Journal of University of Science and Technology Beijing
Volume 10, Number 2, April 2003, Page 39

Development of grain boundary allotriomorphic ferrite/granular bainite duplex steel

Pingguang Xu[1], *Bingzhe Bai*[1], *Hongsheng Fang*[1], *Zhenjia Wang*[2], *Jianping Wang*[1], *and Yongkun Pan*[2]

1) Department of Materials Science and Engineering, Laboratory of Advanced Materials, Materials Research Center, Tsinghua University, Beijing 100084, China

2) Department of Mechanical Engineering, Tsinghua University, Beijing 100084, China

(Received 2002-01-20)

Abstract: A new hot-rolled low alloy high strength steel with grain boundary allotriomorphic ferrite/granular bainite duplex microstructure has been developed through novel microstructure and alloying designs without any noble metal elements such as nickel and molybdenum. Its as-rolled microstructure and mechanical properties, fatigue crack propagation behavior compared with single granular bainitic steel as well as continuous cooling transformation, were investigated in detail. The measured result of CCT (continuous cooling transformation) curve shows that such duplex microstructure can be easily obtained within a wide air-cooling rate range. More importantly, this duplex microstructure has much better combination of toughness and strength than the single granular bainite microstructure. It is found that the grain boundary allotriomorphic ferrite in this duplex microstructure can blunt the microcrack tip, cause fatigue crack propagation route branching and curving, and thus it increases the resistance to fatigue crack propagation, improves steel toughness. The mechanical properties of the above commercial duplex steel plates have achieved or exceeded 870 MPa ultimate tensile strength, 570 MPa yield strength, 18% elongation and 34 J Charpy V-notch impact energy at −40℃, showing good development potential.

Key words: grain boundary allotriomorphic ferrite; granular bainite; duplex microstructure; high strength steel

1 Introduction

High strength steel plates are required to have better strength-toughness and weldability than ever, which speeds up many new technology applications such as ladle refining, microalloying, thermomechanical controlled process (TMCP) [1] *et al.*, and promotes the development of high yield strength steel (HY) [2], copper-precipitation-strengthened high strength low alloy steel (HSLA) [2], ultra-low carbon bainitic steel (ULCB) [3], ultrafine grained ferrite-pearlite steel [4] and so on. However, these high strength steels either contain much noble metal elements Ni, Mo or highly depend on complex process facilities, like controlled rolling with large pass strain and low rolling temperature, accelerated cooling, quenching and tempering/ageing.

In order to meet the industrial requirement on structural steel plates with high strength, toughness and good weldability and partially substitute for the conventional low alloy high strength steel Q345, the authors have developed a new bainitic duplex steel by means of proper duplex microstructure designing and cheap element alloying, which is also suitable for the conventional production line of steel plates. Its as-rolled microstructure, mechanical properties and fatigue crack propagation behavior compared with single granular bainitic steel as well as continuous cooling transformation were investigated and discussed in this paper.

2 Materials and experimental procedures

2.1 New duplex microstructure and its alloy composition

If the two ductile phases in two-phase alloy have same magnitude grain size, according to the classic strengthening-toughening theory of aggregated structure [5], the strength of this duplex microstructure increases as the volume fraction of the stronger phase increases, and the plasticity and toughness of this duplex microstructure increases as the softer phase increases. Since the strength of the granular bainite (Bg) structure is much higher than that of the pearlite structure containing the same carbon content, substituting pearlite with Bg in ferrite-pearlite (F/P) steel will evidently increase the steel strength. Though Bg or martensite-austenite (M/A) constituent in the con-

Corresponding author: Pingguang Xu, **E-mail:** xupg98@mails.tsinghua.edu.cn

tinuously air-cooled low alloy steels was usually thought as a detrimental phase to steel toughness and ductility [6], in fact, granular bainitic steels may obtain good strength-toughness by means of composition optimization, cooling process control and/or properly tempering treatment [7]. On the other hand, the pro-eutectoid ferrite except for ferrite net and Widmannstätten ferrite in F/Bg steel is able to maintain the good plasticity and toughness of F/P steel. So discontinuous grain boundary allotriomorphic ferrite (F_{GBA}) as one constituent phase is adopted to improve steel toughness. The grain boundary allotriomorphic ferrite/granular bainite (F_{GBA}/Bg) duplex microstructure is schematically illustrated in **figure 1**.

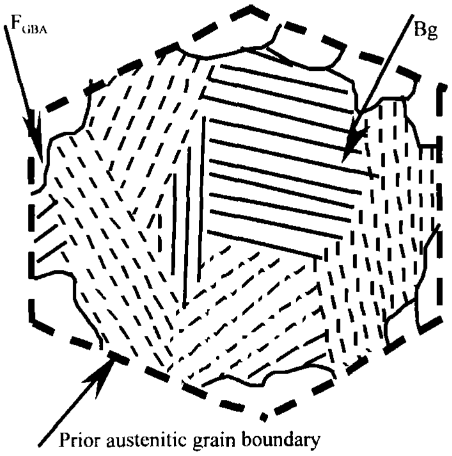

Figure 1 Schematic illustration of grain boundary allotriomorphic ferrite/granular bainite duplex microstructure.

In order to obtain good weldability, the carbon content of experimental steels should not be more than 0.12% (mass fraction, so as the follows). The pancake-shape pro-eutectoid ferrite transformation curve is preferred to obtain proper F_{GBA} within a wide air-cooling rate range, and the lower bainitic transformation start temperature ($T_{B,S}$) for fine bainitic ferrite laths and the higher martensite transformation start temperature ($T_{M,S}$) for lower cold crack susceptibility during post-welded cooling should be considered together, so proper amount of Mn and little amount of Cr were added into this new steel considering the alloy cost. Furthermore, more than 1.8% Mn addition to Fe-C alloys could delay the isothermal transformation and cause its TTT (time temperature transformation) curve to separate into two separate C-curves [8]. Since silicon is well known to suppress the formation of carbide during bainitic transformation, increase the residual austenite stability, and improve the resistance to temper, about 1.2% silicon was added to this new steel.

2.2 Experimental methods

The experimental steel A and steel B with similar composition except for more than 2.5% Mn in steel A and 2.0%-2.5% Mn in steel B were melted by vacuum induction furnace. The samples (ϕ3mm×10mm) were austenized at 940℃ for 10 min, and cooled at different rate in Formastor-digital thermal dilatometry to study the characteristic of its CCT curve. Steel B was also forged into a 35 mm-thick slab, further rolled into a 12 mm-thick plate, then directly air-cooled to ambient temperature, and it obtained F_{GBA}/Bg duplex microstructure in the end. In order to be compared with steel B, steel plate A with single granular bainite microstructure was also prepared through forced air-cooling after hot rolling. Based on the above experimental results, the industrial production of this F_{GBA}/Bg duplex steel plate (steel C) was carried out in a steel plant, with the same composition as steel B. The 12 mm × 1900 mm × 6000 mm plates of steel C rolled from 180 mm thick continuously casting slab were air-cooled in the production line and then off-line-stacked in several groups at about 250℃.

The grain boundary allotriomorphic ferrite amount in F_{GBA}/Bg duplex steel B and C was measured with grid quantitative metallography method, and the retained austenite amount in above three steels was examined on Rigaku D/max-RB X-ray diffractometer. The fatigue crack propagation samples were prepared on Instron-1603 high-frequency-fatigue testing machine and protected by nickel-plating for SEM analysis. Thin foil specimens of the above three steels were prepared according to standard procedures, and examined on JEM 200CX transmission electron microscope at 200 kV.

3 Results and discussion

3.1 Continuous cooling transformation

Figures 2 and 3 show the microstructures of steel B at various cooling rates and its CCT curve respectively. It is easily found that the microstructure at the cooling rate of 75℃/s is essentially low carbon lath martensite. When the cooling rate is reduced to 10℃/s, a little F_{GBA} appears, though the predominant microstructure is bainite/martensite. The F_{GBA}/Bg duplex microstructure occurs at the cooling rate range of 5.0 to 0.5 ℃/s, and the volume fraction of ferrite gradually increases with the decrease of cooling rate, especially at the cooling rate of 0.055℃/s, there is more than 90% ferrite in the steel. In order to achieve 800 MPa ultimate tensile strength according to Vickers hardness in figure 3, the average cooling rate of experimental steel should be no less than 0.5℃/s. The temperature measurement and the simulated computerization both indicate that this cooling rate is easily to meet for up to 20 mm-thick steel plates during air-cooling.

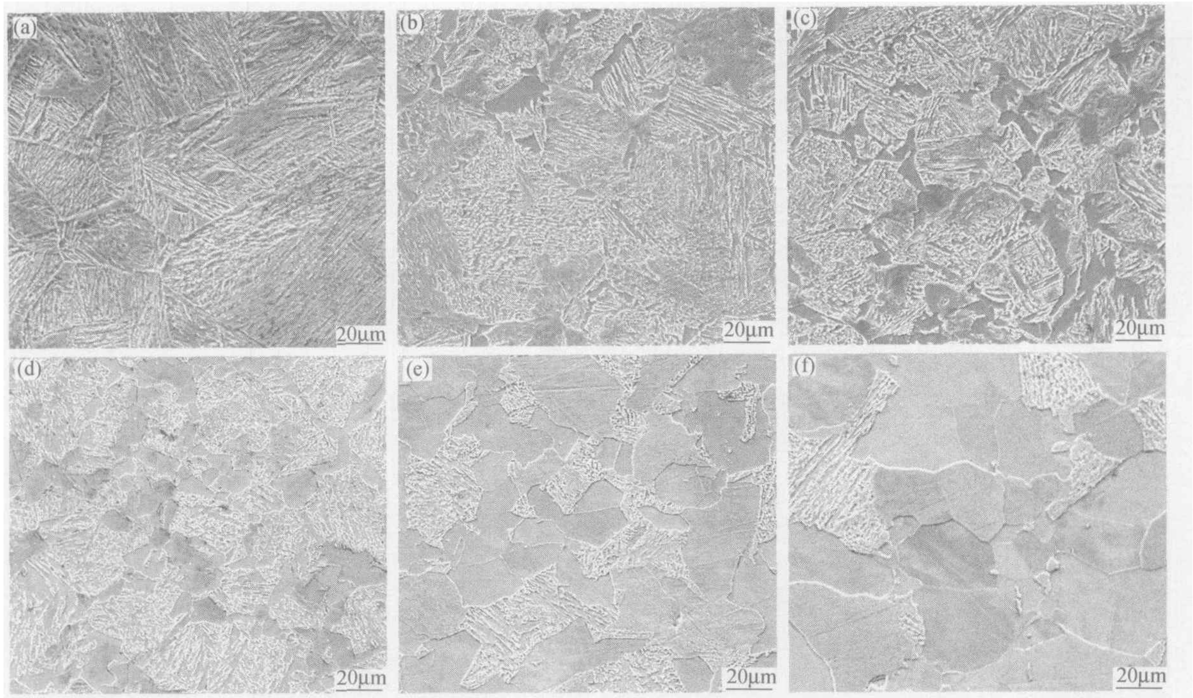

Figure 2 SEM microstructure of continuous cooling transformation of experimental steel B, (a)75℃/s; (b)10℃/s; (c) 5.0℃/s; (d) 2.5℃/s; (e) 0.5℃/s; (f) 0.055℃/s.

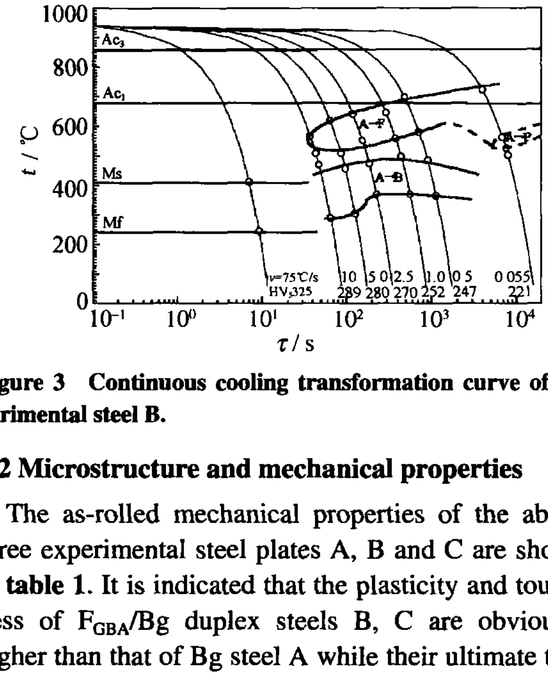

Figure 3 Continuous cooling transformation curve of experimental steel B.

3.2 Microstructure and mechanical properties

The as-rolled mechanical properties of the above three experimental steel plates A, B and C are shown in **table 1**. It is indicated that the plasticity and toughness of F_{GBA}/Bg duplex steels B, C are obviously higher than that of Bg steel A while their ultimate tensile strengths are approximately at the same strength level. The instrumented impact results of the above three experimental steels are shown in **table 2**. The maximum resistance to impact load (P_{max}) of F_{GBA}/Bg duplex steels B and C is close to that of Bg steel A, however, both steel B and C have higher crack initiation energy (E_I) than steel A.

Figure 4 shows the typical micrographs of the three experimental steels, and there are 2.5%, 2.6% and 6.5% retained austenite in steel A, B and C respectively. From figure 4(d)-4(f), it is found that the microstructure of steel A mainly consists of bainite ferrite lath matrix of high dislocation density (gray) and elongated M/A islands (black). The microstructure of steel B and steel C both consist of allotriomorphic ferrite of low dislocation density (white), bainite ferrite laths (gray) and M/A islands (black). The lower ductility and toughness of granular bainitic steel A are mainly due to the existence of non-tempered M/A islands composed of almost high-carbon martensite, which easily initiate micro-cracks along the boundary between M/A islands and bainitic ferrite matrix and then develop into cleavage cracks [9]. The certain amount F_{GBA} in F_{GBA}/Bg duplex steels B and C allows the dislocations to slip easily, relaxes the local stress concentration, delays the microcrack initiation, and therefore improves the steel plasticity and toughness except for lower yield strength. In addition to the smaller size of F_{GBA} and M/A islands due to large accumulative reduction in steel C, the offline-stacking process from 250℃ to ambient temperature has certain self-tempering effect on the high-carbon martensite and thermal stability effect on the retained austenite in M/A islands so that the strength, plasticity and toughness of steel C are all higher than that of steel B. As shown in **figure** 5(a) and 5(b), there are some retained austenite films in the M/A islands to separate high-carbon martensite into several small domains, thereby improving the cooperative deformation capability between M/A islands and bainitic ferrite matrix, and delaying the microcrack initiation near

M/A islands.

Table 1 As-rolled mechanical properties of three experimental steel plates

Symbol	Structure	$w(F_{GBA})$ / %	σ_b / MPa	$\sigma_{0.2}$ / MPa	δ_5 / %	ψ / %	A_{kV} / J	180° bending
A	Bg	0	895	660	15	39	31(20℃), 13(−40℃)	$d = 3a$
B	F_{GBA}/Bg	15	813	515	19	52	43(20℃), 23(−40℃)	$d = 2a$
C	F_{GBA}/Bg	18	875	575	18	41	76(20℃), 34(−40℃)	$d = 2a$

Note: d—bending head diameter, a—sample thickness; A_{kV}—charpy V-notch impact energy, w—mass fraction.

Table 2 As-rolled instrumented impact results of three experimental steels

Symbol	E_t / J	E_I / J	E_P / J	P_{max} / kN
A	30.96	27.86	3.10	21.51
B	45.97	37.94	8.03	20.34
C	62.24	35.72	26.52	21.72

Note: E_t—total energy; E_I—crack initiation energy; E_p—crack propagation energy; P_{max}—maximum impact load.

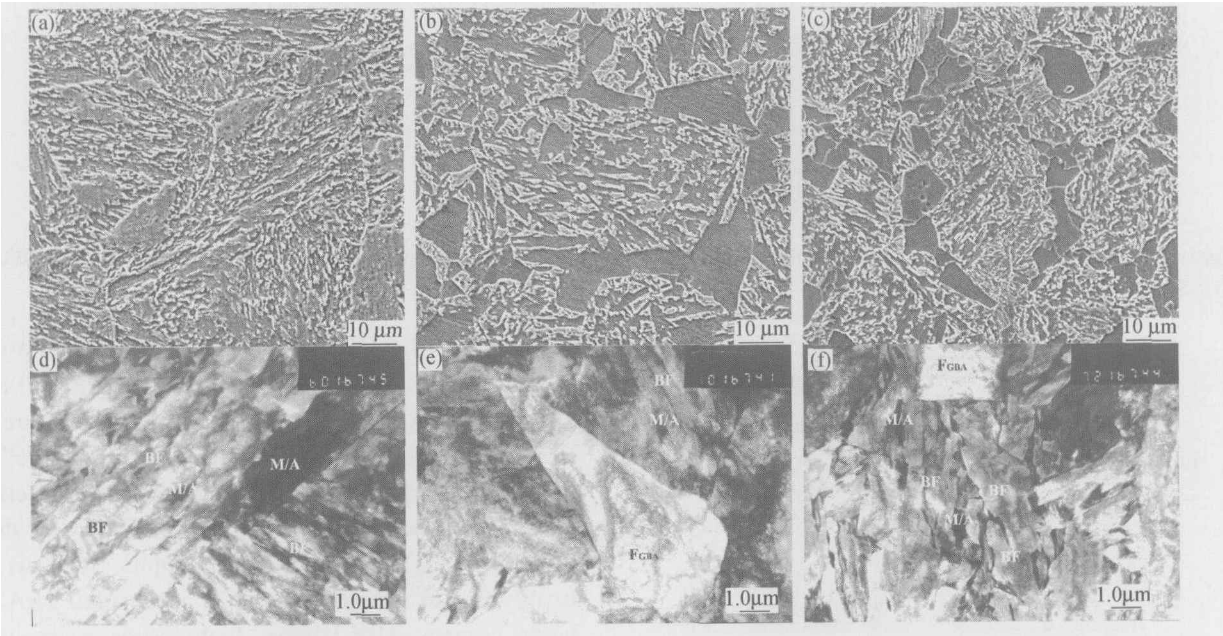

Figure 4 As-rolled microstructure of three experimental steels, (a)-(c) SEM microstructure of steel A, steel B and steel C respectively; (d)-(f) TEM bright field image of steel A, steel B, steel C respectively.

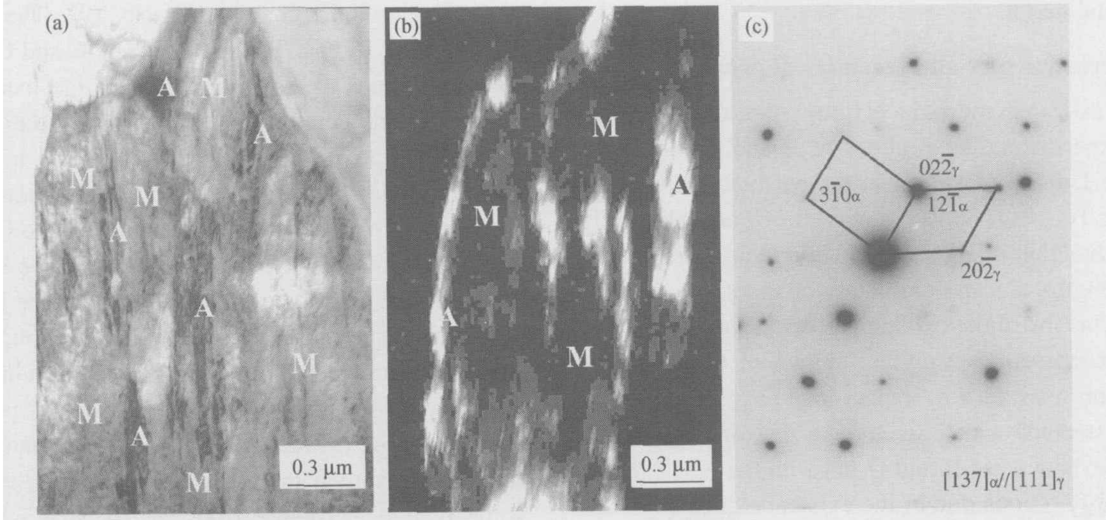

Figure 5 TEM fine structure of M/A island in the offline-stacked F_{GBA}/Bg duplex steel C, (a) bright field image of M/A island; (b) dark field image of M/A island; (c) its selected area diffraction pattern.

3.3 The effect of various microstructures on fatigue crack propagation behaviors

The propagation routes of fatigue microcracks in Bg and F_{GBA}/Bg duplex microstructure are shown in

figure 6. In granular bainite, the elongated M/A islands are distributed between the bainitic ferrite laths or along the prior austenite grain boundary. When a fatigue microcrack meets the granular bainite ferrite laths at a small angle as shown in figure 6(a), the resistance to crack propagate in its original direction is higher than that along the bainitic ferrite laths due to the existence of stronger elongated M/A islands, so that the microcrack may change its propagation direction, extend its propagation route. However, the main crack has to come back to its original direction by cutting off the bainitic ferrite laths and elongated M/A islands due to the existence of external main stress direction. Though the changes of microcrack propagation direction have some good effects on the toughness of Bg steel, the propagation route of main crack is much straight. Therefore, once the main crack has initiated and then begins to propagate forward, the local plastic deformation amount is too little to reduce the concentrative stress at the crack tip. As a result, the rapid propagation of the crack soon occurs, which obviously decreases the crack propagation energy, steel plasticity and toughness as shown in table 2.

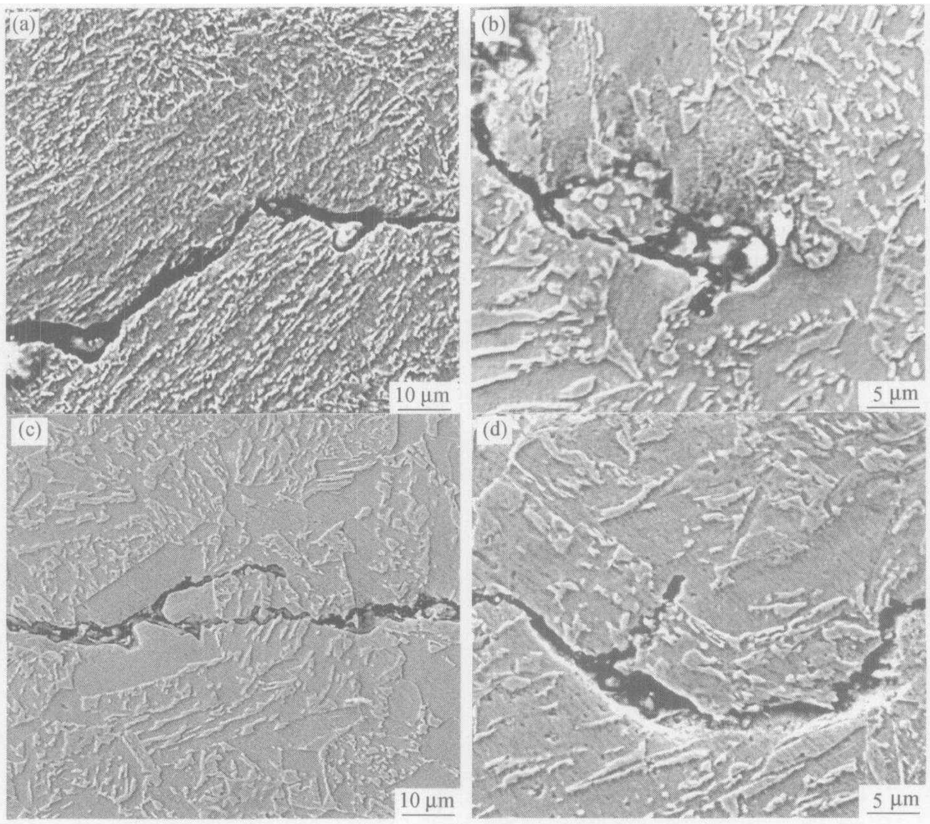

Figure 6 Effect of various microstructures on microcrack propagation behaviors, (a) Crack zigzag propagation route in single granular bainite (Steel A); (b) Blunting effect of F_{GBA} on crack tip in F_{GBA}/Bg duplex microstructure (Steel B); (c) Blunting effect of F_{GBA} on secondary crack tip and crack branching phenomena in F_{GBA}/Bg duplex microstructure (Steel B); (d) Blunting effect of F_{GBA} on secondary crack tip, branching and curving phenomena of crack propagation route in F_{GBA}/Bg duplex microstructure (Steel C).

In the F_{GBA}/Bg duplex microstructure, the crack propagation route is completely different. The softer phase F_{GBA} occurs large plastic deformation to relax local stress concentration of crack tip and reduce three-dimensional tensile stress, and it results in crack tip blunting as shown in figure 6(b). At the same time, some cracks may continue their propagation by crack branching or curving their propagation route to pass by the obstructions due to the existence of dislocation multiple slip or cross slip as shown in figure 6(c). Meanwhile, some cracks have to pass through many Bg and F_{GBA} repeatedly to continue propagating since F_{GBA} does not always connect with each other, contributing to the higher toughness. In addition, many secondary cracks also usually stop at F_{GBA} phase during the propagation process as shown in figure 6(d), which would consume some extra energy compared with Bg structure and be beneficial to steel toughness.

4 Conclusions

(1) The grain boundary allotriomorphic ferrite/granular bainite duplex microstructure has been proved to be a good approach of microstructure strengthening and toughening under the condition of conventional rolling and air-cooling, and the alloy design of duplex steels is also practicable with cheap

alloying elements.

(2) The grain boundary allotriomorphic ferrite in F_{GBA}/Bg duplex microstructure can blunt the microcrack tip, render crack propagation route branching and curving and increase the resistance to fatigue crack propagation.

(3) F_{GBA}/Bg duplex steel possesses better strength-toughness combination than single granular bainitic steel with the same nominal chemical composition.

(4) F_{GBA}/Bg duplex steel plate (12 mm-thick) obtained by offline-stacking at 250℃ and slowly-air-cooling process achieves good mechanical properties: σ_b, 870 Mpa; $\sigma_{0.2}$, 570 MPa, δ_5, 18%; $A_{kV(-40℃)}$, 34 J; together with low cost.

References

[1] ASTM A841/A841M-95, *Standard Specification for Steel Plates for Pressure Vessels, Produced by Thermo-Mechanical Control Process* (TMCP) [S], ASTM Committee on Standards, 1995.8.

[2] E.J. Czyryca, Advances in high strength steel technology for naval hull construction [J], *Key Engineering Materials*, 84-85(1993), p.491.

[3] A.J. McEvily, R.G. Davies, C.L. Magee, and T.L. Johnston, Structure, hardenability and toughness of low-carbon high-strength steels, [in] *Transformation and Hardenability in Steels* [C], Ann Arbor, Climax Molybdenum Company, MI 48105, 1967, p.179.

[4] K. Nagai, Creation of 800 MPa-class ferrite-pearlite steel for welded structures, [in] *3rd Workshop on Ultra Steels: Dreams and Tasks for Materialization* (in Japanese) [C], 1998, p.9.

[5] G.E. Dieter, *Mechanical Metallurgy* (Second Ed.) [M], Singapore: McGraw-Hill Book Co., 1976, p.216.

[6] L.J. Habraken and M. Economopoulos, Bainitic microstructures in Low-carbon alloy steels and their mechanical properties, [in] *Transformation and Hardenability in Steels* [C], Climax Molybdenum Co., Ann Arbor, MI, 1967, p.69.

[7] H.S. Fang and H.J. Deng, Granular bainite microstructure in low carbon Fe-Mn-B steel and its strength-toughness (in Chinese), *Materials For Mechanical Engineering* [J], 5(1981), No.1, p.5.

[8] L. Yang, H.S. Fang, and Z.H. Meng, Kinetics of austenitic isothermal decomposition and mn partition in Fe-C-Mn-B Alloys (in Chinese), *Acta Metallurgica Sinica* [J], 28(1992), No.1, p.A16.

[9] P.G. Xu, *Study on Grain Boundary Allotriomorphic Ferrite/Granular Bainite Steel* (in Chinese) [D], Tsinghua University, Beijing, 2002, p.57.