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FATIGUE CRACK FORMATION ASSOCIATED WITH CYCLIC SLIP DEFORMATION ALONG PRIOR AUSTENITE GRAIN BOUNDARIES

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INTRODUCTION

The fatigue process of high-hardened materials such as a martensitic steel has not yet been well understood in comparison with that of usual ductile materials. It is essential to study whether the crack initiation in the martensitic structure is basically due to the repetition of slip or it is due to the unknown mechanisms without slip deformation. The fatigue process of the martensitic structure is quite localized in the regions of structural discontinuity, such as the martensitic plate, packet and Prior Austenite Grain (PAG) boundaries [1 - 8]. Though much effort has been made in order to specify the characteristics of the fatigue crack initiation associated with the complicated nature of the martensitic structure, a more complete understanding of the mechanism of crack initiation is still needed.

An unusual microstructural change along the PAG boundaries due to fatigue stressing which leads to the fatal crack initiation, has been reported recently [2 - 5]. This change appears along the PAG boundaries on the electro-polished smooth surface of the specimen after the repetition of stress cycles. With further repetition of stress cycling, this change grows into a network structure on the surface of specimen. Fatigue cracks are developed in the network structure, and finally lead to the fatal growth by their coalescence as shown in Figure 1. The mechanism of the formation of this network structure is still left unsolved, though the reconfirmations [9] of such phenomenon have been made. Therefore, it is very important to understand the feature of microstructural change in relation to the fatigue process of the low carbon martensitic structure.

In this paper, an emphasis is placed upon the understanding of the nature of the fatigue process involving the unusual microstructural change mentioned above. Unless otherwise stated, the term "network structure" is referred to the unusual microstructural change due to fatigue stressing.

CHARACTERISTICS OF MICROSTRUCTURAL CHANGE DUE TO FATIGUE STRESSING

Evidences already reported with respect to the formation of the network structure are summarized as follows.

The microstructural change has been observed in the fatigue process of the martensitic steel with the carbon content ranging from 0.15 to 0.35%. The development of such microstructural change into the clear appearance of network structure in the fatigue process would be dependent on the carbon content in the material [2]. For example, the network structure of the martensitic steel having 0.15% C leads to the nucleation of the fatal

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crack, while the similar structure, which is also observed in the martensitic steel with 0.25% C does not always lead to the nucleation of the fatal crack [2]. In addition, PAG size has an influence on the easiness of the formation of the network structure in a given carbon content; for example, in 0.15% C steel specimens having various PAG sizes smaller than about 65 μm, the network structure is observed more clearly in the extent of small PAG sizes [4, 5].

The following experiments were carried out in order to obtain more fundamental evidences on the formation of the network structure. Annealed 0.20% C steel was machined into the unnotched hollow rotating bending and reversed torsional fatigue specimens which have outer diameter of 8 mm and wall thickness of 1 mm. The specimens were induction-hardened with the conditions shown in Table 1, thus martensitic structure with small PAG size of 8 μm was obtained. Chemical compositions of the material and other mechanical properties are also given in Table 1. No ferritic constituent was detected under the optical microscopic observation. The fatigue process of the specimen was obtained mainly by the optical microscope with magnification of 400. Both 98 N.m rotating bending and 98 N.m reversed torsional fatigue testing machines were employed under 50 Hz and 33 Hz, respectively.

The cyclic stress above the endurance limit was applied to this specimen and it was observed that the similar network structure as shown in Figure 1 which were followed by the fatal crack nucleation appeared on the electro-polished surface both in rotating bending and reversed torsion. For the specimen having a PAG size of 38 μm, however, the fatal crack was not nucleated in such a way, but it was nucleated within a grain as shown in Figure 2, thus the formation of the network structure leading to the fatal crack nucleation along the PAG boundaries was reconfirmed to occur in the specimen having relatively small PAG size. Furthermore, it was confirmed that the network structure which had been observed mainly on the surface of induction-hardened specimen was also observed on the specimen surface heat-treated by electric-furnace in which the distribution of carbon would be expected to be uniform. Thus, the formation of the network structure seems to be independent of the heating procedure.

NATURE OF FATIGUE PROCESS INVOLVING FORMATION OF NETWORK STRUCTURE

In order to make clear the critical stress required to form the network structure, cyclic stress below the endurance limit was applied. Figure 3 shows the appearances of the network structure under the cyclic stressing of σ = 412 MN/m^2 (42 kg/mm^2) for rotating bending and of τ = 245 MN/m^2 (25 kg/mm^2) for reversed torsion. It can be recognized that the network structure in both cases have already appeared under the stresses below the endurance limits. However, no structural changes were observed both in rotating bending and reversed torsion under the stresses of σ = 392 MN/m^2 (40 kg/mm^2) and τ = 206 MN/m^2 (21 kg/mm^2), respectively. Thus, the critical stresses required to form the network structure were estimated to be σ_c = 402 MN/m^2 (41 kg/mm^2) for rotating bending and τ_c = 216 $\sim$ 235 MN/m^2 (22 $\sim$ 24 kg/mm^2) for reversed torsion, therefore, it may be concluded that the formation of this structure is mainly governed by the cyclic shear stress because of the ratio $\tau_c/\frac{1}{2}\sigma_c$ being 1.07 $\sim$ 1.17 [11].

Since an optical microscopic observation is not sufficient to investigate the nature of the above structure in more detail, a transmission electron microscopic observation was carried out on the formation process of the

network structure under the stress below the endurance limit. Figure 4 shows an example of the electron micrographs on the specimen surface after the application of 10^7 cycles of torsional stress corresponding to the 86% of the endurance limit. The network structure can be seen to appear not on the PAG boundaries but within the narrow bands near the boundaries. Further, it can be also observed that the scratched line, which was marked on the surface with fine aluminum oxide particle before cyclic stressing, is displaced within the bands near the PAG boundary and that the direction of displacement coincides with that of the maximum shear stress in torsion. Therefore, it may be considered from the results of Figure 3 and Figure 4 that the network structure was brought about by the cyclic slip deformation in the vicinity of PAG boundaries.

In the fatigue process of ductile metals such as copper and brass, it is well known that the microscopic surface irregularity due to the cyclic stressing occurs on the specimen surface and this surface irregularity is recognized as notch and peak which is observed by taper sectioning [12 - 14]. Boettner [8] has reported that even in the tempered martensitic steel, the similar peak could be formed. If the network structure is produced by the cyclic slip deformation, the surface irregularity due to the deformation should be expected to be observed. So, the micro-surface contour on the specimen, where the network structure had clearly come out, was examined by taper sectioning method. Figure 5 shows a taper sectioned irregularities on the surface illustrating the nature of deformation due to the cyclic stressing. It can be seen that the material was extruded at a heigth of about 0.1 μm and a special interest arises from the similarity of this peak to that previously reported. Therefore, it would be concluded that the fatigue crack is caused by the ductile fatigue process which was developed preferentially along the narrow bands near the PAG boundaries.

DISCUSSIONS ON EXISTENCE OF LOW RESISTANCE REGION ALONG PAG BOUNDARIES FOR PLASTIC DEFORMATION

From the evidence that slip induced microstructural change along the PAG boundaries occurs in this specimen, it would be reasonable to expect that the PAG boundaries have a relatively low resistance region for plastic deformation than that of the transgranular region.

In order to make positively a similar low resistance region for plastic deformation along the PAG boundaries by the controlled quenching process, the delayed quenching technique was employed. The special specimens with the proeutectoid ferrite along the PAG boundaries were prepared using this delayed quenching technique whose delayed time is 10 sec. The slip deformation was observed to occur preferentially in this proeutectoid ferrite along the PAG boundaries after the fatigue stressing as shown in Figure 6. It should be noted that the appearance of the microstructural change due to slip deformation is quite similar to that of the network structure in Figure 1.

The traverse measurements of micro-Vickers hardness across the PAG boundaries for the present specimen has revealed that the region along the PAG boundaries has relatively low hardness [15].

It is natural to lead our interests into the consideration why the low resistance region are formed along the PAG boundaries. In order to understand this point, we introduce the idea that the conventional quenching operation is a kind of delayed quenching involving a very short duration of delayed time before the rapid cooling. An optical microscopic observation has revealed that a precipitation of proeutectoid ferrite occurs at the PAG boundary by the introduction of very short delayed time of 2 sec. into the present quenching procedure. No ideal quenching procedure would be expected in practical heat treatment.

Thus, we arrive at the consideration that the low resistance region along the PAG boundaries, whose formation may involve the possibility of very fine ferrite precipitation, could be expected by the conventional heat treatment.

In general, the critical condition for precipitation of ferrite would be dependent on the size of PAG and austenitizing temperature since a small PAG size and low austenitizing temperature have an advantage for easy nucleation of proeutectoid ferrite. Following this consideration, the network structure could be expected to be formed even in the microstructure having the PAG size of 38 μm (see Figure 2) by the introduction of an appropriate delayed time to the quenching operation. Figure 7 shows the evidence that selection of delayed time of 20 sec. brings about the appearance of network structure on the same specimen as that shown in Figure 2 where no network structure was recognized previously because of the absence of intentional delayed time. Figure 8 indicates that addition of more 5 sec. of delayed time to the quenching procedure in Figure 7 produces the proeutectoid ferrite visible by an optical microscope.

Thus, a reasonable explanation is given from the above consideration for the understanding of the fact that the unusual microstructural change under fatigue stressing is markedly observed in the low carbon steel specimen having small PAG size.

CONCLUSIONS

The formation process of the network structure which leads to fatigue crack initiation was studied and the following results were obtained.

1) The microstructural change in the form of a network on the structure which leads to the initiation of a micro-crack was found to be brought about by the cyclic slip deformation within the narrow bands of low resistance for plastic deformation along the PAG boundaries.
2) A reasonable concept which is able to explain the previous findings on the microstructural change was proposed by regarding the conventional quenching procedure as a kind of delayed quenching whose delayed time is extremely short.

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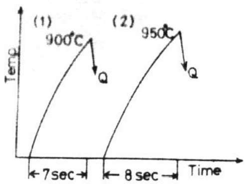
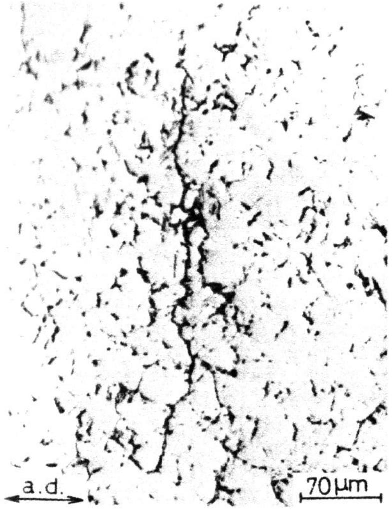
Table 1 Chemical Compositions and Mechanical Data

Material	Chemical composition (%)	Rapidly heat treatment condition	Hv (mean) (100g)	Grain size No (ASTM)	Average grain diameter (μm)	Endurance limit (MN/m²)	
						Rotating bending	Reversed torsion
S 20C	C: 0.20 Si: 0.24 Mn: 0.47 P: 0.016 S: 0.017	Pre-heat treatment 900 ℃ normarizing 1. 900℃ 7sec 2. 950℃ 8sec	474	11	8	451	412



Schematic Illustration of Repetitive Austenitizing (See Table 1)



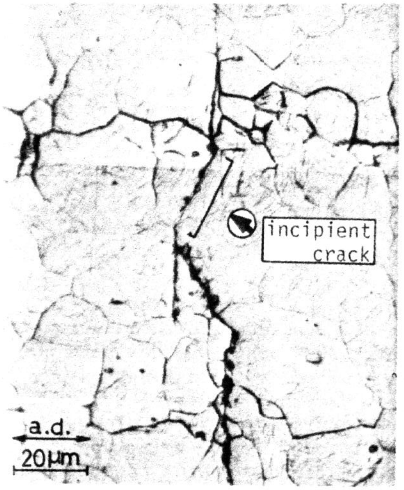
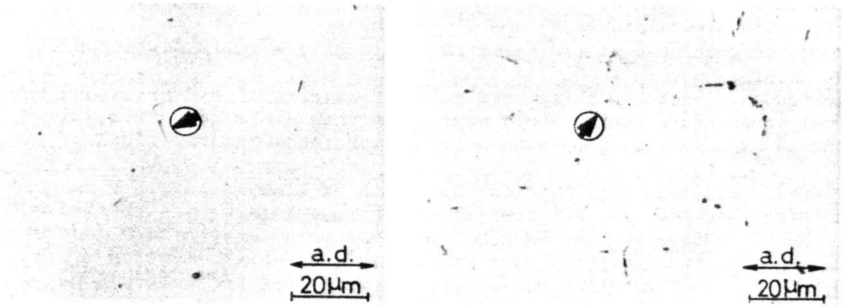


Figure 1 Characteristic Feature of Microstructural Change along PAG Boundaries followed by Fatal Crack Formation [3]. 0.15% C Steel, Rotating Bending, σ=530 MN/m^2, N=1.2x10^5

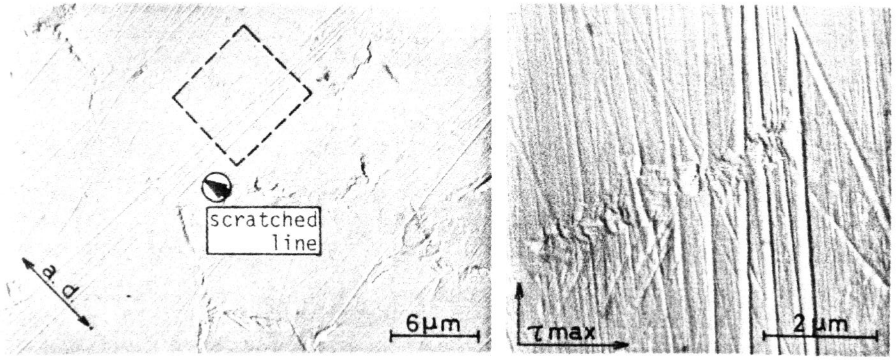
Figure 2 Formation of Transgranular Microcrack in Martensitic Structures* with PAG size of 38 µm. Rotating Bending, σ=687 MN/m^2, N=1.35x10^5.
*PAG Boundaries have been revealed by new Etchant [10] after Cyclic Stressing. Induction-Hardening; 1150° C 20 sec. Endurance Limit; σ=530 MN/m^2



Rotating Bending, σ=412 MN/m^2 (91% of the Endurance Limit), N=1x10^7

Reversed Torsion, τ=245 MN/m^2 (60% of the Endurance Limit), N=1x10^7

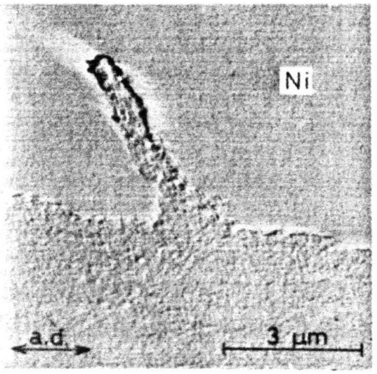
Figure 3 Appearances of Structural Change by Repetition of Stress Below Endurance Limit



Detailed Feature of Network Structure. Reversed Torsion, τ=353 MN/m^2, N=1x10^7

Enlargement of a Dotted Square in the Left Figure

Figure 4 An Example of Network Structure by TEM Observation (Two Stage Replica)



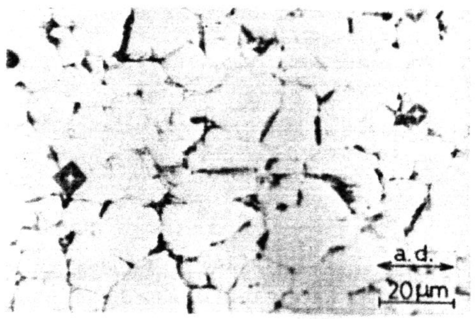


Figure 5 Electron Micrograph of a Peak Associated with Surface Irregularity due to Fatigue. Taper Magnification 45. Rotating Bending, σ=637 MN/m^2, N=6.5x10^4

Figure 6 Slip Band Formation in Proeutectoid Ferrite Along PAG Boundaries of the Specimen* Induction-Hardened with Delayed Time of 10 sec. Reversed Torsion, τ=275 MN/m^2, N=1x10^7.
*PAG Boundaries have been revealed by new etchant Prior to Cyclic Stressing

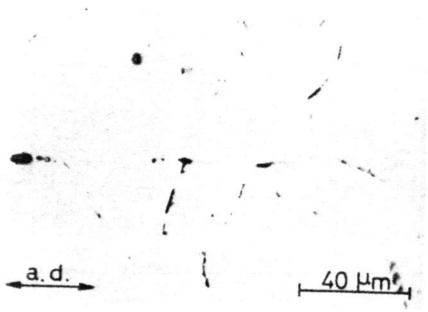


Figure 7 Structural Change Appeared along PAG Boundaries on Electro-Polished Surface of the Same Specimen as in Figure 2 except for Introduction of 20 sec. delayed time. Rotating Bending, σ=637 MN/m^2, N=2.5x10^4

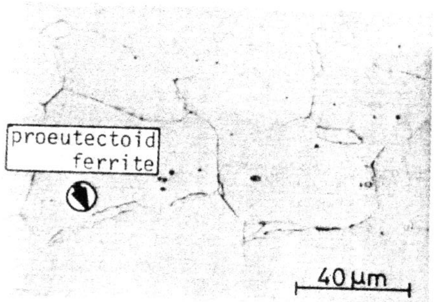


Figure 8 Precipitation of Proeutectoid Ferrite Brought about by Additional 5 sec. of delayed time to the Specimen in Figure 7