Communications

Low-Temperature Mechanical Behavior of an "Acicular" Ferrite HSLA (High Strength-Low Alloy) Line Pipe Steel

M. R. KRISHNADEV AND R. GHOSH

In response to the need for strong and tough steels to carry oil and gas from remote arctic regions several HSLA steels have been developed.[1] One of the prime contenders for these applications is the "acicular ferrite" steel[2] which derives its good combination of strength and toughness due to its extremely fine grained and highly dislocated substructure (see Fig. 1). Although several studies[3-5] have been made on the effect of thermomechanical processing on their microstructure and conventional tensile and impact properties, there is hardly any information in the literature on the effect of low temperature on their ductility, notch sensitivity ratio, and fatigue properties. Such information is necessary to ensure the stability of the structures under the severe climatic conditions of the arctic and to characterize, to a greater degree, the mechanical behavior of the acicular ferrite matrix. For this reason, a comparative study of the effect of low temperature on the mechanical behavior of various candidate steels for Arctic line pipe applications has been undertaken and this note discusses the preliminary results obtained for an acicular ferrite steel.

Fig. 1—Microstructure of "acicular ferrite" HSLA Steel, magnified 17,600 times.

Experimental. The steel used in the study was supplied by Steel Company of Canada (Stelco) and has the composition: 0.05C, 1.93Mn, 0.26Ni, 0.43Mo, 0.065Nb. The material was in the form of 19 mm thick plates. For tensile testing, 6.3 mm diam round specimens with a 25 mm gage length were used; for notched tensile testing, the round specimens had 60 deg notch so that the cross-sectional area at the root of the notch was equal to half the cross-sectional area of the unnotched specimens. For high cycle fatigue tests, 17 mm thick, 355 mm long constant stress tapered specimens (with a polished surface) were used. Testing was carried out in a Fatigue Dynamics Model 500 plate bending fatigue machine at a speed of 1800 cpm. Controlled evaporation of liquid nitrogen was used to obtain low temperatures. A temperature of −129°C was chosen for the low temperature fatigue tests as it was i) lower than the transition temperature (−110°C) of the steel and ii) lower than any temperature that the steel would encounter in its service in the arctic. Tensile tests were carried out at different temperatures from *RT* (21°C) down to −196°C.

Results. 1) Tensile Properties. Figure 2 shows the variation in the yield (0.2 pct offset) and ultimate strength of the steel as a function of testing temperature. It is apparent from the figure that both the strength values increase with a decrease in the testing temperature, the increase being dramatic below about −80°C. The overall increase in the yield strength when the temperature is lowered to −196°C, is about 400 MPa and the increase in the ultimate strength is about 372 MPa. This increase is in the normal range of strength increment which has been reported for low carbon polycrystalline irons[6] and iron base alloys.[2] Thus "acicular ferrite" shows similar temperature dependence of yield strength as low carbon iron alloys and iron base alloys. This dependence has been largely attributed[8] to the larger effective stress needed to move dislocations at low temperatures past short range obstacles (Peirels stress, impurity atoms, and forest dislocation) as sufficient thermal energy for dislocation motion is lacking. Although the low temperature strength properties follow the classical pattern, the ductility (elongation at fracture) values show rather a surprising trend: there is an increase in the percentage elongation values with a decrease in the testing temperature, particularly around −160°C (see Fig. 3). The authors are not aware of any earlier study in the literature regarding such an observation in HSLA steels. The reasons for this phenomenon are not very clear. This steel by virtue of its high TS/YS ratios would appear to have high work hardening rate; therefore, be resistant to necking in tension test and exhibit good elongation values. One contributing factor, certainly, is the large number of mobile dislocations present in

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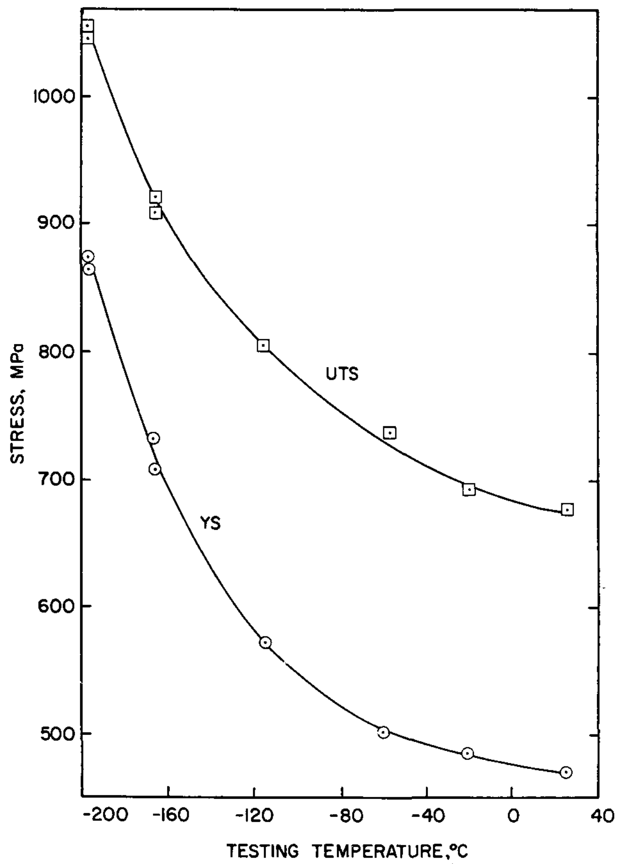

Fig. 2—Strength values as a function of testing temperature.

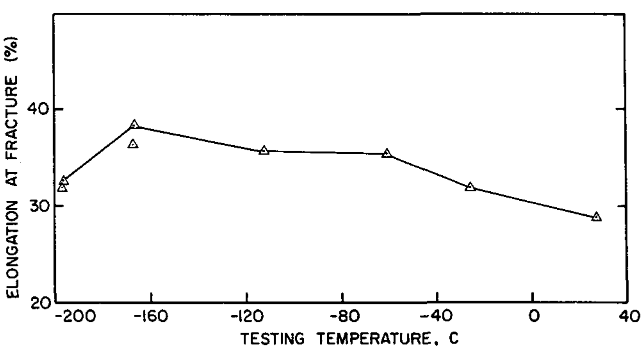

Fig. 3—Elongation values as a function of testing temperature.

the acicular ferrite matrix. Since the matrix is highly dislocated it would have a large number of mobile dislocations even at low temperatures. The presence of these mobile dislocations would tend to limit the tendency for cleavage and permit slip to take place.[9] One conclusion that can be drawn from this observation is that dislocations in the acicular ferrite matrix are mobile at low temperatures. Although no thin foil work has been carried out on specimens deformed in tension, electron microscopy has been carried out on specimens fatigued at -129°C which show dislocation structures (cells, tangles, loops, dislocation generation at grain boundaries, and so forth) indicative of considerable dislocation motion (see Fig. 4).

On plotting the logarithm of the yield strength as a function of the inverse of the absolute testing temperature (Arrhenius plot) a liner relationship was obtained (see Fig. 5). It is apparent from the figure that yield strength obeys the relation $\sigma = Ae^{-K/T}$ where σ is the

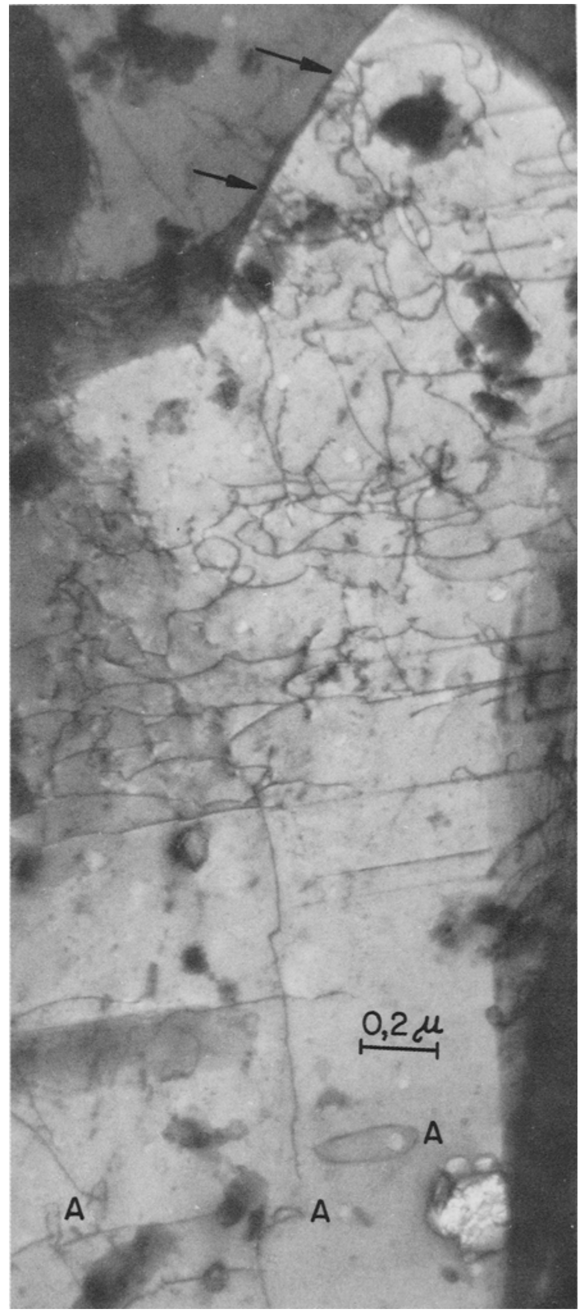

Fig. 4—Microstructure of a specimen fatigued at -129°C (stress level 414 MPa) showing dislocation tangles, loops (as at A) and dislocation generation at grain boundaries (as shown by arrows), magnified 50,000 times.

yield strength, T the absolute testing temperature, A and K are constants. The plot may be used to estimate the yield strength of the steel at other temperatures.

2) Notched-Tensile Properties. Figure 6 shows the notch sensitivity ratio (NSR) as a function of testing temperature. NSR is the ratio of notch strength to the ultimate tensile strength of the unnotched specimen and is a measure of the notch sensitivity. If it is less than unity the material is notch brittle.[10] In the present case values are greater than 1.4, even at low temperatures, indicating good notch ductility.

3) High Cycle Fatigue Properties. Results of the fatigue tests are shown in the form of S-N curves

(Fig. 7). The curves exhibit classical shape (distinct knee and an apparent fatigue limit) familiar in the case of mild steel. At high stress amplitudes lives are shorter. With a decrease in the stress amplitude lives increase, and below a certain stress amplitude there is no apparent failure of the specimens. By decreasing the test temperature, the whole curve is shifted upwards indicating a general improvement in the fatigue resistance with a decrease in the test temperature. The fatigue limit increases from about 310 to 345 MPa. The improvement in fatigue strength with a decrease in the test temperature has been generally attributed to the increase in static strength[11] and also to the homogenization of slip[12,13] and curtailment of substructure formation.[14] In the present case it is very likely that substructure formation plays only a minor role as similar dislocation structures have been observed at room temperature as well as at – 129°C (for example, see Fig. 4). It is worth pointing out here that yield and ultimate strength increased by more than 127 MPa while the fatigue limit increased by 35 MPa. However, based on the fatigue ratios (*i.e.* the ratio of fatigue limit to the ultimate tensile strength), the low temperature appears to affect the monotonic and cyclic properties to about the same extent as the fatigue ratio is still 0.42 at – 129°C compared to 0.45 at room temperature. These values of fatigue ratios are within the normal range (0.4 to 0.6) encountered in steels.

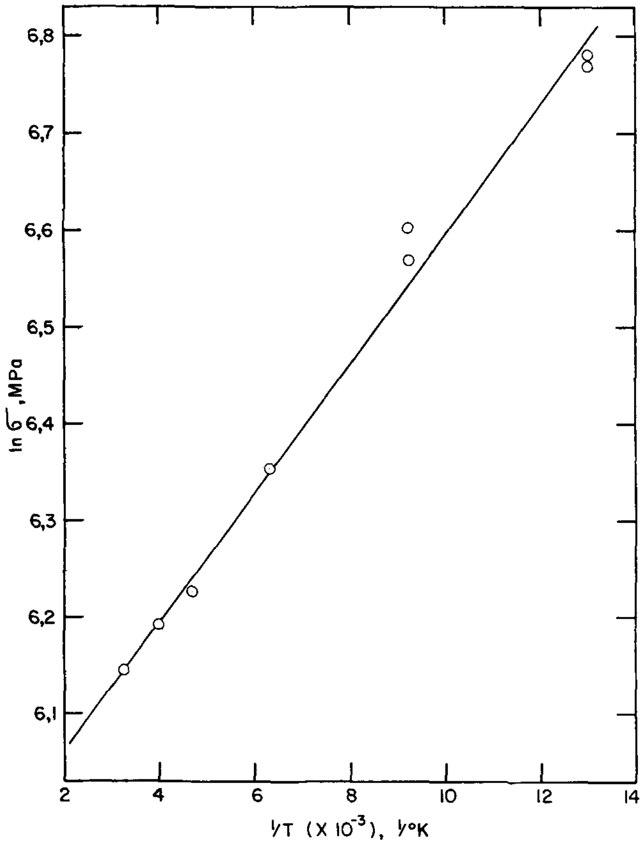

Fig. 5—Arrhenius plot of yield strength and testing temperature.

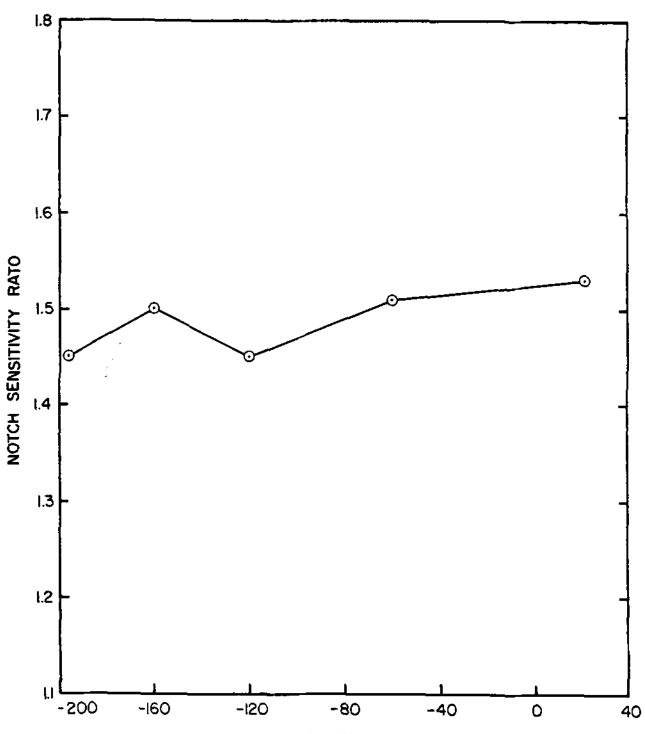

Fig. 6—NSR values as a function of testing temperature.

Conclusions. In an "acicular ferrite" steel i) the temperature dependence of yield and ultimate stress follows the general trend observed in metals of bcc structure, ii) yield strength obeys an Arrhenius type of relation with respect to temperature, iii) elongation values increase with a decrease in the test temperature (up to – 160°C), iv) NSR values are greater than 1.4 even at very low temperatures indicating good notch ductility, v) there is a general improvement in the fatigue resistance with a decrease in the test temperature, and vi) substructure formation plays only a minor role in the prolongation of fatigue resistance at low temperatures.

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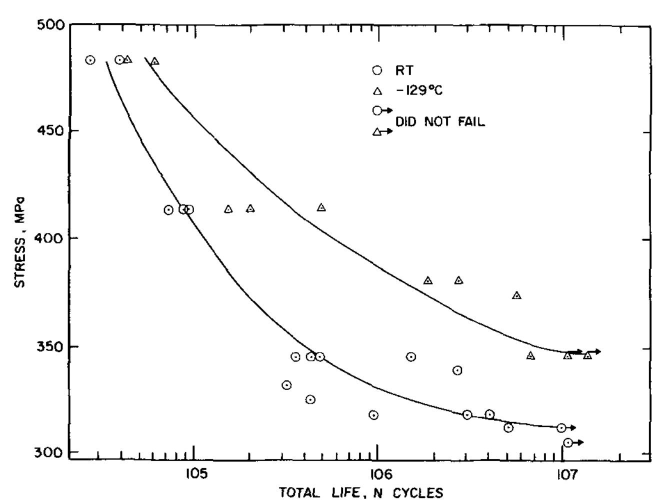

Fig. 7—Effect of low temperature on the high cycle fatigue behavior.

Mr. J. R. Prince and Mr. G. Hamel for carrying out some of the tests.

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Rheological Analysis of High Temperature Creep of Some bccβ Phases: Existence of a Master-Curve

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Few experimental studies have been devoted to the creep of concentrated solid solutions. Many problems, which will be examined further on, arise when this creep is analyzed.[1] The synthesis of experimental results from studies of different bcc β (A_2) phases of a W type and stable at high temperature will be presented. Results concerned with the solid solutions of the equiatomic alloys Ag-Cd,[2] Ag-Zn,[1,3] Cu-Zn,[4,5] Fe-Co-V (Ref. 6) and of Cu 25 pct Al (Ref. 7) are reviewed.

Their constant stress flow is characterized by extensive steady creep with a creep rate $\dot{\epsilon}_s$, followed by a tertiary zone leading to rupture. The lack of primary creep is a general feature of these phases. Besides, internal stress measurements performed with different methods give[8,9] values of σ_i very close to zero whatever the stress field and temperature range[2-7] may be. Therefore, as far as the stress or temperature dependences of the steady creep rate $\dot{\epsilon}_s$ are concerned, one obtains:

$$m \simeq m^* \text{ with } m = \left(\frac{\partial \ln \dot{\epsilon}_s}{\partial \ln \sigma}\right)_T \text{ and } m^* = \left(\frac{\partial \ln \dot{\epsilon}_s}{\partial \ln (\sigma - \sigma_i)}\right)_T$$

and

$$Q \simeq Q^* \text{ with } Q/Q^* = \left(\frac{\partial \ln \dot{\epsilon}_s}{\partial \left(-\frac{1}{kT}\right)}\right)_{\sigma/\sigma - \sigma_i}.$$

Hence, no problem of a choice of a state variable arises when the rheological analysis of creep is considered.[10]

These alloys have m values of about 3.5, except Fe-Co-V which shows *an* parameter increasing with stress for high values of σ.[6] The values of the activation energy for creep are found to be close to the average one of bulk diffusion, except in some cases for the lowest stresses where diffusion short circuits seem to operate.[4]

In an attempt to take into account every physical parameter of creep, the semi-empirical Dorn equation[11] has been used to rationalize the experimental results obtained with the different alloys. This equation takes the form:

$$\frac{\dot{\epsilon}_s kT}{D\,Gb} = A\left(\frac{\sigma}{G}\right)^n \qquad [1]$$

in which A and n are material constants, kT and b have their usual meanings, D is a diffusion coefficient, the determination of which will be discussed later, and G is the Coulomb modulus.

The experimental conditions for each alloy are summarized in Table I.

The evolution of the shear modulus with temperature has been picked up in literature starting from Young's Modulus measurements ($G = 0.4\ E$) (Ag-Cd, Ag-Zn,[12] Cu-Al (Ref. 13) and Fe-Co-V (Ref. 14) and/or dynamically measured on samples in forced vibrations at a frequency of a few hundred Hz in such a way as to obtain a nonrelaxed modulus.[2,3,5] The measured values are actually close to those available in literature.

The choice of the diffusion coefficient to be used in Eq. [1] is a complex problem. Although several theoretical studies have been devoted to this question no definitive conclusions seem to have been reached about the choice of D. According to Brebec and Poirier,[15] if the alloy remains homogeneous one must have

$$D = D' = \phi\ (n_A D_A^* + n_B D_B^*) \qquad [2]$$

in which n_A and n_B are the atomic fractions of the components A and B of the AB alloy, D_A^* and D_B^* the self diffusion coefficients of A and B in the alloy and ϕ is the thermodynamic factor. If, on the contrary, a stationary stage is reached, then, according to the same authors, D should be:

$$D = \overline{\overline{D}} = \phi\ \left(\frac{D_A^* D_B^*}{n_A D_A^* + n_B D_B^*}\right) \qquad [3]$$

On the other hand Nix *et al*[16] take

$$D = \overline{D} = \frac{D_A^* D_B^*}{n_A D_A^* + n_B D_B^*} \qquad [4]$$

if dislocation climb is the rate controlling process (this formula had already been proposed by Weertman[17] and

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