Tr. J. of Engineering and Environmental Science
23 (1999) , 389 – 401.

On the Creep Strength-Rupture Ductility Behaviour of 1.25Cr-0.5Mo Low Alloy Steel

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Received 23.12.1998

Abstract

It is important to select materials having both high strength and good ductility in most engineering applications. This is also true for high temperature problems and the requirement of high creep strength as well as high rupture ductility is of prime concern. The creep data of ferritic low alloy 1.25Cr-0.5Mo steel, which is extensively used in piping systems of fossil power plants, has shown loss of rupture ductility after certain exposure time depending on its initial microstructure and applied stress. This unexpected loss of ductility is called as rupture ductility trough, and this study covers the attempts to investigate the dependency of creep rupture ductility of 1.25Cr-0.5Mo steel on microstructure and stress, in the temperature range of 773-923 K. An empirical relationship in exponential character was developed to predict the rupture time leading to rupture ductility trough, depending on temperature and the initial structure of steel. Finally, a rupture ductility map was constructed for each steel investigated in this study, in order to present rupture strength-rupture ductility data in a more appropriate way.

Key Words: Low Alloy Ferritic Steel, Rupture Ductility, Creep Strength, Rupture Ductility Trough, Rupture Ductility Map.

Düşük Alaşımlı 1,25Cr-0,5Mo Çeliğinin Sürünme Dayanımı ve Sünekliği Davranışı Üzerine

Özet

Birçok mühendislik uygulamasında yüksek dayanım ve sünekliği birlikte sağlayan malzemelerin seçimi önemlidir. Bu durum yüksek sıcaklık problemleri için de geçerli olup, yüksek sürünme dayanımı ve sünekliğine sahip malzemelerin seçimine dikkat edilmektedir. Özellikle termik santralların boru sistemlerinde yaygın olarak kullanılan ferritik ve düşük alaşımlı 1.25Cr-0.5Mo çeliği, uygulanan gerilme, sıcaklık ve iç yapıya bağlı olarak sürünme sünekliğinde beklenmeyen düşüşler gösterebilmektedir. Bu çalışmada öncelikle söz konusu çeliklerde görülen ve Sürünme Sünekliği Oluğu olarak isimlendirilen bu davranışı etkileyen faktörler 773-923 K sıcaklık aralığında incelenmektedir. Oluğun oluşması muhtemel hasar süresini tahmin edebilmek için, sıcaklığa ve çeliğin başlangıç yapısına bağlı olarak üstel bir ampirik bağıntı geliştirilmiştir. Ayrıca kullanılan çeliğe ait sürünme dayanımı ve sürünme sünekliği özelliklerinin grafik olarak daha uygun şekilde sunulması amacıyla, çalışma kapsamındaki çelikler için Sürünme Süneklik Haritaları oluşturulmuştur.

Anahtar Sözcükler: Düşük Alaşımlı Ferritik Çelik, Sürünme Sünekliği, Sürünme Dayanımı, Sürünme Sünekliği Oluğu, Sürünme Süneklik Haritası.

Introduction

It is well known that the rupture ductility of many alloys varies with temperature, stress and exposure time due to the effects of thermally activated processes producing some microstructural changes such as recovery, recrystallization, precipitation, change in size and distribution of precipitates, during the prolonged testing or service time. Many creep resistant transformable low alloy steels, primarily Cr-Mo and Cr-Mo-V types, exhibit an anomalous change in rupture ductility behaviour with rupture time, which is actually pronounced as rupture ductility trough (Pickering, 1991). It is briefly defined as the drop in rupture ductility with increasing rupture time following an unexpected increase in ductility for longer rupture times (Glen 1955, Stone et al 1965, Viswanathan 1977). This kind of behaviour was also observed in some precipitation strengthened austenitic stainless steels (Irwine et al, 1960). In general, the creep ductility is determined by the superposition of strains accumulated in void formation and growth phases separately (Ashby et al, 1979), and is affected by rupture time, which is depending on applied stress as well as the steady state creep rate of steel. In other words, the decrease in ductility can be tied up to the quick void formation behaviour causing brittle fracture while increase in ductility may be regarded as a result of delayed void formation and the change in fracture mechanism from intergranular to transgranular which are very effective on crack growth characteristics of steel leading to ductile behaviour. Another feature of the decrease in ductility is its strong dependency on chemical composition and prior microstructure of steel as well as the exposure temperature (Purmensky et al, 1993). Ductility drop occurs later yielding lower values of minimum rupture ductility at lower temperatures of creep exposure. When the general features and the results of previous studies are encountered, it can be concluded that the formation of ductility troughs in such steels is depending primarily on the kinetics of carbide formation and growth behaviour together with their subsequent effects on the grain boundaries during creep deformation. The present study was originated from a previously published creep data of 1.25Cr-0.5Mo steel (NRIM Creep Data, 1981), in order to have more insight on the creep strength-rupture ductility changes with temperature and rupture time by investigating the effects of temperature and stress on the occurrence of ductility troughs. An attempt has also been made to present the available creep strength-rupture ductility data of 1.25Cr-0.5Mo steel in the form of Rupture Ductility Maps, in which the equal-ductility regions are defined in a stress-temperature space specific to each steel under investigation.

Materials

The chemical compositions and manufacturing details of the steels under investigation are given in Table-1 and Table-2 respectively. Si-killed steels were produced in plate form having approximately the same mass and dimensions. They are all in normalised and tempered condition, and due to small deviations in compositions and/or the applied thermo-mechanical treatments the hardness of steels vary from 155 to 210 Vickers Hardness Numbers, while the sizes of grains remain almost the same in the range of 6 -7 ASTM Grain Size Number. The difference in the hardness of steels with different batches might probably be attributed to the cooling rates in normalising treatment which may cause probably bainitic and/or ferritic-bainitic structures as well as different carbide dispersions (Purmensky et al, 1993). It should also be expected that following tempering treatment might have some effects and produce wide spectrum of microstructures.

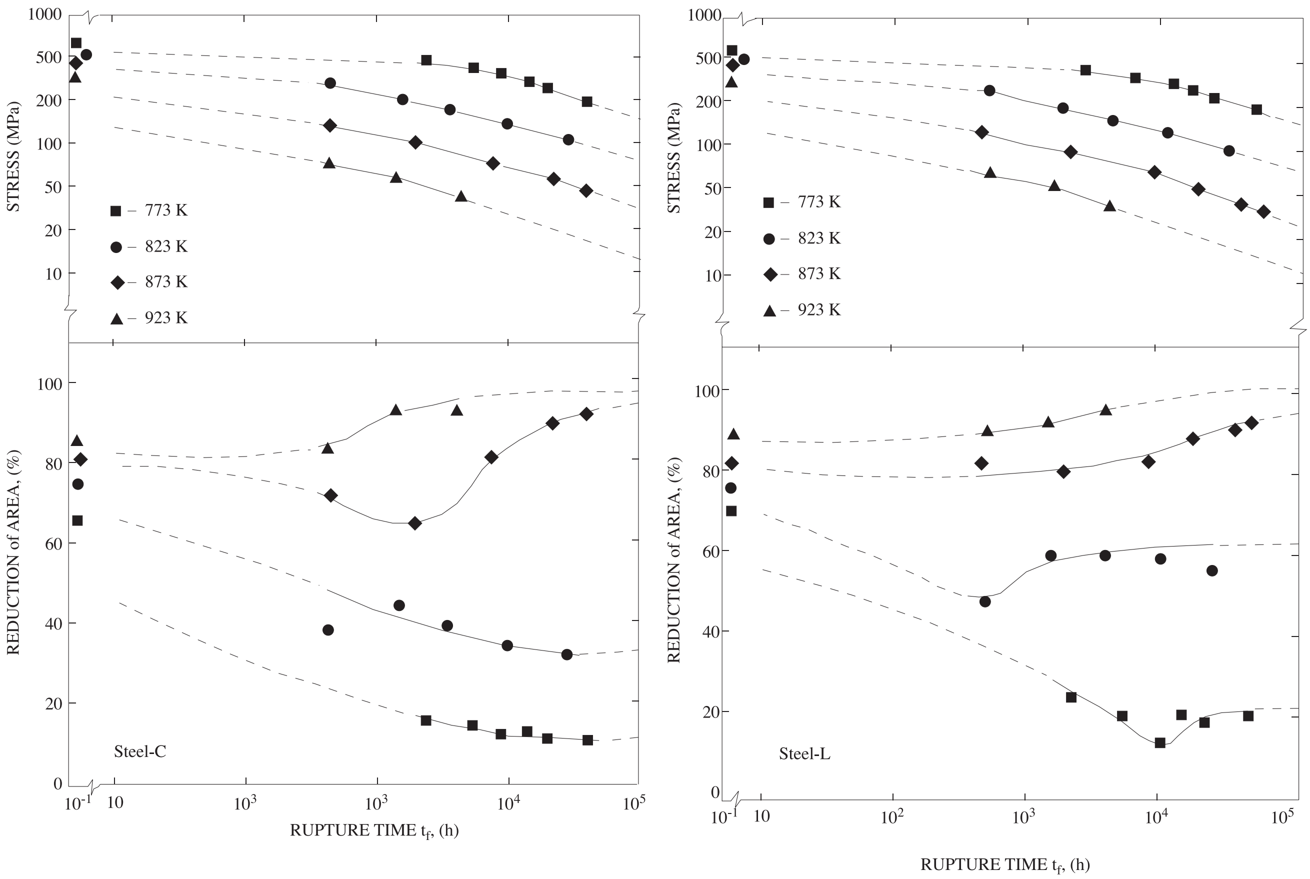
Table 1. Chemical composition of the steels in weight percentage (NRIM Creep Data Sheet, No:21A,1981).

Material	C	Si	Mn	P	S	Cr	Mo	Ni	Cu	Al	N
Steel-C	0.16	0.66	0.56	0.015	0.015	1.27	0.51	0.05	0.04	0.003	0.0072
Steel-L	0.15	0.64	0.58	0.010	0.013	1.27	0.52	0.05	0.03	0.003	0.0062
Steel-V	0.13	0.66	0.58	0.022	0.013	1.24	0.50	0.03	0.03	0.003	0.0095
Steel-E	0.12	0.72	0.50	0.009	0.009	1.20	0.46	0.04	0.04	0.019	0.0043

Table 2. Manufacturing details of the steels (NRIM Creep Data Sheet, No:21A,1981).

Materials	Processing and Heat Treating Details	Grain Size (ASTM No)	Hardness (VHN)
Steel-C	HR / 1223 K, 1.5 hr, AC / 903 K, 2.2 hr, AC	6.4	210
Steel-L	HR / 1223 K, 1.5 hr, AC / 903 K, 2.2 hr, AC	6.3	195
Steel-V	HR / 1223 K, 1.5 hr, AC / 903 K, 2.2 hr, AC	6.6	175
Steel-E	HR / 1183 K, 26 min, AC / 973 K, 77 min, AC / 973 K, 77 min, FC	7.0	155
HR: Hot Rolled, AC: Air Cooled, FC: Furnace Cooled, VHN: Vickers Hardness Number			

The graphical presentation of data on creep strength and rupture ductility at 773, 823, 873 and 923 K for the steels covered in the study are given in Figure-1. In these graphs, the solid lines connect the data points, while the dashed lines are based on the extrapolation of the data considering the tendency represented by final two data points for long rupture times and the tensile test results for the shortest rupture time. The reduction of area values was selected for representation of creep ductility rather than percent elongation values of steels. This is primarily based on the inconsistency of elongation data comparing with reduction of area, which may be originated from some uncertainties in elongation measurements due to its strong dependency on the geometry of specimens and other side effects in creep testing.



(a) (b)

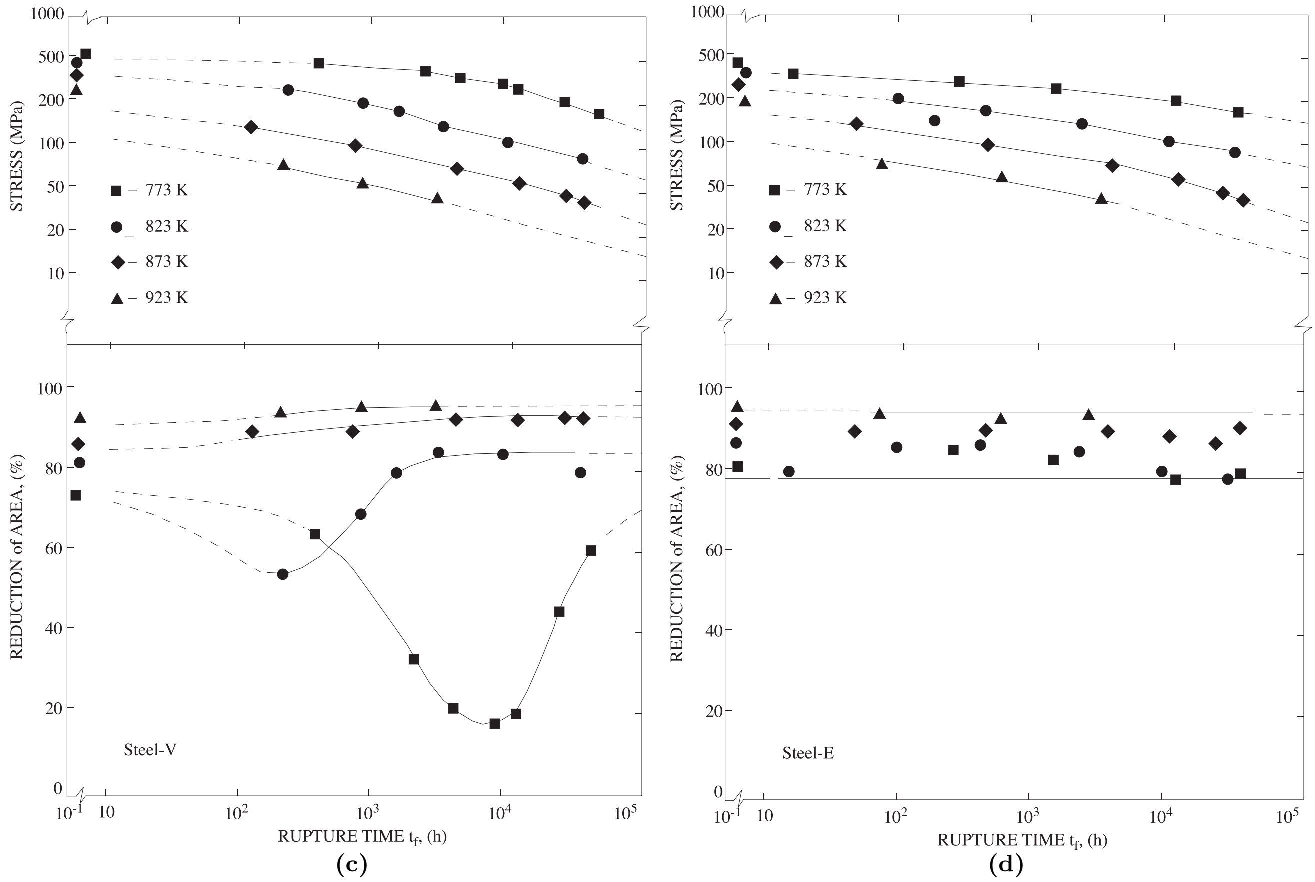


Figure 1. Variation of rupture strength and rupture ductility with rupture time for 1.25Cr-0.5Mo low alloy ferritic steels at different temperatures.

Results and Discussion

In Figure 1, the change in rupture ductility and rupture strength with rupture time is given for all steels covered by this study and the following characteristics can be extracted;

i. all steels exhibit ductility trough except the softest one,

ii. the ductility trough occurs later at lower temperatures,

iii. the ductility trough occurs earlier for softer steels,

iv. the minimum ductility of trough is lower for harder steels at each temperature.

Similar features belong to ductility trough behaviour were reported in previous studies on Cr-Mo steels (Glen 1955, Viswanathan and Fardo 1978). The strong dependency of ductility trough on rupture time and temperature may be regarded as a result of rate controlled processes, which cause some microstructural changes in steel degrading the creep properties respectively. If the chemical compositions and the thermo-mechanical history of the steels were concerned, the most probable microstructural change would be the formation of and growth of hard alloy carbides. To investigate the occurrence of ductility trough of steels depending on temperature and rupture time, Figure 2 was constructed by plotting the logarithmic values of rupture time for minimum ductility against the inverse of absolute exposure temperatures for each steel excluding Steel-E.

A linear relationship holds in semi-logarithmic scale for all steels showing ductility trough behaviour. The results of regression analysis gives an empirical relationship describing the change of rupture times obtained at minimum ductility with temperature such that;

$$t_{fcr} = D^m \cdot \exp\{1000 \cdot m/T\} \quad (1)$$

where t_{fcr} is the rupture time at minimum ductility, D and m are material constants and T is the absolute temperature of exposure. The empirical relationship and the constants of D and m are given in Table-3 for the steels exhibiting distinct ductility trough behaviour.

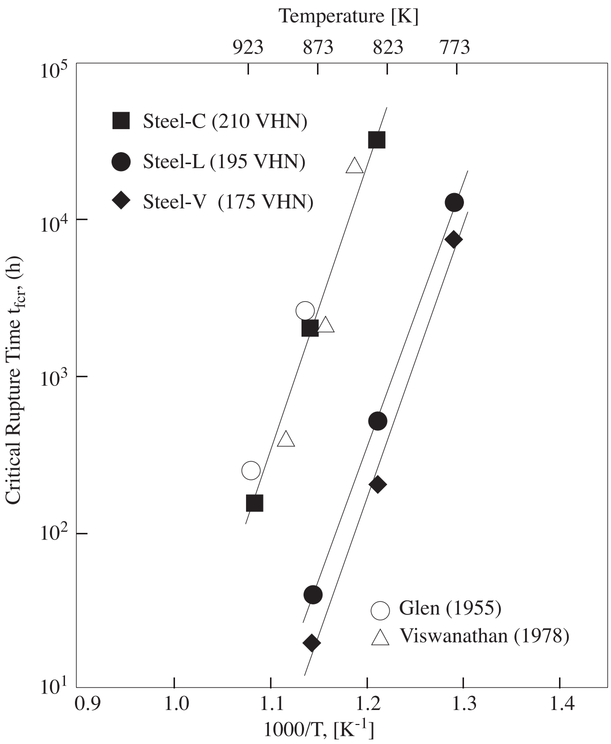


Figure 2. The charge in critical rupture time depending on exposure temperature and the initial hardness of 1.25Cr-0.5Mo low ferritic steels.

Since the power m is almost common for all steels having an average value of 40, the relationship can be expressed in the form of

$$t_{fcr} = D_c \cdot \exp\{4x10^4/T\} \quad (2)$$

where D_c is a preexponential constant depending on the initial structure of steels. If the well-known Arrhenius type function giving the general relationship between temperature and creep rate is considered (Evans and Wilshire 1985), creep rate $\dot{\varepsilon}$ can be expressed in the form

$$\dot{\varepsilon} \sim \exp\{-Q_c/RT\} \quad (3)$$

where Q_c is the activation energy for creep ($J.mol^{-1}$) and R is the universal gas constant (8.3 $J.mol^{-1}.K^{-1}$). Using the assumption of creep rate is depending inversely on the rupture time (Mukherjee et al, 1969), the rupture time causing ductility trough can be expressed by rearranging Equation-3 such that;

$$\frac{1}{\dot{\varepsilon}} \sim t_{fcr} = D_{dt} \cdot \exp\{Q_{dt}/RT\} \quad (4)$$

where D_{dt} is a constant and Q_{dt} is the activation energy controlling the formation of ductility trough which is assumed to be a consequence of primarily the growth of alloy carbides and has an approximate value of 330 kJ.mol $^{-1}$ for the 1.25Cr-0.5Mo steels covered in the study.

The distribution and size of metastable alloy carbides play a major role in the creep behaviour of creep-resistant steels, and the initial particle size, diffusion rate and growth kinetics of carbides are of prime importance (Williams, 1981). When the results of previous studies on the growth behaviour of carbides in Cr-Mo and Cr-Mo-V steels (Williams 1981, Sellars 1974) are considered, it is seen that the magnitude of activation energy for the coarsening of Mo_2C in 2.25Cr-1Mo steel is 176 kJ.mol $^{-1}$, and that for 0.5Cr-0.5M0-0.25V steel is 168 $kJ.mol^{-1}$, which are in the same order of magnitude as the activation energy calculated for the formation of the ductility trough of 1.25Cr-0.5Mo steel. This can be regarded as evidence that the rupture of the ductility trough is closely related to the growth of metastable alloy carbides. Since the stability of alloy carbides can be ranked as M_3C, M_6C, Mo_2C and VC in ascending order, the coarsening of carbide type will preferably be Mo_2C for 1.25Cr-0.5Mo steel, in the absence of vanadium. Figure 2 also includes data obtained from other studies. The data points shown by in triangles belong to the results of a study conducted on 1.25Cr-0.5Mo steel in normalised and tempered conditions with the initial hardness of 210 VHN (Viswanathan and Fardo 1978). The results of another experimental work on 1Cr-0.5Mo steel in only normalised conditions, are given by void circle data points in the same figure (Glen, 1955). Although the hardness of the steel was not reported, it can be expected that it is near 200 VHN owing to its untempered structure. The data representing the effect of exposure temperature on time causing the rupture ductility trough, extracted from different experimental works, show good agreement for the steels having similar initial hardness.

Table 3. The results of regression analysis of the relationship to predict the critical rupture time leading to rupture ductility trough as a function of temperature and initial hardness of 1.25Cr-0.5Mo ferritic low alloy steels.

Materials	D	m	Coefficient of Correlation, R^2	Empirical Relationship[1]
Steel-C	0.384	40.1	0.999	t_{fcr}=2.4x10^{-17}.exp{4x10^4/T}
Steel-L	0.346	40.7	0.999	t_{fcr}=2.4x10^{-19}.exp{4x10^4/T}
Steel-V	0.341	40.0	0.991	t_{fcr}=2.4x10^{-19}.exp{4x10^4/T}

1) Time in hours and temperature in K.

The change in minimum rupture ductility values with temperature is given in Figure 3 for all steels including Steel-E, which is the softest and shows no remarkable ductility trough in the range of 773-923 K. It can be concluded that the level of minimum ductility is higher for softer steels, which have more stable structures, and minimum ductility increases continuously with the temperature of exposure.

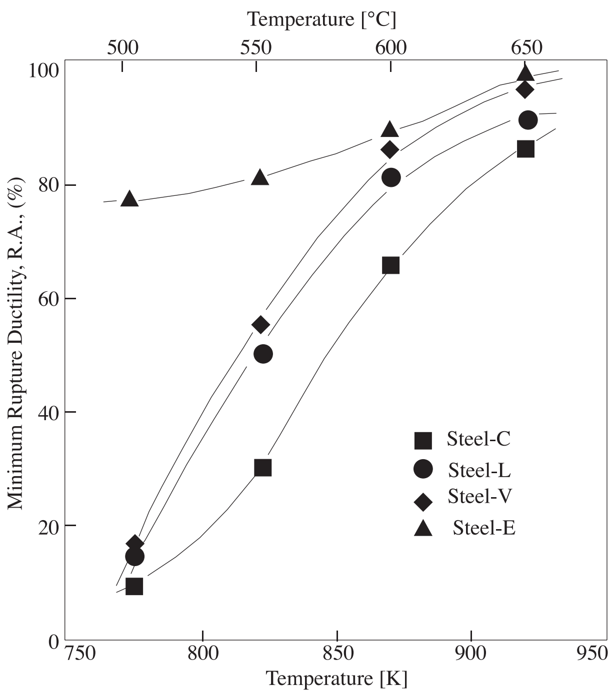


Figure 3. The change in reduction of area at rupture as a measure of rupture ductility with test temperature and initial microstructure of 1.25Cr-0.5Mo low alloy ferritic steels.

The effect of initial hardness of steels, which can be regarded as an effective scale representing the initial microstructure of steels, on the constant D proposed in Equation-1, is given in Figure 4. The value of D has an approximate range between 0.34 for the softest steel reflecting higher stability, and 0.38 for the hardest steel presenting a metastable structure. In other words, in consideration of the lowest value of D, the expected rupture time is too short to obtain microstructural changes causing rupture ductility trough, and for the highest value of D time is too long which practically has no meaning in design considerations for engineering applications. Therefore, it is possible to determine the minimum rupture ductility and the rupture time corresponding to ductility trough by using Equation-2 with the aid of Figure 2, 3 and 4 for normalised and tempered 1.25Cr-0.5Mo steel having initial hardness between 175-210 VHN and in the temperature range of 773-923 K.

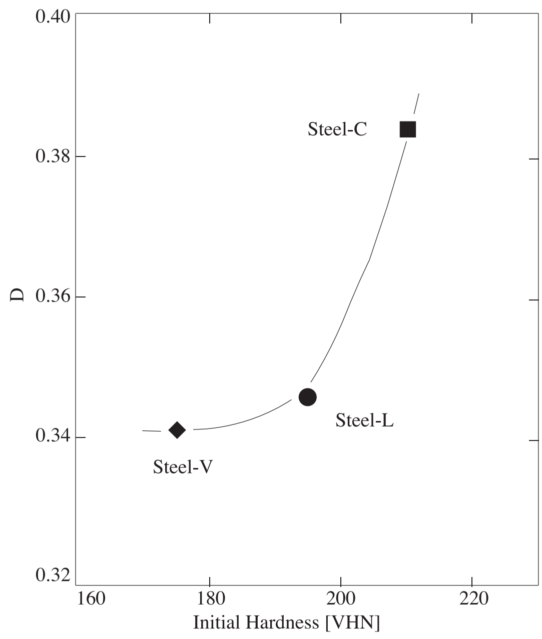


Figure 4. Variation of constant D with the initial hardness of steels exhibiting distinct rupture ductility trough.

Rupture time is inversely related to the minimum creep rate, which can be regarded as an effective indicator of strength of materials at elevated temperatures, by well-known Monkman-Grant relationship

$$\dot{\varepsilon} \cdot t_f = C \tag{5}$$

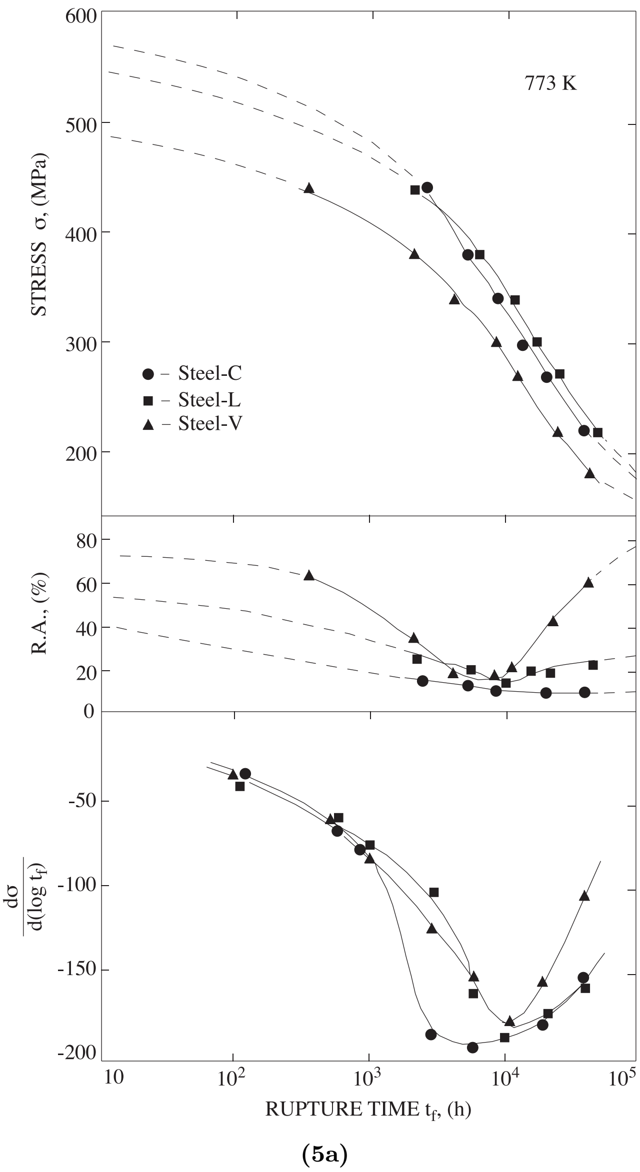
where C is a material constant, and lower creep rates result in longer rupture lives of materials owing to their higher strength (Evans and Wilshire, 1985). Although the change in creep rate depends on various factors, it is generally agreed upon that the most important factor is the size and distribution of alloy carbides in the microstructure of low alloy heat resisting steels (Pickering 1991, Purmensky et al., 1993, Sellars 1974). The finer distribution of carbides results in lower creep rates and longer rupture times respectively, and a change in the size and distribution of alloy carbides should be expected as a consequence of high temperature creep exposure, which is generally pronounced as the degradation of properties.

The damage of creep rupture can be considered as a result of the combination of 2 mechanisms: softening of the structure, and formation and growth of cavities at grain boundaries. It is possible to speak of a critical value of particle size and/or interparticle spacing for particle strengthened alloys generating maximum strength and minimum creep rate of the material. If the structure contains fine distribution of carbides initially, strengthening will occur due to the growth of particles because of the increased distance for dislocations to climb during creep. At the same time, the mechanism of easy bowing of dislocations between and around the loops may take place, which softens the structure. Therefore, it can be assumed that there is a critical size of carbides giving maximum strength under the influence of these mechanisms, which are competing with each other. On the other hand, some delayed precipitation in the grains may also be expected during creep exposure, resulting in some increase in the strength of matrix.

The increase of strength in the matrix or grains will cause relative weakness of the grain boundaries and intergranular damage may be favoured, producing brittle or low ductility fracture. The growth behaviour of carbides is not sufficient to explain the low ductility fracture; it is also necessary to consider the effect of applied stress level. It is possible to define a critical level of stress, which is able to create and grow of cavities at grain boundaries. Stresses below that of the critical value result in longer times for rupture, and the creep rate is accelerated due to the coarsening of carbides exceeding critical size. Therefore, the matrix or grains will be less resistant to creep deformation. The danger of localising the deformation at grain boundaries will diminish gradually and the final fracture tends to be transgranular rather than intergranular, leading to higher rupture ductility. If a low alloy heat resisting steel exceeds the critical values of rupture time without being fractured; it can be interpreted by the insufficient grain boundary damage related to intergranular cavitation at a given stress and temperature state. Since the growth of cavities is basically driven by the mechanism of diffusional flow, the creep rate created by lower stresses may also be insufficient to induce sufficient vacancy flux, which is necessary for the formation and growth of cavities to a critical size. Small-sized cavities at grain boundaries have a tendency to a cause transgranular type of fracture rather than intergranular, yielding higher rupture ductility. At the same time, it is also possible to introduce the effect of sintering, which may cause the annihilation of very small cavities during long time and high temperature creep exposure.

Therefore, it can be concluded that the rupture ductility trough of low alloy creep resistant steels is based primarily on the nucleation and growth behaviour of alloy carbides, as well as voids at grain boundaries. The risk of grain-boundary type damage is dominant when the material is stressed over a critical value, but if the stress too high the ductile and transgranular type of fracture will be favoured and there will be no unexpected change in creep rupture ductility. The growth of carbides to reach a critical size depends on thermal activation, time and initial carbide distribution. Hence, it is quite normal to expect that the ductility trough should occur earlier for higher temperatures of creep exposure, and since the stress leading to a short-time rupture must be high enough, ductile transgranular type of damage will be dominant in at the end. The degree of metastability of steels for the growth of alloy carbides is determined by the microstructural aspect, which is in direct consequence of normalising and tempering heat treatments, if small differences in chemical compositions are ignored. Fast cooling from austenitizing temperature has a tendency to produce finer carbide distribution, and in such a case longer time is necessary to reach the critical size for the occurrence of ductility trough according to the relationship given in Equation 2. In other words, the rupture time for the ductility trough is very short when the initial carbide size is large enough, which indicates the situation in highly tempered low alloy creep resistant steels.

When the creep strength - rupture time data of the heat resisting low alloy steels are plotted on semi-logarithmic scale by designating the stress in linear and rupture time in logarithmic scales, a distinct inflection point behaviour can be realised indicating the change in slope of stress versus rupture time curves with increasing time. The graphical representation of creep strength- rupture time data are given in Figure 5 at 773, 823 and 873 K for the Steels C, L and V of the study. In order to emphasise the inflection behaviour, the values of slope were plotted against rupture time, and the location of the inflection point was determined in a quantitative manner. This is more evident for the steels having less stability at temperatures below 823 K. The creep strength rupture ductility characteristics of steels C, L and V at 773 K are given in Figure 5a. The hardest steel, which is assumed to be the least stable exhibits quicker degradation of creep strength after the exposure of 3000 hours. This can be seen more clearly in the bottom graph of the figure, which indicates that the change of the slope for Steel-C oc-



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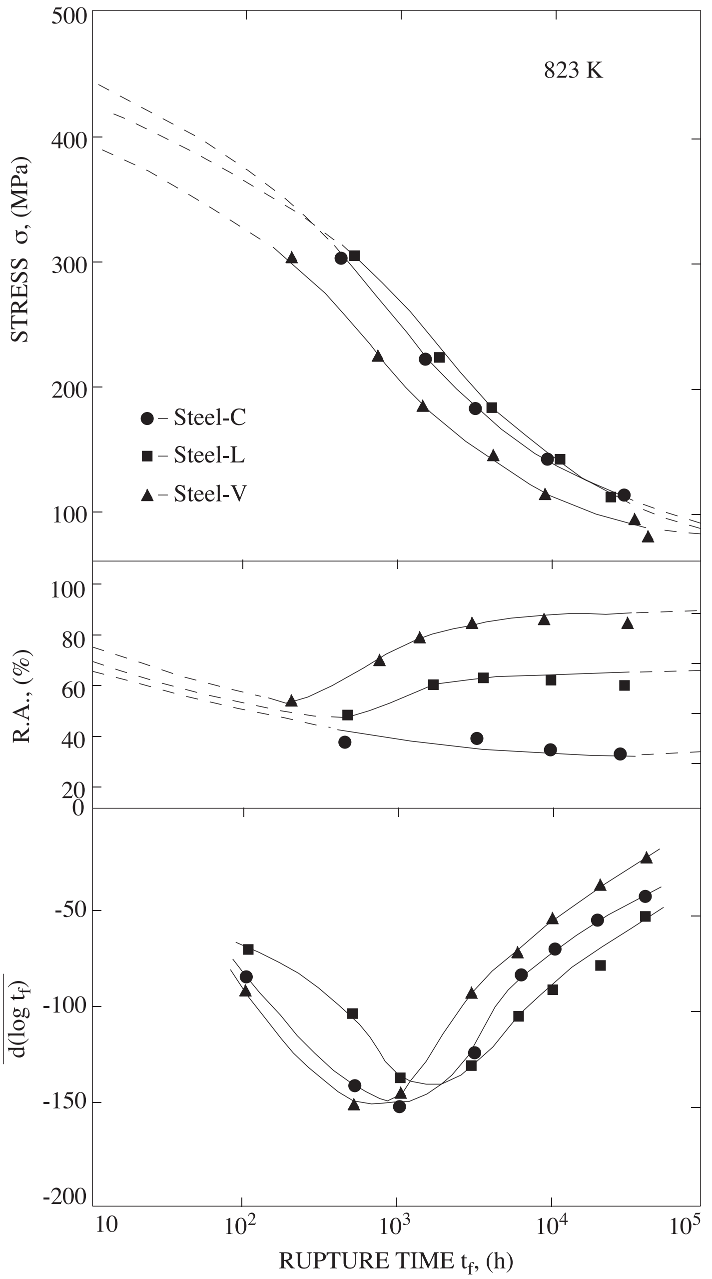
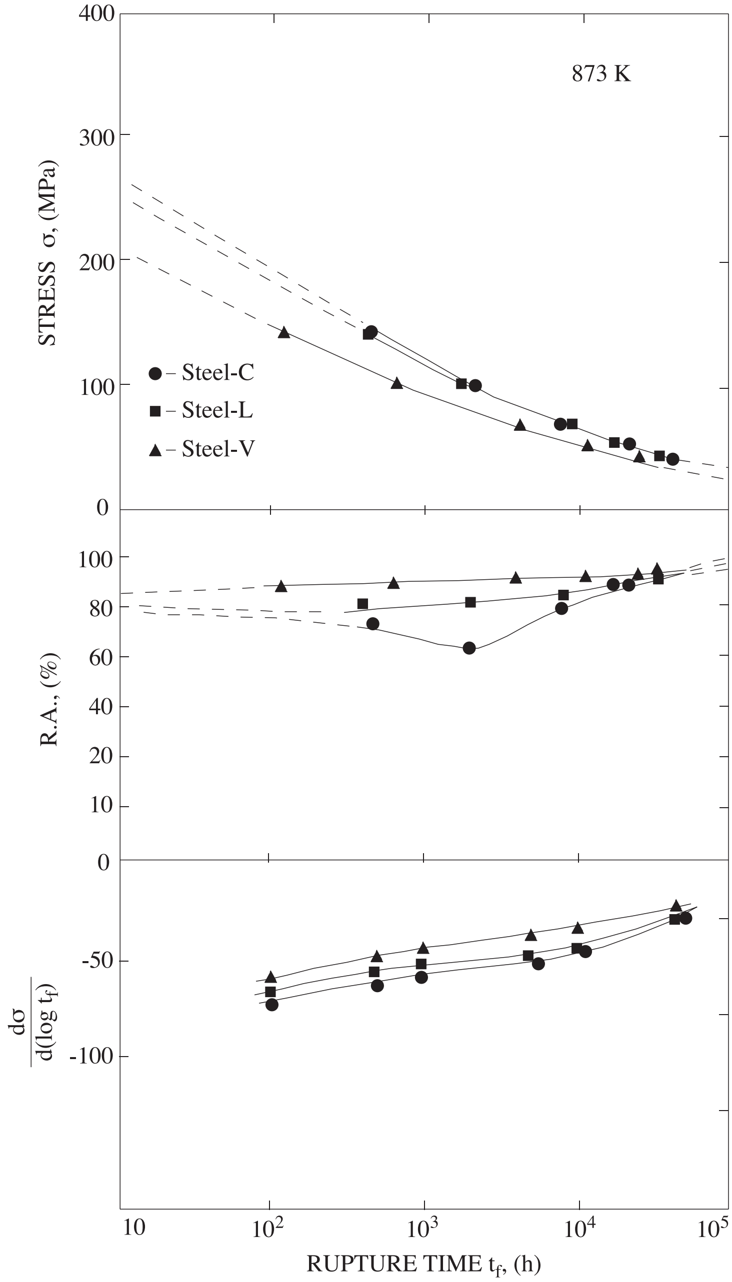


(5b)

curred earlier and steeper than that of other steels. The minimum value of slope coincides with the rupture time for ductility trough, which can be claimed as a consequence of the precipitation and growth behaviour of alloy carbides. The decrease in the creep strength is explained by the loss of precipitation strengthening, which in a decrease in the slope of the curve, and the increase in slope was attributed the the fully tempered structure of the steels in which the creep strength approaches to inherent strength of the matrix (Kimura et al., 1993, Bolton et al 1980). The strength-ductility properties at 823 K and 873 K are given in Figure 5b and 5c respectively, for the same steels. First of all, the degradation of creep properties commences earlier because of higher temperature exposure; secondly the rupture time showing minimum slope again coincides with the range of rupture time domain of ductility trough. The degradation of creep properties occurs very rapidly, probably during the first 50 hours of exposure, since the thermal activation is very high at 873 K, and this may be a reason for the absence of minimum of slope or inflection point after 100 hours of creep exposure, as seen in Figure 5c.



(5c)

Figure 5. The variation of stress, reduction of area and slope of rupture strength curve with rupture time at 773, 823, and 873 K, for 1.25Cr-0.5Mo low alloy ferritic steels.

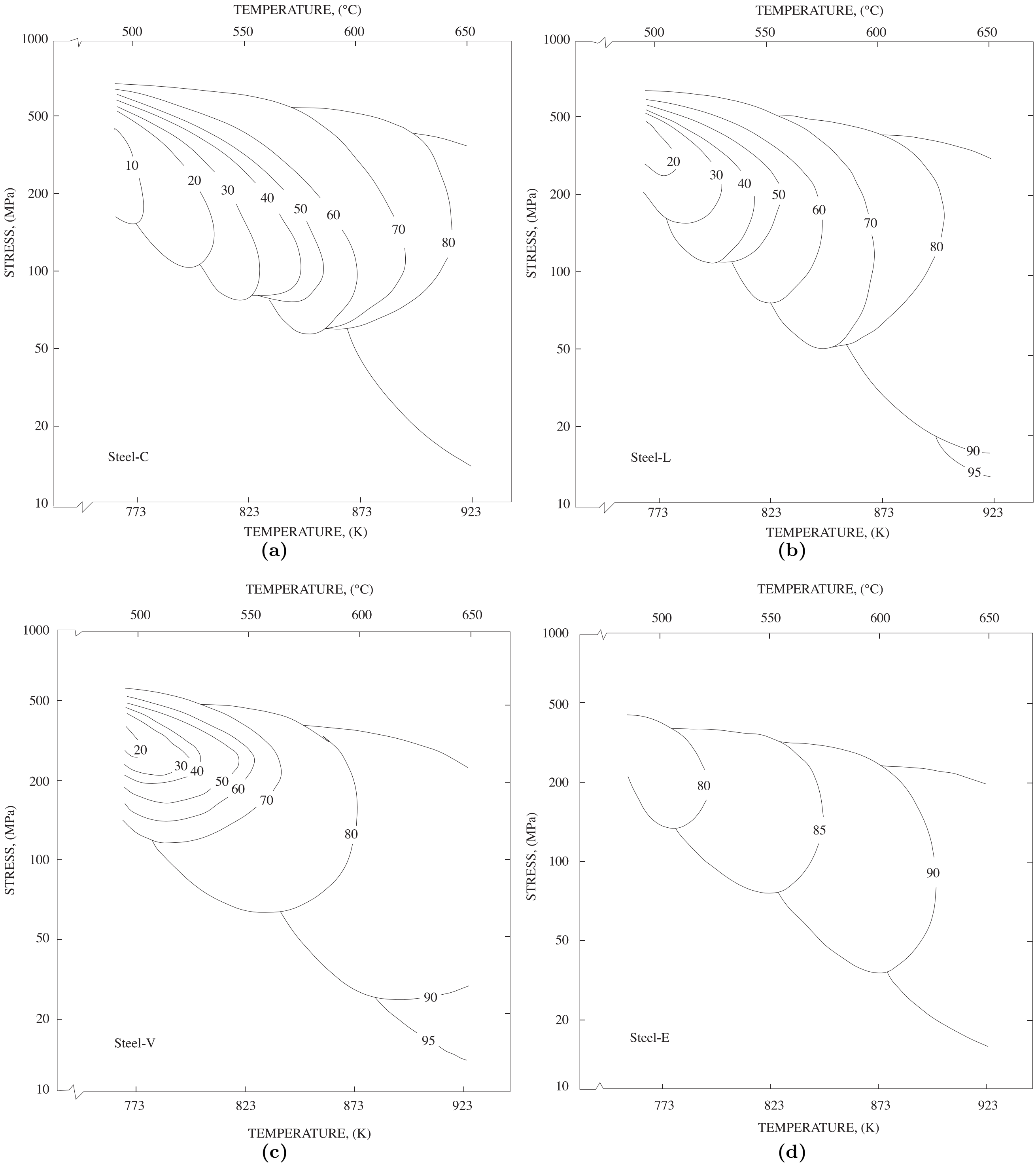


Figure 6. Rupture ductility maps of 1.25Cr-0.5Mo low alloy ferritic steels. (Numbers on equi-ductility contours indicate the rupture of ductility at rupture, in terms of percent reduction of area)

Finally, some attempts have been made to present the creep strength-rupture ductility data as a function of temperature in a more convenient style, which has some similarities with the work published previously (Ashby et al., 1979). Therefore, the creep data of 1.25Cr-0.5Mo steel can be given in the form of a *Rupture Ductility Map* specific to each steel covered by the study in the temperature range of 773-923 K. The rupture ductility maps of the steels are given in Figure 6, in which the applied stress and exposure temperature are represented by vertical and horizontal scales, respectively. Reduction of area at rupture data at different temperatures was located and the points having equal ductility are enveloped by iso-ductility contours producing distinct regions of the same rupture ductility for each steel. The upper bounds of iso-ductility contours are determined by considering the shortest creep-rupture test data of the steels, which were extracted from tensile test results at the temperatures of creep exposure, and the lower bound is based on the extrapolated creep strength and rupture ductility for 105 hours of creep exposure, which can be regarded as reasonably long for most engineering applications. Any point in each region has a ductility lower than the value of the right limiting contour and higher than the value of left limiting contour. Therefore, it can be possible to predict the rupture ductility of a 1.25Cr-0.5Mo steel, depending on stress and temperature with an uncertainty of maximum 10 percent by using the rupture ductility map of each steel. When the maps are compared with each other, it may easily be recognised that there is a strong dependency of iso-ductility contours on the initial microstructure or thermomechanical history of steel. For the steels with higher hardness, which is attributed to the fine distribution of metastable alloy carbides in the structure, low ductility regions are taking place in the temperature range of 750-850 K, as seen in Figure-6a. On the other hand, for the steels with more stable carbide distribution, low ductility regions are shifted to left of the map, and for the softest steel they are completely disappeared, as seen in Figure 6b, Figure 6c and Figure 6d respectively.

Conclusions

When the above results and discussion are taken into consideration, the study has primarily yielded the following conclusions:

1. The ductility troughs, which have been observed principally in creep deformation of low alloy, creep resistant ferritic steels are depending on primarily on the nucleation and growth behaviour of metastable alloy carbides.

2. The rupture time for the formation of ductility trough can be predicted by a logarithmic relationship developed in the study, as a function of temperature for 1.25Cr-0.5Mo steels with different initial hardness.

3. 1.25Cr-0.5Mo heat-resisting low alloy steels exhibit distinct inflection characteristics in their creep strength-rupture time curves, which are coinciding well with the rupture time values creating rupture ductility troughs.

4. Rupture ductility maps can be constructed and used for 1.25Cr-0.5Mo steels in order to predict the rupture ductility at a given state of stress and temperature with reasonable uncertainty.

Acknowledgement

The author would like to thank to British Council for financial support of the study, and Dr. Jonathan Parker for his valuable discussions on the subject.

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