The Microstructure and Mechanical Properties of a Modified 2.25Cr-1Mo Steel

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Tensile and creep properties were determined on a V-Ti-B-modified 2.25Cr-1Mo steel. The modified 2.25Cr-1Mo steel had about 0.2 pct V added for improved elevated-temperature strength. Boron was added to improve the hardenability, thus allowing thicker sections to be quenched or normalized to completely bainitic microstructures. Lower carbon and silicon concentrations were used (~0.1 pct C and 0.02 pct Si) than in standard 2.25Cr-1Mo steel. The modified steel had substantially better stress-rupture properties than did a standard 2.25Cr-1Mo steel (both with bainitic microstructures) with equivalent tensile properties — especially at the lowest stresses and highest temperatures. Comparative transmission electron microscopy studies of the standard and modified 2.25Cr-1Mo steels indicated that the differences involved the carbide precipitates and the dislocation substructures present in the steels.

I. INTRODUCTION

ISHIGURO et al. sought an improvement in the strength of 2.25Cr-1Mo steel by microalloying additions.[1] The objective was a pressure-vessel steel with better creep strength for thick sections for use in coal dissolver vessels, desulfurization vessels, and hydrocrackers. These investigators found that a low-silicon 2.25 pct Cr-1 pct Mo-0.1 pct C-0.25 pct V-0.02 pct Ti-0.002 pct B steel*

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*All compositions are in weight percent.

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gave improved creep strength and impact toughness over those for commercial 2.25Cr-1Mo steel in either the annealed or normalized-and-tempered condition. It was also claimed that the steel had superior resistance to temper embrittlement and improved weld heat-affected zone properties. Low silicon (0.02 pct vs 0.2 to 0.4 pct for most commercial 2.25Cr-1Mo steels) was chosen to increase resistance to temper embrittlement, and carbon content was restricted to 0.1 pct so as to avoid possible welding problems.

The addition of 0.002 pct B was credited with increasing the hardenability of the alloy, allowing for the formation of "martensite or bainite or a mixture of the two at a cooling rate as slow as 0.17 °C/s (10 °C/min)."[1] (The 0.02 pct Ti ties up nitrogen, and boron is not removed from solution as boron nitride.) The start of the proeutectoid ferrite transformation at 700 °C was moved from about 100 to about 2000 seconds.[1]

Precipitate stability is of prime importance for elevated-temperature strength. Ishiguro et al. stated that the steel contains vanadium and titanium carbides which are more stable than the chromium and molybdenum carbides formed in 2.25Cr-1Mo steel.[2] No proof of such a precipitate identification was given. When the constitution diagrams for Cr-Mo-V steels are consulted,[3] it appears that at equilibrium a 2.25Cr-1Mo steel containing 0.25 pct V would contain M_6C (a molybdenum-rich carbide), M_7C_3 (a chromium-rich carbide), and possibly MC which could contain vanadium and titanium.

Ishiguro et al. compared the creep properties of the modified and standard 2.25Cr-1Mo steel taken from the 1/4-thickness (1/4 T) position of a 200-mm plate.[1] The problem with such a comparison is that it was for different microstructures: the modified steel was entirely bainite, and the standard material contained large amounts of polygonal ferrite. The question, then, is whether the mechanical property differences are due to the titanium and vanadium additions to the modified steel or to the different microstructure caused by the increased hardenability of the modified steel. If the difference is due only to the difference in hardenability, boron (or manganese, or nickel) could be added to the standard 2.25Cr-1Mo steel to increase its hardenability.[4,5]

We determined tensile and creep-rupture properties on the modified 2.25Cr-1Mo steel. The results were compared with properties that had previously been determined on bainitic 2.25Cr-1Mo steel.[6,7] Transmission electron microscopy (TEM) was used to determine how the properties may be affected by the microstructure.

II. EXPERIMENTAL

The steel from the Japanese Steel Works came from a 25-mm plate from a 50-kg vacuum-induction melt with the chemical composition given in Table I. The plate was received in the normalized-and-tempered condition, which involved 4 hours at 1000 °C, air cooling, tempering 5 hours at 650 °C, followed by air cooling; this was designated the as-received condition. The plate was sectioned, and part of it given a second normalize-and-temper heat treatment — a "quality" heat treatment.[1] That treatment was 5 hours at 950 °C, air cool, followed by 20 hours at 690 °C, air cool.

The mechanical property test results from the modified 2.25Cr-1Mo steel were compared with results obtained on a heat of standard 2.25Cr-1Mo steel with the composition given in Table I. This steel was from a 25-mm-thick plate that was heat treated (normalized and tempered) as follows: 1 hour at 927 °C, air cool; 1 hour at 704 °C, air cool.

Tensile and creep tests were made on specimens that had a 6.35-mm-diameter by 31.8-mm-long reduced section. The tensile testing methods conformed to ASTM Recommended Practice E 21. The creep tests were made in air on lever-arm

Table I. Chemical Composition of Steels

Element	Composition, Wt Pct												
	C	Si	Mn	P	S	Ni	Cr	Mo	V	Ti	B	Cu	N
Modified 2.25 Cr-1Mo	0.14	0.07	0.49	0.008	0.007	0.16	2.34	1.00	0.26	0.032	0.0024	0.08	0.0085
Standard 2.25 Cr-1Mo	0.11	0.39	0.60	0.012	0.016	0.05	2.10	0.90	0.02	0.01	—	0.20	0.0112

creep frames with 12:1 ratios. Testing methods conformed to ASTM Recommended Practice E 132.

Optical microscopy, electrolytic carbide extractions, and TEM studies were conducted on material cut from the unaffected portion of test specimens. The carbide extractions were produced on samples with an approximate mass of 0.3 g in a solution of 10 pct HCl-90 pct methanol at 1.5 V for 5 hours. The extracted carbides were analyzed by X-ray diffraction. Electron microscopy was performed on JEOL 100C and 100CX transmission electron microscopes; the 100CX microscope was equipped with an X-ray energy-dispersive spectrometer (EDS) and modified for analytical work. Precipitates were identified by a combination of electron diffraction and EDS analysis.

III. RESULTS

A. Heat Treatment and Microstructure

All the comparisons of mechanical properties of the modified and standard 2.25Cr-1Mo steels were made for qualitatively similar bainitic microstructures. In Figure 1, the microstructures for the modified steel in the as-received condition [Figure 1(a)] and after the quality heat treatment [Figure 1(b)] are compared with the normalized-and-tempered 2.25Cr-1Mo steel [Figure 1(c)]. The prior austenite grain size of the modified steel was ASTM No. 6 in the as-received condition and No. 7 after the quality heat

treatment. The standard steel had an ASTM grain-size number of 5 to 6.

When the steels were examined by TEM, it was obvious that, despite the optical microstructures of Figure 1, there is no "uniform" microstructure. Considerable microstructural variation from grain to grain and sometimes within a given grain was found with any specimen.

The microstructure of untempered bainite is a highly dislocated lath structure.[8,9] We examined only the normalized-and-tempered structure (i.e., the microstructure tested). After the quality heat treatment, the modified 2.25Cr-1Mo steel still showed evidence of the high dislocation density as well as some indications of a remaining lath structure [Figure 2(a)]. The other distinctive feature of the microstructure was the precipitate distribution [Figures 2(b) and 3]. We observed basically two types of precipitate morphologies: a relatively small number of large, often polygonal, precipitate particles and a high density of much smaller particles. The larger particles were often, but not exclusively, found on grain boundaries. Precipitates were also often observed on lath boundaries (Figure 4).

When the distribution of small platelet-shaped precipitate particles was examined in more detail, it appeared that they were often on dislocations and, as discussed later, may have played a role in the evolution of the lath structure. It appeared that much of the fine precipitate was not picked up by the replica. Not all prior-austenite grains contained an indication of the lath structure as clear as that shown in the
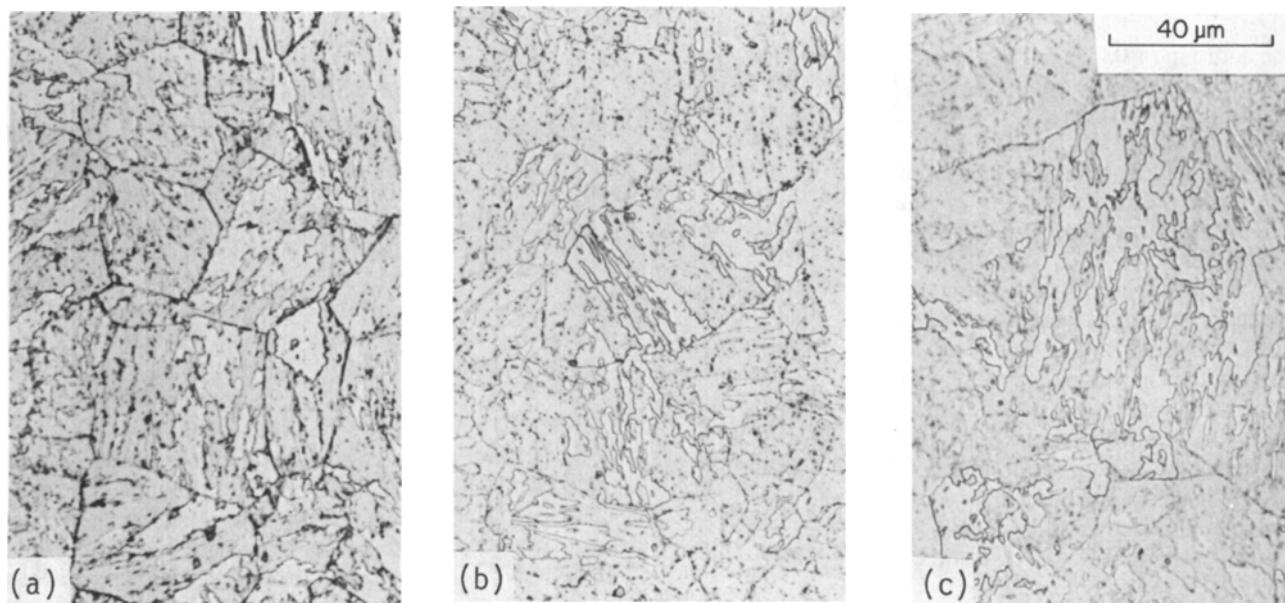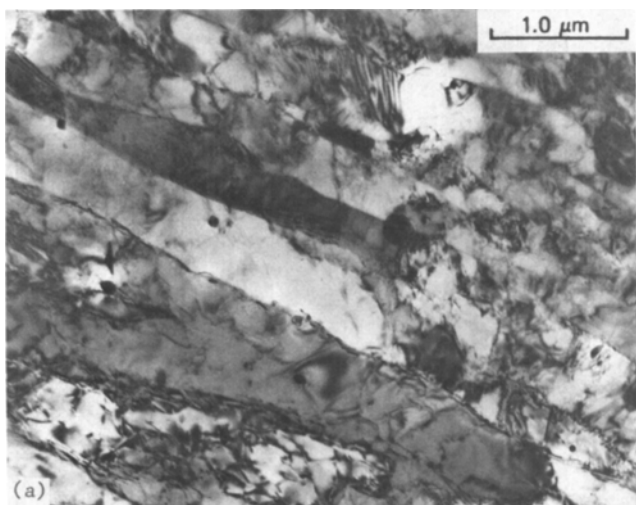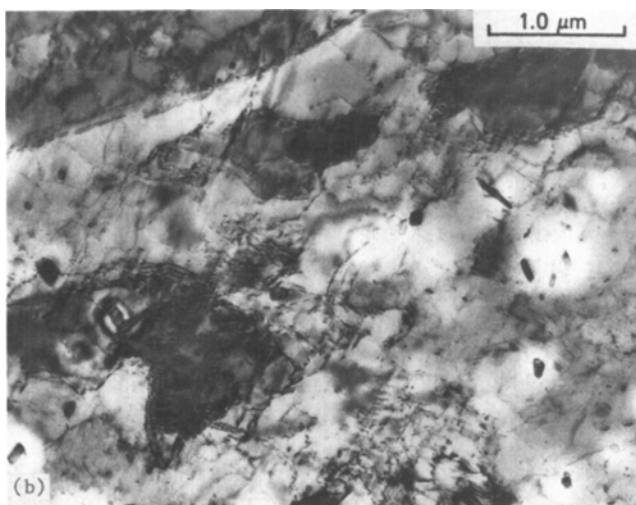

Fig. 1—Microstructures of V-Ti-B-modified 2.25Cr-1Mo steel in (a) the "as-received" condition and (b) the "quality" heat-treated condition. (c) Microstructure of a 25-mm-thick plate of standard 2.25Cr-1Mo steel in the normalized-and-tempered condition.

(a)

(b)

Fig. 2 — Transmission electron micrographs that show the (a) dislocation and (b) precipitate structure of V-Ti-B-modified 2.25Cr-1Mo steel.

Fig. 3 — Transmission electron micrograph of an extraction replica of V-Ti-B-modified 2.25Cr-1Mo steel.

Fig. 4 — Transmission electron micrograph that shows the precipitates on the lath boundaries of the V-Ti-B-modified 2.25Cr-1Mo steel.

upper part of Figure 2(a). In some cases, an elongated cell structure was observed within grains (Figure 4); this substructure was also associated with fine precipitates. The width of both the lathlike features and the elongated cell structure ranged from 0.3 to 0.7 μm.

The standard 2.25Cr-1Mo steel had a somewhat different appearance (Figure 5). Large polygonal precipitate particles were again present, but in this case most of the smaller particles were needle-like. On the replicas, the precipitate density in this specimen appeared larger, but this may be because much of the very fine precipitate in the modified 2.25Cr-1Mo steel was not picked up by the replication technique. Extensive precipitation on dislocations within the grains did not appear to have taken place as in the other steel.

When the carbides were chemically extracted, slightly more precipitate had formed in the standard 2.25Cr-1Mo steel (about 1.6 wt pct was extracted from the standard steel compared to about 1.4 wt pct from the modified steel). Again, we do not know whether all the precipitate was extracted. However, TEM indicated that the average pre-
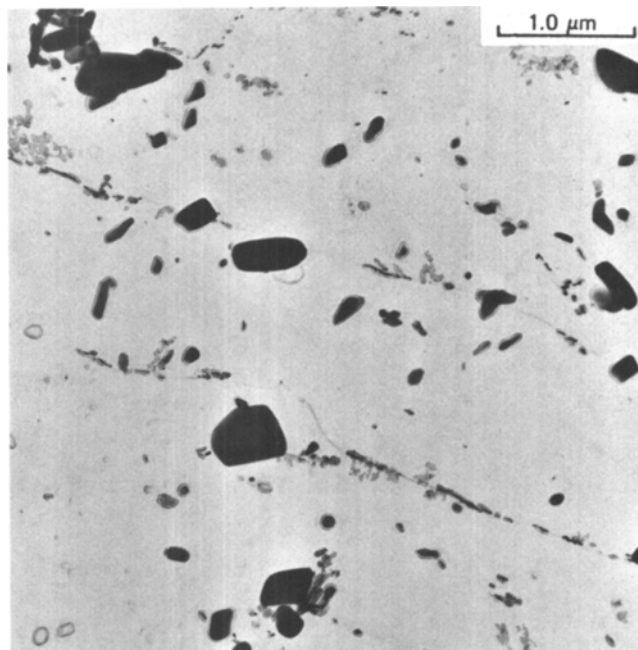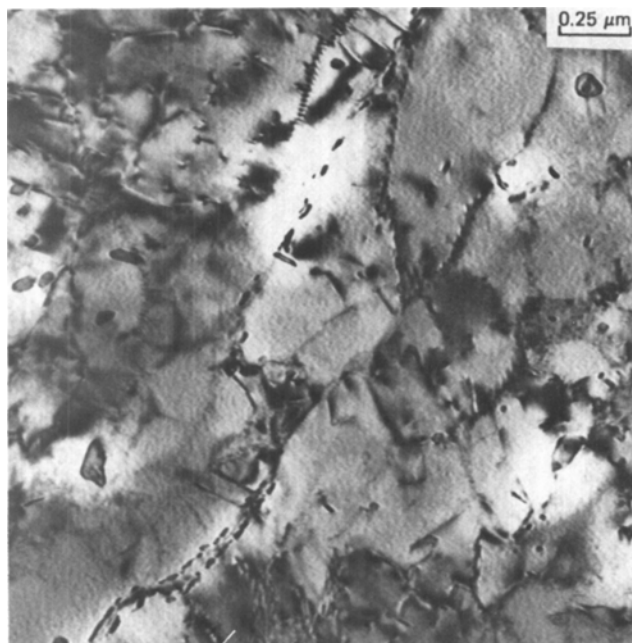
cipitate particle size in the standard 2.25Cr-1Mo steel was larger than that in the modified steel.

We tried to identify the extracted carbides by X-ray diffraction. Chromium-rich M_7C_3 was the only carbide detected in the standard steel. The X-ray diffraction of the precipitates from the V-Ti-B-modified 2.25Cr-1Mo steel indicated the presence of $M_{23}C_6$ ($\sim$80 pct) and M_7C_3 ($\sim$20 pct).

The EDS studies on the extraction replicas showed that the large particles in the standard steel were chromium rich, although many of them contained large amounts of iron,
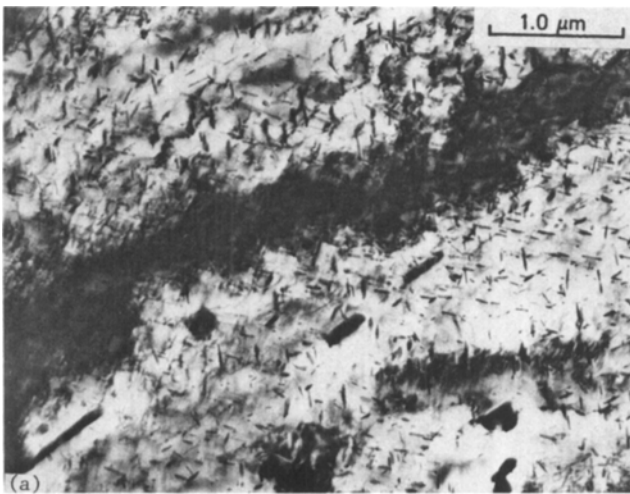
(a)
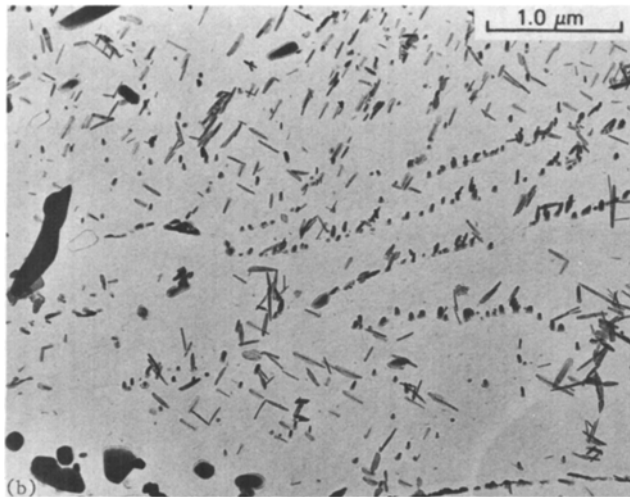
(b)

Fig. 5 — Transmission electron micrographs of bainitic standard 2.25Cr-1Mo steel from (a) thin film and (b) extraction replica.

which is typical for M_7C_3,[10] which was detected by X-ray analysis of the extracted carbides. A few of the smaller needles were found to be molybdenum rich, but with some chromium present. These particles are probably Mo_2C, which may not be present in large enough quantities to be identified in the extracted carbides. The EDS results on the modified steel indicated that all the large precipitate particles (Figure 5) were chromium rich, usually containing iron and sometimes molybdenum. Both iron and molybdenum can dissolve in large quantities in the chromium-rich $M_{23}C_6$ and M_7C_3.[10] The EDS analysis of the very small precipitate particles indicated that they contained primarily vanadium and molybdenum; several of these were found to be rich in vanadium and titanium with significant amounts of molybdenum also present. These precipitates are assumed to be the MC type of carbide.

B. Mechanical Properties

Figure 6 compares the tensile properties between room temperature and 600 °C for the normalized-and-tempered commercial 2.25Cr-1Mo steel[6] and the modified steel with
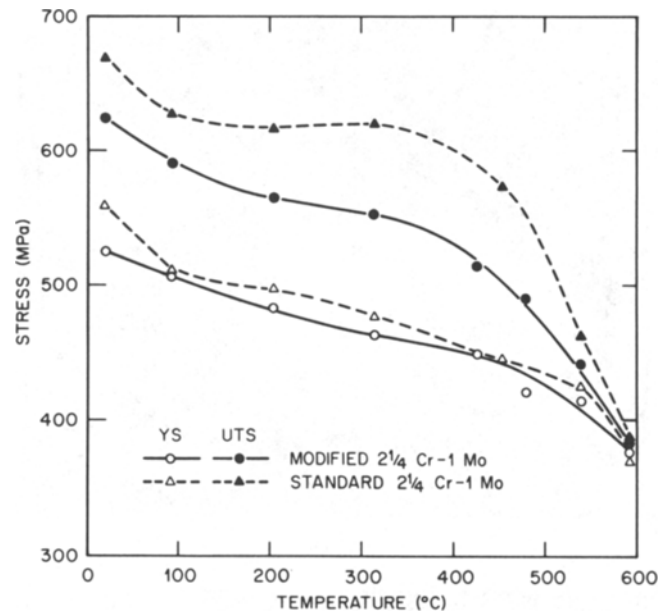

Fig. 6 — The 0.2 pct yield stress and ultimate tensile strength as functions of temperature for the V-Ti-B-modified 2.25Cr-1Mo steel with the quality heat treatment and a normalized-and-tempered standard 2.25Cr-1Mo steel, both in the tempered bainitic condition.

the quality heat treatment (both have tempered bainite microstructures). The 0.2 pct yield stresses for the two steels were quite similar over the entire temperature range, but the ultimate tensile strength was considerably higher for the standard steel, especially between room temperature and 500 °C (Figure 6). Ductilities of the two steels were similar: the total elongation of the modified steel was slightly greater than that of the standard steel; however, the modified steel had a slightly larger reduction of area.

Creep-rupture properties were determined at 482, 510, and 538 °C on the modified steel in the quality heat-treatment condition. Figures 7 through 9 present the stress-rupture curves, stress-minimum creep rate curves, and ductility-rupture life data. Over the range of temperatures and stresses of these tests, the steel exhibited good strength and adequate ductility.
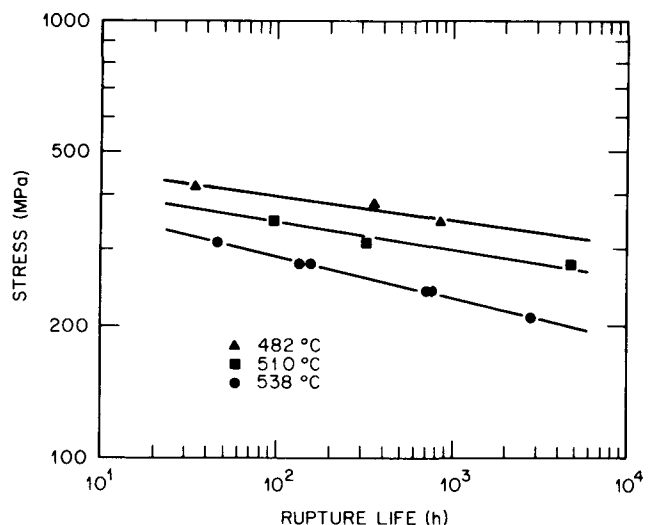

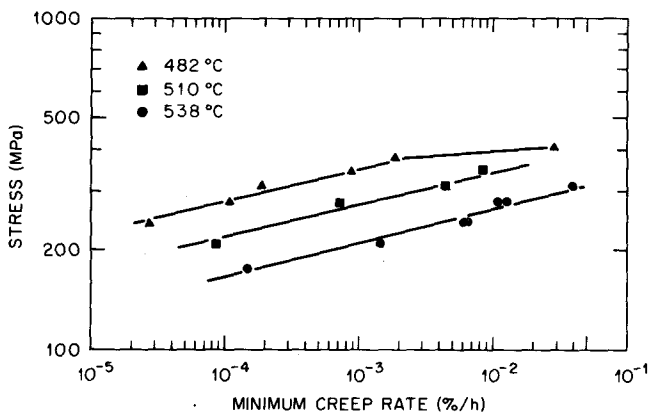Fig. 7 — Stress-rupture curves for V-Ti-B-modified 2.25Cr-1Mo steel.

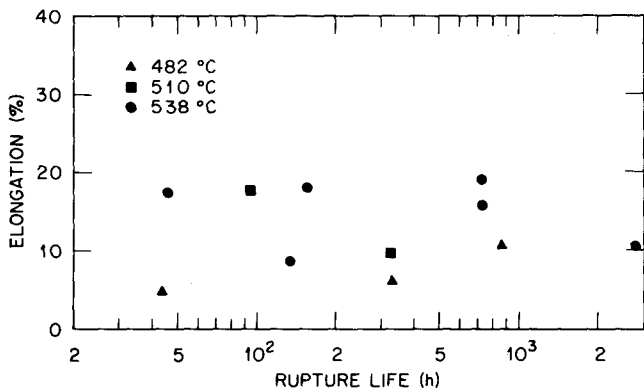Fig. 8—Stress-minimum creep rate curves for V-Ti-B-modified 2.25Cr-1Mo steel.

The stress-rupture curves for the modified and standard 2.25Cr-1Mo steels at 482, 510, and 538 °C show a definite advantage for the modified steel (Figure 10). At 482 °C, the properties were similar at high stresses (short rupture times), but at low stresses a large difference was observed. A similar divergence of properties was noted at 510 °C. At 538 °C, this difference was found at all stresses over the range tested.

Observations on creep-curve shape of the modified 2.25Cr-1Mo steel revealed two interesting characteristics that have also been observed on standard 2.25Cr-1Mo steel (Figure 11). Most of the creep occurs with an increasing creep rate (i.e., relatively small primary and secondary stages and an extended tertiary creep stage). Second, the

Fig. 9—Variation in rupture elongation with rupture life for V-Ti-B-modified 2.25Cr-1Mo steel.
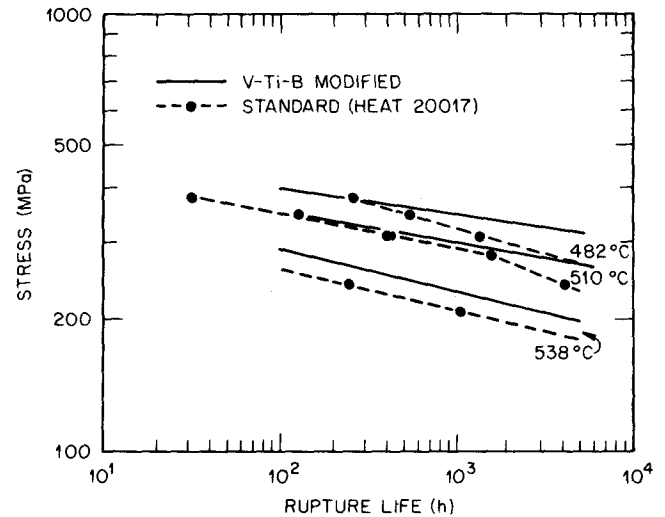

Fig. 10—Comparison of the stress-rupture curves for V-Ti-B-modified and standard 2.25Cr-1Mo steels with similar tensile properties.
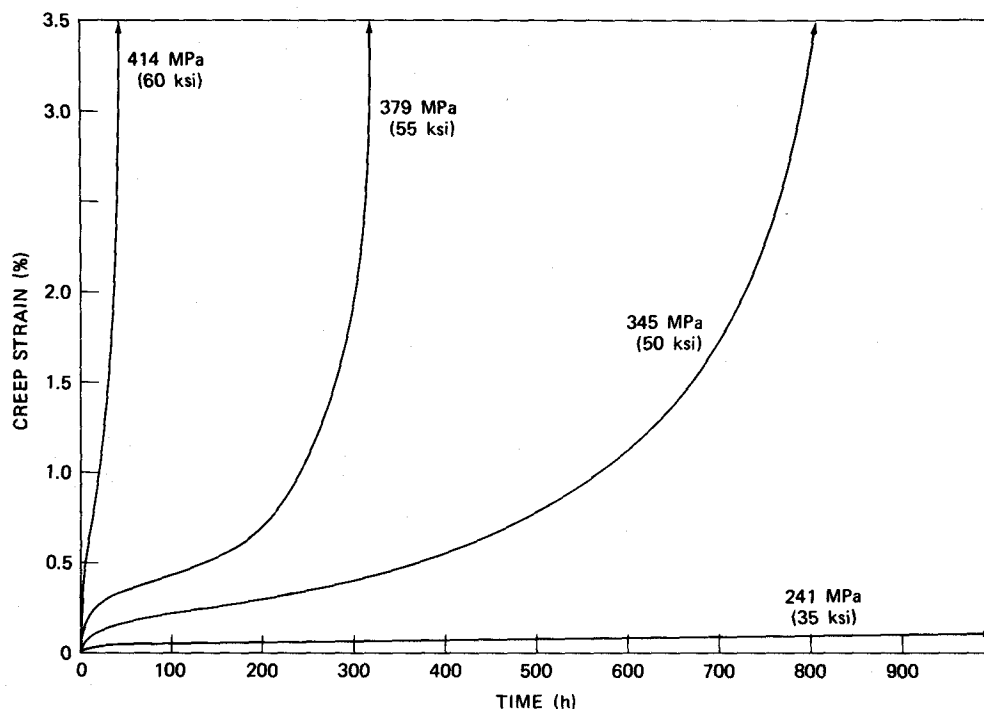

Fig. 11—Creep curves for V-Ti-B-modified 2.25Cr-1Mo steel at 482 °C.

specimen tested at 482 °C and 241 MPa had a nonclassical creep curve; that is, this curve had two steady-state stages.

Visual observations of fractured specimens indicated that the fractures were of a cup-cone type. The transgranular nature of the fractures was verified by metallographic examination of selected specimens.

IV. DISCUSSION

A. *Microstructure*

The TEM observations on the V-Ti-B-modified 2.25Cr-1Mo steel indicated that in those grains in which a lath structure remained, the lath width was 0.3 to 0.8 μm, with carbides along lath boundaries (Figures 2 and 3), features that are indicative of an upper bainite morphology.[8,9] It should be noted, however, that the carbides along the lath boundaries could have formed from retained austenite or martensite during tempering, thus making this granular bainite. Granular bainite is a variation of upper bainite that retains high-carbon austenite, which may then form martensite when cooled to room temperature.[11] The retained austenite or martensite then transforms during tempering. However, granular bainite is generally less lath-like and has a more blocky appearance than observed here — thus, the term "granular."[11] Further studies in the untempered condition are required to establish the nature of the bainite. Previous work indicated that granular bainite forms in the 25.4-mm plate of standard 2.25Cr-1Mo steel when normalized.[12]

The major differences in the microstructure of the modified and the standard 2.25Cr-1Mo steels involved the fine matrix precipitates. The modified steel contained a distribution of small particles that were vanadium rich with significant amounts of molybdenum and, in some cases, also contained titanium. They were assumed to be vanadium carbide or MC. Although vanadium carbide is generally given as V_4C_3, it is the MC-type carbide,[10] with an NaCl-cubic structure, but with composition limits extending to V_4C_3. Todd, Chung, and Parker studied this same steel and concluded that these small particles were vanadium carbide.[13] They found that the large particles in the microstructure were $M_{23}C_6$ but gave no indication of the presence of M_7C_3.

The smaller particles of the standard 2.25Cr-1Mo steel were considerably larger than the small particles of the modified steel. These were identified by X-ray diffraction as M_7C_3; the EDS studies indicated that Mo_2C needles were also present. An M_7C_3 precipitate is known to form in bainitic 2.25Cr-1Mo steel, as shown in the isothermal-transformation diagrams determined by Baker and Nutting,[2] and more recently in studies by Hippsley[14] and Pilling and Ridley.[15] According to these authors, M_7C_3 is generally preceded by Mo_2C, and, for a steel tempered as this one was, Mo_2C, and probably M_3C, should be present. The needle-like precipitates observed by TEM are indicative of the Mo_2C morphology, although M_7C_3 is also known to form in such a morphology.[10]

The fine precipitates within the matrix of the modified steel were often associated with dislocations and with a substructure present within the laths of the tempered bainite. Although large numbers of dislocations were present in the matrix of the standard 2.25Cr-1Mo steel, the density was much less than that in the modified steel; also, the matrix precipitates in the standard steel did not appear to be associated with the dislocations.

The X-ray diffraction analysis of the precipitates extracted from the modified steel showed them to consist of $M_{23}C_6$ (~80 pct) and M_7C_3 (~20 pct). The $M_{23}C_6$ precipitate agrees with previous observations,[13] although no M_7C_3 was observed in that study. Our analysis of the precipitates did not include a detailed determination of their distribution in the microstructure (Figure 5) relative to their morphology and size. Indications are that the large blocky precipitates are the $M_{23}C_6$ particles and that the M_7C_3 particles are the medium-size precipitate particles — primarily of an elongated rectangular shape.

When the constitution diagrams for Cr-Mo-V steels are consulted,[3] the presence of M_7C_3 in these steels is expected. For a 2.25Cr-0.25 V (~0.1 to 0.15 pct C) steel annealed 1000 hours at 700 °C, M_7C_3 and M_4C_3 (V_4C_3) were found; for this same heat treatment in an alloy with 1 pct Mo, M_7C_3 and M_6C were found.[3] This uncertainty concerning the presence of V_4C_3 follows from the limited number of alloys studied.[3] Thus, the existence or nonexistence of V_4C_3 at equilibrium at the tempering temperature is not established. Furthermore, no information is available for the service temperatures for this steel.

The existence of a fine MC carbide could greatly affect the long-term creep properties of this steel. It should be noted that the alloys used to construct the equilibrium diagrams[3] differ from the V-Ti-B-modified 2.25Cr-1Mo steel. Titanium also forms the MC carbide with an NaCl-cubic structure.[10] The small amount of titanium present could possibly stabilize the MC.

The observations indicate that during the tempering of the modified steel the fine vanadium carbide forms on dislocations, thus locking in the upper bainite lath structure or locking in an elongated cell structure, which develops from the lath structure. In the standard steel, on the other hand, tempering apparently allows for an overall coarsening of the structure, probably because of the difference in the nature of the precipitation process that takes place when the matrix precipitates form. This coarser substructure for the standard steel superimposed on the originally coarser prior-austenite grain size undoubtedly plays a significant role in the difference in mechanical properties.

A cursory TEM examination of the as-received steel revealed a very high density of fine precipitate particles but gave little indication of a substructure or remnants of the original lath structure such as that observed after the quality heat treatment. This could be interpreted to mean that at the lower tempering temperature (650 °C) much of the dislocation structure is "annealed out" before precipitate formation and stabilization, whereas at the higher tempering temperature (690 °C) used in the quality heat treatment, more rapid precipitate growth stabilizes that substructure. Another possibility is that the 950 °C austenitization temperature during the quality heat treatment (as opposed to 1000 °C for the first normalization treatment) is not high enough for all the fine vanadium carbide formed at the 650 °C tempering temperature to dissolve. Thus, the undissolved precipitates stabilize a dislocation structure in the bainite when the steel is tempered at 690 °C. More TEM work would be needed to verify this.

B. *Mechanical Properties*

The elevated-temperature strength of standard 2.25Cr-1Mo steel has been attributed to two processes: interaction solid-solution hardening[16,17] and precipitate dispersion strengthening, primarily by Mo_2C.[18,19] The effect of these processes on the tensile and creep behavior has been discussed in some detail for a polygonal ferrite[18,19] and a bainitic ferrite[7,18] microstructure. Because both these processes may play a role in strengthening the V-Ti-B-modified 2.25Cr-1Mo steel, they are briefly discussed here.

Interaction solid-solution hardening is a solid-solution-hardening effect attributed to the strengthening that results from the simultaneous presence in solution of substitutional and interstitial atoms with an affinity for each other (*e.g.*, C and Mo or C and Cr). As a result of this strong attraction, atom pairs or atom clusters form and subsequently interact with dislocations to form atmospheres.[16] Interaction solid-solution-hardening effects have been attributed to carbon-molybdenum interactions in 2.25Cr-1Mo steel with a polygonal ferrite microstructure and to carbon-chromium interactions in the bainitic condition.[17]

Interaction solid-solution hardening can affect creep curve shape.[19] The resulting curves have been classified as classical and nonclassical. Classical curves have primary, secondary, and tertiary creep stages. Nonclassical curves have two steady-state stages. A primary and steady-state stage are followed by a stage with an increasing creep rate; instead of increasing to rupture, however, the rate again decreases, going into a second steady-state stage, which finally gives way to an increasing creep rate and rupture. One such curve was observed for the modified steel. For standard 2.25Cr-1Mo steel, it was concluded that for the first steady-state stage, creep is controlled by interaction solid-solution hardening.[17] Eventually, precipitation removes the supersaturated solution of interacting atoms, and the creep rate increases to a new steady state, in which creep is controlled by atmosphere-free dislocations moving through a precipitate field.

The observation of a nonclassical creep curve for the V-Ti-B-modified steel at 482 °C indicates that interaction solid-solution hardening may also occur in that steel. The nature of the interaction solid-solution-hardening process in this steel could possibly be different from that observed in standard 2.25Cr-1Mo steel. For the modified steel, the strongest interaction between interstitial and substitutional solutes should occur between C and V and/or Ti, as opposed to C and Mo and/or Cr in the standard steel. The Fe-V-C and Fe-Ti-C alloys have not been studied as have Fe-Mo-C and Fe-Cr-C.[16]

The observation that a large portion of the creep curves for the V-Ti-B-modified 2.25Cr-1Mo steel consisted mostly of an increasing creep rate (tertiary creep) is similar to observations on annealed 2.25Cr-1Mo steel.[20] In that case the observation was explained as being caused by interaction solid-solution hardening.[17] Such creep curve shapes were attributed to nonclassical creep in which only the first steady-state stage was delineated. That is, the first steady-state stage was so long that the second steady-state stage was never observed; by the time the first steady-state stage was finished, the second steady-state stage was so short that it occurred by itself over a period of time too short to permit

identification. For standard 2.25Cr-1Mo steel at 510 °C, this type of curve was observed at the highest stresses used. At low stresses, nonclassical curves with two steady-state stages were observed. It is the similarity in the high-stress curves for modified 2.25Cr-1Mo steel to those observed for the standard 2.25Cr-1Mo steel that leads to the conclusion that a similar effect occurs for modified 2.25Cr-1Mo steel. Furthermore, although only a single nonclassical curve was observed, it occurred for the lowest stress test (241 MPa), in agreement with the observations on the standard 2.25Cr-1Mo steel.

Interaction solid-solution hardening also affects the tensile properties of standard 2.25Cr-1Mo steel. The effect is exhibited as a dynamic strain-aging peak when the ultimate tensile strength is plotted against temperature.[17] The peak occurs between 300 and 450 °C, depending on the strain rate. A slight peak is noted for the standard steel in Figure 6; the peak height for this steel becomes more pronounced as the strain rate is decreased.[17] The modified steel shows no indication of a peak at the strain rate (2.7×10^{-4}/s) used for the present tests (Figure 6). This may be a manifestation of the difference in the V-C and/or Ti-C interaction, as opposed to the standard steel, in which Cr-C and Mo-C interactions occur.[17] Vanadium and titanium form more stable carbides than chromium and molybdenum and would be expected to interact more strongly with carbon. The reduced dynamic strain-aging peak may be the result of a small supersaturation of vanadium and titanium relative to their carbides when the steel was cooled from the tempering temperature. Further studies are required to understand these differences. Such studies could prove worthwhile, for, if interaction solid-solution hardening under these conditions were understood in more detail, it could perhaps be applied in the further development of chromium-molybdenum steels.

Although interaction solid-solution hardening is important to the difference in tensile behavior and high-stress creep-rupture behavior of the modified and standard steels, the difference in long-time creep properties depends on the difference in precipitate particle size and distribution in the two steels.[17] As the rupture life increases, the relative amount of time that interaction solid-solution hardening controls the creep process decreases. For such tests, dispersion strengthening by the fine vanadium carbide in the modified steel and molybdenum and chromium carbides in the standard steel determines the creep strength.

The difference in creep-rupture properties of the two steels becomes greatest at the lowest stresses (longest rupture times) (Figure 11). This observation is similar to observations by Todd, Chung, and Parker[13] and Ishiguro *et al.*[1] on these same two steels.[13] The finer distribution of small precipitate particles in the modified steel and the distribution on dislocations would be expected to improve the elevated-temperature strength of the modified steel relative to the standard steel.

Previous work has indicated that the addition of chromium beyond 2.25 pct has no marked strengthening effect.[21] The only reason for adding more chromium is to improve the oxidation and corrosion resistance of the steel (the chromium also increases hardenability). Therefore, for many applications to 550 °C the present work again demonstrates that it is unnecessary to increase the chromium to 9 to 12 pct as is often done.

V. SUMMARY AND CONCLUSIONS

Tensile and creep properties have been determined for a V-Ti-B-modified 2.25Cr-1Mo steel (0.25 pct V, 0.02 pct Ti, 0.002 pct B). The results of the mechanical property determinations were compared with results on standard 2.25Cr-1Mo steel. However, although previous investigators compared steels given similar cooling rates during a normalizing heat treatment, we compared steels having similar bainitic microstructures. The difference in microstructure in the Japanese studies was the result of the 0.002 pct B added to the modified steel. If the boron had been present in the standard steel, a similar hardenability and, hence, a similar basic microstructure should result.

The results generally verified the observations of the Japanese investigators concerning the excellent elevated-temperature strength of the modified 2.25Cr-1Mo steel. The improvement in the mechanical properties of the modified steel was attributed to a fine, relatively stable vanadium carbide precipitate that formed, primarily on dislocations, when the steel was normalized and tempered in accordance with the recommended procedure. This precipitate stabilizes a fine dislocation substructure within the bainite lath structure that developed during normalization, as opposed to the precipitate structure that forms in bainitic standard 2.25Cr-1Mo steel, which is much coarser and less stable. More work is required to understand the evolution of the carbides during elevated-temperature exposure, because the evolution of the carbides will determine the usefulness of this steel at elevated temperatures.

ACKNOWLEDGMENTS

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