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THE RESPONSE OF FERRITIC STEELS TO NONSTEADY LOADING AT ELEVATED TEMPERATURES*

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ABSTRACT

In order to meet industrial requirements for heavy wall process vessels that operate at temperatures in the creep range, the strength level of low-alloy steels is being upgraded by means of modified heat treatments, restricted chemistry specifications, and micro-alloying. Materials with room-temperature ultimate-strength levels in the range 586 to 758 MPa (85—110 ksi) have been produced. Typically, these alloys have been developed and evaluated on the basis of the properties required for incorporation into Section I, VIII, and IX of the *ASME Boiler and Pressure Vessel Code*. High-temperature operating experience is lacking in vessel materials that have strength levels above 586 MPa. Because of their tendency toward strain softening, we have been concerned about their behavior under nonsteady loading. Testing was undertaken to explore the extent of softening produced by monotonic and cyclic strains.

The specific materials included bainitic 2 1/4 Cr-1 Mo steel, a micro-alloyed version of 2 1/4 Cr-1 Mo steel containing vanadium, titanium, and boron, and a martensitic 9 Cr-1 Mo-V-Nb steel. Tests included tensile,

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creep, variable stress creep, relaxation, strain cycling, stress cycling, and non-isothermal creep ratchetting experiments. We found that these steels had very low uniform elongation and exhibited small strains to the onset of tertiary creep compared to annealed 2 1/4 Cr-1 Mo steel. Repeated relaxation test data also indicated a limited capacity for strain hardening. Reversal strains produced softening. The degree of softening increased with increased initial strength level. We concluded that the high strength bainitic and martensitic steels should perform well when used under conditions where severe cyclic operation does not occur.

INTRODUCTION

Experience with 2 1/4 Cr-1 Mo steel in high-temperature pressure vessel and steam piping has been excellent for the last thirty years. This alloy has remarkable versatility and can be heat-treated to produce a large range of strength and toughness properties.[1] Currently, only two strength classes are approved for construction of pressure vessels under Sect. VIII, Divs. 1 and 2 of the *ASME Boiler and Pressure Vessel Code*. These classes are the annealed material (SA 387 grade 22 class 2 with a 207 MPa minimum yield and 414 MPa minimum ultimate strength) and the normalized-and-tempered material (SA 387 grade 22 class 2 with a 310 MPa minimum yield and 517 MPa minimum ultimate strength). The annealed material has a microstructure that is predominately proeutectoid ferrite and has been favored for service in the creep range because of its superior microstructural stability and better long-time rupture strength[2,3] while the normalized-and-tempered material has a tempered bainite microstructure and has been favored for heavy-wall pressure vessel service at temperatures below 480°C because of its better tensile properties and toughness.[1,4,5] Efforts to improve the cleanliness and long-time stability of annealed 2 1/4 Cr-1 Mo steel for nuclear steam generator applications produced a material with marginal short-time strength,[6] and as a result, there has been considerable interest in substituting a 9 Cr-1 Mo-V-Nb steel in some critical nuclear components.[7]

Meanwhile, engineers in the petroleum and petrochemical industries perceived a need for a higher strength bainitic steel with adequate creep-strength and hydrogen attack resistance for service at 480°C and higher.[8] Candidates include (a) restricted chemistry versions of A 542 class 3,[8] (b) modified versions of 2 1/4 Cr-1 Mo steel with nickel or boron additions for improved hardenability,[9,10] (c) micro-alloyed versions of 2 1/4 Cr-1 Mo steel containing strong carbide formers such as vanadium, titanium, and niobium,[11-13] (d) numerous modifications of 3 Cr-Mo alloys,[12,14,15] and (e) new 9 Cr-Mo alloys.[7,16,17] The metallurgy of these steels is quite complicated. The basic transformation products range from ferrite/pearlite through upper and lower bainite to martensite with various amounts of retained austenite or ferrite. The multiplicity of carbides that develops during tempering, postweld heat treating, and subsequent high-temperature service adds to the complexity of the problem in selecting any new alloy for commercialization and codification.[4] One feature common to all the advanced alloys is a high yield-to-ultimate-strength ratio. Concomitant with this feature is a tendency toward low uniform elongation and strain softening. High-temperature design and operating experience with these classes of material is lacking in the

United States, and we believe that there is a need to explore for the conditions that give rise to softening and to examine the implications of this phenomenon relative to design requirements.

MATERIALS

Data were collected on two heats of 2 1/4 Cr-1 Mo steel, two heats of V-Ti-B modified 2 1/4 Cr-1 Mo steel, and one heat of Nb-V-Ti modified 9 Cr-1 Mo steel. The product forms, compositions, and heat treatments are provided in Table 1. The vacuum-arc-remelted (VAR) heat of 2 1/4 Cr-1 Mo steel (heat 56448) was produced as bar, plate, and forging; however, all our tests were performed on a 25-mm-diam bar product. The heat treatment (annealed plus PWHT) was designed to maximize long-time stability and produced a microstructure consisting of proeutectoid ferrite plus regions of decomposed pearlite and bainite [see Fig. 1(a)]. The tensile properties were near the minimum acceptable values.

The second heat of 2 1/4 Cr-1 Mo steel (heat A6660) was produced by the electric furnace process and rolled to 150 and 300-mm plate. We tested the 150-mm product in two conditions: normalized and tempered (662°C for 6 h) and after a simulated postweld heat treatment (PWHT) 690°C for 28 h. The normalized and tempered condition was near the upper limit of the strength range for SA 387 grade 22 class 2 while the PWHT condition produced average properties. In this paper, we will only report on the properties of the PWHT material. The microstructure of the PWHT material is illustrated in Fig. 1(b).

Table 1. Chemical compositions of alloys

Alloy	Heat	C	Si	Mn	P	S	Ni	Cr	Mo	V	Ti	Nb	B
2 1/4 Cr-1 Mo	56448	0.10	0.26	0.54	0.01	0.006	0.15	2.14	1.01	<0.01	<0.01		
2 1/4 Cr-1 Mo	A6660	0.13	0.05	0.33	0.012	0.02		2.34	0.99				
2 1/4 Cr-1 Mo-V-Ti-B	JSWI	0.12	0.02	0.51	0.007	0.008	0.10	2.28	0.99	0.25	0.21		0.002
2 1/4 Cr-1 Mo-V-Ti-B	JSWII	0.14	0.07	0.49	0.008	0.007	0.16	2.34	1.00	0.26	0.32		0.0024
9 Cr-1 Mo-V-Nb	30383	0.083	0.35	0.46	0.007	0.015	0.09	8.46	1.02	0.20	0.005	0.072	0.001

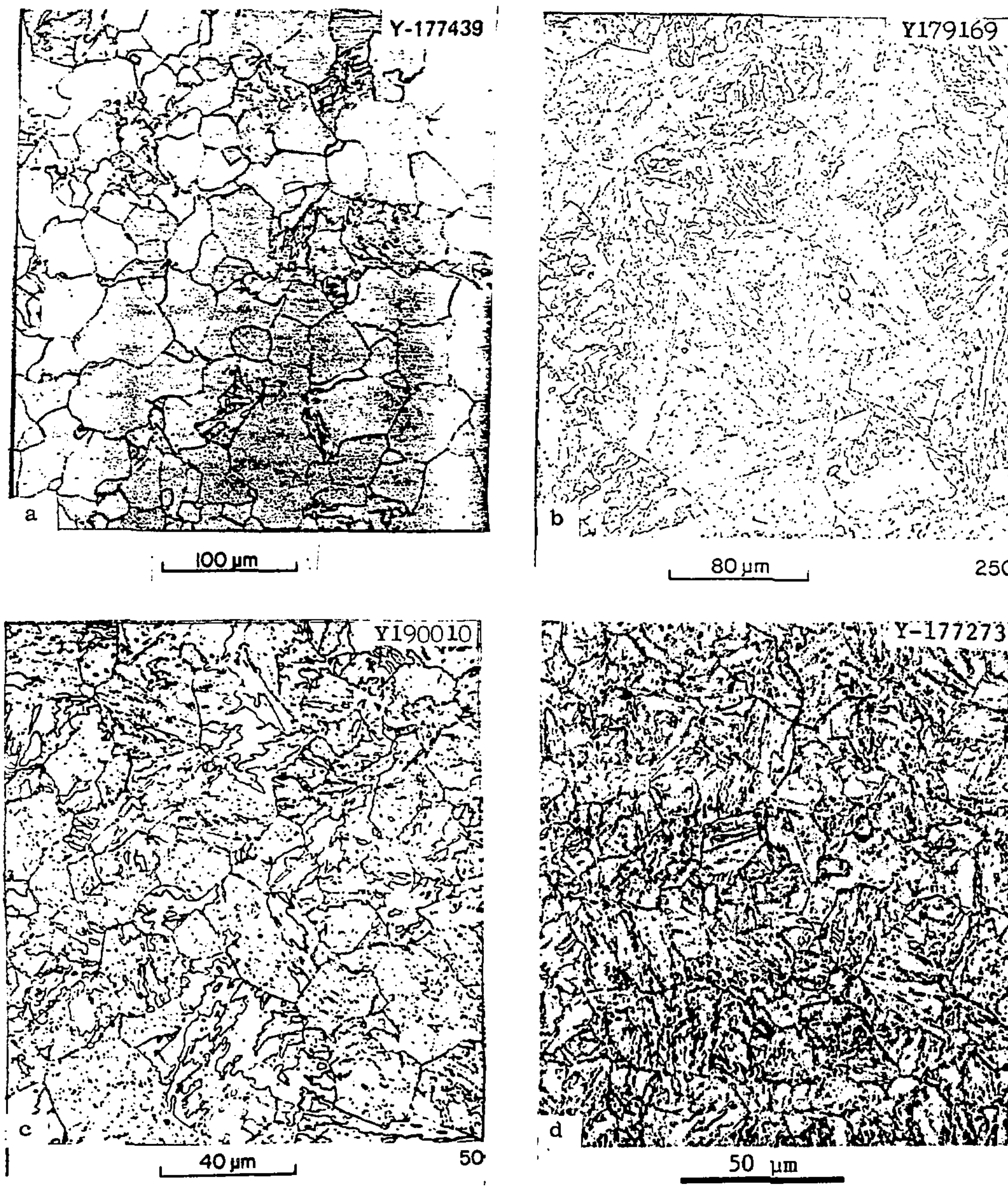

Fig. 1. Microstructures of four steels: (a) annealed 2 1/4 Cr-1 Mo steel heat 56448, (b) normalized and tempered 2 1/4 Cr-1 Mo steel heat A6660, (c) normalized and tempered 2 1/4 Cr-1 Mo-V-Ti-B steel heat JSWI, and (d) normalized and tempered 9 Cr-1 Mo-V-Nb steel heat 30383.

One plate of the V-Ti-B modified 2 1/4 Cr-1 Mo steel (identified here as JSWI) was forged from a 50 kg vacuum induction melt. The 25-mm plate was normalized at 1000°C for 4 h, air cooled, tempered 650°C for 5 h, reannealed at 690°C for 20 h, and air cooled. The resulting microstructure was fully bainitic, as illustrated in Fig. 1(c).

The second heat of V-Ti-B modified 2 1/4 Cr-1 Mo steel (identified as JSWII) produced by the basic oxygen process and forged to four thickness levels ranging from 235 to 525 mm. We studied the properties of a 425-mm-thick section. The material was normalized at 950°C for 7 h, water quenched, tempered at 650°C for 15 h, and subjected to a simulated postweld heat treatment at 690°C for 26 h. The material was fully bainitic at the quarter thickness and had a prior austenite grain size somewhat coarser than the 25-mm plate shown in Fig. 1(c).

The 9 Cr-1 Mo-V-Nb steel was procured in the form of 51-mm plate from a 15-ton heat (heat 30383) produced by the AOD process. The material was normalized at 1040°C, air cooled, tempered 1 h at 760°C, and subjected to a simulated postweld heat treatment for 4 h at 732°C. The microstructure was tempered martensite with a relatively fine prior austenite grain size as illustrated in Fig. 1(d).

TEST METHODS

Specimen blanks were taken from locations representative of the quarter thickness microstructure of the plate products. The tensile test specimen design was a standard threaded-end bar with a 6.3-mm diameter by 32-mm-long uniform test section. Tensile test methods conformed to ASTM recommended practice E21. The creep test specimen design was identical to the tensile specimen and the testing methods conformed to ASTM recommended practice E138. The cyclic test specimen design was a button-end bar type as described in ASTM recommended practice E606. The uniform gage cyclic specimen had a 6.3-mm diameter and 19-mm testing section length.

TENSILE PROPERTIES

Typical tensile curves for the four materials at room temperature are compared in Fig. 2. The martensitic steel (heat 30383) and the bainitic steels (heat A6660 and JSWI) exhibited initially high rates of work hardening and reached the ultimate strength at strains around 6%. The weaker annealed material (heat 56448) exhibited a yield plateau and a much greater strain at the ultimate strength (18%). With increasing temperature, this pattern was qualitatively the same. The work hardening rate just after yielding remained high for the bainitic and martensitic steels but the strain at the ultimate strength decreased to 4% or less. The yield plateau for annealed material (heat 56448) disappeared and the strain at the ultimate strength dropped slightly.

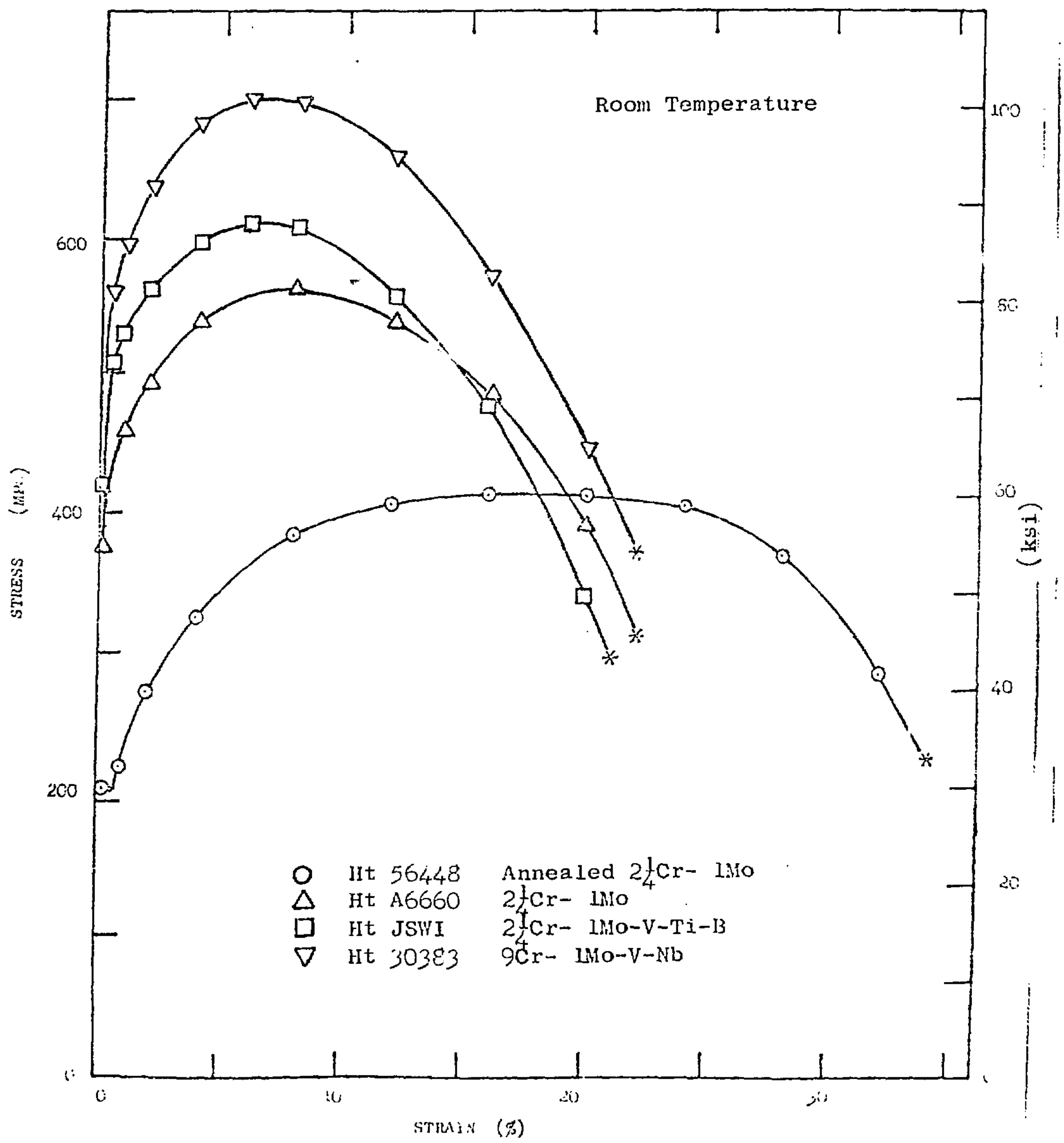

Fig. 2. Room-temperature tensile curves for four steels.

In Figs. 3 through 5, we show typical trends with respect to the influence of temperature on the tensile properties of the four alloys. The tensile strain rate was 6.7×10^{-5}/s for the two 2 1/4 Cr-1 Mo steels and the 9 Cr-1 Mo-V-Nb steel but 2.7×10^{-4}/s for the 2 1/4 Cr-1 Mo-V-Ti-B steel. With respect to strength (Fig. 3), the salient features were: (1) the yield strength of the annealed material (heat 56448) was relatively insensitive to temperature, and (2) all steels rapidly lost strength above 450°C. With respect to the strain at the ultimate strength (Fig. 4), the annealed material exhibited much higher values at all temperatures than the other steels. The bainitic and martensitic steels showed precipitous decreases in the strain at the ultimate strength once the temperature exceeded 450°C. All steels exhibited good elongations and excellent reduction-of-area data as shown in Fig. 5. Values were reasonably constant up to 400°C or so; after which they increased sharply with increasing temperature. Strain rate became an important factor at temperatures above 450°C; and, as might be expected, higher strain rates increased the ultimate strain and strain at the ultimate but decreased elongation.

STRAIN CYCLING RESPONSE

Strain cycling tests were performed on all four materials but not always under conditions that permitted valid comparisons. Nevertheless, it was apparent that the annealed material (heat 56448) exhibited relatively stable cyclic behavior while the bainitic and martensitic steels exhibited stability at low strain ranges and appreciable softening

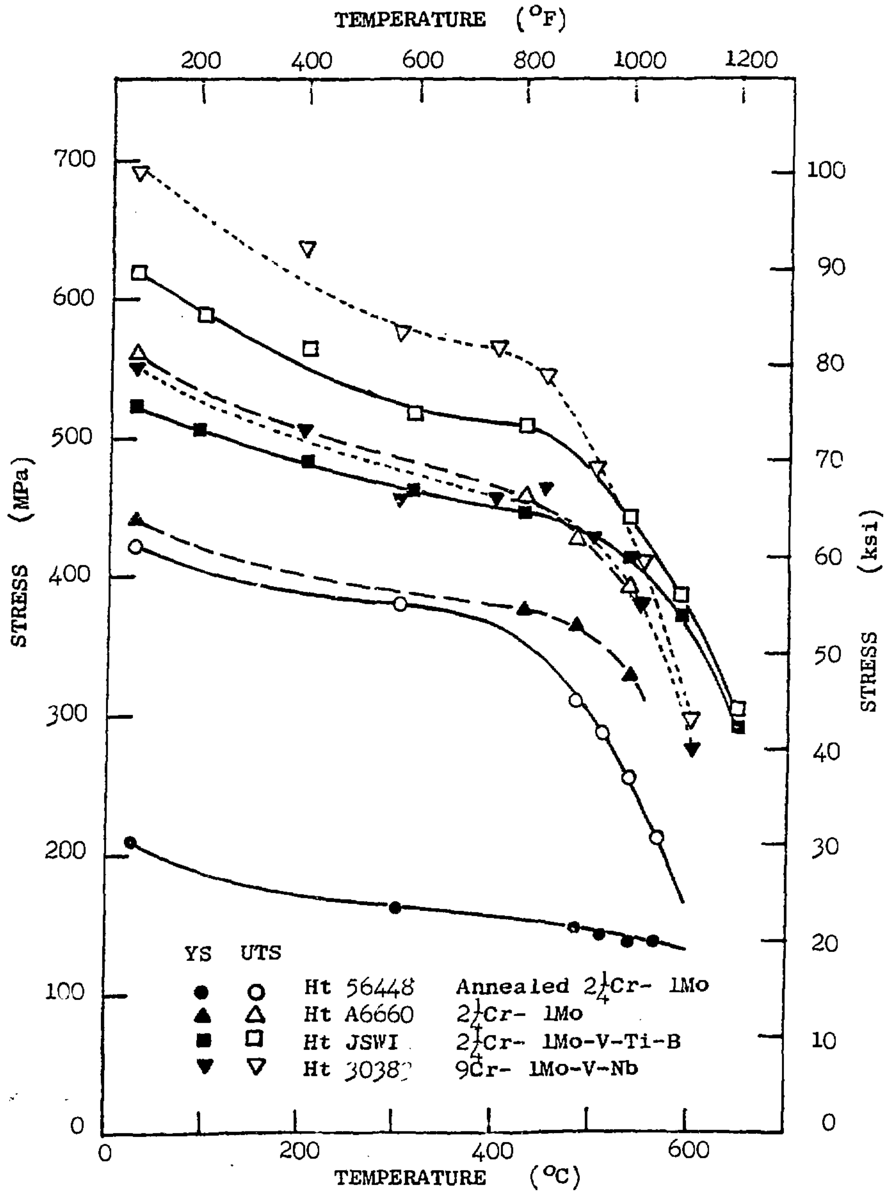

Fig. 3. Yield strength and ultimate strength versus temperature for four steels.

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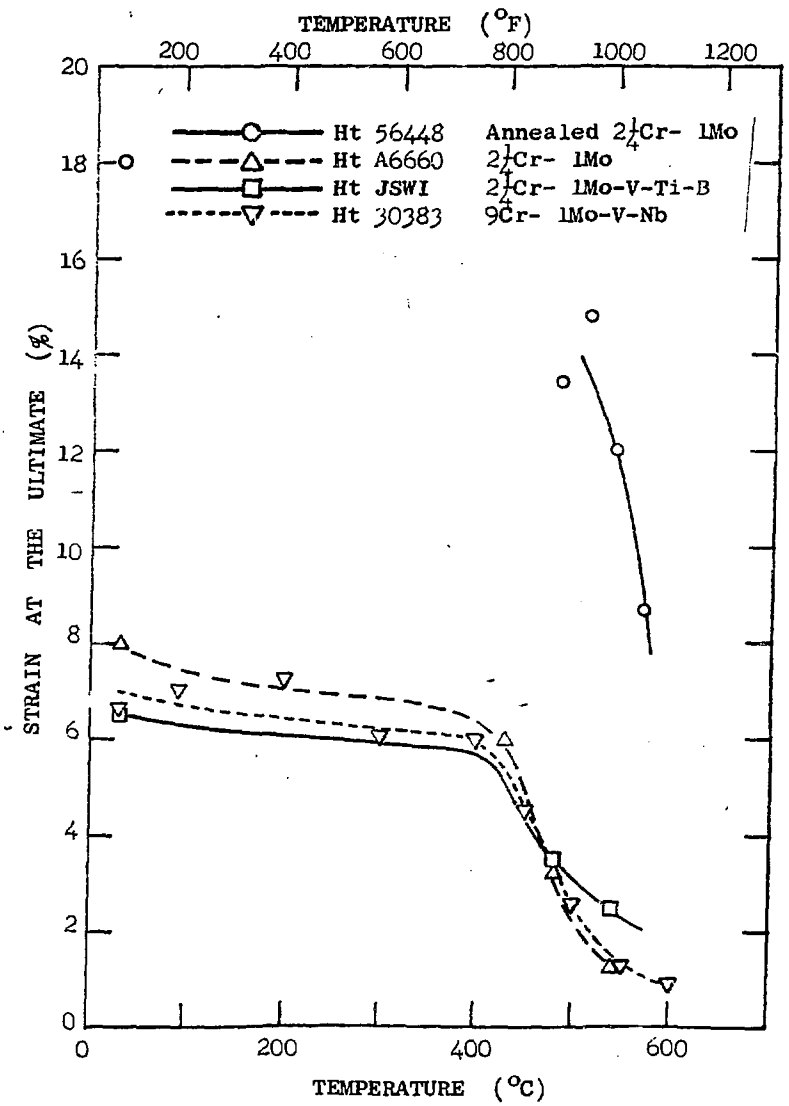


Fig. 4. Strain at the ultimate strength versus temperature for four steels.

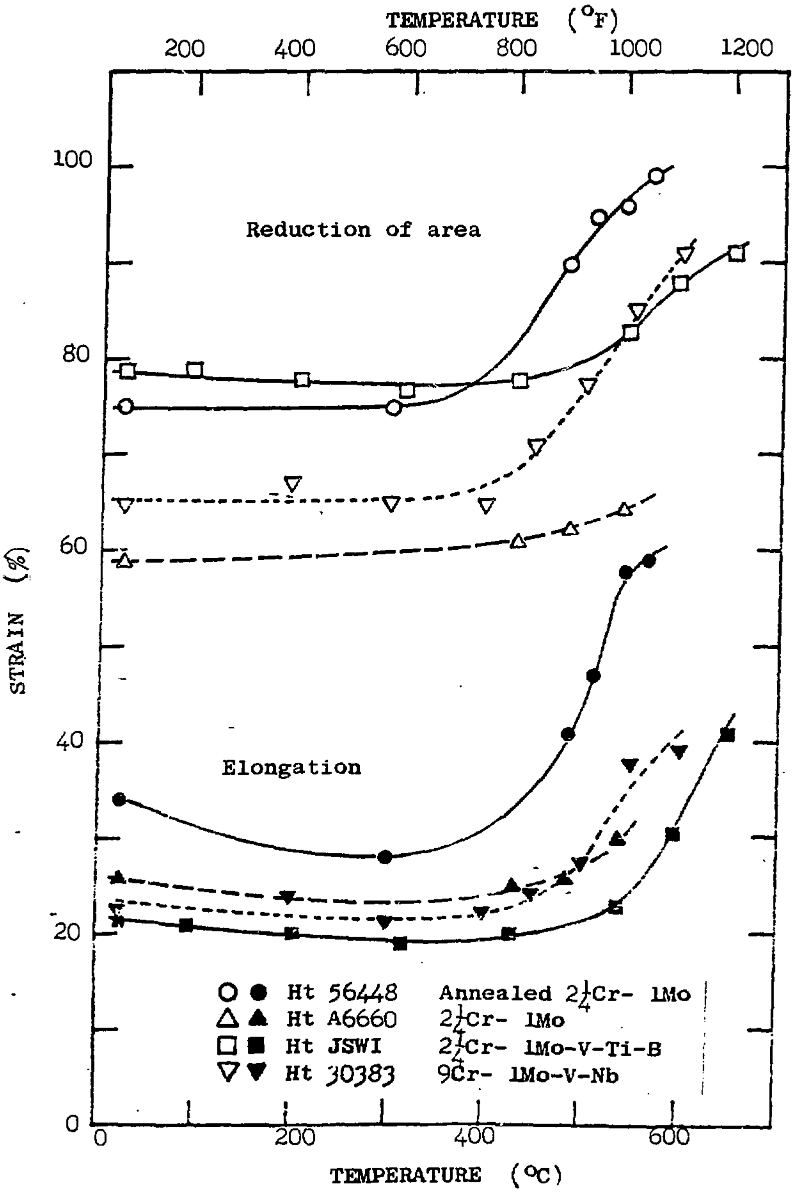


Fig. 5. Elongation and reduction of area versus temperature for four steels.

at high strain ranges. Some trends are illustrated in plot of stress range versus cycles for typical tests at cyclic strain ranges around 0.6% and temperatures in the range 480 to 540°C (Fig. 6).

The high-strength steels exhibited appreciable losses in cyclic strength (up to 30% at half life) but were always stronger than the annealed material. The relative loss in stress range increased with strain range and temperature; and as a result the cyclic hardening curves constructed by plotting stress range, $\Delta\sigma$, against strain range, Δe, tended toward a very low plastic modulus after 1000 cycles or so. These cyclic hardening curves are compared in Fig. 7 where it can be seen that hardening curves for the 1st, 10th, 100th, and 1000th cycles fan out from a stress near twice the yield strength for the starting material. Again, we observed strain rate effects in the cyclic hardening behavior. The cyclic stress range decreased with decreasing strain rate and increasing temperature. High strain range tests produced more strain rate sensitivity than low strain range tests.

CREEP BEHAVIOR

The character of the creep curves for the different alloys varied widely and in Fig. 8 we compare curves for the alloys. Tests were selected that lasted for approximately the same time. The annealed 2 1/4 Cr-1 Mo steel (heat 56448) crept rapidly compared to the other materials but exhibited only small changes in the creep rate over very large strains. In fact, the creep rate was still decreasing slightly at strains above 30%. The bainitic steels, on the other hand, had very small primary creep periods and very large tertiary creep periods that often started at strains below 1%.

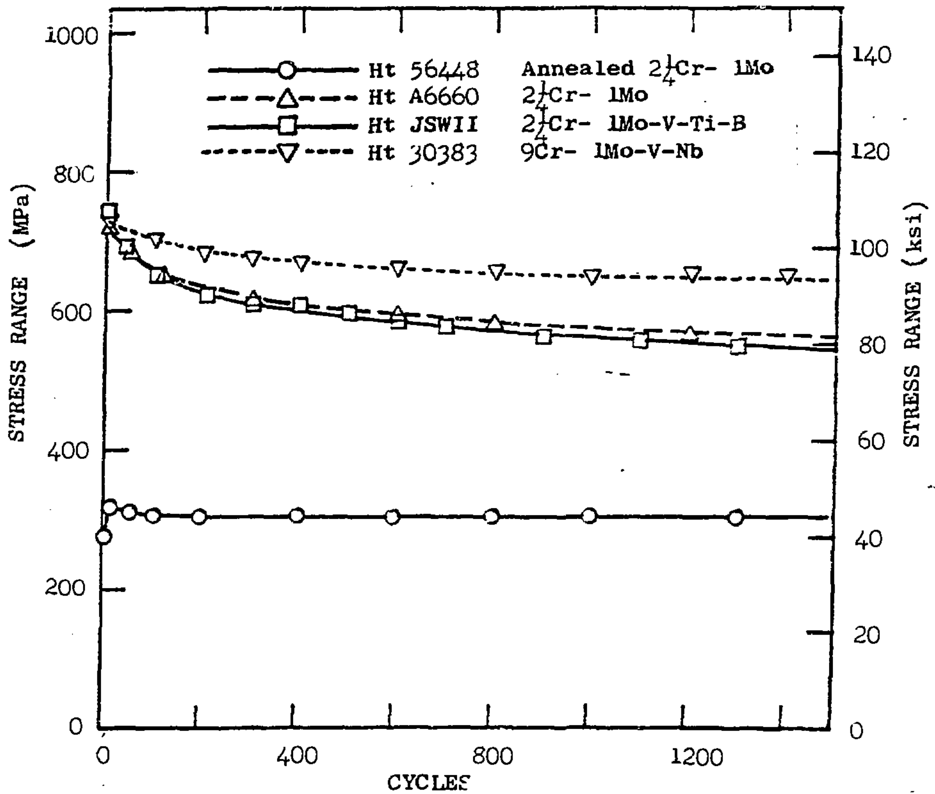


Fig. 6. Cyclic stress range versus cycles for four steels at 480 to 500°C and strain range near 0.6%.

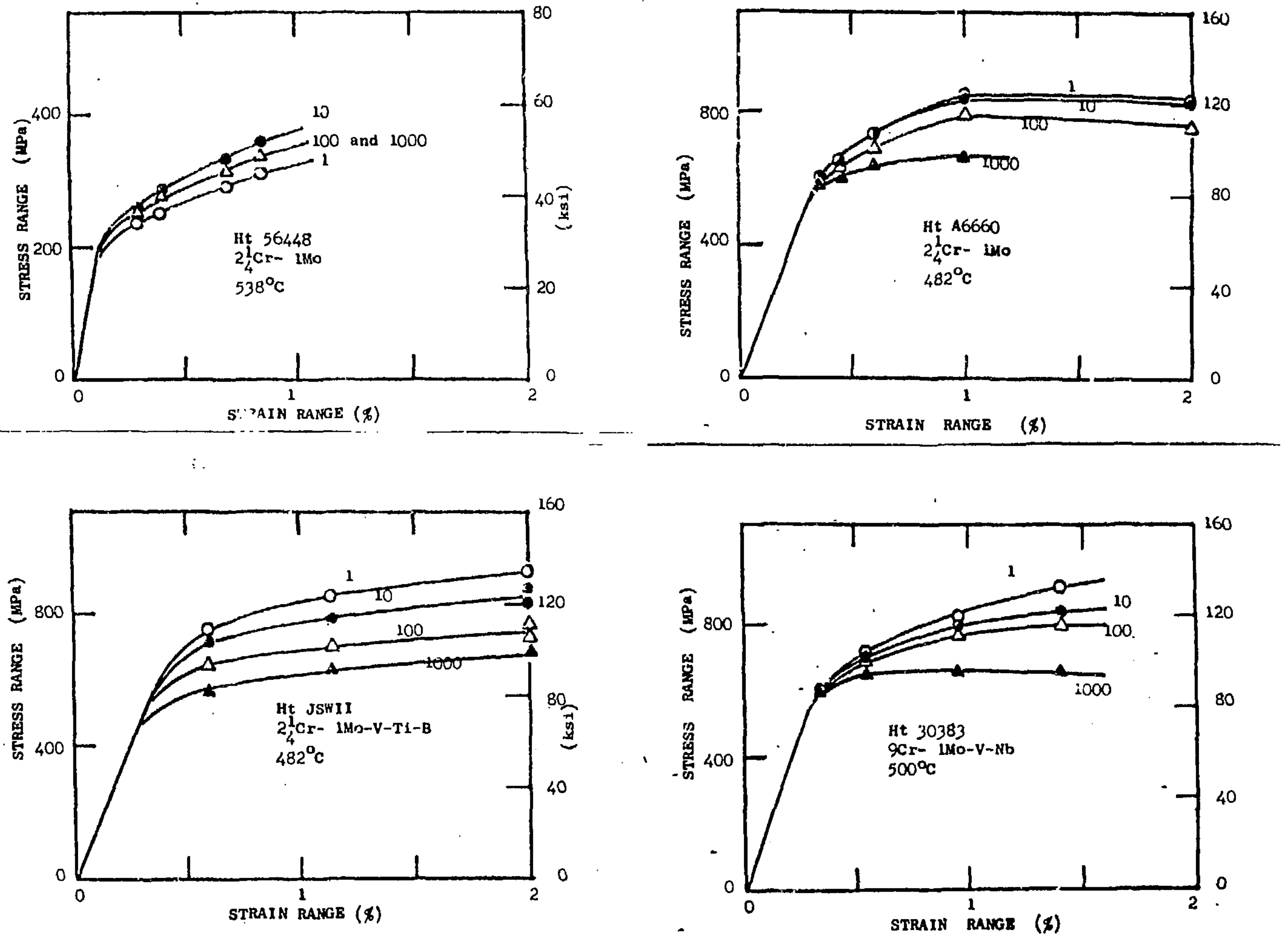


Fig. 7. Stress range versus strain range at 1, 10, 100, and 1000 cycles for (a) annealed 2 1/4 Cr-1 Mo steel heat 56448 at 538°C, (b) 2 1/4 Cr-1 Mo steel heat A6660 at 482°C, (c) 2 1/4 Cr-1 Mo-V-Ti-B steel heat JSWI at 482°C, and (d) 9 Cr-1 Mo-V-Nb steel heat 30383 at 500°C.

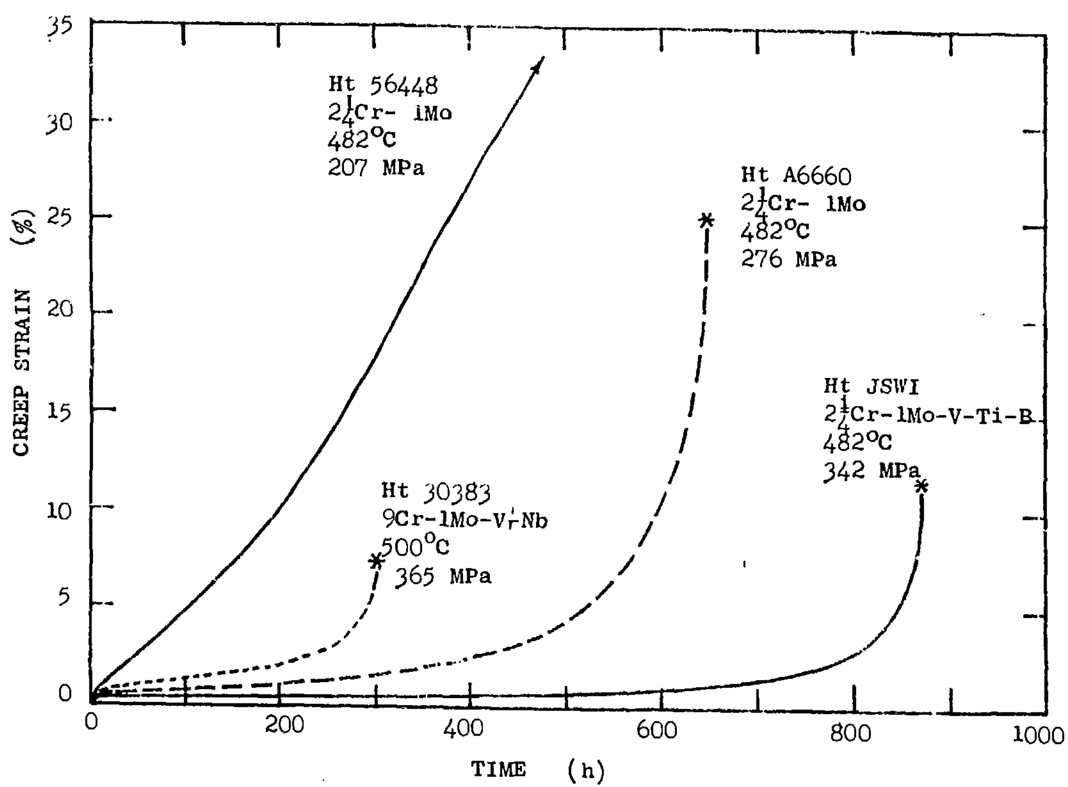


Fig. 8. Shape of short-life creep curves for four steels at temperatures in the range 480 to 500°C.

The creep behavior of the four alloys could best be compared on the basis of the effect of stress on the log creep rate versus log creep strain pattern as shown in Fig. 9 for data at temperatures in the range 480 to 500°C. For example, the results of creep tests on the annealed 2 1/4 Cr-1 Mo steel (heat 56488) at 482°C are shown in Fig. 9(a). The log creep rate for the test at 207 MPa decreased almost linearly with log creep strain then reached a minimum around 3%. Tests at lower stresses produced rapidly decreasing creep rates that attained minimum rates around 0.1% creep strain. After this, creep rate increased but eventually gave way to a new period of constant creep rate that lasted until large strains developed. This type of behavior has been often observed and has been analyzed in some detail by Klueh[18] for annealed 2 1/4 Cr-1 Mo steel.

The log creep rate versus log creep strain data for the bainitic 2 1/4 Cr-1 Mo steel (heat A6660) were limited to results from only a few short-time tests. Data are plotted in Fig. 9 (b) and show that the minimum rate developed around 0.5 to 1%. After a few percent strain accumulation the creep rate at high stresses accelerated and produced a tertiary creep stage that continued until rupture. Behavior at low stresses was not experimentally determined for this material.

The log creep rate versus log creep strain behavior of the bainitic 2 1/4 Cr-1 Mo-V-Ti-B steel (heat JSWI) is illustrated in Fig. 9(c) for tests at 482°C. The creep rate decreased rapidly with creep strain and reached a minimum at a strain that decreased with stress. For example, the minimum rate was reached around 0.7% creep strain at 413 MPa stress

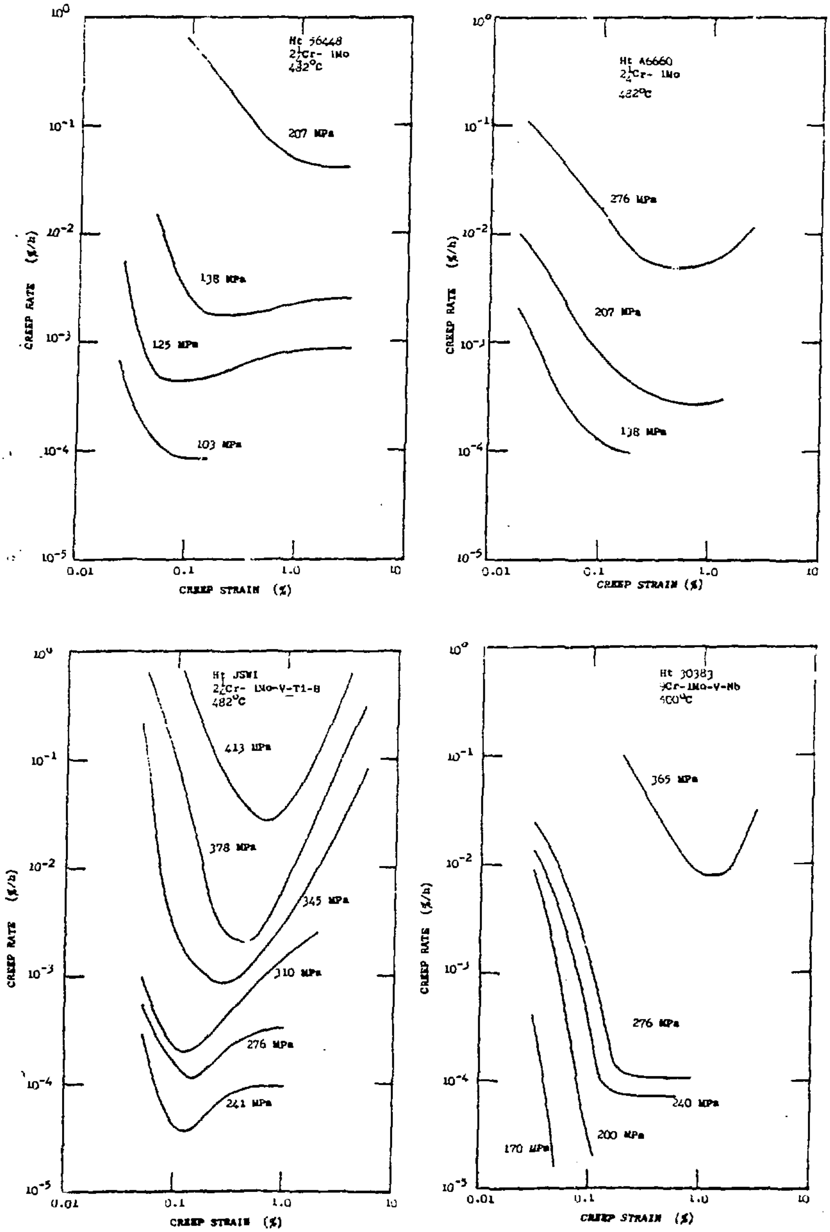


Fig. 9. Log creep rate versus log creep strain for (a) annealed 2 1/4 Cr-1 Mo steel heat 56448 at 482°C, (b) 2 1/4 Cr-1 Mo steel heat A6660 at 482°C, (c) 2 1/4 Cr-1 Mo-V-Ti-B steel at 482°C, and (d) 9 Cr-1 Mo-V-Nb steel heat 30383 at 500°C.

while the minimum occurred near 0.1% at 241 MPa stress. High stress tests accelerated into tertiary creep after reaching the minimum but low stress tests were qualitatively similar to the tests on annealed 2 1/4 Cr-1 Mo steel material in that a period occurred after the minimum creep rate in which a linear creep stage was established at a higher creep rate than the minimum.

The log creep rate versus log creep strain behavior of the martensitic 9 Cr-1 Mo-V-Nb steel at 500°C appeared to be qualitatively different from the lower chromium steels. Most data plotted in Fig. 9(d) revealed a very rapid decrease in creep rate with creep strain that reached a minimum around 0.5% at low stresses. After this, the creep rate remained relatively constant or decreased slightly for almost a decade of strain accumulation. More higher stress data are needed for a complete comparison of behavior with the other materials in the temperature range 480 to 500°C.

With increasing temperature, we found changes in the creep behavior of all four alloys. Generally, the rate of hardening at the start of creep tests diminished and the linear creep stage became more pronounced. Further, the strains at which tertiary creep started increased significantly, especially for the high-strength alloys.

Variable stress creep tests were performed on all four alloys. Testing was of an exploratory nature designed to examine general trends. We found that short transients were associated with both upward and downward stress increments. However, after a few tenths of a percent creep strain accumulation, the creep rate versus creep strain data for the stress change tests usually fell close to the trends illustrated in Fig. 9 for the appropriate conditions.

RELAXATION

Relaxation tests were performed on all four alloys in the temperature range 480 to 600°C. Data for three of the steels are compared in Fig. 10, which shows stress versus log time curves for tests on the two bainitic steels and the 9 Cr-1 Mo-V-Nb steel that were started at stresses near 275 MPa. The 2 1/4 Cr-1 Mo-V-Ti-B steel exhibited the least relaxation after 100 h, and the 9 Cr-1 Mo-V-Nb steel relaxed the most. The trends were consistent with the creep data in Fig. 9 that indicated initially higher creep rates in the 9 Cr-1 Mo-V-Nb steel. Repeated relaxation testing of the sample, however, produced hardening in the 9 Cr-1 Mo-V-Nb steel and eventually it proved to have greater relaxation strength than the other steels (as shown by the dashed curves in Fig. 10). This behavior pattern is consistent with expectations based on the trends shown in Fig. 9.

Although relaxation of the stresses in all four alloys continued well beyond 100 h, the rate of relaxation diminished rapidly. We found, as suggested by Penney and Marriott,[19] that the relaxation strength at 100 h was close to the stress producing a creep rate of 10^{-5}%/h for temperatures in the range 480 to 600°C. Figure 11 shows comparisons of relaxation and creep strengths for the four alloys. For the case of the 9 Cr-1 Mo-V-Nb alloy, we used the relaxation strength of the second loading rather than the first since this seemed to be more consistent with the general strength level of the steel.

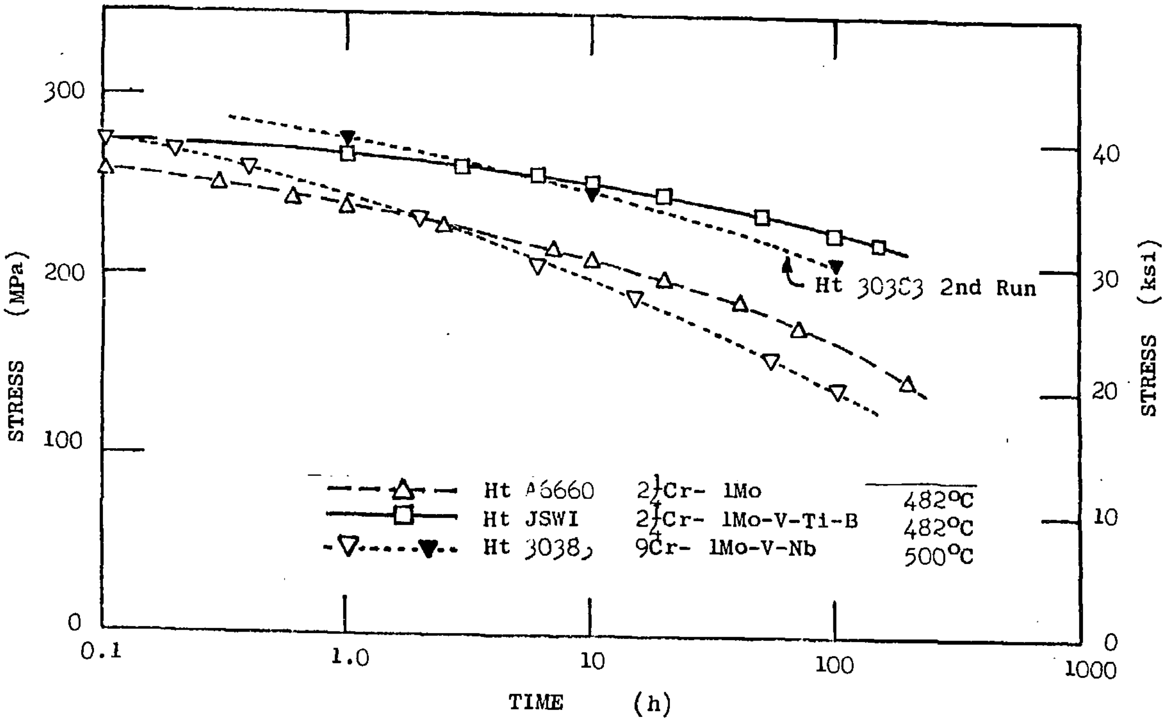


Fig. 10. Relaxation curves for three steels.

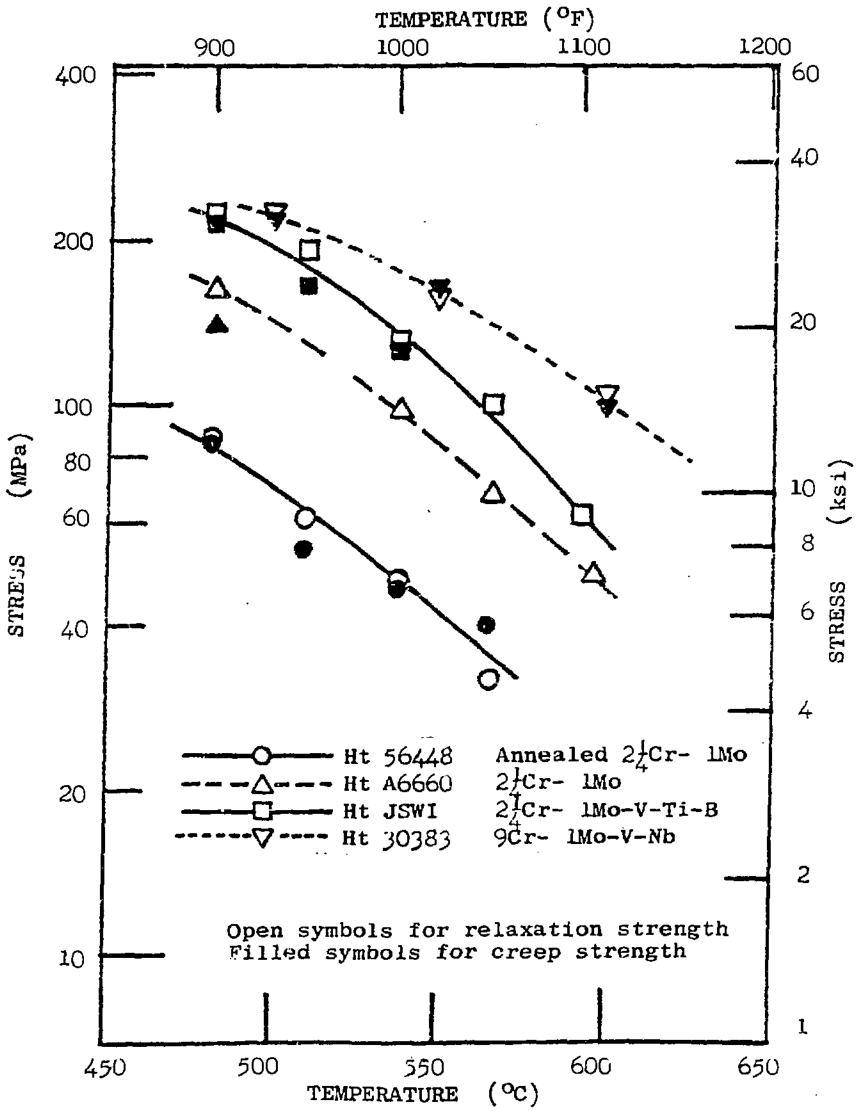


Fig. 11. The 100 h relaxation strength and the stress to produce 10^{-5}%/h creep rate versus temperature for four alloys.

CREEP PLASTICITY INTERACTIONS

We were primarily concerned with two aspects of creep-plasticity interactions. First, how do periods of creep or relaxation affect the cyclic hardening behavior described earlier and illustrated in Fig. 7? Second, how does cyclic plasticity affect the creep or relaxation behavior? There are, of course, other aspects of creep-plasticity interactions that could produce loses of ultimate tensile strength or creep ductility but such issues are beyond the scope of this work.

The most commonly performed creep-plasticity interaction test is probably the creep-fatigue test in which a relaxation hold time is introduced at one or both strain limits of a strain-controlled fatigue test. Vital information can be gleaned from just a few such tests in which long-time relaxation periods are followed by reloading to reveal the change in flow stress or by reversed loading to examine the isotropic and kinematic aspects of the change in the flow stress. Variables such as temperature, ramp strain rate, and cyclic strain range also influence the behavior but a great number of tests is needed to fully characterize the constitutive behavior of each of the materials.

The annealed 2 1/4 Cr-1 Mo steel that we tested exhibited a relatively simple and consistent pattern. First, as indicated in Fig. 7(a), only slight changes occurred in the cyclic stress range with continuous cycling, especially at temperatures above 500°C. The introduction of relaxation hold periods in the cyclic test produced only small recovery effects. As the material relaxed, the center of the kinematic yield surface tended toward zero stress, and the material

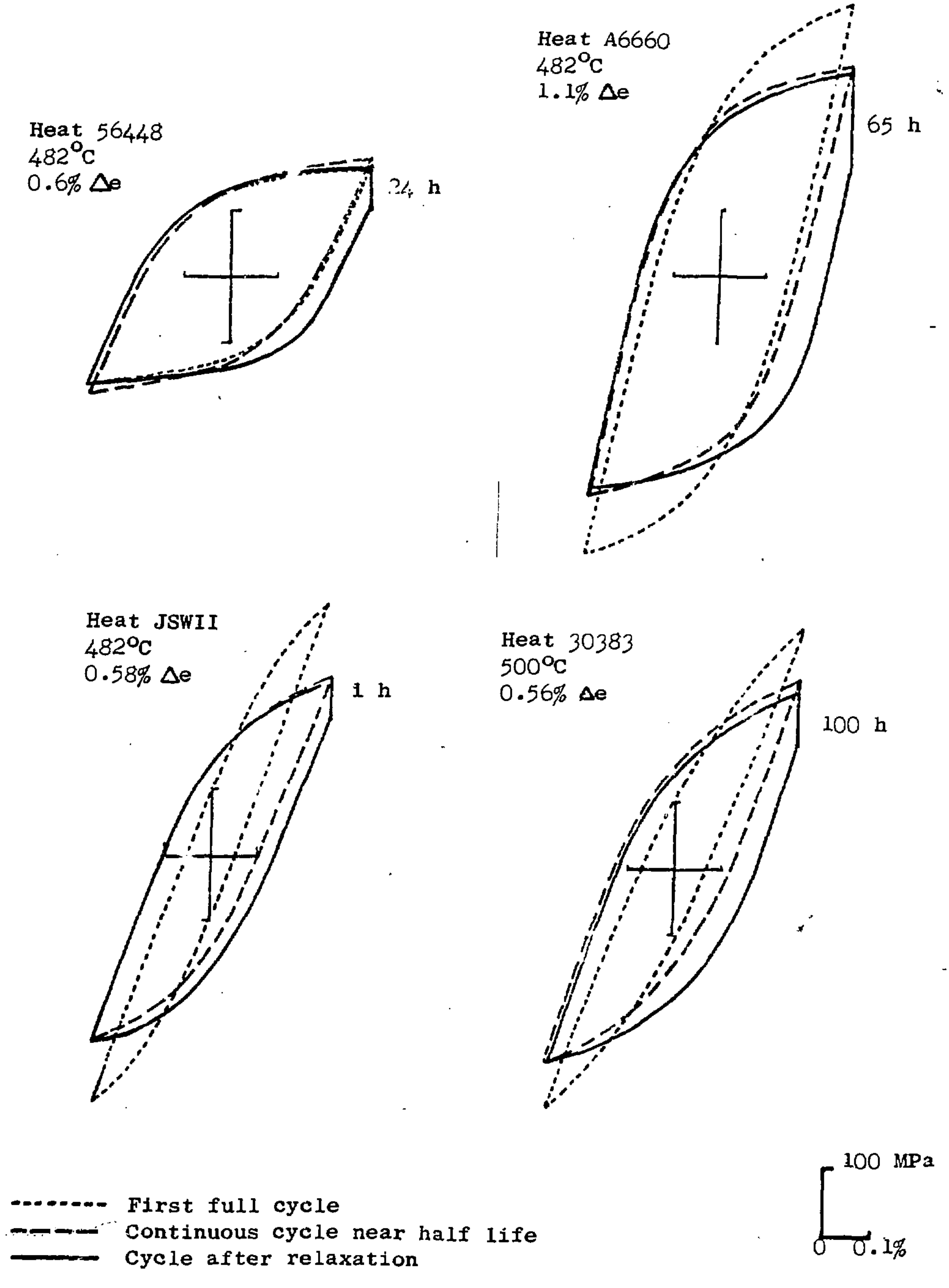


Fig. 12. Effect of relaxation on the hysteretic loop around half the fatigue life of four steels (a) annealed 2 1/4 Cr-1 Mo steel heat 56448 at 482°C and 0.6% strain range, (b) 2 1/4 Cr-1 Mo steel heat A6660 at 482°C and 1.1% strain range, (c) 2 1/4 Cr-1 Mo-V-Ti-B steel at 482°C and 0.58% strain range and (d) 9 Cr-1 Mo-V-Nb steel heat 30383 at 500°C and 0.56% strain range.

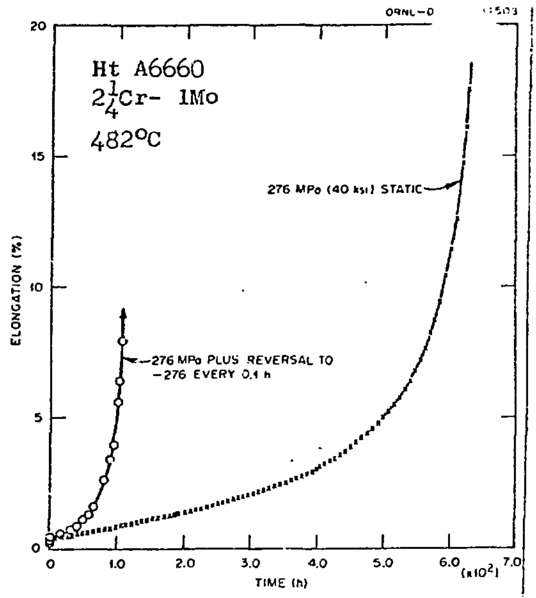
essentially returned to its original unstressed condition. Relaxation periods of 100 h were adequate to erase history effects and restore the material to the response pattern exhibited in the first stress versus strain hysteretic loop. Stress versus strain curves that compare the first complete cycle of the continuous cycling test with results obtained from a cyclic test that involved relaxation hold periods have been constructed in Fig. 12(a). These curves reveal only a small change in the cyclic stress amplitude as a result of relaxation. Generally, we found the annealed 2 1/4 Cr-1 Mo steel to quickly re-adjust to changes in strain rate, strain range, relaxation hold times or temperature changes, and follow patterns established from tests in which conditions were maintained constant throughout the test.

The bainitic steels were sensitive to history. Since the steels cyclically softened, the stress versus strain response after a creep or relaxation period was linked to cyclic history. In Fig. 12(b), we show some curves that were produced by examining the flow stress and relaxation strength of a bainitic 2 1/4 Cr-1 Mo specimen tested to cyclic conditions typical of half of the continuous-cycling fatigue life. The stress range was reduced appreciably. The relaxation strength was also below that observed in monotonic tests.

The 2 1/4 Cr-1 Mo-V-Ti-B steel and the 9 Cr-1 Mo-V-Nb steel were qualitatively similar to the bainitic 2 1/4 Cr-1 Mo steel in regard to response to conditions of creep or relaxation introduced between strain cycles. Figures 12(c) and 12(d) show stress versus strain curves constructed to show trends. Again, the cyclic stress amplitudes after 10 or 50 percent of life were well below the first cycle behavior.

A few tests were performed in which creep was periodically interrupted by stress cycling. Tests on the annealed 2 1/4 Cr-1 Mo steel were at several temperatures and included several different load histograms designed to study the constitutive behavior in greater detail than we can discuss in this paper. The essential feature of the behavior of the annealed material was that after relatively short transients, the creep behavior always returned to the trend expected from constant load test. That is, the "steady-state" creep behavior typified by the creep curve in Fig. 8 represented the dominant trend.

Exploratory tests on the bainitic and martensitic steels revealed losses in creep strength when cyclic stresses were introduced at temperatures in the range 480 to 600°C. Typical results from tests are shown in Fig. 13. Here a creep stress of 276 MPa was imposed in tension but periodically reversed to compression and immediately returned to tension. The cycling produced about 10^{-5} inelastic cyclic strain at first but with the accumulation of cycling, the inelastic component of cycling strain increased and the subsequent creep rate during the constant stress creep period increased. The tertiary creep stage occurred much sooner. The life of the bainitic 2 1/4 Cr-1 Mo steel was shortened by a factor of six [Fig. 13(a)] while the 2 1/4 Cr-1 Mo-V-Ti-B steel lost more than a factor of ten in life and the 9 Cr-1 Mo-V-Nb lost at least a factor of twenty in life [Fig. 13(b)]. These testing conditions, of course, were very severe relative to service conditions, and we cannot say that similar losses would occur at lower stresses and longer times between cycles.



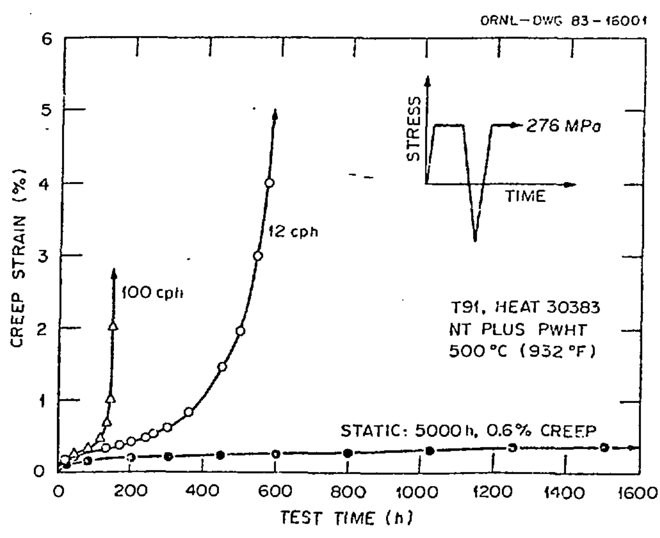


Fig. 13. Effect of stress reversals to −276 MPa on creep of two steels at 276 MPa (a) 2 1/4 Cr-1 Mo steel heat A6660 at 482°C and (b) 9 Cr-1 Mo steel at 500°C

A few non-isothermal tests were performed on the alloys. Tests of particular interest were two-bar ratchetting experiments[20] in which a mean load was imposed on a pair of specimens tested such that both specimens were forced to strain the same and maintain an average stress equal to the mean stress. The specimens were then thermal cycled by cooling one specimen from 538 to 300°C in 10 min, then the other. The two specimens were heated uniformly to 538°C for 10 min, held for 11.5 h, and the pattern repeated. Stress and strain were measured throughout the tests and ratchetting strains determined. We found that the annealed 2 1/4 Cr-1 Mo steel experienced plastic strains during the cycle and relaxation strains during the hold period. This produced a mean ratchetting in each cycle as shown in Fig. 14. The bainitic 2 1/4 Cr-1 Mo steel, however, experienced only elastic loading and unloading. The mean stress was too small to produce significant creep or relaxation after the first cycle or two hence the material did not ratchet. Similar behavior is expected for the 2 1/4 Cr-1 Mo-V-Ti-B and 9 Cr-1 Mo-V-Nb steels.

DISCUSSION

The allowable stresses specified in Sect. VIII of the code for the design of pressure vessels are based largely on tensile and creep-rupture properties with some consideration given to other information such as that derived from component operating experiences. We can compensate for the lack of operating experience for new alloys by expanding the data base for these steels to include information on strain rate effects, aging effects, notched bar strengths, behavior under cyclic stress or strain, and multiaxially stressed deformation and fracture behavior. However, supplementary data must recognize the particular applications for the

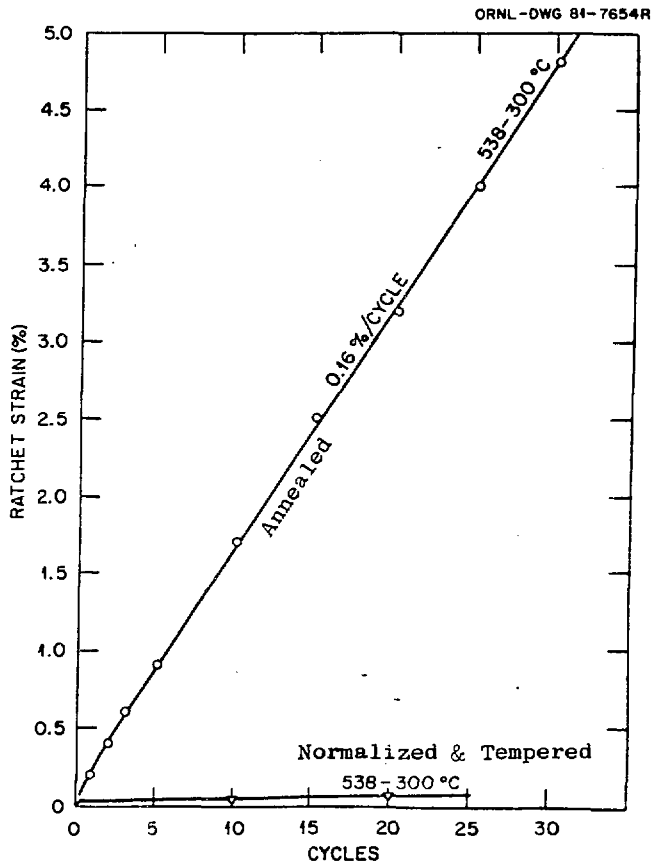


Fig. 14. Two bar creep ratchetting data for annealed 2 1/4 Cr-1 Mo steel and normalized and tempered 2 1/4 Cr-1 Mo steel.

alloys. For example, heavy wall pressure vessels used for petroleum and synfuels vessels operating at high pressures and temperatures will be designed to the maximum allowable primary stresses. Thermal transients will be slow, occurring over hours, if they occur at all, because of the low heat transfer properties of the process materials and heavy walls of the vessels. Low strain rate tensile data would be of value in assessing the relative merits of the new micro-alloyed steels since they may prove to be more strain rate sensitive than annealed 2 1/4 Cr-1 Mo steel at temperatures above 480°C. Our data are very incomplete in this respect, but data available in the literature suggest that significant decreases in the ultimate strength will occur as the strain rate falls from the standard 5×10^{-1}/min often used for tensile testing to a 1×10^{-6}/min rate representing typical service transients.

The high strain rate sensitivity and high primary creep rates of the bainitic and martensitic steels would be beneficial in promoting a rapid relaxation of secondary and residual stresses to values approaching the primary values. These features in combination with high yield strengths, low thermal expansion coefficients, and high thermal conductivities make the bainitic and martensitic steels attractive for high-temperature pressure vessel applications.

In contrast to petroleum vessels, nuclear vessels could have low primary stresses but undergo more severe transients. The work softening character of bainitic and martensitic steels should be of concern in complicated structures that contain regions of stress concentrations. Here it would be possible to severely degrade the properties of the micro-alloyed steels if the component experiences a significant number of cyclic loadings that exceed the proportional limit. Much more work needs to be done before we can identify what we mean by a significant number of cyclic loadings, but the initial indications are that the number will decrease with increasing strength level.

CONCLUSIONS

1. The 2 1/4 Cr-1 Mo-V-Ti-B and 9 Cr-1 Mo-V-Nb steels have excellent high-temperature tensile and creep-rupture properties.

2. At temperatures in the range 480 to 600°C, the steels relax in 100 h to strengths close to the stress producing creep rates of 10^{-5}%/h.

3. The bainitic and martensitic steels exhibit low elongation at the ultimate strength of tensile tests and softening in strain cycling tests. The softening is strain dependent rather than time dependent and is more severe for large cyclic strain conditions.

4. The bainitic steels are more resistant to ratchetting than the annealed 2 1/4 Cr-1 Mo steel.

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