JOURNAL OF MATERIALS SCIENCE **34** (1999) 2513–2517

Stress dependence on stress relaxation creep rate during tensile holding under creep-fatigue interaction in 1Cr-Mo-V steel

C. Y. JEONG, S. W. NAM*
Department of Materials Science and Engineering, Korea Advanced Institute of Science and Technology, 373-1 Kusong-dong, Yusong-gu, Taejon 305-701, Korea

J. GINSZTLER
Department of Electrical Engineering Materials, Technical University of Budapest, H-1111 Budapest Goldmann György, Tér 3., Hungary

A quantitative analysis of the stress dependence on stress relaxation creep rate during hold time under creep-fatigue interaction conditions has been conducted for 1Cr-Mo-V steel. It was shown that the transient behavior of the Norton power law relation is observed in the early stage of stress relaxation in which the instantaneous stress is relaxed drastically, which occurs due to the initial loading condition. But after the initial transient response in a 5 hour tensile hold time, the relations between strain rate and instantaneous stress represented the same creep behavior, which is independent of the initial strain level. The value of stress exponent after transition was 17 which is the same as that of the typical monotonic creep suggested from several studies for 1Cr-Mo-V steel. Considering the value of the activation energy for the saturated relaxation stage, it is suggested that the creep rate is related to instantaneous stress and temperature by the Arrhenius type power law.

1. Introduction

High-temperature low cycle fatigue (HTLCF) is of considerable interest in the design of many components used in power generation and turbine rotors, which are subjected to complex cyclic loading conditions. Therefore, HTLCF experiments with hold time can be very meaningful tests for understanding creep-fatigue interaction phenomenon.

In HTLCF, it has been shown that the fatigue life is decreased as the hold time which results in the creep effect of stress relaxation is increased at a given temperature [1–6], but the exact reason for the reduction of the fatigue life due to the creep damage formation during stress relaxation has not been identified yet.

Under the test condition of total strain controlled LCF with hold, the total strain, made up of the elastic (ε_e) and plastic (ε_p) strain components, remains constant during hold time. Hence the sum of the each strain rate is given by,

$$\dot{\varepsilon}_e + \dot{\varepsilon}_p = 0 \quad (1)$$

where $\dot{\varepsilon}_e$ and $\dot{\varepsilon}_p$ represent the elastic and plastic strain rates, respectively. Thus, the plastic strain rate can be expressed in terms of elastic modulus (E) and a stress relaxation rate by,

$$\dot{\varepsilon}_p = -\frac{1}{E}\frac{d\sigma_r}{dt} \quad (2)$$

where σ_r and t are the instantaneous relaxation stress and time, respectively and the value of $d\sigma_r$ is negative during stress relaxation. Thus, the plastic strain rate is determined from the rate of change in stress during stress relaxation, and therefore its value changes throughout the hold time.

Recently, the present investigators [7] have studied the activation processes for stress relaxation under creep-fatigue interaction conditions in 1Cr-Mo-V steel. It has been found that the apparent activation energy for stress relaxation at the saturated stage, in which the stress is hardly relaxed, is the same as that for the lattice diffusion activation energy of iron i.e., 251 kJ/mole [8] independent of the total strain range. Also, analyzing the value of the activation volume for the initial transient relaxation behavior in which the stress is relaxed drastically, it has been suggested that the rate controlling the dislocation mechanism is either cross slip [9], or overcoming Peierls-Nabarro stress [10, 11]. Thus, the temperature dependence of creep rate was identified during stress relaxation. It was shown that the creep mechanism is identical with that of the steady state in monotonic creep [12–15] after a long enough hold time, that is the dislocation climb which is controlled by self diffusion.

For most of the solid materials, it has been shown that the steady state creep rate $\dot{\varepsilon}$ is related to the applied stress and temperature by [16],

* Jointly Appointed at the Center for the Advanced Aerospace Materials.

$$\dot{\varepsilon} = A\sigma^n \exp\left(-\frac{Q_{app}}{RT}\right) \quad (3)$$

where Q_{app} is the apparent activation energy for creep, T is the absolute temperature, and A and n are the structure factor and the creep exponent, respectively. So if dislocation creep is considered, the strong dependence of creep rate on the applied stress is observed and is very important from an engineering point of view. The main difference is that the total strain is constant in tensile hold test and the stress is changed with time, whereas the stress is constant in monotonic creep. Therefore, it may be possible to interpret the stress dependence of the rate change throughout the hold time.

In this study, the creep behaviors of stress relaxation in 1Cr-Mo-V steel during hold time cycled at high temperature have been analyzed and the stress dependence on the creep rate for relaxation behavior has been compared with that of monotonic creep. Finally, a constitutive equation for creep deformation of stress relaxation in 1Cr-Mo-V steel is suggested in Arrhenius type power law relation.

2. Experimental

The chemical composition and the heat treatments of 1Cr-Mo-V steel used in this investigation are shown in Table I. The heat treated microstructure was observed to be upper bainite and the prior austenite grain size was measured to be approximately 110 μm as shown in Fig. 1.

The specimen was of a cylindrical type with a diameter of 7 mm and a gauge length of 8 mm. Total strain-controlled uniaxial fatigue tests with hold time were carried out in an argon atmosphere (99.99%) by a dynamic Instron machine model 1362. Stress relaxation behaviors of 10 min and 5 hour tensile hold were analyzed over a temperature range from 813 to 833 K under the total strain range of $\pm 1.7\%$ and $\pm 2.5\%$. All the test results were obtained from a half fatigue life in which the stable peak stress is maintained. And all the data were recorded by a computer data acquisition system. The final transmission electron microscopy (TEM) foils were prepared by the twinjet electropolishing with a solution of 95 vol % acetic acid +5 vol % perchloric acid at 50 V and 5 °C. TEM observation was conducted in Philips CM-20 microscope operating at 200 kV.

3. Results and discussion

3.1. Stress dependence on creep rate during stress relaxation

Typical relaxation stress versus hold time resulting from 10 min tensile hold with initial total strain range of $\pm 1.7\%$ and $\pm 2.5\%$ are shown in Fig. 2. The stress relaxation curves at the half of the fatigue lives are chosen to obtain the stabilized fatigue properties in various test temperatures. In all the tests, the curves are found to be similar with those of monotonic normal primary creep in which the creep rate is decreased with the creep strain, and this stress versus time can be converted to strain rate versus stress i.e., the stress relaxation rate

TABLE I The chemical composition and the heat treatment steps of 1Cr-Mo-V steel (all in wt %)

C	Si	Mn	P	S	Ni	Cr	Mo	V	Ti	Co	W	Nb
0.272	0.278	0.710	0.016	0.007	0.534	1.189	1.263	0.265	0.004	0.015	0.012	0.012

Heat treatment
Austenitizing: 1241 K/26 h → forced air cooling
Tempering: 948 K/37 h → 643 K/forced air cooling → air cooling

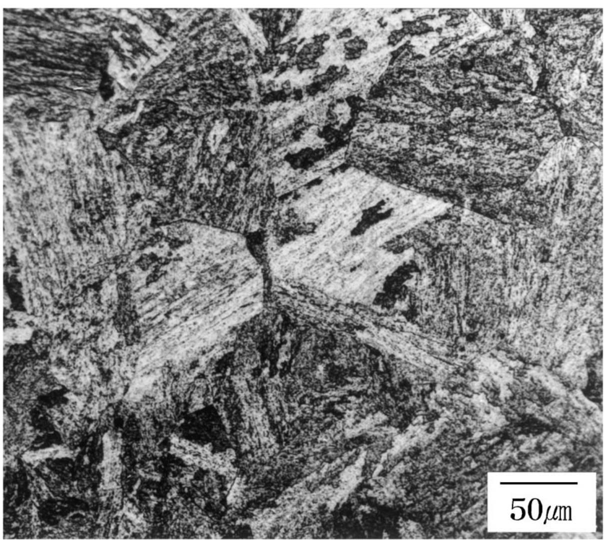

Figure 1 Optical microstructure of 1Cr-Mo-V steel showing bainite structure.

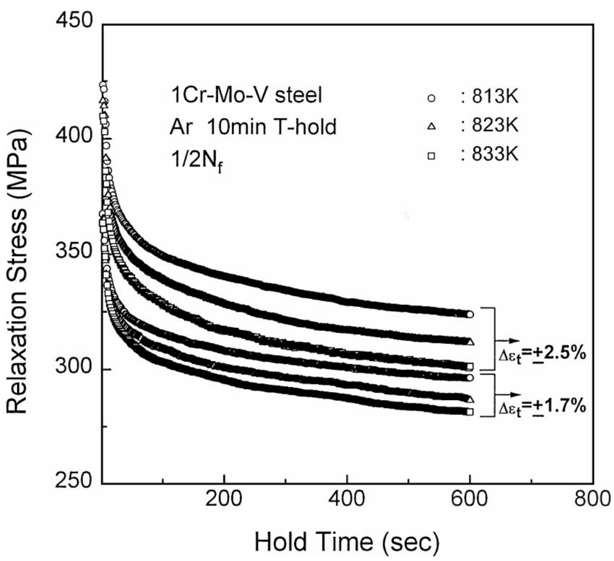

Figure 2 Stress relaxation curves at three test temperatures in a ± 1.7 and $\pm 2.5\%$ total strain range.

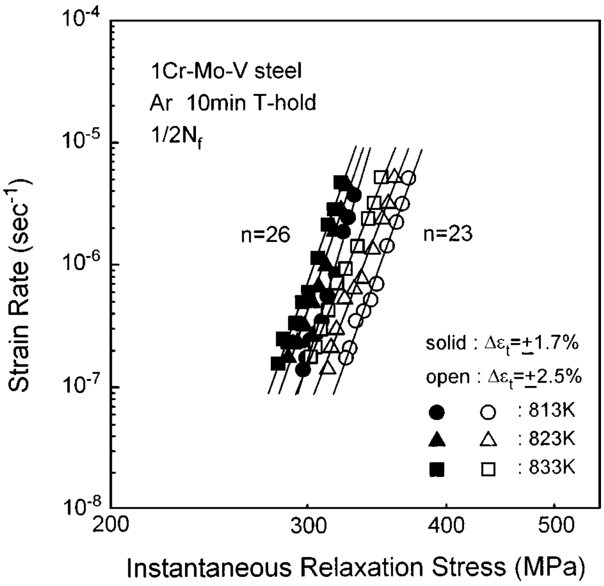

Figure 3 Stress dependence of creep rate for stress relaxation at three test temperatures in a ±1.7 and ±2.5% total strain range, and 10 min tensile hold.

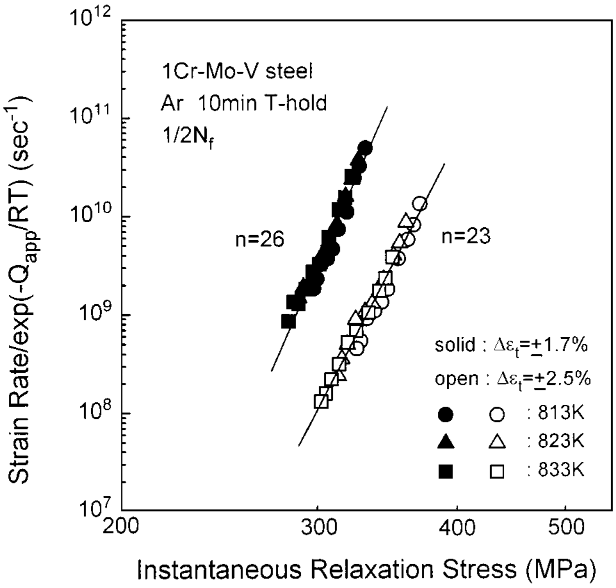

Figure 4 Stress dependence of temperature-compensated creep rate for stress relaxation at three test temperatures in a ±1.7 and ±2.5% total strain range.

$d\sigma/dt$ is converted to the creep strain rate $\dot{\varepsilon}_p$ as mentioned in the previous introduction.

Fig. 3 shows the stress dependence of creep rate calculated from Equation 2 for the two different initial strain levels. First, it is observed that the creep rate is increased with an increasing temperature at a given instantaneous relaxation stress and this trend is a general behavior in that the creep rate is accelerated by the thermal activation process. Also the values of stress exponent n are the same as 23–26 regardless of the test temperature. But the non-unique behavior is observed in the early stage of the curves and it has been related to the recoverable anelastic strain. Similar results were obtained for aluminium [17] and titanium alloys [18], where it was shown that the transient behavior is due to the accumulation of anelastic strain resulting from the initial loading. This characteristic of the anelastic effect will be discussed later.

It was already found that the activation energy for stress relaxation of 1Cr-Mo-V steel increases with the increasing relaxed stress, and is saturated to the value of 251 kJ/mole which is same as the activation energy for lattice diffusion [7]. So using the concept of temperature term in Equation 3 temperature-compensated creep rate is analyzed. The stress dependence of a temperature-compensated creep rate is shown in Fig. 4. It can be seen that the data at 813–833 K are in good agreement for a given total strain range, respectively. As mentioned above, the transient behavior is observed in a relatively short time such as a 10 min hold time, and the accumulation of anelastic strain is dominant in the early stage of stress relaxation.

Fig. 5 shows the plots of 5 hour hold data, as well as 10 min hold data. An important point shown in Fig. 5 is that the results are self-consistent and repeatable as the transient effects are eliminated. After the initial transient, the relations represent the same creep behavior, which is independent of the initial stress or strain level, and similar results have been reported by other researchers [17–19].

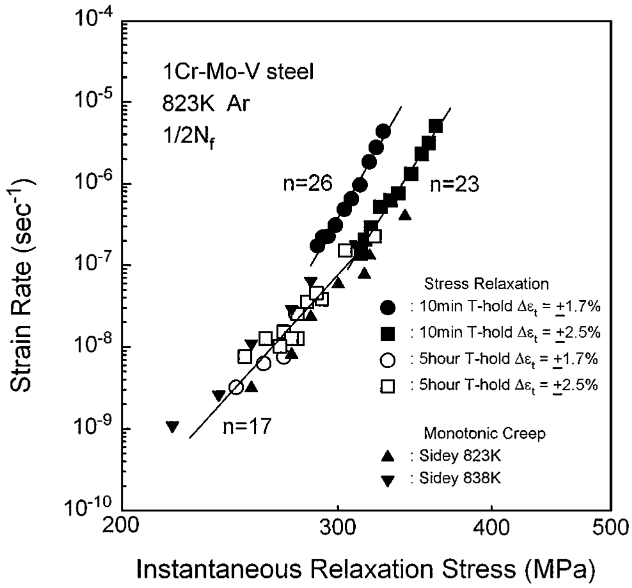

Figure 5 Stress dependence of creep rate for stress relaxation at 823 K in a ±1.7 and ±2.5% total strain range, and 10 min and 5 hour tensile hold.

In this study, the creep behavior determined from the stress relaxation is compared to the monotonic constant-stress creep test using existing data from the literature. In Fig. 5 the results from the present study are compared with the general creep behavior from Sidey [12]. The data from the monotonic creep at 823–838 K are in good agreement with the data from the stress relaxation, indicating that the stress exponent is found to be 17. This exponent value is the same as the values suggested from the conventional creep of several studies [12–15] in the high-stress region for 1Cr-Mo-V steel. High values of the stress exponent are commonly obtained with precipitation and dispersion-hardened alloy [12, 16], even though recovery processes are known to play an important role in creep of such alloys. Therefore, it is known that the creep mechanisms of stress relaxation which are the main reasons for fatigue life reduction under creep-fatigue interactions after a long enough time are the same as that of long-term monotonic creep. It has a benefit in that a large amount of creep information can be obtained from the short term of a relaxation test.

From Fig. 5, the stress dependence of the creep rate after transient behavior can be represented by,

$$\dot{\varepsilon} = A_0\sigma^{17} \qquad (4)$$

where A_0 is a temperature dependent constant. It was already known that the apparent activation energy for stress relaxation at the saturated stage is 251 kJ/mole, thus the creep rate can be related to instantaneous stress and temperature from Equation 3 by,

$$\dot{\varepsilon} = A\sigma^{17} \exp\left(-\frac{251,000}{RT}\right) \qquad (5)$$

where A is a structure factor which is regarded as a material constant.

3.2. Effects of microstructural change on creep behavior

1Cr-Mo-V steel which is a general precipitation hardened alloy by carbides formation. It is understood that the particles introduce a kind of creep resistance opposing a dislocation motion which is an additive to the normal material resistance. In this study, microstructural investigation of thin foils was carried out by TEM. The initial microstructure is upper bainite as shown in Fig. 6a. It consisted of many carbides in both the matrix and grain boundaries. After creep-fatigue deformation under 10 min tensile hold at ±1.7% and ±2.5% total strain range, it was observed that the carbides have coarsened. Their spacing had also increased relative to that at the initial microstructure shown in Fig. 6b and c, respectively. Examination of Fig. 6 also shows that the carbides at ±2.5% condition have coarsened considerably in comparison to that tested at ±1.7%.

Above all, it is shown that the relaxation creep process during hold time affects on the material degradation. The effect of carbides growth reflects on the creep behavior of stress relaxation in 1Cr-Mo-V steel under creep-fatigue interaction conditions. As mentioned in the introduction, the fatigue life is decreased under creep-fatigue interaction conditions. The reason for life reduction in HTLCF with hold is discussed by the creep effect during the stress relaxation. Under creep-fatigue interaction conditions, the size and the distribution of second particles can act as effective obstacles to dislocation motion. If the chemical composition of an alloy is properly chosen, an optimum dispersion of the strengthening phase particles would be created by a thermal treatment involving a solid solution decomposition as its final phase. In this investigation, the dimension and spacing of the carbides are observed to be increased during a creep-fatigue test in 1Cr-Mo-V steel, although the particles are recommended to be sufficiently stable with respect to coalescence. A progressive microstructural degradation is therefore to be expected. During creep-fatigue conditions in precipitation hardened alloys, the fatigue effect of prestrain before hold, as well as the period of hold time, have a role of creep-fatigue strength loss as shown in Fig. 6. According to the previous works [14, 20], the loss of creep strength during creep is caused by aging. Whereas the stress assisted growth of carbides is another dominant factor in the creep-fatigue situation. Therefore, in the life extension problem of LCF with hold, the consideration of hold time which is a typical simulation method of operation time in the power plants, may not be effective for a given material. But the efforts on the prestrain before and after the hold time would give a solution of deceleration to the material degradation. The effects of unloading and reloading would be main causes of the particle growth and the life reduction, which may be due to the acceleration of the crack growth rate after the hold time. As shown in Fig. 6, the growth of carbides is increased when imposing the strain deformation and increasing the initial total strain range. Thus, the long

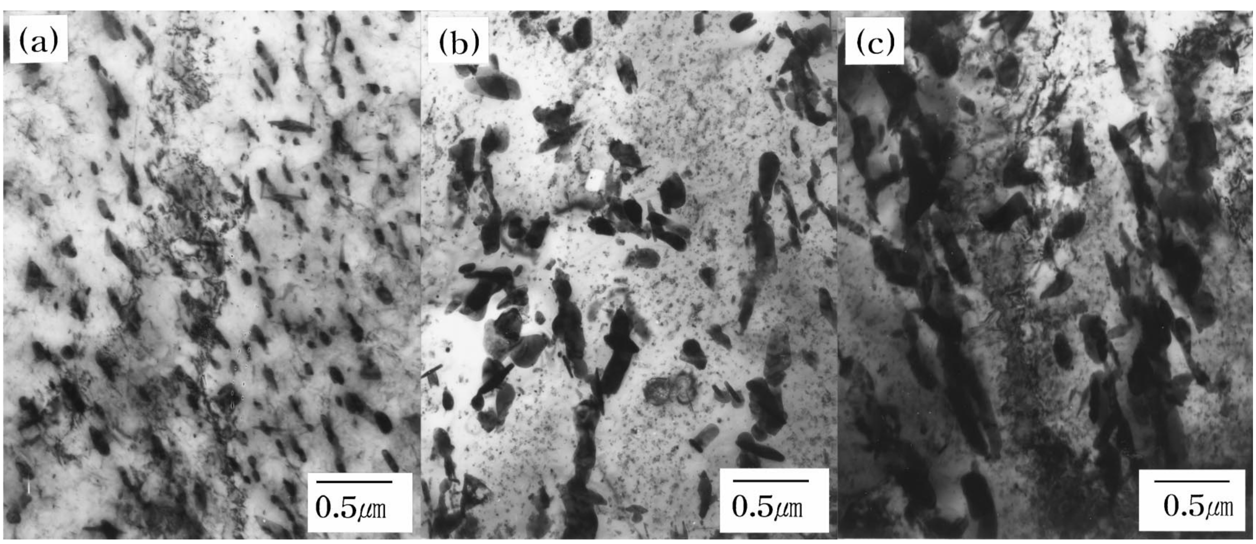

Figure 6 TEM photographs showing the carbides size and distribution. (a) Before creep-fatigue test; (b) $\Delta\varepsilon_t = \pm 1.7\%$, 10 min tensile hold and (c) $\Delta\varepsilon_t = \pm 2.5\%$, 10 min tensile hold.

time operation of a power plant system and the control of the heating rate which satisfies a recommendable efficiency can give an extension of remnant life for this given precipitation hardened alloy.

4. Conclusions

1. In the dependence of creep rate for stress relaxation on instantaneous stress, the transient phenomenon is observed in the early stage of relaxation depending on the initial loading conditions, but the consistent Norton power relation is shown after the initial transient stage independent of the applied strain level.

2. The creep behavior analyzed from the stress relaxation is in good agreement with that from the typical monotonic creep, showing the stress exponent values of 17.

3. The creep rate of stress relaxation at the saturated stage is related to instantaneous stress and temperature by, $\dot{\varepsilon} = A\sigma^{17} \exp(-\frac{251,000}{RT})$.

Acknowledgements

This work was sponsored by the EESRI (Electrical Engineering and Science Research Institute, Project, No. 97-016) and MOST (Korean-Hungarian Joint Project, No. I-1-055). Thanks are given for their financial support.

References

1. J. W. HONG, S. W. NAM and K.-T. RIE, *J. Mater. Sci.* **20** (1985) 3763.
2. S. W. NAM, J. W. HONG and K.-T. RIE, *Metall. Trans. A* **19A** (1988) 121.
3. S. W. NAM, Y. J. OH and B. O. KONG, in Proceedings of 3rd Conference on Low Cycle Fatigue and Elasto-Plastic Behaviour of Materials, Berlin, September 1992, p. 216.
4. J. H. RYU and S. W. NAM, *Int'l. J. Fatigue* **11** (1989) 433.
5. S. W. NAM, Y. C. YOON, B. G. CHOI, J. M. LEE and J. W. HONG, *Metall. Mater. Trans. A* **27A** (1996) 1273.
6. C. Y. JEONG and S. W. NAM, in Proceedings of 13th International Forgemaster Meeting, Pusan, October 1997, p. 173.
7. C. Y. JEONG, S. W. NAM and J. GINSZTLER, "Activation Processes of Stress Relaxation during Hold Time in 1Cr-Mo-V Steel." *Mater. Sci. Eng. A* (1998), accepted.
8. F. S. BUFFINGTON, K. HIRANO and M. COHEN, *Acta Metall.* **9** (1961) 434.
9. N. BROWN and R. A. EKVALL, *Acta Metall.* **10** (1962) 1101.
10. J. HESLOP and N. J. PETCH, *Phil. Mag.* **1** (1956) 866.
11. H. CONRAD and S. FREDERICK, *Acta Metall.* **10** (1962) 1013.
12. D. SIDEY, *Metall. Trans. A* **2** (1976) 1785.
13. J. MYERS, G. WILLOUGHBY and R. H. HAM, *Met. Sci. J.* **2** (1968) 210.
14. D. A. MILLER, W. J. PLUMBRIDGE and R. A. BARTLETT, *Met. Sci.* **15** (1981) 413.
15. D. A. MILLER, *Mater. Sci. Eng.* **54** (1982) 169.
16. O. D. SHERBY and P. M. BURKE, *Prog. Mater. Sci.* **13** (1968) 325.
17. E. W. HART and H. D. SOLOMON, *Acta Metall.* **21** (1973) 295.
18. A. H. ROSENBERGER, A. MADSEN and H. GHONEM, *J. Mater. Sci. Perf.* **4** (1995) 182.
19. D. A. WOODFORD, D. R. VAN STEELE, K. AMBERGE and D. STILES, in "Creep Strength Evaluation for IN 738 Based on Stress Relaxation," edited by S. D. Antolovich (Supralloys 1992, TMS, 1992) p. 657.
20. K. R. WILLIAMS and B. WILSHIRE, *Mater. Sci. Eng.* **28** (1977) 289.

Received 23 September 1998
and accepted 13 January 1999