to make use of his unpublished results. One of us (L. R. O. S.) is indebted to the Chief Scientist, Ministry of Supply, for permission to contribute to this communication.

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Change of Elastic Constants in a Superconductor

IT is known[1] that there must be a small difference between the elastic constants of a superconductor in the superconducting and the normal states. The difference expected is of the order of 1 part in 10^5, but no actual measurement has so far been reported. We have now used a statical method to measure the change in the modulus of rigidity of a chemically pure polycrystalline tin wire when superconductivity is destroyed by a longitudinal magnetic field.

The specimen was placed vertically in liquid helium, and fixed at the lower end. The upper end was connected by a rigid rod to a phosphor bronze wire at room temperature. A twist applied to the top of the bronze wire then distributed itself between the two wires according to their torsional constants. This equilibrium distribution of strain changed on destruction of superconductivity, and the change was observed by means of an optical lever, as described by Jones[2]. Too much torsion of the tin caused excessive creep, so that the deflexion changes actually measured were less than 5×10^{-7} radians.

The results of several runs are plotted in Fig. 1 in the form $(G_n - G_s)/G_n$, where G_n and G_s are the rigidities in the normal and superconducting states respectively.

Absolute calibrations are only available for some of the runs. Results for the uncalibrated runs have been normalized to fit the calibrated runs at 2° K., and they are thus of help in establishing the dependence of the effect on temperature. The full curve represents the function $A\{1 - (T/T_c)^4\}$, where T_c is the transition temperature, and the constant A has been chosen for agreement with the observations at 2° K. It can be deduced from work on the change of critical field of tin with stress[3] that the difference in volume between the superconducting and normal states may be expected to be dependent on temperature in this way. This is also in agreement with the work of Lasarew and Sudovstov[4]. It can easily be shown that, in any event for a lattice with central forces, the difference in the shear moduli of the two states will depend on temperature in the same way as the volume difference.

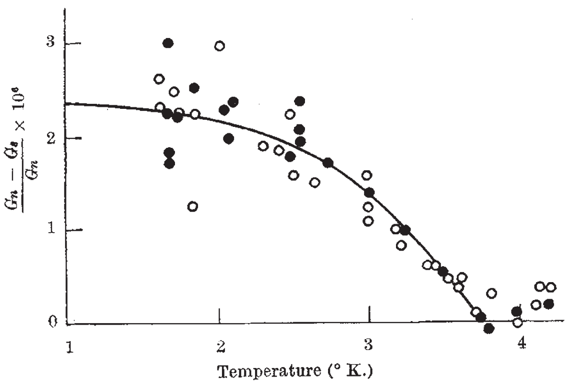

Fig. 1. Temperature variation of $(G_n - G_s)/G_n$ for tin. ●, Runs with direct calibration; ○, uncalibrated runs normalized to fit at 2° K. The full curve represents the function $A\{1 - (T/T_c)^4\}$ with A chosen to fit the results at 2° K.

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A Phase Diagram for 1 per cent Carbon–Iron Alloys containing up to 16 per cent Nickel

DURING the course of an investigation into the austenite stabilization phenomena of iron–nickel alloys containing 1 per cent carbon, it became necessary to obtain some knowledge of the limiting solubility of the carbide in the austenite. There appeared to be no specific information available concerning the phase relationships of such alloys and, due to the promotion of graphitization by high nickel contents[1], a great deal of uncertainty attends the nature of the iron–carbon–nickel ternary system.

Casts weighing 18 lb. were made in a high-frequency furnace; the ingots, the analyses of which are given in Table 1, were forged to $\frac{5}{16}$-in. diameter rods which were normalized from 1,050° C. Specimens were machined, nickel-plated and tempered for 24 hr. at 600° C. before testing. The progress of transformation was followed dilatometrically, using a dilatometer previously described[2], and temperatures measured by means of platinum/platinum–rhodium thermocouples peened centrally into the specimens. Readings were taken at 5 deg. C. intervals with a heating-rate of 1 deg. C. per min.

The results, presented in Fig. 1, agree very well with the 1 per cent carbon section which may be

Table 1. PRINCIPAL ALLOYING ELEMENTS OF THE NICKEL STEELS EXAMINED

No.	% C	% Ni	% Si	% Mn
1	1·02	0·03	0·23	0·32
2	1·04	0·66	0·19	0·36
3	0·98	1·03	0·13	0·39
4	1·02	2·01	0·15	0·31
5	1·02	3·06	0·21	0·34
6	1·00	4·09	0·13	0·32
7	1·00	5·06	0·26	0·25
8	0·93	7·10	0·30	0·19
9	1·04	9·9	0·22	0·26
10	0·99	15·8	0·47	0·42