Smooth-type specimens, shown in Figures 4(b) and (d), were used for the ST and PR materials, respectively. Cyclic stress-strain data at shorter lifetimes were obtained in the tester cycled between fixed load limits, where the extensometer was attached onto the cylindrical part of the round bar specimen. Loop recordings were made periodically at a frequency of 0.1 Hz, although a higher frequency of up to 1 Hz was used between loop recordings. The general appearance of the loops for the loading conditions is illustrated in Figure 5.

C. Microscopy and Characterization

The microstructure and fatigue-crack-initiation sites and fracture surfaces were studied by scanning electron microscopy (SEM). A JEOL* JSM-6400 microscope equipped with a

*JEOL is a trademark of Japan Electron Optics Ltd., Tokyo.

LaB₆-type gun was used at 20 keV. Electron backscattered diffraction pattern (EBSP) analysis with SEM was performed

Fig. 5—Schematic illustration of loop configurations.

to determine the matrix structure. A data set of point analyses with every 0.2 μm beam scanning in hexagonal grids yielded image and orientation maps characterizing the matrix structure. The microstructure and dislocation structure were also studied by transmission electron microscopy (TEM). A JEOL 2000FXII transmission electron microscope equipped with a double-tilt goniometer stage and energy-dispersive X-ray spectroscopy (EDS) was employed at 200 keV. Quantitative analysis with the Cliff–Lorimer ratio was performed on thin sections for the elements of Fe, Mn, Si, and Cr. For fatigued specimens, TEM disks about 600- μm thick were sectioned from beneath the fracture surface. They were cut perpendicular to the principal stress axis and were reduced to about 100 μm in thickness by abrasion with No. 600 emery paper. Each disk diameter was measured, and its stress amplitude (σ_a) was calibrated, since the fatigue-test specimens had an hourglass shape. Both the TEM foils and EBSD samples were prepared by a conventional electrical-jet polishing technique in a stirred solution of 10 pct perchloric acid and 90 pct acetic acid at about 290 K. The nature of the dislocation was determined using the standard $g \cdot b$ analysis technique under the two-beam condition.

Phase identification by an X-ray diffractometer was made on 1-mm-thick sheet specimens cut from the ST, CR, and PR materials, using filtered $\text{Cu-K}\alpha$ radiation at 40 kV and 300 mA. The diffraction patterns were compared with the ASTM cards. The equilibrium in the pseudoternary Fe–Mn–Cr system was calculated with a thermodynamic application, ThermoCalc, for the chemical compositions of the 32 Mn–7Cr steel at 973 K.

III. RESULTS

A. Tensile Properties and Microstructure

Figure 6 shows the tensile properties and the Charpy absorbed energy (CVN) for the ST and PR materials. Stress-strain curves of the PR material at 77 K show a higher work-hardening rate than that of the ST one. The PR material exhibits a higher ultimate tensile strength and 0.2 pct proof

stress, but a lower elongation than the ST one. The PR material, especially, displays a quasi-brittle manner at 77 K, such as an extremely low reduction of area and CVN. Figure 7 shows the tensile fracture surface of the PR material. Quasi-brittle features such as microdimples and tong are observed. These features suggest poor resistance for crack propagation and the reason for the early fracture in the tensile test.

Figure 8 shows an optical micrograph in the longitudinal section of the PR material. Prior-matrix grains are elongated to the RD. In the EBSD image-quality map in Figure 9, the microstructure of the PR material is classified into two grain types, which are called the recrystallized grain and the recovered grain. Since the image quality of the EBSD reflects the perfection of a crystal, strain and defects such as dislocations cause poor quality (*i.e.*, darkness in the image-quality maps). Therefore, the image is clearer in the recrystallized grain than in the recovered one. In fact, the recovered-grain region is a darker gray than the recrystallized one shown in Figure 9. The grain diameter in the recovered-grain region is about a few microns, compared to about 10 μm in the recrystallized one. The PR material showed a $\langle 111 \rangle$ fiber texture, where the RD is parallel to $\langle 111 \rangle$. Recrystallized grains had a large-angle boundary between neighboring grains.

Figure 10 shows recovered grains of the PR material in the transverse section. Dislocations are observed in the recovered grains; however, only a few dislocations were observed in the recrystallized grains. The dislocations are not tangled, and no slip bands are observed. Dislocations are aligned mainly along the primary glide planes, but they do not have the same characteristics (Figure 10(b)). Thus, dislocations having various characteristics are dispersed in a grain. As a result of the cold rolling and annealed condition, a partially recrystallized phase structure is obtained. Figure 10 shows the distribution of the particles that are less than a few hundred nanometers in diameter in grains and grain boundaries, although only the γ phase was detected by the X-ray diffraction. The compositions of both the particle and the matrix were evaluated by EDS. The average value is plotted in the calculated pseudoternary phase diagram (Figure 11). The concentration of Cr is enriched in the particle, and its compositions are in the $\alpha + \gamma + \sigma$ -phase region at 973 K. The particles are estimated to be σ -phase precipitates.

B. Fatigue Strength and Location of Crack Initiation Site

Figure 12 shows the S–N curves of 32Mn–7Cr steel and also describes the data from a previous work.^[2] The S–N data plots in Figure 12 are distinguished between the specimen surface initiation and the subsurface initiation. The subsurface crack initiation was only detected in the ST material. In the ST material, the fatigue-crack-initiation site shifts from the specimen surface to the specimen interior at about 10^6 cycles. The sharp drop in the fatigue strength over 10^6 cycles is related to the subsurface crack-initiation failure. On the other hand, all the samples of the PR material show surface crack initiation. The fatigue strength of the PR material gradually decreases with increasing cycles to failure.

For most steels, fatigue strength is proportional to the tensile strength at room temperature.^[11] If this relationship applies at 77 K, the fatigue strength of the PR material should increase. In fact, the fatigue strength of the PR material is

Fig. 6—Tensile properties and toughness of ST and PR materials: (a) stress-strain curves at 77 K, (b) strength, (c) ductility, and (d) toughness.

Fig. 7—(a) Tensile fracture surface of the PR material at 77 K and (b) its magnified image of a tongue.

Fig. 8—Microstructure of the PR material in the longitudinal section.

Fig. 9—EBSP image quality maps in the transverse section.

superior to that of the ST one (shown in Figure 12(a)), but its tensile strength is higher. Figure 12(b) arranges the S-N data by the ratio of fatigue strength to tensile strength. The S-N curves of the PR and ST materials at 77 K appear to cross each other. The PR material shows a lower normalized fatigue strength than the ST material below 2×10^6 cycles, while the fatigue strength of the PR material is higher than that of the ST material, and the difference in the S-N curves becomes wider with increasing number of cycles to failure, N_f . Partial recrystallization can be applied to overcome the sharp drop in the fatigue strength of the ST material and improve the high-cycle fatigue strength over 5×10^6 cycles. The ratio of the 10^7 cycle fatigue strength to σ_b for the PR material is about 0.65.

Fig. 10—Bright-field images (a) of the recovered grains in the PR material and (b) of the dislocation structure in a recovered grain.

Fig. 11—A part of the calculated phase diagram for the pseudoternary Fe-Mn-Cr system at 973 K, indicating the composition of the EDS analysis results.

C. Crack-Initiation Site and Fracture Surface

The subsurface crack initiation was only detected in the ST material. No defects such as inclusions or pores were detected at the subsurface crack-initiation sites; however, facets formed an initiation site in the material. Intergranular cracking was employed to describe the fractographic features in the ST material.^[2] Figure 13 shows the subsurface crack-initiation site on the opposite fracture surfaces of the ST sample. Facets at the site match in an interlocking manner on opposite fracture

Fig. 12—(a) S-N data of ST and PR materials and (b) S-N curves normalized by tensile strength.

surfaces. The region of the subsurface crack-initiation site reaches about $300\text{ }\mu\text{m}$ in diameter. In the authors' previous article,^[2] the dependence of the subsurface crack size on the peak cyclic stress was evaluated. The present result agrees with the tendency that the peak stress is lower with increasing size of the subsurface crack-initiation site.

The SEM photographs in Figure 14 show the vicinity of the fatigue-crack-initiation site for the PR material. The fatigue crack propagates radially from the specimen surface. In the PR material, crystallographic facets and tongs were observed in the higher- ΔK region of the crack-propagation plane. These features show an embrittlement of the PR material, which exhibits low plasticity and planar-slip deformation.

IV. DISCUSSION

A. Deformation Structure

Figure 15 shows the deformation structure of a fatigued PR material. Dislocations are introduced in most of the

Fig. 13—Matching fracture surfaces and their subsurface crack initiation sites of a fatigued ST material ($\sigma_{\max} = 591\text{ MPa}$, $N_f = 15,054,500$).

Fig. 14—(a) Fracture surface and (b) surface crack initiation site of a fatigued PR material ($\sigma_{\max} = 933\text{ MPa}$, $N_f = 5,895,320$).

grains, and their densities are higher than that of the fatigued ST samples at the same stress level. During the cycling of the PR specimens, the planar structure does not undergo any change as well as the cycling of predeformed specimens.^[12] However, the coplanar structure disappears. A simple explanation is that the pre-existing dislocations in the grains multiply themselves and that the dislocation structures are rearranged through cyclic deformation. Figure 15(c) shows that primary, secondary, and/or conjugate slips are operative,

Fig. 15—(a) Bright-field image of the dislocation structure in the vicinity of $\{111\}$ twins and (b) their selected area diffraction pattern for a fatigued PR material ($\sigma_a = 353$ MPa, $N_f = 142,350$); beam direction is $[01\bar{1}]$ zone normal. (c) The dislocation structure for the same sample; beam direction is near $[111]$. Arrows (white color) indicate the σ phase.

although the dislocation tangle that developed in the tensile-fracture specimen can hardly be seen. A few slip bands are observed, but dislocations with different characteristics develop a band. Figure 16 shows a high-resolution image of a fatigued ST specimen at 4 K. The stress level of the sample is almost the same as that of the PR sample shown in Figure 15. Moiré fringe by distortion with deformation twins are seen; thus, the slip bands observed in the PR specimen are different from the microbands that developed in the ST specimens, from which the coplanar fault accompanied distortion. Piled-up dislocations impinged on a grain boundary are not clearly detected in the PR material. Consequently the plastic deformation of the PR specimens is developed much more uniformly than that of the ST one. Thus, the strain incompatibility at the grain boundary may be relieved, since the dislocation structure in the PR material assists homogeneous and multidirectional plastic deformation and results in the change of the shape of the grains.

B. Cyclic Stress-Strain Behavior

Steele and McEvily^[13] compared the cyclic stress-strain behavior of a Ti-6Al-4V alloy and that of copper and suggested that the number of mobile dislocations increased and caused a cyclic softening for the Ti-6Al-4V alloy in the high-cycle range at a stress well below the macroyield range. However, few studies have been done to clarify the relationship between the cyclic stress-strain behavior and the microstructural characteristics of subsurface crack generation. Then, the cyclic stress-strain behavior of the 32Mn-7Cr steel was examined. Figure 17(a) shows that the cyclic plastic strain, (ϵ_p) gradually increases with cycles in the ST materials. The cyclic plastic strain in the higher-stress level becomes stable after a certain number of cycles. However, for the samples showing subsurface crack generation, an increase in cyclic plastic strains up to 1000 cycles reveals a cyclic creep. Since cyclic creep occurs in the direction of

Fig. 16—High-resolution image of a fatigued ST material at 4 K ($\sigma_a = 392$ MPa, $N_f = 874,010$) shows moiré fringe by distortion with deformation twins.

increasing tensile strain, the accumulation of damage is accelerated by cyclic softening and higher mean strain.^[14] Furthermore, cyclic softening is only detected in the ST samples which showed subsurface crack generation. The plastic strain range ($\Delta\epsilon_p$) of the samples that caused subsurface crack initiation is extremely low and decreases slightly, as shown in Figure 17(b). Cyclic softening itself is not substantial for the subsurface crack generation, but becomes a specific incompatibility at the grain boundary due to a planar slip

Fig. 17—Cyclic strain at 77 K as a function of the number of cycles under the load-control mode.

in only the weak grain.^[3] On the other hand, a small plastic strain during cycling for the samples is detected macroscopically, meaning that cyclic deformation does not produce a change in the shape for the fatigue specimen and that the generated dislocation arrays are geometrically necessary (GN) dislocations.^[15]

For the PR material, the plastic strain is almost the same up to 1000 cycles. At a higher stress level generating a surface crack, both the plastic strain and the plastic strain range are only one-tenth as high as those at the lower stress level. Thus, the difference of cyclic stress-strain behavior in the PR and ST materials agrees with each dislocation structure and is consistent with the proposed model of subsurface crack generation based on the local strain incompatibility in

Figure 1, where screw dislocations on a single-slip system cause a shear-strain field against neighboring grains near the boundaries.

C. Suppressing Subsurface Microcracking and Its Growth

Freudenthal^[15] defined high-strength materials showing a deformation manner such as coplanar arrays as *pseudoelastic* materials. In contrast, *plastic* materials such as copper are defined as those in which the microslip is extensive and diffuse and where the surface distortions produced by the reversed slip are the primary nuclei of the macrocrack generation. The extrusion-intrusion (persistent slip-band) models^[16] provide a good explanation for the *plastic* materials, but not for the *pseudoelastic* materials in the high-cycle regime.

The features of the subsurface crack-initiation site shown in Figure 13 sustain the proposed model in Figure 1. The strain incompatibility at the grain boundary is believed to cause intergranular or transgranular cracking, since a coplanar dislocation configuration is developed. Intergranular fracture has been observed at low temperatures in high-Mn austenitic steels that are associated with a ductile-brittle transition^[17] and high-cycle-fatigue fracture.^[2] The steel containing Mn provides a weak link of the grain boundary against strain incompatibility and microcrack generation, although the reason for the grain-boundary embrittlement has not been clear yet. The main ideas regarding the embrittlement of the steels are summarized in References 18 and 19. Tomota and Shibuki^[18] noted that the intergranular fracture in low-temperature tensile tests is suppressed by grain refining. However, in the high-cycle fatigue, the growth or the coalescence of microcracks forms facets to adopted a critical size as a subsurface crack-initiation site for propagation. Thus, in addition to grain refinement, it is also necessary to lower the strain incompatibility to increase the resistance to microcrack growth.

The initial structure after grooved rolling in the present study is, basically, a severely deformed structure, because the applied strain level is 1.77. There are several recent works on the fatigue behavior of severely deformed materials. Most of these are on *plastic* materials such as copper. An attempt to account for the effect of grain refinement on fatigue performance to copper has been made by Mughrabi^[20] from the general standpoint of fatigue-damage phenomenology. Vinogradov and Hashimoto^[21] have also studied the fatigue behavior of ultrafine-grained materials produced by severe deformation such as equal-channel angular pressing (ECAP). A copper and aluminum alloy serves as a popular model of fcc, *wavy-slip* materials. The test material is a low-stacking-fault fcc material showing a *planar-slip* manner equal to that of titanium (hcp). It is pointed out that not only grain size, but also the slip mode, affects the cyclic response.^[21] Thus, discussion has been made to compare their results with the present ones.

The severely deformed materials show an excellent endurance limit in the high-cycle fatigue regime, but poor low-cycle fatigue properties.^[21] The present results show a similar tendency; the PR material shows an excellent fatigue strength in the high-cycle regime, but poor strength in the low-cycle regime (Figure 12(b)). The fine-grained materials are less ductile than their coarse-grained counterparts. In addition, low structural stability, susceptibility to cyclic softening, and early

strain localization have been pointed out as other reasons for this.^[21] Several causes have been proposed in Reference 20 to account for the phenomenon: (1) dislocation recovery and recovery of highly distributed grain boundaries, (2) grain coarsening, recrystallization, and abnormal grain growth promoted by cyclic stresses, and (3) strain localization in the large-scale shear bands appearing during mechanical testing and microcracking. However, shear banding and microcrack nucleation are strongly dependent on the material (particularly, on the slip mode). In the case of titanium, the defects created during fatigue on the specimen surface were associated with nonpropagating microcracks and were hardly distinguishable from the microscopic shear bands.^[21]

In this study, both the refinement in grain structure and the introduction of movable dislocations on various kinds of slip systems in grains resulted in a higher fatigue resistance to subsurface crack generation and a tendency to promote surface crack initiation as slip processes became more active at the higher stress levels. Figure 18 summarizes the results with the illustration of S-N curves.

1. As mentioned in the Introduction, the S-N curve of the ST material can be classified into the two parts that are fitted to the different mechanisms of fatigue-crack generation.
2. Grain refinement and remaining damage provided a higher tensile strength in the PR material. However, the increase of fatigue strength in the low-cycle regime is relatively low and is not proportional to the increase of tensile strength. The damage of the PR material may accelerate plastic-strain accumulation as well as strain of the severely deformed *plastic* materials.
3. Partial recrystallization was successful in lowering the maximum grain size. The dislocation structure that developed in the fatigued PR samples showed multidirectional plastic deformation and resulted in the disappearance of coplanar arrays. Since the weakest-link model is adopted for the subsurface fatigue-crack generation, grain refinement and pre-existing dislocations may assist the homogeneous deformation and relieve the strain incompatibility at the grain boundaries. Then, the subsurface crack generation due to intergranular cracking disappears, and the fatigue

strength in the high-cycle regime increases. However, no shear bands were detected. Dislocation mobility is controlled by the slip mode, and the general stability of the fine-grained structure in hcp or low-stacking-fault fcc materials is supposed to be more stable than that in *plastic* materials such as copper.

Thus, both the grain refinement and the work hardening due to the increase of the average dislocation density may commonly play an important role in the fatigue behavior of severely deformed materials. To gain a better understanding of the cyclic behavior of fine-grain-structure materials, the effects of (1) randomly distributed crystal orientation, (2) fine second-phase dispersion, and (3) lowering the local yield stress of *pseudoelastic* grains on high resistance to microcrack growth will be considered in the forthcoming communication, since the distribution of microcracks is closely related to the grain structure and/or second-phase structure in titanium alloys.^[1,5]

V. CONCLUSIONS

Partial recrystallization was conducted on the 32Mn-7Cr steel, and the low-temperature, high-cyclic-fatigue characteristics were examined to study the relationship between strain incompatibility and subsurface fatigue-crack generation. The microstructural modification to overcome subsurface crack generation was proven by the following results.

1. The PR material showed a fine duplex structure that consisted of a recrystallized grain and a recovered one. Movable dislocations were dispersed in those grains. Fine σ -phase precipitates were distributed in the grains and grain boundaries.
2. The PR material exhibited higher strength and lower ductility than the ST material. Actually, the PR material showed a quasi-brittle manner in both the tensile fracture and fatigue-crack propagation.
3. The PR material showed higher fatigue strength than the ST material, especially in the high-cycle regime. No subsurface crack generation was detected for the PR material, although it appeared in the lower peak stress and/or in the longer-life range for the ST material. Intergranular facets formed a subsurface crack-initiation site in the ST material.
4. Since the dislocation structure that developed in the fatigued PR material assisted homogeneous and multidirectional plastic deformation, the localized deformation and/or stress concentration at the grain boundaries by coplanar arrays were believed to be relieved. Therefore, the subsurface crack generation due to intergranular cracking may disappear.

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Fig. 18—Schematic S-N curves showing modification by partial recrystallization.

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