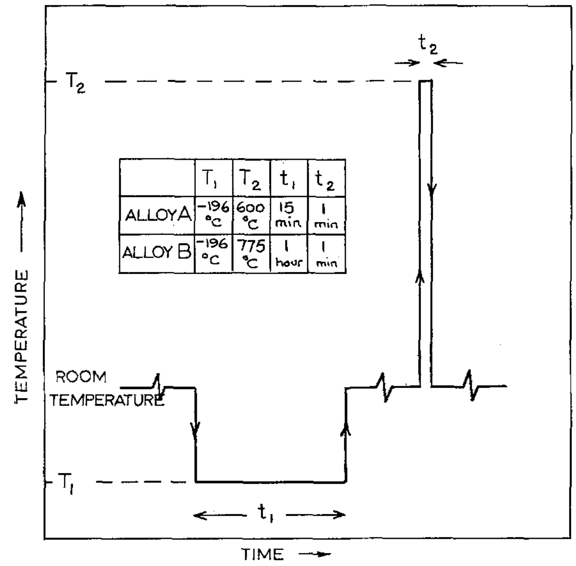
Cyclic phase transformations in certain stainless steels

The reversion of martensite (α') to austenite (γ) in steels can strengthen the γ and also, to a lesser extent, the α' produced by cooling the "reversed γ". Detailed studies of the effect of repeated cycles of transformation ($\gamma \rightarrow \alpha' \rightarrow \gamma$) using iron–nickel based alloys containing large amounts of acicular martensite have shown that the main strengthening is achieved by the first cycle [1-5]. Less information is available for chromium–nickel semi-austenitic steels, but the dominance of the first cycle in strengthening has been shown [6-8] together with stabilizing effects of one or more reversion cycles on subsequent α' formation [8-10].

The investigation reported here has explored cumulative strengthening and stabilization in two semi-austenitic stainless steels of the following compositions (wt%): alloy A; Fe, 16.1 Cr, 11.7 Ni, ~ 0.003 C, 0.009 N, and alloy B; Fe, 14.9 Cr, 8.7 Ni, 2.05 Mo, 0.009 C, 0.01 N. Solution treatments consisted of 20 min at 1050°C for alloy A and 1 h at 1250°C for alloy B, followed by water quenching in each case; the M_s temperatures were − 50 and − 40°C respectively. Lath type martensite, α', together with ϵ and stacking faults, formed on cooling to − 196°C [11]. The transformation cycles, involving refrigeration and reversion are illustrated in Fig. 1. The A_f temperatures were ~ 600 and 775°C respectively [12]. Other aspects of reversion phenomena, together with details of the experimental procedure have been previously reported for these alloys [12-14].



	T1	T2	t1	t2
ALLOY A	−196 °C	600 °C	15 min	1 min
ALLOY B	−196 °C	775 °C	1 hour	1 min

Figure 1 Diagrammatic representation of one transformation cycle.

The changes in hardness and α' content during five transformation cycles are shown in Fig. 2. In alloy A (Fig. 2a) the hardness of the austenite was doubled by five cycles, about a half of this increment being achieved in one cycle. The hardness after cooling to − 196°C increased slightly during the first cycle, and remained unchanged thereafter. The amount of α' formed on refrigeration decreased progressively with increasing numbers of cycles, and nearly complete stabilization resulted after five cycles. In alloy B (Fig. 2b) the effect of one cycle on hardening of the austenite was less than in alloy A, but there was a substantial progressive increase with continued cycling. This observation appears to be contrary to published data [6] for a similar steel (15.0 Cr, 7.8 Ni, 1.7 Mo, 0.04 C) where very little strengthening of the austenite was achieved after the first cycle. This difference can possibly be explained on the basis of the faster heating and shorter reversion times employed in the present study; a further possible contributing factor is the smaller martensite content of the steels reported here. The hardness of alloy B after cooling to − 196°C showed a progressive increase up to two cyles with little subsequent change. Stabilization of the martensite transformation, $\gamma \rightarrow \alpha'$, occurred to a smaller extent that in alloy A. In neither alloy was α' detected at room temperature after the reversion treatments. In alloy B, containing 0.09% C, no carbide precipitation was detected by transmission electron microscopy after one complete transformation cycle.

Structural observations to determine the nucleation sites of α' formed from reversed γ were not made on samples subjected to repeated transformation cycles. However, in a study of a single crystal of Fe–16 Cr–12 Ni alloy, Breedis and Robertson [15], found that the α' formed in the same sites on requenching after reversion. Structural features of "second generation" α' have also been investigated in a similar alloy subjected to reversion for 2 min at 600°C [10]; a stabilization effect similar to that shown in Fig. 2a was found, associated with a decrease in M_s. The evidence obtained suggested that most of the α', formed in a series of cycles, lay within the original "bands" ("sheets") of first generation

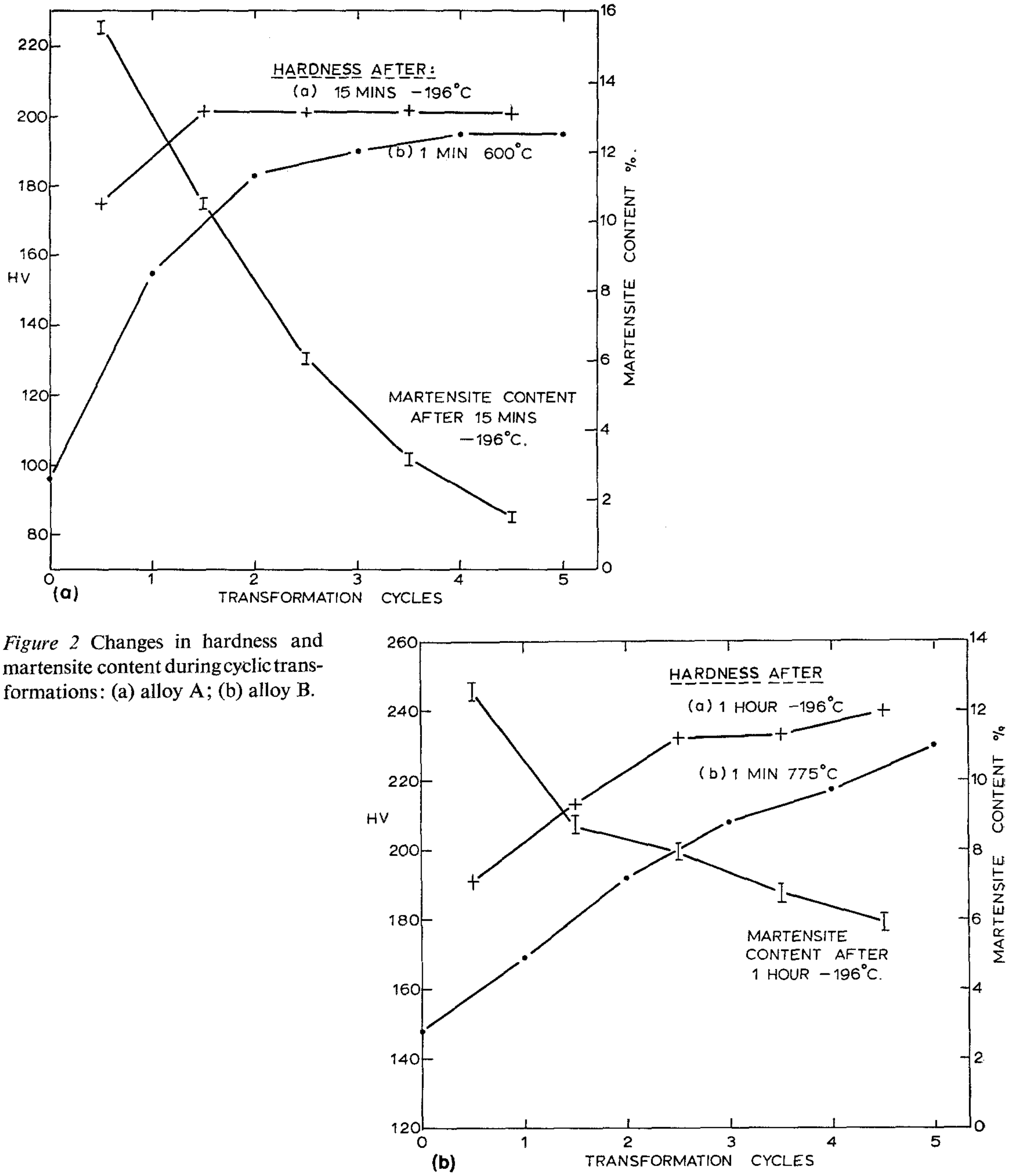


Figure 2 Changes in hardness and martensite content during cyclic transformations: (a) alloy A; (b) alloy B.

transformation products. The regions of ϵ and stacking faults in these bands are thought to provide nucleation sites for α' laths, but lath growth is restricted by the defect structure of the reversed γ causing stabilization. The progressive stabilization observed with successive cycles is attributable to the increase in the volume of reversed γ within the bands. The smaller degree of stabilization in alloy B may result from carbide precipitation depleting the reversed γ in carbon and hence raising M_S [16]; another possible factor is the higher reversion temperature giving a greater degree of recovery in the reversed γ [13].

In alloy A, strengthening of the austenite is limited by the virtually complete stabilization which occurs; this may be attributed to the original transformation "bands" becoming almost fully transformed to reversed γ [10]. The smaller strengthening in alloy B may be due to the higher reversion temperature, but the smaller stabilization effect allows progressive strengthening to proceed throughout the series of transformation cycles studied.

In general, the strengthening of austenitic stainless steels by the reverse martensite transformation appears to be limited by three main factors. Firstly, little deformation of the retained austenite is produced by the direct and reverse martensite transformations and hence, unlike iron–nickel alloys, significant strengthening occurs only in those regions which actually transform. Secondly, only small amounts of martensite are produced in these steels by cooling below the M_S temperature, although in some steels prolonged isothermal holding below M_S can produce substantial amounts of martensite [6]. Finally, the cumulative strengthening action of repeated transformation cycles is restricted by the development of austenite stabilization.

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Variation of MC carbide geometry with local solidification time in cast Inconel 713 C alloy

In as-cast polycrystalline nickel-base superalloys, MC-type carbides contribute to high temperature grain-boundary strengthening. In these carbides, M is tantalum, niobium, zirconium, titanium or molybdenum [1-2]. Among them TaC and NbC are very stable, whereas multicomponent carbides are metastable with a tendency towards dissociation or dissolution during long service at high temperature. In the as-cast alloy, no orientation relationship has been established between MC carbide and matrix [3-5].

In unidirectionally solidified monocrystalline or polycrystalline superalloys, the MC carbide morphology has been described as dendritic and unconnected near the chill, with tendencies to form an interconnected, interdendritic network away from the chill [5, 6]. An increase in particle size was observed with distance from the chill [5, 6].

The purpose of the present investigation was to establish quantitatively the dependence of MC carbide geometry on local solidification time [7] in unidirectionally solidified Inconel 713 C alloy.

The alloy was vacuum-melted and solidified in moulds (4.50 cm × 4.50 cm × 15 cm) made of "fiberchrome", an insulating material, with open bottom placed against a water-cooled copper chill [8]. All surfaces of the mould cavity, runner and risers were coated with "zircon" wash. Cooling curves were recorded using Pt/Pt–10% Rh thermocouples inserted horizontally along the height of the ingot at 1, 2, 3