Hot Ductility in Nb-bearing High-strength Low-alloy Steels*

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Synopsis
The effect of Nb content in HSLA steels on hot ductility was investigated
under various thermal cycles. The increase of Nb content was found to
deteriorate hot ductility in all the thermal cycles investigated. The change
in hot ductility with temperature gave rise to only one trough in the tempera-ture region from 700 to 900 °C, which covered the lower temperature region of the austenite (I), and the Y-a region (II). The strain concentration in the ferrite primarily reduced hot ductility in the region II, and the increase of Nb content reduced hot ductility remarkably in region I, where grain boundary precipitates such as Nb(CN) primarily controlled hot ductility. The latter fact was confirmed through the studies of fractography and reversibility of hot ductility behavior. The improvement of hot duc-tility in the Nb-bearing steels was shown to be attained by adoption of the thermal cycle corresponding to hot charging process of the ingot or the slab, and also by the decrease of N content or the addition of the small amount of Ti. Relation between hot ductility and hot deformation mechanism, such as dynamic recrystallization or grain boundary sliding, is discussed.

I. Introduction

It has become increasingly important and neces-sary to prevent the surface cracking of slabs from large ingots or from continuous casting (CC) in order to

practice the direct rolling or the hot charging for energy saving, where the conditioning of the slab by means of the cold scarfing cannot be taken. Niobium which is very popularly used in the high strength low alloy steel is known as one of the alloying elements to

promote hot cracking susceptibility of the ingot or the slab. Cause of the cracking of the large ingot of Nb-bearing steels, experienced in the past, has been at-tributed to the large inclusion type precipitates of Nb(CN) formed in the center portion of the ingot.'-5) However the surface defects or crackings of the slab should be originated from the surface layer of the ingot or the slab where such eutectic precipitates of Nb

(CN) are not formed, and actually the cracking just beneath the surface in the CC slab is observed in the

Nb-bearing steels.6) Therefore another cause of poor hot ductility in these steels has to be considered. Many studies have been undertaken about the effect of Al

on hot ductility,'-12> and it has been made clear that

precipitation of A1N in austenitic grain boundary or proeutectoid ferrite causes the decrease of hot duc-tility. Although Nb is considered to have the similar influence, very few studies have been carried out.13-15) The objective of this study is to examine hot duc-tility of the Nb-bearing steels by hot tensile testing in the course of various thermal cycles, and to make clear the mechanism which controls hot ductility. The thermal cycles were determined from the cycles that

actually occurred in the ingot or the CC slab of the Al-killed steels. Possible ways for improving the hot ductility in the Nb-bearing steels were examined from two aspects-control of the thermal cycle, and control of chemical composition such as decrease of the N content or Ti addition.16) Regarding the relation be-tween hot ductility and hot deformation mechanism, behavior of grain boundary sliding was investigated systematically together with the fractographic studies of the fracture surface.

II. Experimental Procedures

Chemical compositions of the steels used are shown in Table 1. These were melted by a high frequency induction heating furnace and poured into the 50 kg ingots. Hot tensile specimens were prepared from the as-cast ingots. High temperature tensile testing was carried out in a hot working simulator equipped with oil pressure servomechanism. In this equipment strain rate is variable from 10-4 to l0/sec, with the maximum load of 2 000 kg. The various thermal cycles of heat-ing and cooling as well as the deformation conditions were programmed and tensile testing was carried out by following the program. Specimens of 8 mm di-ameter were heated under vacuum by a high frequen-cy induction coil of 30 mm length. Helium gas was used for cooling of the specimen during the thermal cycles and after tensile testing. The load, displace-ment and temperature were recorded with a usual

pen recorder or with an electromagnetic oscillograph for the higher strain rate deformation above 10-1/sec. The three kinds of the thermal cycles employed in this investigation are shown schematically in Fig. 1. From the microstructural observations in the speci-mens quenched after holding for 5 min between 700 °C and 800 °C and also after tensile testing in the thermal cycle (A), it was made clear that the proeutectoid fer-rite was not formed above 800 °C in all steels. There-

fore the temperature above 800 °C is designated here as r (austenite) and below 750 °C is r+a (ferrite). The correspondence of these thermal cycles to the cooling and reheating stages of the ingot or the CC slab is as follows. The cycle (A) corresponds to the cooling stage of the ingot or the CC slab after solidi-fication. The cycles (B) and (C) are the reheating stage for hot and cold charging in the reheating fur-nace, respectively. In the case of the hot charging, the temperature of the surface layer of the ingot or the CC slab may decrease down to the lower temperature

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region of r or to the r+a region, and the cycle (B) represents these thermal cycles. On the other hand, the cycle (C), where the solution treatment at 1 300°C is not involved, enables one to investigate hot ductility during reheating of the cold ingot.
Tensile testing was performed with the strain rate of 4 X 10-3/sec, except the experiment for investiga-tion of the effect of strain rate on hot ductility, where it was varied from 10_4 to l0/sec. Hot ductility was measured from the variation of the value of reduction

of area (RA). The specimen used for the measure-ment of the grain boundary sliding had a flat polished

part on the surface of the round shape specimen, which was scratched by a diamond needle. Observa-tion of the fracture surface were carried out by a scan-ning electron microscope (JEOL-U3), while an ana-lytical electron microscope (JEOL-200 CX) was used for identification of the precipitates on the ex-tracted replica taken from the fractured surface.

III. Results

1. The Effects of Nb Content on Hot Ductility
The effects of Nb content on hot ductility in the thermal cycle (A) are shown in Fig. 2. Hot ductility decreased with the increase of Nb content particularly in the lower temperature region of austenite r, because of the formation of Nb(CN) in the r grain boundaries

(see Section III. 3). The results of the testing in the r+a region (750 °C) indicate that hot ductility was not influenced so much by Nb content, but was reduced due to the formation of the very thin proeutectoid fer-rite along the r grain boundaries. While hot ductili-ty showed recovery at 700 °C, it again exhibited de-

pendence on Nb content; the increase in Nb deterio-rated hot ductility. Figure 3 (A) shows the effect of carbon content on hot ductility in the 0.03 % Nb steel. Variation of carbon content from 0.08 % to

0.20 % had little influence in hot ductility in the whole range of the testing temperatures. It was also con-firmed that the change of Mn content from 0.65 % to 1.50 % or Si content from 0.04 % to 0.40 % in the

Nb-bearing steels had little influence on hot ductility at the temperature above 800 °C, although the in-crease of Mn or Si content slightly decreased hot duc-tility below 750 °C. B. Mintz and J. M. Arrowsmith reported that poor hot ductility of the Nb-bearing

Table 1. Chemical composition of the steels. (wt%)

Fig. 1. The thermal cycles used for investigation of hot ductility.

Fig. 2. The effect of Nb content and
temperature on the tensile
the thermal cycle (A).

deformation
properties in

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steel might be influenced by phosphorus.15~ There-fore the effect of P content was investigated under the thermal cycle (A), where P content was varied from 0.004% to 0.026% in the 0.14%C-1.45%Mn-0.035%Nb-0.075%V steel. It is clear from Fig. 3 (B) that P content does not affect hot ductility of the Nb-bearing steels.
The results of the thermal cycle (B) are shown in Figs. 4 and 5. These cycles, where the temperature was once dropped to 800 °C or 700 °C after reheating at 1 300 °C, caused a prominent hot ductility reduc-tion compared with the results of the cycle (A) as shown in Fig. 4. Hot ductility of the Nb-free steel decreased for the holding temperature of 700 °C, but not so much for 800 °C. On the other hand, hot duc-tility of the Nb-bearing steels decreased for the both holding temperatures, particularly for the holding temperature of 800 °C. A complete recovery of hot ductility was observed at 1 100 °C.
In the above investigations, the tensile testing tem-perature was varied. Figure 6 shows the effects of the thermal cycles and the Nb addition on hot duc-tility at a fixed temperature. The tensile testing tem-perature of 1 000 °C was chosen, because very high ductility was obtained at this temperature in the ther-mal cycle (A) even in the Nb-bearing steels. The Nb-free steel exhibited very high ductility for the holding temperature from 700 to 900 °C, but hot ductility in the Nb-bearing steels decreased with decreases of the holding temperature. That is, the hot ductility at constant testing temperature and strain rate is in-fluenced by variation of the metallurgical factors caused by the prior thermal cycle. The metallurgical factor in the present case is precipitation of Nb (CN) or A1N on the r grain boundary or in the proeutectoid ferrite. The precipitation of Nb(CN) or A1N takes place in the ferrite formed along the r grain boundary during holding at 750 °C or 700 °C, and the prefer-ential precipitation of Nb(CN) on the r grain bound-ary also occurs during holding in the lower tempera-ture region of r from 800 to 900 °C. The size and the amount of the precipitates will depend on the con-dition of the thermal cycle and the Nb content. The

precipitates of Nb(CN) remaining undissolved in r on reheating at 1 000 °C primarily control hot ductility at this temperature. The Nb-free steel with a given content of Al and N causes similar precipitation and re-dissolution behavior of A1N. The results from

Figs. 4 to 6 indicate that the precipitates of A1N can re-disslove into r on reheating at 1 000 °C much more rapidly than Nb(CN), although AIN tends to precipi-tate in the low temperature region of r and also in the r+a region similarly to Nb(CN).

Fig. 3. The effect of (A) C and (B) P contents on hot duc-tility in the steel with 0.03 % Nb and 0.035%Nb-0.075%V, respectively.
Fig. 4. Comparison of hot
(A) and (B).

ductility in the thermal cycles

Fig. 5. The effect of Nb
thermal cycle (B).

content on hot ductility in the

Fig. 6.
The effect of Nb content on
hot ductility in the thermal
cycle (B) where the holding
temperatures were varied.

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The effect of the holding time in the thermal cycle

(B) was investigated in the 0.054 % Nb steel. The holding time at each dropped temperature and also the holding time before tensile testing at the repeat-ing temperature was varied as shown in Figs. 7 (A) and (B), respectively. As shown in Fig. 7 (A), the in-crease of the holding time in the temperature region of r (800N900 °C) deteriorated hot ductility after reheating at 1 000 °C, which appear to be due to the increase of the Nb(CN) precipitates in the r grain boundary during the longer holding time. The in-crease of the holding time at 750 °C improved hot duc-tility, because the increase of the volume fraction of the ferrite produced a larger volume fraction of refined

r grains after reheating at 1 000 °C. This decreased the tendency of intergranular fracture along the r

grain boundary. As shown in Fig. 7 (B), the change of the holding time before tensile testing little influ-enced hot ductility, when the testing temperatures were below 1 000 °C. However, the holding time at 1 050 °C resulted in peculiar changes of hot ductility, which appears to be caused by the coarsening and dis-solution of Nb (CN) at this temperature. That is, the

prominent coarsening of Nb (CN) precipitates that started at this temperature initially deteriorated hot ductility, and then ductility was improved by re-dis-solution into r that took place at the longer holding time.

The result of the thermal cycle C is shown in Fig.
8. The reheating temperature was varied to cover the range of the a, r+a and r phases. These changes of the testing temperature and the Nb content re-sulted in the variation of not only the strength and the microstructure but also the amount and size of the

precipitates such as A1N or Nb (CN). The increase of the Nb content decreased hot ductility at each tem-

perature. Particularly the increase of the undis-solved precipitates of Nb(CN) in r deteriorated hot ductility as shown in the results of 900 °C.

2. The Effect of Al and N Contents on Hot Ductility Figure 9(A) shows the effects of N content on hot ductility of the Nb-bearing steels in the thermal cycle (A). Figure 9(B) shows the effect of a small amount of Ti addition on the similar steels that was enough to scavenge the nitrogen as TiN. The decrease of N content and Ti addition improved hot ductility par-ticularly below 900 °C. Figure 10 shows the similar results in the Nb-free steels, where Al and N contents were varied. The increase of Al content up to 0.080 % deteriorated hot ductility below 850 °C. It is clear that the amount of A1N precipitates controls hot ductility of the Nb-free steel. With regard to hot ductility in the lower temperature region of r, the duc-tility trough of the higher Al steel was observed in a narrow temperature region compared with that of the higher Nb steels (see Figs. 2 and 10(B)). This ap-

pears to be due to the difference of precipitation be-havior of A1N and Nb (CN) in the r region.
The recent studies by one of the authors made clear that NbCo,S5 is formed in the Nb steels with lower N
content around 0.0020 % or with 0.020 % Ti addi-tion, while NbCo•soNo.25 was formed in the steel with N content higher than 0.0050 %. These different

Fig. 7. The effects of holding time on hot ductility 0.054 % Nb steel. The holding temperatures
(A) and (B) were shown in the thermal cycle.

of
of

Fig. 8. The effect of Nb content on hot thermal cycle (C).

ductility in the

Fig. 9. The effects of (A) N content and (B) Ti hot ductility of Nb-bearing steels.

addition on

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compositions of the Nb precipitates may cause the dif-ferent precipitation kinetics in r, that is, NbCo.s0No.25 may precipitate more easily than NbC0,85 in the lower temperature region of r. Besides this, the addition of Ti has another effect. The stable precipitates of TiN

caused the refinement of the r grain size which was attained by reheating at 1 300 °C. Both factors low-er the density of precipitates in the r grain boundary that area formed in the lower temperature region of r. All of these factors contribute to improve hot duc-tility of the Nb-bearing steels. The important obser-vation in Figs. 9 and 10 was that improvement of hot ductility in the r-a region (750 °C) due to the decrease of N content or Ti addition was not nearly as great as the improvement observed in the lower temperature region of 1.

3. Fractographic Studies
The fracture mode observed in region I (lower temperature region oft) was intergranular. The var-iation of the RA values due to the various parameters was directly proportional to the fraction of grain boundary area on the fracture surface. The charac-teristic feature of this fracture mode observed by scan-ning electron microscope was the wavy pattern on the intergranular fracture surface as shown in Photo. 1. Observation of the fracture surface in the higher mag-nification was performed by means of the carbon extraction method, using an analytical electron

Fig. 10. The effects of (A) N content and (B) Al content on hot ductility in Nb-free steels.

Photo. 1.
Intergranular fracture appearance
observed by SEM. 0.054 % Nb
steel was fractured at 850 °C in the
thermal cycle (A).

Photo. 2.
Fracture surface by means of carbon
extraction replica.
Nb (CN) precipitates shown by
arrow were analyzed by analytical
electron microscope.
Cu peak on EDS chart appeared
from filter.

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microscope that enabled one to identify the very fine precipitates. Photograph 2 represents an example, where very fine precipitates around 200 to 500 A were observed, and these were identified as Nb(CN) by EDS. The precipitates observed in the Nb-free steel were A1N, which were around 500 to 1 000 A. On the other hand, the fracture mode in the region II (r-a region) was shallow dimple fracture, although it was macroscopically intergranular. Hot ductility in this case was primarily influenced by thickness of proeutectoid ferrite formed along the r grain boundary. As shown in Fig. 11, it was varied by changing the holding time before tensile testing at 750 °C in the thermal cycle (A) in the 0.032 % Nb steel. The thin-ner ferrite layer gave rise to a larger reduction of hot ductility. Photograph 3 shows the fracture surface observed by SEM. It was found that depth of each dimple increases with the increase of ferrite thickness, while normal dimple fracture occurs in the thicker ferrite layer. As the void formation in the shallow

dimple fracture originated from the precipitates, the increase in density of the precipitates such as Nb(CN) promotes the void formation and link up. However, hot ductility at 750 °C where the worst hot ductility in the r-a region was observed, was influenced rela-tively little by the content of Nb, N or Al. That is, the primary factor controlling hot ductility in the r-a region is the thickness of the ferrite layer (or volume fraction of ferrite) ; poor ductility is caused by non-uniform deformation between the r and a phases due to their strength difference at a given temperature. A non-uniform deformation behavior in the r-a region depends on strain rate, which remarkably affects hot ductility in this region as shown in the following section.

Iv. Discussion

1. Relation between Hot Ductility and Hot Deformation Mechanism

The characteristic feature of the hot ductility in the Nb-bearing steels was that the poor hot ductility was caused in the lower temperature region of r (800 900 °C, region I) in addition to the r-a region (7O0'- 750 °C, region II), although hot ductility variation with the temperature showed only one trough which was continuous in regions I and II. The largest re-duction of hot ductility in the carbon or the Al-killed steel was observed in the region II, but the in-crease of the Nb content in the HSLA steels was found

to shift the temperature of the minimum hot ductility to the region I (800 °C) as shown in Fig. 2. While the mechanism of reduced hot ductility should be surely different in the regions I and II, most of the past

Fig. 11. The
0.032

effect of ferrite thickness
% Nb steel.

on hot ductility of

Photo. 3. The change of d imple pattern with the increase of ferrite thickness.

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studies have been concerned with the region II,17,18) in which the mechanism is relatively clear as described in the previous section. Therefore discussion here is mostly focused on the case of region I.
Several mechanisms for intergranular fracture at the higher temperature have been proposed. They are related to the hot deformation mechanisms such

as dynamic recrystallization14) or grain boundary slid-ing,23) both of which are strain-rate dependent pro-cesses. Therefore the effect of strain rate on hot duc-

tility was investigated in the Nb-free and Nb-bearing steels under the thermal cycle (A). As shown in Fig. 12, strain rate was widely changed, particularly at 800 °C. Hot ductility of both steels was improved with increase of strain rate, but the temperature of the minimum of the hot ductility trough did not change. The improvement of hot ductility in the region II with the increase of strain rate indicates that it suppresses strain concentration in ferrite during deformation, re-sulting in homogeneous deformation in both phases r and a.

With regard to hot ductility in the region I, G. Bernard et a1.14) proposed that a reduced hot ductility in the Nb-bearing steel was due to suppression of dy-namic recrystallization. It has been well confirmed that a critical strain (E~) is necessary for the onset of dynamic recrystallization, which corresponds to a peak strain in the stress-strain curve. The nucleation of

dynamic recrystallization originates in the local bulg-ing of the grain boundary, and then the recrystallized grains cover the whole of the original boundary as dynamic recrystallization progresses.20) This phe-nomenon appears to be beneficial to suppress the in-tergranular fracture. The increase of the Nb content retards dynamic recrystallization so as to increase ec,20'21) and apparently to reduce the hot ductility. However the results of the present studies do not sup-port this hypothesis. First, the true fracture strain (E f) for poor ductility in the region I is much smaller than e~. Secondly, increase of strain rate increases e~,20) nevertheless it improves hot ductility. Thirdly,

hot ductility of the steels widely changed with the thermal cycles under a given deformation condition and a given r grain size, which primarily control dy-namic recrystallization behavior. As reported by Norstrom,22) hot ductility would be influenced by dy-namic recrystallization only for the case of E f>e~. Nb suppresses not only dynamic recrystallization but also dynamic recovery in the strain range below e~ and the latter may be one of factors promoting in-tergranular fracture for the case of e f < e~.
The austenitic grain boundary sliding behavior was investigated in the several steels under the thermal cycle (A). The strain of grain boundary sliding (egb) was obtained by means of a method taken by A. W. Mullendore and N. J. Grant.23) When the strain of tensile testing (et) was varied from 0.026 to 0.18 with strain rate of 6 x 10-4/sec and at the temperature of 900 °C, the value of egb/et decreased from 0.40 to 0.15. The effects of strain rate, temperature and chemistry of the steels on grain boundary sliding were investigated under a given et of 0.26. Figure 13 gives one of these results, where the effects of strain rate on egb/et and hot ductility in the 0.054 % Nb steel are shown. The increase of strain rate accompanying with im-provement of hot ductility decreases Egb'Et. However, the value of egb/rt depended neither on the Nb or N content, nor on the temperatures between 900 °C and 1 100 °C. The whole of these results indicates that

austenitic grain boundary sliding may be a necessary condition for initiation of a grain boundary crack, but is not the controlling factor of hot ductility.

2. Reversibility of Hot Ductility Behavior
It seems reasonable to consider from the results

obtained here that hot ductility in the region I is mostly controlled by the grain boundary precipitates such as Nb(CN) or A1N. These precipitations can be caused by static and dynamic processes. The static precipitation will take place preferentially on the r grain boundary for the holding time taken in this investigation and the dynamic precipitation will be on the dislocations or substructure developed during deformation. Both precipitation processes become prominent in the lower temperature region of r. The results shown from Figs. 4 to 7 indicate that static
precipitation is enough for explaining hot ductility
variation. That is, the static precipitation of Nb(CN)
in the r grain boundary occurs during holding in the

Fig. 12. The effect of strain rate on hot free and Nb-bearing steels.

ductility in Nb- Fig. 13. The effect of strain rate on hot ductility and grain boundary sliding in 0.054 % Nb steel.

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lower temperature region of r, and these precipitates which remain undissolved deteriorate hot ductility at 1 000 °C, where the static and the dynamic precipi-tation become very slow. The fact that precipitates control hot ductility was confirmed through reversibili-ty study of hot ductility behavior as shown in Table 2. Hot ductility of the three steels were investigated at a fixed tensile testing temperature of 1 000 °C. The Nb and the high Al steels showed complete recovery of hot ductility by reheating at the temperature above 1200 °C in the thermal cycle (C) shown in this table. These results indicate that grain boundary precipi-tates such as A1N or Nb(CN) formed at 800 °C coars-ened during reheating above 900 °C and completely re-dissolved into r above 1200 °C.

The other evidence of precipitates controlling the fracture phenomena is shown in Fig. 14, where the effect of the maximum reheating temperature on hot ductility at 800 °C or 900 °C is demonstrated. Even though the r grain size increased continuously with the increase of the reheating temperature above 1200 °C, hot ductility did not change at all in the temperature region from 1 250 to 1 400 °C. Howev-er, ductility was improved on reheating to the tem-peratures below 1200 °C, where the change of hot ductility corresponded to dissolution behavior of A1N or Nb(CN) rather than the variation of the r grain size. The condition of complete dissolution of these precipitates into r is fulfilled at the reheating tempera-ture of 1 250 °C, heating to the higher temperatures has no further effect. The Nb or Al in soultion has

potential to re-precipitate preferentially in the r grain boundary in the lower temperature region of r, which causes a reduction of hot ductility. These results also explain the effects of Ti. The stable TiN that pre-cipitates at the reheating temperature around 1 300 °C reduces the amount of N which can be combined with

Nb as Nb (CN) precipitates at the lower temperature region of r, and this gives rise to modification of the compositions of Nb carbo-nitride. The r grain size attained by reheating at 1 300 °C for 5 min was around 450600 pm in all the steels, except the Ti-added steel of which grain size was around 200 pm. Although the beneficial effect of a small amount of Ti addition

in the HSLA steels on hot ductility may be partially brought about by refinement effect of the r grain size, the results of Fig. 14 indicate that the dissolution and precipitation behaviors of the precipitates mostly con-trol hot ductility.

3. The Causes and Prevention of Cracking
When the surface cracking problem arises in the slab or the hot rolled products, it is very important to find out whether it is caused during the process of steel making or hot rolling. The result observed on the effect of strain rate on hot ductility is very sug-gestive with regard to this point. As shown in Fig. 12, the deformation condition with higher strain rate around l0/sec, that corresponds to that of the various hot rolling processes, gave rise to the excellent hot ductility. This means that cracking of the slab or the hot rolled products in the Nb-bearing steels is not

caused during slabbing or the other hot rolling pro-cesses. On the other hand, the deformation caused in the slab during continuous casting or in the large ingot by the thermal stress during cooling or repeat-ing has relatively lower strain rate ranging from 10-2 to 10-5,14 which can deteriorate hot ductility of the steels. Actually cracking problems in the Al-killed or the Nb-bearing steels have been caused in the steel making process. The results from Figs. 4 to 7 indicate that hot charging process is beneficial to solve these cracking problems. The large ingot stripped from an ingot mold or the slab coming out from strand casting must be put into the reheating furnace keeping the surface temperature of the ingot or the slab, above 900 °C. This hot charging process that primarily aims at energy saving enables one to avoid the tem-perature region of reduced hot ductility of the steels. Improvement of hot ductility of the Nb-bearing steels, besides this modification of the thermal cycle, is also attained by adjustment of the chemistry of the steel, decrease of nitrogen content or Ti addition.

V. Conclusions
The effects of the Nb content in the HSLA steels

on hot ductility were investigated by the high tem-perature tensile testing under the various thermal

Table 2. The effect of thermal
terms of RA value.

cycles on hot ductility in

Fig. 14. The change of hot ductility and r grain size with
the maximum reheating temperature.

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cycles. The following results were obtained.
(1) The increase of the Nb content deteriorated hot ductility under all the thermal cycles investigated. The ductility trough covered the lower temperature region of r (800-900 °C; region I) and also the r+a region (700-750 °C; region II).

(2) The causes of poor ductility in each region were found to be based on the different mechanisms.

In region II, the strain concentration in ferrite re-sulted in poor ductility, and the thickness of the pro-eutectoid ferrite along the r grain boundary primarily controlled hot ductility. The precipitates of A1N or Nb(CN) were minor factors.

(3) The Nb addition in the HSLA steels deterio-rated hot ductility particularly in the region I. This was caused by the precipitation of Nb (CN) on the r grain boundaries. The static precipitation of Nb(CN) became prominent with decrease of the temperature below 1 000 °C. The re-dissolution of the Nb(CN) precipitates formed during holding in region I was at-tained by reheating above 1 100 °C.

(4) The improvement of hot ductility in Nb-bear-ing steels was attained by decrease of the N content or small amount of Ti addition so as to scavenge the nitrogen as TiN. High hot ductility in the Nb-bear-ing steel was also obtained under the thermal cycles corresponding to the hot charging of the ingots or the cc slabs, where the surface layer temperature are maintained above 900 °C during tracking.

(5) The increase of strain rate improved hot duc-tility over the whole temperature region, and the highest strain rate around 10/sec that almost corre-sponds to the deformation condition of hot rolling gave rise to the excellent hot ductility in the Nb-bear-ing steels.

(6) It was concluded that dynamic recrystalliza-tion was not related to poor ductility in the region I in the Nb-bearing steels. Although grain boundary sliding appeared to be a necessary condition for nu-cleation of grain boundary cracking, it was not a con-trolling factor of hot ductility.

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