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# Interpretation of Charpy impact energy characteristics by microstructural evolution of dynamically compressed specimens in three tempered martensitic steels

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ABSTRACT

In this study, Charpy impact energy characteristics of three tempered martensitic steels were evaluated at room and low temperatures by instrumented Charpy impact tests, and were interpreted by fracture initiation and propagation mechanisms in relation with microstructural evolution of dynamically compressed specimens. As the tempering temperature increased, the volume fraction of total carbides increased, while carbides were spheroidized, and the overall Charpy absorbed energy increased at both 25 °C and −50 °C. At −50 °C, the fracture initiation energy largely increased in the tempering temperature range of 400–500 °C, while the propagation energy increased greatly (about 20 times) in the range of 500–600 °C. According to microstructural analyses of dynamically compressed specimens, adiabatic shear bands were formed in the 400 °C- and 500 °C-tempered steels, and worked as preferred fracture propagation paths to critically reduce the fracture propagation energy. In the 600 °C-tempered steel, the deformation energy due to the pendulum impact was effectively absorbed by the combination of fine spheroidized carbides and softened tempered martensitic matrix without forming adiabatic shear bands, thereby leading to the relatively homogeneous deformation in the pendulum-impacted region and the very large increase of fracture propagation energy.

## 1. Introduction

Current models for characterizing toughness in conventional steels generally use upper shelf energy (USE) of Charpy V-notch impact test as fracture resistance values [1–8]. The toughness of steels is often predicted by a simplified semi-empirical formula obtained from the correlation of fracture speed with crack propagation speed [9,10]. In the case of the high-strength steels such as tempered martensitic steels, however, the correlation of Charpy USE with fundamental fracture process or crack speed is less obvious [11–13]. This is because a considerable amount of fracture initiation energy, which is not sufficiently related with actual resistance to fracture propagation as plastic deformation significantly increases at a notch tip, is involved [11–13]. Also, the impact testing method is limited by specimen geometry and reliability, although the evaluation of the absorbed energy has provided reliable standards.

Since the Charpy impact specimen is rapidly fractured by crack initiation and propagation processes during the impact of a heavy pendulum, the plastically deformed area is quite small even at the notch-tip region, which leads to difficulties in the observation of plastic deformation zone [14–16]. Norris Jr. [14], Mathur et al. [15], and Nazari et al. [16]. calculated various stress and strain states occurring in the Charpy impact specimen by using computational simulations. At the time of pendulum impact, the plastic strain field is strongly acting at the notch-tip region, and the fracture initiates along the direction, in which the critical resolved shear stress applied by the tensile stress is highly activated. Since the dynamic compressive stress state is strongly working in the pendulum-impacted region, the compressive strain of 5–10% is built up together with some strain hardening [14–16]. These strain and strain hardening affect directly to the fracture propagation of the Charpy impact specimen and the resultant Charpy impact energy. Furthermore, under a dynamic loading condition of the Charpy impact test, the microstructural evolution occurring near the notch-tip region and pendulum-impacted region, in which the dynamic fracture initiation and propagation processes are predominated, respectively, becomes more important than the quasi-static loading cases. Thus, studies on dynamic deformation mechanisms are essential for the evaluation of alloy designing in relation with microstructural evolution and process control in

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order to improve the Charpy impact energy as well as strength and ductility, but only limited information is available.

In this study, Charpy impact energy characteristics of high-strength tempered martensitic steels were analyzed by dividing total absorbed energy into fracture initiation energy and propagation energy in load-displacement curves obtained from the instrumented Charpy impact test, and effects of tempering and test temperatures on the fracture initiation energy and propagation energy were examined. Deformation mechanisms were also investigated in relation with dynamically deformed microstructures at room and low temperatures, and the correlation between microstructural evolution process and Charpy impact energy was clarified.

## 2. Experimental

Chemical composition of the high-strength tempered martensitic steel used in this study is 0.45 C–0.2Si–0.02 P–0.75Mn–1.0Cr–0.2Mo–Fe (wt%, provided by POSCO). A bloom of $300 \times 400 \times 6400$ mm$^3$ in size made by continuous casting was homogenized at 1250 °C for 5 h in a furnace, and was rolled to make billets of $160 \times 160$ mm$^2$ in cross-sectional area. The billets were heat-treated at 1200 °C for 3 h, and were rolled again to make rods of 15 mm in diameter. These rods were austenitized at 900 °C for 1 h, and were quenched into oil. They were then tempered at 400–600 °C for 30 min to obtain tempered martensitic structures. For convenience, the steel specimens tempered at 400, 500, and 600 °C are referred to as '400 T', '500 T', and '600 T', respectively.

The steel specimens were polished and etched in a 2% nital solution, and microstructures of longitudinal planes were observed by a scanning electron microscope (SEM, model; JSM-6330F, JEOL, Japan). Tensile and Charpy impact specimens were obtained from the center location of the rod. Round tensile specimens (gauge length; 25 mm, gauge diameter; 6 mm) were prepared in the longitudinal direction. Three tensile specimens were tested at least for each datum point at 25 °C and $-50$ °C at a strain rate of $2 \times 10^{-3}$ s$^{-1}$ by a universal testing machine of 100 kN capacity (model; 8801, Instron Corp., Canton, MA, USA). In the case of the low-temperature tensile test, a low-temperature chamber (size; $20 \times 15 \times 15$ cm$^3$) was attached into the universal testing machine. The 0.2% offset stress was determined to be the yield strength in the specimens showing continuous yielding behavior. Charpy impact tests were performed on standard Charpy V-notch specimens (size; $10 \times 10 \times 55$ mm$^3$) three times at least for each datum point at 25 °C and $-50$ °C by a Tinius Olsen impact tester of 500 J capacity (model: FAHC-J-500-01, JT Toshi, Tokyo, Japan). Load-displacement curves were also obtained from the instrumented Charpy impact system attached into the impact tester. After the test, fracture surfaces and cross-sectional areas of the Charpy impact specimens were examined by a SEM to observe fracture modes and fracture initiation and propagation paths.

A split Hopkinson pressure bar was used for dynamic compressive tests [17,18], whose schematic diagrams are presented in Fig. 1. Cylindrical specimens (size; $5\Phi \times 5$ mm) used for dynamic compressive tests were prepared in parallel to the rolling direction so that the specimen orientation was matched with the pendulum-impact direction of the Charpy impact test. For the dynamic compressive tests, the specimens situated between incident and transmitter bars was compressed and loaded by a striker bar (diameter; 19 mm) projected at a high speed using an air pressure of 0.2 MPa (impact velocity; 26.7 m/s), and the strain rate could be controlled by varying the compressive pressure. During the dynamic compression, the incident wave, reflective wave, and transmitted wave were detected at strain gages, and were recorded at an oscilloscope. Among the recorded wave signals, average compressive or tensile strain rate expressed as a function of time was measured from the reflected wave, while compressive or tensile stress expressed as a function of time was measured from the transmitted wave. Dynamic compressive stress-strain curves were obtained from these two parameters by eliminating the time term. Compressive strain rate during the test was about 2000–2500 s$^{-1}$, and dynamic compressive tests were performed three times for each datum point. Detailed descriptions of the dynamic compressive tests are provided in references [17,18].

## 3. Results and discussion

### 3.1. Microstructure and quasi-static tensile properties

Fig. 2(a) through (c) shows SEM micrographs of the 400 T, 500 T, and 600 T steels. All the steels consist of lath-type tempered martensite, and precipitated carbides are classified into interlath and intralath carbides, depending on the precipitation location [19,20]. The average length and volume fraction of interlath and intralath carbides were measured from three different SEM micrographs at least for each steel by using an image analysis software (model; SigmaScan Pro ver. 4.0, Jandel Scientific Co., Erkrath, Germany), and the results are shown in Table 1, although these quantitative data might be overestimated. In the 400 T steel, needle-shaped interlath carbides mainly formed along lath or packet boundaries are predominantly found (Fig. 2(a)). As the tempering temperature increases, the volume fraction of total carbides increases, particularly in the 600 T steel, and carbides tend to be spheroidized (Fig. 2(b) and (c)). The rapid increase in carbide volume fraction in the 600 T steel (about 59% increase) is associated with the increase in volume fraction of intralath carbides whose shapes are spherical or elliptical [21–23]. Thus, morphologies of prior austenite or martensite packet are not clearly defined.

The yield strength, ultimate tensile strength, and elongation of the 400 T, 500 T, and 600 T steels tested at 25 °C and $-50$ °C are summarized in Table 2. At room temperature, the yield strength, ultimate tensile strength, and elongation of the 400 T steel are

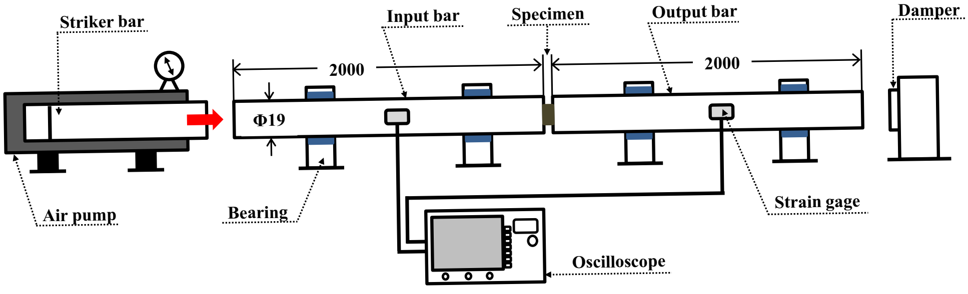

**Fig. 1.** Schematic diagram of the split Hopkinson pressure bar.