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THERMAL CYCLING TREATMENTS AND MICROSTRUCTURES
FOR IMPROVED PROPERTIES OF Fe-0.12%C-0.5%Mn STEELS

J. Y. Koo and G. Thomas

SUMMARY

A non-standard heat-treating process has been successfully applied for improving the mechanical properties of low carbon Fe-0.12%C-0.5%Mn steels. The process consists of alternate thermal cycling in the γ range and two phase ($\alpha+\gamma$) range with intermediate quenching so as to produce a microduplex fine grained mixture of martensite and proeutectoid ferrite. Some retained austenite is also present. These structures were characterized using optical and transmission electron microscopy.

The microduplex structures and the related tensile properties resulting from the heat treatment were compared with those after conventional and rapid heat treatment methods. The tensile properties of the microduplex structures are related to the volume fractions of ferrite and harder martensite phases as a composite, and the refined grain size. Considerable increases in strength are obtained which indicate potential applications of these heat treatments for strengthening steels e.g. in automobiles, thereby achieving weight reductions and fuel savings.

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INTRODUCTION

A great deal of work has been carried out to increase toughness and strength of medium and high carbon alloy steels¹⁻³ while comparatively little emphasis has been directed towards a fuller understanding of the ductility and strength of low carbon steel, particularly plain-low carbon steels.

It appears that there is considerable scope for exploiting the potential ductility of plain-low carbon steel with adequate strength. In steels containing ferrite and pearlite the increase in strength found by increasing carbon is very effective,⁴ but there are major disadvantages associated with this increase, namely, inadequate ductility and weldability. For a given composition of ferritic and pearlitic steel, it has been shown that the most important factor controlling strength and associated ductility is grain size.^{5,6} Following the work of Grange,⁷ a rapid heat treatment process was developed. This consisted of repeated cycles of rapid austenitizing and cooling and has now become a practical method to achieve ultrafine grain size. However, the increase in strength due to ultrafine grain size in steels with a mixture of ferrite and pearlite is not sufficient to reach the required strength level in some engineering applications, e.g. transportation in which achievement of weight reduction is an urgent task for design engineers.

If a plain-low carbon steel, subjected to thermal cycling so as to obtain fine grain sizes by conventional or rapid heat treatment techniques, is then quenched to form martensite so as to increase

strength, the ductility drops drastically and undesirable microstructures result due to low hardenability. Appreciable savings in weight and cost can be obtained if the strength can be increased without sacrificing ductility (so as to allow forming to be done).

The use of the $(\alpha+\gamma)$ two phase field for improving mechanical properties of steels has been recognized by other investigators e.g., in the maraging steels by Jin et al,⁸ and for low alloy steels by Snape and Church.⁹ However, the transformation behaviour and the beneficial effect resulting from using two phase decomposition are dissimilar in different systems and conditions.

The aim of this work is to extend the thermal cycling techniques to existing commercial 1010 type steel, by manipulating variables such as grain size and transformation substructures, so as to increase the strength at adequate ductility levels without the necessity of using mechanical or thermal-mechanical treatments.

EXPERIMENTAL PROCEDURE

Materials and Initial Heat Treatment

The materials used in this investigation were commercial automobile structural steels (INNA) and 1010 steel whose alloy compositions are given in Table I. The steel alloys, hereforth denoted "A" and "B" respectively, were donated by General Motors Research Laboratories, courtesy of Dr. D.J. Bailey.

"A" steel was received in the form of 1/8 in. pieces of rolled sheet in the nitrogenized and spheroidized condition. "B" steel was received as 1/8 in. hot rolled sheet. Tensile and transmission electron microscopy specimens were prepared by cutting the sheets into slightly bigger dimensions than the exact tensile specimens, homogenizing under an argon atmosphere at 1100°C for two hours, and then furnace cooling. The specimens were austenitized in an argon atmosphere at 1100°C for 30 minutes and then quenched directly into agitated iced brine. Later, transmission electron microscopic examination showed that the as-quenched specimens were free of undissolved carbides and were almost completely martensitic. The details of the thermal cycling treatments are given in the following section.

Thermal Cycling1. Conventional and Experimental Heat Treatments

All the samples, initially martensitic as a desirable starting microstructure⁷ for grain size refinement, were thermally cycled according to the cycling program shown schematically in Figs. 1 and 2.

Figure 1 represents a somewhat idealized experimental cycle in comparison with a reasonably representative conventional austenitizing treatment. In an experimental cycle the samples were heated alternately in the austenite range and two phase ($\alpha+\gamma$) range with intermediate iced brine quenching. Specimens were kept at a constant temperature in a tube furnace, which was controlled within $\pm 5^\circ\text{C}$. The principle of the treatments is to obtain a microduplex mixture of ferrite (low carbon-soft) and martensite (higher carbon-hard) as determined by the tie line associated with the annealing temperature in the ($\alpha+\gamma$) field. Thermal cycling then gives grain refinement and dislocated ferrite.

2.. Conventional and Experimental Rapid Heat Treatments

Figure 2 shows the experimental rapid thermal cycling method in comparison with a rapid austenitizing treatment for developing ultrafine grain size. Rapid heating was accomplished by immersing the specimens in a neutral salt bath kept at a desired temperature. The neutral salt bath was adopted because the desired temperature could be closely controlled and accurately measured automatically. The rapid heat treatment process, originally developed by Grange,⁷ consists of repeated cycles of rapid austenitizing each of very short duration at a temperature barely sufficient to austenitize, and intermediate drop quenching. The thermal cycle profile is identical for the experimental rapid heat treatment, shown in Fig. 2, and the experimental heat treatment shown in Fig. 1 except for the heating rate and duration at peak temperature.

Tensile Testing

Tensile properties were determined using the flat tensile specimens of 0.1 in. gage section and 1 in. gage length. The specimens were originally machined oversize, heat treated and subsequently ground to final dimensions. Approximately 0.012 in. layer was removed from each surface in order to eliminate the possible effects of surface decarburization. Optical microscopy and microhardness examinations revealed that decarburization was not deeper than 0.007 in.

Tensile tests were performed at room temperature in an Instron machine with a cross-head speed of 0.05 cm/min and full scale load of 1000 Kg. Uniform elongation was measured from the stress-strain curve and minor corrections were made with a one-in. extensometer capable of measuring strains up to 60%, the output of the extensometer being fed into the chart drive system of the Instron control console. Values of percent elongation reported are averages of at least three tests.

Metallography

It was decided to concentrate this investigation mainly on the microstructures and related tensile properties of the alloy "A", since there is little difference in compositions between them except for the amount of interstitial solutes (especially nitrogen).

Sections to be used for optical microscopic observation were cut from the tensile specimens. In metallographic investigations the standard etchant 2% nital was used; to reveal the prior austenite

grain boundaries in quenched samples, an etch solution of 1.4g picric acid, 1g sodium tridecylbenzene sulfonate and 100 ml water was utilized. The prior austenite grain sizes were measured by the linear intercept method from optical micrographs, and are given in Table 2.

Specimens of 0.125 in. thickness for electron microscopy were heat treated together with the tensile specimens. Each side of the specimens was carefully ground to 30~40 mils thickness. Flood cooling was employed to minimize specimen heating during grinding. The specimens were then chemically thinned to about 5 mils thick in H_2O_2 containing 2% HF at room temperature. Discs of the size of the standard Siemens specimen holder (2.3 mm dia.) were punched and mechanically sanded to 1~2 mils and then electropolished in a twin-jet polishing apparatus using a chromic-acetic acid solution (75 gms CrO_3 + 400 ml CH_3COOH + 21 ml H_2O). The resultant foils were examined in a Siemens Elmiskop IA at an accelerating voltage of 100 kV.

RESULTS AND DISCUSSIONS

Grain Size Refinement1. Comparison of Standard and Non-Standard Thermal Cyclings

For many reasons¹⁰⁻¹² the degree of grain refinement becomes more pronounced as the carbon content in steel increases. Therefore, it is expected that, for alloys A and B, their response to various heat treatments would be rather poor. However, this is desirable for the present purposes, because variables which influence grain size and mechanical properties in steels of such low hardenability are accentuated. The starting microstructure is an important variable in determining the extent of grain refinement by thermal cycling. Speich and Szirmai¹² have shown that the preferred starting structure for carbon steels is a hot rolled structure, whereas Grange¹¹ reported that martensite or fine ferrite-carbide aggregates are desirable.

In this investigation, martensite was chosen, by quenching from high austenitizing temperature (1100°C), as the initial microstructure. This was partly to minimize possible segregation of embrittling impurity elements into the prior austenite grain boundaries on slow cooling, which as a result might lead to intergranular embrittlement. Therefore, it was necessary to use intermediate quenching rather than slow cooling on subsequent thermal cycling steps (as expected in commercial structural steels, preliminary experiments showed that the tensile properties of the specimens which had undergone the thermal cycling with intermediate air cooling and had finer grains were

nevertheless poorer than those of as-received specimens with larger grain size, presumably due to intergranular embrittlement).

The initially martensitic samples (grain size $\sim 80\mu$, ASTM #2) were subjected to the thermal cycling treatments shown in Figs. 1 and 2. Extensive isothermal experiments were made, by varying the temperature at fixed holding time and then varying the holding time at constant temperature. Based on optical metallography, the A_3 temperature was found to be 860°C for alloy A and 880°C for alloy B. In order to determine suitable austenitizing periods, isothermal experiments were made at 20°C above their respective A_3 temperatures. Insufficient holding time results in undissolved ferrite which can be identified by its "blotchy" appearance in the optical micrograph of an as-quenched specimen; excessive grain growth sets the upper limits. Following these guiding principles of obtaining optimum peak temperature and holding time, the conditions for subsequent steps were also optimized to achieve maximum grain refinement. For specimens A and B subjected to conventional heat treatment, a suitable austenitizing time was found to be seven minutes for the first cycle (C1) and six minutes for the second cycle (C2), producing (after C2) microduplex structures consisting of mainly martensite and proeutectoid ferrite (Fig. 3a). For the experimental heat treatment, the annealing temperature in the two phase ($\alpha+\gamma$) range was chosen at 20°C below the A_3 temperature for alloy A, and at 30°C below A_3 for alloy B. The optimum annealing time in the ($\alpha+\gamma$) range was found to be nine minutes and seven minutes for the cycles T2 and T4 respectively. The grain size

achieved after heat treatment T2 is 18μ as shown in Fig. 3b, illustrating that carbon enriched austenite, produced by partitioning in the two phase field, transformed to martensite as evidenced by dark etching. Similarly, for rapid heat treatment the austenitizing time at the peak temperature was determined to be 40 seconds in the first cycle (RC1) and successively shorter times 38, 33 and 30 seconds for the RC2, RC3 and RC4 heat treatments respectively, to bring about the structures shown in Fig. 4a (RC4). For the experimental rapid heat treatment, optimized annealing times in the two phase field were found to be 39 and 32 seconds for RT2 and RT4 respectively. The resulting microduplex structures are illustrated in Fig. 4b (for RT4). In order to make meaningful comparisons of microstructures and tensile properties in each thermal cycling technique, only the specimens which had undergone second and fourth cycles were mainly investigated. As may be seen in Figs. 3 and 4, most of the microduplex structures showed a fairly uneven grain size distribution throughout the samples. Presumably this is due to the existence of proeutectoid ferrite inevitably present because of low hardenability, and due to the presence of nonuniform carbide morphology (see Fig. 7). This would lead to more grain refinement in some areas than others, since the nucleation of the $\alpha \rightarrow \gamma$ transformation on heating is usually believed to be precipitate-correlated.¹² Above two cycles, further reduction in grain size by slow heating methods (in comparison with rapid heat

treatment) did not occur, demonstrating the limitations of developing a substantially finer grain size by the above heat treatments. The grain size ($\sim 18\mu$, ASTM #9) obtained by heat treatment RC2 suggests that rapid heat treatment does not provide a significant improvement in grain size refinement over conventional methods (24μ , ASTM #8) in the alloy investigated. Furthermore, no appreciable change in grain size was observed with two more cycles (heat treatment RC4, Fig. 4a) after the second cycle. However, the experimental heat treatments, T and RT provided finer grain sizes (though not pronounced) than those obtained by conventional (C) and rapid heat treatments (RC), respectively. It appears that the second phase (proeutectoid ferrite) may effectively inhibit grain growth of austenite in the two phase field. Moreover, the second phase rendered the grain size refinement to be less sensitive to holding time, whereas the degree of refinement by the repeated austenitizing method is very sensitive to slight changes in holding times at the peak temperature.¹¹

Microstructural Changes on Thermal Cycling

The various ferrite, martensite and carbide morphologies encountered with each of the thermal cycling treatments were examined using transmission electron microscopy techniques. The most significant feature in all the as-quenched specimens, except the starting microstructure, was the microduplex two phase structure consisting mainly of proeutectoid ferrite and martensite, each amount depending on the

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specific heat treatment and the number of cycles. After careful visual examination of many electronmicrographs and optical micrographs the approximate fraction of martensite was determined by standard quantitative metallography using the linear intercept method.¹³ Some estimated values from selected heat treatments are given in Table 2. The typical microstructure of martensite consisting of dislocated laths in this low carbon steel is shown in Fig. 5. Adjacent laths were sometimes twin related or separated by small angle boundaries. Another prominent feature in the martensite was the occurrence of autotempering. The density of the autotempered carbides was highest in the starting microstructures and decreased remarkably in the specimens after two cycles. Negligible amounts of autotempered carbides were present in all the samples after 4 cycles. The reason for this may be that the specimens subjected to thermal cycling always contained some undissolved carbides with the % undissolved carbide increasing with additional cycles. Any additional carbon in solution was probably segregated to lattice defects during quenching^{14,15} since most of these sites have a lower energy than those available at the precipitate,¹⁶ so there is little driving force for precipitation in the martensite.

A surprising observation was the detection of extremely narrow films of retained austenite along the martensite lath boundaries, the amount depending upon heat treatment.

The 0.5% Mn, 0.12% C steel is hypo-eutectoid in composition, and has very low hardenability; thus, pro-eutectoid ferrite followed by precipitation of (α + carbide) according to the $\gamma \rightarrow \alpha$ + carbide reaction should be expected at a high transformation temperature upon quenching. The ferrite regions in all the specimens subjected to the various heat treatments were heavily dislocated. However, the significant differences in proeutectoid ferrite morphologies between specimen A4 (heat treatment C2 + IBQ) and specimen A5 (T2 + IBQ) were subgrain size and dislocation density. In specimen A4, the high density of dislocations in ferrite, which are probably produced as a result of the $\gamma \rightarrow$ martensite transformation strain, would appear to have been reduced by self annihilation and subsequently annealed out on heating, since the crystal structure changes as the $\alpha \rightarrow \gamma$ transformation proceeds. Thus, in this structure the dislocations were mostly fresh dislocations produced in the last step of quenching in the heat treatment sequence. Moreover, the dislocations were frequently concentrated in the vicinity of the α /martensite boundaries.

On the other hand, the proeutectoid ferrite in the specimens A5, subjected to two phase decomposition, does not transform to austenite when annealed in the two phase region. As a result, the dislocations generated by partial martensitic transformation on quenching would rearrange themselves into dislocation walls, thus forming sub-boundaries during annealing in the two phase field. This is shown in Figs. 6(a) and (b). The electron microscopy examination revealed that the specimens subjected to heat treatment T2 contained sub-boundaries spaced on

average 0.5 to 1 μ apart (Fig. 6(b)) in some areas, and larger sizes than these in other areas.

The ferrite in the immediate vicinity of the martensite/ α boundaries showed a high density of carbide particles resulting from the $\gamma \rightarrow \alpha$ + carbide reaction after the completion of the nucleation and growth of proeutectoid ferrite upon quenching. This is illustrated in Fig. 7. This type of structure was frequently observed in all the specimens observed, and is similar¹⁷ to that observed in the isothermal decomposition of austenite in Cr steels. The zone of the localized carbides was 1 to 3 μ wide. It appears that the carbide adopted both a fibrous morphology (Fig. 7d) in which the prior γ/α interfaces are heavily faceted, and an interface precipitation morphology (Figs. 7a and c) where the interfaces are mostly smooth. The prior γ/α interfaces seem to be devoid of carbide precipitates. This can be shown by dark field microscopy e.g., Fig. 7b.

Formation of Retained Austenite

Careful diffraction and dark field electron microscopy²⁰ revealed retained austenite that was trapped between two growing martensite laths in the form of narrow thin films. This is an interesting observation since the occurrence of this phase in such low carbon steels has not been reported by earlier investigators.

The stabilization of austenite has been the subject of a number of investigations. Austenite retention at room temperature can be explained partly by the lowering of M_f temperature below room temperature

due to carbon enrichment. Its retention can also be understood in terms of thermal and mechanical stabilization.^{18,19}

In applying these concepts, it seems plausible that retained austenite should form in a low carbon steel. Figure 8(a) shows an example of retained austenite trapped between martensite laths in the specimen A5 and Fig. 8(b) is the dark field micrograph where the retained austenite reverses contrast. (For more details of identification of retained austenite by these techniques see ref. 20). The crystallographic relation between austenite and martensite exhibited the Kurdjumov-Sachs orientation relationship, $\langle 111 \rangle_{\alpha} \parallel \langle 110 \rangle_{\gamma}$. Another interesting observation was that the specimens given alternate thermal cycling (T2 and RT2) revealed more retained austenite than those given repeated austenitization (C2 and RC2). It seems that the increase in the amount of retained austenite may result from stabilization due to carbon enrichment in the austenite produced by annealing in the $\gamma+\alpha$ region. However, it was impossible to obtain quantitative data on the exact amount of retained γ by X-ray analysis.

Relation Between Microstructures and Tensile Properties

The room temperature tensile tests data are summarized in Table 2. Comparing the values of yield strength and uniform elongation between the specimens given conventional treatments and non-conventional heat treatment, for example, specimens A2 and A3, A4 and A5, B4 and B5, and so on, one observes that the latter provided better combinations of yield strength and uniform elongation than the former treatments.

In addition, there is no significant advantage accrued through rapid heating techniques (e.g., RC2 and RT2) over slow heating methods (e.g., C2 and T2). Figure 9 shows the plot of yield strength vs. uniform elongation for the specimens A1, A2 and A4 (conventional), and A1, A3 and A5 (experimental). The superior combinations of yield strength and uniform elongation should be understood in terms of the substructural and morphological differences. The following parameters appear to be important: (a) fraction of martensite, (b) the prior austenite grain size, (c) dislocation substructure, and (e) retained austenite. The results of Fig. 10 indicate that of these parameters, the fraction of martensite seems to be the most significant.

Some consideration must also be given to the effects of grain size and the retained austenite. Firstly, grain size: for a number of metallic polycrystalline aggregates, it has been shown experimentally that σ_f , the flow stress at constant strain, is related to the grain diameter D by the Hall-Petch equation

$$\sigma_f = \sigma_o + kD^{-1/2} \quad (1)$$

where σ_o and k are material constants.

As described previously, the experimental heat treatments provided finer prior γ grain size than can be attained by other treatments, hence this may contribute to the improved tensile properties. However, since the decrease in the grain size was not great, the contribution, if any, would be small. In order to isolate the effect of grain size

on the properties, separate experiments were performed by austenitizing at 1100°C and quenching followed by direct annealing in the two phase field. In this way, a grain size of 30 μ was obtained, and the microstructure was almost identical to that of specimen A5 (heat treatment T2 and IBQ; grain size 18 μ). The specimens after one thermal cycle showed about 3 ksi drop in yield and tensile strength compared to specimen A5 at similar uniform elongations.

Warrington²¹ and Baird²² reported that the Eq. (1) can be applied to the effect of subgrain size on the flow stress of iron. Since the ferrite region in the specimen A5 involved ultrafine subgrains (though not distributed evenly), the ferrite would be effectively strengthened according to Eq. (1). This effect appears to be mainly responsible for increasing the yield strength of the non-conventionally treated composite materials.

Secondly, retained austenite (γ) is known to influence a variety of properties, but the way in which the presence of retained γ aids in improving these properties is not completely clear at this time. However, the beneficial effects of small amounts of retained austenite on the mechanical properties have been reported by many investigators (e.g., refs. 1,23). All the specimens A6 (A5 + refrigeration at LN) exhibited smaller values of uniform elongation (though not pronounced) than specimens A7, possibly due to the reduced retained γ as described before. However, the differences were very small; therefore, the effect of retained γ on the improved properties is expected to be very small, but further experiments on this could be interesting.

There is no difference in tensile strength between the specimens given conventional and non-conventional treatments e.g., A4 and A5, B4 and B5 and so on in spite of the fact that the fraction of martensite in each specimen is different. This could be understood from the well-known rule of two phase mixtures for the condition that the microstructural constituents (ferrite and martensite) are equally strained, using the following equation (as a good approximation):

$$\sigma_{uc} = \sigma_{uf} V_f + \sigma_m (1 - V_f) \quad (2)$$

where

σ_{uc} = tensile strength of thermally cycled specimen.

σ_{uf} = tensile strength of martensite

σ_m = stress of the matrix (ferrite) at the ultimate tensile strain of martensite.

V_f = volume fraction of martensite

Consider two representative specimens A4 and A5. If both σ_m and σ_{uf} are the same for A4 and A5 then it would be expected that σ_{uc} for A4 would be higher than that for A5 due to its higher volume fraction of the higher strength martensite. However, σ_m for A5 is higher than σ_m for A4 due to the ultrafine subgrain size. Thus the increase in the strength (σ_m) in A5 just balances the increase in V_f in A4, resulting in the same value of σ_{uc} for both specimens.

CONCLUSIONS

1. 1010 is a commercial steel which when conventionally treated (hot rolled) is specified to the following properties; $\sigma_y \sim 40$ ksi, $\sigma_{uts} \sim 50$ ksi, and uniform elongation $\sim 23\%$. As a result of the present work these properties can be changed to: $\sigma_y \sim 70$ ksi, $\sigma_{uts} \sim 100$ ksi, and uniform elongation $\sim 10\%$ by simple multiple-cycling heat treatments.
2. These multiple heat treatments gave better combinations of yield strength and uniform elongation than are achieved by conventional treatments, i.e. 15 $\sim$ 20 ksi increase in yield strength at similar elongations or 2 $\sim$ 3% increase in uniform elongation at similar yield strengths.
3. The heat treatments investigated provided a better refinement in grain size compared to that from conventional heat treatments. It appears that proeutectoid ferrite effectively inhibits grain growth of austenite and renders grain size refinement to be less sensitive to holding time at peak temperatures.
4. The heat treatments arrived at in this research ideally combine the beneficial effects of finer grain size, dislocation substructure, and retained austenite. In particular the ferrite region of the specimens subjected to the experimental heat treatment seemed to be effectively strengthened mainly by the formation of ultrafine subgrains.
5. Multicycling more than two cycles did not provide additional grain refinement, nor did it provide additional improvements in tensile properties.

6. As described before, a high density of carbide particles in the immediate vicinity of the martensite/ α boundaries appears to be detrimental to tensile properties. Therefore, it is suggested that alloying so as to inhibit the growth of carbides should improve the mechanical properties.

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Table 1. Compositions of Alloys used (in Wt. Pct.)

Alloy	C	Mn	S	N	Fe
A	0.12	0.49	0.012	0.011	bal.
B	0.08	0.46	0.024	0.006	bal.

TABLE 2. SUMMARY OF TREATMENTS, MECHANICAL PROPERTIES AND OTHER DATA

Heat Treat- ment	Final Quench	Specimen Number*	Yield Strength** ksi (Mn/m ²)	Tensile Strength ksi (Mn/m ²)	Uniform Elonga- tion (%)	Estimated Fraction of Martensite (%)	Grain size (μ)
CO [†]	IBQ	A1	148 (1020)	174 (1200)	2	100	80
		B1	104 (717)	118 (814)	4		
C2	WQ [‡]	A2	76 (524)	112 (772)	7	55	25
T2	WQ	A3	78 (538)	111 (765)	9	45	18
C2	IBQ	A4	101 (696)	130 (896)	4	75	25
		B2	71 (490)	100 (689)	7		
T2	IBQ	A5	97 (669)	130 (896)	6	60	18
		B3	71 (490)	100 (689)	10		
	IBQ+LN ^{‡‡}	A6	95 (655)	130 (896)	5		
RC2	IBQ	A7	103 (710)	139 (958)	5		18
		B4	76 (524)	106 (731)	6		
RT2		A8	95 (655)	137 (945)	7		10
		B5	71 (490)	105 (724)	9		
RC4	IBQ	A11	103 (710)	137 (945)	4		16
		B8	72 (496)	103 (710)	7		
RT4		A12	95 (655)	134 (924)	7		10
		B9	70 (483)	103 (710)	9		

* A refers to alloy A, and B to alloy B.

** 0.2% offset

[†] Starting microstructure.

[‡] Water quench

^{‡‡} Ice brine quench + liquid nitrogen refrigeration

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Figure Caption

- Fig. 1. Schematic representation of experimental thermal cycles in comparison with conventional heat treating cycles, referred to part of the Fe/C equilibrium phase diagram.
- Fig. 2. Schematic representation of conventional and experimental rapid heat treating cycles.
- Fig. 3. Optical microstructures. 2% nital etch (a) Specimen A4 (heat treatment C2), lower half: same specimen, but with different etching solution to reveal prior austenite grain boundaries. (b) Specimen A5 (T2).
- Fig. 4. Optical microstructures (as for Fig. 3) (a) Specimen A11 (RC4). (b) Specimen A12 (RT4).
- Fig. 5. Transmission electron micrograph showing the region of typical dislocated martensite of the as-quenched microduplex structure.
- Fig. 6. Dislocation substructures in ferritic regions of specimen A5. (a,b) Dislocations rearranged to form a sub-boundary. (c) Subgrains spaced on an average 0.5μ to 1μ wide.
- Fig. 7. Morphology of heavy precipitation of carbides in the immediate vicinity of α /prior γ boundaries. (a) Bright field image and dark field (b) shows reversal of carbide contrast. (c) and (d) other regions of carbide precipitation.
- Fig. 8. (a) Bright field image of retained γ trapped between two martensite laths in specimen A5. ($(200)_{\gamma}$ spot is marked by image of objective aperture in SAD). (b) Dark field of retained austenite.
- Fig. 9. The values of uniform elongation are plotted against yield strength for the specimens subjected to conventional and experimental heat treatments (refer to Fig. 1).
- Fig. 10. Yield strength vs. percent martensite for the specimens A1, A2 and A4 (conventional cycling treatment) and A3 and A5 (experimental cycling treatment).

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Fig. 1. Schematic representation of experimental thermal cycles in comparison with conventional heat treating cycles, referred to part of the Fe-C equilibrium phase diagram.

Fig. 2. Schematic representation of conventional and experimental rapid heat treating cycles.

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Fig. 3

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Fig. 4

0 0 0 0 4 3 0 6 5 1 6

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Fig. 5

Fig. 6

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Fig. 7

Fig. 9. The values of uniform elongation are plotted against yield strength for the specimens subjected to conventional and experimental heat treatments (refer to Fig. 1).

Fig. 10. Yield strength vs percent martensite for the specimens A1, A2, and A4 (conventional cycling treatment) and A3 and A5 (experimental cycling treatment).

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