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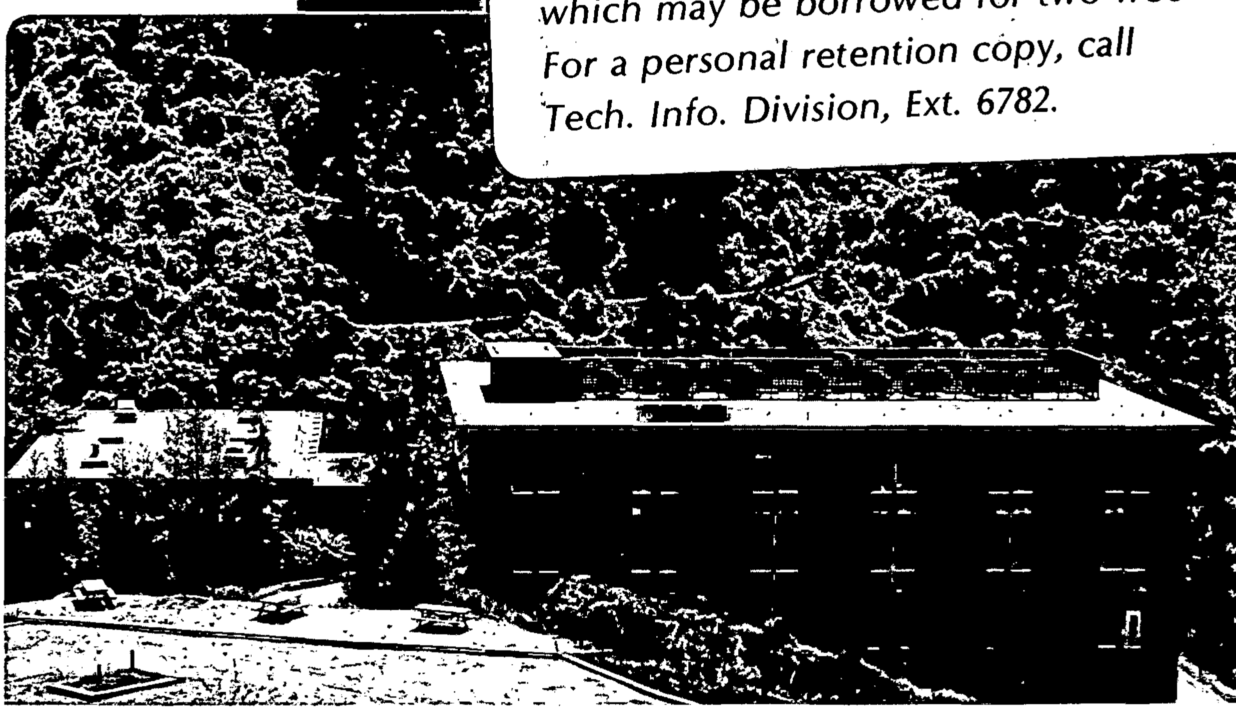

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INTRODUCTION

AISI 304 stainless steels are the most widely used structural alloys in cryogenic service. To the cryogenic engineer they offer good fracture toughness, good fatigue properties, and adequate strength at 4 K. Additionally, their paramagnetism and corrosion resistance are often desirable properties. Recently 304 stainless steels have been used in superconducting magnet structures for magnetic fusion energy devices. Large amounts of steel are required in these structures, for example nearly 300 tonnes of AISI 304LN were used in the yin-yang magnet cases for the Mirror Fusion Test Facility at the Lawrence Livermore National Laboratory.

It is well-known that austenitic (γ-phase) 304 stainless steels are metastable with respect to the martensite transformation at low temperatures. Mechanistic details of the phase transformation to bcc martensite (α'-phase), such as the importance of an intermediate hexagonal phase (ε-phase) are controversial. It is well-known, however, that mechanical stress will promote the $\gamma \rightarrow \alpha'$ transformation in this material[1,2]. The effects of this transformation on work-hardening and fatigue crack growth rate are discussed below.

High magnetic fields are known to promote the $\gamma \rightarrow \alpha'$ transformation in metastable austenitic steels. Even when the $\gamma \rightarrow \alpha'$ transformation is athermal, all experimental evidence (mostly from Soviet studies, e.g.[3,4]) suggests that the additional $\gamma \rightarrow \alpha'$ transformation due to high magnetic fields correlates to the increased free energy difference between the γ and α-phases; the thermodynamic potential of the α-phase is lowered when its magnetic moments are aligned in an applied magnetic field. For austenitic alloys of many different compositions, it appears that a 1 T magnetic field biases the material in favor of the formation of α′-martensite by the about same amount as does a reduction in temperature of 2 K[3,4]. A thermodynamic study of the role of an applied magnetic field on the $\gamma \rightarrow \alpha'$ transformation was not a goal of the present work. Here we report initial measurements of tensile and fatigue crack growth rate properties of 304 stainless steels exposed to 8 T magnetic fields. Recently, no effect of a 7 T magnetic field on the tensile properties of 304 stainless steel at 4 K was observed[5], but observations of small effects at 4 K could be masked by the jerky plastic flow due to the Basinski mechanism. However, in another study at 77 K employing pulsed 16 T magnetic fields[6], plastically strained specimens of 304L and 304LN exhibited increased elongation in the presence of the magnetic field, together with an increased amount of α′-martensite.

RELATIONS BETWEEN THE MARTENSITE TRANSFORMATION AND WORK-HARDENING BEHAVIOR IN TENSILE TESTS

This section describes the possible effects of the $\gamma \rightarrow \alpha'$ transformation on the stress-strain curves of 304 stainless steels. Although a variety of metallurgical mechanisms have been proposed to account for mechanical behavior at low strains, here we will be concerned only with the significant work-hardening observed after strains of more than 0.1. This work-hardening consistently correlates with the formation of α′-martensite[1,2]. Our quantitative formulations are limited to mechanical behavior during uniform elongation, for which the stress is homogeneous, and will invoke only general assumptions about the phenomenology of work-hardening and the $\gamma \rightarrow \alpha'$ transformation.

If we assume that our material has a fixed and finite capacity to elongate by slip, then the transformation strain of the $\gamma \rightarrow \alpha'$ transformation can help accomodate an increase in length of the material. Accomodating an increase in length by the $\gamma \rightarrow \alpha'$ transformation strain will slow the rate of work-hardening, and may thereby prolong the operation of the slip systems of the material. The macroscopic true strain of the material, ε, may be written as[7]:

$$\varepsilon = \varepsilon_{el} + \varepsilon_p + \phi \cdot f \qquad (1)$$

where ε_{el} is the elastic strain, ε_p is the strain arising from slip, f is the fractional extent of the $\gamma \rightarrow \alpha'$ transformation, and ϕ is an average coupling between the transformation strain (mostly shear) and the increase in length of the material. Assuming conservation of volume during deformation, the Considerè criterion for the necking instability becomes[7] :

$$d\sigma = \sigma(\varepsilon_p) \{ d\varepsilon_p + \Phi \cdot df \} \quad (2)$$

where $\sigma(\varepsilon_p)$ is the stress required to elongate the material after it has been plastically strained an amount ε_p. All differentials are taken at the true strain ε.

At this point in our development of the necking criterion we allow the $\gamma \rightarrow \alpha'$ transformation to interact with the mechanisms of slip. Previous work[6] suggested that the $\gamma \rightarrow \alpha'$ transformation served to induce additional slip in the material to relieve the applied stress. However, we hypothesize that the $\gamma \rightarrow \alpha'$ transformation also induces some slip that helps to accomodate the $\gamma \rightarrow \alpha'$ transformation strains, but does not accomodate the macroscopic elongation of the material. The second factor on the right hand side of Eq. 2 is a mechanical geometry factor, and consequently depends only on the macroscopic elongation of the material, and not on these microscopic processes of slip. However, the flow stress, σ, depends on all slip, both microscopic and macroscopic, that contributes to the work-hardening. We cannot continue to use a general plastic strain, ε_p, for both factors in Eq. 2, so we distinguish the contributions to plastic flow. The variable ε_{p1} is the macroscopic plastic strain that would occur without the $\gamma \rightarrow \alpha'$ transformation. The term $\Theta \cdot f$ is the plastic strain induced by the $\gamma \rightarrow \alpha'$ transformation that accomodates the elongation of the material; Θ is the amount of coupling between the $\gamma \rightarrow \alpha'$ transformation and the slip which contributes to the elongation of the specimen. Likewise, the additional slip which does not contribute to the elongation is: $\Omega \cdot f$. Assuming that Φ, Θ, and Ω are not functions of ε, it is straightforward to show that the necking criterion becomes:

$$d\sigma = \sigma(\varepsilon_{p1} + (\Theta+\Omega) \cdot f) \{ d\varepsilon_{p1} + (\Theta+\Phi) \cdot df \} \quad (3)$$

These considerations of the necking criterion illustrate three relationships between the $\gamma \rightarrow \alpha'$ transformation and work-hardening behavior: 1) The transformation strain can reduce the large work-hardening rate of 304 stainless steels and thereby reserve the slip systems for further deformation at large strains. The $\gamma \rightarrow \alpha'$ transformation may cause two effects through direct modification of the slip behavior: 2) The induced slip which accomodates the elongation of the specimen has the same effect as the slip that would occur without the $\gamma \rightarrow \alpha'$ transformation. 3) By obstructing the operative slip systems, or by causing slip that does not accomodate the elongation of the specimen, the $\gamma \rightarrow \alpha'$ transformation will use up the work-hardening capacity of the material. The changes in the stress-strain curve associated with extra $\gamma \rightarrow \alpha'$ transformation will be determined by a competition of effects 1) and 3). The relative importance of these two effects may vary from material to material.

EFFECTS OF THE MARTENSITE TRANSFORMATION ON FATIGUE CRACK GROWTH RATE

Changes in the fatigue crack growth rate caused by the $\gamma \rightarrow \alpha'$ transformation are far more difficult to predict than changes in the stress-strain behavior; the stress distribution is much more complicated, and detailed mechanisms of fatigue crack growth are unknown. Nevertheless, it is well-known that in the Paris regime, data on the fatigue crack growth rate per cycle, da/dN, can usually be related to the applied range of stress intensity, ΔK, by a power law: $da/dN = D \cdot (\Delta K)^m$. Here D and m are material-dependent constants. Many theories and models of fatigue crack growth rate have been proposed to justify values of m from 2 to 4. For 304 stainless steels with various carbon and nitrogen contents, m = 4 [8], and this is consistent with the predictions of dislocation models for Mode III loading, and also for theories based on the energy in the plastic zone at the crack tip[9]. In these theories and others[10], the assumption of work-hardening in front of the crack tip will serve to reduce da/dN. Recently the effect of the $\gamma \rightarrow \alpha'$ transformation strain on the fatigue crack growth rate has been treated in a semiquantitative way[11]. As in more rigorous treatments of the effects on fracture toughness[12], it was argued that the dilatational part of the $\gamma \rightarrow \alpha'$ transformation strain will develop compressive stresses in front of the crack tip, making it more difficult to propagate the crack through regions of extensive transformation.

Unlike the situation for tensile tests, both the work-hardening and the transformation strain associated with the $\gamma \rightarrow \alpha'$ transformation should serve to reduce da/dN. Tobler and Reed found that reducing the carbon plus nitrogen content of 304 stainless steel from 0.128 wt.% to 0.067 wt.% reduced da/dN by a factor of 4 at 77 K and by a factor of 8 at 4 K, in apparent correlation to the increased amount of $\gamma \rightarrow \alpha'$ transformation at 4 K [8]. We take a quantitative risk and now predict magneto-mechanical effects on the fatigue grack growth rate in terms of these composition effects by an equilibrium thermodynamic argument. The effect of an 8 T magnetic field on the $\gamma \rightarrow \alpha'$ transformation should be equivalent to $-16^{\circ}C$. It is less likely that the effect of C+N content on the temperature dependence of the $\gamma \rightarrow \alpha'$ transformation results from changes in the equilibrium thermodynamic driving force. However, if we assume that it does, and since we know that the change in the martensite transformation temperature is about $473^{\circ}C/\%C+N$, we expect an 8 T magnetic field to be equivalent to a change in C+N content of 0.034%. This implies that an 8 T magnetic field should reduce the fatigue crack growth rate by perhaps a factor of two at 77 K.

EXPERIMENTAL METHODS

Our plates of 304L and 304LN stainless steel had carbon plus nitrogen contents of 0.098 wt.% and 0.160 wt.%, respectively. Mechanical test specimens were machined from the 304LN plate directly, but the 304L material was given the heat treatment: 1050°C 2 hr / W.Q. Tensile specimens of 25.4 mm gauge length and 3.2 mm gauge diameter were machined from the plates with their axes parallel to the rolling direction. Tensile specimens were tested in pairs (one in the field, one out of the field), so care was taken to match them in dimensions, heat treatment, region of plate, and machining procedure. One pair of 304L specimens was heat-treated in an evacuated quartz ampule after machining: 960°C 1 hr / air cooling of ampule; we refer to these as the 'annealed pair'. Tensile tests employed the magneto-mechanical cryogenic research facility described previously[6]. Because of anticipated problems with operating strain gauges in 8 T magnetic fields, all tensile tests were performed under stroke control with a hydraulic actuator velocity of 0.001 cm/sec. The compliance of the mechanical linkage is included in the data of Fig 1.

Significant effort was directed towards accurate measurements of load and displacement during operation of the 8 T magnet. Preliminary tests showed that the load cell was affected by the fringing fields of the magnet, and the linear velocity and displacement transducer was affected to a smaller extent. An independent load measurement was obtained from the differential hydraulic pressure across the actuator piston; in this scheme the differential pressure transducer was well-removed from the fringing fields of the magnet. Surrounding the upper part of the load frame with steel I-beams was also effective in magnetically shielding the load and stroke transducers. A pair of brass tensile specimens tested with and without the applied magnetic field showed the same stress-strain curves to within ±1% in stress; we believe this to be the accuracy of our load measurements, and the stroke measurements were approximately this accurate.

Constant load fatigue crack growth rate tests also employed the MMCR facility. Compact tension specimens of 44.5 mm width and 23.2 mm thickness were used, and the crack plane orientation was L-T. Peak loads were controlled to within ±1%. The load cycle was a 10 Hz sinusoid, and the ratio of minimum to maximum load was 0.1. A C-shaped clip gauge was used to measure the load-line compliance of the specimen, from which the crack length was derived[13]. Stress intensity ranges were determined by interpolating numerical results for compact tension specimens[14]. Crack growth rates were determined with the secant method recommended by ASTM standard E-647. Two specimens of each material were tested. One of the specimens was tested entirely without the magnetic field, for the other the 8 T magnetic field was applied for a chosen number of cycles at three times during the test. Since the clip gauge readings in the presence of the 8 T magnetic field were originally suspect, the crack growth rate during the period of magnetic exposure was determined from the compliance measurements immediately before and after the application of the magnetic field; this total amount of crack growth was compared to the crack growth for the specimen tested entirely without the magnetic field.

Fatigue fracture surfaces were examined with an ISI scanning electron microscope operated at 25 kV with secondary electron detection. Phase analysis of the larger samples was performed with a Picker Nuclear x-ray diffractometer using Fe Kα radiation. A hand-held magnetic permeameter was used for surface measurements of the amount of α'-martensite. Phase analysis of electropolished 5-10 mg sections of the tensile specimens was performed at room temperature with a S.H.E. VTS-805 SQUID magnetometer using a 2kG applied magnetic field.

RESULTS AND DISCUSSION

Tensile Tests

Three matched pairs of 304L in the 'as-received' condition were pulled with and without the magnetic field. No magneto-mechanical effects were seen in the elastic region or in the first part of the plastic region. However for every pair, the specimen exposed to the 8 T magnetic field showed a more rapid rate of work-hardening, a larger tensile strength, and usually a smaller uniform elongation (see fig 1a). Similar, but less pronounced effects were seen in two pairs of as-received 304LN specimens (see fig 1b). Annealed 304L pulled in the magnetic field did not show its small increase in work-hardening until larger strains, and showed an increased elongation. The pair of 66% cold-worked 304L specimens showed the largest magneto-mechanical effects, including the earliest increase in work-hardening. Although the magnetically-induced increases in flow strength were small (3-7%), they were observed for all seven pairs of specimens tested. The total elongations of the three pairs of 304L specimens varied by $\pm.01$.
The dependence of the magneto-mechanical effect on the cold-work of the 304L specimens can be explained by the work-hardening effects of extra α'-martensite formed during magnetic exposure. The explanation requires the assumption that the prior cold-work causes the magnetically-induced $\gamma \rightarrow \alpha'$ transformation to further work harden the material without contributing effectively to the elongation (i.e. the prior cold-work causes effect 3 of the RELATIONS... section to become more important with respect to effect 1). Although this seems intuitively reasonable, our assumption of a fixed capacity for work-hardening is challenged by the small increase in

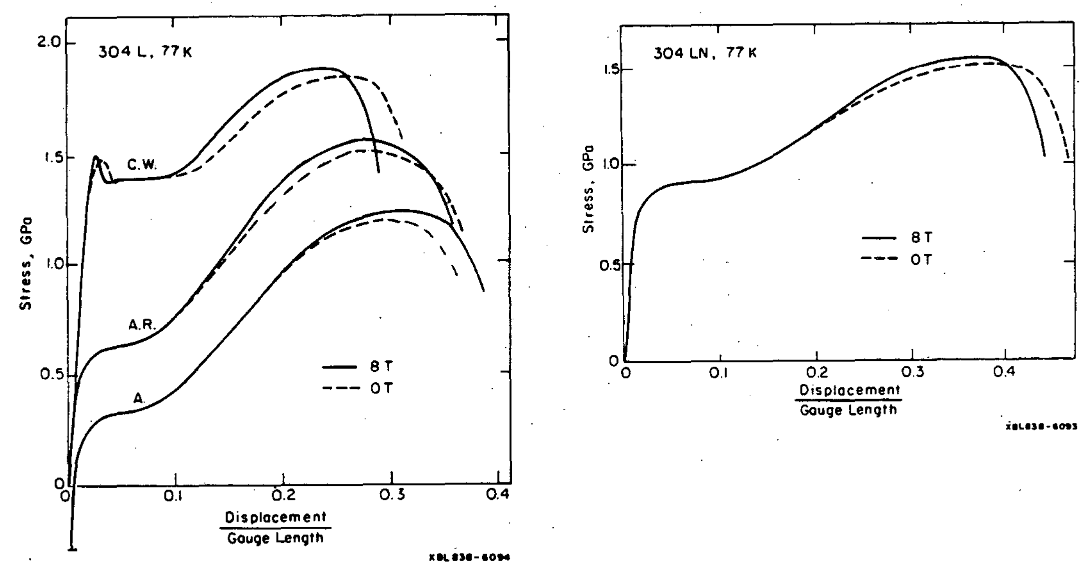

Fig. 1a. Engineering stress-stroke curves for 304L: C.W. - 66% cold worked, A.R. - 1050°C treatment only, A. - annealed (curves are shifted down by 0.321 GPa). Fig. 1b. Engineering stress-stroke curves for as-received 304LN.

ultimate tensile strength observed for the specimens tested in the magnetic field.

Figure 1a suggests that prior cold-work may help promote the magnetically-induced $\gamma \rightarrow \alpha'$ transformation. This is reasonable if some features of the dislocation sub-structure serve as nucleation sites for the $\gamma \rightarrow \alpha'$ transformation. However, within the estimated error of the SQUID magnetometer data, all specimens of 304L showed 50% of α'-martensite in their uniformly elongated regions. The amounts of martensite formed in the 304LN specimens were also indistinguishable, but they were over 10% less than the amounts formed in the 304L specimens.

Fatigue Crack Growth Rate

Our first data indicate that magneto-mechanical effects on the fatigue crack growth rates of 304L and 304LN stainless steel at 77 K are very small (see Fig. 2). Compliance measurements taken immediately before and after the operation of the 8 T magnet showed a decrease in the amount of fatigue crack growth with respect to the same number of cycles without magnetic exposure. For the as-received 304L this decrease was 15% at $\Delta K = 28 MPa\sqrt{m}$, 14% at $\Delta K = 35 MPa\sqrt{m}$, and 4% at $\Delta K = 45 MPa\sqrt{m}$. For less reliable data from the as-received 304LN, this decrease was -15% at $\Delta K = 32 MPa\sqrt{m}$, and 50% at $\Delta K = 50 MPa\sqrt{m}$. Although these effects are proportionately larger than the effects seen in the tensile tests, we must be cautious in claiming the existence of a magneto-mechanical effect from these data.

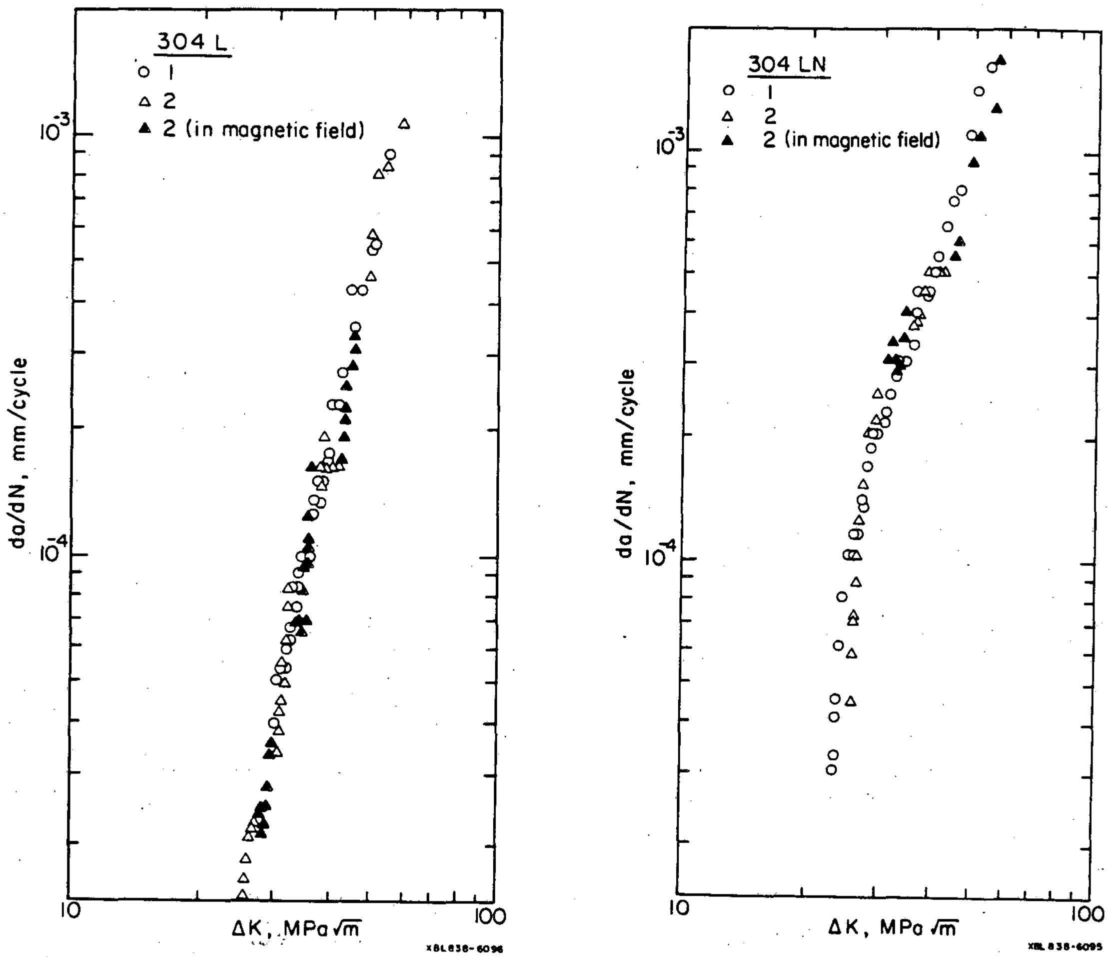

Fig. 2a. Fatigue crack growth rate for 304L: 1 - without field, 2 - with and without 8T field. Fig. 2b. Fatigue crack growth rate for 304LN: 1 - without field, 2 - with and without field.

Scanning electron fractography failed to reveal any difference between those parts of the fatigue fracture surface formed with and without magnetic exposure. The average amount of α′-martensite formed in the center of the fatigue fracture surface was > 85% for the 304LN specimen, and > 90% for the 304L specimen, as determined by x-ray diffractometry. Perhaps the magnetic field is ineffective in inducing further $\gamma \rightarrow \alpha'$ transformation during fatigue crack growth when the transformation is already so extensive.

CONCLUSIONS

From an engineering standpoint, the effects of an 8 T magnetic field on the strength and fatigue crack growth rate of 304L and

304LN stainless steels at 77 K are small. Nevertheless, the work-hardening behavior of 304L is consistent with an increased amount of $\gamma \rightarrow \alpha'$ transformation in the presence of the magnetic field. It appears that both an increase in the amount of prior cold-work and an increase in the amount of carbon plus nitrogen in this material will cause the $\gamma \rightarrow \alpha'$ transformation to effect more work-hardening which does not contribute to the elongation of the material. Preliminary data show that the effect of an 8 T magnetic field on the fatigue crack growth rates of 304L and 304LN at 77 K is barely measurable, but a decrease in the fatigue crack growth rate is suggested.

In the opinion of the authors, a magnetic field of 8 T is a practical threshold for observing magneto-mechanical effects in 304 stainless steels. If 304 stainless steels are proposed as structural materials for superconducting magnets with higher operating fields, magneto-mechanical effects may become a subject of more engineering concern.

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REFERENCES

1. F. Lecroisey and A. Pineau, Metall. Trans. 3A:387 (1972).
2. R. L. Tobler, D. T. Read, and R. P. Reed, Materials Studies for Magnetic Fusion Energy at Low Temperatures-IV:37 NBSIR 81-1645 (April, 1981).
3. M. A. Krivoglaz and V. D. Sadovskii, Fiz. Met. Metalloved. 18:502 (1964).
4. Ye. A. Fokina, L. V. Smirnov, and V. D. Sadovskii, Fiz. metal. metalloved. 19:772 (1965), and succeeding papers in this journal.
5. R. P. Reed et. al., Proc. ICMC Conf., Kobe, Butterworths, Surrey, England (1982).
6. B. Fultz, G. M. Chang, and J. W. Morris, Jr. in Austenite Steels at Low Temperatures, R. P. Reed and T. Horiuchi, eds., Plenum, New York (1983).
7. J. R. C. Guimarães and R. J. De Angelis, Mater. Sci. Eng. 13:109 (1974), and J. R. C. Guimarães and R. J. De Angelis, Met. Trans. 4A:2379 (1973).
8. R. L. Tobler and R. P. Reed, Materials Studies for Magnetic Fusion Energy at Low Temperatures-IV:101 NBSIR 81-1645 (April, 1981).

9. N. E. Frost, K. J. Marsh, and L. P. Pook, Metal Fatigue, Chapter 5, Clarendon Press, Oxford, England (1974).
10. A. G. Pineau and R. M. Pelloux, Metall. Trans. 5A:1103 (1974).
11. E. Hornbogen, Acta. Met. 26:147 (1978).
12. R. M. McMeeking and A. G. Evans, J. Am. Ceram. Soc. 65:242 (1982).
13. A. Saxena and S. J. Hudak, Jr., Int. J. Fract. 14:453 (1978).
14. G. C. Sih, Handbook of Stress Intensity Factors, Institute of Fracture and Solid Mechanics, Lehigh Univ., Bethlehem, PA (1973).

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