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C. K. Syn, D. R. Leusuer, O. D. Sherby, E. M. Taleff

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STRESS-STRAIN RATE RELATIONS IN ULTRAHIGH CARBON STEELS DEFORMED IN THE FERRITE RANGE OF TEMPERATURE

C.K. Syn[1], D.R. Lesuer[1], O.D. Sherby[2], and E.M. Taleff[3]

[1]Lawrence Livermore National Laboratory, PO Box 808, Livermore, CA 94551, USA;
[2]Dep't of Materials Science and Engineering, Stanford University, Stanford, CA 94305, USA; [3]Dep't of Mechanical Eng., The University of Texas at Austin, Austin, TX 78712, USA

Keywords: Ultrahigh carbon steel, creep strength, strain rate effect, strain aging, ferrite, ferromagnetism, lattice diffusion, dislocation pipe diffusion, elastic modulus, activation energy, dislocation climb, strengthening mechanisms, Cottrell locking.

Abstract. The stress-strain rate relations in ultrahigh carbon steels (UHCSs) have been analyzed at high temperatures in the ferrite range where dislocation slip is the principal deformation mechanism. Specifically, the present investigation centers on the factors influencing the strength of UHCS in the ferrite range of 500 to 700°C and in the strain rate range of 10^{-3} to 10^{3} s^{-1}. These steels contain iron carbide as a second phase either in the form of spheroidite or pearlite and vary in the grain size according to the processing history. The new variables that need to be taken into account are the effect of ferromagnetism and its change with temperature on the creep strength of the ferritic UHCSs. Recent studies have shown that ferromagnetism strongly influences the lattice and dislocation pipe diffusion coefficient and the elastic modulus of iron in the ferrite range of temperature. These variables are shown to influence the creep strength of ferrite-base steels and explain the high activation energies that are observed in many of the ferritic UHCSs.

Introduction

Ultrahigh-carbon steels (UHCSs) are an emerging class of steels (containing 1–2.1%C) with high ambient-temperature strength, hardness, wear resistance and ductility [1]. It has been shown that the UHCSs with fine grain microstructure obtained through many processing routes are superplastic at elevated temperature and suitable for near net shape forming [1, 2]. In the present study, we examine the creep response of UHCSs in the ferrite range at strain rates and temperatures characteristic of hot metal forming. The strain rate sensitivity is low and deformation is controlled by dislocation slip mechanisms and not by grain boundary sliding. Previous studies have centered on the stress-strain rate relations at high temperatures in the austenite range in the UHCSs [3]. These studies have shown that lattice and dislocation pipe diffusion control the flow stress, and that the amount of carbon in solution and the stacking fault energy also contribute to the strength. Specifically, the present investigation centers on the factors influencing the strength of UHCS in the ferrite range of 500 to 700°C and in the strain rate range of 10^{-3} to 10^{3} s^{-1}. These steels contain iron carbide as a second phase either in the form of spheroidite or pearlite and vary in the grain size according to the processing history. The new variables that need to be taken into account are the effect of ferromagnetism and its change with temperature on the creep strength of the ferritic UHCSs. This is because recent studies have shown that ferromagnetism strongly influences the lattice and dislocation pipe diffusion coefficient and the elastic modulus of iron in the ferrite range of temperature. These variables are shown to influence the creep strength of ferrite-base steels and explain the high activation energies that are observed in many of the ferritic UHCSs. In the following sections, we report new stress – strain rate test results of a UHCS and compare them with the results available in the literature.

Materials and Experimental Procedures

Material used in the present investigation had the composition of 1.25%C, 1.6%Al, 1.5%Cr, 0.5%Mn and balance iron. The material was melted and processed at the Bethlehem Steel Homer Research Laboratory. A 75-mm bar was reduced to a 7.5-mm plate by warm rolling in 8 passes with a start temperature of ~780°C and a finish temperature of ~705°C. The microstructure was essentially in the fully spheroidized condition with fine ferrite grains (2 µm) and cementite particles less than 1 µm in diameter.

Specimens were tested in compression from 650 to 800°C in 50°C increments. The specimens were right circular cylinders of 7.6 mm height and 5.1 mm in diameter with an aspect ratio of 1.5. The strain rates were varied from slightly above the conventional test rates (10^{-2} s^{-1}) up to commercial forming rates (20 s^{-1}). The load was applied by using a servo-controlled, electro-hydraulic test machine and the specimens were heated in a quartz lamp furnace.

Experimental Results and Analysis

The true stress – true strain curves were obtained for different temperatures and strain rates. Figure 1 shows examples of the stress – strain curves at 700°C for five different strain rates and one at 750°C for the maximum strain rate of $2x10^1$ s^{-1}. The reduction in flow stress in the plastic flow ranges observed at high strain rates is caused by adiabatic heating. Figure 1 allows us to determine the stress exponent and the activation energy. The stress exponent was about seven. An activation energy for creep was calculated, corrected by the modulus as done for the austenitic UHCSs [3]. The approximate value was found to be 280 kJ/mol. It was decided to compare the results on the UHCS – 1.25%C material with other high carbon and ultrahigh carbon steels deformed in the slip creep region in the ferrite range. Table 1 lists available sources of high strain rate creep data in other low and high carbon steels including UHCSs. Data in Table 1 also include information on composition, test method, and microstructure.

Table 1. Composition, test method, microstructure, temperature - strain rate range, and activation energies for creep at a fixed modulus-normalized stress, $Q_C|_{\sigma/E}$, lattice diffusion, Q_L, and dislocation pipe diffusion, Q_P.

Material, Reference	Test Method	Microstructure	Temperature, Strain rate	$Q_C\|_{\sigma/E}$, kJ/mol	Q_L, kJ/mol	Q_P, kJ/mol
Mild Steel 0.3%C Manjoine [4]	Tension	Unknown	400-600°C, 10^{-3} - 10^3 s^{-1}	245	~350	240
High Purity Eutectoid Steel-0.8C% Harrigan and Sherby [5]	Compression	Spheroidized 150 µm austenite grain size	500-700°C	295	~350	240
1.25%C-1.5%Cr Rosenberg et al [6]	Hopkinson Bar	Quenched & Tempered	500-745°C, $1x10^3$ s^{-1}	-N/A	-N/A	-N/A
UHCS-1.25C-1.5Cr-1.6Al-0.5Mn Present investigation	Compression	Ferrite grain size 2 µm, cementite <1 µm	700-750°C, 10^{-2} - 20 s^{-1}	280	~400	240
UHCS-1.25C% -1.5C-3%Si-0.5%Mn Oyama et al [7]	Tensiom	Spheroidized, Grain size, 2.5 µm Cementite, 1 µm	650C-700°C, $4x10^{-3}$ to $1.5x10^{-2}$ s^{-1}	300	~400	240

Discussion

In order to establish the factors influencing plastic flow of UHCSs in the ferrite range, it is necessary to know the diffusion rate in body-centered-cubic iron in the lattice and in the dislocation site. Such data on diffusion are shown in Figure 2 [8]. Two unusual features are observed. The logarithm of the lattice diffusion rate is shown to be a non-linear function of the reciprocal absolute temperature. This is because iron is ferromagnetic and it loses its ferromagnetic strength as a function of temperature. This leads to an increasing rate of diffusion as a function of increasing temperature up to the Curie temperature (770°C, 1043K) when it becomes paramagnetic. Thus, an increase in the activation energy for lattice diffusion will take place with an increase in temperature. This activation energy is better described as an apparent activation energy since the ferromagnetic state is different at each temperature. A value of 300 kJ/mol is observed at low temperature increasing to about 430 kJ/mol near the Curie temperature.

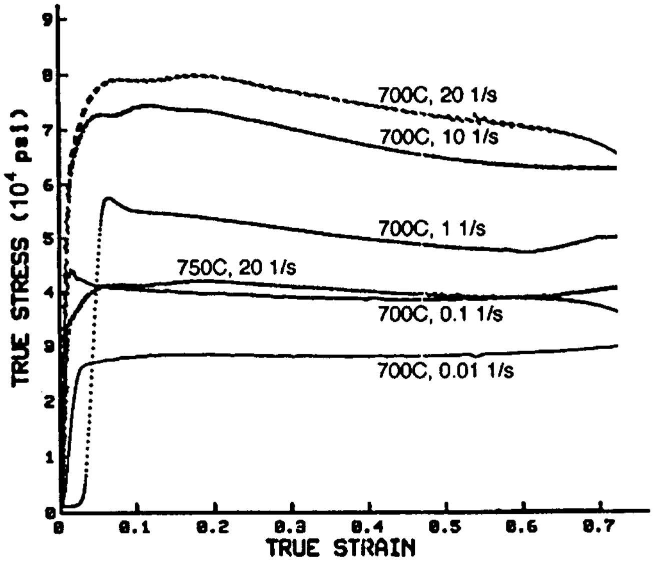

Figure 1. True stress – true strain curves for five different strain rates at 700°C and one curve for the highest strain rate of $2x10^1$ s^{-1} at 750°C.

The data for dislocation pipe diffusion are shown from two different sources in Figure 2. The diffusion data of Shima et al [8] is for super-purity iron containing only 0.5 mass ppm carbon. The observed diffusion rate is higher than that reported by Mehrer et al [9] for iron containing 16 mass ppm carbon. The difference is likely related to influence of carbon in solution on the ferromagnetic state of iron atoms located at the dislocation site. It was decided to select the higher carbon iron diffusion data by Mehrer et al since UHCSs would have the maximum amount of carbon in solution.

The lattice and dislocation pipe diffusion data are used to calculate an effective diffusion coefficient established for describing the high-temperature creep behavior in metallic polycrystalline solids [10]. The equation is given as

$$D_{eff} = D_L + 50\,(\sigma/E)\,D_P \qquad (1)$$

where D_L, σ, E, and D_P are the lattice diffusion coefficient, flow stress, Young's modulus, and dislocation pipe diffusion coefficient respectively.

The creep data were analyzed by means of the following relation:

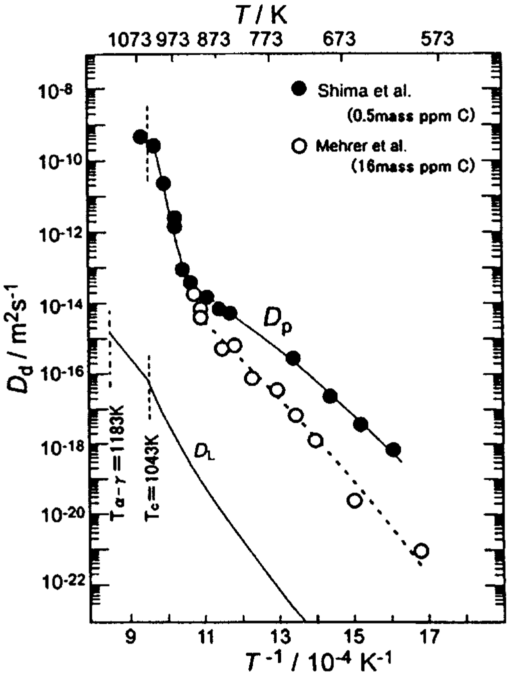

Figure 2. Data for lattice diffusion (D_L) and dislocation pipe diffusion (D_P) for Fe atoms in Fe-C alloys.

$$\dot{\varepsilon} = A\cdot D_{eff}\cdot(\sigma/E)^n \qquad (2)$$

where $\dot{\varepsilon}$ is the deformation rate, n is the stress exponent and A is constant for a given structural state. The modulus data were by Köster [11]. Figure 3 shows a plot of the creep rate normalized by D_{eff} as a function of the modulus compensated stress for the studies documented in Table 1.

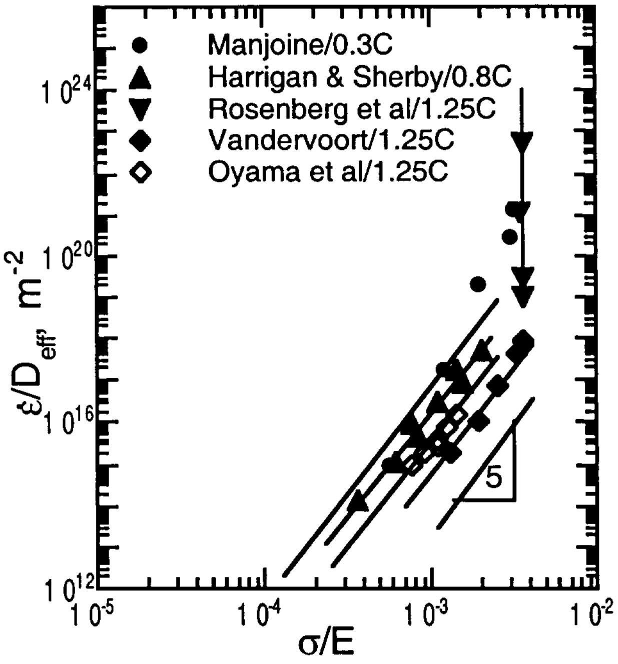

Figure 3. Stress – strain relationship in several steels including UHCSs for deformation at temperature in the ferritic region

Separate parallel lines are seen for four of the five investigations. The factors influencing the strength difference will be discussed later. The stress exponents are between five and eight for these four steels, indicating that creep is controlled by dislocation climb. The fifth investigation is that of Rosenberg et al [6] who conducted dynamic deformation tests ($\dot{\varepsilon} \sim 10^3$ s^{-1}) in the temperature range of 500 to 745°C. The data showed athermal type behavior. That is, the modulus-compensated flow stress was independent of temperature and strain rate equal to ~$2.5\text{x}10^{-3}$. It is proposed that this is a result of dynamic strain aging. That is, the condition where the velocity of dislocations during dynamic deformation is controlled by the rate of carbon atom diffusion. Support for this proposal is the work of Manjoine [4] on strain rate – temperature effects in deformation of mild steel. His data, in the temperature range of 400 to 600°C are shown in Figure 3. At low stresses the data are contained within the power law region with power law breakdown occurring at increasing stresses. Remarkably, at the highest strain rate, the mild steel data merge into the data of Rosenberg et al for an ultrahigh carbon steel. It is accepted that so-called athermal effects in mild steel are a result of strain aging [4] that is explained by Cottrell locking of dislocations by solute carbon atoms.

The difference in strength of the materials tested in the power law region, Figure 3, is related to the microstructure of the respective materials. The two highest strengths are observed in the UHCS-1.25%C-1.6%Al and UHCS-1.25%C-3%Si alloys. These UHCSs consisted of ultrafine ferrite grains (2 and 2.5 μm respectively) and ultrafine carbide particles (~1 μm). The pure Fe-C eutectoid composition material revealed a coarse austenite grain size of 150μm. This would imply a ferrite grain size (prior pearlite colony size) no smaller than 15 μm. The mild steel data show the lowest strength and would be expected to have the coarsest structure. The activation energies for the four materials are documented in Table 1. The average value is $Q_C = 280$ kJ/mol. The value of the activation energy for lattice diffusion is about $Q_L = 375$ kJ/mol and the activation energy for dislocation pipe diffusion is $Q_P = 240$ kJ/mol. The fact that the creep activation energy is somewhat between the two diffusion activation energies gives support to the validity of an effective diffusion coefficient.

Figure 4 illustrates the stress – strain rate relationship for UHCSs in the austenite range of temperature [3] utilizing the relation given by Eq. (2). A wide range of carbon content and dilute alloying additions was evaluated. In the austenite range, carbon is fully dissolved and plays an opposite role to that in ferrite. The carbon contributes to solid solution weakening, because carbon lowers the iron bond strength and increases the rate of iron self-diffusion. The spread in the creep data for UHCSs in austenite range is attributed to substitutional atoms present in UHCS that control the stacking fault energy [3]. A decrease in the stacking fault energy will lead to a decrease in the creep rate. This is because the separation width of the partial dislocations is increased making dislocation climb more difficult.

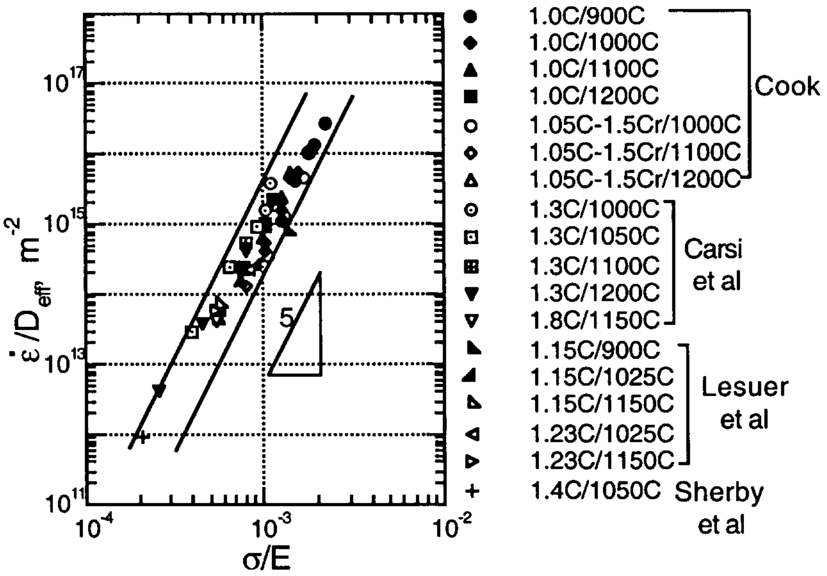

Figure 4. Stress – strain relationship in UHCSs for deformation at temperature in the austenite region [3].

The effect of carbon in the case of UHCSs deformed in the ferrite range is the opposite to that in the austenite as stated earlier. The small amount of carbon in solution lead to solute - dislocation interaction by Cottrell locking. The remainder of carbon contributes to strengthening by the formation of iron carbide particles. These particles are not only barriers to plastic flow but they also contribute to strengthening by grain size refinement.

The present analysis on the creep of iron-carbon materials in the ferrite range of temperatures illustrates the importance of ferromagnetism in creep when deformed in the range from 400°C to the Curie temperature. The activation energies for lattice and dislocation pipe diffusion are exceptionally high at Q_L = 300 to 430 kJ/mol and Q_P = 240 kJ/mol. This information has not been noted in the literature. As an example, the activation energy for lattice and dislocation pipe diffusion in the ferromagnetic range is commonly quoted as Q_L = 251 kJ/mol and Q_P = 174 kJ/mol respectively [12]. These values have been accepted mostly because diffusion information in iron has been guided by the activation energy for paramagnetic alpha and gamma iron. The values are listed as Q_L = 239 kJ/mol for lattice diffusion in paramagnetic BCC alpha and delta iron and as Q_P = 174 kJ/mol [12]. The use of the later knowledge has misled investigators who find high activation energies for creep of iron base materials in the ferrite range. As a result the conclusion has been made that the creep of such a steel is not a dislocation climb creep process controlled by lattice diffusion [13]. Diffusion along the dislocation line in iron is also strongly influenced by ferromagnetism. This is clearly seen in Figure 2 for the case of super-high purity iron containing only 0.5 mass ppm carbon [8]. A sharp increase in the activation energy for pipe diffusion occurs at about 600°C. This occurs because of the continuous change of electronic structure of iron atoms with the change in the ferromagnetic. The sharp decrease in the elastic modulus with increase in temperature beginning at 600°C [11] is further proof of this gradual change in electronic structure.

Figure 5 compares the results of the creep behavior of UHCSs in the ferrite range with those of UHCSs in the austenite range. That is, the data of Figures 3 and 4 are replotted as one with the use of Eq. (2). All the data for the austenite results (in Figure 4) are given as small open circles. The major observation is that the creep data obtained in the austenite range are located in the lower portion of the data for the data obtained in the ferrite range. The faster average creep rate of UHCSs in the ferrite range than in the austenite range maybe be a result of the higher stacking fault energy in ferrite. This is possible because it is accepted that the stacking fault energy in body-centered-cubic metals is very high and in the order of 300 mJ/m^2 and higher. Another observation with respect to the data given in Figure 5 is that the average slopes in the Fe-C steels tested in ferrite are a little higher than the slope of five noted for the austenite data. The stress exponent is 6 to 7.5 especially at the higher stress end of the data. A stress exponent higher than five has been related to constant structure creep [14]. That is, to creep where the structure does not change with stress. When the stress exponent is five, the structure is often related to creep where the structure changes as a function of stress. The structure change refers to subgrain formation where the subgrain size decreases with a decrease in stress. This trend has been proposed for the case of creep of steels when tested in the austenite range. On the other hand, the possibility of constant-structure creep conditions for creep in the ferrite region are in agreement with the present observations given in Figures 3 and 5. The constant structure is proposed to be unique for each Fe-C steel with the different processing steps used to prepare these low, high and ultrahigh carbon steels.

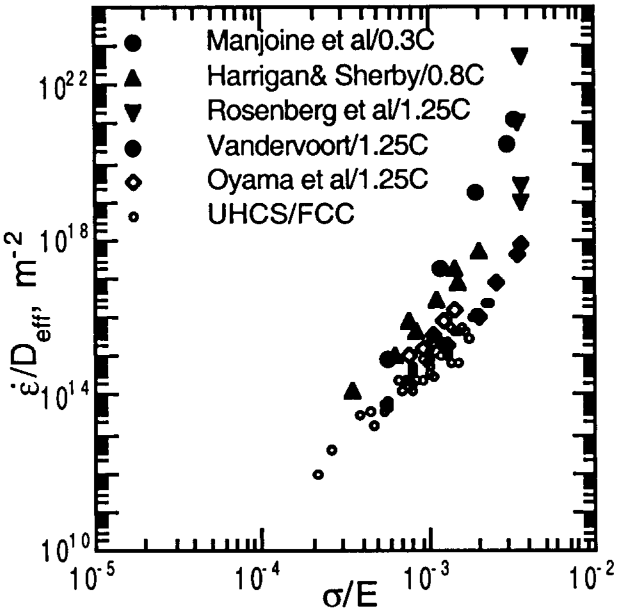

Figure 5. Comparison of the stress – strain rate relationship for deformation in ferrite and austenite ranges of temperature by plotting data included in Figures 3 and 4.

Acknowledgments

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