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Technical note

Very high cycle fatigue behaviors of bainite/martensite multiphase steel treated by quenching-partitioning-tempering process

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ABSTRACT

A bainite/martensite (B/M) multiphase steel was treated by a novel bainite-based Q-P-T treatment with various process parameters to obtain multiphase microstructure composed of bainite, martensite and different types of retained austenite (nanometer-sized film-like and micrometer-sized blocky retained austenite). We investigate here the effect of retained austenite on the very high cycle fatigue (VHCF) behavior, especially the fatigue crack initiation in VHCF regime. Results show that the non-inclusion induced crack initiation is the main fatigue failure mechanism in the B/M steel with blocky retained austenite. The blocky retained austenite transforms easily to martensite due to local plastic deformation under cyclic loading, which is prone to induce the formation of micro-cracks. In contrast, the film-like retained austenite is beneficial to the VHCF property of B/M steel due to the dislocation absorption and transformation induced plasticity effects. Therefore, the B/M steel with film-like retained austenite exhibits excellent VHCF property, namely, the fatigue limit in VHCF regime reaches 770 MPa while the tensile strength is 1410 MPa.

1. Introduction

Very high cycle fatigue (VHCF, i.e., enduring cyclic stress in excess of 10^7 cycles) behavior of high-strength steels has become an important subject to ensure the long-term safety of the structural components such as railway wheels and rails [1,2]. The failure mechanism for VHCF is basically different from that for low cycle fatigue (LCF) and high-cycle fatigue (HCF). The fatigue crack initiation within VHCF regimes is generally divided in two types [3–5]: (i) the cracks initiate from surface of the specimens due to the defects, pores, heterogeneities and other potential stress concentration region at surface [3]; (ii) the cracks initiate from the subsurface or interior of specimens, which is characterized by so-called "fish-eye" surrounding a "fine granular area (FGA)" [1] in the vicinity of an inclusion (also called optically dark area, ODA [6] and bright granular facet, GBF [2]). In general, the inclusions in steels have a significant influence on the fatigue crack initiation and fatigue life [7–10]. Hence, many works in VHCF behaviors of high-strength steels have focused on improving the VHCF properties by reducing the inclusion level and size [7–10].

Recently, the "non-inclusion induced crack initiation" has been observed by some researchers. The crack did not initiate from inclusions but within matrix microstructure [11–15]. For instance, Chai reported this phenomenon (termed as subsurface non-defect crack origins-SNDFCO) in ferrite/martensite and austenite/martensite two phase steels [11,15]. The formation of SNDFCO was caused by local plastic deformation of soft phase (such as ferrite and austenite) [11,12,15]. Our previous work indicated that the VHCF properties of the designed bainite/martensite (B/M) multiphase steels are less sensitive to inclusions compared to tempered martensite steels, and "non-inclusion induced crack initiation" could be one of the most important fatigue failure mechanisms in VHCF regime [13,14,16]. By addition of Si, the carbide is replaced by retained austenite in B/M steels [17]. The microstructures of B/M multiphase steels consist of bainitic ferrite, martensite and retained austenite. Until now, the effect of retained austenite on VHCF behaviors of B/M multiphase high strength steels is still controversial [16,18–21]. On one hand, compared with bainitic ferrite and martensite, the retained austenite is softer phase, which is prone to subject to local plastic deformation and become the crack

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initiation site. On the other hand, the deformation-induced martensitic transformation (DIMT) from austenite might occur near the crack tip and lead to the transformation induced plasticity (TRIP) effect [18]. The TRIP effect could provide a high work hardening capacity and postpone plastic localization [22]. In addition, the martensite formed from DIMT might also affect the fatigue crack initiation from two aspects: (i) the martensite as hard phase offsets the softening of retained austenite; (ii) the martensite is relatively brittle, which might induce the crack initiation and provide a path for crack propagation. Therefore, it is essential to understand the role of retained austenite on VHCF behaviors and ensure excellent service performance of the B/M multiphase steel involving retained austenite.

A novel heat treatment named quenching and partitioning (Q&P) process was employed to tailor the fraction, size and morphology of retained austenite in steels. Q&P process was proposed by Speer et al. for achieving martensite and austenite multiphase [23,24]. The Q&P concept involves an initial cooling step to a quenching temperature (T_q) between the martensite-start (M_s) temperature and the martensite-finish (M_f) temperature, followed by an isothermal holding at T_q (or higher than T_q) with purpose of promoting carbon diffuse from martensite to austenite, thereby obtaining more retained austenite at room temperature. It should be noted that the process parameters (e.g., T_q) greatly affect the final morphology and fraction of retained austenite [17,23,24]. In the present work, a B/M multiphase steel was treated by two bainite-based Q-P-T (BQ-P-T) [25] treatments to obtain different types of retained austenite (with different volume fraction, size and morphology). The effect of retained austenite on VHCF behavior of the B/M multiphase steel was discussed in detail.

2. Test materials and experimental methods

The chemical composition of experimental steel in wt% was Fe-0.2C-2.0Mn-1.0Si-0.8Cr-0.2Mo-0.6Ni. Manganese was added to avoid the austenite to ferrite transformation at high temperatures. Silicon was added to suppress carbide precipitation during quenching, partitioning and tempering steps. The steel was melted in a vacuum induction furnace, followed by electroslag remelting, aiming to reduce the size and number density of non-metallic inclusions in steel [26,27]. The ingot was reheated at 1200 °C and forged to 30 mm thickness with a finish-forging temperature at around 950 °C. The forged plate was then annealed at 950 °C followed by furnace cooling. The as-received microstructure prior to the following experiments (i.e. heat treatment and dilatometer tests) was ferrite plus pearlite.

The continuous cooling transformation (CCT) and martensite-start (M_s) temperature of the steel (~340 °C) was determined by dilatometer tests (on a Bähr D805L quenching device). During continuous cooling, the full martensite microstructure can be obtained when the cooling rate is higher than 15 °C/s, whereas the bainite/martensite (B/M) multiphase microstructure is achieved within a wide cooling rate range (0.1–10 °C/s). Therefore, in order to realize bainite formation during the initial quenching step in this study, the cooing rate during the BQ step was set as ~2 °C/s. The following BQ-P-T processes were designed with different quenching and partitioning temperatures: (i) cooled at ~2 °C/s to 200 °C (below M_s) after austenitization at 880 °C for 60 min, partitioned at 280 °C for 45 min, and tempered at 250 °C for 120 min (abbreviated as BQ200 sample); (ii) cooled at ~2 °C/s to 320 °C (below M_s) after austenitization at 880 °C for 60 min, partitioned at 360 °C for 45 min, and tempered at 250 °C for 120 min (abbreviated as BQ320 sample).

The tensile properties of the steel subjected to BQ-P-T treatment (using cylinder specimens with a gage diameter of 5 mm and a gage length of 25 mm, standard GB/T 228-2002) and impact toughness (with standard Charpy V-notch specimens, $10 \times 10 \times 55\ mm^3$, standard GB/T 229-2007) were measured at room temperature before fatigue test. Three tensile specimens and five impact specimens were tested for each condition and the average values were recorded. The repeatability of the testing data is proved very well and the standard deviation can be acceptable, as shown in Table 1. Microstructures were characterized using optical microscope (OM), scanning electron microscope (SEM, ZEISS EVO18, 20 kV) equipped with electron backscatter diffraction (EBSD, with step size: 0.1 μm) device and transmission electron microscope (TEM, JEOL 2010, 200 kV) before and/or after fatigue test. The volume fraction of retained austenite (RA vol.%) was measured by X-ray diffractometer (Rigaku Smartlab, Cu Kα radiation) at a step of 0.01° and a counting time of 2 s/step.

The geometry and dimensions of the smooth hour-glass type specimens for VHCF testing are shown in Fig. 1a and the real fatigue specimen installed on the ultrasonic fatigue testing machine is shown in Fig. 1b. The specimen geometry is dependent on the Young's modulus and density of the steel, so that the natural frequency of specimens matches the vibration frequency of ultrasonic testing machine. The VHCF testing was conducted on a Shimadzu USF-2000 type ultrasonic fatigue testing machine (Fig. 1b) with a resonance frequency of 20 kHz and stress ratio of $R = -1$. Compressed air was introduced to avoid significant "self-heating" in specimens during the VHCF testing [16]. The ultrasonic fatigue

Table 1
Mechanical properties of BQ200 and BQ320 samples.

Samples	Tensile strength, R_m (MPa)	Yield strength, R_p (MPa)	Elongation, A (%)	Impact ductility, A_{KV} (J)
BQ200	1410 ± 3	1130 ± 7	15.2 ± 0.5	80 ± 5
BQ320	1320 ± 10	995 ± 5	18.8 ± 0.3	30 ± 4

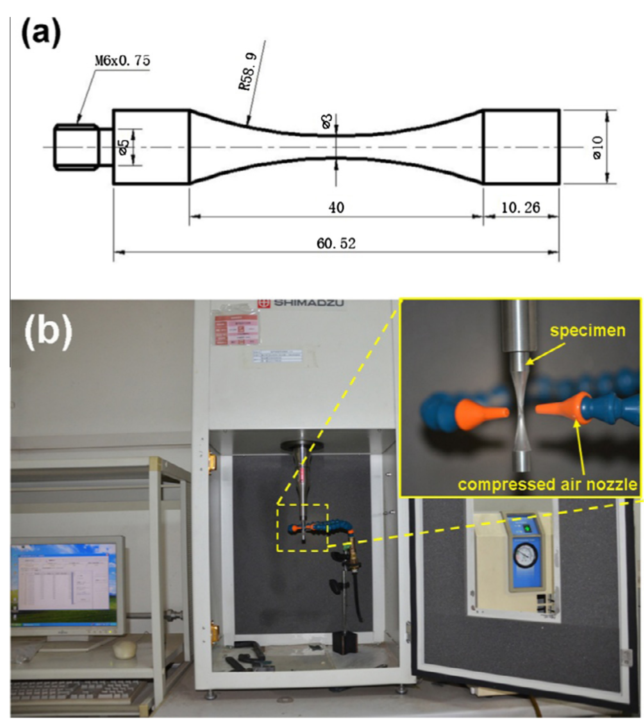

Fig. 1. (a) The geometry and dimensions of the smooth hour-glass type specimens and (b) Shimadzu USF-2000 type ultrasonic fatigue testing machine for VHCF testing.