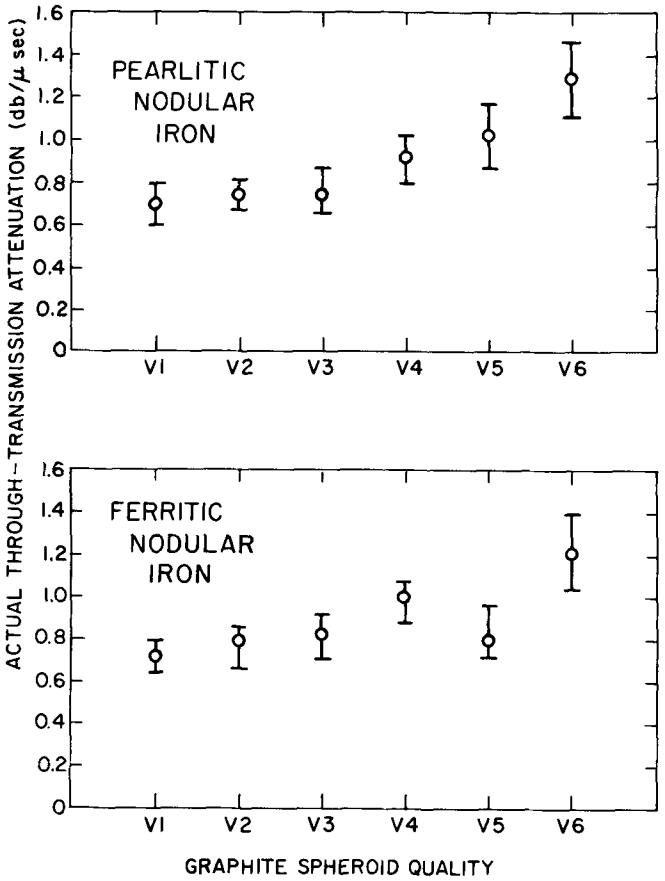


Fig. 3—Actual through-transmission ultrasonic attenuation as a function of spheroid quality.

range from V1 for spheres to V6 for stubby flakes, Fig. 1.

Fig. 2 shows that as spherodicity degenerates, relative attenuation increases for both the ferritic and pearlitic matrices. This is an expected result since a greater scattering of the acoustic wave should occur for irregularly shaped particles. In this frequency range, the scattering loss appears to be independent of matrix. The attenuation is rather insensitive to spheroid quality from V2 to V4, since microstructural changes are fairly small in this range.

It appears that surface ultrasonic waves do provide a reliable nondestructive test of microstructure. To substantiate this conclusion, the ultrasonic attenuation of the samples has also been obtained with the more familiar through-transmission technique described earlier. An Automation Industries type SFZ, 2.25 MHz transducer is bonded to the samples with the same couplants used in the surface wave tests. The plots of actual attenuation against spheroid quality are given in Fig. 3, where it is again evident that poorer spheroids give rise to higher damping losses.

If a reliable set of microstructural standards can be developed for a material of interest, there is little doubt that surface ultrasonics provide a usable technique for microstructure determination.

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# Comment on the Strength Differential (SD) Effect in TD Nickel

B. A. WILCOX, A. H. CLAUER, AND G. LUETJERING

WE wish to emphasize an important point brought out in a recent paper by Olsen and Ansell[1] concerning the strength differential (SD) effect in two-phase alloys, *i.e.*, the observation that compression yield strengths are higher than tension yield strengths. The point of interest is embodied in the last sentence of their Discussion:

> ". . . one might expect variations in the SD effect in samples from different heats of a given alloy or from samples of the same heat that have been processed differently."

The purpose of this communication is to provide evidence for this suspected lack of reproducibility of the SD effect in TD nickel (Ni-2 pct $ThO_2$) bar.

Tension and compression tests were conducted at two laboratories by different investigators and the results are listed in Table I, together with the original results reported by Olsen and Ansell. The strain rates used were about $10^{-3}$ to $10^{-4}$ $sec^{-1}$, and were the same for compression and tension tests on a given material. Each of the materials A through E was from a different heat of bar stock. It is seen that materials A and D each have a compression yield stress significantly higher than the tensile yield stress. However, for materials B and C, there is essentially no difference in the compression and tension yield strengths, and the tensile yield strength of material E is higher than the compressive yield strength. In the latter case the same specimen configuration was used for tension and compression specimens to avoid any possible effects of specimen design on the yield strength.

We thus conclude that the SD effect in TD nickel bar is not reproducible. This may be the result of different levels of impurities from one material to another, as suggested by Olsen and Ansell. If the particle/matrix inferfacial strength influences the SD effect, then different impurities segregated to the interfaces could affect

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Table I. Room Temperature Tension and Compression Tests on TD Nickel Bar to Test the Reproducibility of the SD Effect

| Material* | Tension 0.2 pct Offset Yield Strength, psi | Compression 0.2 pct Offset Yield Strength, psi | Reference |
|---|---|---|---|
| A) $\frac{1}{4}$ in. diam bar | 45,000 | 65,000 | Olsen and Ansell[1] |
| B) $\frac{1}{2}$ in. diam bar | 55,600 | 54,000 | Work at Battelle-Columbus |
| C) $\frac{1}{4}$ in. diam bar | 62,000 | 60,000 | Work at Battelle-Columbus |
| D) $\frac{1}{4}$ in. diam bar | 72,000 | 93,500 | Work at Battelle-Columbus |
| E) $\frac{5}{8}$ in. diam bar | 82,000 | 76,000 | Work at Allied Chemical |

*The difference in magnitude of the strength levels from material to material probably arises from the fact that somewhat different thermomechanical processing treatments were used for the various heats.

the tension and compression yielding behavior. It is clear that additional research is necessary to determine the cause of this irreproducibility.

1. R. J. Olsen and G. S. Ansell: *Trans. ASM*, 1969, vol. 62, p. 711.

# The Influence of Titanium Additions on the Fracture Behavior of Iron

H. J. RACK

RELLICK and McMahon[1] have recently indicated that titanium is not as effective as aluminum in preventing low energy intergranular brittleness in iron. However, their examination was limited to rather low titanium concentrations (0.12 wt pct). Jolly[2] has also indicated that as the titanium concentration increases a minimum in the transition temperature, as measured by the Charpy impact energy, occurs. Unfortunately, no grain sizes were reported by the latter investigator, and since this is an important factor in establishing the low temperature properties of iron, the exact influence of titanium in this regard is still in doubt. This work was undertaken to clarify the role titanium additions play in the deformation and fracture behavior of ferrous alloys.

The compositions of the alloys examined are given in Table I, together with a record of their thermomechanical history prior to testing. It should be noted that all of the alloys possessed the same grain size (~0.150 mm). Tensile testing was carried out on electropolished cylindrical specimens 0.125 in. diam by 0.75 in. gage section at a nominal strain rate of $1.67 \times 10^{-3}$ $sec^{-1}$ between 25°C and –195°C.[3]

At –195°C, all the alloys, with the exception of the Fe-3.16Ti alloy, deformed by a mixture of slip and twinning while no evidence for twinning, unassociated with the main fracture, was found in the latter alloy. The amount of twinning sharply decreased with increasing test temperature; at –120°C only the Fe-0.003C exhibited any twin deformation. No twinning was found in any of the alloys at a test temperature of –60°C or higher, in agreement with the previous observations

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Table I. Materials

| Designation | Chemistry Weight Percent C | O | N | Ti | Heat Treatment Schedule* Homogenization† Temp., °C | Time, hr | Recrystallization‡ Prior Strain | Temp., °C | Time, hr |
|---|---|---|---|---|---|---|---|---|---|
| 1 | 0.003 | 0.006 | 0.0008 | – | 850 | 72 | 0.45 | 700 | 3.0 |
| 2 | 0.004 | 0.004 | 0.0008 | 0.12 | 850 | 72 | 0.67 | 700 | 2.5 |
| 3 | 0.001 | 0.005 | 0.001 | 0.49 | 850 | 72 | 0.45 | 800 | 2.5 |
| 4 | 0.004 | 0.005 | 0.003 | 1.08 | 850 | 72 | 0.45 | 800 | 1.0 |
| 5 | 0.005 | 0.007 | 0.003 | 3.16 | 1000 | 48 | 0.45 | 900 | 1.75 |

*All heat treatments were conducted under a dynamic vacuum of $1 \times 10^{-5}$ mm Hg.
†Air cooled.
‡Furnace cooled.

of McMahon and Cohen[4] on a ferrite of similar composition.

The fracture mode is also a function of the titanium content and test temperature. Failure of the Fe-0.003C at –195°C is initiated intergranularly, *i.e.*, the observed microcracks lie along prior boundaries. Their characteristic morphology is identical to that observed by McMahon[5] and Gilbert *et al.*;[6] microcracks apparently being nucleated by twins at grain boundaries. Propagation of the microcracks in Fe-0.003C at –195°C was intergranular. Interestingly, numerous intergranular microcracks were observed in this material indicating that, at least in the present instance, crack propagation rather than crack nucleation, was the critical event leading to ultimate failure.

The microcracks observed in the Fe-Ti alloys were transgranular in character, Fig. 2. Those in the solid-solution alloys appeared to be initiated at a twin-grain boundary interaction. In addition, the number of microcracks in these alloys decreased with increasing test temperature; none were observed at –120°C. Floreen *et al.*[7] recently showed that cleavage initiation in Fe-Ni alloys originates from cracked Ti(C,N) particles located at grain boundaries. These observations suggest that, at –195°C, twinning and microcrack formation in the solid-solution Fe-Ti alloys are intimately related and that the Ti(C,N) particles lying on grain bound-