

The Strengthening of Ti-55 at. % Ni by Formation of a Ti_2Ni_3 Intermetallic Phase

Normally ductile and impact resistant β (bcc)-TiNi is known to become hardened and embrittled when quenched from 1000°C and aged at 625°C [1, 2]. The subsequent ageing at 625°C results in the formation of the newly found ϵ - Ti_2Ni_3 phase [3] (DO_{24} structure with stacking variants). The specific role played by such precipitation in the hardening mechanism has not been established.

It has been reported [3, 4] that 1, the equilibrium β -TiNi structure has a narrow homogeneity range below 500°C; 2, a structure of approximate composition Ti_2Ni_3 forms peritectoidally at 625°C and 3, a martensitic transformation occurs at 60°C in the equilibrium β -TiNi. The formation of the TiNi, Ti_2Ni_3 intermetallic phases at the corresponding equilibrium temperatures is shown in fig. 1. These observations by Wasilewski [3] have been duplicated in the present investigation of a Ti-55 at. % Ni alloy. The effect of such ϵ formation on the hardness and tensile properties

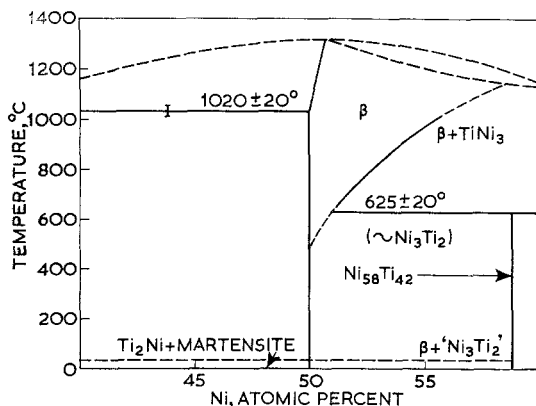


Figure 1 Equilibrium phase diagram of the Ti-Ni system

of Ti-55Ni is summarised in table I where all specimens were heat-treated as indicated.

The substantial increase in strength associated with the Ti_2Ni_3 reaction is not accompanied by any degree of plasticity prior to failure which occurs at maximum load. The enhancement in strength with some ductility is possible by a two-stage heat-treatment (case G) where TiNi₃ is formed at a higher temperature prior to annealing at 500°C (1000°C, 150 h at 650°C, quenched plus 300 h at 500°C, quenched). The presence of

TABLE I. Effect of Ti_2Ni_3 on tensile properties of Ti-55Ni.

Treatment	Tensile strength ksi	Uniform elongation %*	Hardness Rc
A 30 h at 1000°C, quenched	180	>1	61
B 1000°C, 70 h at 700°C, quenched	121	4-5	47
C 1000°C, 70 h at 650°C, quenched	130	1.3	48
D 1000°C, 300 h at 400°C, quenched	161	>1	60
E 1000°C, 300 h at 500°C, quenched	158	>1	59
F 1000°C, 300 h at 625°C, quenched	155	>1	59
G 1000°C, 150 h at 650°C, 300 h at 500°C, quenched†	170	10-12	57

†See fig. 2 for microstructure and fig. 3 for fractography. *Gauge length for elongation measurements was 3.75 cm.

TiNi₃ prior to ageing at 500°C leads to a uniform dispersion of Ti_2Ni_3 throughout the β matrix as shown in fig. 2. The formation of TiNi₃ is reported to initiate at grain boundaries.

Fractographic examination of tensile specimens annealed at temperatures between 300 and 625°C to produce Ti_2Ni_3 revealed microscopic ductility while failure occurred with no macroscopic plastic strain. The regions denoted by A in fig. 3 exhibit the ductile deformation structure.

The brittleness of alloys with the Ti_2Ni_3 phase can be associated with either crack initiation in

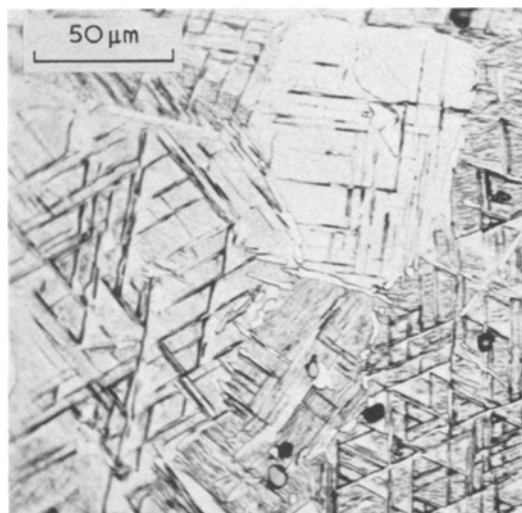


Figure 2 The effect of anneal treatment on the structure Ti-55-Ni. The specimen was annealed 30 h at 1000°C plus 150°C, and then water quenched.

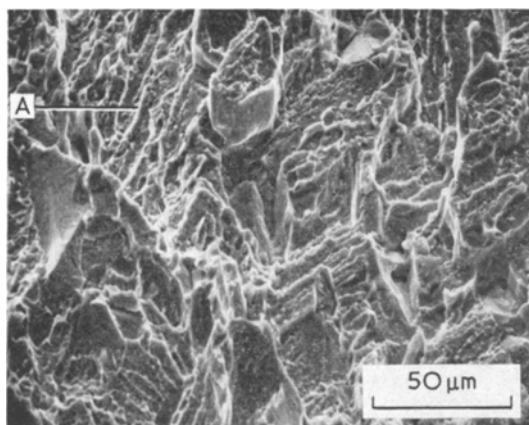


Figure 3 Scanning electron micrograph of the fracture after heat-treatment G to form nearly 50 vol % Ti_2Ni_3 .

the inherently brittle phase or the obstruction of slip in the β -phase. While the mechanical behaviour of the Ti_2Ni_3 phase is unknown, it is reasonable to expect that plastic deformation of Ti_2Ni_3 occurs with difficulty. Since failure in β -TiNi is largely of a ductile nature, the operative mechanism is then most likely associated with crack propagation in the grain boundaries and the barrier to slip imposed by the Ti_2Ni_3 phase. In discussing the second phase barrier to slip, Silcock [5] has shown that if three of the $\langle 111 \rangle$ slip directions in a given bcc matrix are not parallel with the likely slip-directions in a given second phase, then rupture of strong bonds are required for the remaining slip-direction. Ashby [6] has shown that the resultant increase in work hardening associated with the interaction of primary glide-dislocations with the forest-dislocations created about the second phase can be expressed as:

$$\tau - \tau_0 = KG_m \left[\frac{bV\gamma\rho}{d} \right]^{1/2}$$

where G_m is the shear modulus of the matrix, V is the volume fraction of the second phase, $\gamma\rho$ is the plastic shear strain on the primary slip-system, and d is an effective diameter of the second phase particles. The constant K has been

found experimentally to be near 0.2. By equating the shear stress difference $\tau - \tau_0$ to the difference in tensile stress between fracture and yielding of polycrystalline β -TiNi, $(\sigma - \sigma_0)/m$ where m is 2.7 for bcc metals) then the plastic shear strain required on the primary system is in the vicinity of macroscopic yielding (10^{-2} to 10^{-3}). Thus, the apparent brittle behaviour of ϵ -containing alloys is related both to the hard Ti_2Ni_3 phase which increases the rate of work hardening of the β -phase and to crack propagation along brittle grain boundaries. Failure occurs by crack propagation in the grain boundaries and some type of "interphase" fracture which is quasi-ductile on a microscopic scale. It is important to note that 85% of the fractured surfaces were intergranular and 15% were transgranular.

The attainment of high strengths in Ti-55Ni at room temperature with the retention of some ductility should be possible by controlling the volume fraction of ϵ by appropriate heat-treatment, or through alloying. In general, good tensile properties may be obtained if TiNi_3 can be dispersed throughout the β -matrix. The value of the duplex heat-treatment is to produce a fine dispersion of TiNi_3 which subsequently transforms to Ti_2Ni_3 . Thus, it is possible to attain attractive tensile properties through such duplex heat-treatment.

References

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Direct measurements of the reverse stress asymmetry in the yielding of anisotropic polypropylene

In our previous work we have recognized an asymmetry between the tensile and compressive

yield behaviour of oriented polymers [1]. Most of our previous measurements have been made either in tension or simple shear on drawn polymer sheets and similar results have been obtained by other workers [2-4]. The measurements provide only an indirect measure of the