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Fatigue crack initiation and propagation of a TiNi shape memory alloy

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In this paper, fatigue crack initiation and the propagation stages of a TiNi shape memory alloy are examined using a low cycle fatigue interrupted test. Submitted to fatigue cyclic loading, the response of the alloy presents a classical pseudoelastic response. Two potential initiation crack areas are highlighted: at the phase interfaces and at the grain boundaries. Propagation results from the coalescence of many microscopic cracks. These two stages are detectable in the last 20% of the total fatigue life.

Keywords: Low cycle fatigue; Pseudoelastic behavior; Fatigue crack initiation; Fatigue crack propagation; Dissipated energy

Shape memory alloys (SMA) are fascinating materials. Unlike the usual metallic alloys, they exhibit very specific thermo-mechanical behaviors, including shape memory effect [1,2], two-way shape memory effect [2], superelasticity [3–5] and damping [6]. Many engineering applications have been developed using these properties in various fields (medical, aeronautic, automobile, domestic or civil engineering).

Within the frame of design and reliability of systems using SMA, it is essential to have phenomenological models representing thermo-mechanical behavior as closely as possible fatigue. Understanding the physical mechanisms governing the cyclic behavior and leading to the degradation is a necessary step in making such models.

In this paper, we focus on superelasticity, also called pseudoelasticity, which describes the capacity of an SMA to support very significant recoverable deformation when a mechanical effort is applied. This deformation results from an austenite–martensite transformation under isothermal conditions, at a temperature greater than the finishing temperature of the reverse martensite–austenite transformation. The aim of this paper is to detect fatigue crack initiation and the crack propagation stages, and to link these observations with cyclic behavior.

The material investigated here is a near-equiatomic TiNi alloy. More precisely, sample analyses showed that the composition is 51.3 at.% Ti and 48.7 at.% Ni.

In order to develop pseudoelastic behavior, the following thermo-mechanical treatment was carried out:

- A first heat treatment, consisting of heating at 850 °C for 1 h followed by water quenching. At this temperature, which is above the recrystallization temperature in the austenitic field [7,8], work hardening due to machining is thus eliminated.
- A mechanical treatment, consisting of work hardening from 0 to 550 MPa at 200 °C. A high dislocation density is thus introduced. This implies an increase in the resistance for slip deformation by raising the critical stress for slip [9–11].
- Finally, a second heat treatment, consisting of heating at 400 °C for 1 h. At this temperature, which is below the recrystallization temperature, the high dislocation density is maintained [11].

The resulting microstructure is mainly made up of fine grains with an average size of 25 μm (Fig. 1a).

The transformation temperatures can be obtained by differential scanning calorimetry (DSC) measurements [12]. Figure 1b displays a DSC thermogram (heat flow vs. temperature). Initially, the material specimen was heated from approximately −40 to 60 °C (lower part of the curve). At low temperatures the material is in the martensite phase. The heat flow peak at a temperature of approximately 16 °C corresponds to the endothermic transition to the austenitic phase. The area

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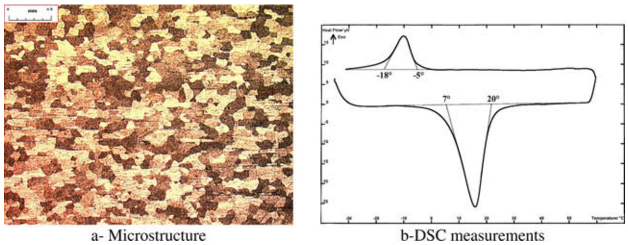

Figure 1. Properties of the TiNi alloy.

under this peak represents the latent heat of transformation and the construction lines indicate the values of the starting ($A_s = 7$ °C) and finishing ($A_f = 20$ °C) temperatures of the transition. The material was then cooled from approximately 60 to −40 °C (upper part of the curve). The heat flow peak at a temperature of approximately −10 °C corresponds to the exothermic transition to the martensite phase. The starting temperature of the transition is $M_s = -5$ °C and the finishing temperature is $M_f = -18$ °C.

The test specimen used was cylindrical, with a gauge section 8 mm in diameter and 20 mm in length, and with a total length of 120 mm. The gauge length was polished mechanically with silicon carbon paper in order to minimize the effects of any surface irregularities, like work hardening due to machining or oxide layers developed at 850 °C. The final surface preparation was then achieved by an electrolytic polishing.

A low cycle fatigue test was conducted on a servo-hydraulic machine (MTS 810) under controlled force (from 0 to 23.4 kN) at 1 Hz. In order to work on the austenitic phase, the test was performed at 50 °C. The signal was sinusoidal in shape, with a null stress ratio.

The strain amplitude was measured using an EPSILON extensometer with a root of 10 mm, placed on the gauge of the test specimen.

The aim was to observe the initiation of cracks (size between 5 and 10 μm) and to follow their growth. The test was stopped at various stages of cycling and disconnected from the servo-hydraulic machine. At each stop, observations of the specimen surface were carried out in a JEOL scanning electron microscope (JSM 5910 LV). The observations reported here were noticed in a homogeneous way on the test specimen surface.

In this paper, the results of the cyclic behavior are presented in two ways:

- As hysteresis loops (Fig. 2a and b), which describe the stress–strain behavior.
- As dissipated energy vs. number of cycles (Fig. 2c), the dissipated energy being equal to the surface of the hysteresis loop in the stress–strain curve.

During its fatigue life, the material investigated here showed two different behaviors:

- From cycles 1 to 1259 (Fig. 2a), a classical pseudoelastic response can be observed. Although the test is stopped, the cyclic behavior looks like those described by Moumni et al. [4] and by Miyazaki et al. [11] on non-stopped tests. Namely, the stress–strain response is characterized by hysteresis loops which evolve during cycling, changing their form and become smaller. Nevertheless, this change tends to stabilize with increasing number of cycles. As many authors [6,11,13,14] have shown, this stabilization effect occurs around the 100th cycle. Contrary to Moumni et al. [4] and Paradis et al. [13], no residual strain is noticed after the first unloading. Residual strain starts to appear at the end of the second unloading, increases during cycling and tends to saturate with a value of 0.4%. Note that in order to observe the surface of the specimen the test is stopped and the specimen is disconnected from the machine. However, following this pause, at the resumption of cycling, the form of the hysteresis loop remains unchanged. This behavior,

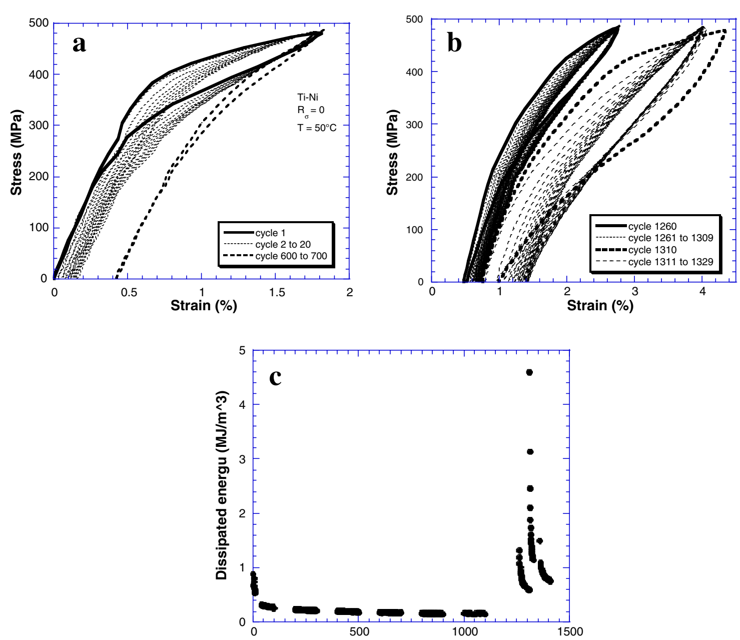

Figure 2. Cyclic behavior. (a) Pseudoelastic behavior, cycles 1–1250; (b) pseudoelastic behavior, cycles 1260–1329; (c) dissipated energy vs. number of cycles.

shown in a diagram of the dissipated energy vs. the number of cycles (Fig. 2c), shows a rapid decrease during the first 10 cycles followed by a slight decrease until around the 100th cycle and a complete stabilization until cycle 1250.

– From cycle 1260 to fracture, the TiNi alloy still presents a classical pseudoelastic response (Fig. 2b), though now cycling interruptions have a significant impact on the shape of the hysteresis loops and thus on the dissipated energy (Fig. 2c). Immediately after a pause (cycle 1260, 1310 or 1360) there is a discontinuity: the value of the dissipated energy increases after each stop in the cyclic loading. However, the global variation in the dissipated energy remains unchanged if cycling allows it: there is a rapid decrease then a stabilization (Fig. 2c).

Many authors [5,13,15] have reported the same kind of behavior for TiNi. For them, pausing between sets of cycles implies a dissipation of energy after each resumption of cycling. When cyclic loading is paused, a phenomenon called "strain recovery" occurs. Paradis et al. [5] explained that the residual strain, believed to be irreversible, can be partially recovered.

In this study, an increase in the dissipated energy after a pause in cycling is observed only after the 1260th cycle and not as soon as the cyclic loading is stopped, contrary to authors previously quoted [5,13,15]. The phenomenon of "strain recovery" could not be highlighted in this study. Indeed, for each scanning electron microscopic (SEM) observation, the test specimen was disconnected from the servo-hydraulic machine, thus concealing any "strain recovery". Here the increase in the dissipated energy observed after a pause could be due to damage, namely crack initiation and growth.

Various SEM observations of the specimen were performed at various stages, in order to detect crack initiation and to follow crack growth. Until cycle 1260 no crack was observed, only the formation of line traces in the grains (Fig. 3a). These line traces indicate the presence of stress-induced martensite plates. As reported by Miyazaki et al. [11], these residual martensite plates increase with the number of cycles and are considered to be one of the causes of the formation of residual strain during cyclic loading. The present observations confirm this statement: the higher the number of cycles, the more martensite plates appear.

In this study, for this material, two crack initiation areas could be distinguished:

– At the martensite–martensite or austenite–martensite interfaces (Fig. 3b), as already mentioned by Miyazaki et al. [11]. For these authors, martensite–martensite or austenite–martensite interface motion causes the formation of defects in the grain, these defects becoming potential crack initiation areas.
– At grain boundaries (Fig. 3c). This observation is in contradiction with the results of Melton and Mercier [16], for whom TiNi memory shape alloy did not exhibit any brittleness with grain boundaries.

Nevertheless, phase interfaces seem to be the preferential areas for crack initiation. Here we observed, over the whole test specimen, more crack initiations in the grains, at the martensite–martensite interfaces, than at the grain boundaries.

After the first crack initiation was observed (at cycle 1260) during a set of cycles, the microcracks increased much more quickly in number than in size. Hornbogen [17] found that, during cycling in steels, the martensite transformation delayed the growth of cracks. Indeed, this transformation induced internal compressive stresses ahead of the crack tip, thus delaying its growth. The same phenomenon could take place in TiNi. The martensite transformation blocks the crack propagation to the detriment of microscopic crack nucleation. It is only at about the 1400th cycle that all microscopic cracks join together, causing a rapid growth of the crack size (Fig. 3d).

The fracture surface presents a brittle feature, namely a smoother fracture surface without secondary cracks. Looking at this more closely, the initiation site (solid

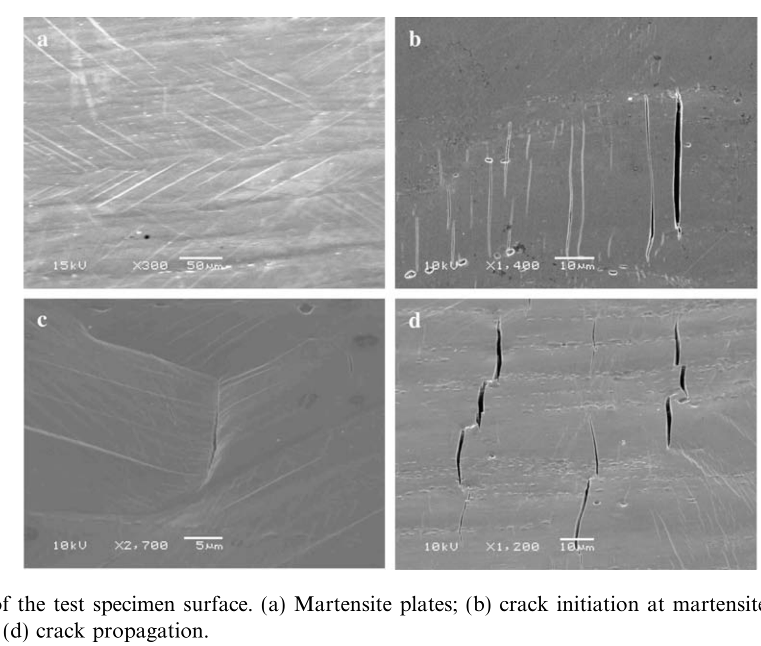

Figure 3. SEM observations of the test specimen surface. (a) Martensite plates; (b) crack initiation at martensite–martensite interfaces; (c) crack initiation at grain boundaries; (d) crack propagation.

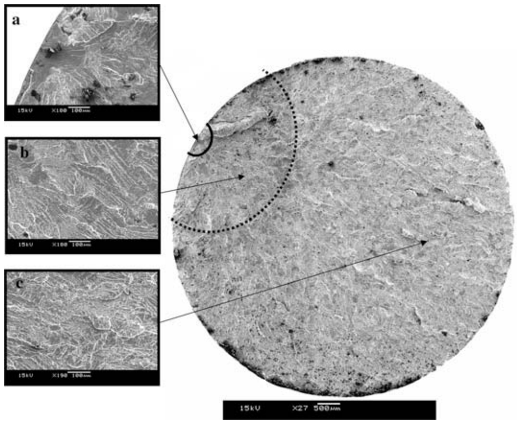

Figure 4. Observations of the fracture surface. (a) The initiation site; (b) the propagation stage; (c) the final fracture.

line in Fig. 4), the propagation stage (dotted line in Fig. 4) and the final fracture are clearly identifiable. The initiation site is identified by the presence of converging river patterns towards the initiation site (Fig. 4a) at the surface of the specimen. As previously explained, initiation is the result of the coalescence of several microscopic cracks. Observations at higher magnification indicate that the crack path is a mixture of transgranular and intergranular propagation (Fig. 4b and c). The specimen broke at the cycle 1576, meaning that the visible damage, namely the crack initiation and propagation stages, occurred in the last 20% of the fatigue life.

Fatigue crack initiation and propagation have been investigated on a TiNi shape memory alloy using a low cycle fatigue interrupted test. In order to observe the initiation of cracks and to follow their growth, many observations of the test specimen were carried out between sets of cycles. The following conclusions can be made:

1. During its fatigue life, the TiNi investigated here presents a classical pseudoelastic behavior. Namely, the stress–strain response is characterized by hysteresis loops which evolve during cycling, changing their form and becoming smaller. Nevertheless, this change tends to stabilize with increasing number of cycles.
2. If the stress–strain response is displayed in a diagram of the dissipated energy vs. the number of cycles, the evolution of the dissipated energy presents two different behaviors:
   – First, from cycles 1 to 1259, a rapid decrease during the first 10 cycles followed by a slight decrease until around the 100th cycle and a complete stabilization until cycle 1250. No discontinuity was observed although the test was stopped.
   – Secondly, from cycle 1260 to fracture, there are many discontinuities. Indeed, an increase in the dissipated energy is observed after each resumption of cycling.
3. Because of the test procedure, no “strain recovery” has been clearly seen in this study. Nevertheless, SEM observations reveal that the increase in the dissipated energy may be linked to microscopic crack initiation. Two sites of initiation have been reported, the first at the martensite–martensite or austenite–martensite interfaces and the second one at the grain boundaries.
   Note that complementary tests will soon be carried out. Low cycle fatigue tests will be performed at different stress/strain levels, at different temperatures and at different stress/strain rates in order to evaluate the influence of these extrinsic parameters on the crack initiation and propagation. The observations of the specimen surface will be realized with a numerical microscope, without disconnecting the specimen and the servo-hydraulic machine.
4. The visible damage, namely the crack initiation and propagation stages, occurs only in the last 20% of the total fatigue life of the TiNi specimen.

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