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Publication Date

1969-10-01

Submitted to Journal of Materials Science

UCRL-19080
Preprint

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October 1969

AEC Contract No. W-7405-eng-48

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ON THE RESISTANCE OF TRIP STEEL TO HYDROGEN EMBRITTLEMENT

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INTRODUCTION

TRIP steels (TRansformation Induced Plasticity), recently developed by Zackay, et al.,^(1,2) demonstrate a useful combination of high strength, high ductility and good fracture toughness. By a careful balance of composition and thermomechanical treatment, TRIP steel is metastable austenite at room temperature with its M_D slightly above and its M_S well below room temperature. The high yield strength of the austenite is produced by a combination of solid solution strengthening, precipitation strengthening, and work hardening. The high ductility and toughness result from the austenite to martensite transformation occurring during plastic deformation at regions of potential necking as in uniaxial behavior or at a crack tip as in crack propagation behavior.

Hydrogen embrittlement is an especially serious problem to high-strength martensitic steels because both the time required for failure and the minimum stress for failure usually decrease as the strength is increased. TRIP steels are of interest with respect to this phenomenon since, in the austenitic state, they should be relatively immune to hydrogen embrittlement, whereas, after the strain-induced transformation, adverse effects might be expected. Therefore, an investigation to determine under what conditions TRIP steel might be susceptible to hydrogen embrittlement was undertaken.

PROCEDURE

A representative high-strength TRIP steel and a tempered martensitic steel (to evaluate the hydrogen charging procedure) were selected for testing. The compositions and heat treatments of these steels are shown in Table 1. From sheets of these steels, single-edge-notch (SEN) specimens, approximately 2X1X0.0 in., were machined. The eccentricity (distance between the center of the pin holes and the centerline of the specimen) was about 0.16 in. The method of calculating the stress intensity factor employed throughout this investigation is based upon the work of Srawley, et al.⁽³⁻⁵⁾ The stress intensity factor at which crack propagation (at the onset of fracture) becomes unstable is the critical stress intensity factor K_{Cr} . The K_{Cr} values of the SEN specimens used in this investigation were higher than the plane-strain critical stress intensity factor K_{IC} because the thickness of the SEN specimens used was not large enough to ensure plane strain conditions throughout the fracture region. Because all the SEN specimens in this investigation had approximately the same dimensions, however, comparisons of their K_{Cr} values relative to each other are justified.

Prior to testing, several operations were performed upon these specimens.

1. A fatigue crack, about 0.04 in. long, was produced at root of the notch by cycling the specimens between an upper stress intensity factor of 28 ksi-in^{1/2} and a lower stress intensity factor of 5 ksi-in^{1/2} for about 10,000 cycles. The length of the notch plus the fatigue crack measured about 0.33 in. from the edge.

2. In most of the specimens, a relatively large C-shaped plastic zone ahead of the tip of the fatigue crack was produced by applying for about 30 seconds a stress corresponding to 60, 80, or 90% of K_{Cr} . Since this plastic zone contained a large amount of strain-induced martensite, the effect of martensite being present at the crack tip upon the tests for hydrogen embrittlement could be evaluated.

3. Most specimens were cathodically charged with hydrogen in a 4% by weight H_2SO_4 electrolyte containing a phosphorus poison (white phosphorous dissolved in CS_2). The charging conditions (time and current density) were varied to adjust the hydrogen concentration in the specimens.

4. To obtain a homogeneous concentration of hydrogen throughout the specimen, five of the specimens were cadmium plated for two hours at a current density of 0.15 amps/in² depositing a 2-mil layer of cadmium and subsequently baked at 300°F. The baking time was varied as another means of adjusting the hydrogen concentration.

To test the charged specimens for hydrogen embrittlement, they were held at a constant load (e.g. 80% K_{Cr}) for about eight hours. The load was maintained constant by periodically readjusting the displacement of the Instron machine. If the specimen failed, the time under load up to failure was noted. If the specimen neither failed nor displayed evidence of significant crack growth after about eight hours under load, the load was removed and the microstructure around the crack tip was examined metallographically.

RESULTS AND DISCUSSION

The conditions and results of the tests for the various specimens are given in Table 2. In ten tests on cathodically charged TRIP steel

specimens with and without the plastic zone prior to charging, there were no instances of hydrogen-induced, delayed brittle failures. In fact, even after eight hours at a stress intensity level of $115 \text{ ksi-in}^{1/2}$ ($80\% K_{Cr}$), the amount of crack growth was nil. In comparison, the charged martensitic steel specimen failed brittly at one pin-loading hole after 34 minutes at a load corresponding to an initial stress intensity level of $42 \text{ ksi-in}^{1/2}$, 45% of the uncharged K_{Cr} value. At failure, the crack had grown 0.18 in., corresponding to a stress intensity level of $74 \text{ ksi-in}^{1/2}$.

In all the TRIP steel specimens, an etchant of 5 gm cupric chloride, 100 ml HCl, 100 ml H_2O , and 200 ml methanol revealed martensite plates within the plastic zone. The plastic zone near the surface, as shown in Fig. 1A, was about three times larger than the plastic zone at the mid-thickness, as shown in Fig. 1B. This is what might be expected if the surface were in a relatively plane-stress state and the mid-thickness were in a relatively plane-strain state. Both calculation and observation of the surface plastic zone show that it is slightly smaller than the thickness and so the above is a reasonable interpretation to make.

Tested TRIP steel specimens that had no homogenization treatment revealed a crack network near the specimen surface and within the plastic zone, as shown in Fig. 2A. This C-shaped network extended about 0.05 in. ahead of the fatigue crack. After etching, it was evident that the cracks followed along paths of martensite plates, as seen in Fig. 2B. This suggests that while the martensite was hydrogen embrittled, the surrounding austenite was not. The fact that in many alloys of the transition metals the body-centered cubic structure but not the face-

centered cubic structure is susceptible to hydrogen embrittlement⁽⁶⁾ is consistent with this observation. Around the mid-thickness of the unplated specimens, the amount of cracking occurring in the hydrogen-bearing martensite was far less than at the surface and the total amount of crack extension was on the order of 0.01 in., as shown in Fig. 2C. This smaller amount of cracking at the center may be due either to a smaller amount of martensite or to a lower level of hydrogen concentration. Considering the lack of significant crack propagation passed the original plastic zone in eight hours, the freshly strain-induced martensite apparently is not embrittled sufficiently by the residual hydrogen to extend the crack further.

Instead of the relatively large surface crack networks of the unhomogenized specimens (Fig. 2A), only slight crack extensions of less than 0.01 in. (Fig. 1A) were observed on homogenized specimens. Thus, the crack network of the unhomogenized specimen is probably due to the embrittling effect of the relatively high concentration of hydrogen remaining at the surface of the specimen after charging. As it was intended, the homogenization treatment apparently lowers this relatively high hydrogen concentration, greatly reducing further cracking in this region.

In summary, TRIP steel could not be made to fail in a delayed brittle manner subsequent to cathodically charging with hydrogen. Relatively high concentrations of hydrogen at the specimen surface did embrittle pre-existing martensite that had been introduced at the crack tip by plastic deformation. However, either the inability of existing hydrogen to migrate across austenite-martensite interfaces or the lack of

sufficient hydrogen build-up in the newly created martensite at the fresh crack tip prevented further crack growth.

ACKNOWLEDGEMENT

This work was supported by the United States Atomic Energy Commission.

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Table 1. Compositions and Heat Treatments of the Steels Tested

Steel	Composition (wt.%)						Heat treatment
	C	Mn	Si	Cr	Ni	Other	
TRIP	0.25	2.1	2.0	8.8	8.3	3.7 Mo	Solution annealed at 1175°C for 1 hr., ice brine quenched, and 80% warm rolled at 425°C.
Martensite	0.46	0.4	0.1	0.5	12.0	4.0 Co	Austenitized at 1000°C for 1/2 hr., oil quenched, and tempered at 450°C for 2 hr + 2 hr.

Table II. Conditions and Results of Hydrogen Embrittlement Tests

Specimen	t_o (mils)	K_{Cr} (ksi-in ^{1/2})	Plastic Zone Formed at This % K_{Cr}	Hydrogen Charging Conditions		Cd Plated	Bake Time (min)	Dead Weight Holding Conditions		
				Time (min)	Current Density (ma/in ²)			Time (min)	Load (lbs)	% K_{Cr}
1	102.6	139	-	50	125	No	0	60	3300	70
2	101.3	145	-	50	125	No	0	240	4100	80
3	103.0	144	-	-	-	No	0	-	-	-
4	102.2	148	80	-	-	No	0	-	-	-
5	102.6	-	80	50	125	No	0	210	4150	80
TRIP Steel	6	97.7	-	50	125	No	0	480	3700	76
7	95.2	-	60	50	125	No	0	420	3870	80
Y.S. $\approx$ 220 ksi	10	99.0	-	120	500	Yes	0	420	3920	80
11	94.0	-	80	120	500	Yes	2520	450	3870	80
12	95.8	-	80	120	500	Yes	2520	420	3900	80
13	100.4	-	80	120	500	Yes	50	480	3960	80
14	95.2	-	80	120	500	Yes	120	510	3900	80
Marten- sitic Steel	1	75.7	48 ^a	-	50	125	No	0	-	-
2	75.5	74 ^b	-	50	125	No	0	34 ^c	1050	45
3	71.8	96	-	-	-	No	0	-	-	-
Y.S. $\approx$ 225 ksi	4	74.2	89	-	-	No	0	-	-	-

^a Based upon the original crack length (notch plus fatigue crack). Slow crack growth confined to the region near the surface made the calculation of the true K_{Cr} impossible.

^b Stress intensity after the crack grew 0.18 in., at which point failure occurred at one pin-loading hole.

^c Time to failure at one pin-loading hole.

FIGURE CAPTIONS

Figure 1. Plastic zone containing martensite at fatigue crack tip of hydrogenated, tested, and etched TRIP steel specimen #12; (A) near surface and (B) near mid-thickness.

Figure 2. Crack network at fatigue crack tip of hydrogenated and tested TRIP steel specimen #5; (A) near surface unetched, (B) near surface etched (revealing martensite), and (C) near mid-thickness unetched.

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Fig. 1

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Fig. 2

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