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LOW CARBON DUAL PHASE STEELS FOR HIGH STRENGTH WIRE

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I. Introduction

Conventionally, high tensile strength wire has been made by cold drawing wire rod which is manufactured by hot rolling carbon steels of varying carbon content (e.g. "Stelmore, EDC etc. processes). The microstructure of such rods consists of ferrite and pearlite and the properties depend on carbon content and cooling rate, which control the pearlite volume fraction, and interlamellar spacing of the pearlite colonies. More recently, industrial practice has tended towards microalloying additions, e.g. Cr, Mo, V in order to refine the pearlite structure in attempts to achieve high tensile strength drawn wire, e.g. for steel ropes, prestressed concrete strands, tire cord, etc. without involving patenting heat treatments (1,2). However, so far the attainment of steel wire with 400,000 psi or greater, tensile strength involves at least one patenting heat treatment. Clearly the elimination of this processing stage as well as microalloying additions would be of significant economic value.

An alternative to pearlitic steel is the so-called ferritic-martensitic (or bainitic) dual phase steels (DFM) initially developed for cold formable sheet especially in automotive applications. Since the introduction of dual phase steel (3), considerable research effort has been directed at UC Berkeley toward understanding and improving these steels, in particular the effects of different metallurgical variables such as volume fraction, morphology, and carbon content of the martensite and alloying elements (4-10). The potential of dual phase steel for high strength steel wire has been recognized and optimized in recent years and the first results on dual phase steel wire have been published by Nakagawa and Thomas (11). This work indicated the potential of DFM rods for high strength wire, e.g. tire cord. For this to be economical, it is necessary that the initial wire rod be produced using existing rod mills, without additional heat treatment. It is the object of the present paper to demonstrate this feasibility.
In addition, it is useful to point out the advantages of DMF steels over existing carbon, HSLA or microalloyed steels. They are as follows:

1. The major source of strengthening in the DFM structure arises from the presence of inherently strong martensite as a load carrying constituent in a soft ferrite matrix which supplies the system with the essential element of ductility. The resulting mixture is analogous to that of a composite and the concept of fiber-composite strengthening is thus useful in qualitatively understanding dual phase steels.

2. The required composite microstructures can be produced solely by simple heat treatment or directly on-line in a rod or bar mill. The alloy composition and heat treatment may be varied to give different transformation paths and hence different morphologies and mechanical properties. The composition can be tailored to the particular steel making process, e.g. scrap melting, continuous or ingot casting, etc.

3. Dual phase steels have mechanical properties which are characterized by continuous yielding with high initial work hardening rates, large uniform strains and high tensile to yield ratios. These factors account for their better cold formability compared to that of ferritic, pearlitic or HSLA steels of similar strengths and facilitate a wide range of combinations of strength and ductility. A single composition can thus replace a range of carbon steels (as currently needed for different strengths) thus simplifying inventory and processing.

4. Carbide forming elements as in commercial HSLA are not essential so simple alloy chemistries are involved.

5. The strong phase, if controlled to be lath martensite, deforms homogeneously along with the ferrite matrix. This means large strains can be sustained without decohesion.

6. Corrosion and fatigue resistance are superior to those of conventional steels.

7. The low carbon levels mean easier weldability.
II. Production of DFM Rod

The principle of the method is as follows. Starting with a low carbon steel ($C \leq 0.1\text{wt}\%$), a microstructure of ferrite and austenite is obtained by heating to the desired temperature in the $(\alpha + \gamma)$ field. This is followed by rapid quenching which yields a composite ferrite-lath martensite ($<0.4\text{wt}\%C$) structure. The latter results from the austenite to martensite shear transformation.

In order to obtain the high strength steel wire in one continuous, multipass, cold drawing operation, it is a prerequisite that the starting dual phase steel for wire drawing has a good combination of strength and formability. This is achieved by producing an alloy containing fibrous lath martensite in a fine grained ferrite matrix. During wire drawing, load transfer is most efficient when the second phase (martensite) is present in the form of deformable fibers rather than spheres, which would tend to reduce void formation. This is primarily because the transfer of load occurs by shear acting along the martensite/ferrite interfaces and, for a given volume fraction and the same number of martensite particles, more interfacial area is available in the case of a fibrous morphology. In addition, the carbon content of martensite must be controlled not to exceed $-0.4\text{wt}\%$ or else undesirable plate martensite will be formed which will adversely affect drawability and ductility. The basic physical metallurgy for attaining these objectives is detailed in ref. 10 and need not be repeated here.

Two types of processing have been investigated to produce desirable dual phase steels for cold drawing into high strength wire:

1. Controlled Rolling

Controlled rolling is one of the processing techniques that simultaneously improves the strength, ductility and toughness of the steel (12) and is important for rod mill rolling practice. This improvement in properties is mainly due to the refinement of ferrite grain size and in the present case to control the dual phase morphology. The controlled rolling
is illustrated in Fig. 1 and consists of soaking at the optimum temperature, deforming above and below the austenite recrystallization temperature, then rolling in the \((\alpha + \gamma)\) field below the \(A_r_3\) temperature. The optimum microstructure can be obtained by careful control of processing variables, especially the amount of deformation and the deformation temperature. Moreover, when the alloy is deformed in the \((\alpha + \gamma)\) region and directly quenched, the dual phase structure is developed in which the martensite islands are more or less unidirectionally aligned fibers in the fine grained ferrite matrix. This method is suitable for commercial rod mills provided a quench system is available. The particular finish rolling (no twist mill) temperature will depend upon composition since the latter determines the \(A_r_1\) and \(A_r_3\) temperatures (Table 1). Silicon raises \(A_r_3\), carbon lowers \(A_r_3\), but Mn has little effect.

2. Intermediate Quenching

This heat treatment consists of austenitizing and quenching to 100% martensite, followed by annealing in the \((\alpha + \gamma)\) region and then quenching as shown in Fig. 2. This treatment was developed so as to fully exploit the characteristic nature of the initial martensite structure prior to subsequent annealing in the \((\alpha + \gamma)\) range. The initial martensite structure provides sufficient heterogeneous nucleation sites for austenite during two phase annealing. The austenite grows along the martensite lath boundaries resulting, after quenching, in a fine fibrous distribution of martensite in a ferrite matrix. The heat treatment process is, of course, less economical than controlled rolling but it does provide great flexibility in control of structure and properties.

III. Experimental Procedure

1. Material Preparation

The materials used in this investigation are based on Fe/0.1C/X where X is Si and/or Mn. No microalloying elements are added. Supplied billets were cut into smaller pieces
and hot open die forged into rods of 0.65" dia. The compositions of the alloys are listed in Table 1.

2. Measurement of Transformation Temperatures

A Theta Dilatronic IIIR dilatometer was used to measure phase transformation temperatures in the high purity Fe/Si/C rods. Specimens were heated to 1150°C and held for 5 min. before cooling to room temperature. A programmed linear heating and cooling rate of 125°C/min. was used. The strain-free phase transformation temperatures for this alloy are $A_{c1} = 780°C$, $A_{c3} = 1030°C$, $A_{r1} = 700°C$ and $A_{r3} = 940°C$. The phase transformation temperatures for the other alloys were calculated using known formulae (13), and are listed in Table 1.

However, deformation of austenite raises the temperature of ferrite formation ($A_{r3}$) by strain-induced transformation in the controlled rolling treatment. In this study, the strain-induced $\gamma \rightarrow \alpha$ transformation temperature was assumed to be 50°C higher than the strain free $A_{r3}$ temperature for a deformation of 30% (14).

3. Heat Treatment

A. Controlled Rolling Process. The 0.65" dia. rods were soaked in the austenite region for 20 min., then rolled down to 0.25" dia. in three successive passes on a two high reversing bar mill, followed immediately by water quenching (Fig. 1). Finish rolling was done just below the $A_{r3}$ temperature, measured under an unstrained condition, to obtain the desired dual phase microstructure.

B. Intermediate Quenching Treatment. Oversized blanks (0.25" dia.) were machined for drawing into wire and tensile test specimens. These were then heat treated in a vertical tube furnace under flowing argon atmosphere according to the schedules as shown in Fig. 2. The two phase annealing temperatures were chosen so as to obtain the required volume fraction of martensite as shown in Table 2.

4. Wire Drawing

The DFM rods were machined down to a diameter of .217" (5.5mm) and then drawn without coating through 6° - 8° semi-die angle conical carbide and diamond dies lubricated
with Dupont Vydax Freon-Teflon dispersion. The usual reduction in area per pass decreased from 35% in the early stage to approximately 20% in the later stages. This technique differs from conventional wire drawing of pearlitic steel because of the high initial work hardening characteristics of DFM steels (10). If the reduction in area is too small, central burst cracking may occur (15).

5. Tensile Testing

Tests of the steels in the as-heat treated condition were conducted on cylindrical specimens of 0.113" gauge diameter on an Instron machine. Drawn wires between 0.119" and 0.0526" dia. were machined in the gauge section to give approximately 90% of the original cross sectional area. Smaller wires less than 0.0526" dia. were tested either in flat, serrated hardened steel jaws, or in drum type grips, in which the fine wire was wrapped around a drum, then tightened with a screw.

IV. Results and Discussion

1. Microstructure

For controlled rolled steels, all of the microstructures show more or less unidirectionally aligned fine martensite islands in the fine grained ferrite matrix. A typical microstructure of dual phase steel wire rod in this process is shown in Fig. 3. During the controlled rolling process, the coarse austenite obtained by soaking in the austenite region is broken into small recrystallized austenite through deformation in the roughing stage. By continuous deformation at lower austenite temperatures, austenite grains are elongated producing deformation bands. These regions are the nucleation sites for the austenite to ferrite transformation, thus allowing fine ferrite grains to form during finish rolling in the \((\alpha+\gamma)\) phase field, i.e. just below the \(\text{Ar}_3\) temperature. When the finished rod is directly quenched, a dual phase structure is produced in which the second phase, now martensite, appears unidirectionally aligned in the fine, equiaxed ferrite matrix, and is present at about 20% volume fraction.
Figure 4 shows the typical microstructure which was obtained by the intermediate quenching treatment of a commercial welding rod steel. The microstructural features are acicular martensite particles in the ferrite matrix, which show the influence of the initial martensite structure before two phase annealing. Comparing Figs. 3 and 4 shows the superior refinement of ferrite effected by the controlled rolling treatment. The volume fraction of martensite for each process is given in Table 2. Fig. 5 shows a typical transmission electron micrograph of the dual phase structure, consisting of lath martensite in the ferrite matrix.

The progression of the dual phase structure as the reduction in area by wire drawing was increased to 97% is shown in Fig. 6. These micrographs clearly show that martensite deforms continuously with the ferrite matrix, and there is no observable void formation at the interface between two phases. The initial microstructure was produced by the intercritical annealing treatment to produce a more equiaxed structure. A high dislocation density of the drawn wire was observed by transmission electron microscopy and an example is shown in Fig. 7. It is quite difficult to distinguish between the ferrite and martensite. This structure is similar to that of drawn pearlite and ferrite in Fig. 8.

2. Drawability

The drawing limit for all the steels corresponds to the maximum drawing strain the wire can sustain before failure at the die. Generally, all the dual phase steels produced by the controlled rolling process and the intermediate quenching treatment could be drawn to large strains (ε > 4, i.e. >98% reduction in area) without any patenting (intermediate) heat treatments or fracture. The drawability limits for the different specimens are listed in Table 2. A comparison of the three intermediate quenching treated silicon containing alloys shows that the drawing limit of the specimen with the highest volume fraction of martensite is the lowest, as can be expected.

Metallurgical control is essential. For example, the effect of martensite on drawability is particularly important if the microstructure contains plate (twinned)
martensite rather than lath martensite. This can happen due to (a) too low a quench temperature, or (b) inefficient cooling or (c) carbon segregation (10). These factors cause the carbon content of austenite to exceed 0.4wt% which on subsequent transformation leads to the high carbon plate martensite. The latter is non-deformable leading to microvoids and coalescence and subsequently to shear failure (Fig. 9).

3. Mechanical Properties

A. Initial Properties. The tensile properties of the as-heat treated steels are summarized in Table 3. Data show that these properties follow the law of mixtures, viz. the strength increases while the ductility generally decreases with increasing the martensite volume fraction.

B. Drawn Properties. From Table 4 it is seen that dual phase steel wire rods can be drawn to total true strain, \( \varepsilon = 3.6 \), i.e. 97% reduction in area for a strength level of 270 - 300 ksi (1860 - 2070 MPa) or drawn to \( \varepsilon = 6 \), i.e. 99.8% reduction in area for higher strength level of 380 - 400 ksi (2620 - 2760 MPa).

Figures 10-14 show drawing data for some of the steels investigated. It can be seen that very high tensile strengths are achieved even though the carbon content is low (compare to conventional 0.7%C wire, Fig. 14). The results show that the flow stress of dual phase steel wires follow the empirical equation initially proposed by Embury and Fisher (16)

\[
\sigma_f = \sigma_i + \frac{k}{\sqrt{r_0}} \exp \left( \frac{\varepsilon}{4} \right) = \sigma_i + \frac{k}{\sqrt{r_0}} \sqrt{\frac{D_0}{D}}
\]

where \( \sigma_i, k \): constants

- \( \sigma_f \): flow stress of wire after being drawn to a strain \( \varepsilon \)
- \( D_0 \): initial wire rod diameter before drawing
- \( D \): wire diameter after being drawn to a strain \( \varepsilon \)
- \( r_0 \): mean spacing of dislocation barrier of wire rod
It is apparent from the above equation that to obtain higher strength wire with a given
deformation, \( r_0 \) should be as small as possible. That is, the effective grain size (the scale
of the DFM structure) should be refined.

The effect of volume fraction of martensite on the tensile properties of drawn wire
is shown in Fig. 11. However, it is difficult to vary the volume fraction in controlled
rolling processing because the finish temperature does not affect this parameter
appreciably over the range which is needed to obtain lath martensite. What is critical is
that the quench be rapid enough to prevent further \( \gamma \rightarrow \alpha + \gamma \) decomposition above the
Ms temperature which would result in high carbon plate martensite or bainite formation.
As shown in Fig. 9, this is undesirable.

The hardenability of the austenite phase is determined by partitioning of alloying
elements, especially carbon, during the time the steel is in the \((\alpha + \gamma)\) phase field.
Independent research using microanalytical techniques (17) indicates slow partitioning of
Mn and no significant partitioning of silicon but it is estimated from CCT diagrams
corresponding to the estimated austenite composition that water quenching should be
effective for producing the desired dual phase steels as rods or bars up to 3/4" diameter.

In conclusion, the present research shows that dual phase steels can be designed and
processed as new, economical low carbon steels for cold drawing into high tensile strength
steel wires. Current work indicates wires of tensile strengths up to 400,000 psi can be
obtained (Figs. 10-13). Potential applications for dual phase steel wire include bead wire,
tire cord, wire rope and prestressed concrete (Fig. 14). It should be possible to produce
wire rods in existing rod mills by adapting the controlled rolling and quenching procedures
outlined in this paper.

Acknowledgements

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Canada and Arbed Saarstahl, W. Germany for providing the steels used in this investigation. We also thank Dr. R. M. Fisher, Center for Advanced Materials, Lawrence Berkeley Laboratory, Berkeley, CA for his helpful discussions.

References
17. M. Ohmura; Ph.D. Thesis, Univ. of California, Berkeley, in progress.
Table 1. Alloy Compositions (wt%) and Phase Transformation Temperatures

<table>
<thead>
<tr>
<th>Alloy</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Fe</th>
<th>Ac₃(°C)</th>
<th>Ar₃(°C)</th>
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<td>A</td>
<td>0.08</td>
<td>1.89</td>
<td>0.32</td>
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<td>0.004</td>
<td>bal</td>
<td>1030</td>
<td>940</td>
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<tr>
<td>B*</td>
<td>0.084</td>
<td>1.05</td>
<td>1.62</td>
<td>0.016</td>
<td>0.021</td>
<td>bal</td>
<td>850</td>
<td>-</td>
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<tr>
<td>C</td>
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<td>0.25</td>
<td>1.08</td>
<td>0.004</td>
<td>0.005</td>
<td>bal</td>
<td>830</td>
<td>740</td>
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</table>

*Commercial Welding Rod: samples provided by Stelco, Canada.

Table 2. Drawability

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<tr>
<td>A1</td>
<td>CHR</td>
<td>~20%</td>
<td>&lt;0.0105</td>
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<tr>
<td>A2</td>
<td>IQ</td>
<td>~20%</td>
<td>&lt;0.0105</td>
</tr>
<tr>
<td>A3</td>
<td>IQ</td>
<td>~40%</td>
<td>0.0136</td>
</tr>
<tr>
<td>A4</td>
<td>IQ</td>
<td>~50%</td>
<td>0.0284</td>
</tr>
<tr>
<td>B</td>
<td>IQ</td>
<td>~30%</td>
<td>0.0136</td>
</tr>
<tr>
<td>C</td>
<td>CHR</td>
<td>~20%</td>
<td>&lt;0.0105</td>
</tr>
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</table>

'CHR' refers to the controlled hot rolling treatment
'IQ' refers to the intermediate quenching treatment
Table 3. Tensile Properties
(as heat treated)

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<tr>
<td>A1</td>
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<td>79.2</td>
<td>119</td>
<td>25.5</td>
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<td>IQ</td>
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<td>117</td>
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<tr>
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<td>97</td>
<td>30</td>
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Table 4. Tensile Strengths of Drawn Dual Phase Wire (ksi)

<table>
<thead>
<tr>
<th>Wire dia. (in.)</th>
<th>Total True Strain</th>
<th>A1</th>
<th>A2</th>
<th>A3</th>
<th>B</th>
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<td>119</td>
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<td>201</td>
<td>215</td>
<td>208</td>
<td>176</td>
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<td>195</td>
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<tr>
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<td>245</td>
<td>245</td>
<td>260</td>
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<td>-</td>
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<td>271</td>
<td>287</td>
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<td>-</td>
<td>-</td>
<td>-</td>
<td>307</td>
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<td>369</td>
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<td>.0105</td>
<td>6.06</td>
<td>407</td>
<td>403</td>
<td>-</td>
<td>-</td>
<td>380</td>
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Figure Captions

Fig. 1. Schematic of the controlled rolling treatment. (1) Rough rolling in austenite recrystallization region. (2) Intermediate rolling in austenite non-recrystallization region. (3) Finish rolling (e.g. no twist block).

Fig. 2. Schematic of the intermediate quenching treatment.

Fig. 3. Scanning electron micrograph of dual phase wire rod obtained by the controlled rolling process. Specimen A1.

Fig. 4. Scanning electron micrograph of dual phase wire rod obtained by the intermediate quenching treatment. Specimen B.

Fig. 5. Transmission electron micrograph of initial wire rod showing lath martensite in ferrite matrix developed in the specimen A2. M: martensite F: ferrite.

Fig. 6. Scanning electron micrographs of dual phase steel wire as a function of reduction in area (R.A.) by cold drawing (A) 0% R.A. (ε=0); (B) 70% R.A. (ε=1.2); (C) 88% R.A. (ε=2.1); (D) 97% R.A. (ε=3.6). Alloy A.

Fig. 7. Transmission electron micrograph of dual phase steel wire after a total drawing strain of 3.6. Specimen A2.

Fig. 8. Transmission electron micrograph of pearlitic wire drawn to 0.005 in dia. (Courtesy of R. M. Fisher)

Fig. 9. Scanning electron micrograph of highly deformed ferrite around non-deforming martensite particles. (ε=3.6). Notice the void formation near the martensite-ferrite interface.

Fig. 10. Plot of tensile strength of dual phase steel wire as a function of wire diameter. Specimen A1.

Fig. 11. Plot of tensile strength of dual phase steel wire as a function of wire diameter. Specimen A2, A3.

Fig. 12. Plot of tensile strength of dual phase steel wire as a function of wire diameter. Specimen B.
Fig. 13. Plot of tensile strength of dual phase steel wire as a function of wire diameter. Specimen C.

Fig. 14. Comparison of the drawing schedule and resulting tensile strength for dual-phase wire and patented pearlitic wire. (Courtesy of R. M. Fisher).
Fig. 1
Fig. 7
CONTROLLED ROLLING TREATMENT

\[ ~20\% \text{ Ms} \]
Fig. 11

Wire Diameter (mm)

Tensile Strength (ksi)

40% Ms

20% Ms

INTERMEDIATE QUENCHING

\[ \sqrt{D_0/D} \]
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