Effects of Sizes and Aggregate of Bainite Laths on Strength and Elongation in Bainitic Steels*

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Both strength and ductility in bainitic steels have been analyzed as functions of size of bainite lath nucleated within austenite grain, hereafter referred to as BWING, the number of BWINGS contained in an aggregate of bainite laths with almost parallel slip systems between neighboring bainite laths, hereafter referred to as ALPS, and related factors. 0.2% proof stress increased with decreasing size of BWING. Dislocation network could be introduced into austenite formed with isothermal holding at 1023 K before austenitization above 1673 K and would act as nucleation site for BWING, resulting in size of BWING being smaller. It could be suggested that almost parallel slip systems in neighboring BWINGS induced large homogeneous deformation, and dense BWINGS in an ALPS increased total elongation. As the number of BWINGS in an ALPS increases, many BWINGS having the larger Schmid factors would exist at many spots, resulting in many initiation sites for deformation and in inducing sufficient deformation of an ALPS. Conclusion is that main factors controlling 0.2% proof stress and total elongation are size of BWING, parallelism of slip systems in neighboring BWINGS, size of ALPS under a fixed size of BWING and the number of BWINGS within an ALPS.

Key Words: BWING, ALPS, Strength, Homogeneous Elongation, Local Elongation, Dislocation Network in γ , Pretreatment, Coherent Stress, Transformation Stress

1. Introduction

The controlling factors of strength and ductility in materials have been discussed for long period. In polycrystals, finer grains induce higher strength as suggested by Hall-Petch relation, $\sigma = \sigma_0 + kd^{-1/2}$, where σ is the yield strength, d the grain size, and σ_0 and k the constants. The above relation is greatly controlled with the propagation of deformation between neighboring grains. In order to propagate deformation between neighboring grains, increase in the strain at the grain boundary due to both multiplication and accumulation of dislocations is indispensable for the activation of dislocation source in the neighboring grain. In order to accumulate dislocations at grain boundary, external stress is necessary to overcome the back-stress from the accumulated dislocations. In the case of smaller grain size, the back-stress might become larger easily with small strain compared with the case of larger size of grain. It might lead higher yield strength.

On the other hand, if relaxation of deformation strain within a grain occurs through activation of deformation in neighboring grain without cracking at grain boundary due to accumulation of dislocations at a local position, both elongation and toughness might be improved^{1, 2)}.

Bainitic steel has been utilized to improve both strength and ductility, because bainite lath nucleated within austenite (γ) grain, hereafter referred to as BWING, has its small size, $\sim 0.1 \times 1 \times 5 \mu\text{m}^3$, and peculiar crystallographic orientation relationship, which might easily induce the relaxation of deformation strain in a bainite lath due to deformation of neighboring bainite lath. In the case of bainite laths at γ grain boundary (intergranular bainite), the aggregate of bainite laths forms a region of nearly single crystal, resulting in the decrease of both σ and elongation³⁻¹⁰⁾.

The homogeneous elongation would be controlled by the relaxation of deformation strain in a BWING with glide of dislocations in neighboring BWINGS. Under good parallelism of slip systems, both slip plane and slip direction, between neighboring BWINGS, the relaxation of the deformation strain might be enhanced smoothly, and elongation would be increased. It is important that bainitic transformation induces the good parallelism of slip systems between neighboring BWINGS⁷⁾. As for local elongation, the propagation of crack should be prevented. The aggregate of finer BWINGS composed of various variants is preferable to prevent the propagation of crack, resulting in the increase in local elongation.

In the present investigation, the aggregate of BWINGS with good parallelism of slip systems between neighboring BWINGS, hereafter referred to as ALPS, is also noticed to improve ductility. Both strength^{11, 12)} and ductility have been known to be controlled by the sizes of BWING and ALPS³⁻¹⁰⁾. It has been suggested that microstructure is one of substantial controlling factors also from a standpoint of crystallography for the mechanical property.

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Therefore, the microstructure consisting of BWINGs was prepared using large austenite (γ) grain as well as pretreatments before austenitization. Fine BWINGs inducing high strength with good parallelism of slip systems between neighboring themselves are investigated how effective for both the strength and ductility in bainitic steels.

2. Experimental procedure

2.1 Preparation of specimens and procedures for both heat treatment and microstructure analysis

Chemical compositions of the steels are listed in Table 1. The steels containing poor sulfur, hereafter referred to as SP, and rich sulfur, hereafter referred to as SR, were prepared to examine effects of sulfur on mechanical properties in steels. In order to enhance the nucleation of BWING, dislocation network was introduced into γ using pretreatments before austenitization, and the temperature of austenitization was set at a high temperature of 1673 K to enlarge grain size of γ for both making bainite lath nucleate inside γ grain and suppression of nucleation of bainite lath at γ grain boundary by decrease of area of γ grain boundary.

Table 1 Chemical compositions of the steels (mass%).

Steel	C	Si	Mn	P	S	N	O
SP-1	0.075	0.25	1.45	0.001	<0.001	0.003	0.002
SP-2	0.14	0.29	1.44	0.001	0.001	0.0004	0.002
SR	0.13	0.29	1.44	0.001	0.062	0.0003	0.002

Size of ALPS and/or the number of BWINGs being contained within an ALPS were controlled by the pretreatment. The size of ALPS was determined based on analyses of slip systems in neighboring BWINGs using electron back-scattered diffraction pattern, hereafter referred to as EBSP. The parallelism between slip systems in neighboring BWINGs, which could increase homogeneous elongation in steel, was analyzed using EBSP. The number of BWINGs within an ALPS, which might control both homogeneous and local elongations in steel, was also analyzed with EBSP.

Microstructures were observed with optical microscope and scanning electron microscope (SEM).

2.2 Pretreatment before austenitization

The process of pretreatment (—) before austenitization is illustrated in Fig. 1. Pretreatment was done for introducing dislocation network into γ , which scarcely absorbs vacancies even in the γ region and acts as nucleation site for BWING.

Tensile tests were performed at room temperature under a strain rate of 5×10^{-4} /s using the test piece as shown in Fig. 2.

Fig. 1 Schematic illustration for heat treatment process. Austenitization at 1673 or 1723 K was done for the suppression of nucleation of bainite lath at γ grain boundary. The solution treatment at 1673 K keeps elongated MnS, which had existed in the as-supplied steel. The increase in the temperature for austenitization to 1723 K is making MnS decompose almost completely into γ . Holdings at 1023 and 773 K are for formations of γ at ferrite (α) grain boundary and BWING in γ , respectively.

Fig. 2 The tensile test piece 2 mm thick. Tensile tests were done at room temperature under a strain rate of 5×10^{-4} /s. All values in the figure have a unit of mm.

3. Result and discussion

3.1 Introduction of dislocations into γ

Cold rolling hardens ferrite (α), whose hardness would be kept during isothermal holding at 1023 K, because the duration for the isothermal holding (for reverse transformation, $\alpha \rightarrow \gamma$) is so short about 100 s. As shown in Fig. 3, reverse transformation, $\alpha \rightarrow \gamma$, occurs mainly from α grain boundary and/or annealing-twin boundary. Newly formed thin γ region might be deformed by both coherent stress generated at γ/α boundary and $\alpha \rightarrow \gamma$ transformation stress. Dislocations might be introduced into the thin γ region. The introduction of dislocations into γ might be enhanced under the cold-rolled α , because the both coherent stress and $\alpha \rightarrow \gamma$ transformation stress are hardly relaxed with

Fig. 3 SEM micrographs taken from (a) as-received and (b) pretreated (isothermal holding at 1023 K for 100 s) and quenched steel SP-1. In (a), α grain boundaries and annealing-twin boundaries within α grain exist. White contrast (indicated with arrow) in (b) corresponds to martensite region, where had been γ region due to the pretreatment.

deformation of α , namely, dislocations could be easily introduced into the thin γ region through the deformation of γ .

In the case of quenching after isothermal holding at 1023 K, martensite would fill the thin γ region in Fig. 3 and the reverse transformation would occur quickly only in the thin γ region, because of the pipe diffusion along high density of dislocations in martensite. Also in this case, the thin region quickly transformed to γ might be deformed due to both the coherent stress and $\alpha \rightarrow \gamma$ transformation stress. As shown in Fig. 4, α is hardened with cold rolling. Reverse transformation, $\alpha \rightarrow \gamma$, might start at grain boundary and/or twin boundary. Dislocations are introduced with both coherent stress at (γ/α) -interface and transformation stress into thin and soft γ to relax the stresses. The hardened α would be difficult to relax the both stresses. Intersection and reaction between dislocations in γ might form their network. The network would be possible to be stable even in the γ region.

Fig. 4 Formation process of dislocation network in austenite (γ). Thin γ is formed along ferrite (α) grain boundary.

3.2 Formation of dislocation network in γ ¹³⁾

It could be considered that dislocations form their network and/or dislocation cell wall through their reactions and/or accumulation of dislocations around the dislocation network. The dislocation network would be stable even in the high temperature region of γ , because the dislocation segment constituting the network have different edge-component of Burgers vector (b_e), resulting in each of the dislocation segments constituting dislocation network having different speed of climbing motion due to absorption of vacancies. Under such condition, the bowing of the dislocation segment constituting the network might occur, and energies of themselves might increase. Therefore, dislocation network scarcely absorbs vacancy, and the climbing motion of dislocation network could be suppressed, resulting in the dislocation network being stable even in the high temperature region of γ . Some of multiplied dislocations might be tangled with the dislocation network, and form a kind of dislocation-cell wall, resulting in inducing the variation of shape of dislocations as well as size of b_e .

3.3 Nucleation of BWINGs and homogeneous elongation

As shown in the Fig. 7 of Ref. 7, it could be suggested that dislocation network was formed in γ . Furthermore, BWING might nucleate at (BWING/ γ)-interface continuously to relax the transformation stress around (BWING/ γ)-interface (cooperative transformation)^{7, 9, 10)}. In this case, slip systems in neighboring BWINGs are almost parallel^{7 - 10)}. Homogeneous elongation would be increased by parallelism of slip systems A and B (both slip plane and slip direction) in Fig. 5. In Fig. 5, when slip system A operates, shear strain generates at the interface, $\underline{ab}$, resulting in the interface between laths 1 and 2 being $\underline{acdb}$. When slip systems A and B are nearly parallel each other, the deformation at the interface occurs smoothly, resulting in the increase in the area of interface from $\underline{ab}$ to $\underline{acdb}$. It leads that dislocation density per unit area of interface decreases with increasing in the area of interface. Accumulation of dislocations at interface could be suppressed, resulting in the suppression of cracking. Then, size of ALPS should be large under size of BWING being constant to increase in homogeneous elongation⁷⁾. During homogeneous deformation, cracking scarcely forms and BWINGs elongate as suggested in Fig.5. Stress vs. strain curves taken from steel SP-1 pretreated at 1023 K for (a) 15 and (b) 100 s are shown in Fig. 6. It is clear that 0.2 % proof stress, tensile strength, homogeneous elongation and total elongation of the steel SP-1 pretreated under (b) is superior to those pretreated under (a). Average thickness of BWINGs in the steel pretreated under (b), 2.1 μm , is thinner than that under (a), 2.5 μm .

Fig. 5 Illustration of deformed neighboring bainite laths. Shear strain of Lath 1 is relaxed with deformation of neighboring bainite Lath 2, in the case of slip systems, A and B, being almost parallel each other. Original interface is $\underline{ab}$, and deformed interface is $\underline{acdb}$. $\underline{acdb} > \underline{ab}$.

Fig. 6 Stress vs. strain curves taken from the steel SP-1 pretreated at 1023 K for (a) 15 and (b) 100 s followed by quenching into iced brine. 0.2% proof stress, tensile strength, homogeneous elongation and total elongation in the steel pretreated (b), 538, 709 MPa, 22.7, 38.1 %, respectively, are larger than those in the steel pretreated (a), 524, 650 MPa, 16.0, 26.1 %, respectively. The mechanical properties in the steel pretreated (b) are superior to those in the steel pretreated (a).

It could be suggested that dimple on fracture surface corresponds to ALPS¹⁰⁾. Average sizes of dimples pretreated under (a) and (b) are almost same, $\sim 3.2 \mu\text{m}$. Therefore, the number of BWINGs contained in an ALPS pretreated under (b) is larger compared with that pretreated under (a).

3.3 Major factor controlling elongations

It could be suggested that the number of BWINGs contained in an ALPS is one of major factors, which controls homogeneous- and local-elongations. In order to increase homogeneous elongation, localization of dislocations should be suppressed. It has been clarified that size of dimple on ductile fracture surface is larger than that of a BWING, and nearly equal to that of an ALPS by EBSD measurements under the condition of the angle, θ ,

between slip systems A and B in Fig. 5 being 15° . These results suggest that cracking might generate inside an ALPS through dislocation tangling and edge of dimple is almost coincident with the boundary of ALPS. As increasing in the number of BWINGs within an ALPS, the number of BWINGs having large Schmid factor, that is, being easy to deform under a constant shear stress, might increase. In order to deform until getting a fixed strain, dislocations have to sweep a fixed area of slip planes. In the case of many BWINGs having large Schmid factor within an ALPS, the large number of BWINGs are swept by dislocation glides. Hereafter, this case is referred to as Case of Many BWINGs. However, in the case of the small number of BWINGs having large Schmid factor within an ALPS, hereafter this case is referred to as Case of Small BWINGs, dislocation glides might be localized to the small number of BWINGs, namely, the smaller region compared with the Case of Many BWINGs. In order to get a fixed strain, the areas being swept with dislocation glides are nearly same in both the cases.

Then, in the Case of Small BWINGs, dislocations are easy to localize to the BWING having large Schmid factor and to around the BWING, namely, to the small region, because dislocations glide mainly at/around the BWING having large Schmid factor. It leads that localization of high density of dislocations might induce crack, and decrease homogeneous elongation in the Case of Small BWINGs as shown in Fig. 7.

Fig. 7 Illustration of typical cases of ALPSs. The ALPS is the aggregate of BWINGs. The ALPSs consisting of relatively (a) large and (b) small BWINGs having the largest (●), middle (●) and small (○) Schmid factors are illustrated. In (b), many BWINGs having the largest Schmid factor exist, inducing the BWINGs having the middle Schmid factor through phase transformation, $\gamma \rightarrow \text{BWING}$. Comparing with both cases, (a) and (b), occupied area with BWINGs having the largest and middle Schmid factors is larger in (b) than that in (a). It suggests that dislocation density would be homogeneous in (b), compared with the case in (a). In (a), high density of dislocations might localize, resulting in inducing crack.

3.5 Local elongation

Local elongation after necking may be controlled by the propagation of cracking. As mentioned above, in the Case of

Many BWINGs, many kinds of variants would be contained within an ALPS, resulting in the suppression of propagation of crack. The bainitic steel consisting of the ALPS under the Case of Many BWINGs is superior in both homogeneous- and local-elongations.

3.6 Effects of sulfur or carbon on mechanical property

In steel SR, MnS exists remarkably. It was obviously observed that BWINGs nucleate at MnS and grow radially at every fixed angle from MnS. It suggests that BWING grows along fixed habit planes. In high density of MnS steel, 0.2 % proof stress was relatively low compared with that in low sulfur steel, and crack easily nucleates at the interface between MnS/BWING^{7, 9)}.

In this case, nucleation of BWING at MnS may be fast, resulting in making BWINGs grow fast and having large size of BWINGs as well as decreasing 0.2 % proof stress. However, the case that total elongation is relatively equal compared with low sulfur steel exists⁷⁾. It could be suggested that large size of BWINGs having large Schmid factor were formed at many spots.

As for the effect of carbon, the existence of carbon at grain- and/or twin-boundaries enhance the reverse phase transformation, $\alpha \rightarrow \gamma$, because in order to induce the $\alpha \rightarrow \gamma$, carbon has to accumulate to grain- and/or twin-boundaries up to the concentration of ~ 0.5 wt. %. Depending on carbon concentration, amount and/or thickness of γ would be different. In order to introduce dislocation network into γ , a critical thickness of γ is necessary, because dislocations having different Burgers vectors have to tangle and react with each other. Therefore, dislocations on different slip systems have to glide to some extent.

The thickness of γ formed through the reverse phase transformation, $\alpha \rightarrow \gamma$, is important factor to control the density of dislocation network in γ . The carbon concentration, therefore, is one of controlling factors for nucleation of BWINGs, namely, size of ALPS.

4. Conclusions

1. Thin γ region formed through reverse phase transformation, α or $\alpha' \rightarrow \gamma$, could be deformed by coherent stress around the γ/α interface and transformation stress, resulting in the formation of dislocation network in γ .
2. The dislocation network would be stable even in γ , because suppression of climbing motion due to absorption of vacancies might be caused by the difference of b_e of component dislocations in the dislocation network.
3. It could be suggested that piling dislocations up locally at interface between BWINGs within an ALPS scarcely occur, resulting in the increase in homogeneous elongation.
4. Many BWINGs having large Schmid factor should exist

within an ALPS to suppress localization of dislocations. The ALPS consisting of many BWINGs is greatly effective for the improvement of both homogeneous- and local-elongations.

5. Localization of dislocations through homogeneous deformation inducing crack might be formed inside each ALPS, because dimples on fracture surface is larger than a BWING.
6. Flow stress might be increased with decreasing size of BWING. Under keeping a fixed strength, size of ALPS should be larger to obtain large elongation. An ideal bainitic steel should consist of one ALPS having small BWINGs.
7. The BWING nucleated at MnS/ γ -interface would grow fast, because of its speedy nucleation. The grown bainite laths with large Schmid factor increases homogeneous elongation.
8. The critical thickness of γ is indispensable to make dislocations react with each other and form dislocation network.

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