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Authors
Kim, N.J.
Yang, A.J.
Thomas, G.

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MICROSTRUCTURE-PROPERTIES OF DIRECTLY QUENCHED Nb CONTAINING LOW CARBON STEEL

N.J. Kim, A.J. Yang and G. Thomas

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Materials and Molecular Research Division
Lawrence Berkeley Laboratory
University of California
Berkeley, CA 94720

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MICROSTRUCTURE-PROPERTIES OF DIRECTLY QUENCHED Nb CONTAINING LOW CARBON STEEL

N.J. Kim, A.J. Yang and G. Thomas

An investigation has been made on the microstructure and mechanical properties of Nb containing low carbon steel directly quenched after controlled rolling. The microstructure and mechanical properties of this steel are strongly affected by variations in the finish rolling temperature. Finish rolling above the Ar₃ temperature produces a duplex acicular ferrite-martensite structure which has good combinations of strength and impact toughness. When the alloy is further deformed in the two phase (α + γ) region, a duplex polygonal ferrite-martensite structure is produced. The ferrite matrix has a high dislocation density due to the deformation in the (α + γ) region. There is an abrupt increase in yield and tensile strength with a small deterioration in impact toughness. Lowering the finish rolling temperature decreases the volume fraction of martensite, thus decreasing the strength.

I. INTRODUCTION

Recently, there has been considerable research effort directed towards the development of high strength low alloy (HSLA) steels. These HSLA steels are mainly used for structural applications such as buildings, bridges, ships, pressure vessels, tube and pipelines, and for automotive applications. For such uses the important mechanical properties are strength, ductility, and toughness at low temperatures, ductile-brittle transition temperature, and weldability. The basic difficulty is that these properties are often incompatible in HSLA steels.
Controlled rolling is one of the processing techniques that simultaneously affects the strength, ductility, and toughness properties of steel. This improvement in properties is mainly due to the refinement of ferrite grain size. The controlled rolling procedure consists of heating the slab to an optimum temperature, deforming above and below the austenite recrystallization temperature, and/or deforming in the austenite-ferrite two-phase range. The optimum microstructure can only be obtained by careful control of processing variables, especially the amount of deformation and the temperature. So far, the utilization of this controlled rolling has been mostly done for the development of steels with ferrite-pearlite structures. These steels contain microalloying elements such as Nb, V to control the grain size and/or promote the precipitation strengthening. However, these steels have limitations in meeting the requirements for higher strength and toughness. Recently, transformation strengthening has been introduced and acicular ferrite or bainitic steels have been developed.3,4 Recent work by Kim and Thomas5 showed that controlled rolling of Fe/Mn/C alloy followed by direct quenching produces a duplex ferrite-bainite steel which has attractive mechanical properties, e.g., for pipeline use.5 The present research is mainly concerned with the effects of direct quenching after controlled rolling on the microstructure and mechanical properties of a similar steel containing Nb.

II. EXPERIMENTAL PROCEDURE

The composition in wt. pct. of the steel examined in this investigation is 0.06 pct. C, 1.49 pct. Mn, and 0.05 pct. Nb. The alloy was vacuum induction melted and cast into a round ingot (11.3 Kg.). This ingot was then homogenized in argon at 1200°C for 24 hours, and upset,
cross-forged at 1100°C into 3 cm x 6.4 cm slabs. These steel slabs were rolled in the laboratory rolling mill as shown schematically in Figure 1. After reheating, the slabs were given either a two-pass rolling or a three-pass rolling treatment. The third rolling pass was done in the (α + γ) two-phase range. After finish rolling, the plates were directly quenched into water. The temperatures were monitored by inserting a thermocouple into the midthickness of the slab prior to reheating.

Tensile specimens, 13 mm gauge length with 3.2 mm gauge diameter, and 3/4 subsize Charpy specimens were machined from the plates in the longitudinal direction. Tensile testing was done on an Instron machine with a crosshead speed of 0.05 cm/min. at room temperature. Total elongations were determined by measuring the gauge length before and after testing. The rest of the properties were determined from the stress-strain curves. Low temperature Charpy tests were performed following the ASTM-23-72 specifications. The ductile-brittle transition temperature was taken at the temperature corresponding to the midpoint of the upper and lower shelf energies.

Specimens for optical and transmission electron microscopy were cut from the broken Charpy specimens. 5% Nital etching was used for general observation of the resulting microstructure. In order to differentiate the second phase from the matrix, an etchant of 4% picric acid saturated with K₂S₂O₅ was used. The use of a hot etch (50-70°C) for 10-15 minutes was found to be necessary to reveal the prior austenite grain boundaries. Transmission electron microscopy specimens were prepared and examined in the usual way.
RESULTS AND DISCUSSION

A. Microstructure

The grain structure obtained after various rolling conditions is shown in Figure 2, from which it can be seen that the reheating temperature has a strong influence on the grain size of the recrystallized austenite. Niobium, depending on the reheating temperature, can act either as a grain refining agent or as a precipitation strengthener or both. At high reheating temperatures, most of the Nb will be in solution in austenite and, on cooling, will precipitate both in austenite and in ferrite as NbC(N), thereby increasing the strength of the steels. At lower reheating temperatures, a larger proportion of NbC will remain undissolved in the austenite and these NbC particles will act as barriers to grain boundary movement. Usually, in controlled rolling processing, the reheating temperature tends to be high for effective dissolution of NbC, resulting in greater precipitation strengthening, but considerable grain growth would take place during reheating. After rolling, the recrystallized austenite grain size will be coarser than that after a low reheating temperature. This is clearly shown in the present study. Since fine recrystallized austenite is a prerequisite for a fine ferrite grain size, low reheating temperature is used. Further deformation of the recrystallized austenite within the lower austenite temperature range (750°C-850°C) produced elongated grains. Changes in the second rolling temperature did not produce any detectable changes in the morphology of this elongated austenite. Direct quenching following these second rolling procedures produced mostly nonpolygonal structures in which the morphology of the second phase is revealed (Figure 3). Transmission electron microscopy analysis showed that the nature of the second phase was mostly martensite but with some bainite, e.g., Figure 4. Figure 5 shows the details of the matrix which is called "acicular ferrite" because of
its high dislocation density and elongated shape. Another interesting feature encountered in these structures is the presence of retained austenite, as shown in Figure 6. This type of retained austenite has been often found in microalloyed steels containing Nb, V. However, X-ray diffraction analysis failed to detect this phase, indicating that the volume fraction of retained austenite was less than 1%. Changes in the second rolling temperature (above 750°C) do not produce any remarkable changes in the resulting microstructure.

When the alloy is further deformed in the (α + γ) two-phase range and directly quenched, the dual phase (ferrite-martensite) structure is developed in which the martensite islands are more or less unidirectionally aligned in the ferrite matrix (Figure 7). Transmission electron microscopy shows that these martensite islands consist of mostly lath martensite (Figure 8).

The volume fraction of martensite decreases as the finish rolling temperature decreases. Deformation in the (α + γ) two-phase range results in elongated ferrite grains (Figure 7) containing a high dislocation density (Figure 9a). In some areas, recovery and small angle sub-boundaries can be seen (Figure 9b). It is probable that there is not enough time for significant recovery to occur because of fast cooling directly after deformation.

B. Mechanical Properties

The mechanical properties are plotted in Figure 10 as a function of finish rolling temperature. The data shows that the variation in the finish rolling temperature above Ar₃ (∼750°C) does not have a significant influence on the resulting mechanical properties. This can be understood in view of the fact that there is no essential change in the microstructure with finish rolling temperature when this temperature is above Ar₃. The attractive mechanical properties following these treatments result from the duplex microstructure in
which the fine grain size of the acicular ferrite matrix is responsible for low ductile-brittle transition temperature while the martensite islands contribute to the high strength.

Finish rolling below $Ar_3$ temperature produces considerable changes in the mechanical properties which can be correlated with the changes in microstructure. There is a large increase in strength and ductile-brittle transition temperature in steel finish rolled just below $Ar_3$. These properties decrease gradually as the finish rolling temperature is decreased. The high strength of the ferrite-martensite structure developed by this treatment compared to the acicular ferrite-martensite structure in steel finish rolled above $Ar_3$, is presumably because of the larger volume fraction of martensite and increased dislocation hardening in ferrite. The decrease in strength after lower finish rolling temperatures is due to the decrease in volume fraction of martensite (as observed in many dual phase steels $^{11,12}$). The higher ductile-brittle transition temperature of ferrite-martensite structure than that of acicular ferrite-martensite structure might be due to the coarsening of effective grain size when the matrix changes to coarse polygonal ferrite from fine acicular ferrite.

The interesting features of this ferrite-martensite steel are its small uniform elongation, and high yield strength-to-ultimate tensile strength ratio, while the characteristics of "conventional" dual phase (ferrite-martensite) steels are large uniform elongation and low yield strength-to-ultimate strength ratio $^{11,12,13}$ This behavior is presumably due to the high dislocation density in ferrite inherited from the deformation in the $(\alpha + \gamma)$ two-phase field. It is generally understood that yield strength and uniform elongation of dual phase steels are mainly determined by the properties of ferrite. The low yield strength of dual phase steels is due to the presence of mobile dislocations in
the ferrite region near the martensite-ferrite interface.\textsuperscript{13,14} The ferrite is constrained by the adjacent martensite and so the work hardening rate increases thus giving rise to a large uniform elongation.\textsuperscript{15} However, deformation of ferrite in the $(\alpha + \gamma)$ region gives a high dislocation density within the ferrite making it already work hardened. As pointed out earlier, most of these dislocations are not recovered following quenching, and this is presumably the origin of the high yield strength and small uniform elongation.

CONCLUSIONS

Based on this investigation, it has been found that the microstructure and mechanical properties of directly quenched Fe/1.49 Mn/0.06C/0.05Nb steel are influenced by the finish rolling temperature as follows.

1. Finish rolling above the $\text{Ar}_3$ temperature produces a duplex acicular ferrite-martensite structure. Variation in the finish rolling temperature above $\text{Ar}_3$ does not produce remarkable changes in the microstructure and mechanical properties.

2. A duplex polygonal ferrite-martensite structure is produced when finish rolling is done in the two-phase $(\alpha + \gamma)$ region. The ferrite matrix has an unusually high dislocation density due to this deformation. There is an abrupt increase in strength with a small deterioration in impact toughness when the alloy is finish rolled just below $\text{Ar}_3$ temperature. The strength decreases with further lowering of the finish rolling temperature, presumably due to the decreasing volume fraction of martensite.

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REFERENCES


FIGURE CAPTIONS

Fig. 1. Schematic illustration of rolling process.

Fig. 2. Optical micrographs of prior austenite grain with various rolling conditions. Etchant: 4% picric acid + K_2S_2O_5 (hot etch)

Fig. 3. Optical micrograph of the duplex structure developed by direct quenching from finish rolling above Ar_3 temperature. Etchant: 4% picric acid + K_2S_2O_5 (supersaturated).

Fig. 4. Bright field electron micrograph showing lath martensite formed by direct quenching after finish rolling above Ar_3.

Fig. 5. Bright field electron micrograph of the matrix showing acicular ferrite formed by direct quenching after finish rolling above Ar_3.

Fig. 6. Bright and dark field images showing retained austenite developed by direct quenching after finish rolling above Ar_3.

Fig. 7. Optical micrographs of ferrite-martensite structure developed after direct quenching following finish rolling below Ar_3. a) 740°C b) 725°C c) 710°C.

Fig. 8. Bright field electron micrograph showing martensitic region developed by direct quenching after finish rolling below Ar_3.

Fig. 9. Bright field electron micrograph showing substructures in ferrite. a) high dislocation density b) recovered region showing subgrains.

Fig. 10. Variation of the mechanical properties of directly quenched Fe/1.49Mn/.06C/.05Nb steel as a function of finish rolling temperature.
N.J. KIM is a postdoctoral Research Associate with the Lawrence Berkeley Laboratory, Materials and Molecular Research Division, Berkeley, California, 94720. A.J. YANG was a Visiting Scholar at the University of California, Berkeley, California, and has now returned to the Jiaotong University, Department of Metallurgical Science, People's Republic of China. G. THOMAS is Principal Investigator for the Lawrence Berkeley Laboratory, Materials and Molecular Research Division, and Professor of Materials Science, Department of Materials Science and Mineral Engineering, University of California, Berkeley, California, 92720.
TEMP.

AFTER ROUGH ROLLING (R.R.)

R.H. 1120°C (REHEATING)

R.H. 1200°C

AFTER FINISH ROLLING (F.R.)

F.R. 850°C

F.R. 800°C

F.R. 750°C

FIG. 2
FIG. 10
This report was done with support from the Department of Energy. Any conclusions or opinions expressed in this report represent solely those of the author(s) and not necessarily those of The Regents of the University of California, the Lawrence Berkeley Laboratory or the Department of Energy.

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