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Microstructural effects on high-cycle fatigue properties of microalloyed medium carbon steel 38MnVS

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ABSTRACT

High-cycle fatigue properties of medium carbon microalloyed (MA) steel 38MnVS with different microstructural characteristics were studied by rotating bending fatigue test. Low alloy steel 40Cr with quenched and tempered (QT) microstructure was also used for comparison. The results show MA steel in the as-forged condition has the best fatigue property mainly owing to the significant precipitation strengthening of fine V(C,N) precipitates and much fine ferrite–pearlite microstructure, which is even better than that of the QT steel 40Cr. The formation of film-like ferrite along coarse prior austenite grain boundary and coarse bainitic type microstructure as well as the coarsening of the initial fine precipitates significantly deteriorated the MA steel's fatigue properties. SEM examination of fatigue fracture surfaces revealed that the fatigue fracture mechanism of MA steel is different from that of QT steel. It is concluded that microstructure of ferrite–pearlite type MA steels should be carefully controlled after rolling and/or forging to obtain improved fatigue properties.

1. Introduction

In recent years, there are increasingly uses of medium-carbon microalloyed (MA) steels for the production of automotive and some other components for the purpose of higher productivity and thus cost saving compared with conventional quenched and tempered (QT) steels. MA steels are characterized by a small addition of microalloying elements such as V, Nb and Ti to a basic C–Mn composition [1]. These steels are designed to exploit the precipitation strengthening that occurs when a component is cooled after hot forging. By making proper use of this strengthening mechanism, hot-forged components can be obtained which have the same level of strength as those manufactured from low or medium-alloy steels quenched and tempered after forging [2]. However, toughness is a factor that may limit the general use of MA steels, which is inferior to that of QT steels at same strength level. Therefore, the development of MA steels has been considered largely from the point of view of structure–property especially toughness relationships [3–10].

Most engineering components made of MA steels experience cyclic loading in service. Therefore, special attention should also be given to the fatigue properties of MA steels for a proper selection or replacement of materials [2,11–21]. Moreover, lighter cars are in great demand for the purpose of increasing energy efficiency and reducing exhaust. As a result, there are increasing demands for increasing strength, especially for increasing fatigue strength, of MA steels. At the same time, such steels must also be machinable to a certain extent so as to control manufacturing costs. To meet these requirements, the fatigue strength of MA steels should be increased while at the same time maintaining strength level below a given value to permit machinability [13]. In other words, fatigue strength ratio (the ratio of fatigue strength to tensile strength) should be increased.

The forging parameters such as preheating temperature, soaking time, start and finish forging temperatures, strain (or degree of upset) utilized and the post-forging cooling rate as well as compositions of MA steel could highly influence on the microstructure and therefore play an important role on its final mechanical properties [6–8,10,14,15,17,20,22–25]. However, most of these studies were focused on the effects of forging conditions on the microstructure and mechanical properties of MA steels [6–8,10,22–25], only a few focused on the effects on fatigue properties [14,15,17,20,21]. For example, Gündüz et al. studied the effect of different microstructures, which were obtained through post-forging controlled cooling, on the fatigue behavior of MA steel 38MnVS6 and showed that cooling rate had a remarkable effect on the microstructure, hardness and fatigue behavior [20]. Sankaran et al. developed a multiphase microstructure in a V-bearing medium-carbon MA steel using a two-step cooling and annealing treatment following finish forging in an attempt to improve

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fatigue resistance [14,15,17]. It should be noticed that in actual forging experience, the fluctuations of forging parameters and post-forging cooling rate are inevitable and thus may cause microstructural variations. Therefore, in this paper, the high-cycle fatigue properties of widely used medium-carbon MA steel 38MnVS having different microstructural characteristics were compared. Low alloy structural steel 40Cr with quenched and tempered (QT) microstructure was also used for comparison.

2. Material and experimental procedure

2.1. Materials and specimen preparation

The materials used were a commercial medium-carbon MA forging grade steel 38MnVS and low-alloy structural steel 40Cr for comparison and their chemical compositions are listed in Table 1. All steels were supplied in the form of as-hot rolled 90 mm (38MnVS) or 25 mm (40Cr) diameter round bars. One group of the 90 mm diameter round bars were reheated to 1200–1220 °C for at least 1 h and then forged to rods with diameter of 18 mm. The finish forging temperature was around 850–900 °C and then still air cooled. To investigate the effects of microstructures on fatigue properties, part of the 18 mm diameter rods of 38MnVS steel were reheated and soaked at 900 °C or 1200 °C for 45 min or 60 min and then control cooled to room temperature, details of these treatment processes are given in Table 2.

Specimens for fatigue and tensile testing of the 38MnVS steel were machined from the as-forged and treated 18 mm diameter rods or from half the radius of the 90 mm diameter as-rolled round bars in the longitudinal direction. Specimens of the 40Cr steel were machined from the 25 mm diameter rods and then were austenitized at 860 °C for 30 min, oil quenched and reheated to 630 °C for 120 min followed by still air cooling to obtain QT microstructure. Smooth round bar specimens used to evaluate fatigue properties have a minimum diameter of 5.97 mm and a gauge length of 30 mm. Specimens for tensile testing are standard round bar with minimum diameter of 5 mm and gauge length of 25 mm.

2.2. Fatigue testing

The surfaces of all the fatigue specimens were polished in the axial direction using No.1000 abrasive papers after final finishing. Fatigue tests were carried out in laboratory air at room temperature using PQ1-6 type rotating bar two-point bending fatigue testing machine, with the cyclic speed of 5000 rpm and stress ratio $R=-1$. The test was stopped when the specimens did not fail at a cycle number of $N_f=1\times10^7$. The fatigue strength was figured out by the staircase method of at least six pairs in order to raise the confidence. The fracture surfaces of the failed fatigue specimens were examined on a Hitachi S-4300 type scanning electron microscope (SEM).

2.3. Microstructural and mechanical evaluation

Olympus GX51 optical microscope (OP) and SEM were used for microstructural characterization. The specimens were etched with a 3% nital solution after polishing, and the volume fraction of ferrite and pearlite interlamellar spacing were measured by using digital image processor software of SISC IAS V8.0. The specimens for transmission electron microscope (TEM) were sliced into 0.5 mm thick plate and subsequently ground down to a thickness of about 50 μm. These foils were finally electropolished in a twin-jet electropolishing apparatus using a standard chromium trioxide–acetic acid solution. Thin foils were examined in Hitachi H-800 TEM and FEI Tecnal G2 F20 TEM with energy dispersive X-ray spectrometer (EDS) at an operating voltage of 200 kV to study precipitates.

Table 1
Chemical compositions of the tested steels (wt%).

Steel	C	Si	Mn	S	P	Cr	Ni	V	Al	N
38MnVS	0.39	0.20	1.38	0.059	0.016	0.17	0.08	0.11	0.025	0.011
40Cr	0.38	0.28	0.71	0.008	0.007	0.87	0.12	–	0.021	0.009

Tensile tests were conducted on a MTS 810 type universal testing machine with a constant cross-head speed of 1 mm/min at room temperature. Vickers hardness of the specimens were measured with a 5 kg load and Vickers microhardness of both the ferrite and pearlite portions were also measured separately with 10 g and 50 g load, respectively, and the results were the average of at least 10 measurements.

3. Results and discussion

3.1. Microstructures

Table 3 summarizes the microstructural parameters and hardness of the samples. The microstructures corresponding to different conditions of the tested steels are given, respectively, in Fig. 1 (a)–(f). As seen in Fig. 1(a), the as-rolled sample (38-AR) exhibits a large equiaxed prior austenite grain size, and proeutectoid ferrite appears as a thin, nearly continuous network at prior austenite grain boundaries, causing a coarse ferrite network plus pearlite microstructure. This is attributed to the comparatively lower cooling rate after hot rolling because of the large diameter of the as-rolled bar. The microstructure of the as-forged sample (38-F) is mainly composed of very fine and more evenly distributed ferrite and pearlite (Fig. 1(b)), mainly owing to much higher forging deformation ratio (~25:1), low finish forging temperature and high post-forging cooling rate of small size rod. In general, low forging temperature retards significantly the recrystallization process, and thus leads to a decrease of the austenite grain size [26]. Heavy deformation at low temperature increases remarkably the degree of pancaking and the creation of more deformation bands. Therefore, the proeutectoid ferrite nucleates not only on austenite grain boundaries, but also on deformation bands, causing significant microstructural refinement.

For the high temperature annealed sample 38-HA, its microstructure is similar to that of the as-rolled sample 38-AR (Fig. 1(c)), whereas for the low temperature annealed sample 38-LA, its microstructure is similar to that of the as-forged sample except its grain size is larger (Fig. 1(d)). These results could be related to the influence of austenitizing temperature. When the sample was soaked at 1200 °C, almost all the V(C,N) precipitates were dissolved in austenite, thus they did not exert a pinning effect on austenite grain boundaries to prevent grain growth [6]. Moreover, slow furnace cooling till 900 °C (~5 °C/min) also favors grain coarsening. Therefore, rather coarse microstructure was obtained for the 38-HA sample. However, when the sample was soaked at 900 °C, part of the V(C,N) precipitates still remained undissolved in austenite, thus these undissolved precipitates could effectively inhibit the growth of austenite grains during soaking and subsequent furnace cooling. Therefore, the 38-LA sample possesses a finer microstructure.

Normalizing treatment resulted in a coarse bainitic or acicular ferrite plate type microstructure (Fig. 1(e)). This phenomenon is generally associated with the influence of austenitizing