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BAINITE MORPHOLOGY IN TWO EXPERIMENTAL Mo-Cr and Mo-Cr-V-Ti STEELS

ABSTRACT

The paper presents an investigation of the bainite morphology in two experimental Mo-Cr and Mo-Cr-V-Ti steels using TEM, high speed dilatometry backed by thermodynamic analysis. The microstructure was investigated using metallography and TEM method. After austenitisation at 1200°C followed by bainitic reaction in upper and lower temperatures of isothermal transformation the bainite was in the form of classical sheaves. The amount, distribution and morphology of retained austenite and bainitic ferrite depend on prior austenitisation and isothermal transformation temperatures within the bainitic range.

Keywords: *bainite morphology, bainite reaction, low alloy experimental steels*

INTRODUCTION

Until now many topics on bainite transformation, such as the transformation mechanism of ferritic component and the sources of carbide precipitates, remain controversial [1,2]. The present paper attempts to investigate bainite morphology in two laboratory prepared low alloy Cr-Mo and Cr-Mo-V-Ti steels and to clarify some problems of the formation mechanism of bainite.

This type of high strength low alloy (HSLA) steels are supposed to have the requisite good combination of strength, toughness and weldability for many industrial applications. Low carbon content promotes good toughness and weldability by elimination the volume fraction of carbide containing microconstituents (e.g. pearlite). Addition of chromium increases hardenability of steels and additionally in second steel vanadium and titanium are present to refine the austenite grain size, resulting in a fine grained ferritic structure.

The main aim of the present investigation was to study the influence of austenitising temperature (changing the volume of dissolved carbides and from this the chemical composition of austenite) and temperature of isothermal transformation on bainite morphology. In Fe-C-X steels, when X is a strong carbide forming alloying element, interphase boundary carbides and fibrous carbides were present after isothermal reaction at temperatures above the bay temperature [3]. In the present study attention was paid to a much more detailed delineation of bainitic ferrite and retained austenite morphologies formed at considerably lower temperatures, e.g. the temperatures of bainitic range.

EXPERIMENTAL PROCEDURE

The compositions of the alloys investigated are given in Table 1. The ingots were forged to 15mm dia. rod. The specimens were austenitised for 20 min. at 1000 and 1200°C. Subsequent isothermal heat treatments were carried out by quenching the specimens into a tin bath covered with a layer of charcoal. A coating layers were used to protect the specimens against decarburization. After isothermal transformation, the specimens were water quenched.

Table 1. Chemical compositions of the alloys investigated (wt. %)

Alloy	C	Si	Cr	Mn	Mo	V	Ti	P	S	Al
Mo-Cr	0.14	0.23	0.88	0.73	2.4	-	-	0.021	0.017	0.03
Mo-Cr-V-Ti	0.15	0.24	0.84	0.92	2.6	0.24	0.12	0.022	0.019	0.06

Dilatometric analysis was carried out on a Leitz-Wetzlar vacuum dilatometer using a 20 mm by 3.5 mm dia. specimens to establish the A_1 and A_3 temperatures. A high-speed LK-02 Adamel Lhomargy dilatometer was used to establish the M_s temperatures (Table 2). In order to ensure rapid quenching (300 K s^{-1}) the specimens were 12mm in length and 1.0mm in diameter. The specimens were austenitised for 10 min. at 1000°C in dilatometry furnace and subsequently gas quenched to the isothermal transformation temperature by an automatically controlled high-pressure helium jet.

Quantitative X-ray analysis was used to determine the retained austenite content after isothermal transformation. For this purpose, samples were examined in a DRON-1,5 X-ray diffractometer using Fe-filtered $\text{CoK}\alpha$ radiation. The machine was operated at 38 kV and 10 mA. The retained austenite content was calculated from the X-ray intensities. A peak separation technique was used to separate the (111) γ -reflection from the overlapping (110) α -reflection.

The specimens for transmission electron microscopy (TEM) were machined to 3mm dia. rod, carefully avoiding any heating. The rods were sliced into 0.35mm thick discs while being kerosene cooled. The discs were subsequently ground down to a thickness of 40-50 μm on 500 grid paper. These foils were finally electropolished in a twin-jet at room temperature and at 55-60V using a 25-pct glycerol, 5 pct perchloric acid and 70 pct ethanol mixture. Thin foils were stored in ethanol and subsequently examined in a Tesla BS-540 electron microscope at an operating voltage of 120 kV. Optical microscopy was used to examine etched structures. Specimens were etched in 2% nital solution.

RESULTS AND DISCUSSION

To plan the heat treatments a time-temperature-transformation (TTT) diagrams were calculated for the alloys given in Table 1. The calculation was done taking into account the carbon, silicon, manganese, nickel, molybdenum, vanadium and chromium concentrations, using a method developed by Bhadeshia [4,5,6]. In that method the TTT diagram has been

treated as being composed of two separate C curves, one of which represents reconstructive (diffusional) transformations and the other, displacive reactions (Widmanstätten ferrite and bainite). The technique is capable of correctly generating the bay region of time-temperature-transformation diagrams, while at the same time allowing relative shifts in the shear and diffusional C curves, as a function of alloying element content. The bainitic ferrite B_S and martensitic M_S start temperatures were also calculated using the same method. The bainite start temperature B_S is the temperature corresponding to the upper part of the shear transformation C curve and it strictly refers to the point at which the nucleation of ferrite that grows displacively first becomes possible [5,6]. The transformation temperature data are given in Table 2.

Table 2. A_1 , A_3 , B_S and M_S temperatures

Alloy	Temperature, °C				Experimentally determined ***		Thermodynamically calculated			
	A_{c1}	A_{c3}	A_{r1}	A_{r3}	M_S	B_S	M_S^*	B_S^*	M_S^{**}	B_S^{**}
Mo-Cr	793	941	771	914	467	-	445	574	486	607
Mo-Cr-V-Ti	807	969	730	896	453	-	423	555	479	601

* Assuming full austenitisation (1200°C) ensuring complete dissolving of carbides in austenite

** Assuming partial dissolving of carbides in austenite (1000°C)

*** After austenitisation at 1000°C/10min.

The dilatometric investigations were used to verify thermodynamic calculations. The dilatometric specimens were cooled to the isothermal transformation temperatures and were kept for an appropriately long time (until the bainite reaction ceased) for a variety of temperatures below the bainite start temperature B_S and above the martensite start temperature M_S . The volume change associated with a phase transformation was registered and could provide information about the reaction kinetics. The time-temperature-transformation diagrams showing calculated reaction start curves are shown for Mo-Cr steel in Figure 1 and for Mo-Cr-V-Ti steel in Figure 2. In general, the bainite transformation was faster as the isothermal temperature was decreased. The bainite transformation is faster after austenitising at 1000°C because it is known that the rate of bainite formation increases with decreasing carbon and alloying elements content in austenite. The lower M_S and B_S temperatures after austenitising at 1200°C can also be attributed to the higher carbon content in austenite as compared with the austenitising at 1000°C.

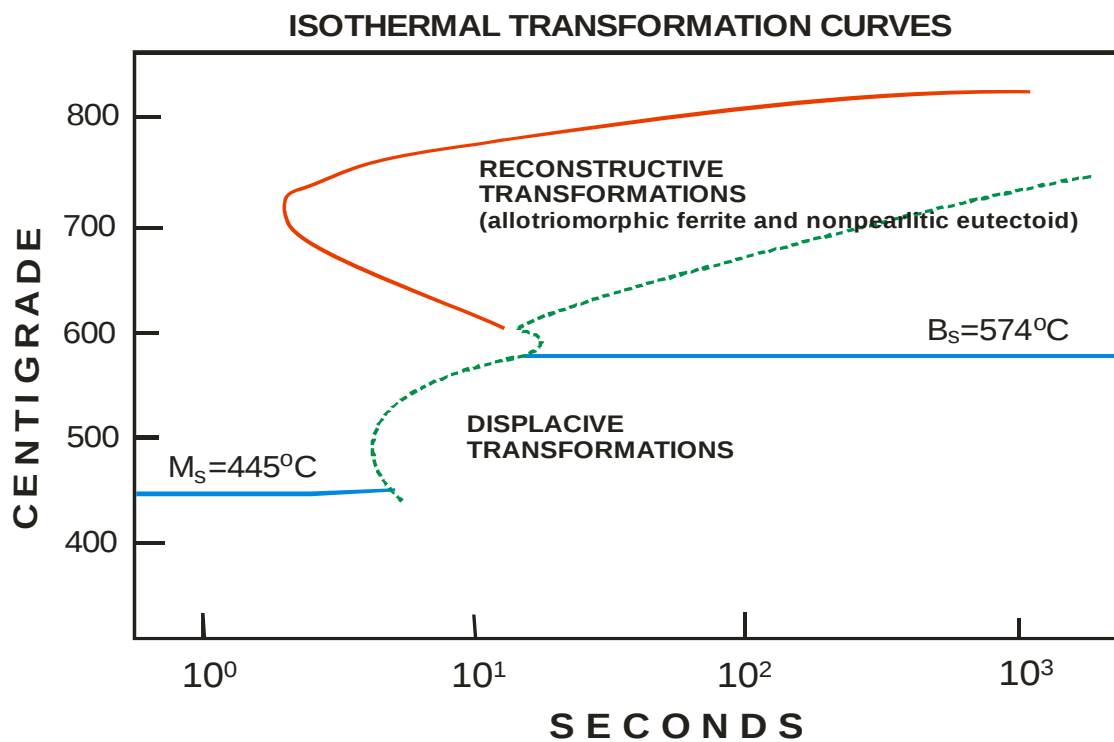


Fig. 1. Calculated time-temperature-transformation (TTT) diagram showing the reaction initiation C curves for investigated Mo-Cr steel and chemical composition given in Table 1

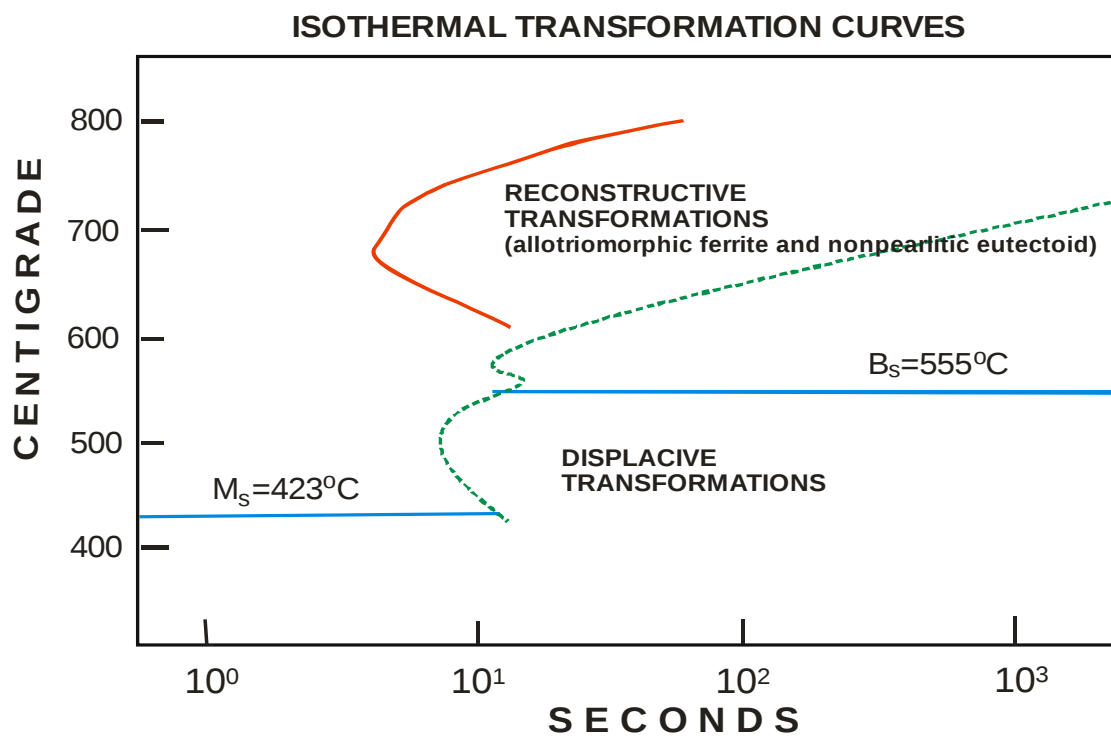


Fig. 2. Calculated time-temperature-transformation (TTT) diagram showing the reaction initiation C curves for investigated Mo-Cr-V-Ti steel and chemical composition given in Table 1

Based on the dilatometry data and the data from the calculated TTT diagrams heat treatment of Mo-Cr steel was planned as follows: two austenitising temperatures 1000 and 1200°C/20min. and two utmost temperatures of bainite range, e.g. 575 – 500°C and 550 – 450°C respectively. Austenitising temperature 1200°C was required for complete dissolution of VC carbides [3,7,8]. Isothermal reaction temperatures were chosen to lie above the M_s temperature to avoid formation of martensite during the isothermal heat treatment, and below the bay temperature of investigated steel. In case of Mo-Cr-V-Ti steel austenitisation was carried out at 1000 and 1200°C/20 min. as well, and temperatures of isothermal transformation were chosen the same as for previous steel, e.g. 575-500 °C and 550-450°C respectively.

Fig.3 shows a microstructure after eutectoidal decomposition of Mo-Cr steel composed of ferrite and two different carbide morphologies, e.g. fibrous carbides and fine carbide particles (interphase precipitation). During eutectoidal decomposition of the molybdenum, vanadium and titanium alloys, Mo_2C , VC or TiC carbides are precipitated respectively, as a result of non-pearlitic reaction with an opposition to pearlitic transformation which is a classical form of eutectoidal transformation in steels [3,7-9].

Thus, in steels containing strong carbide forming elements fibrous and interphase precipitation are the alternative forms of eutectoidal transformation.

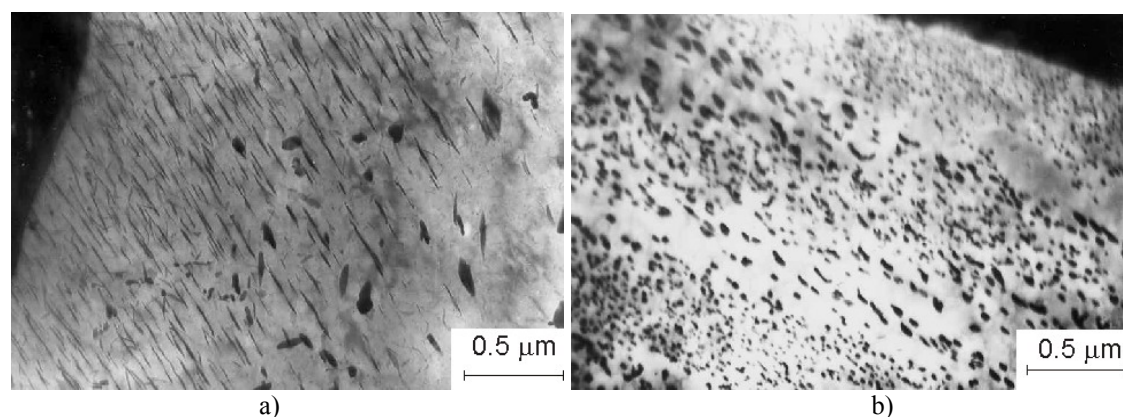


Fig. 3. Microstructure of non-pearlitic eutectoid in Mo-Cr steel after isothermal transformation at 750°C above the bay region: a) fibrous carbides, mainly Mo_2C , b) interphase precipitation. Thin foil. TEM

Optical microscopy was carried out in order to observe any variations in the gross features of microstructure, e.g. prior austenite grain size, bainitic packet size and distribution of blocky austenite with heat treatment.

Optical micrographs of the partially reacted specimens of the Mo-Cr-V-Ti steel are shown in Figure 4. Only large individual carbides TiC, regardless of austenitization temperature, can be detected in micrograph of the Mo-Cr-V-Ti steel (TiC carbides are arrowed). In observed microstructures the bainite reaction has not proceeded to completion, and large pools of residual austenite have subsequently transformed to martensite on water quenching. After reaction for longer time at temperature 550°C there were observed significant changes of extend of transformation (Fig. 4).

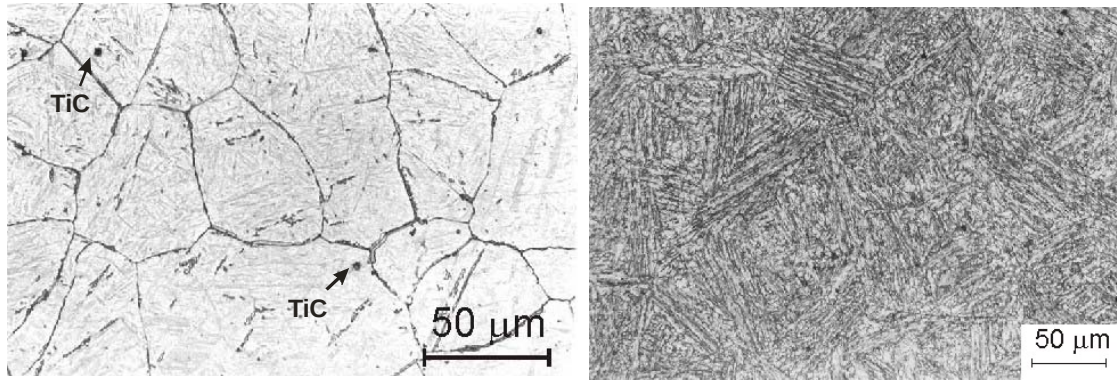


Fig. 4. Microstructure of Mo-Cr-V-Ti steel austenitized at 1200°C and austempered at: a) 550°C/1.5h, b) 550°C/6h. Etched with 2% nital

The fine details of the microstructure were examined by TEM. Figure 5 shows the general morphology of the microstructure in Mo-Cr-V-Ti steel after austenitisation at 1200°C and isothermal transformation at 550°C for: a) 5400s, and b) 7200s.

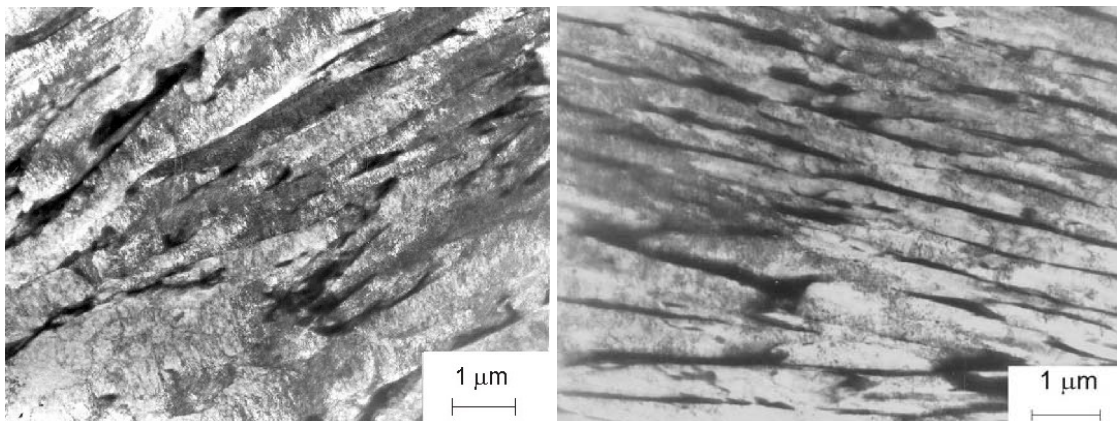


Fig. 5. Microstructure in Mo-Cr-V-Ti steel after austenitisation at 1200°C and partial isothermal transformation at 550°C for a) 5400s, and b) 7200s. Thin foil. TEM

The morphology of the bainite is similar to low carbon lath martensite, where dislocated laths are separated by films of retained austenite. No blocky austenite was observed in this structure. Careful examination of this microstructure shows no evidence of carbides precipitation. No significant changes in morphology with increasing reaction time were observed after isothermal transformation at 550°C. This structure, therefore, belongs to upper bainite assuming that upper bainite in this steel is a structure composed of carbide free bainitic laths with interlath retained austenite films [8,13]. Sometimes this kind of bainitic microstructure is called as carbide free bainite.

After austenitisation at 1200°C and at lower transformation temperature (450°C), bainite changes into a morphology without carbides within the laths (Figure 6).

During a progress of bainitic reaction the excess of carbon in bainitic ferrite may partition eventually into the residual austenite or precipitate from the ferrite in the form of carbides. If the latter process is absent, then upper bainite is obtained [10,11]

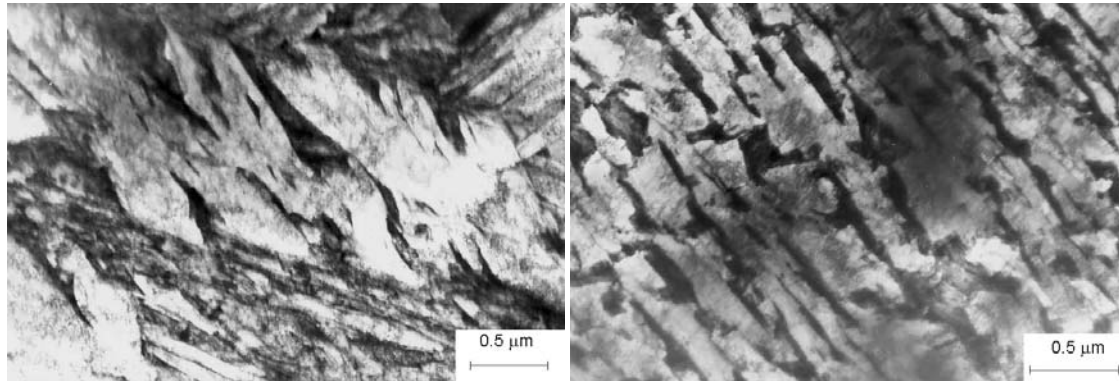


Fig. 6. Microstructure of Mo-Cr-V-Ti steel after austenitisation at 1200°C and isothermal transformation at 450°C for: a) 900s, b) 7200s. Thin foil. TEM

Figure 7 shows the general morphology of the microstructure in Mo-Cr steel after austenitisation at 1200°C and isothermal transformation at 550°C. The morphology of the bainite is significant different to that after transformation Mo-Cr-V-Ti steel at the same temperature.

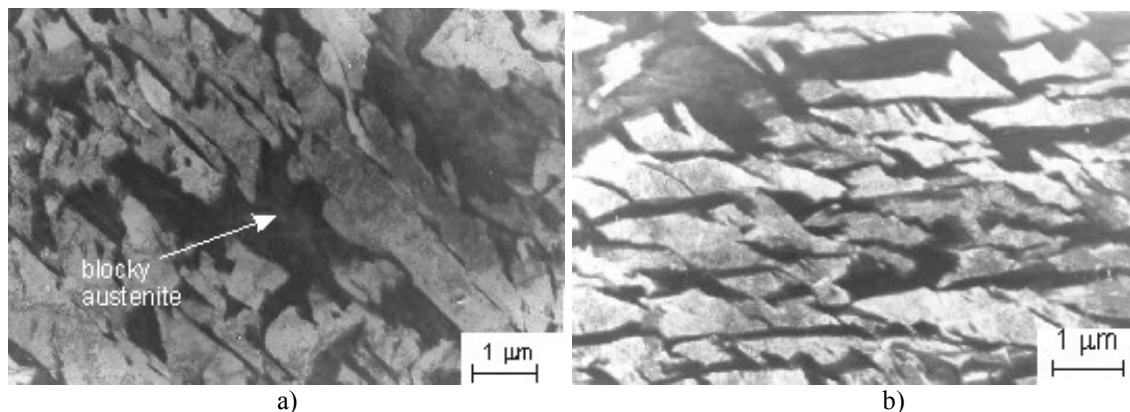


Fig. 7. Microstructure in Mo-Cr steel after austenitisation at 1200°C and isothermal transformation at 550°C for: a) 3600s, and b) 550°C/6h. Thin foil. TEM

Figure 6a shows a morphology of blocky residual austenite in Mo-Cr steel after heat treatment at 1200°C following isothermal transformation at 550°C. After transformation to bainite at higher temperature retained austenite exhibited a blocky, triangular shape (Fig. 7). Beside blocky austenite, interlath films of retained austenite in the bainitic regions were also observed after this treatment. In Mo-Cr steel after prolonged time of isothermal transformation retained austenite occurs mainly as interlath films (Fig.7b).

As the temperature of isothermal reaction for Mo-Cr steel decreases from 550 to 450°C the structure changes from upper (granular) bainite to upper lath bainite (Fig. 8). It is assumed that upper bainite in these steels is a structure composed of carbide free bainitic ferrite laths

with interlath retained austenite films replacing the interlath cementite of the classical upper bainitic microstructure. The interlath retained austenite films are stable and there is no apparent decomposition to martensite.

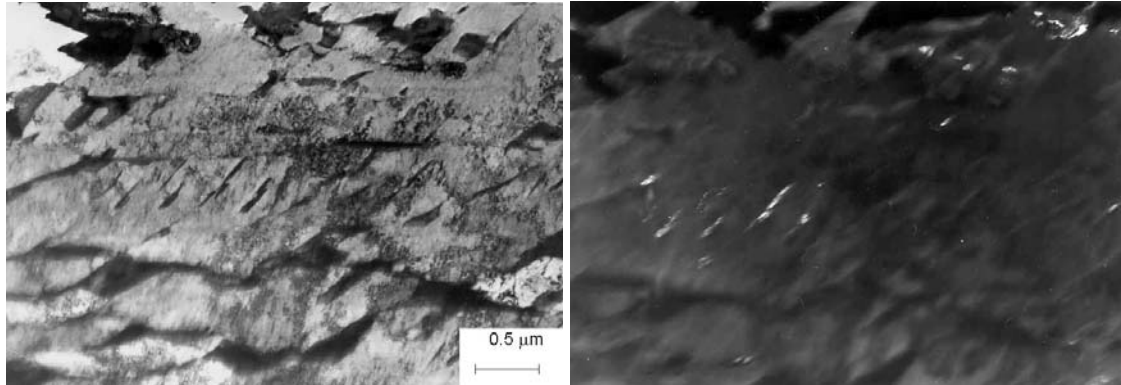


Fig. 8. Microstructure of Mo-Cr steel after austenitisation at 1200°C and isothermal transformation at 450°C for 7200s. Bright and dark field from residual austenite. Thin foil. TEM

In Figure 8 it may be seen from the dark field image that discrete residual austenite films are distributed between ferrite laths, and no fine carbide particles are within the laths. This structure, therefore, belongs to typical upper bainite. It is also interesting to note that these fine austenite films are lying along bainitic ferrite laths, where the carbon content of the retained austenite is likely to be highest [12,13]. It is generally accepted that the residual austenite is a consequence of the partitioning of carbon from the bainitic ferrite to the austenite that accompanies the bainite reaction [6,8,10]. Figure 8 illustrates the results discussed.

The microstructural dimensions are given in Table 3. There is a general tendency for structural refinement with decreasing austenitising temperature. The packet size of Mo-Cr steel is coarser than that of Mo-Cr-V-Ti steel, which can be related to the larger austenite grain size of Mo-Cr steel.

It was found that for both Mo-Cr and Mo-Cr-V-Ti steels the volume fraction of retained austenite decreased with decreasing of the isothermal transformation temperature (Table 3). The volume fraction of retained austenite after austenitisation at 1000°C was beneath the level of resolution of applied X-ray diffractometer. When temperature of isothermal transformation was decreased the blocky morphology of retained austenite was replaced by films of retained austenite separated by platelets of bainitic ferrite. This film austenite, since it is trapped in the immediate vicinity of bainite laths is known to contain a higher carbon content than blocky austenite has and it is difficult to transform to martensite [6,10]. Therefore an increase in the volume fraction of retained austenite with an increase in isothermal transformation temperature results as a higher amount of residual austenite with lower carbon content.

Thus the amount, distribution and morphology of retained austenite and occurrence of carbides vary with the austenitising and transformation temperatures. The dispersed and ductile austenite films between the ferrite platelets can be expected to have a crack blunting effect. But the blocks of austenite tend to transform to high carbon untempered martensite under the influence of small stresses and consequently have an embrittling effect.

Table 3. Microstructural parameters

Microstructural parameters	Steel Mo-Cr and steel (Mo-Cr-V-Ti) *			
	Austenitisation temperature: 1200°C		Austenitisation temperature: 1000°C	
	Transformation temperature, °C		Transformation temperature, °C	
	550	450	575	500
Average austenite grain size, μm	75 (50)	75 (50)	30 (17)	30 (17)
Volume of bainite in the form of packets, %	70 (45)	80 (50)	40	45
Packet length, μm	60 (40)	64 (40)	-	-
Packet width, μm	30	36	-	-
Width of bainite lath, μm	0,6 (0,5)	0,3 (0,4)	0,7 (0,6)	0,4 (0,5)
Morphology of retained austenite	thin films, also irregular islands	mainly thin films	irregular islands and thin films	irregular islands and thin films
Volume fraction of retained austenite, % **	7,0 (5,0)	14,5 (8,6)	-	-

* Microstructural parameters for Mo-Cr-V-Ti steel are given in parenthesis

** Volume fraction of retained austenite were determined by rtg diffractometer on samples quenched to ambient temperature after isothermal transformation for 6h at temperature 550°C and for 3h at temperature 450°C

CONCLUSIONS

The development of bainitic transformation was studied using TEM, dilatometry and thermodynamic calculations, both from a fully austenitic microstructure ($T_A=1200^\circ\text{C}$) and from a microstructure containing a mixture of austenite and undissolved carbides ($T_A=1000^\circ\text{C}$).

Transmission electron microscopy revealed that after isothermal transformation the microstructure within the prior austenite grains consists of a mixture of ferritic bainite, residual austenite and/or carbides. After austenitisation at 1200°C followed by bainitic reaction, the bainite was in the form of classical sheaves. In Mo-Cr-V-Ti steel for upper and lower temperatures of isothermal transformation (550, 450°C) the bainite was found to contain ferrite laths and mainly austenite films, identical to that observed in conventional upper bainite. The amount and morphology of retained austenite and bainitic ferrite strongly depends on prior austenitisation and isothermal transformation temperatures within the bainitic range.

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