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Microstructure evolution of a hypereutectoid pearlite steel under rolling-sliding contact loading

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A B S T R A C T

The microstructure evolution on the rolling surface of GCr15 steel subjected to rolling-sliding and pure rolling contact loading was systematically investigated. Experimental results showed that the pearlite structure of the surface layer in the rolling-sliding sample transformed into nanocrystalline α-Fe–C alloy in which cementite underwent severe decomposition while the pearlite lamellae appeared unperturbed in the pure rolling sample. A white etching layer (WEL) was also detected in the surface of the rolling-sliding sample. The WEL formation was found to be due to cyclic shear plastic deformation instead of frictional heating. A surface layer of Fe_3O_4 was detected in the pure rolling samples. Microhardness depth profiles of the rolling-sliding and pure rolling samples also showed different trends.

1. Introduction

Over the last few decades, a so called white etching layer (WEL) has been frequently observed on the rail surface due to the heavy load and high speed of trains. Due to the practical importance of WEL, much effort has been made to simulate the microstructural evolution and WEL formation of pearlite steels using severe deformation processes (e.g. mechanical milling [1,2], surface mechanical attrition treatment [3,4] and high pressure torsion (HPT) [5]) and fast heating processes such as laser-treatment [6,7].

Newcomb et al. [8] reported that the martensitic microstructure of WEL with supersaturated carbon in solution was formed by cyclic shear deformation of the surface of rails. Lojkowski et al. [9] also stated that the presence of WEL is due to the transformation from pearlite to a nanostructured Fe–C alloy layer after cyclic plastic deformation, which is deduced to take place at rail-wheel contact temperature of less than 230 °C. However, Pyzalla et al. [10] claimed that WEL was formed by austenization and subsequent martensitic transformation, based on their observation of retained austenite within WEL. By shape analysis of the manganese-enriched zones in three-dimensional atom probe (3DAP), Takahashi et al. [11] concluded that the flash temperature could reach the austenization temperature, which supported the hypothesis of austenization and subsequent martensitic transformation of WEL. So far, the mechanism of WEL formation is still a highly controversial issue [1,2,7–13]. Therefore, it is crucial to investigate the microstructural evolution, especially the formation mechanism of WEL of pearlite steel under cyclic deformation, and this is one of the aims of the present work.

Apart from microstructure characterization, mechanical property such as microhardness and the traction coefficient is another key point in the research of WEL [7,11,13]. Österle's [7] measurement of hardness yielded values between 600 and 700 HV in WEL of rails as well as in the martensitic zone of the laser-treated specimen. However, Zhang et al. [13] found that the highest hardness value (950 HV) appeared not on the topmost surface but near the interface between WEL and the transition layer. Takahashi et al. [11] further explained the relative low hardness of the topmost surface by tempering due to the passage of trains after WEL formation. The study of Krause and Demirci [14] showed that the behavior of the traction coefficient under dry rolling-sliding conditions is affected by the structural evolutions, such as changes of the crystal orientation and subgrain size of the deformed surface layer. However, there is still a lack of detailed experimental analysis and illustration of mechanism of the microhardness profile and the traction coefficient during the formation of WEL. Moreover, the microhardness profile and the traction coefficient of the sample subjected to rolling-sliding may be different from

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those of the sample subjected to pure rolling [15], and another aim of the present work is to make clear the differences in the microhardness profile and the traction coefficient between the samples subjected to rolling-sliding and those subjected to pure rolling and to find the underlying microstructural differences.

The rolling contact fatigue machine is regarded as an effective tool for the study of wear and rolling contact fatigue behavior of rail steels, since all the parameters of the real rail-wheel contact such as tensile, shear and compress loading cycle [16,17] can be obtained easily using this process. By measuring the applied load and traction coefficient during the experiment and then characterizing the microstructure and microhardness after the experiment, the relationship of these parameters can be investigated.

In the present experiments, GCr15 (similar to AISI 52100) was chosen as the substance upon which to study the microstructural evolution of pearlite steel using a rolling contact fatigue machine for the following reasons [18]: (1) The ever-increasing strength requirements on rails for heavy-haul railways have recently mandated the use of steels with hypereutectoid compositions. (2) Chromium known to promote the refinement of pearlite lamella is the main additive alloy in rail steel. (3) Cementite content is larger in GCr15 steel, which facilitates the characterization of cementite deformation behavior. In this paper, we use the rolling contact fatigue machine to simulate the contact condition between rail and wheel. By varying experimental parameters of the rolling contact condition, such as contact stress, sliding ratios and experiment time, samples deformed to different degrees can be obtained. The microstructural and mechanical properties of the samples are then studied.

2. Experimental methods

GCr15 steel with the composition of Fe–0.98C–1.61Cr–0.34Mn–0.27Si (mass%) is used here. To get full pearlite structure, the sample is heated up to the austenitizing temperature of 1000 °C and held for 30 min followed by air cooling to room temperature. The as-normalized sample contains mainly pearlite and a very small fraction of pro-eutectoid cementite at the prior austenite grain boundary, with a hardness of 380–390 HV. The original mean true interlamellar spacing measured from ten different TEM micrographs is 155 nm. Fig. 1 shows a typical TEM micrograph of the as-normalized sample, in which the dislocation density of ferrite lamella is very low.

The steel is cut into cylindrical disc with 60 mm in diameter and 10mm in thickness. Twin-disc specimens schematically shown in Fig. 2 are used to simulate rail-wheel contact by using rolling contact fatigue machine [19]. The top and bottom discs are driven by different AC motors, so it is convenient to adopt varying sliding ratios between those two discs, which gives an advantage to model the real rail-wheel contact. During the experiment, torque and traction coefficient are recorded by torque transducer and the surface temperature of the disc is measured by the infrared thermometer.

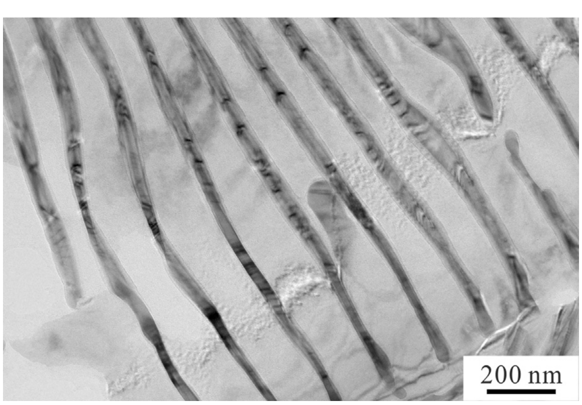

Fig. 1. TEM micrograph of the as-normalized sample.

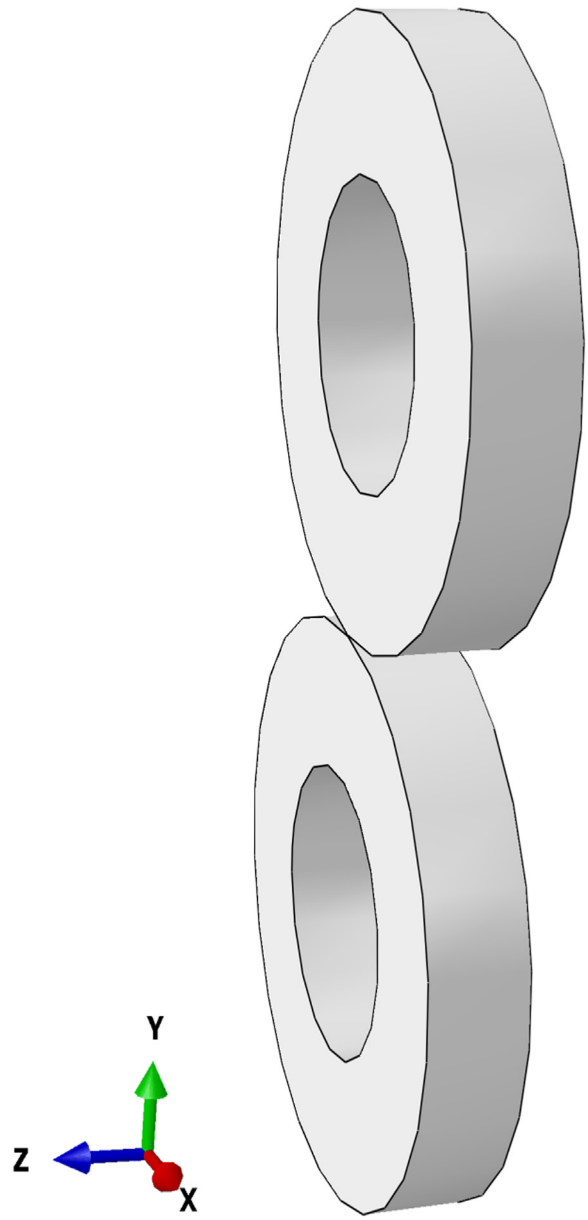

Fig. 2. Twin-disc contact model for the rolling-sliding contact experiment.

The maximum contact stress (normal stress) is calculated according to Hertz contact theory by using the equation [20]

$$\sigma_{\max} = 0.418\sqrt{\frac{FE}{L}\left(\frac{1}{R_W} + \frac{1}{R_R}\right)} \quad (1)$$

where F is the applying force, E is the modulus of elasticity of steel, L is the line contact length, R_w and R_R are the radii of the top and bottom disc, respectively.

The relative sliding ratio between the top and bottom disc is calculated by equation