Ti-2007 Science and Technology, edited by M. Ninomi, S. Akiyama, M. Ikeda, M. Hagiwara, K. Maruyama
The Japan Institute of Metals (2007)

Phase Transformations in Ti-5Al-5Mo-5V-3Cr-0.5Fe

James D. Cotton[1], Rodney R. Boyer[1], Robert D. Briggs[1], Roy G. Baggerly[1], Christopher A. Meyer[1], Matt D. Carter[1], William Wood[2], Graham Tewksbury[2], Victor Li[2] and Xin Yao[2]

[1]*The Boeing Company, PO Box 3707, Seattle, Washington 98124-2207, USA.*
[2]*Portland State University, Mechanical Engineering, PO Box 751, Portland, Oregon 97207, USA.*

The near-beta alloy, Ti-5Al-5Mo-5V-3Cr-0.5Fe (weight percent) (Ti-5553), is being developed for high-performance airframe applications. Although derived from established Russian alloy VT-22, additional investigations to clarify phase transformations, phase equilibria and microstructural evolution in this alloy are continuing. This understanding is important to enable smooth production implementation of Ti-5553. The purpose of this paper is to review on-going efforts to achieve this end. Preliminary time-temperature-transformation and continuous-cooling-transformation diagrams are presented, based on dilatometry and related isothermal heat treatment experiments. Initial data indicate that $\beta \rightarrow \alpha$ precipitation ensues at the nose of the C-curve within 1-2 minutes of quenching from above the beta transus, and that the precipitation reaction is largely complete within an hour. Reaction start is slightly accelerated by beta enrichment following subtransus solution treatment. Omega-like phonon scattering was observed within the beta phase, and precipitation of α_2 (Ti_3Al) from the alpha phase is possible in this system under certain conditions. The bulk $\beta \rightarrow \alpha$ transformation begins at 600-700 ˚C for industrially-accessible cooling rates, and is suppressed completely at rates exceeding about 0.25 ˚C/sec.

Keyword: *titanium (Ti), Ti-5553, 555, phase transformations, precipitation, CCT, TTT*

1. Introduction

Ti-5Al-5Mo-5V-3Cr-0.5Fe (Ti-5553) is a heat-treatable titanium alloy with an improved strength-toughness combination at thicker gauges relative to competing alloys, such as Ti-10V-2Fe-3Al. This alloy is capable of minimum ultimate strength values of 1100 MPa (160 ksi) in the high toughness condition and of values exceeding 1240 MPa (180 ksi) in the high-strength condition[1)]. The nominal composition is given in Table 1. Based on Russian alloy VT-22, it achieves this improvement through high alloying levels of slow diffusers that delay precipitation onset, and inhibit microstructural coarsening, but has the added benefit of reduced Fe to help avoid beta fleck melt defects. Higher Cr compensates for the lower Fe content as a beta stabilizer.

Little has been published on phase transformations and equilibria specific to Ti-5553. Early work in VT-22 only revealed alpha and beta phases, although unusual instabilities in the beta phase were reported[2)]. Harper et al[3)] noted that both bimodal (globular + lamellar alpha) and lamellar microstructures can be produced in Ti-5553, depending on whether the worked structure is subjected to solution treatment below or above the β-transus, respectively. They also noted the occurrence of unusual " saw-tooth" morphology alpha plates. In related work[4)], it was shown that athermal omega phase forms during oil quenching from the beta, and that subsequent aging leads to sympathetic nucleation of alpha on the fine omega and low ductility. Further, beta recrystallization kinetics were reportedly sluggish in this material and the window between full recrystallization and grain growth relatively small.

More recently, Clement, Lenain and Jacques[5)] published that intragranular alpha precipitation in Ti-5553 was significant only below 700 °C, with the resulting strength relatively low when aged above this temperature. This was presumably due to fairly coarse precipitated alpha phase The best balance of tensile properties, 1200 MPa ultimate strength and 7% elongation, were achieved for material quenched and aged at about 700 °C, and significant ductility loss occurred if aged below 600 °C. The ductility loss was ascribed to the presence of omega phase. Faint <112> β diffuse scattering was reported in electron diffraction patterns for the as-quenched condition, a precursor to omega precipitation, but no discrete omega precipitates. It was also noted that, although the microstructures and properties are highly adjustable in metastable beta alloys, a lack of control of the phase transformations can result in deleterious changes to the mechanical properties.

2. Microstructures

As with other near-beta titanium alloys, strengthening is primarily achieved via precipitation of alpha in the parent beta. For Ti-5553, the β-transus temperature occurs in the range of 850-880 °C. Thus, heating above this temperature creates a single phase beta solid solution; subsequent cooling below the β-transus leads to alpha precipitation in the α+β field. See Figure 1. With sufficient driving force (stored deformation energy), sub-transus recrystallization will create globular alpha particles and somewhat enriched beta phase (with Fe, Cr and Mo), which can then be aged at a lower temperature (typically, 540-600 °C). This is the solution treated and aged (STA) condition for which a typical light optical microstructure is shown in Figure 2a.

If super-transus beta annealing, slow cooling and aging at 540-600 °C (BASCA) is employed, a fully lamellar structure with grain boundary alpha, as shown in Figure 2b, is produced. Slow cooling is utilized to coarsen the transformed structure, which typically provides

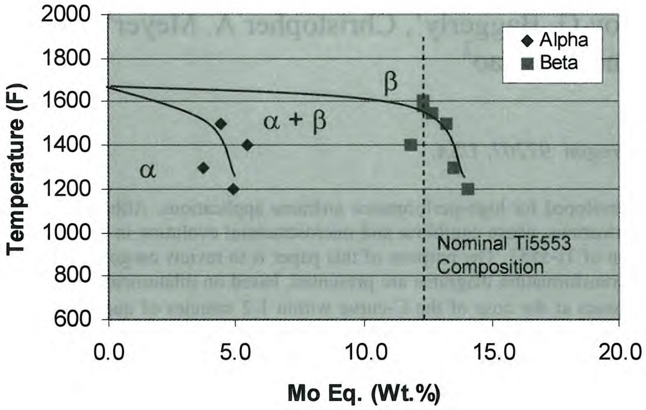
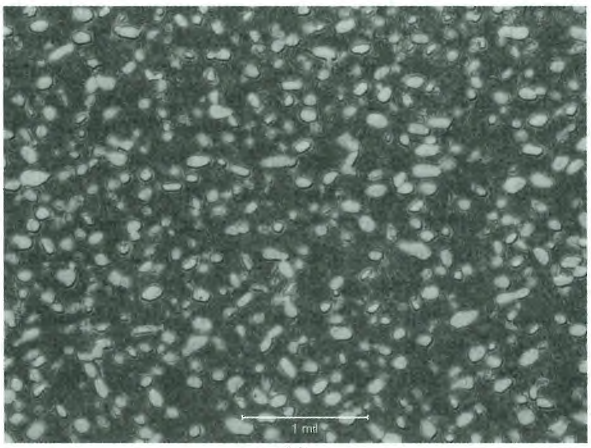
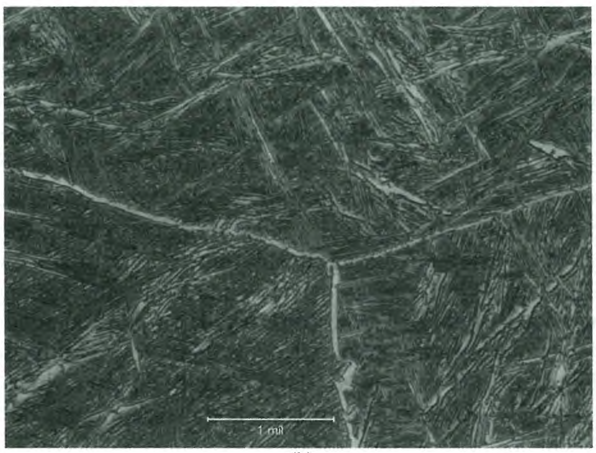


Figure 1. Approximate Pseudobinary Phase Diagram for Ti-5553. Small specimens were held for 40 minutes in an air furnace at the designated temperature and air cooled. Compositions were based on quantitative electron microprobe analysis (EMPA) of the resulting polished specimens, and Mo equivalent[8].

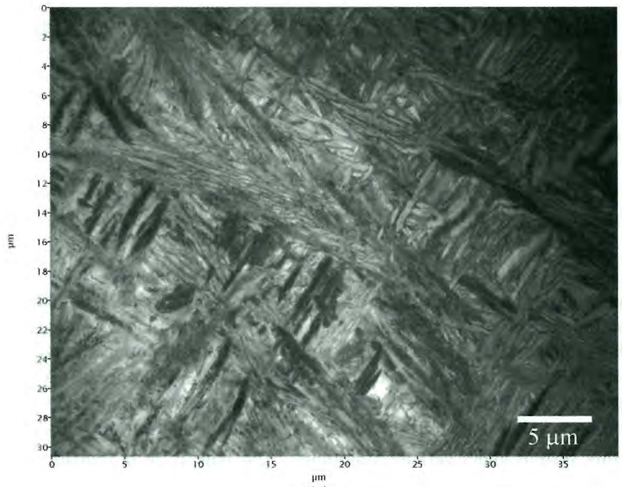


(a)

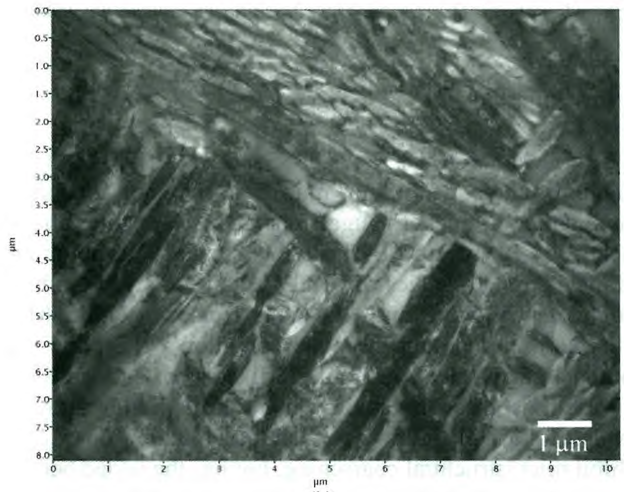


(b)

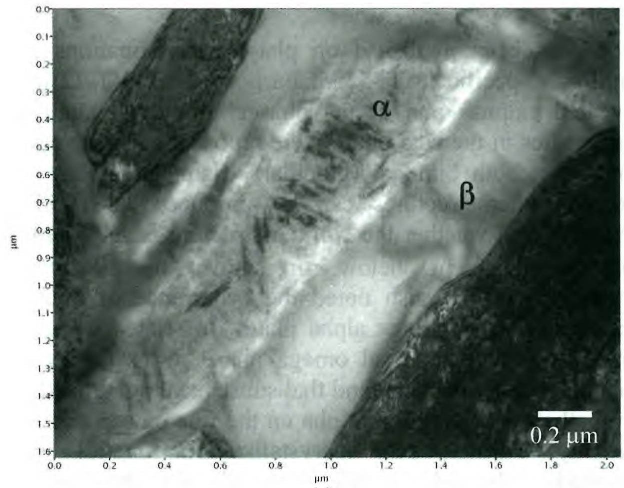
Figure 2. Typical Light Optical Microstructures of Ti-5553 in the (a) STA and (b) BASCA Conditions.



(a)



(b)



(c)

Figure 3. Transmission Electron Micrographs, Bright Field, of Typical Lamellar Alpha Structure following BASCA Heat Treatment. Selected area electron diffraction confirmed the precipitating lamellar phase as alpha in a beta matrix. Weak α_2 (Ti_3Al) diffraction spots (Figure 4) were also observed to originate within the alpha.

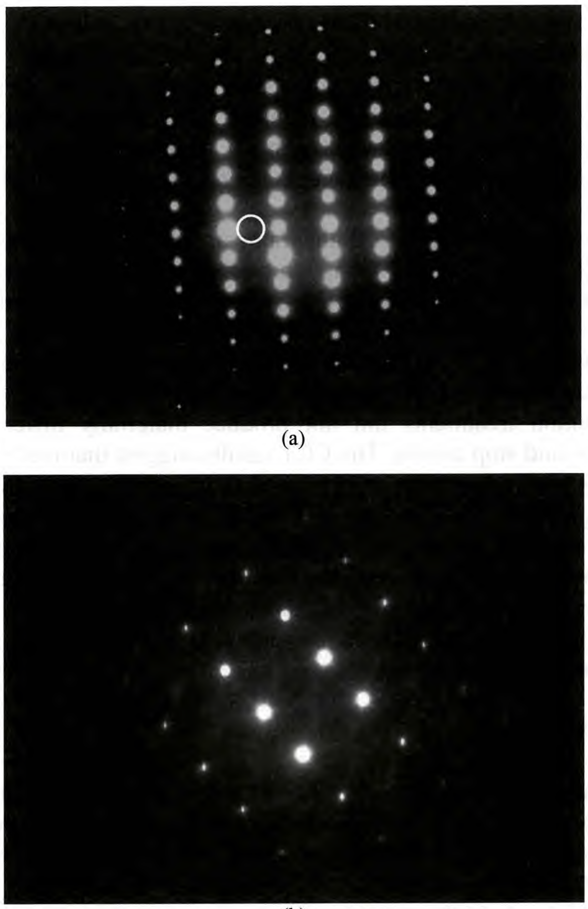


Figure 4. Selected Area Diffraction Patterns from Figure 3 Regions: (a) Alpha (HCP), zone axis [$2\bar{1}\bar{1}0$]. The faint spots (circled) are superlattice reflections due to α_2 (Ti_3Al). (b) Beta (BCC), zone axis [110]. The streaking is due to omega-precursor diffuse scattering.

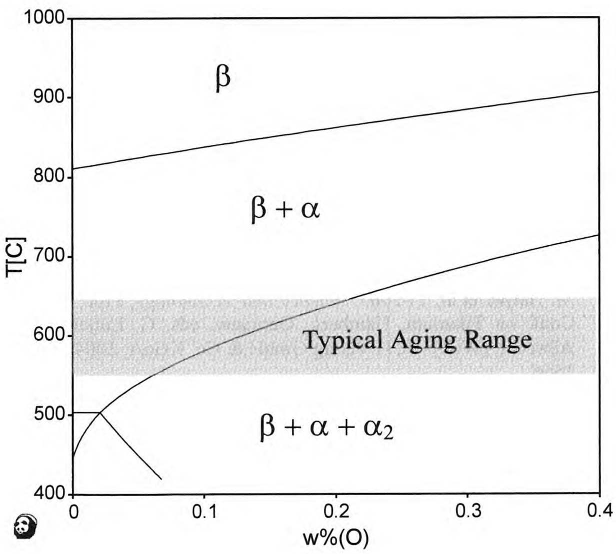


Figure 5. Ti5553 Calculated Isopleth of Phase Boundaries vs. Oxygen Content at Nominal Chemistry (5.0 Al, 5.0 V, 5.0 Mo, 3.0 Cr, 0.4 Fe and 0.01 N, Wt.%). Note the stability of α_2 rises with oxygen content.

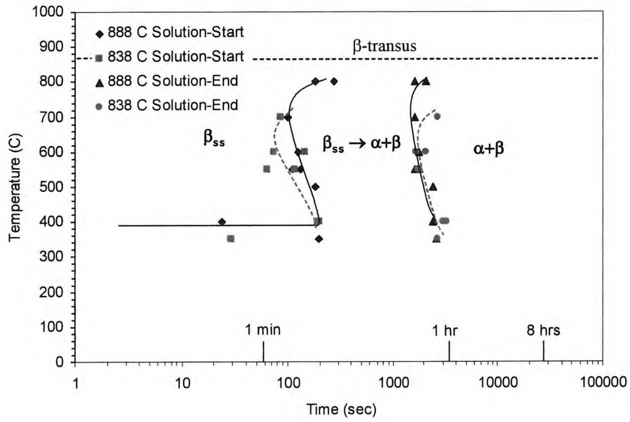


Figure 6. Time-Temperature-Transformation Diagram for Ti-5553 as Determined by Dilatometry. Specimens were held three hours at 838 or 888 °C prior to cooling. Curves for grain boundary α and α_2 are not indicated.

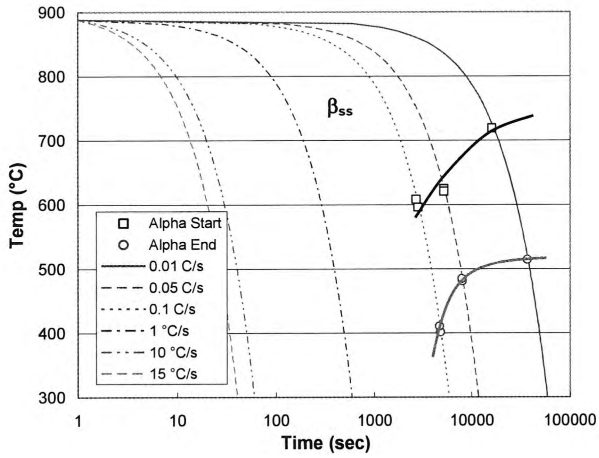


Figure 7. Continuous-Cooling-Transformation Diagram for Bulk Alpha Phase in Ti-5553 as Determined by Dilatometry. Specimens were held for 60 minutes at 888 °C prior to cooling at the indicated rates.

improvement in the damage tolerance characteristics. Closer analysis of the microstructure, Figure 3, shows fine alpha lamellae in a beta matrix. Aging at, or below, 600 °C for several hours resulted in minor amounts of α_2 (Ti_3Al) within the alpha—see Figure 4a. This was not unexpected, since when the Al equivalent [Al-eq = Al 10*(O+N+C)] exceeds about 6.5 wt.%, α_2 is favored[6-7]. In this case, the alloy heat chemistry was Heat A, as shown in Table 1, which had alloy equivalents of Al-eq = 7.05 and Mo-eq = 13.20[8]. Thermodynamic modeling[9] further illustrates the effect of Al-eq, via the strong effect of oxygen content, on α_2 stability, as shown by the pseudobinary phase diagram in Figure 5. This effect was corroborated by additional aging at 620 °C for eight hours, which eliminated the α_2 and suggested the α_2 solvus is around 550-600 °C. (Generally, α_2 is to be avoided, since it can result in a notable drop in ductility and stress corrosion resistance[6]). The beta phase was unremarkable, except that faint omega-precursor scattering was visible in the selected area diffraction patterns (Figure 4b).

Table 1. Compositions of Ti-5553 in this Study (wt.%)

Heat	Ti	Al	V	Fe	Mo	Cr	O	N	C
Nominal	Bal	5.0	4.7	0.4	4.7	3.0	0.15	0.03	0.01
A	Bal	5.16	5.20	0.42	5.15	2.83	0.17	0.02	0.01
B	Bal	5.38	4.85	0.31	5.20	2.93	0.17	0.01	--

From these observations, it is clear that microstructural development in Ti-5553 depends mostly on the diffusional $\beta \rightarrow \alpha$ precipitation reaction. In industrial practice, for large parts, cooling rates from the beta will by nature be relatively slow, i.e. < 0.1 °C/sec (10.8 °F/min). Therefore, effort was undertaken to generate a continuous-cooling-transformation (CCT) diagram for Ti-5553 in this range, and accompanying time-temperature-transformation (TTT) diagrams for both super-transus and subtransus annealing.

3. TTT and CCT Diagrams

High-Temperature dilatometry was utilized, via a Duffers Gleeble thermomechanical simulator[10], to investigate alpha precipitation kinetics. Heat chemistry is indicated by Heat B in Table 1. The TTT experiments were conducted by programmed resistance-heating 0 mm diameter by 120 mm long bars machined from alpha-beta forged Ti-5553 to 28 °C above, or below, the β transus (about 832 °C), and then rapidly quenching and holding at six (6) temperatures (350, 400, 500, 600, 700 and 800 °C) for times sufficient to capture beta decomposition kinetics. All runs were conducted in vacuum. The specimens were held at temperature until the transformation rate was not discernable by differentiation, to a maximum of three hours. The CCT experiments were carried out in similar fashion, by beta annealing at 888 °C for 60 minutes, and then cooling linearly at 0.01, 0.05, 0.1, 1.0, 10 and 15 °C/sec from 800 to 300 °C under vacuum in the Gleeble unit.

Changes in specimen diameter were likewise collected and analyzed using Dynamic Systems Inc.'s CCT software[10] to identify the start of the bulk $\beta \rightarrow \alpha$ transformation during cooling via the slope-line method. Note: Grain boundary alpha precipitation was deemed undetectable by this technique and equipment and is not considered in this effort.

Resulting TTT and CCT diagrams are shown in Figures 6 and 7, respectively. Although there is an apparent nose to the alpha-start C-curve at 600-700 °C, the data indicate that the time to transformation start did not depend strongly on temperature. Additional transformation was essentially undetectable within roughly one hour. The sub-transus solution treatments did not produce materially different start and stop curves. The CCT results suggest that the bulk $\beta \rightarrow \alpha$ transformation during cooling is not significant until about 650 °C. Cooling at a linear rate of 0.05 °C/sec (5.4 °F/min), which is achievable for large alloy sections, would take about one hour to reach this temperature. Cooling rates faster than about 0.25 °C/sec appear to suppress the transformation entirely, but can result in athermal omega phase.

4. Summary

Phase transformations in Ti-5553 have been reviewed and assessed. Common solution treatment and aging heat treatments result in lamellar alpha in a beta matrix. The alpha may contain fine α_2 precipitates if the Al-eq is sufficiently high, and the aging temperature is sufficiently low, below about 600 °C. No omega precipitates were found under any conditions, although omega-precursor diffuse scattering was common. On quenching and aging, isothermal precipitation of alpha begins within 1-2 minutes at 600 °C, and only slightly longer times for higher and lower temperatures. Subtransus solution treatment prior to aging appeared to slightly accelerate the transformation start, and delay the end. During continuous cooling, the $\beta \rightarrow \alpha$ transformation does not ensue until about 650 °C. Cooling rates faster than 0.25 °C/sec retain 100% beta.

REFERENCES

1) S.L. Nyakana, J.C. Fanning and R.R. Boyer: JMEPEG **14** (2005) pp. 799-811.

2) A.A. Popov, L.I. Anisimova and V.D. Kibal'nik: Phys. Met. Metall. **52** (4) (1981) pp. 129-136.

3) M. Harper et al, Ti-*2003: Science and Technology*, Proc. 10th World Conf. on Titanium, Hamburg, Germany, eds. G. Lutjering and J. Albrecht, (Wiley-VCH Verlag GmbH & Co. KGaA, 2004) pp. 15591566.

4) M.L. Harper: "A Study of the Microstructural and Phase Evolutions in TIMETAL 555", M.Sc. thesis, Ohio State Univ. (2004).

5) N. Clement, A. Lenain and P.J. Jacques: JOM **59**(1) (2007) pp. 50-53.

6) R.E. Curtis, R.R. Boyer and J. C. Williams: ASM Trans. Quart. **62** (1969) pp. 457.

7) M. J. Blackburn: ASM Trans. Quart. **59,** (1966) pp. 694.

8) See, for example, the Al-eq and Mo-eq equations in Materials Properties Handbook: Titanium Alloys, ASM (1994) p. 10.

9) F. Zhang and J.D Cotton: unpublished work, (May 2006).

10) Manufactured by Dynamic Systems Inc., P.O. Box 1234, 323 Route 355, Poestenkill, NY 12140 USA